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Effects of Structural Steels Microstructure and Waveform on
Corrosion-Fatigue Behaviour of Offshore Wind Turbine Foundations

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ABSTRACT

Marine structures in the offshore environment are subjected to constant cyclic wave and wind forces. Also, the foundations are in direct contact with seawater (SW), hence introducing corrosion damage in the structure. Due to the cyclic loading cracks could nucleate and grow, or existing cracks could propagate at loads far less than the maximum design load, and the problem is worsened in marine environment. Offshore Wind Turbines (WTs) are relatively new structures and their long-term corrosion-fatigue performance data are very scarce. In fact, presently there are little or no public data on corrosion-fatigue performance of these sub-grades of S355 steel used in the design of extra-large (XL) Wind turbine support structures (WTSSs). This experimental research primarily seeks to understand the rate at which a crack grows in the modern normalised-rolled (NR) and thermomechanical control process (TMCP) S355 steel subgrades designated as S355G10+M, S355G8+M and S355J2+N under the influence of: *fatigue waveforms, change in fatigue load level and material microstructure* in seawater (SW) at frequencies within the range of commercial Wind Turbine (WT) operating condition.

The test programme employed a soft-stiff frequency range of commercial WTs in the North Sea. The waveforms considered in this study are constant amplitude sinewave and trapezoid waveform (generally referred to here as hold-time) and the experiments were conducted under frequencies of 0.2Hz, 0.3Hz, 0.5Hz. The stress ratio in all tests was 0.1 and the maximum applied loads were 9kN and 10kN. All calculations were done under Linear Elastic Fracture Mechanics (LEFM) and all tests and investigations were limited to the Paris Region or the Stage II of the da/dN vs. ΔK plot, where LEFM applies. All the tests have been performed in accordance with BS 7910:2013+A1:2015 and ASTM E647-15.

This study found that frequency has no obvious effect on the FCGR of ferrite-pearlite steels in air. There was enhancement of the FCGR in seawater by a factor that depended on the waveform and load level. The CFCGR of the sine waveform was found to be higher than that of the holdtime for all the load levels and frequencies used. Using the mean curve, SW enhanced the FCGR by an

average factor of 1.48 under hold-time and 2.17 under sinewave throughout the Paris Region in the range of 20 - 35 MPa $\sqrt{m}$ with reference to the air mean curve. Similar trend was obtained if mean +2SD is used, giving an average factor of increase of 1.48 under hold-time and 2.30 under sinewave.

Comparison of the experimental results on S355G8+M, S355G10+M and S355J2+N has shown that under both sinewave and hold-time waveforms, the CFCG trend in normalised steels (e.g. S355J2+N) is consistently higher than that of the TMCP steels (e.g. S355G8+M, S355G10+M) in both air and sea environments. Change in load level and frequency did not affect the CFCGR of the TMCP steels in SW under sinewave, but slight change was observed with decrease in the load level for the holdtime. The P_{max} is found to have a profound influence on CFCGR than the cyclic frequency (in the range of 0.2Hz to 0.4Hz). Decreasing the load level reduces the effect of frequency in SW and the difference in CFCGR for waveforms diminishes with increasing frequency and ΔK .

In SW, increase in the fatigue load and decrease in the frequency, especially for holdtime has the highest retarding effect under corrosion dominated regime. This is a consequence of what we referred to here as the *microplastic* zone size. High load level increases the *microplastic* zone size closest and ahead of the propagating crack tip and opens up the crack tip region to allow the entry of more damaging chemical species into the plastically deformed regions causing fast attack by the time-dependent corrosion process resulting to rapid crack tip dissolution and blunting. When the main active crack tip is blunted the crack propagation process is considerably reduced or even arrested for a very long time as observed especially for the hold time test. This appears to suggest that for same ΔK value, combination of low stress range and long crack length, a will be more damaging than combination of high stress and short a in SW. Corrosion products are also found to build up in and around the blunted crack tip leading to further retardation of the crack growth. Extensive blunting of the crack front due to availability of time explains why the CFCGR is generally lower in holdtime as compared to sinewave.

Fractographic and metallographic analyses were carried out to understand the disparity in the CFCGR between the waveforms under the test conditions. Fractographic examination of the fractured surface of the corrosion-fatigue specimens showed that the CFCG mechanism is by ductile striations both in air and SW. The crack path was always non-planar with complex crack front. Generally, three phenomena were identified that primarily retarded crack growth in the ferrite-pearlite steels in air. These are crack diversion, crack bifurcation and metal crumbs. The three factors retarded the crack growth by reducing or re-distributing the effective driving stress at the main propagating active crack tips. It was found that the main crack tip blunting process is the primary factor controlling the CFCGR of steel at high ΔK and low frequency in a ferrite-pearlite steel in SW. Other fundamental factors are crack angle diversion, branching of crack front and formation of metal crumbs along the crack path. The extent of formation of the aforementioned phenomena is a strong function of the steel microstructure. This implies that microstructure has a strong effect on the FCGR of ferrite-pearlite steels in the Paris Region, both in air and SW. This conclusion is in contrary to current theory that microstructure has little or no effect in the Paris Region of the da/dN vs ΔK sigmoidal curve

Transgranular and *quasi-intergranular* modes of propagation in both environments were observed. The *quasi-intergranular* mode is a situation where the fatigue-crack propagated through a thin layer of high solute ferrite ribbon, α_{HA} adjacent to the low relief, α_{LR} or the pearlite, P phases. The morphology and chemistry of the phases local to the main crack front and the load level appeared to determine which mode the crack growth would adopt. The angle the crack front made with the least resistant path ahead of it seemed to determine if it would propagate by transgranular or *quasi-intergranular* mode.

Keywords:

Offshore wind turbine, monopile, fatigue, corrosion-fatigue, fatigue crack growth, microstructure, seawater, S355 steel, BS7910

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NUMENCLATURE

a_o	Initial crack length
a_f	Final crack length
σ_y	Yield stress
σ_{UTS}	Ultimate tensile stress
$\Delta\sigma$	Cyclic stress range
ΔK	Cyclic SIFR
K	Stress intensity factor (SIF)
da/dN	Fatigue crack growth rate (in m/cycle)
P_{max}	Maximum load
P_{min}	Minimum load
R	Stress ratio
α	Ferrite
P	Pearlite
$\alpha-P$	Ferrite-P
γ	Austenite
θ	Cementite (Fe_3C)
α_{LR}	Low relief (LR) ferrite
α_{HR}	High relief (HR) ferrite
α_{RB}	Ferrite ribbon (RB)
$FCGR$	Fatigue crack growth rate, da/dN
CF	Corrosion-fatigue
$CFCGR$	Corrosion-fatigue crack growth rate
$CFCG$	Corrosion-fatigue crack growth
FCG	Fatigue crack growth
$LEFM$	Linear Elastic Fracture Mechanics
PS	Present study
$SIFR$	Stress intensity factor range
$TMCP$	Thermomechanical control process
QT	Quenched and tempered
NR	Normalized-rolled
FID	Final Investment Decision
$HAWT$	Horizontal axis wind turbine
WT	Wind turbine
OWT	Offshore WT
$SLIC$	Structural Lifecycle Industry Collaboration Joint Industry Project
XL	Extra Large
$WTSSs$	Wind turbine support structures
SW	Seawater

1 Introduction

1.1 The wind energy potential and growth

Wind power is recognized as having enormous potential in meeting the European Union 2020 and 2030 renewable energy targets [1][2] and the long term reduction of greenhouse gas emissions by at least 80% by the year 2050 as compared with 1990 levels [3][4]. Wind energy accounted for about 49.2% of all the renewables generated in UK for the second quarter of 2018 [5]. A modern single 1.5MW WT can generate sufficient energy to meet the annual needs of 332 homes, and a 2.5MW WT can generate enough electricity to power over 1,400 homes for a year, enough to power an average computer for more than 2,000 years, or sufficient to make 230 million cups of tea [6]. A 1GW wind farm can power just over one million homes while a 10GW wind farm serves the energy needs of over 7.5 million UK households a year [6][7]. Due to the capacity of this form of renewable energy source, it has witnessed increased use to power, from small and medium sized businesses, and local communities to large commercial environments in countries with enough wind resources.

The growth of offshore wind industry will depend essentially on reductions in wind energy cost. Experience with the technology has revealed that environmentally friendly and economical wind energy can be produced by increasing the size of the turbine [8]. This simply means that the larger the turbine, the greener the electricity [9]. Consequently, offshore wind turbines (OWTs) are growing bigger because of the need to push down the total cost of wind energy to a minimum. Study has shown that further cost reduction could be achieved through efficient, economic and optimized turbine support structure [10].

Monopiles are the most commonly used wind turbine support structure (WTSS) due to their simplicity in design compared to other foundation concepts [11][12]. At water depths up to 30m, monopile has clearly more commercial and technical

advantages. It is well-suited for mass production, installation method based on conventional impact driving and robust in most soil conditions [10]. Study has shown that about 96% of the commissioned offshore wind structures currently are supported on monopile structures while the remaining 4% are supported on jacket structures [13][14]

For newly developed and licensed wind farms in Europe especially in UK and Germany, monopile structures are still gaining more attention from developers and from 10MW concept designs [15][16][17][18][14]. The Gemini Wind Farm of 150 WTs of 4.0-MW turbines, Veja Mate Wind Farm of 67 turbines, Burbo Bank Extension Offshore Wind Farm of 32 MHI Vestas V164-8.0MW turbines and Dong Energy (now Ørsted)/ MHI Vestas 8MW WTs were all installed on monopiles.

The steel structure of a WT is one of the most critical parts in the wind energy development system [19]. This is because mistakes in steel selection and fabrication considerations will impact the cost and lead to the destruction of expensive blades and the nacelle/rotor system. Given the huge contribution of steel material cost in the overall cost of a WT, a material efficient design with satisfactory performance becomes an important step in the manufacture of support structures. Also, the challenges that the large thickness of these XL monopiles support structures can introduce in the structural integrity assessment are of major concern.

WTs are under highly and continuous cyclic loads due to the constant exertion of wind and wave forces. This situation becomes worse if they are in a very turbulent wind climate. The components which are subjected to repeated bending stress may eventually fail by initiation and growth of cracks – in a process commonly referred to as fatigue damage. Fatigue analysis simply involves understanding how component parts of the WT would perform in withstanding continuously varying loads for a particular period of time – along the lifespan scale of the WT.

The monopile structures under wind load moves with a certain natural frequency. The frequency distribution is a significant factor in the structural fatigue life of WT components. The primary excitation frequency due to the rotational speed of the rotor is commonly called the $1P$ - in fact, $1P$ is the rotor frequency. The second excitation frequency of WT can be explained using **Figure 1-1** [20][21]. In **Figure 1-1**, the vertical tower creates wind shade. Anytime a single blade of the rotor passes the wind shade of the vertical tower there is reduction of the wind loading on the tower causing additional cyclic load on the structure at the frequency equal to $1P \times \text{number of blades}$. Thus, $3P$ is the blade passing frequency for a three-bladed WT. Considering all the frequency contents of the applied loads, to design the support structure of WT, the design engineer targets support structure frequency range (i.e. the global frequency of the overall WT) which lies outside the frequency ranges of the rotor, blade passage of tower shade, wind and wave to avoid system resonance. **Table 1-1** shows the $1P$ and $3P$ frequencies for some theoretical and commercial WTs.

In WT design, there are three possible intervals where the natural frequency of the support structure should lie. The frequency of the support structure should not fall into the $1P$ and $3P$ ranges. DNV guideline on design of offshore WT structures [22], stipulates that the natural frequency of WT lies in a narrow curve between the $1P$ and $3P$ frequencies which are specifically at least $\pm 10\%$ away from the $1P$ and $2P/3P$ frequencies. Under these design concepts, the frequency condition of the WT is said to be soft-stiff [23]. **Figure 1-2** shows typical allowable frequency range and excitation frequencies for commercial and reference WTs. The common choice for the frequency position of the support structures of present commercial WTs is in the *Soft-Stiff* region and this informed the use of test frequency that falls within soft-stiff in this research.

Temperature is one of the most important extrinsic variables that influence the mechanical, fatigue and fracture behaviour of steels. Studies show that fatigue life reduces with increase in temperature [24].

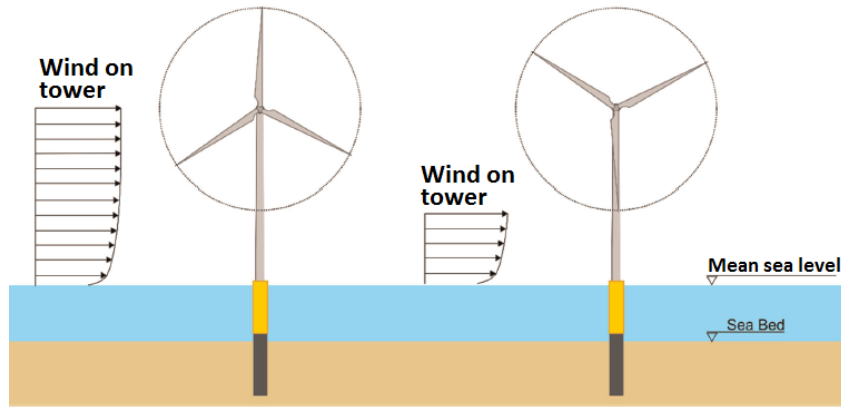


Figure 1-1 Cyclic/dynamic load acting on the tower due to blade shadowing effect (3P load).

Table 1-1 1P and 3P frequencies for some operational and reference offshore WTs

Name	Rotor Freq. Range (1P) [Hz]	Blade passing Freq. Range (3P) [Hz]	Ref.
Vestas V66 2 MW turbine	0.18 - 0.41	0.54 - 1.23	[23]
Vestas V90 3 MW turbine	0.14 - 0.31	0.42 - 0.90	[23]
Siemens SWT-3.6(MW)-107	0.08 - 0.22	0.24 - 0.66	[23]
Vestas V120 4.5 MW	0.17 - 0.25	0.50 - 0.75	[25], [26]
NREL 5MW WT	0.12 - 0.20	0.35 - 0.60	[27]
DTU 10MW RWT	0.10 - 0.16	0.30 - 0.48	[15]
3MW Sinovel WT	0.14 - 0.32	0.41 - 0.95	[20]

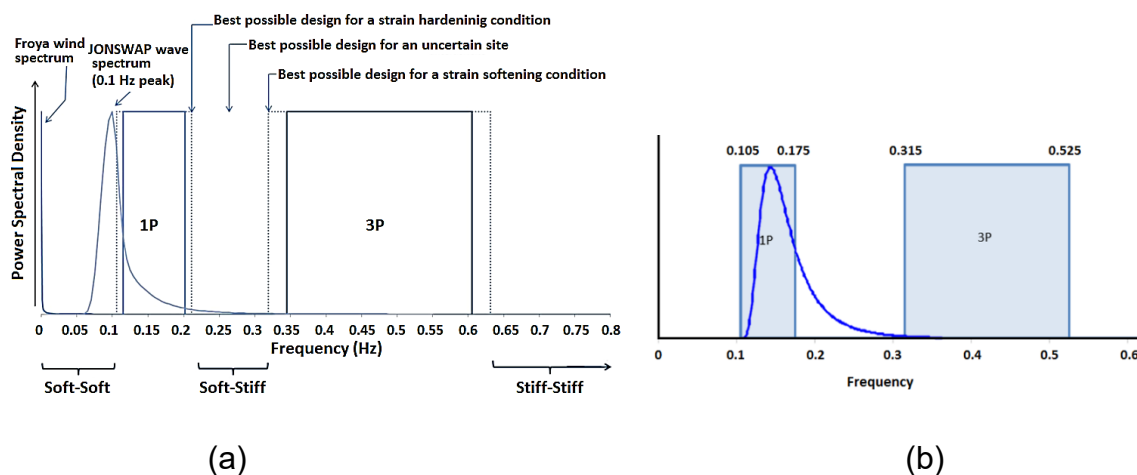


Figure 1-2 The 1P and 3P frequency range for WTs (a) for a 3 bladed NREL standard 5MW WT [27] (b) for the LEANWIND 8 MW reference turbine [28].

But temperature effect on fatigue crack growth rate (FCGR) is rather not sequential and complex, however, it tends to have upper and lower limits [29]. Study found that at free corrosion potential, rise in seawater (SW) temperature from 5°C to 20°C doubled the rate of crack growth [24]. Substantial reduction in fracture toughness is seen at low temperatures, which reduces the critical sizes at fracture [29]. To eliminate ambiguities in corrosion-fatigue test result and to compare the results with published data, a range of temperature pertinent to offshore environment was maintained. The reported range for some of the offshore WTs are shown in **Table 1-2**. A map of the monthly sea surface temperature fluctuations around the UK shows an annual temperature range of 6 to 18°C in January and August, respectively [30].

Table 1-2 Service temperature for some commercial WT

Name	Operational Temperature range	Extreme Temperature range	Ref.
Vestas V164-8 MW turbine	-10 to 25 °C	-15 to 35 °C	[31]
SeaTitan 10MW	-10 to 40 °C	-20 to 50 °C	[16]
Sway ST10	-10 to 30 °C	-20 to +40 °C	[32]

1.2 Steel Materials for offshore application

The magnitude of stresses acting on WT is of primary concern to the design engineer because they directly determine the functionality and lifespan of the WT. Traditionally, European Standard EN10025 S355 (or EN 10225 S355) is the main structural steel typically in use in the design of WT structure [33][34]. The S355 steel subgrades which are given in this standard for design of offshore structures are shown in **Table 1-3**.

As seen in **Table 1-3**, the minimum yield stress for S355 structural steel is 355MPa which explains the origin of the steel's standard name. The minimum yield strength value is for relatively small thicknesses of between 6 and 16mm.

However, the value of the yield strength decreases with increase in plate thickness as seen in **Table 1-3**. The ultimate tensile strength (UTS) for S355 steel is generally between 470 – 630 MPa for thickness of up to 100mm [35][36][37][38].

Table 1-3 Common structural steel grades for offshore design [33][34][39][40]

Non-alloy structural steels		Minimum yield strength in MPa at thickness t (mm)		
European	British			
EN10025 Part 2: 2004	BS4360:1990	$3 \leq t \leq 16$	$16 < t \leq 40$	$40 < t \leq 63$
S355 JR	50B	355	345	335
S355 JO	50C	355	345	335
S335 J2	-	355	345	335
S355J2+N	50D	355	345	335
S355 K2	50EE	355	345	335
S355K2+N	50DD	355	345	335
S355N	50	355	345	335
S355NL	50EE	355	345	335
S355M	-	355	345	335
S355ML	-	355	345	335

The general chemical composition of S355 is given in EN10025 and ASTM 572 standards and summarized in **Table 1-4**. **Table 1-4** shows that S355 steel is generally a low carbon steel whose specifications offer 'high' yield strength and the value varies with the steel grade. Limited grades contain also the compositions that are in bold. Different sub-grades of S355 are selected for use in offshore applications which are often the M series grade. S355M and S355ML are referred to as Thermo-mechanically rolled weldable fine grain structural steels where M stands for Thermo-Mechanical control rolled/(or process) (TMCR or TMCP).

Table 1-4 General composition wt% (max) EN10025 S355 [ASTM 572]
[40][35][36][37][41]

C	Mn	Si	Cu	N	P	S	Cr	Ni	Mo	V	Al	Fe
0.22	1.60	0.55	0.55	0.02	0.06	0.035	0.03	0.50	0.10	0.12	0.02(min)	Bal

Some advanced M series grades regarded as the modern material choice for XL wind turbine design [42] are shown in **Table 1-5**. These steel grades have been manufactured specifically for offshore pipelines, platforms, pressure vessels, and modern WT installations [42]. The material is essentially a very low carbon C-Mn steel. The advantage the steel seems to have over traditional S355 is its very low Sulphur, Phosphorus and minute increase in Ni. Also, amounts of the alloying elements Si, Cu, Mo, N, and V were reduced and this makes it highly ductile. Additional advantage, though obtainable in other class of steel, is that it is hot rolled to give fine-grained microstructure.

Table 1-5 Mechanical properties of advanced M series S355 steel subgrades for XL offshore WT design

Offshore Structural Steel Plates							
European	British	Typical application Min. yield strength R_{eH} (MPa) for thickness t (mm)					
Plates & tubulars							
EN10225	BS7191	t ≤ 25	25 < t ≤ 40	40 < t ≤ 63	63 < t ≤ 100	100 < t ≤ 150	Typical application
S355G7+N	355EM	355	345	335	325	320	Primary Structure
S355G7+M	355EM	355	345	335	325	-	Primary Structure
S355G8+N	355EMZ	355	345	335	325	-	Critical Joints
S355G8+M	355EMZ	355	345	335	325	320	Critical Joints
S355G9+N	-	355	345	335	325	320	Primary Structure
S355G9+M	-	355	345	335	325	-	Primary Structure
S355G10+N	-	355	345	335	325	320	Critical Joints
S355G10+M	-	355	345	335	325	-	Critical Joints
Tensile properties: R_m t(mm) ≤ 100mm: 470 – 630 MPa; t(mm) > 100mm: 460 – 620 MPa							
Minimum elongation = 22%, Minimum average longitudinal Charpy V-notch impact test value = 50J at - 40 °, for all the sub-grade. Max CEV: 0.41 or 0.42							

The corrosion-fatigue resistance of mild steel in seawater (SW) has been found to degrade with dissolution of oxygen in the steel [43]. Generally, the advanced TMCP steels are vacuum degassed, fully killed and control rolled. Fully killed because some alloying elements have been added in very small amount to completely deoxidize or remove oxygen from the steel melt before casting. This prevents release of gases during solidification and so eliminate gas porosity in the product. This practice gives the steel high degree of chemical homogeneity.

TMCP steel is generally used for WT components and auxiliary attachments. *Dillinger* supplied around 94,500 tonnes of S355G8+M plates with thicknesses of up to 95 mm for the Gemini Offshore Windfarm steel foundations [44]. The currently estimated cost per tonne of some sub-grades of S355 steel is given in **Table 1-6**. The implication is that 94,500 tonnes of S355G8+M steel plates supplied for the Gemini windpark cost at least about £56.7 million. Thus, informed and accurate steel selection has the capacity of saving millions of pounds for offshore windfarm.

Table 1-6 Cost of WT steel as at September 2017

S355 steel grade	£/tonne
EN10025: S355JR, S355J0, S355J2	£303 - 380
EN10025: S355JR, S355J0, S355J2 with normalization.	£349 - £644
EN10025: S355G8+M, S355G10+M - TMCP steels	£600 – 680

In the calculation of levelized cost of energy, LCOE to compare the costs of energy generation (£/MWh), WT operating lifetime is a very important parameter. The higher the lifespan, the lower the LCOE. To be able to predict the remaining life of a WT, having structural defects, is not only important to the client but to the insurance company. Lowering LCOE has economic and positive consequence for wind energy sector. Therefore, for WTs to be economically desirable it must not only function and yield enough energy within the design life to cover every single investment into it, but also deliver good profit for business. From the foregoing, important steps towards achieving maximum energy cost reduction in offshore wind is efficient and economic corrosion-fatigue resistant material selection and accurate determination of WT operational life when defect is present.

The first offshore wind farm in the world, Vindeby was developed in 1991 [45] and since then the offshore wind industry has enormously expanded specially due to recent significant cost reductions in the levelised cost of offshore wind energy [46][47][48]. Knowing that offshore OWTs have a typical life span of between 20

and 25 years, a considerable number of offshore wind farms are coming to the end of their design life. The sensitivity of the CFCGR to frequency, load level and waveforms is pertinent in predicting the lifespan of these monopile support structures under forcing conditions. Hence, accurate characterisation of CFCGR in these structures in direct contact with the sea water is vital for inspection, maintenance and potential life extension.

1.3 Aims, objectives and expected industrial benefits of this study

WTs in the offshore environment are subjected to constant exertion of cyclic wave and wind forces. Also, the foundations are in direct contact with seawater (SW), hence introducing corrosion damage in the structure. Although, WTs are coated, but there are periods when the painting peels off giving rise to high rate of pitting and eventual crack formation. There are also many regions of the WT that are unprotected, even in protected areas, cracks could nucleate and grow. The presence of cracks could cause the WT components to fail catastrophically by crack propagation at loads far less than the maximum design load, and the problem is worsened for cyclic loading (fatigue) in marine environment. Thus, it is essential to design the OWT foundations against fatigue failure. The fatigue assessment will ensure that these structures perform optimally throughout their design life. Hence, there is need to inspect and measure the tendency of cracks of particular sizes to propagate under a given stress condition and remedial action taken to ensure they are free from such flaws.

Fracture mechanics approach is frequently employed in the prediction of the loads at which cracks of given length will grow and cause failure. This knowledge enables safe design and aid in economic and effective preventive maintenance. Offshore WTs are relatively new structures and their long term corrosion-fatigue performance data are scarce, or simply non-existent [20]. Presently, there are little or no public data on corrosion-fatigue performance of these so-called

advanced sub-grades of S355 steel used in the design of extra-large (XL) wind turbine support structures (WTSSs).

1.3.1 Overall aim of this research:

In this plethora of modern S335 steels, this experimental research seeks to understand how the modern normalised-rolled (NR) and thermomechanical control process (TMCP) ferrite-pearlite S355 steel subgrades compares in terms of tolerance for crack defects and the rate at which the defect grows under the influence of: *fatigue waveforms, change in fatigue load level, material microstructure and seawater (SW)*.

1.3.2 Main objectives of this research:

To investigate under *WT* loading condition in air and seawater:

1. Sensitivity of FCGR to *cyclic waveform, frequency and change in the fatigue load level*
2. Influence of *microstructure* on the FCGR
3. *Mechanism* of FCG

in S355G10+M, S355G8+M and S355J2+N steel subgrades

1.3.3 Expected industrial application of this research:

1. Economic selection of advanced extra-large steel plates for the design of modern offshore structures.
2. Improved prediction of operating lifetime of offshore structures, especially WTs designed with advanced S355 steels.
3. Provision of guidance on inspection requirements for S355 TMCP steels.
4. Better understanding of fatigue crack propagating mechanism in ferrite-pearlite steels
5. Possibility of development of better fatigue resistant steel grade

1.4 Scope of study

This research investigated corrosion-fatigue crack growth rate (*CFCGR*) in three modern S355 steel subgrades of 90 mm thickness designated as S355G10+M, S355G8+M and S355J2+N. The test programme employed a soft-stiff frequency range (0.2Hz, 0.3Hz and 0.5Hz) of commercial WTs in the North Sea. The test was carried out under constant amplitude sine and hold-time waveforms and stress ratio (*R*) of 0.1. Linear Elastic Fracture Mechanics (LEFM) technique was employed in this research and all tests and investigations were limited to the Paris Region or the Stage II of the da/dN vs. ΔK plot, where LEFM applies.

1.5 Thesis Structure

This thesis is by paper publication.

Chapter 1 presents a general introduction to this research – highlighting the potentials of the wind energy sector and the fatigue challenges confronting these marine structures. Presented also are the aim, objectives and expected industrial benefits of this research.

The literature review is presented in two parts - Chapter 2 and Chapter 3. A journal publication on the review of the current trend in offshore wind energy sector and material requirements for fatigue resistance improvement in large wind turbine support structures is presented in Chapter 2. Chapter 3 (i.e. the Part two) reviews the fatigue loading and operational conditions of wind turbines in marine environment. Also presented is the previous studies on the sensitivity of corrosion-fatigue crack growth rate (*CFCGR*) to fatigue waveforms. The experimental conditions, measurements and analyses procedures in this research are introduced in Chapter 4. Chapter 5 presents a journal publication on the corrosion-fatigue perspective for the selection of XL steel materials for wind turbine support structures (WTSSs). This study was done at 0.3Hz and the

environmental reduction factor (ERF) for the material of study presented. A publication on the critical assessment of the fatigue crack growth rate sensitivity to material microstructure in ferrite-pearlite steels in air and marine environment is presented in Chapter 6. Chapter 7 discussed the waveform and frequency effects on CFCG behaviour in the experimental materials. Also, the metallography and fractography studies on the fractured samples are presented. Chapter 8 presents a study on the effect of microstructure on the CFCGR in the advanced ferrite-pearlite steels in the Paris Region. Detailed metallographic study on the microstructural phases influencing corrosion-fatigue is presented. The role of plastic zone in the corrosion process is also discussed.

1.6 Publications

1. V. C. Igwemezie, A. Mehmanparast. Waveform and Frequency Effects on CFCG Behaviour in Modern Marine Steels, submitted for publication
2. Igwemezie V., Dirisu P., Mehmanparast A. Critical assessment of the fatigue crack growth rate sensitivity to material microstructure in ferrite-pearlite steels in air and marine environment. *Materials Science and Engineering A*. 2019; 754: 750–765.
3. Igwemezie V., Mehmanparast A., Kolios A. Materials selection for XL wind turbine support structures. A corrosion-fatigue perspective. *Marine Structures*. 2018; 61: 381–397. Available at: DOI: 10.1016/J.MARSTRUC.2018.06.008
4. Ali Mehmanparast, Pietro Albani, Victor Igwemezie. Crack growth monitoring in corrosion-fatigue tests using back face strain measurement technique. *Structural Integrity Procedia - ECF22 - Loading and Environmental effects on Structural Integrity*. 2018; 13: 261-266

5. Igwemezie V, Mehmanparast A, Kolios A. Current trend in offshore wind energy sector and material requirements for fatigue resistance improvement in large wind turbine support structures – A review. *Renew Sustain Energy Rev* 2019. doi:10.1016/j.rser.2018.11.002.

2 Literature Review – part one: Current trend in offshore wind energy sector and material requirements for fatigue resistance improvement in large wind turbine support structures – A review

2.1 Abstract

At present, the UK government is driving the survival of the wind energy industry by using interventions that encourage investment in the sector. The use of a Contract for Difference (CfD)/Strike price model by the UK government supports the wind industry and guarantees that wind energy generators have a stable premium over a period of 15–20 years; however, this may not last forever. The growth and stability of the wind industry will depend essentially on continued reductions in wind energy cost, even below that of fossil-fuel based energy sources. Huge cost reduction beyond the present strike price of £57.50/MWh for some projects to be delivered in 2022/2023 may be achieved quickly through efficient and optimized turbine support structure. Consequently, the offshore wind industry is currently making enormous efforts to upscale wind turbines (WTs) from 8MW to 9.5MW, 10MW and then 12MW HAWT (Horizontal Axis Wind Turbine). This level of upscaling no doubt creates tough challenges because the mass of the turbine increases linearly with the cube of the rotor radius. Monopiles having diameters larger than 7m have been proposed, with a wall thickness section in the range of 70–110mm. It is generally thought that Thermo-Mechanical Controlled Process (TMCP) steels are well suited for extra-large (XL-WTs). This paper reviews the present status of WTs and critically assesses the material factors in the structural integrity concerns that may confront the use of XL steel plates in the design of XL-WT support structures.

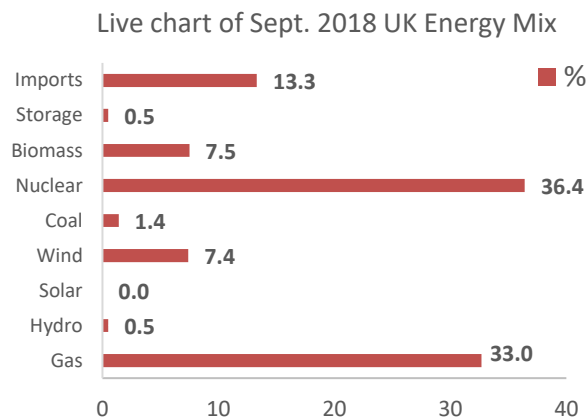
2.2 Introduction

This paper reviews the present status of wind turbines (WTs) and the economic perception of extra-large wind turbines (XL-WTs). It also presents some critical structural integrity concerns relating to XL steel plates, which need to be evaluated to enable technical and cost-effective decisions to be made for the design of next generation WT support structures.

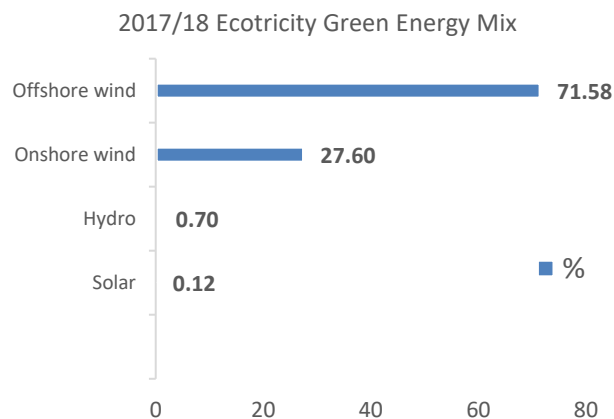
Wind power is recognized as having enormous potential in meeting the European Union 2020 and 2030 renewable energy targets [1][2] and the long term reduction of greenhouse gas emissions by at least 80% by the year 2050 as compared with 1990 levels [3][4]. Wind energy accounted for about 49.2% of all the renewables generated in UK for the second quarter of 2018 [5]. A modern single 1.5MW WT can generate sufficient energy to meet the annual needs of 332 homes, and a 2.5MW WT can generate enough electricity to power over 1,400 homes for a year, enough to power an average computer for more than 2,000 years, or sufficient to make 230 million cups of tea [6]. A 1GW wind farm can power just over one million homes while a 10GW wind farm serves the energy needs of over 7.5 million UK households a year [6][7]. Due to the capacity of this form of renewable energy source, it has witnessed increased use to power, from small and medium sized businesses, and local communities to large commercial environments in countries with enough wind resources.

A typical WT is reported to generate about 70 – 85% electricity at a time, depending on the speed of the prevailing wind. The wind speed is a limiting factor to the amount of energy that can be produced. Nevertheless, the sector is now well established as a major part of the UK's low carbon energy mix. **Figure 2-1 (a)** shows a live chart of energy mix as extracted from MyGridGB [49] on 03 September 2018. **Figure 2-1 (b)** shows electricity supplied by the Renewable Energy Company Ltd, trading as Ecotricity. About 98TWh of renewable electricity was generated in 2017 which is an increase of 18.8% against 2016 records [50]. This increase has been attributed to bigger wind turbines and higher wind speeds.

Onshore wind is said to generate over 50% of renewable power and there are large numbers of projects under development for offshore wind power [2][51]. The UK government is very much committed to the development of this sector by increasing subsidies and planning for a biannual contract for difference (CfD) auction to increase offshore wind capacity [2].



(a)



(b)

Figure 2-1: (a) Plot of Live Chart of energy supply mix in the UK, Sept. 2018 (MyGridGB) [49] (b) 2017/2018 Ecotricity green energy supply mix (Adapted from [52])

As at the end of 2017, UK has the world's largest offshore wind market and with the highest installed global capacity of about 35.7% (5.8GW) for operating

offshore wind capacity [53][54]. If China is excluded, UK then has the largest global capacity of offshore wind projects under construction (2.6 GW) [53]. The UK appears to have succeeded because its government is driving the survival of the OWT industry by using an intervention scheme [55], thereby making it somewhat competitive with non-renewable based energy sources. In fact, towards the end of 2017, BEIS published the outcome of the second Contracts for Difference (CFD) allocation round and three offshore wind developers won a strike price as low as £57.50/MWh for projects to be delivered between 2022 and 2023 in England and Scotland [56].

From the foregoing, continued investment in wind energy will be determined largely by the ability of the industry to continue to achieve reductions in wind energy cost. A study on the Long-term Trends in Wind Energy LCOE [57] shows a steady decline in the cost of energy over time – in fact up to 35%-40% reduction in LCOE by 2030. This hypothetical exercise can only be achieved by considering some factors listed in the report, which include performance improvements of the turbines and support structure/tower concepts. A large wind energy cost contribution comes from the turbines [58]. Although it is ahead of the set milestones, substantial cost reduction can be made in this area by efficient and optimized turbine design.

2.3 The current status of wind turbines

The largest WTs currently installed and functional with power ratings of 8MW are the Siemens Gamesa SG 8.0-167 DD, MHI Vestas V164-8.0MW and Areva 3-bladed 8MW [59][60][61]. This 8 MW WT has been used in the Burbo Bank Extension wind farm [62][63] (see also [8][60]). To stay ahead and to achieve high design efficiency, the industry is currently making intense efforts to upscale WTs from 8MW to 9.5MW, 10MW and then 12MW HAWT. The strike price of £57.50/MWh won by Moray Offshore Windfarm (East) Limited in 2017 is partly due to anticipated use of MHI Vestas V164-9.5MW WT for the 950MW capacity Moray East wind farm, expected to be commissioned around 2022 and 2023.

Currently, General Electric is investing in the development of 12MW Haliade-X – which will be recorded as the world’s biggest OWT, to be available in 2021 [64][65]. This level of upscaling no doubt creates tough challenges because the mass of the turbine increases linearly with the cube of the rotor radius [66]. This trend can be seen in **Figure 2-2**, which shows the increase in the size of the support structure with rotor diameter.

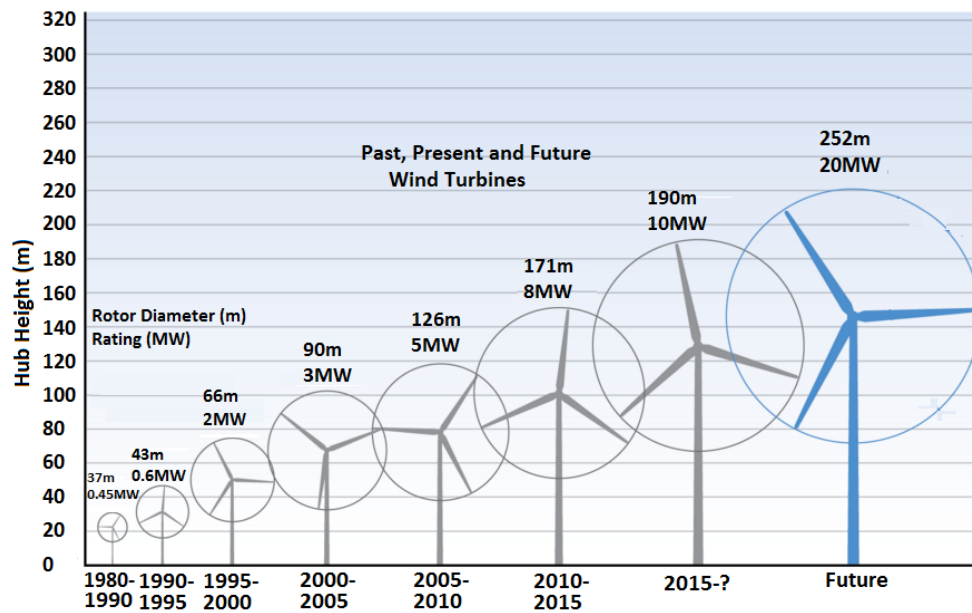


Figure 2-2: Growth in size (and power) of horizontal axis wind turbine (HAWT) (Adapted from [67][68]).

The update is informed by **Table 2-1** which is mainly theoretical concepts found in the literatures and works of [16][17][18][14] [69][70].

Table 2-1: 3 Bladed 10MW HAWT under development

3 Bladed 10MW HAWT	Rated Wind speed (m/s)	Rotor diameter (m)	Ref.
DTU 10MW Reference Wind Turbine	11.4	178.3	[15]
AMSC SeaTitan 10MW Wind Turbine Model	11.5	190	[16]
10MW Wind Turbine Model	12	170	[17]
Rotor Design 10MW Offshore Wind Turbine Model	13.25	141	[18]
*SWAY 10MW-1-bladed Wind Turbine Model	13.4 (Approx.)	164	[14]

From the foregoing, high level industry achievement has been made so far and the offshore wind sector has delivered 20.9TWh of electricity onto the national grid in 2017, while that of onshore wind delivered 29.1TWh [71]. This capacity can generate enough electricity to meet the needs of over 3.2 million UK households. In 2017, the wind energy sector is reported to have supplied the electricity needs of 5.3 million homes in the UK. Analysis has shown that deep offshore designs will contribute immensely in lowering the LCOE in offshore wind energy by unlocking the offshore market potential in the Atlantic, Mediterranean and deep North Sea waters [72]. It is reported [72] that deep waters in the North Sea alone could provide turbine energy capable of meeting the EU's electricity consumption four times over. To harness better energy resources at sea, the offshore energy industry is frantically researching and developing concepts for offshore wind farms in deeper waters. **Figure 2-3** shows global offshore wind projects at various water depths and distances from the shore with the project phase colour-coded [73]. The figures show that some projects are being planned for water depths up to 50m.

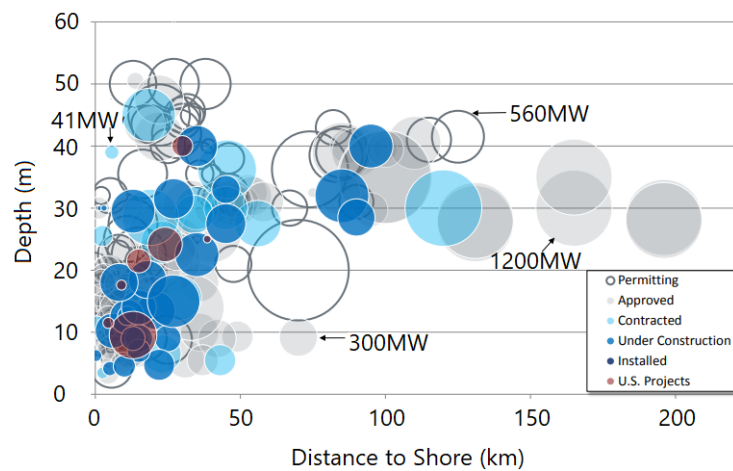


Figure 2-3: Average water depth vs. distance to shore for global offshore wind projects where the bubble size represents project rated capacity (in MW) [73]. (Reproduced with permission)

2.4 Support structure for large wind turbines

As aforementioned, wind energy providers are assiduously pushing towards reduction of the cost of energy. Experience with the technology has revealed that environmentally friendly and economical wind energy can be produced by increasing the size of the turbine [8]. This simply means that as the turbines become bigger more green energy is generated and our environment is the better for it [9].

The monopile is the commonly used support structure for WTs. This is due to their ease of design and manufacture, as compared with other foundation concepts. Also, monopile is very suitable for water depths up to 30m and well-suited to mass production, has a conventional impact driving installation method and remains rigidly in position in most soil conditions [10]. At up to 30m depth, the monopile has clearly more commercial and technical advantages. As offshore activity is now moving into deeper waters, there is ongoing research on which support structure is suitable.

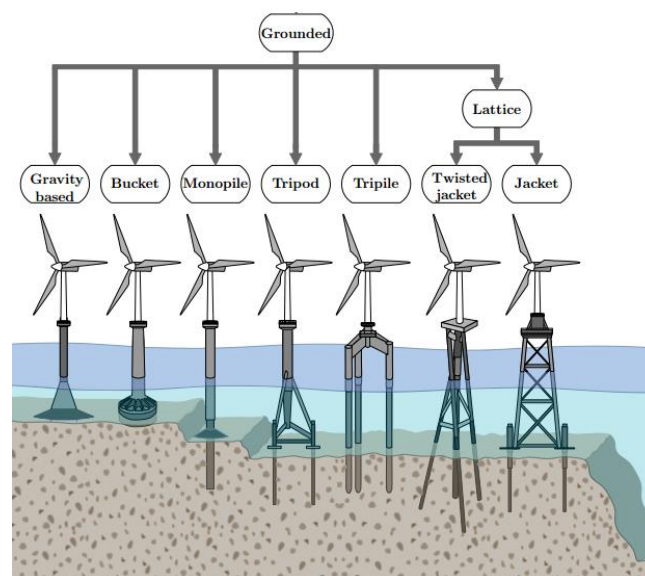


Figure 2-4: Most common grounded turbine support structures and some of their modifications [70] See also [72][74]. [This figure is used under MDPI Open Access Information and Policy]

There are four common types of OWT foundations – gravity based, monopile, tripod and jacket – other modifications are simply based on these four, as shown in **Figure 2-4**. Beyond 30m water depth, other types of foundation structure start gaining preference over the monopiles. One report states that about 96% of the presently commissioned OWTs are supported on monopile structures while the remaining 4% are supported on jacket structures [14]. Some of the existing and conceptualized solutions for water depth above 30m are shown in **Figure 2-5**.

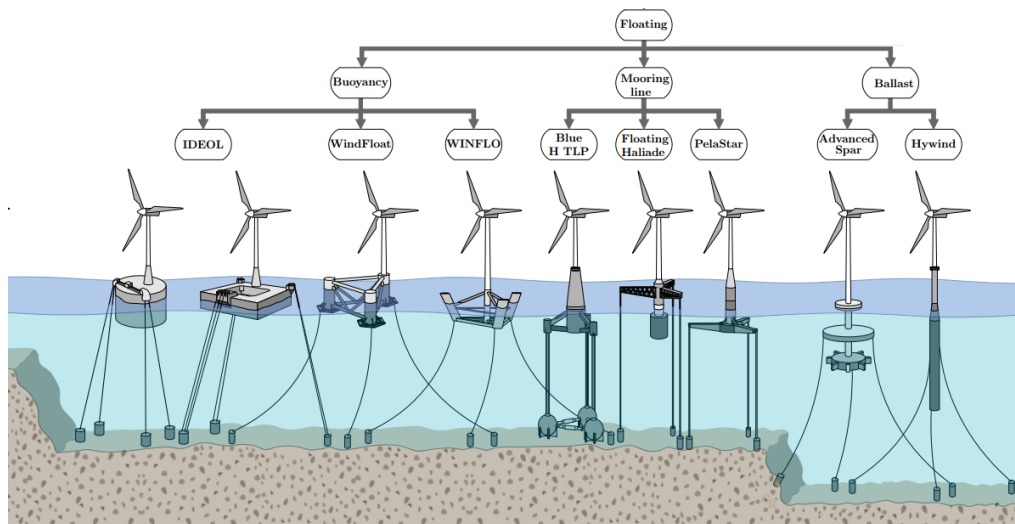


Figure 2-5: Existing floating solutions and proposed deep water support structures for WTs [70] see also [72][75]. [This figure is used under MDPI Open Access Information and Policy]

These floating WTs are increasingly gaining acceptance and deployment. They offer great means of expanding the wind energy potential and reducing energy cost. Since the successful commissioning of the first floating wind turbine - Hywind Scotland, in 2017 the technology has seen an increasing interest by investors. Other pioneering floating WTs in operation are the Fukushima Phase 1&2 5MW that started operation in 2017 [76]; Fukuoka with total 7.5MW that started operation in 2018 [76]; The WindFloat Atlantic project, Viana do Castelo Portugal [77]; Dounreay Tri Scotland, started in 2017 and expected to be commissioned in 2020; Floatgen, France's first offshore floating WT commissioned in 2017 [78]; and Kincardine Floating Offshore Wind Farm projects

are expected to be completed in 2020 [79]. The 24MW Groix offshore wind and 24.6MW Elomed Gruissan in France are expected to go commercial in 2020 and 2021 respectively. The Provence Grand Large in Faraman, France is to be deployed in 2020 and will use 8MW Siemens WTs [73]. This indicates an increase in the size of the support structure for modern floating WTs.

Bak et al. [66] presented studies on the dynamic loading and long-term cyclic WT loading on these alternative foundations. Consequently, jacket structures tend to be favoured beyond 30m water depths, up to 50m. These types of studies are giving impetus to the shift of attention to alternatives such as multi-pod foundations (e.g. jacket, tetrapod and tripods) supported on shallow foundations to reduce the environmental effects of piling noise [27]. They are reported to offer about 50% reduction in the quantity of steel used for their manufacture compared with the monopile structure [80]. This may form one of the reasons for their serious consideration and related foundation type for larger turbines in deeper waters [13]. However, the choice of substructure type largely depends on the nature of loads to be supported, water depth, soil conditions, storage requirements, manufacturing, installation, and transportation cost.

2.5 Offshore wind turbines and size of monopile

Many of the newly licensed wind farms in Europe, especially in UK Round 3 and Germany, to be developed before 2020 are at a water depth of 40m or more. Despite the studies, it appears that monopile structures are still gaining more attention from developers and from 10MW concept designs (see **Table 2-1** references). Following the results of advanced finite element (FE) modelling, highly optimized design methods and experiences from full-scale measurements, suggestions have been made regarding the possibility of using monopiles for water depth above 30m [12].

The Gemini Wind Farm of 150 WTs of the Siemens SWT-4.0-130 model with a production capacity of 4MW each and a total capacity of 600MW, sited 85km from

the coast of Groningen in the Dutch part of the North Sea in water depths up to 37m, is reported to be one of the world's largest wind farms currently in existence [81]. The wind farm is fully commissioned and is expected to serve the energy needs of about 785,000 households or 1.5 million people, reducing CO₂ emissions by 1.25 million tons per annum [81][82][83]. The yearly power production is expected to reach 2.6TWh and ultimately will add towards reducing the cost of wind energy in the Netherlands and Europe. The WT in this project is supported by a monopile with a base diameter of 7m and top diameter of 5.5m, with each steel monopile weighing 670 – 916 tons [83][84].

Veja Mate Wind Farm has 67 turbines of the Siemens SWT-6.0-154 model with 6MW each and a total capacity of 402MW, sited in the North Sea at a distance of 95km from the shore. It is capable of powering about 285,000 homes with a reduction in CO₂ emissions of about 575,000 tons per annum and SO₄ of about 13,400 tons per year [85]. The monopile diameter for this 6MW turbine capacity is expected to be from about 7.8m [83] and each steel monopile weighs about 1,300 tons. Also, the Gode Wind Offshore Wind Farm in the German North Sea is expected to use 97 Siemens 6MW WTs with a 7.5m diameter monopile [86].

The Burbo Bank Extension Offshore Wind Farm has 32 MHI Vestas V164-8.0MW turbines with a production capacity of 8MW each and a total capacity of 254.2MW. It is sited in the Irish Sea in a water depth a little above 30m and 11.5km distance from the shore. It is reported to have the capability of powering 180,000 homes with a reduction in CO₂ emissions of about 364,000 tons per annum and SO₄ of about 8,000 tons per year [87]. The monopile diameter for this 8MW turbine capacity is about 8 to 9m [15][88]. The V164-8.0MW turbine with same monopile diameter is expected to be used for the Moray Firth Eastern Development Area 2, with 93 turbines, and many others under construction across the UK, the Netherlands, Germany, etc.

Monopiles having a diameter larger than 7m are today referred to as extra-large (XL). Their general thickness is usually in the range of 70 to 110mm [83]. XL

monopiles, with diameters up to 10m, are claimed to be feasible in water depths up to 60m [12][89]. There are current considerations to venture into production of monopiles with 10m diameter and 95mm thick [89]. However, the economic and technical feasibilities of XL monopiles in deeper water will pose enormous challenges as wave action will increasingly interfere with the dynamics of the turbine structure. XL monopiles can also be commercially unviable due to storage, transportation and installation logistics, although there are fewer transportation and installation constraints for offshore projects than onshore projects [73]. The challenges that the large thickness of these XL monopiles support structures can introduce in the structural integrity assessment are the major motivation for this paper and will be discussed shortly.

Economic efficiency is one of the most important considerations in the design of a WT support structure. Hence, there is a need to have a balance in the system components. **Figure 2-6** shows a typical costs breakdown for an OWT [90].

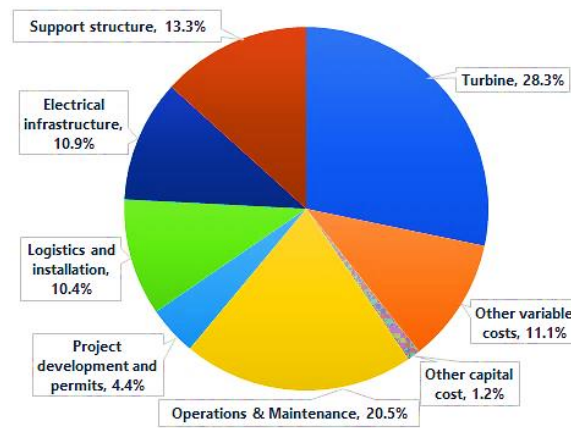


Figure 2-6: (a) Estimated life-cycle cost breakdown for a typical offshore wind project, adapted from [90]. See also [91]

The support structure is typically made of 'high' strength steel and the price contribution is as high as 13.3%. The cost of transportation and erection of the components is a function of the materials used and subsequently the weight of the system. Since the height of a tower directly determines the energy yield, it is

therefore quite important for the final cost of energy to build towers to be optimized and hence should be determined before the design process commences [92]. The steel structure of a WT is one of the most critical parts in the wind energy development system [19]. This is because mistakes in steel selection and fabrication considerations will impact the cost and lead to the destruction of expensive blades and the nacelle/rotor system. Given the huge contribution of steel material cost in the overall cost of a WT, a material efficient design with satisfactory performance becomes an important step in the manufacture of support structures.

2.6 Steel selection for offshore application

Some of the many grades of structural steel in Europe are: S195, S235, S275, S355, S420 and S460. Grades S235, S275 and S355 are the common structural steel grades in use across the EU for many construction projects [93]. **Table 2-2** shows the typical steel grades and properties in use across the UK.

Table 2-2: 3 EN10025: Parts 2 and 4 UK steel grades [39]

EN10025:Part2				
Grade	Yield (R_{eH}^a) min Strength at t=16mm (N/mm ²)	Tensile R_m^a Temp(°C)	Charpy v-notch longitudinal Energy (J)	
S275JR	275	410/560	20	27
S275J0	275	410/560	0	27
S275J2	275	410/560	-20	27
S355JR	355	470/630	20	27
S355J0	355	470/630	0	27
S355J2	355	470/630	-20	27
S355K2	355	470/630	-20	40

EN10025: Part4

Grade	Yield (R_{eH}^a) min Strength at t=16mm (N/mm ²)	Tensile R_m^a Strength at t=16mm (N/mm ²)	Charpy v-notch longitudinal Temp(°C) Energy (J)	
S355M	355	470/630	-20	40
S355ML	355	470/630	-50	27
S420M	420	520/680	-20	40
S420ML	420	520/680	-50	27
S460M	460	540/720	-20	40
S460ML	460	540/720	-50	27

In 'European Standard classifications', structural steels are sub-graded and referenced using standard symbols such as, but not limited to: S, JR, J0, J2, K2, W, Z, C, etc., where S denotes structural steel. The sub-grades JR, J0, J2, K2 refer to material toughness at a particular temperature, W denotes weathering steel, Z structural steel with improved strength perpendicular to the surface and C structural steel formed by cold working. The numerical digits following the symbol S shows the minimum yield strength of the steel. For instance, typical sub-grades of S355 steel are S355K2, S355J2 and S355JR. S355K2 is a structural steel with an impact resistant testing strength of 40J at a testing temperature of -20°C (K2); if given as S355K2W, the W means that this sub-grade has been designed with a chemical composition to resist increased atmospheric corrosion (weathering) (W) [93].

Further letters and classifications based on chemical composition, manufacturing process and relevant application may be used to reference particular grades/products of structural steel. For example, S355J2+N is a structural steel with an impact resistant testing strength of 20J at a testing temperature of -20°C (J2) and has been given a normalized heat treatment (+N). S355JR steel is a high tensile strength structural steel which can be readily welded to other weldable steel. J denotes the notch impact test (done at; JR: room temp, J0: 0°C; J2: -20°C). The simple meaning is that S355JR can withstand an impact energy of 27J at 20°C, S355J0 can withstand an impact energy of 27J at 0°C, and S355J2 can withstand an impact energy of 27J at -20°C. The sub-grades are sometimes

followed by a letter H, e.g. S355J2H, where H stands for hollow section. The *M-series* are TMCP steel equivalents where *M* denotes Thermo-mechanical controlled rolled.

The minimum yield for S355 steel is 355N/mm² (MPa) hence the name S355. The general chemical composition and mechanical properties of S355 are as shown in **Table 2-3**. **Table 2-4** shows that S355 steel is generally a low carbon steel whose specifications offer ‘high’ yield strength and the value varies with the steel grade. Limited grades also contain the compositions that are in bold. The impact strength at different temperatures is seen to decrease due to the fact that steel tends to become brittle when the temperature drops. If a structure is likely to experience temperatures down to -20°C, it is better and safer to then choose S355J2 rather than S355JR or S355J0.

Table 2-3: (a) General composition of S355 (b) Mechanical Properties of EN 10025 S355. [35][36][37][40][41].

Composition wt% (max) EN 10025 S355 [ASTM 572]							
Fe	C	Mn	Si	Cu	N	P	S
~ 97	~0.22	1.60	0.55	0.55	0.02	0.06	0.035
Cr	Ni	Mo	V	Al			
0.03	0.50	0.10	0.12	0.02 (min)			

(a)

Steel Designation (Density 7.85 g/cm ³ {7850 kg/m ³ })							
Mechanical properties (strength at 16mm)	EN 10025 S355 – [ASTM 572] [BS 4360]						
	S355JR [A572]	S355J0	S355J2	S355K2 A656	S355M	S420M	S460M
σ_y (MPa)	355	355	355	355	355	420	460
σ_{UTS} (MPa)	470 -630	470 -630	470 -630	470 -630	470 -630	520-680	540 - 720
EL %	20	22	23	22			17
E (GPa)	210	210	210	210	210	210	210
ν	0.3	0.3	0.3	0.3	0.3	0.3	0.3
Notch toughness longitudinal Energy (J)	27 @ 20 °C	27 @ 0 °C	27 @ -20 °C	40 @ -20 °C	40 @ -20 °C 27 @ -30 °C	40 @ -20 °C	40 @ -20 °C

(b)

Table 2-4: Variation of mechanical properties of S355 steel with thickness [40][38]

Steel Designation (Density 7.85 g/cm ³ {7850 kg/m ³ })			
Mechanical properties <i>t</i> = thickness	EN 10025 S355 – [ASTM 572]		
	Normalized 870°C, air cooled		
	<i>t</i> = 3-16	<i>t</i> = 16 - 40	<i>t</i> = 40 - 63
σ_y (MPa)	355	345	335
σ_{UTS} (MPa)	470 -630	470 -630	470 -630

The minimum yield strength 355MPa is up to 16mm thickness. This means that this test was carried out on a specimen whose thickness is 16mm. The value of the yield strength decreases with an increase in plate thickness. It is pertinent for designers to note this. The tensile strength is generally between 470 – 630MPa for thickness up to 100mm.

Traditionally, European Standard EN10025 S355 (or EN 10225 S355) is the main structural steel typically in use for WT support structures [33][34] and different sub-grades of S355 are selected for use in offshore applications. The so-called modern material choices for offshore applications are S335G8+M, S355G10+M, S420G2+M, and S460G2+M steel sub-grades. **Table 2-4** also shows how the mechanical properties of S355 steel changes with an increase in material thickness for a normalized condition from an austenitization temperature of 870°C.

These steel grades have been manufactured specifically for offshore pipelines, platforms, pressure vessels, and modern WT installations [42]. The G grades are hot rolled to give a fine-grained microstructure, vacuum degassed, fully killed. A typical chemical composition of this class of steel S355G10+M (manufactured by Dillinger Hütte) and the mechanical properties of different offshore grades are shown in **Table 2-5**.

Table 2-5: The chemical composition and mechanical properties of offshore steel grades [94]

Composition <i>wt%</i> S355G10+M								
Fe	C	Mn	Ni	Si	Cu	Cr	NB	P
~97.5	~0.06	1.57	0.33	0.27	0.24	0.034	0.022	0.013
Mo	N	Ti	V	S				
0.006	0.004	0.003	0.001	0.0009				

Steel Designation (Density 7.85 g/cm ³ {7850 kg/m ³ })				
Mechanical properties Hot rolled plate	EN10225 S355G10+M			
	S355G8+M	S355G10+M	S420G2+M	S460G2+M
σ_y (MPa)				
<Ø16mm	355	355	420	460
Ø16 - Ø25 mm	355	355	400	440
Ø25 - Ø40 mm	345	345	390	420
Ø40 - Ø63 mm	335	335	380	415
Ø63 - Ø100 mm	325	325	380	406 (63 -80) /400 (80 -100)
σ_{UTS} (MPa)				
<Ø40mm	470 - 630	470 - 630	500 - 660	~520 - 700
Ø40 < 100 mm	470 - 630	470 - 630	480 - 640	~500 - 675
BH	285	285	-	-
EL % (min)	22	22	19	17
E (GPa)	205	205	-	-
ν	0.29	0.29	-	-
M %	55	55	-	-
Charpy V-notch (J)	50 @ - 40 °C	50 @ - 40 °C	50 @ - 40 °C	-
$\varnothing$ (mm) = specimen thickness; σ_y = Tensile Yield strength; σ_{UTS} = Ultimate Tensile Strength; BH = Brinell Hardness; EL = Elongation at break; E = Modulus of Elasticity; ν = Poisson's Ratio ; M = Machinability				

In general, the grading system follows the kind of heat treatment and/or thermo-mechanical processing given to steel. So, we have normalized, normalized-rolled weldable fine grain structural steel while others are thermo-mechanically rolled weldable fine grain structural steels. It is extremely important that these steels are correctly specified, covering the strength grade and the steel sub-grade, in order to avoid brittle failure [39]. In other words, the intention of the structural

classification is to serve as a guide in selecting the appropriate steel and a suitable inspection to avoid brittle fracture.

Selection criteria for steel material for marine structures are commonly based on properties such as strength, toughness and weldability. There are challenges in the optimum combination of high steel strength and large diameter tower in an attempt to reduce total steel tonnage per tower. The tensile ductility requirement considers the gap between yield strength and ultimate tensile strength. The wall thickness, δ is an important criterion for tower stability and must be sufficient to withstand buckling of the tower wall. The optimum combination of tower wall thickness and shell diameter is a cost target.

The welding requirement considers the production of a good weld, which is critical, because there are long-term problems with *fatigue* and *cracks* in welded structures, and an increase in the thickness of steel plate increases this problem. The huge tower is also constrained by the physical capacity of specialist plate manufacturers and challenges associated with the welding of very thick steel plate [19].

2.7 Environmental loading on wind turbine support structure

WT support structures are exposed to high cyclic loading caused by wind and normal operation. If the structure is in an offshore environment, forces due to waves and sea current (tides) will be present. Furthermore, forces due to mass, installation and maintenance activities also act on the structure. Gravitational and inertia forces are due to blade, tower and nacelle masses. The rotor blades experience bending moments in the edgewise direction caused by gravitational loads. For a pitch-controlled turbine, the bending may be in the flapwise direction. The bending moments vary cyclically due to operational rotation of the blades and the magnitude increases with the rotor diameter [22]. Other load conditions that may be considered are those due to operational centrifugal forces, Coriolis

forces and gyroscopic forces. Centrifugal force acts outwardly due to rotation of the blade. Forces acting on a typical WT support structure are axial loads, lateral loads and bending moments, as shown in **Figure 2-7**.

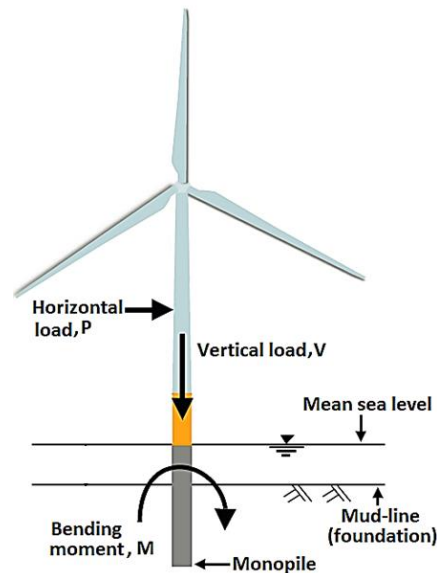


Figure 2-7: Nature of loads on a monopile supported WT [27][23]

Torsional loading is sometimes considered. In other words, axial, lateral, bending and torsional vibrations are all present [13]. The predominant external loading problems come from wind and wave, particularly from wind loads [22].

OWT support structures also have to withstand some extreme loading events during their design life. External loading has two categories, the loading due to normal operating conditions and the extreme loading event occurring in a particular time, say once in 50 years of a wave height say 30m (i.e. 15m wave amplitude) or wind gust. These two conditions are usually put into consideration during the design. The loading situations give rise to fatigue as the dominating mode of structural failure and this is a major concern. The fatigue phenomenon is enhanced by the corrosion process in the offshore environment.

Typical fatigue loading histories on structures are diverse, ranging from simple to repetitive to completely random. In other words, real load conditions are complex.

The loading may be quasi-stationary, periodic, stochastic or transient, varying in time and origin [22][95]. In fact, the stochastic nature of fatigue loading on WT structures is not easily defined [13]. The design of the support structure is fundamentally based on load assumptions and simplifications; the simplification permits numerical modelling and simulations [80]. For a design life of about 20 years, the structure is expected to withstand some 10^9 cycles of fatigue loading [80]. However, the design life and number of cycles the structure of a WT can sustain depends on the structural dimensions, soil-structure interaction, natural frequency, stiffness and damping, water depth, turbine size [23][27][96][97] and material properties of the support structures.

The overturning capacity under extreme conditions is basically what determines the length of a monopile support structure, or the maximum allowable tilt of the turbine over its life time due to accumulated rotations from cyclic loading [11]. The diameter of the monopile is typically driven by requirements for the natural frequency of the turbine, and the soil stiffness is an important factor that is strongly linked to the dimensions of the monopile [11]. Fatigue and shell buckling are the two fundamental parameters that determine the thickness of a WT structure. Fatigue is of most concern in the welded regions.

The common analytical equations often employed in the design of a WT support structure, especially the tower, are given in **Figure 2-6**. It can be seen from these equations that the buckling and bending that must be designed against have a direct relation to the yield strength, diameter and thickness of the steel used for the support structure. Thus, the main concern is to ensure sufficient stiffness to withstand the dynamic force of wind and waves [89].

According to DNV Standard [98], tubular members with low D/t ratios are generally not subject to local buckling under axial compression and can be designed on the basis of material yielding. This simply means that the yield strength of the steel can be taken to equal the local buckling stress. If the ratio

D/t increases, the elastic local buckling strength decreases, making local buckling checks necessary. In other words, for that simple ratio, increasing the thickness of the steel section eliminates elastic buckling tendencies. Increasing the thickness this way to check the buckling of XL monopiles has a severe penalty that must be paid in terms of the structural integrity of the thick steel section.

2.8 Selection of strengthening mechanism for fatigue resistance of support structure of large wind turbine

The aim of the fatigue design of a WT is to ensure that it has adequate cyclic life. Calculated fatigue lives also form the basis for efficient inspection programs during fabrication and the operational life of the structure. DNV-RP-C203 [99] recommends the use of *fatigue tests* and *fracture mechanics* for fatigue analyses of WT structures. Potential fatigue crack could come from welded joints, points of change in geometry, and from other parts of the structure, requiring fatigue analysis to be extended to all parts of the structure or component.

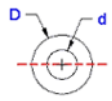
In welded joints, fatigue cracks may start and grow from the weld toe, weld end, HAZ, weld root or even at weld metal. The fatigue strength of welded joints is found to depend to some extent on plate thickness. This effect is said to be due to the local geometry of the weld toe in relation to the thickness of the adjoining plates. It is also dependent on the stress gradient over the thickness. To account for the thickness effect on fatigue, a modification to the stress is made such that the design S-N curve for thickness other than the reference thickness reads [[99]; pp. 17-18]:

$$\log N = \log \bar{a} - m \log \left[\Delta \sigma \left(\frac{t}{t_{ref}} \right)^k \right] \quad (2-1)$$

where m is the negative inverse slope of the S-N curve; $\log \bar{a}$ the intercept of the $\log \bar{N}$ axis; t_{ref} is the reference thickness which equals 32mm for welded tubular

joints, 25mm for bolts, t is the thickness through which a crack will most likely grow; and k the thickness exponent of fatigue strength.

Table 2-6: Common analytical expressions for the design of a wind turbine support structure, especially the tower [15] [22][99][100][101][102]

Parameter	Equation
Area moment of inertia (or second moment of area), $I_{tower}(m^4)$	$I = \frac{\pi}{64}(D^4 - d^4)$  D = outside diameter of the tower d = inner diameter of the tower
Radius of gyration $r_g (m^2)$	$r_g = \frac{1}{4}\sqrt{D^2 + d^2}$
Torsional stiffness constant $K(m^4)$	$K = \frac{\pi}{32}(D^4 - d^4)$
Cross-sectional area of the conical tower, $A (m^2)$	$A = \frac{\pi}{4}(D^2 - d^2)$
Mass per length, $m(kg/m)$ (where ρ is the mass density)	$m = A\rho$

(a) Geometrical parameter calculations

Parameter	Equation
2. Bending check	
$f_b = \frac{M}{Z_e} \leq \frac{F_b}{\gamma_{R,b}}$	f_b = bending stress due to forces from factored actions $\gamma_{R,b}$ = partial resistance factor for bending strength M = bending moment due to factored actions F_b = representative bending strength $F_b = \left(\frac{Z_p}{Z_e}\right)F_y \text{ for } \frac{F_y D}{Et} \leq 0.0517$ $F_b = \left[1.13 - 2.58\left(\frac{F_y D}{Et}\right)\right]\left(\frac{Z_p}{Z_e}\right)F_y \text{ for } 0.0517 < \frac{F_y D}{Et} \leq 0.1034$ $F_b = \left[0.94 - 0.76\left(\frac{F_y D}{Et}\right)\right]\left(\frac{Z_p}{Z_e}\right)F_y \text{ for } 0.1034 < \frac{F_y D}{Et} \leq 120 \frac{F_y}{E}$ Z_e = elastic section modulus $Z_e = \frac{\pi}{32} \frac{[D^4 - (D - 2t)^4]}{D}$ Z_p = plastic section modulus $Z_p = \frac{1}{6} [D^3 - (D - 2t)^3]$

(b) Bending calculations

Parameter	Equation
1. Axial compression and buckling strength	
$f_c \leq \frac{F_c}{\gamma_{R,c}}$	f_c = axial compressive stress due to forces from factored actions F_c = representative axial compressive strength $F_c = [1.0 - 0.278\lambda^2]F_{yc}$, for $\lambda \leq 1.34$ $F_c = \left(\frac{0.9}{\lambda^2}\right)F_{yc}$, for $\lambda > 1.34$ $\lambda = \frac{KL}{\pi r_g} \sqrt{\frac{F_{yc}}{E}}$ $\gamma_{R,c}$ = partial resistance factor for axial compressive strength (= 1.18) E = Modulus of elasticity λ = column slenderness parameter K = effective length factor F_{yc} = representative local buckling strength $F_{yc} = F_y$ for $\frac{F_y}{F_{xe}} \leq 0.17$ $F_{yc} = \left[1.047 - 0.274\frac{F_y}{F_{xe}}\right]F_y$ for $\frac{F_y}{F_{xe}} > 0.17$ $F_{xe} = \frac{2C_xEt}{D}$ F_y = representative yield strength F_{xe} = representative elastic local buckling strength C_x = critical elastic buckling coefficient = 0.3 t = wall thickness of the tower $N_{sd} \leq Af_c$
	N_{sd} = design axial force

(c) Compression and buckling strength calculations

Again, it can be seen from this equation that the number of cycles to failure depends on the stress range and section thickness of the steel. According to the equation, as thickness increases the number of cycles to failure decreases.

Of particular interest for the design of new generation large WTs in the marine environment is the steel sub-grades listed in **Table 2-5**. They are commonly produced by Dillinger Hütte, Tata Steel or ArcelorMittal, certified by either Lloyds Register of Shipping (LRS) or DNV (Det Norske Veritas), depending on the thickness, and supplied as a hot rolled plate with thickness up to 100mm. Thickness of this magnitude invariably increases the scale of defects – inclusions and segregations (brittle zones) – thereby decreasing the integrity of such a structure.

The ideal structural steel combines high strength with excellent fracture toughness. The strength of steel microstructures is commonly factorized into a number of intrinsic components [103]:

$$\sigma = \sigma_{Fe} + \sum_i x_i \sigma_{SS_i} + x_C \sigma_C + K_L \{L\} + K_D \rho_D^{0.5} \quad (2-2)$$

Where σ_{Fe} is the strength of pure annealed iron, 219MPa (MNm⁻²) at temperature of 27°C; x_i is the concentration of a substitutional solute which is denoted here by a subscript i ; σ_{SS_i} is substitutional solute (i) strengthening; σ_C is solid solution strengthening due to carbon; $K_L \{L\}$ is the function for strengthening due to 'grain' size, given as 115 (MNm⁻¹); K_D is the coefficient for strengthening due to dislocations, 7.34 × 10⁻⁶ MNm⁻¹; ρ_D is dislocation density, typically 10¹⁶ m⁻²; and L is the measure of the ferrite plate size, typically 0.2µm.

The reduced alloying-element contents, including carbon of M-series steel subgrades in **Table 2-5**, are expected to give the steel high weldability. This property will be good for thick shell sections in large WT design. For non-tempering steel, the increase in the strength of the steel can be achieved by solid-solution alloying, strain hardening and grain-size reduction. Note that the dispersion-strengthening process is utilized for tempered steels. The reduced alloying elements in these steels will tend to reduce the strength of the steel due to reduction in the strength contributions from solid solution – substitutional (such as Mn, Cr, Mo, etc.) and interstitial (carbon). Again, the M-series are supposed to be hot-worked in an austenite range and essentially there is little or no strain-hardening strengthening contribution in hot-worked steel. To make up for this, other strengthening methods need to be adopted, one of which is grain refinement. It is well established that fine grained microstructure in a steel can confer high mechanical strength and toughness [104]. In other words, decreasing the grain size can increase both the strength and the toughness of a steel simultaneously [105]. A typical plot illustrating the effect of grain refinement on a mild steel is shown in **Figure 2-8**. In **Figure 2-8**, decrease in D increases the flow stress.

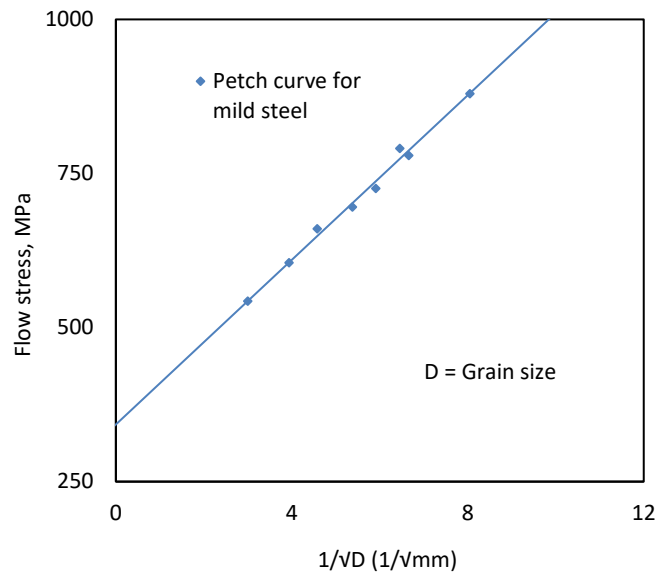


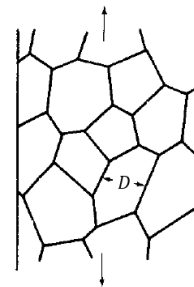
Figure 2-8: The increase in strength with grain size reduction (Adapted from [106][107]).

This strength variation with grain size, at ambient temperature is usually expressed by the Hall-Petch relationship of a polycrystalline material as given in equations 2.3, 2.4 and 2.5.

$$\sigma = \sigma_o + kD^{-0.5} \quad (2-3)$$

$$\sigma_y = \sigma_{oy} + k_y D^{-0.5} \quad (2-4)$$

$$\sigma_F = \sigma_{oF} + k_F D^{-0.5} \quad (2-5)$$



where σ is the yield (or flow) stress; σ_o is the friction resistance for dislocation movement within the polycrystalline grains or simply the friction stress denoting the total resistance of the crystal lattice to movement of dislocation; and k is a measure of the local stress needed at a grain boundary to move the dislocation. If the flow is plastic, it is denoted as k_y or k_c if the flow or fracturing is by cleavage

at a lower temperature. In other words, k is a parameter that measures the relative hardening contribution (dislocation locking) of the grain boundary. D is the average polycrystal grain diameter [106][107][108]. The value of σ_o is not the same for yielding and cleavage. σ_o , k_y and k_c are constants for a particular material.

This equation was proposed by Hall [108] and extensively expanded by Petch [109][110]. The Hall-Petch equation has been found to express the variation of brittle fracture stress with grain size and the dependence of fatigue strength on grain size. Yield stress, σ_y and the "fatigue limit", σ_F depend on steel grain size, D as expressed in equations 2 and 3 respectively [106][111].

This dependence is reported to vary according to whether the fatigue stress is greater than, equal to, or less than the yield stress of the same material for low-carbon steel [112]. The increase in fracture toughness k_c with reduction in grain size as composition and other microstructural variables are kept constant, is shown in **Figure 2-9**. The fine grain size helps to minimize dislocation pile ups stresses.

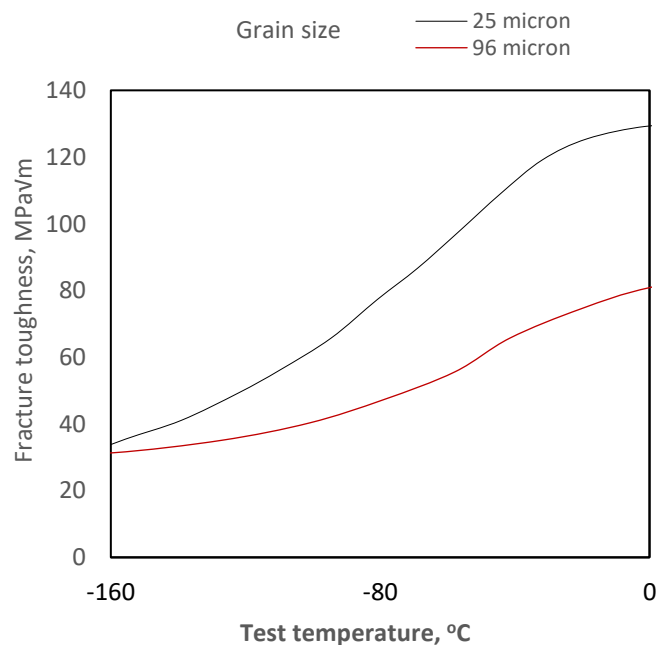


Figure 2-9: Effect of grain size on fracture toughness of St37-3 steel (Adapted from [113]).

The Hall-Petch equation has also been found to apply to other kinds of boundaries, such as ferrite-cementite in pearlite, mechanical twins, and martensite plates. As already noted, a finer grain structure provides higher strength and ductility. This strengthening mechanism is unique, in the sense that other strengthening techniques give increased strength but with a corresponding reduction in toughness. Strengthening by finer grain size is achieved, in part, by the increase in the number of grain-boundary regions as the grain size is reduced. As the grains become finer, the grain boundary area impeding dislocation movement increases.

Grain boundaries are stronger than individual grains at temperatures below an equicohesive temperature. Equicohesive temperature is the temperature below which grain boundaries become stronger than the crystalline grains, or above which grain boundaries become weaker than the grains. Moreover, because a greater number of randomly aligned grains are achieved when grain size is reduced, the stressed material has more opportunity to allow slip and thus improve ductility. In contrast, coarser-grained material tends to exhibit generally poorer mechanical properties, e.g. lower yield strength, higher ductile-brittle transition temperature, and poorer long-life fatigue properties. Thus, as already noted, grain size refinement provides a means for the materials engineer to increase both strength and ductility (toughness) [114]. Hence, strength, including long-life fatigue strength, DBTT, fracture toughness and hardness directly or indirectly depend on grain size, and the dependence is strong enough that commercial mechanical/thermal processing practice is focused on keeping the grain size as small as possible. However, it has been reported that excessive grain refinement to raise yield strength is limited by the onset of grain boundary sliding in the fine-grained microstructure [115].

2.9 Thermomechanical processing of *M-series* steel sub-grades

As already discussed, exceptional strength-toughness combinations can be achieved in even low-alloy steels by refining grain size. TMCP is a manufacturing method through which this kind of grain refinement is achieved. The process is essentially a combination of controlled rolling and controlled cooling operations. **Figure 2-10** presents typical inline methods for cooling hot rolled steel superimposed on a continuous cooling temperature (CCT) diagram. The austenitization is done such that grain growth is suppressed. In the rough rolling stand, recrystallization (strain-free grains with low density of dislocations) takes place and deformed recrystallized grain is obtained in the finishing stand. If the cooling is fast, martensite forms in sufficiently alloyed steel. If moderately cooled, bainite forms, while ferrite and pearlite forms when slow cooling is obtained [37]. The strength of the steel increases from slow cooling to accelerated cooling. The superior point of this process is that high tensile strength can still be obtained with a reduction in the amount of carbon and other alloying elements, as is the case for the M-series steel in **Figure 2-5**.

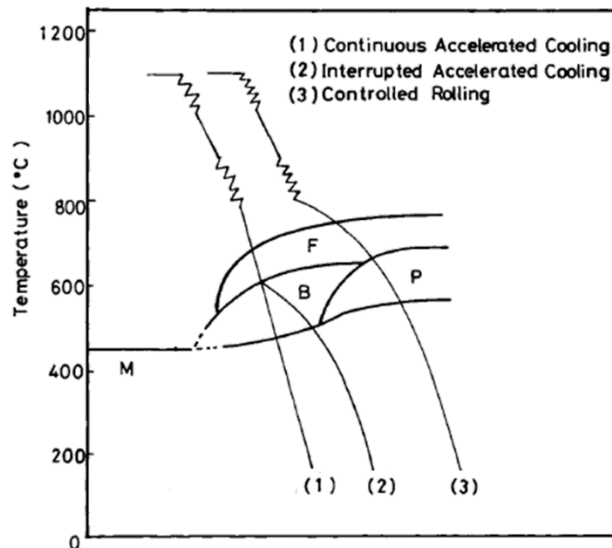


Figure 2-10: Typical methods of inline cooling after hot-rolling for TMCP steel.

The low carbon equivalent (C_{eq}) and fine grains confer two major advantages; very good weldability and improved toughness. The superiority of this process in terms of tensile strength is shown in **Figure 2-11**. TMCP steel gave the highest tensile strength at all carbon equivalents. Hence, the overall essence of thermomechanical processing is to produce the finest possible grain in order to achieve the end results by combining plastic deformation (which is necessary to shape the steel) with controlled heat treatment to give recrystallized grains.

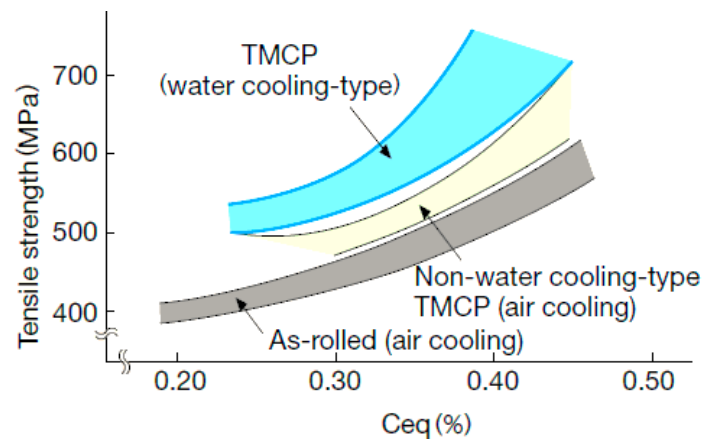


Figure 2-11: Comparison of TMCP and conventional process, in terms of tensile strength and C_{eq} for plate thickness of 20 to 30mm [37].

There are limits to the smallest grain size that can be obtained by applying the theory of phase transformation in steel. Theory has shown that a grain size of allotriomorphic ferrite much smaller than $1\mu\text{m}$ is unlikely in large-scale production processes [116]. In conventional controlled rolling, $5\mu\text{m}$ is reported to be achievable by the transformation of deformed austenite grains to ferrite due to generation of a greater number of nucleation sites in the rolled (pancake-like) austenite. It has been found that the Hall-Petch relation correlates well with experimental results on many metals and alloys, but tends to be invalid for a low strain rate, coarse and very fine grain sizes [117].

A major barrier to the successful implementation of thermomechanical processing on thick plates is that it is difficult to impart large, uniform deformation through the thickness of plate steel [115]. To obtain fine grain size high rolling

deformation (80% or more) is required and since the whole thickness will not be rolled, care must be taken not to assume that uniform property exists within the material. Hence, we may have an inner core that has grain growth which is detrimental to fatigue life – particularly crack initiation.

2.10 Structural metallurgy and integrity of thick TMCP steels for the support structure of XL-WTs

2.10.1 The effect of temperature

Temperature is one of the most important external variables that influence the mechanical behavior of steel. The three basic crystal patterns associated with metals are: (a) the body-centered cubic (*bcc*), (b) the face-centered cubic (*fcc*) and (c) the hexagonal close-packed (*hcp*). Iron or ferrite has a *bcc* structure while austenite is *fcc*. The interstitial solid solution of carbon in γ -Fe is called *austenite* (γ). Temperature dependence of the mechanical properties is found to vary for *bcc* and *fcc* crystal structures of steels. The *fcc* phase (austenite) commonly exhibits plastic deformation and ductile fracture mechanisms, even at low temperature, but *bcc* (ferrite) undergoes cleavage fracture at low temperatures. This temperature effect is commonly illustrated as shown in **Figure 2-12**.

The reduced carbon content in *M*-series steels (**Table 2-5**) will decrease the transition temperature as can be seen in **Figure 2-12 (a)**. Also another advantage of finer grain size is the decrease in the transition temperature [115]. Increase in grain diameter raises the *DBTT* and this trend is usually observed for increases in loading rate [109][114] .

Decreasing grain size and carbon content both decrease the ductile to brittle transition temperature. This is where *M*-series steel sub-grades seems to be superior as they may not suffer from this kind of *CVN* transition shift due to their low alloying contents and finer grain size. However, as the volume fraction of

ferrite (*bcc*) increases due to reduction of carbon, there may be a competing process making ductile to brittle fracture a serious concern.

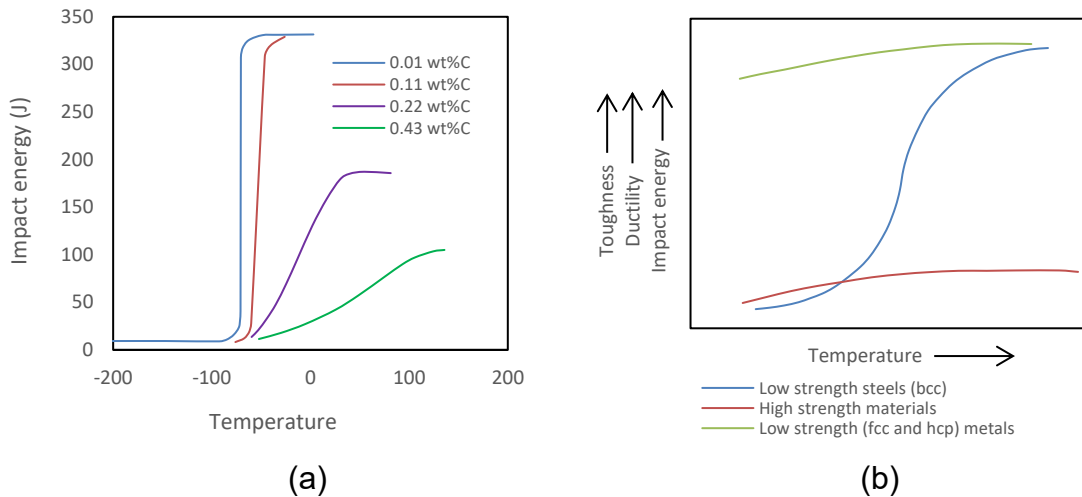


Figure 2-12: (a) Effect of increasing carbon content of steel on the Charpy V-notch energy vs. temperature curve. (b) Effect of temperature on toughness and ductility of *fcc*, *bcc*, and *hcp* metals (Adapted from [118])

2.10.2 The effect of plate thickness

The plate thickness is another important factor in the brittle-ductile transition problem. This effect is commonly associated with the size of the deformation zone in comparison with the plate thickness. It is found that the transition temperature shifts to the left for thin plates. In other words, thick plates are more brittle than thin ones at low temperatures. Plane strain condition associated with an increase in sample thickness imposes a more significant plastic constraint on the plastic zone than does plane stress. The triaxial stress field that exists in thick plates tends to decrease the extent of plastic deformation, thereby increasing local stresses which could lead to brittle failure.

It is practically impossible to obtain uniform fine grains across the thickness of a thick plate (or what the authors may refer to as XL steel plates) due to

recalcsence (liberation of the enthalpy of transformation or latent heat of transformation) [116]. The beneficial mechanical disruption or breakup of non-metallic inclusions and grain refinement during rolling does not occur to the same extent for both the outer regions and inner region of the steel during metalworking. Hence, the fatigue property of thick TMCP steels is not expected to be the same across the thickness of the material – in fact the so-called fine-grained structure may only be obtained at a shallow depth from the metal surface and the interior, which is practically unaffected by the TMCP process, may have significant implications for the design of XL-WTs where an increase in section thickness is used to achieve high stiffness and buckling strengths.

Typical microstructures in a hot rolled, low-alloy steels are shown in **Figure 2-13**. **Figure 2-13 (a & c)** are microstructures of S355J2+N [119], **Figure 2-13 (b)**. S355 (sub-grade unspecified) [120] with composition as shown in **Table 2-7**. **Figure 2-13 (c)** shows the inherent segregation associated with hot rolled ferritic steels. **Figure 2-13 (e & f)** are the microstructures of S355G8+M and S355G10+M respectively.

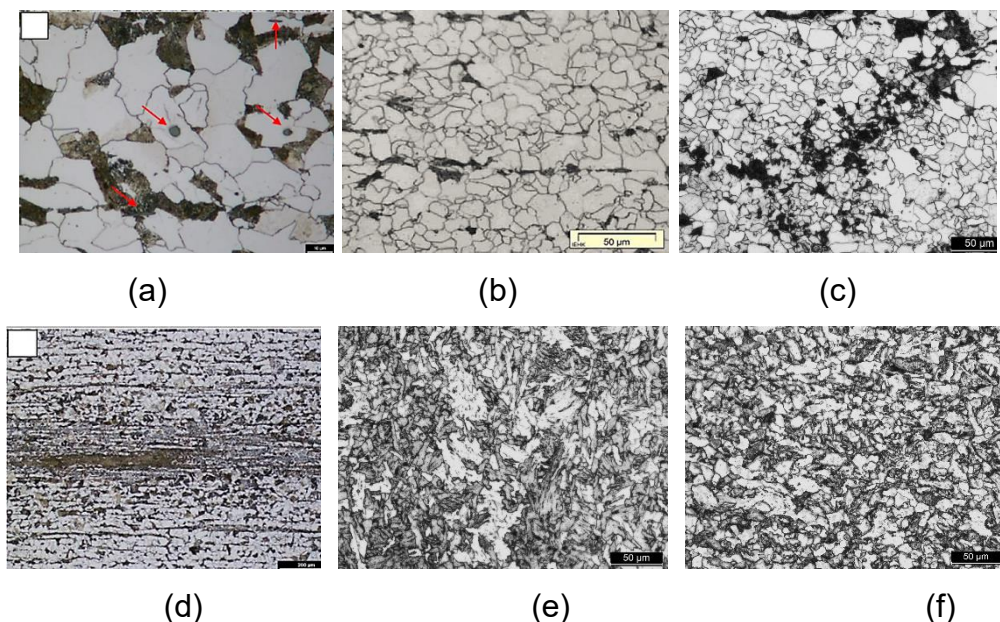


Figure 2-13: (a) Typical microstructures of hot rolled low-alloy steels: (a) S355J2+N [119], (b) S355 (grade unspecified) [120], (c) S355J2+N, (f) segregation

associated with hot rolled S355J2+N steel [119], (e) and (f) are the microstructures of S355G8+M and S355G10+M respectively

Table 2-7: Properties of steels whose microstructures are shown in Figure 2-13.

<i>S355J2+N Chemical composition</i>		<i>Chemical composition of s355 studied</i>	
<i>C</i>	0.22	C	0.054
<i>Si</i>	0.55	SI	0.018
<i>Mn</i>	1.60	MN	1.085
<i>P</i>	0.035	P	0.011
<i>S</i>	0.035	S	<0.001
<i>Cr</i>	0.30	CR	0.015
<i>Mo</i>	0.08	MO	<0.005
<i>Ni</i>	0.30	NI	0.019
<i>Al</i>	0.02	AL	0.021
<i>Cu</i>	0.55	NB	0.023
		CU	0.010
		TI	0.009

A microstructural examination of S355J2+N with properties as shown in **Table 2-7**, shows that 80% of the microstructure is allotriomorphic ferrite and others pearlite, with traces of acicular or Widmanstätten ferrite [119]. Hardness is around 168HV_{0.1} with equiaxed grains of average size of 15µm, measured in the through-thickness direction. The arrows in **Figure 2-13 (a)** are some inclusion such as MnS [120]. **Figure 2-13 (b & c)** confirms that S355 steel is predominantly ferritic. **Figure 2-13 (d)** is the alignment of pearlite in the rolling plane, which is often referred to as microstructural banding. The banding is commonly more evident in sections containing the rolling direction than in those containing the transverse direction [121][122]. Banding can also be attributed to segregation of solutes in the last regions of the steel melt to solidify during the cooling of steel from the molten state. In low-carbon/low-alloy steels prone to banding, solidification starts as δ - ferrite. As the volume fraction of this delta phase increases, solutes such as Mn, Si, P and S partition into the interdendritic regions, which then solidify later with a higher than average concentration of these solutes [123]. Successive

deformation during the hot rolling causes these regions to spread out as bands. The pearlite and ferrite in this segregation zone are usually harder and may extend up to 400 μm [119]. Since other alloying elements are relatively very low in M-series steels, the segregation of concern will be mainly that of Mn – a substitutional element, although other solutes such as Ni, Cr, Si can be found in the pearlite band phase [122]. The presence of unbroken segregation zones in the inner material of thick plates, which increase with plate thickness, may impact negatively on the toughness of the support structure of a WT made with XL steel plates. Micrographs of **Figure 2-13 (c), (e) & (f)** are part of an ongoing project by the authors to ascertain the integrity of S355 thick plate subgrades.

The separation of ferrite and pearlite phases into bands creates local microstructural properties. The presence of a hard phase in a soft matrix is expected to influence fatigue crack growth (*FCG*) behavior in the steel. The difference in local properties due to two different phases with dissimilar stiffnesses would affect the *FCG* behavior of the material. A study found that banding, often seen in hot rolled steel plates, contributes to a superior fatigue strength and higher threshold values [124]. The orientation of the pearlite banding in the studied steel was seen to offer higher resistance to *FCG* [124]. The pearlite band, which is the hard phase, was reported to be responsible for arresting and deviating the crack propagation [124], as shown in **Figure 2-14 (a & b)**.

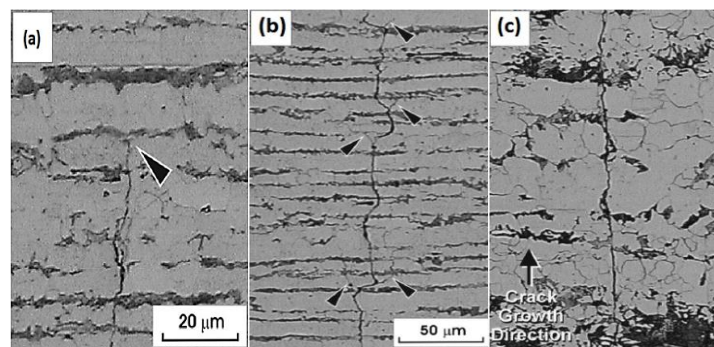


Figure 2-14: (a) Fatigue crack growth path in ferrite-pearlite banding [124]: (a) crack tip arrest by pearlite, (b) crack paths through banded microstructures (L-S) and (c) crack paths through non-banded microstructures

The arrow shows where the crack tip was arrested in **Figure 2-14 (a)**, and the near-threshold region crack branching in **Figure 2-14 (b)**. The crack branching was reported to occur after passage of the crack tip through the pearlite band [124]. It was also reported that the extensive crack branching reduced the fatigue crack propagation driving force at the local crack tip. This effect of banding for hot rolled plate may possibly be seen on the outer region of XL steel plates while the inner material is essentially non-banded. In fact, XL steel plates are not expected to be banded, as can be seen from **Figure 2-13 (c, e & f)**. The microstructures of these ferritic-pearlitic TMCP steels are clearly different and the comparative *FCG* resistance performance study with normalized subgrades have not been carried out or the data are very scarce.

2.10.3 Potential challenges in welding thick plates

As already noted, coarse-grained material tends to exhibit generally poorer mechanical properties, e.g. lower yield strength, higher ductile-brittle transition temperature, and poorer long-life fatigue properties. Decreasing grain size and carbon content both decrease the ductile to brittle transition temperature and increase both the strength and toughness of a steel simultaneously. When these steels are welded, the benefits of the much-celebrated thermo-mechanical processing are lost. Fine-grain pearlite microstructure is tolerant of weld defects. Pearlite is composed largely of the ductile ferrite phase containing about 13% by volume of the hard intermetallic compound, cementite [113]. Fine pearlite is found to offer highest resistance to crack initiation under high cycle fatigue conditions; this finding was attributed to the formation of a more homogeneous distribution of plastic strain in fine pearlite than in a coarse one [125]. The cycling associated with multiple weld passes can cause the formation of coarse pearlite which tends to reduce the fracture toughness of the material.

Residual stress has been defined as that stress which remains in a body that is stationary and at equilibrium with its surroundings [126]. Development of residual

stresses after welding occurs due to localized heating/fusion and non-uniform application of heat energy (sharp thermal gradient). As the weld pool is solidifying it thermally contracts and exerts a pull on the surrounding material and this pull places the region under tensile stress, as shown in **Figure 2-15**. The thermal contraction of the weld metal may be too large to sustain elastically, causing plastic deformation to occur [127]. Having many phases in a microstructure or phase transformation can cause grain scale stresses to exist in a material [126]. If microstructural transformation effects are neglected, significant tensile stresses remain in the near weld region when the welding operation is completed and thermal equilibrium reached by the component.

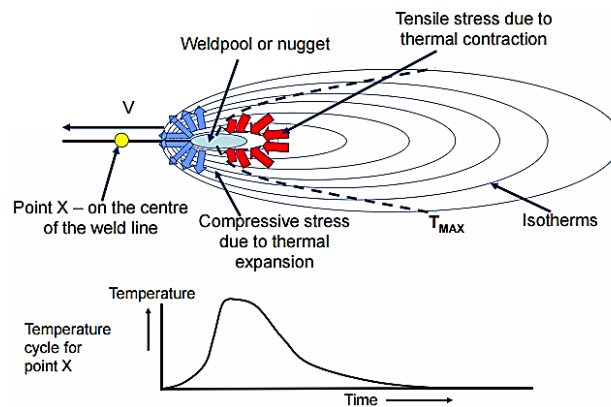


Figure 2-15: Schematic of development of tensile residual stress during fusion welding [128].

Residual stresses are essentially not a part of a material and not needed to maintain equilibrium. They exist as a result of human handling of the material. It has been reported that tensile residual stresses after welding can contribute to fatigue crack development in a structure, even under compressive cyclic loading [129][130]. The tensile residual stress could combine with the service stresses, increasing the mean stress and this seriously reduces the lifespan of structures. Hence, residual tensile stress is very detrimental to the fatigue performance of a steel, or the life of a component.

2.10.4 Change in the weld metallurgy

During fusion welding the heat generated is conducted away by the parent metal. The rate of heat conduction affects the weld metallurgy. **Figure 2-16** is a weld schematic showing various local regions for a single pass weld. The heat affected zone (HAZ) is an area in the material that is not heated high enough to cause melting, but its microstructure is altered by the thermal cycling during welding. It lies immediately after the fusion zone.

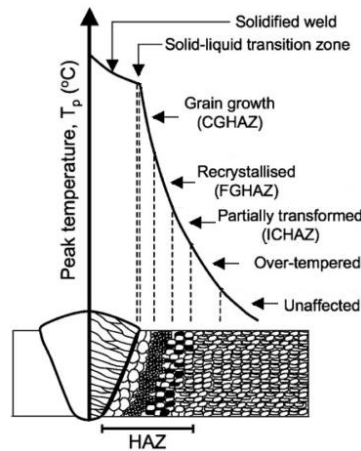


Figure 2-16: Weld metallurgical zones in a single pass weld [123] (see also [131][132])

The HAZ is commonly subdivided into coarse-grained (CG) – region of grain growth; fine-grained (FG) – region of recrystallized grain; inter-critical (IC) – region of partially microstructural transformation; and over-tempered (OT) – region of over tempering on the parent metal. This classification is based on the extent to which grain growth and austenitization occur. To weld thick walled plates, as would be required for modern WT support structures, many welding passes are needed to fill the joint. During the process, the previous weld beads would experience reheating due to deposition of succeeding beads causing the metallurgical zones to be further subdivided.

The CGHAZ zone is associated with reduction in toughness. Reheating due to multi-pass welding at the region of coarse-grained HAZ could make worse the

decline in toughness [133][134]. The constant re-austenitization of certain portions of the weld due to the multiple passes could result in segregation of alloying elements (e.g. C, Mn, Ni, Co,) in steels containing them. The effect of this upon cooling is the formation of local regions that have transformed into hard microstructures, which is detrimental to toughness. These regions are often called local brittle zones.

For XL plates, filler rods with a different composition to that of the parent metal are often used. In a multipass weld, the composition of a previous weld bead can vary from the succeeding weld bead as a result of changes in dilution levels. The resultant weld microstructure is a function of the degree of dilution, which depends on the extent of variation in composition between the filler metal and parent metal. Weld dilution has been found to be detrimental to weld toughness [46][135]. The microstructure at the surface of the structural material influences considerably the fatigue properties [117] and in a corrosive environment this surface is under attack.

Failure during WT service is accompanied by huge economic cost. The need to select the right material cannot be overemphasized in order to reduce capital expenditure (CAPEX) and in-service fatigue failure. Thus, a reliable fatigue study of all sub-grades of S355 – normalized and TMCP steel – in a marine environment must consider the effect of plate thickness, frequency, waveform and corrosion. Hence, research is currently ongoing to provide the fatigue crack growth rates, guidance on inspection requirements and prediction of corrosion-fatigue life of offshore structures designed with ferritic steels, with the focus on comparing normalized and TMCP S355 steels.

2.11 Conclusions

Some of the findings and structural integrity issues raised in this paper can be summarized thus:

1. All indications show that the global energy from wind power will rise due to rapid reduction in the cost of wind energy. With advances in turbine technology, efficient and optimized turbine support structure the outlook shows that wind energy will become the major and critical renewable source of electricity in the UK by end of 2030.
2. TMCP steels are reported to be well suited for XL-WTs. The reduced alloying-elements in this steel will give the steel high weldability and this property will be good for thick sections in XL-WT design. The reduced alloying elements in these steels will tend to reduce the strength of the steel due to reduction in the strength contributions from solid solution. M-series are supposed to be hot-worked in an austenite range and essentially there is little or no strain-hardening strengthening contribution in hot-worked steel. Hence, grain refinement remains the best strengthening method for this steel to attain high mechanical strength and toughness. In general, the low carbon equivalent (C_{eq}) and fine grains confer two major advantages; very good weldability and improved toughness.
3. A major barrier to the successful implementation of thermomechanical processing on thick plates is that it is difficult to obtain large, uniform deformation through the thickness of plate steel. To obtain fine grain size, high rolling deformation (80% or more) is required. Thus, for M-series only a small zone is probably rolled and care must be taken not to assume that uniform property exists within the material. It is likely we will have an inner core that has grain growth which reduces fatigue life. The advantage of this M-series sub-grade of S355 and that of normalized S355 of similar composition in terms of corrosion-fatigue needs to be critically assessed to aid economic material selection decisions and use of steels for various engineering applications in marine environment.
4. Decreasing grain size and carbon content both decrease the ductile to brittle transition temperature. However, as the volume fraction of ferrite (*bcc*) increases due to the reduction of carbon, there may be a competing process making ductile to brittle fracture a serious concern.

5. Increasing the thickness of the steel section eliminates elastic buckling tendencies. Increasing the thickness this way to check buckling of XL monopiles has a severe penalty that must be paid in terms of structural integrity of the thick steel section.
6. Plane strain condition, associated with an increase in sample thickness, imposes significant plastic constraint on the plastic zone than does plane stress. The triaxial stress field that exists in thick plates tends to decrease the extent of plastic deformation, thereby increasing local stresses which could lead to brittle failure. Hence, thick plates are more brittle than thin plates.
7. It is practically impossible to obtain uniform fine grains across the thickness of XL steel plate due to recalcification. The beneficial mechanical breakup of non-metallic inclusions and grain refinement during rolling for XL plates merely occurs at a small layer on the metal surface. Consequently, the fatigue property of TMCP steels is not the same across the thickness of the material. The interior, which is practically unaffected by the TMCP process, may have experienced grain growth and segregations and this could have a negative impact on the toughness of the support structure of XL-WTs, where an increase in section thickness is used to achieve high stiffness and buckling strengths.
8. The design S-N curve, such as given below, shows that as thickness increases the number of cycles to failure decreases exponentially if other variables remain constant.

$$\log N = \log \bar{a} - m \log \left[\Delta \sigma \left(\frac{t}{t_{ref}} \right)^k \right] \quad (2-6)$$

It is likely that the large scale of defects, inclusions, grain growth and segregation (brittle zones) accompanying an increase in thickness could accelerate crack propagation in the inner material of XL-WT support structures.

9. When these steels are welded, the benefits of much celebrated thermo-mechanical processing are lost. Fine pearlite is found to offer highest resistance to crack initiation under high cycle fatigue conditions. The cycling associated with multiple weld passes can cause the formation of coarse pearlite, which is detrimental to fracture toughness.
10. The development of residual stresses after welding occurs due to localized heating/fusion and non-uniform application of heat energy. To weld thick walled plates, as would be required for modern turbine towers, many welding passes are needed to fill the joint. If microstructural transformation effects are neglected, large tensile stresses could remain in the near weld region. The tensile residual stress is very detrimental to the fatigue performance of a steel. Tensile residual stress is reported to contribute to fatigue crack development in a structure, even under compressive cyclic loading.
11. Weld microstructure is a function of the extent of dilution by the filler metal. Weld dilution could be a problem with thick plates and is found to be detrimental to weld toughness.
12. The knowledge of temperature and corrosion effects, fatigue crack growth rate and magnitude of residual stress will enable cost effective design of next generation OWTs.

Finally, apart from consideration of the enumerated design factors, next generation OWT support structures should consider achieving buckling strength and required fatigue life through innovative increases in diameter, rather than thickness, in order to avoid some of these structural integrity concerns.

3 Literature Review – part two: Fatigue loading and corrosion-fatigue in marine steel

Corrosion-fatigue is very important to all offshore industries that use steel in marine environment. This is because fatigue is still the most common mode of failure in mechanical engineering components and structures and can lead to major failures, with severe consequences to plant, equipment, the environment and human lives. The number of cycles needed to propagate a crack to failure under given stress level and environmental conditions forms a major fraction for life assessment of structures. For LEFM, in fact, it is the only criterion to assess fatigue damage as all real engineering structures regardless of how they are made contain pre-existing cracks. No material on earth exists without a defect!

Steel structures in dynamic service environment degrade with time due to fatigue. Offshore engineering structures, such as wind turbines (WTs) are subjected to cyclic loading conditions, due to constant exertion of wind, wave and current forces. The initiation and propagation of cracks in these structures could lead to catastrophic failure at loads far less than the yield stress of the material. The fatigue crack initiation and fatigue crack growth (*FCG*) are accelerated by the *SW* environment. That is, the combination of a corrosive environment and fatigue phenomenon is found to be more detrimental than each of them acting separately. This simultaneous action of cyclic load and chemical attack is commonly referred to as corrosion-fatigue (CF) and it is found to be an important failure mechanism in offshore structures [136][137]. The structural integrity concern for such structures is to ensure that they carry the applied loads during operation without any macro-scale failure throughout the life span. For this reason, it becomes a common engineering practice to measure the tendency of cracks to propagate under a given stress condition.

Corrosion is both time and temperature dependent [138][139]. Two factors that are known to control the extent of corrosion damage contribution to the FCG

behaviour of steel are the rate at which the corrosion is occurring and the time available for the corrosion damage to occur per cycle. In a corrosion-fatigue study [24], these two factors have been used to explain the acceleration of crack propagation rates with decreasing test frequency under constant temperature.

There exist a huge number of publications spanning from the time of Head, Frost, Dugdale, McEvily, Griffith, Irwin, Paris, Barsom, Newman, etc. to the present. The reason in one hand is due to critical role fatigue plays in the failure of engineering structures, especially those in marine environment. As new materials evolve - for example the emergence of modern TMCP steels and additively manufactured parts, their fatigue responses need to be evaluated and this assessment is a continuous one - from the onset of a project to its decommissioning. It makes fatigue study a dynamic and case sensitive undertaken. Hence, the study of fatigue crack growth rate (FCGR) is essentially important in both life prediction of engineering structures and in the development of new materials. A detailed review of modern material requirements for fatigue resistance improvement in large wind turbine support structures (WTSSs) can be found elsewhere [140].

Engineering materials often contain macrocracks, microcracks, or crack-like defects when manufactured and during service. Defect are likely to arise from welds, inclusions in the structural steel, surface damage, machining flaws, defects in manufactured steel like casting pores, solidification cracking and crack initiated during steel rolling, cyclic loads (fatigue) or corrosive attack during service. The size of these pre-existing defects may be very small making them difficult to be detected using existing NDT method. For example, Particles such as precipitates possessing an unfavourable geometry (e.g., being plate-like or sharp-edged) can initiate cracks in the matrix of steel structures. Crack forms easily by detaching from the matrix or by breaking [141]. The number of cycles required to grow these pre-existing defects from their initial size to a critical crack size above which unstable fast fracture occurs is often taken to be the fatigue life of the material.

Fatigue-crack-growth data are used in the fracture-mechanics approach for fatigue analysis. This approach can be employed either (1) as a *life analysis* or (2) as a *damage tolerance analysis*. In *life analysis*, the structure is assumed to contain defects from the start of its life and the severity of these defects is represented by an equivalent crack-like initial flaw. The growth of this equivalent flaw to failure then determines the life of the structure. It is important to realize that this equivalent flaw may not be physically realistic; it is just the initial flaw that gives the appropriate life. If the equivalent flaw size is unique for a given structure and load histories of interest, then this concept is useful. For generalized variable-amplitude loading, this approach seems to work fairly well, but the equivalent initial size is sometimes a function of loading, i.e., it is nonunique.

Damage-Tolerant Design is based on the assumption that finite cracks will exist in an engineering structure. Fracture mechanics are employed to determine whether the cracks will grow large enough to cause failure, between regular inspections. This method is applied to materials with high fracture toughness, slow crack growth and when reliable non-destructive evaluation methods are available. The damage-tolerance concept recognizes that at some time cracks may develop in a structure, and the growth of these cracks is then computed. Using a risk analysis, a critical strength and the associated critical crack size are calculated. The time for a crack to grow from some detectable size to the critical size is determined next. An inspection interval shorter than this time period can be established so that damage can be detected well before failure occurs.

Thus, the damage-tolerance approach assumes that if cracks exist or develop, they will be detected and repaired before failure can take place. This philosophy inherently requires highly reliable methods for inspecting and detecting cracks in the structure. With either the life analysis or the damage-tolerance analysis, the growth of a crack is computed using fracture mechanics and experimentally measured fatigue-crack-growth data. In aggressive aqueous environments, the

crack growth rate normally is accelerated compared with that in less aggressive environments such as air. In other words, the effect of aggressive environment generally is to accelerate the crack growth rate. In some cases, the environmental attack at or near the crack tip is so high that the crack growth rate is actually decreased because of crack-tip blunting by the corrosive action. Since this retardation of crack-growth-rate effect is conservative from a design viewpoint, it is not of as much concern as the more commonly observed acceleration effect.

3.1 Fatigue loading on offshore WTSS

Marine structures are subjected to diverse loading conditions – that are dynamic and cyclic in nature. For offshore OWTs the external environmental forces - e.g. wind, wave and normal service operation, result in the production of the desired energy and at the same time the forces cause development of stresses in the WT support structures. Forces due to mass, installation and maintenance activities also act on the structure. The magnitude of these cyclic stresses directly determines the functionality and lifespan of the WT, thus presents a major fatigue problem.

Figure 3-1(a) shows that forces acting on a typical WT support structure are axial loads, lateral loads and bending moments. Torsional loading is sometimes considered too [13]. The bending moments vary cyclically due to operational rotation of the blades and the magnitude increases with the rotor diameter [22]. Among these forces, dead loads, the meteorological and oceanographic (MetOcean) conditions are considered to be the most significant loads acting on the structure [88]. Dead loads are self-weight of: the pile, the transition piece - if used, the tower and the turbine (rotor – hub and blade, and nacelle). MetOcean loads are wind, waves and current. The thrust force, P_1 in **Figure 3-1(b)** is the lateral wind force acting on the rotor (often taken to act at the hub) of the turbine from the rotating blades, produced by the turbulence in the wind. The magnitude of dynamic component depends on the turbulent wind speed [21].

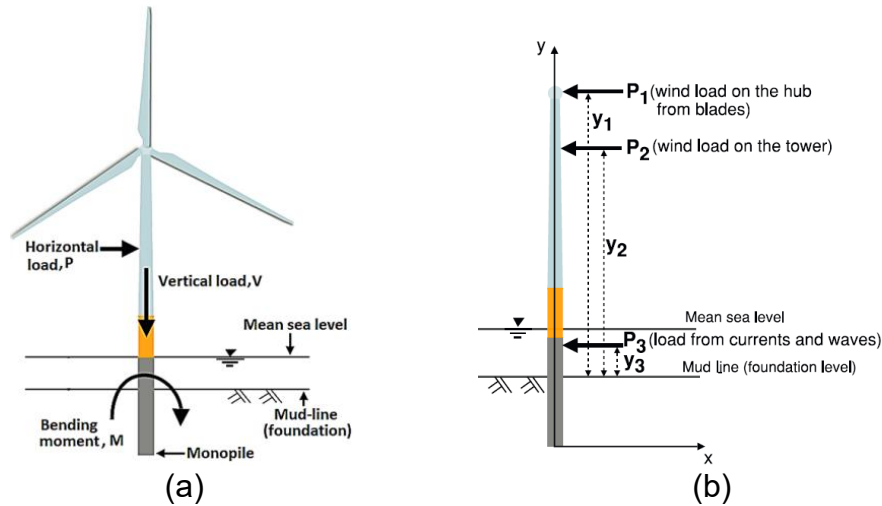


Figure 3-1 Fatigue loads on WT (a) Nature of loads on a monopile supported WT [27] (b) position of the thrust force P_1 on the hub of the rotor [23].

Also shown is wind load on the tower P_2 , wave and current loads P_3 on the substructure as illustrated by [23]. The magnitude of the waves load hitting the substructure very close to the foundation depends on the wave height and wave period. If the mudline is taken to be rigidly fixed, then the highest moment (M) acting on the support structure about the mudline becomes $M = P_1 \cdot Y_1$.

Wind force or moment due to wind loading on the tower structure is small compared with the magnitude of the thrust forces acting on the rotor of the turbine during power production. Hence, wind load on the tower structure is not considered to be a critical component of the fatigue design. Thus, the predominant external fatigue problem on OWTs comes from wind loads that has longer moment lever arm (**Figure 3-1(b)**), acting on the hub and impacting the greatest bending moment on the support structure during normal operation [22].

3.2 The operational frequency of a WT

The natural frequency of a WT is the frequency with which the system oscillates during normal operating condition. The assessment of the first natural frequency of the offshore WT and frequency distribution (or the eigenfrequencies – range of frequencies with which the WT can vibrate in) is a very important part of structural

analysis. The frequencies of the forces that are commonly taken into considerations are that of the wind and wave spectra, and operational intervals of the rotor. The primary excitation frequency due to the rotational speed of the rotor is commonly called the $1P$. For modern OWTs, it is reported that the rotor speed under normal operation is typically about 7–12 revolutions per minute (rpm), which corresponds to $1P$ frequency of 0.12 - 0.20Hz [142]. The second excitation frequency of OWT occurs due to creation of wind shade by the WT tower. Anytime a single blade of the rotor passes across the wind shade of the vertical tower there is reduction of the wind loading on the tower causing additional cyclic load on the support structure at the frequency equal to $1P$ times the number of blades. Thus, $3P$ is the blade passing frequency for a 3-bladed WT. This implies that if the rotor completes 6 rpm, the $1P$ frequency is 0.1Hz while the $3P$ is 0.3Hz.

For wave loading, which is often produced by wind, the range of possible frequencies typical of North Sea offshore waves is given in the range 0.04 and 0.10 Hz [27]. Many of the studies on CFCGR in marine environment used a frequency of 0.1Hz and temperature range of about 4 - 10 °C. The intention was to simulate offshore conditions and obtain fatigue data for application in offshore Oil & Gas production and exploration platforms. This would then enable extrapolation of the laboratory data to field structures. For the gust wind, using Davenport frequency spectrum model, the dominant frequency of a gusty wind is reported to be about 0.02 Hz [143], lower than that of the waves.

Generally, in designing commercial OWTs, frequency of the support structure is such that it lies between $1P$ and $3P$ [144]. This frequency condition of the WT is referred to as soft-stiff design [23]. This implies that fatigue experiments of practical significance to many of the commercial WTs in soft-stiff condition should be carried out in the range 0.16 – 0.54Hz [144]. Thus, the frequencies used in generation of experimental data in this study are 0.2Hz, 0.3Hz and 0.5Hz under stress ratio of 0.1.

3.3 The load frequency waveform of a WT

Wind speed which impacts the highest bending moment on WT structure is highly variable. **Figure 3-2** shows the inherent fluctuations in wind speed on two different locations. It also shows that the load variation on WT is not symmetrical about any reference line and has no deterministic form. **Figure 3-3** shows the wind shear and wind induced wave profile along the support structure of an OWT. The wind speed changes with time for a given height, also with change in height above the sea level or ground surface as shown in **Figure 3-3 (a)** [23]. **Figure 3-3 (b)** shows how the wind changes with height, towards maximum value at the hub in 3D space.

The variation in wind speed is obviously a random or stochastic process and the stochastic nature of fatigue loading on WT structures is not easily defined [13]. Stochastic or random loads cannot be predicted to occur with a certain magnitude at a given time. Wind speed variations on a small timescales of minutes down to seconds or less is known as turbulence [145]. Wind turbulence has very significant impact on the design and performance of the individual WTs [145]. Knowledge of wind turbulence is another aspect that helps achieve realistic data and modelling of fatigue response of WTs.

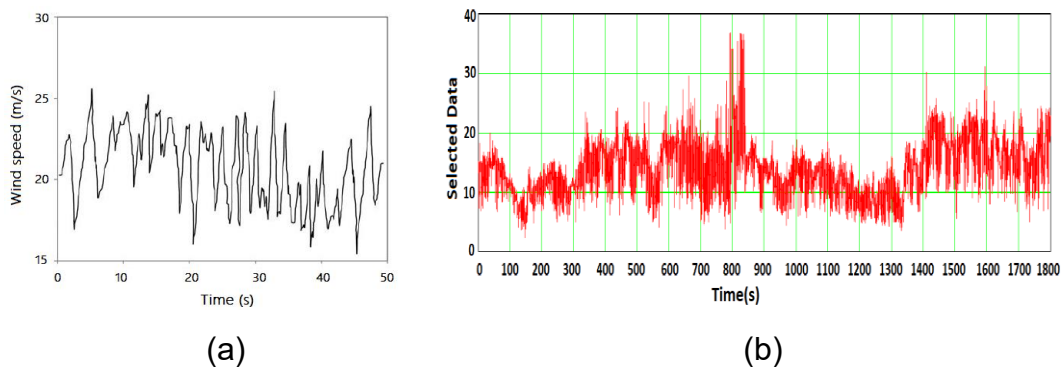


Figure 3-2 Variation of wind speed in time domain at two different sites (a) 0.83-minute wind speed record (b) 30-minute wind speed record [146][147]

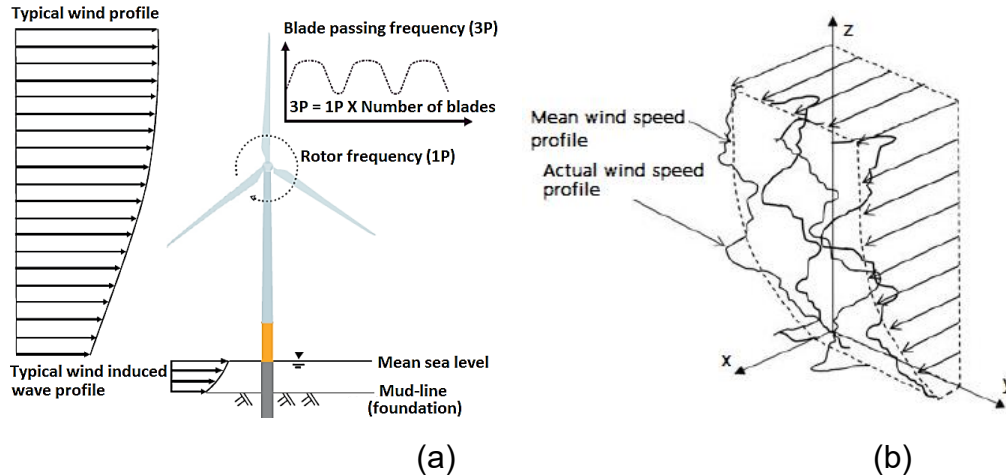


Figure 3-3 Wind shear and wind induced wave profile (a) The main external dynamic and cyclic loading acting on a typical OWT [23] (b) variation of wind speed in 3D space [146].

In reality, we can say that the wind load as a stochastic phenomenon can take at instance any of the waveforms presented in **Figure 3-4** [148] and combination of these waveforms builds up the wind load spectrum. **Figure 3-4** shows various cyclic and hold-time histories or shape of load waveform, that could build up random loading waveform of a wind on a WT. In many fatigue tests, the waveform that is usually employed is sinusoidal wave of constant cyclic loading amplitude [149] and a constant mean load. From above discussion, it can be seen that this condition does not represent the service conditions for WTs. Every structural component or mechanical process has its own specific loading sequence. This makes it difficult to provide information on component's fatigue behaviour for every different variation and combination of load sequence. This has been the reason for the use of the constant amplitude stress range in fatigue design and modelling practical load histories sometimes as constant amplitude.

Fatigue of materials is a consequence of a time-dependent stress–strain loading spectrum. Degradation of material by corrosion is generally known to depend on the corrosion rate and the time available for the corrosion to occur. In fatigue

process where load frequency has effect, it should be expected that the wave shape of the load cycle can also influence crack growth in corrosive environment.

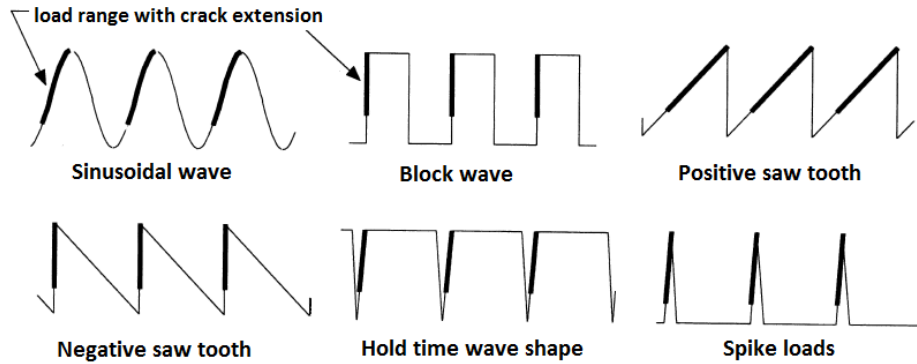


Figure 3-4 Various cyclic and hold-time histories

3.4 Previous studies on waveform effects on corrosion-fatigue

Corrosion is a time dependent material degradation mechanism. In corrosion-fatigue process, the loading frequency and the wave shape or form of the load cycle are expected to have some potential effects on the crack growth behaviour of steel in corrosive environment. Some studies have shown that variation in *CFCGR* exists for different waveforms. Atkinson, et al. [150] studied *FCGR* in Fe-C-Si-Mn structural steel designated BSI501-213-32A under triangular waveform in air, distilled water and natural lake water at temperatures between 25 and 30°C and pH 6.0. They concluded that *FCGR* in aqueous environment was 7 times higher than that obtained in air and that this increment is dependent on the frequency. The magnitude of environmental effect was found to depend on rates of loading. They also observed that negative sawtooth and square wave, which have very rapid rise-time (period of increasing tensile load) of the fatigue cycle, did not increase the *CFCGR* at frequencies as low as 0.01Hz. Reported also is that holding at maximum load in the *SW* for up to 10 minutes did not affect the *FCGR* at room temperature and at high cyclic frequencies the *FCGR* in the *SW* was similar to that in air.

Nakasa, et al. [151] studied CFCGR in JIS SNCM439 steel quenched and tempered at 200 or 500°C. He reported that CFCGR was higher in square load waveform than positive sawtooth for steel with high sensitivity to delayed failure. For the sample tempered at 500°C, which they reported has delayed failure under static stress, the CFCGR was higher under positive sawtooth compared with square (block) waveform. They suggested that stress rise time and hydrogen concentration control the crack growth in corrosion-fatigue. This implies that the interaction between the crack tip and hydrogen atoms controlled the crack growth mechanism.

Bhuyan, et al. [152] studied the effect of waveform on CFCGR in a steel manufactured to CSA G40.21M 350 WT specification in air and *SW* under constant amplitude loading. The pH of the *SW* was in the range 7.25-7.55 while the temperature was in the range 0°C - 21°C in *SW* and -15 to 4°C for the tests in air. The tests were carried out using sinusoidal and sawtooth waveforms. They reported that CFCGR in *SW* was on average 2.7 times higher than that in air for the stress ratio (*R*) of 0.1, frequency of 0.05Hz and in temperature range of 0 to 4°C. They also reported that the CFCGR became almost similar to the value in air as the frequency was increased from 0.05Hz to 0.5Hz at load ratio of 0.1. They also found that CFCGR increased by a factor of 2.3 as frequency was decreased from 0.5 to 0.05Hz. For the air test, the FCGR was almost unaffected in the temperature range of -15 to 4°C and was generally not influenced by the load ratio variation in the range of 0.05 – 0.3.

Achilles, et al. [153] investigated the effect of sinusoidal and positive sawtooth on the FCGR of a forged steel designated as SA508 cl III RPV. The frequencies used were 1Hz and 10Hz and *R* = 0.2. The test environment was laboratory air, PWR (Pressurized water reactor) water and high purity argon. They found that FCGR under positive sawtooth was always higher than sinewave both under 10Hz and 1Hz, and the magnitude is higher for 1Hz than 10Hz. In argon, the

FCGR was slightly higher under 1Hz by a factor of 1.3. In air, at 10Hz, they found that FCGR under the positive sawtooth was 2 times higher than the sinewave, while this value of 2 reduced to 1.5 under 1Hz. The reason for this discrepancy was not given. In PWR water, the waveform effect is more pronounced under 10Hz and sawtooth is higher by a factor of 1.5. Thus, the magnitude of difference depended on the environment and frequency. However, this finding appears to contradict the common observation that corrosion process increases the FCGR with decrease in frequency.

Another important study on FCGR in air and *SW* on hot-rolled and normalized BS4360 Grade 50D (equivalent S355J2+N) steel was done by Scott, et al. [154]. This work has been heavily cited in the literatures. The plate was hot-rolled and normalized. The test was performed in laboratory air condition and in *SW* under free corroding condition using sine, triangular, positive sawtooth, negative sawtooth and square waveforms. The work shows that the cyclic frequency and R ratio (in the range 1.0 – 0.85) had little or no influence on the FCGR in air. 2-stage Paris-law trends was used to describe the FCGR of the whole test in air.

For *SW* environment, they used constant ΔK approach, which involves load reduction technique as the crack grows. They reported that about or above 1Hz, the FCGR was similar to that observed in air for all the waveforms. They found that CFCGR under sinusoidal, triangular and positive sawtooth waveforms were similar and the values were very much higher than square waveform. They also reported that the CFCGR from sinusoidal wave was slightly higher than triangular waveform.

Under cathodic protection at free corroding potential (- 0.65V); - 0.85 and - 1.10 V (Ag/AgCl), they reported that the CFCGR was similar again for sinewave, triangular and positive sawtooth, (particularly in the Paris Region) and the value was higher than that of negative sawtooth. Consequently, they concluded that *SW* only affected the CFCGR during the tensile load rise time. In order words,

the electrochemical reactions contributed to the increase in the crack growth only during the tensile load rise time. However, they pointed out that this conclusion was not enough to explain all the behavioural details of the crack growth. To investigate this result further, the present authors re-plotted their CFCGR data and this is shown in **Figure 3-5**.

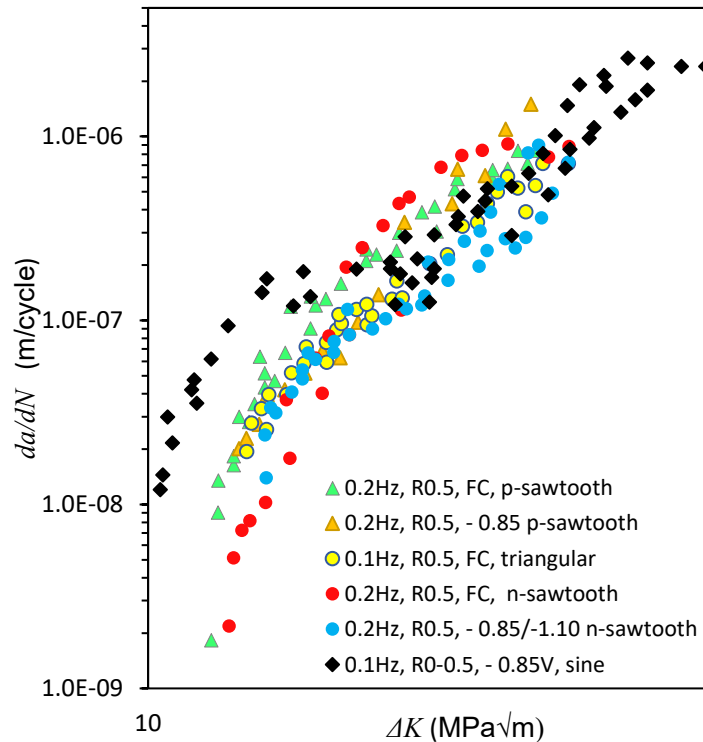


Figure 3-5 Scott, et al. [154] result on the study of BS4360 Grade 50D (equivalent S355J2+N) steel in seawater (p = positive, n = negative)

For the cathodically protected at -0.85V, the values of the negative and positive sawtooth are almost the same below SIFR of 20 MPa√m. The positive sawtooth became higher than negative sawtooth only above 20 MPa√m, which implies that the effect of rise time only happens at high SIFR. It can also be observed that the negative sawtooth in free corrosion (FC) is lower than other waveforms only below about 18MPa√m and above other waveforms beyond this SIFR point. Thus, their conclusion is somewhat flawed. Some other important observations in this study is that the curve of the positive sawtooth under free corroding is

higher than that of triangular waveform under same condition, even though the frequency of the triangular waveform is lower. This may imply that reduction in frequency may not necessarily lead to increase in growth rate. Also, under sinewave, R in the range 0.1 – 0.7 have little or no effect on the CFCGR at cathodic potential of - 0.85, in the temperature range of 5 -10 °C, and frequency 0.1Hz.

In a similar study conducted by Thorpe, et al. [24], the R and frequency were found to have no effect on FCGR in air. The result again shows that similar results were obtained from the test with triangular, positive sawtooth and sinusoidal waveforms. They also showed that in the corrosion-fatigue tests the CFCGR increased as test frequency is decreased, and R increased.

Jones [155] [156] studied FCGR in air and in *SW* of a naval steel and trapezoidal waveform was used. The test cycle period was 2 minutes or a frequency of 0.0083Hz. The R was 0 and pH 7.7- 8.1%. The rise time and time of descent was about 15sec (which was reported to vary from 10 to 22 seconds). This implied that the hold time at maximum and minimum load was about 45 seconds each. The medium was an aerated natural *SW* that was circulated through a chamber at temperature range of 16 - 21 °C. and the test was done under constant load and FCGR obtained using secant method. It is pertinent to note that the method of calculation of stress intensity did not conform to ASTM E647. The result showed that the CFCGR of trapezoidal waveform in *SW* was higher than air test by an average factor of about 5 from the upper scatter bound. This very high rate was not conclusively explained.

Appleton [157] studied the CFCGR in BS4360 grade 50D steel in 3% NaCl at room temperature and found that there was no difference in the CFCGR for the tests done at 0.05Hz, 0.17Hz and 1Hz up to about ΔK of 18MPa $\sqrt{m}$. However, above 18MPa $\sqrt{m}$, CFCGR at 0.17Hz increased steadily above that of 1Hz. A large decrease in the CFCGR was seen at the test done at 5Hz. In general, the

effect of SW started decreasing steadily from 1Hz upwards towards that of air test value. Moreover, the results showed that for 0.17 Hz tests, there was no significant difference in the CFCGR curve in the R range of 0.1 - 0.7. Appleton [157] also performed tests at other different waveforms. The additional waveforms were triangular and square shape at 0.17 Hz. No difference in CFCGR was observed between sinusoidal and triangular waveforms.

Tanaka, K and Wei, R. P. [158] investigated the difference in CFCGR of short and long cracks in HY130 steel using four-point bending specimen. The environments of study were air and 3%NaCl solution. Sinewave was used and frequency of 10Hz was applied in air and 0.1Hz in 3%NaCl solution. The R was 0.1 and in some others 0.45. The result from this study for long cracks (8.9-15.2 mm) shows that below ΔK of $15\text{MPa}\sqrt{\text{m}}$, the CFCGR was similar to that of air test. Above $15\text{MPa}\sqrt{\text{m}}$ the FGCR increased in the 3%NaCl solution by average factor of about 2 with respect to the air test throughout the Paris Region. Also observed is that the test conducted under the same condition, but at R = 0.45 was slightly higher by a factor no more than 1.46 and at $35\text{MPa}\sqrt{\text{m}}$ it tends to similar result with that of the test done at R = 0.1. There was no difference between crack growth rate for short (0.4 to 2.5 mm) and long crack in air. In the NaCl medium, CFCGR of short crack was reported to be higher than that of long crack by a factor of 2. This difference is said to disappear when the short crack grows to about 1.1mm. Note that most of the tests were completed within a day because of use of high stress ranges ($\Delta\sigma = 517$ and 827 MPa) in the experiments and this may not represent adequately the effect of the corrosive medium.

Cheng, Y. W. [159] studied the FCGR in ABS grade EH36 steel in 3.5%NaCl under constant load-amplitude sinewave and spectrum loading. For the sinewave test, the R of 0.1 or 0.5, frequency of 10Hz in air and 0.1Hz in the 3.5%NaCl were used. Two specimen orientations LT and TL were compared in the study. The result shows no difference in FCGR for the two orientations both in air and in the simulated SW. The R had a small influence on the FCGR both in air and in the

corrosive medium. Very small difference can be seen for both conditions as the SIFR decreases. This appears to agree with the finding of [160] that specimen orientation and R have significant effect only at high and low growth rates. Below about $16\text{MPa}\sqrt{\text{m}}$, the FCGR in air and in the 3.5%NaCl were almost similar. Above this SIFR, the FCGR in the 3.5%NaCl was higher than that of air by an average factor of 3.1 in the Paris Region. Using equivalent stress range approach, Cheng found that FCGR under constant load-amplitude and spectrum in the simulated SW were similar.

Barsom J. M. [161] studied waveform effect on the FCGR in a 12Ni-5Cr-3Mo maraging steel and the test was done in air and 3%NaCl and he compared the effect of sine, triangular, and square waveforms loadings. Frequency in the range 0.1 - 10Hz was used. The result showed that similar FCGR results were obtained in air for the three waveforms. In other words, waveform has no effect on crack growth in the air. In the 3%NaCl, the CFCGRs of the square waveform and negative sawtooth were the same and slightly higher than the air data, by a factor of 1.1. For the sinusoidal, triangular and positive-sawtooth waveforms the CFCGRs were similar and higher than FCGR in air by a factor of 3. Barsom, J. M. [162] also showed that there was no significant difference in FCGR for test under random-sequence and ordered sequence load fluctuations. Note that the corrosive environment increased the FCGR by about the same factor for the waveforms. Due to this similar factor and the difference between sine-triangular-positive sawtooth and square waveform, it is often concluded that the increase in CFCGR takes place only during the rise time of the tensile loading of the applied waveform and not during holding time at maximum stress in the cycle period below K_{ISCC} . In other words, corrosive medium enhances FCG when the tensile stress around the crack tip is increasing and not when it is steady or decreasing. However, this conclusion is dependent on the material-environment system.

Knop, M., et al. [163] studied the effect of cycle frequency rise-time on CFCGR in a high-strength steels. They also presented the effects of fall-times and hold

times at maximum and minimum loads on the CFCGRs. They informed that FCG at the Paris Region is generally by plastic blunting and that the crack length (a) increased during the rising of the load towards the cycle peak for each load cycle. They reported that the difference between crack growth in inert and embrittling environments was that less crack tip blunting occurred in the embrittling environment due to localisation of deformation slip and this resulted to increased crack growth rate. They suggested that CFCGR depended on the concentration of hydrogen adsorbed at crack tips and at tips of nanovoids ahead of the cracks tip. They concluded that crack growth is enhanced under increasing rise time and that longer fall times and longer hold time at maximum and minimum loads may enhance crack growth when the rise time is short. In their study, the interval of increased effect of rise time was from 1 to about 10 s and the growth rate within this range depended on the SIFR.

In his report, Austen, I. M. [164] showed that for En24 steel in distilled water, under positive sawtooth waveform at $R = 0.25$, the CFCGR increased with increase in tensile load rise time (from 0.1s to 10 s). For the test under square waveform at $R = 0.25$, the result showed that CFCGR increased gradually with increase in hold time from 0.01 – 1 s. The effect is more marked at lower ΔK . In the distilled water and at a frequency of 5Hz, $R = 0.25$ and under different waveforms, the CFCGR in the En24 steel increased thus; positive sawtooth>triangular>square>negative sawtooth. The rise time for the p-s was 0.2s while that of Tr was 0.1s. It is pertinent to note that the frequencies for these tests are not the same due to different hold times, hence the independent effect of frequency reduction or hold time increase on the waveforms cannot be delineated in this study. They also showed that for 300m steel in distilled water at frequency of 0.45Hz and $R = 0.5$, the CFCGR under sinewave is higher than triangular waveform and the lowest was the square waveform. Note that the results of Knop, M., et al. [163] and Austen, I. M. [164] showed that increase in rise time had the most corrosion-fatigue damaging effect.

Kawai, S., et al. [165] in their study of FCGR in SNCM2 (BS830M31) steel in 3% NaCl solution reported that increase in the stress rise time increased CFCGR to a maximum enhancement factor of 3 over the same test done in air. This constant value was attained after 10s. This strong effect was attributed to corrosive dissolution of fresh exposed surfaces formed during the rise time. In their study, increase in rise time increased the CFCGR. However, at low ΔK , the crack growth rate decreased when the stress rise time increased beyond 1s. They also reported that stress decreasing time and holding at both maximum and minimum stress decreased the CFCGR. The extent of decrease was said to depend on the ΔK , rise time and period of hold time (or Stress decreasing time, minimum stress holding time). Note that in this regard, Knop, M., et al. [163] opined that the long stress decreasing time and long hold time can enhance crack growth when the stress rise time is short. But in Kawai, S., et al. [165] study, a decrease in stress rise time (0.04 s) and long hold time (1.92 s) did not increase the crack growth rate, in fact, it decreased the crack growth rate.

Wand, Z. F. [166] studied the influence of waveform on CFCG in iron in 3.5%NaCl. The yield stress, tensile strength and elongation of the iron are 150MPa, 295MPa and 42% respectively. The test frequency was 0.1 or 1Hz and $R = 0$. The waveforms investigated were sinusoidal, positive sawtooth, negative sawtooth and square waveform. They found that the positive sawtooth and square loading waveforms have very little or no effect on the fatigue crack growth rate in air. This result tends to confirm that, waveform general has no effect on the rate at which a fatigue crack grows in steel in air. Many studies have shown that stress ratio (R) in the range 0.1- 0.85, frequency in the range 0.005Hz – 50Hz and waveforms have little or no effect on the FCGR in the ferrite-pearlite steels in air [150] [153][161][167][168] [169][170]. They reported that the effect of the loading waveform on CFCGR was due to the difference in the straining mode at the crack tip (crack tip geometrical difference) and not as a result of material dissolutions. They found that under anodic dissolution, the amount of material dissolved at the vicinity of the crack tip within one cycle of loading was the same

for sinusoidal, positive sawtooth, negative sawtooth and square waveforms. A continuous loading pattern was reported to give accelerated CFCGR. The CFCGR of the positive sawtooth was higher than that of the sinusoidal and the rates for the two waveforms were higher than that of negative sawtooth and square at free corrosion condition. They also found that for the square waveform, holding for 0.1 s and 0.9 s at maximum load had little influence on the CFCGR. Consequently, they attributed the waveform crack growth rate difference to the continuous deformation of the crack tip material.

In the above discussions, we can infer that the process of environmental enhancement of fatigue is not an instantaneous increase in length, Δa , at maximum stress, σ_{max} point, but one that occurs before the maximum stress is reached. During this stress increasing period, the corrosive environment can enhance the decohesion of the crack tip or cause the dissolution of fresh exposed surfaces and consequently accelerate the crack extension. In other words, the increasing tensile stress at the crack tip appears to be necessary for FCGR to be enhanced by the aggressive environment. Again, the crack tip is required to remain uncovered by corrosion products for the FCGR to be enhanced. The increase in tensile stress at the crack tip is said to prevent protective coating from forming on the freshly exposed tip thereby facilitating increase in FCGR by dissolution mechanism which is known to operate at free corrosion. The retardation of crack growth with square waveform has been attributed to crack tip blunting due to the dissolution reaction.

BS EN ISO 11782 informed that for a given cyclic frequency, the waveform determines the rate at which the protective film on a corroding surface ruptures. And the film rupture at the crack tip may determine the rate of crack growth in corrosive environment. It stated that the holding at minimum, intermediate or maximum load may increase or decrease CFCGR, depending on the failure mode. Holding at maximum load may be expected to increase the CFCGR if the K_{max} is above K value for time-dependent modes. It further informed that CFCGR

may be reduced under holdtime in steels that are resistant to static-load cracking because of time-dependent crack tip blunting due to corrosion of the plastically deformed region around the crack tip.

Generally, the studies above show that CFCGR is sensitive to waveform and frequency. This sensitivity to waveform underscores the need to perform CFCGR tests under practical operating conditions. Hence, in fatigue analysis of OWT high-practical-value fatigue analysis can only be achieved under realistic representation of service loads, actual frequencies and waveforms. In as much as perfect cyclic waveform of wind loading is difficult to obtain and simulate in a fatigue test, near match waveform or conservative one is necessary to generate high quality data for assessment of crack growth rates in modern S355 steel in *SW* similar to OWT service condition. This will increase confidence in prediction of fatigue life and inspection planning of OWT.

In all the studies mentioned above, there was no explicit comparison of waveform effect under the same frequency and load ratio for subgrades of a particular grade of steels with differing microstructures. There was also limited micrographs showing mode of crack growth and the specific roles of microstructural phases under these waveforms in the Paris Region. Such study is expected to show clearly the damage mechanism and the effect of the material microstructure if it exists in the Paris Region. The influence of waveform other than sinusoidal on CFCGR for the steels used for WT design is non-existent, at least to the extent of information available to the authors. The intention of this study, therefore, is to determine the effect of these waveforms under different load frequencies in *SW* and to establish the most conservative results from the two waveforms for reliable assessment of OWT foundations.

This study therefore, seeks to presents new sets of experimental results on the effect of frequency, waveform, load level/ K_{max} and *SW* on the FCGR in normalised-rolled (NR) and thermomechanical control process (TMCP) ferrite-

pearlite steel subgrades used in the fabrication of modern offshore structures, including WT monopile foundations, in the Paris Region of the da/dN vs. ΔK curve. The waveforms under considerations are sine and hold time and the experiments were to be conducted under frequencies of 0.2Hz, 0.3Hz, 0.5Hz and stress ratio of 0.1. The experiments performed in this study are expected to assist towards more reliable assessment of effect of waveform on the *FCG* in *SW* under free-corrosion condition

4 Experimental study

4.1 Materials, specimen geometry and loading conditions

Traditionally, BS4360 Grade 50D was widely used in the design of offshore structures. In recent years, European Standard EN10025 S355 (or EN 10225 S355) has become the main structural steel typically in use for WT structures [33][34] and different sub-grades of S355 are selected for use in offshore applications, including monopiles. In fact, the equivalent grade of BS4360 Grade 50D is S355J2+N. The modern material choice for offshore application and the conventional structural steels are shown in **Table 4-1**.

Table 4-1 S355 structural steels for marine applications, (a) advanced S355 subgrades (b) Conventional S355 structural steel grades

Offshore Structural Steel Plates			
(a)		(b)	
European	Typical offshore application	European	British
EN10225		EN10025 Part 2: 2004	BS4360:1990
S355G7+N	Primary Structure	S355	50A
S355G7+M	Primary Structure	S355JR	50B
S355G8+N	Critical Joints	S355JO	50C
S355G8+M	Critical Joints	S355J2	50D
S355G9+N	Primary Structure	S355J2+N	50D
S355G9+M	Primary Structure	S355K2	50DD
S355G10+N	Critical Joints	S355K2+N	50DD
S355G10+M	Critical Joints	S355NL	50EE

The materials for this study were S355G10+M, S355G8+M and S355J2+N sub-grades. The came as a 90-mm thick plate of EN-10225:01 S355G10+M steel, with inspection certificate of 3.2 as per EN 10204:2004. The subgrade is also known as 1.8813+M. These steels are essentially a C-Mn steels. The S355G10+M is reported to have enhanced through-thickness ductility and impact energy verified at -40°C. The material was reported by the manufacturer to have a Yield Strength, R_P (σ_y) of 460MPa and Ultimate Tensile Strength, R_m (σ_{UTS}) of 561MPa. A confirmatory tensile test performed prior to experiment gave Yield

strength, $R_P (\sigma_y) = 435$ MPa, Tensile strength $R_m = 545$ MPa and tensile strain at failure of 38%. The values quoted is the average value of the two tests performed.

S355G8+M and S355G10+M are thermo-mechanical control processed (TMCP) steels while S355J2+N is normalised-rolled steel (NR). **Table 4-2** gives the chemical and mechanical properties of these sub-grades. Note that The composition and mechanical properties of the subgrades, and S355J2+N in this study is similar to the steel studied by Ref [171], hence the mechanical property of their steel has been reported here.

Table 4-2 Test materials - chemical compositions and mechanical properties S355 structural

Steel Grade	Notation	C	Mn	Ni	Si	Cu	Cr	Mo	E%	YS (MPa)	UTS (MPa)
S355G10+M (TMCP)	G10	0.06	1.57	0.33	0.27	0.24	0.03	0.01	38	435	545
S355G8+M (TMCP)	G8	0.05	1.52	0.32	0.27	0.24	0.03	0.01	35	447	549
S355J2+N (NR) [171]	J2N	0.17	1.54	0.04	0.04	0.08	0.02	0.10	20	385	531

Compact tension C(T) specimens were extracted from the mid-thickness of a 90mm thick plates shown in **Figure 4-1 (a)** and **(b)**. The specimens were extracted in such a way that they were far from near surface residual stress induced regions. **Figure 4-1 (c)** shows the 3D schematic of the experimental specimen manufactured to ASTM E647 –15 [172]. The crack plane and direction of propagation are perpendicular to the rolling direction as schematically shown in **Figure 4-1 (d)**. This orientation is the through-thickness and the most likely plane of crack initiation and propagation for monopiles. The test specimens were pre-cracked in air using force shedding (or K -decreasing) method as described in [172]. The pre-crack length for a given machined V-notch was at least 1.5mm and no more than 3.5mm. The load ratio for pre-cracking was 0.1 and loading frequency of 4 or 5Hz.

The initial conditions of the C(T) specimens before start of test are given in **Table 4-3**. A total of 18 corrosion-fatigue tests were conducted in free-corroding condition. S denoted sinewave, H: holdtime followed by the maximum load P_{max}

and F: frequency, followed by the frequency value. All tests were conducted using the same load ratio of $R = 0.1$.

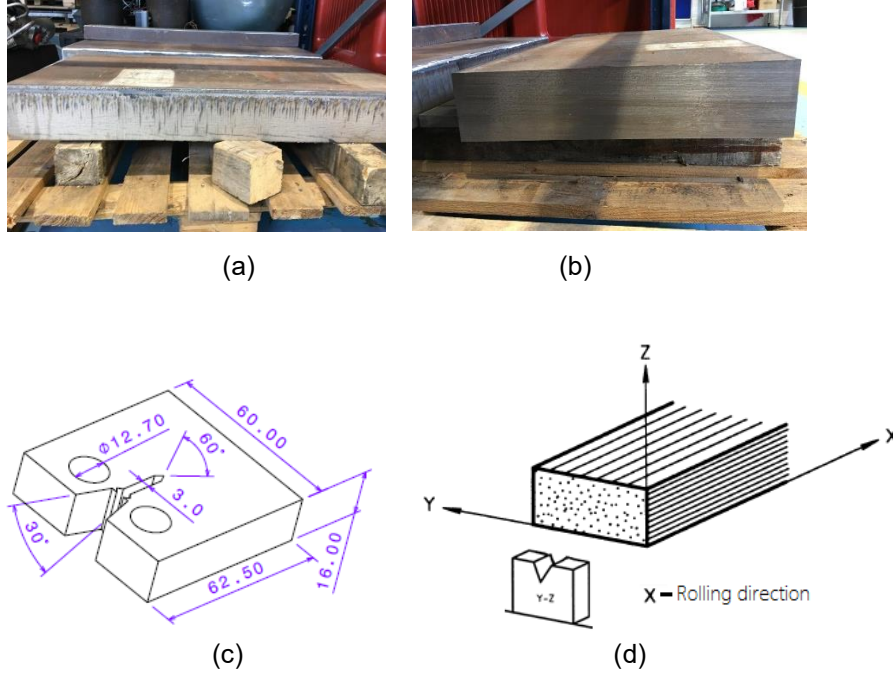


Figure 4-1 Experimental test material: (a) and (b) 90 mm steel plates used in this study, (c) 3D schematic of the C(T) specimen in accordance with ASTM E647 –15 [172], (d) Orientation of the crack plane

The use of $R = 0.1$ was to remove mechanical fatigue contribution due to change in mean stress. All the fatigue tests were conducted in accordance with ASTM E647-15 [172]. The substitute SW was prepared in accordance with ASTM D1141-98 [144][173]. Additional description of the experimental procedure can be found in reference [144].

4.2 Experimental waveform

This study is primarily designed to evaluate the effect of waveform on the CFCGR of S355 steel under conditions pertinent to WT operational environment. In many fatigue tests, the waveform that is usually employed is sinusoidal wave of constant cyclic loading amplitude [149] and a constant mean load. But, the wave

nature of wind is indeterministic. As a simplified approach, the wind loading on a WT can be assumed to occur in three stages – the rising stage, ‘dynamically steady’ stage and relaxation stage.

**Table 4-3 Specimen dimensions and loading conditions (in mm) in seawater (SW).
R = 0.1**

#	Test ID.	W	B	a_o	a_r	f (Hz)	P_{max} (kN)
1	J2N-A12F2	50.00	16.00	14.00	32.50	2	12
2	J2N-A10F2	50.00	16.00	14.00	31.27	2	10
3	J2N-A10F5	50.00	16.00	20.50	29.53	5	10
4	G8-A10F2	50.00	16.00	14.50	35.00	2	10
5	G8-A10F5	50.15	15.90	20.33	32.12	5	10
6	G10-A10F5	50.01	16.00	20.30	29.53	5	10
7	G10-A10F4	50.00	16.00	24.75	32.81	4	10
8	G10-A9F5	50.01	16.03	20.74	30.95	5	9
9	G10-S10F0.2	50.0	16.00	21.26	29.11	0.2	10
10	G10- H10F0.2	50.05	16.08	22.37	29.38	0.2	10
11	G10-S10F0.3	50.03	16.05	21.57	29.44	0.3	10
12	G10- H10F0.3	50.03	16.07	21.87	29.73	0.3	10
13	G10-S10F0.5	50.03	16.04	21.20	29.61	0.5	10
14	G10- H10F0.5	50.03	16.04	21.26	29.33	0.5	10
15	G10-S9F0.2	50.00	16.00	22.68	30.16	0.2	9
16	G10-H9F0.2	50.00	16.00	22.98	29.81	0.2	9
17	G10-S9F0.3	50.01	16.16	21.69	29.30	0.3	9
18	G10-H9F0.3	50.01	16.16	20.22	29.51	0.3	9
19	G10-S9F0.5	50.05	16.06	21.95	30.21	0.5	9
20	G10-H9F0.5	50.05	16.08	22.09	30.30	0.5	9
21	G8-H10F0.2	50.01	16.02	17.03	33.83	0.2	10
22	J2N-S10F0.2	50.00	16.00	18.32	28.02	0.2	10
23	J2N-H10F0.2	50.00	16.00	18.55	31.07	0.2	10

These three stages vary continuously and randomly to build up the wind spectrum. A simplistic view of this idea is shown in **Figure 4-2 (a)** with associated gust loading and turbulence. This waveform is adapted from typical wind loading spectrum acting on an aircraft wings [148]. This wave spectrum can be approximated to a hold time waveform as shown in **Figure 4-2 (b)**. **Figure 4-2 (c)** shows the associated sinewave. Thus, two waveforms were used in this study –

sinusoidal and holdtime with constant amplitude as shown in **Figure 4-2 (b)** and (c). A commercial Instron software was used to programme the two waveforms as shown in **Figure 4-2 (d)** and (e). The load ratio 0.1 and loading frequencies of 0.2Hz, 0.3Hz and 0.5Hz were used.

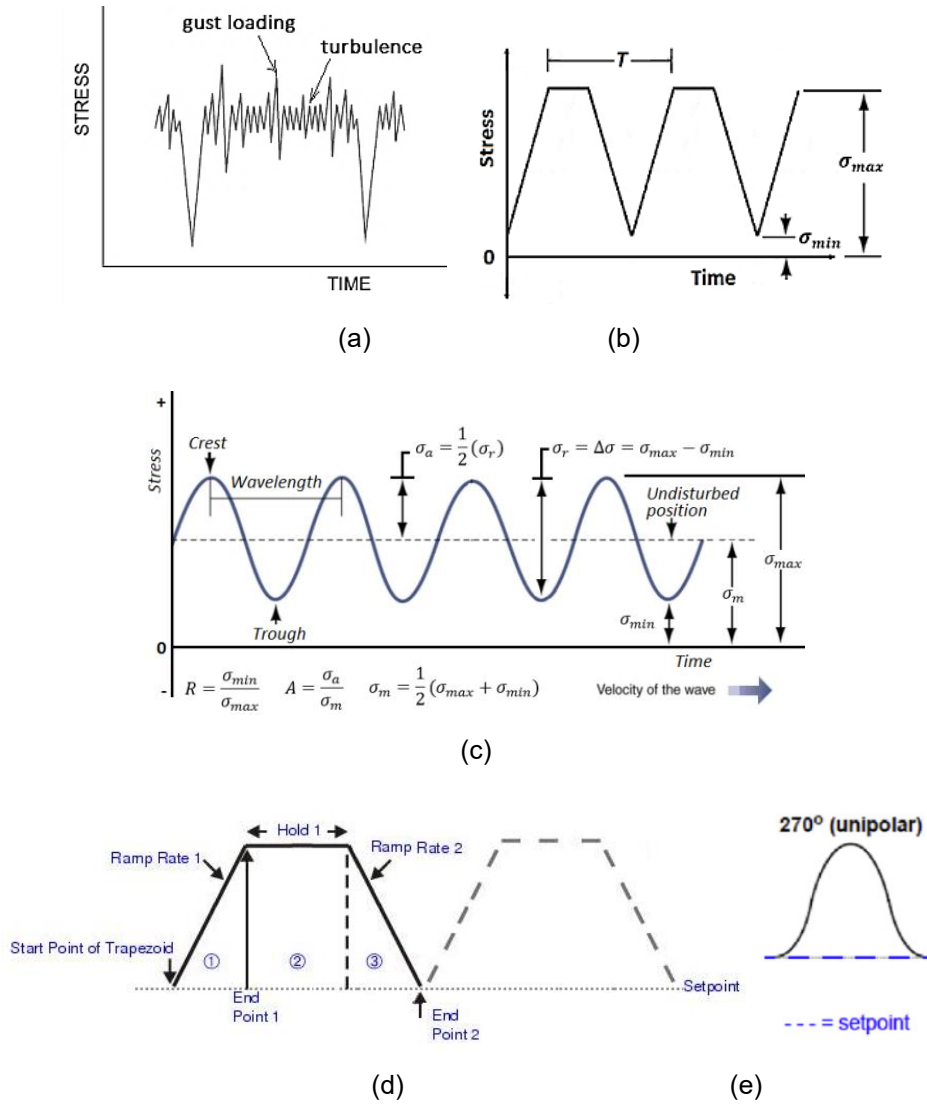


Figure 4-2 Experimental waveforms and loading parameters: (a) A simplistic illustration of wind loading stages on a WT, (b) & (c) holdtime and sine waveforms used in the present study, (d) & (e) programme waveforms

The loading rate for the holdtime waveform was constant for the three frequencies while the holdtime varied. For 10KN for example, the starting point or setpoint

(P_{min}) for the holdtime waveform is 1kN. From the setpoint, the load is ramped to 9kN and cycled above the setpoint at a rate of 48.6kN/s. This implies that the rise time to 9kN for the holdtime cycle is about 0.19s, 10kN is 0.21s; and 12kN is 0.25s. Because the rise times were relatively close for 9kN and 10kN, most of the tests were done under the two loading conditions.

The rise-time is almost equal to the fall-time. The frequencies 0.2Hz, 0.3Hz and 0.5Hz were generated by varying the hold time at P_{max} , but with constant loading rate of 48.6kN/s from the setpoint. Typical waveforms generated for the actual experiment are shown in **Figure 4-3 (a to c)** for the case of 10kN.

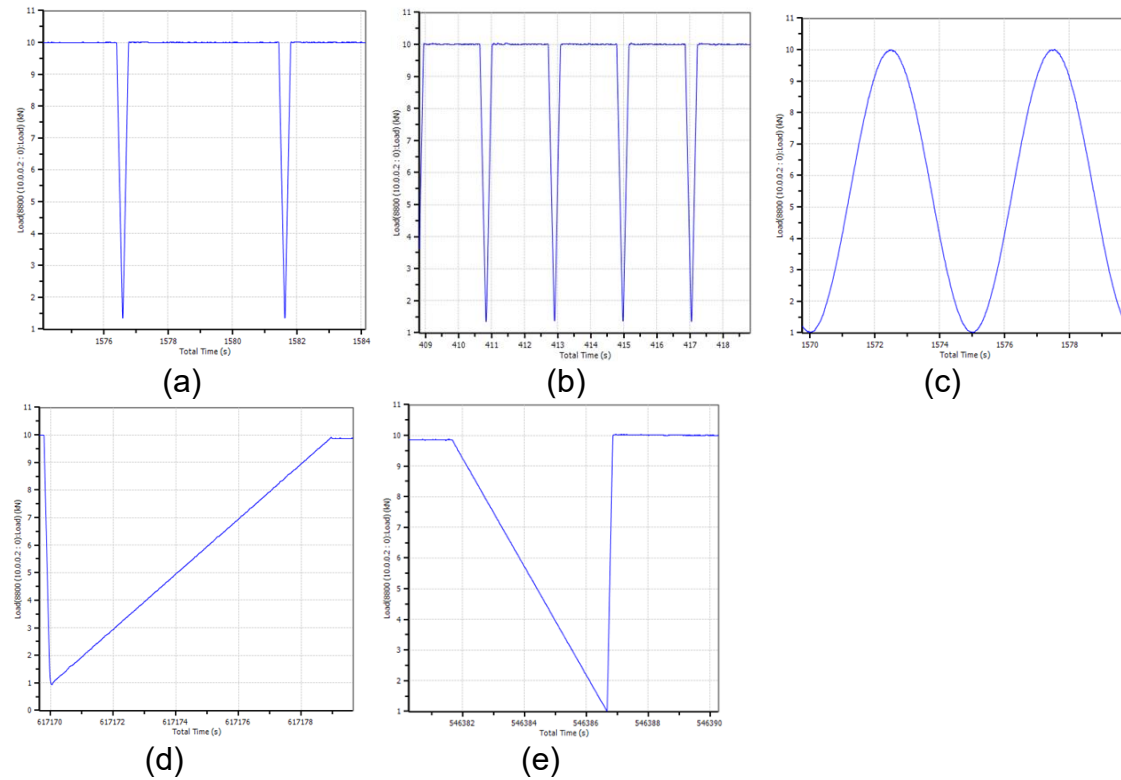


Figure 4-3 Experimental waveforms: (a) holdtime - 0.2Hz,10kN, ht: 4.59s, (b) holdtime - 0.5Hz,10kN, ht: 1.59s, (c) sinewave – 0.2Hz, 10kN, (d) & (e) holding for 10 s at 99.8% to take microstrain value

The summary of the loading conditions is given in **Table 4-4**. **Figure 4-3 (d) and (e)** show the ramping up to 99.8% of P_{max} and holding for about 10s to take readings

and ramping down for continued cycling. This is to avoid the static force overshooting above P_{max} . This is consistent with BS 7910:2013+A1:2015 [174] where it stated that the applied static force must not exceed the maximum fatigue force, else transient effect or crack blunting may occur.

Table 4-4 Holdtime waveform parameters

Load in kN; Loading rate 48.6kN/s; R=0.1 TTRF = Total time to rise and fall (excluding hold time)						
P_{max}	P_{min}	Freq	Wave Period	Rise time	Fall time	Hold time
9	0.9	0.2	5.00	0.19	0.18	4.63
9	0.9	0.3	3.33	0.19	0.18	2.96
9	0.9	0.5	2.00	0.19	0.18	1.63
10	1.0	0.2	5.00	0.21	0.20	4.59
10	1.0	0.3	3.33	0.21	0.20	2.92
10	1.0	0.5	2.00	0.21	0.20	1.59
12	1.2	0.3	3.33	0.25	0.24	2.84

4.3 Test setup in air and crack length (a) measurement

A 100 kN frame capacity, computer-controlled, with modern digital controllers Instron 8801 servo-hydraulic fatigue test system was used to perform the FCG test. It was set-up in accordance with the standards enumerated in Ref. [175]. The machine has an actuator stroke of ± 75 mm and the load cell calibrated dynamically as recommended in BS Standard [176] to a level of 1.5 times the maximum test load. The machine grips were accurately aligned using double pin-clevis arrangement. In general, the fatigue tests complied with these references [172][177]. The test specimens were pre-cracked in air using force shedding (or K-decreasing) method as described in Ref. [172]. The pre-crack length for a given machined V-notch was in the range 1.5 – 3.5 mm. The load ratio for pre-cracking was 0.1 and loading frequency of 4 or 5Hz. Air tests were conducted at laboratory ambient temperature conditions of approximately 20 °C. Tests were conducted under load control mode. Before the start of the experiment, the machine load cells was calibrated dynamically as recommended in BS Standard [176] to a level

far above the maximum test load. The machine grips were accurately aligned using double pin-clevis arrangement.

Before setting C(T) specimen into the fatigue machine's grips, its surface was polished with 320 and 400 grit abrasive papers and using direct lighting to aid in the resolution of the crack tip in air test. The crack length, a were measured in air using digital cameras and, in some cases, travelling microscope. **Figure 4-4 (a)** shows the test setup and crack tip monitoring using two *StreamPix5* digital cameras at both sides of the specimen. **Figure 4-4 (b & c)** shows the actual measurement of a using a measuring commercial software installed in the computer. **Figure 4-4 (d)** shows combination of digital measurement and metric rule to ensure consistency and accuracy.

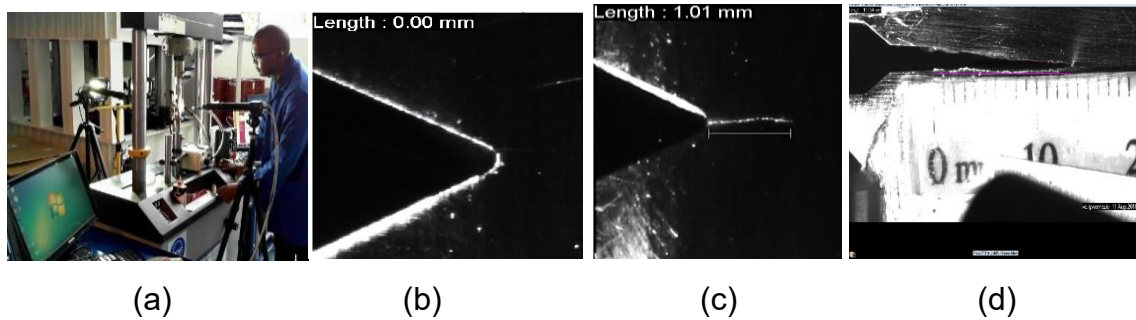


Figure 4-4 Fatigue crack length measurement and calibration technique
(a) Monitoring crack growth using digital cameras (b) and (c) digital measurement of crack growth (d) use of metric rule for calibration and consistency

The BS 7910 recommends taking measurements at interval of $\Delta a \leq 0.04W$ for $0.25 \leq a/W < 0.40$ and $\Delta a \leq 0.02W$ for $0.40 \leq a/W < 0.60$, where a is the crack length, W is C(T) width. This implies that for the first case, if $W = 50$ mm measurement interval will be equal or less than 2mm. This interval is judged to be too large because it might not show the real trend of FCG mechanism. The standard number of cycles for taking a reading was every 2000 interval. As the crack driving force increases, this value is reduced to 1000 and 500. This is to capture as many points as possible along the FCG curve. A continuous stream

of large data was generated with this method, instead of few points that would have been the case if an interval of 2mm was used. Within this large data, equal crack distance of 0.5, 1.0, 1.5, 2.0 mm and corresponding number of cycles (N) can be noted in each case and the curve generated are all the same. Use of 0.5 mm interval simply increases the data points on the plotted curve and better showcased the true trend of the FCG. If a is taken for every 500, 1000 or 2000 cycles, the curve generated is still the same as in the former case. If the number of cycles is taken for every a equal to or above 0.02 mm, the curve trend obtained is still the same as in the above cases, but with many more data points. Measuring a at every 500 cycles was found to be similar to taking number of cycles at every 0.02 or 0.5 mm crack length increase – the effect was simply a production of more data points on the plotted curve

During air test, it was ensured that the a at front and back of the C(T) specimen did not differ by more than 2 mm otherwise the test is terminated. BS Standard recommends rejection of data when the crack lengths at the front and back differ by 4 mm. Throughout the tests, the difference in the a for the two faces was less than 1.0 mm. This is because, it was always ensured that the C(T) specimen was accurately aligned perpendicularly to the loading axis and specimen centralized within grips. BS 7910 also recommends that presence of tunnelling or crack-front curvature be checked at the end of the experiment. If tunnelling is found, then an average through-thickness crack length, a_{avg} must be calculated. This is done by using 3 to 5 equally spaced points across the fatigued surface. The crack curvature correction length, a_{cor} then becomes the difference between the a_{avg} and a recorded during the test. The determination of a_{cor} is supposed to be done at 2 or 3 locations on the fatigued surface. Where the crack curvature correction results in a more than 5% difference in the calculated stress-intensity factor (SIF) at any location, then this correction must be included when analysing the recorded test data, and the effective crack length becomes [174]: $a_{eff} = a_n + a_{fat} + a_{cor}$, where a_n is the notch length and a_{fat} the fatigued length. If the crack curvature correction factor varies with a , then a linear interpolation may be used to

determine the middle data. **Figure 4-5** shows the extent of crack curve in three of the many C(T) specimens used in the establishment of the calibration curves. Note that the fatigue was carried out to complete failure. The colour-inverted picture clearly shows the end of the plain strain fatigued surface. The material has good ductility, and this can be seen from the extensive shear lips that occurred at about 45°. **Figure 4-5 (a & b)** show typical shear lips and procedure of measurement of notch and pre-crack which forms the initial a .

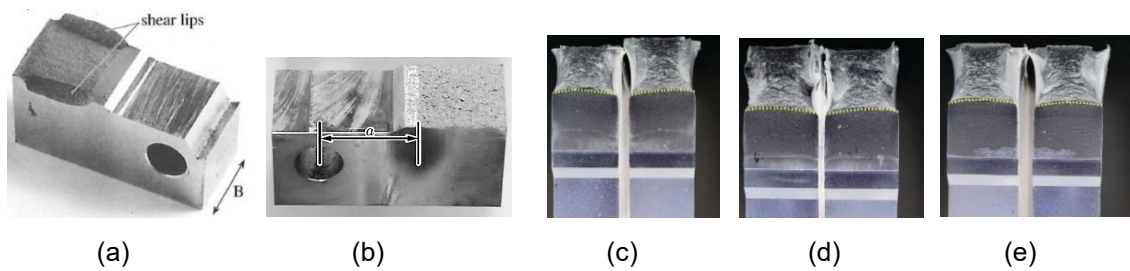


Figure 4-5 Picture of fatigued surfaces for the samples tested in air and used for the calibration curve

It can be seen from **Figure 4-5 (c, d & e)** that the difference between the front and back crack lengths is generally less than 0.5 mm, far less than the 4mm BS7910 maximum. There was no crack curvature and no mixed mode of fatigue. Beyond the yellow dash lines, failure occurred entirely under plain stress condition.

4.4 Calibration curve for seawater (SW) test

For SW test, the presence of an environmental chamber containing an aqueous solution hides the crack tip, necessitating the use of a nonvisual technique as recommended in Ref. [172]. Hence, in SW the Back-Face Strain (BFS) method was used. BFS is shown to be highly reliable in measuring crack growth in SW more than Alternating Current Potential Difference (ACPD), Direct current potential difference (DCPD) and optical measurements [149]. It is simple and cost effective, but thorough understanding on how it is used is needed, because minor error in strain gauge installation could lead to large error in crack growth measurement [178]. To measure the a in SW, a calibration curve is required. For each material, at a particular load level, about three sets of air test are needed to

set the calibration equation right. **Figure 4-6** shows activities needed to make a calibration. Linear electrical resistance strain gauges shown in **Figure 4-6 (a)** were fitted at the back of the C(T) specimens for the air tests. **Figure 4-6 (b)** shows the schematic C(T) specimen with BFS gauge and **Figure 4-6 (c)** the actual bonding of strain gauge to the experimental C(T) specimen. The installation complied with instructions as given in [179][180][181]. The lead wires were soldered to the strain gauge through the connecting tabs in accordance with [182], and the other ends of the wires connected to a Vishay Model P3 Strain Indicator and Recorder [183] shown in **Figure 4-6 (d)**.

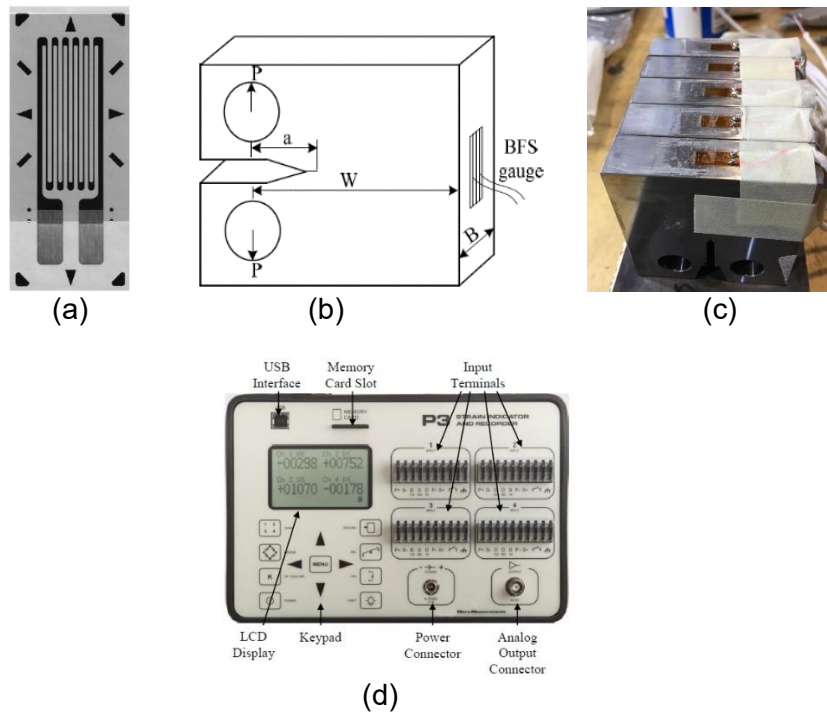


Figure 4-6 Calibration set up (a) linear electrical resistance strain gauge (b) Schematic of a C(T) specimen with bonded strain guage (c) bonding of strain gauge to test samples (d) Vishay Model P3 Strain Indicator and Recorder

The calibration correlates the a in air, number of cycles (N) and amount of microstrain (μm) produced by the strain gauge. In the corrosion-fatigue test, the N vs μm was recorded, hence, this trend was used in the calibration process. The static force was maintained at 99.8% for about 10s to take readings after a predetermined cycle or a . The microstrain reading was made using the

microstrain recorder and the number of cycles was digitally displayed by Instron control console. Gauge factor balancing and overall operation of the strain box can be found in references [183] and [184]. It is important to always quote the strain gauge used for calibration due to, for example, gauge factor. In this test, it was observed that using different linear strain gauges for calibration could cause significant variation in the BFS curve. This underscores the need for consistency in the use of same strain gauge type for a given corrosion-fatigue study. The strain gauge used for this study was CEA-06-240UZ-120 gauge of Vishay Precision Group, having gauge factor of 2.14, grid resistance $120 \pm 0.3\%$ and gauge length of 6mm.

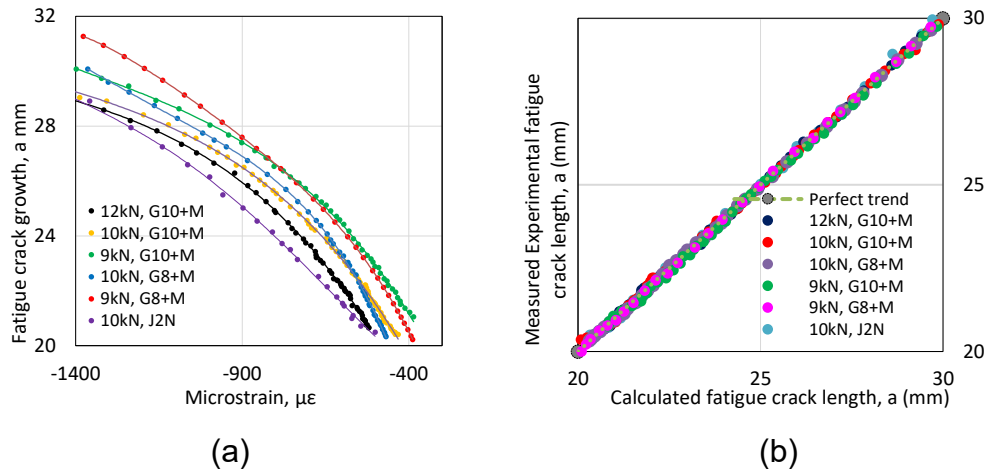


Figure 4-7 (a) Calibration curves (b) Correlation between the experimental data and the calibration equations

It works well in the temperature range of $(-75^{\circ}$ to $+175^{\circ}\text{C})$ [49]. **Figure 4-7 (a)** shows the calibration curves for 12, 10 and 9kN for the three subgrades of the S355 steels. The 9kN, G8+M curve was extracted from Ref [175] and the 10kN, G8+M was generated in the present study. The 10kN, NR (J2N) curve generated had the lowest resistance to compressive stress compared to that of 10kN, G10+M and 10kN, G8+M. This is evident from its lower yield stress value (see **Table 4-2**). This appears to suggest that the mechanical property has an influence on the value of the microstrain. In other words, the calibration curve varied with steel grade and with load level. **Figure 4-7 (b)** is the correlation between the

experimental data and the calibration equations. The calibration equations generated are listed in **Table 4-5**:

Table 4-5 Equations of the BFS calibration curves for various examined materials at different load levels

Material	P_{max} (kN)	Calibration equations
G10	12	$a = -1.50 \times 10^{-12}(\mu\epsilon)^4 - 1.12 \times 10^{-8}(\mu\epsilon)^3 - 3.24 \times 10^{-5}(\mu\epsilon)^2 - 4.48 \times 10^{-2}\mu\epsilon + 4.72$
G10	10	$a = -1.50 \times 10^{-12}(\mu\epsilon)^4 - 1.10 \times 10^{-8}(\mu\epsilon)^3 - 3.03 \times 10^{-5}(\mu\epsilon)^2 - 4.05 \times 10^{-2}\mu\epsilon + 7.58$
G10	9	$a = -1.50 \times 10^{-12}(\mu\epsilon)^4 - 1.15 \times 10^{-8}(\mu\epsilon)^3 - 3.06 \times 10^{-5}(\mu\epsilon)^2 - 3.89 \times 10^{-2}\mu\epsilon + 10.0$
G8	10	$a = -1.23 \times 10^{-11}(\mu\epsilon)^4 - 6.08 \times 10^{-8}(\mu\epsilon)^3 - 1.12 \times 10^{-4}(\mu\epsilon)^2 - 9.77 \times 10^{-2}\mu\epsilon + 6.52$
G8	9	$a = -1.23 \times 10^{-11}(\mu\epsilon)^4 - 5.02 \times 10^{-8}(\mu\epsilon)^3 - 7.93 \times 10^{-5}(\mu\epsilon)^2 - 6.60 \times 10^{-2}\mu\epsilon + 3.90$
J2N	10	$a = 6.42 \times 10^{-12}(\mu\epsilon)^4 + 2.64 \times 10^{-8}(\mu\epsilon)^3 + 3.45 \times 10^{-5}(\mu\epsilon)^2 + 6.20 \times 10^{-3}\mu\epsilon + 17.7$

Using equations in **Table 4-5**, there is an excellent correlation between the measured and calculated crack lengths within a range of 20 to 30mm and this increased the confidence in the predicted a in *SW*.

4.5 Seawater test

It is known that concentrations of ocean water vary with sampling location. To maintain consistency and to compare this work with other studies, substitute *SW* in accordance with ASTM D1141-98 [173] was used. The composition of the artificial *SW* is given in **Table 4-6** [173].

Table 4-6 Chemical composition of substitute seawater water [173]

Compound	Concentration, g/L
NaCl	24.53
MgCl ₂	5.20
Na ₂ SO ₄	4.09
CaCl ₂	1.16
KCl	0.695
NaHCO ₃	0.201
KBr	0.101
H ₃ BO ₃	0.027
SrCl ₂	0.025
NaF	0.003

The pH of the SW was maintained in the range of 8.0 to 8.2. Before the commencement of the tests, pre-cracked specimens were soaked in SW for about 2 days. The bonded strain gauge on the C(T) sample was isolated from the SW by coating as shown in **Figure 4-8 (a)** and the procedure is described in Ref. [185]. The coating also protected the installation from mechanical damage. During the test, the SW temperature was maintained within the range of 7.5°C to 8. 2°C. The specimen was then pin-loaded to the grips of the fatigue testing machine shown in **Figure 4-8 (b)**.

The SW was circulated around the C(T) specimen in a Perpex chamber (**Figure 4-8 (c)**) by means of a pump at a continuous rate of about 4L/min. 70 litres of SW was use for each experimental run.

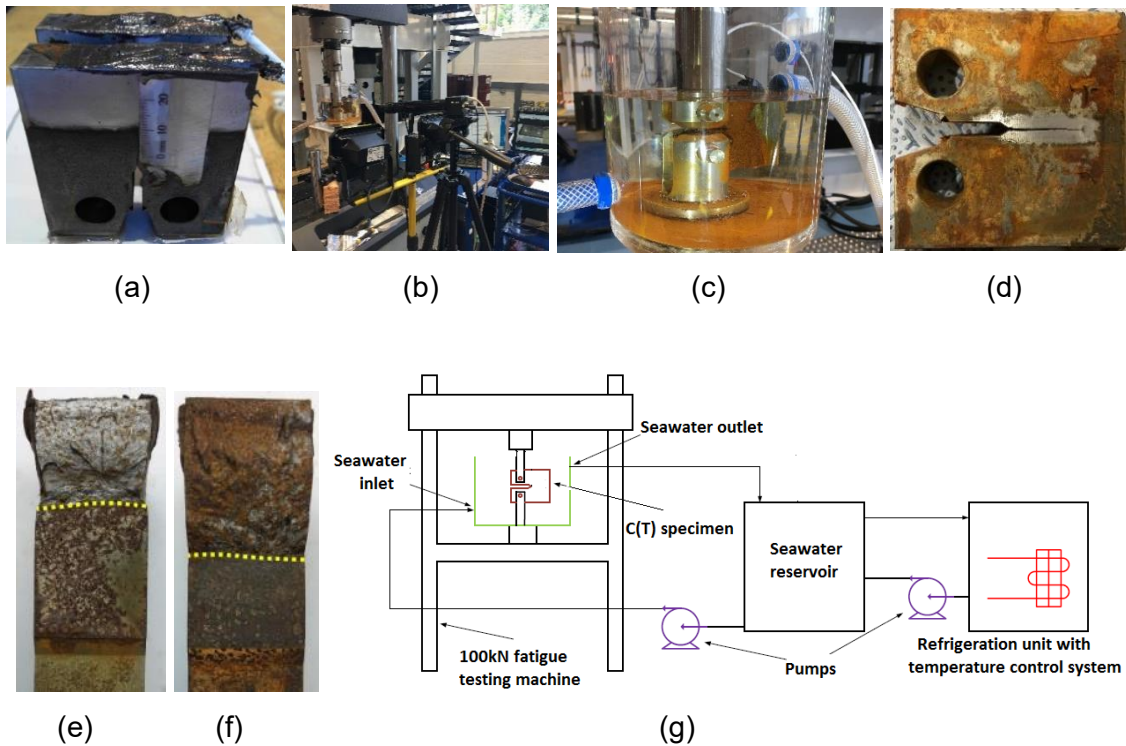


Figure 4-8 (a) Experimental set up: (a) Instron fatigue test system, (b) corrosion-fatigue test chamber, (c) checking validity of crack growth in seawater, (d & e) no tunneling or sides difference in fractured samples (g) Schematic representation of complete fatigue testing circuit.

The volume of the SW was such that the same chemical conditions and pH range were obtained at the end of each test. Duration of each run was from 10 days to 7 weeks. The SW was replaced after every 15 days, and this is simply to conform to ASTM E647. The specimen was completely immersed in the SW throughout the test period.

Before the data is accepted for analysis, the validity of the crack growth path as per ASTM E647 or BS7910 was checked as shown in **Figure 4-8 (d)**. This crack branching or out-of-plane cracking may render the test invalid. **Figure 4-8 (e & f)** show fatigued surface, with no tunnelling and little or no difference in the front and back crack lengths. The samples were broken in a liquid nitrogen, hence no presence of shear lips. **Figure 4-8 (g)** shows the complete experimental set-up for this study. The general experimental procedure is illustrated in a flow diagram in **Figure 4-9**.

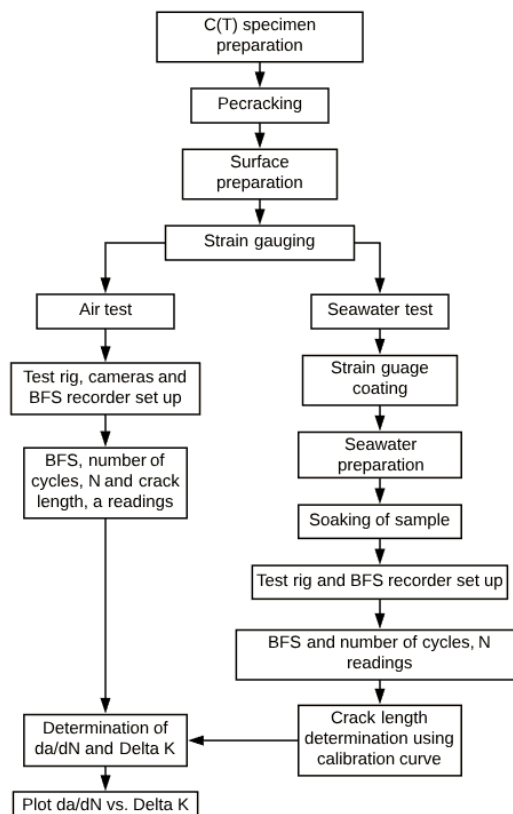


Figure 4-9 Flow chart of experimental procedure

4.6 Fatigue crack growth rate calculations

The fatigue crack growth rate calculation was carried out in accordance with linear elastic fracture mechanics (LEFM) and as specified in ASTM E647-15 [172] and BS EN ISO 11782-2:2008 [177].

The parameters used in air and *SW* tests calculations are applied load, P , crack length, a , number of cycles, N and microstrain (BFS), $\mu\epsilon$. Using the appropriate calibration curve (**Figure 4-7 (a)**), the $\mu\epsilon$ was converted to a and ΔK calculated. The change in the maximum stress intensity factor, K_{max} during the crack growth was calculated thus:

$$K_{max}(\text{or } K_{min}) = \frac{P_{max}(\text{or } P_{min})}{B\sqrt{W}} * Y\left(\frac{a}{W}\right) \quad (4-1)$$

$$\Delta K = K_{max} - K_{min} = \frac{\Delta P}{B\sqrt{W}} * Y\left(\frac{a}{W}\right) \quad (4-2)$$

where P_{max} is the maximum applied load, W is the specimen width and B is the specimen thickness. The change in stress intensity factor range (ΔK) during cyclic loading is ΔK , and $Y(a/W)$ is given as:

$$Y\left(\frac{a}{W}\right) = \frac{(2+\alpha)}{(1-\alpha)^{3/2}} * (0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4) \quad (4-3)$$

$$\alpha = \frac{a}{W} \quad \text{and this equation is for } \frac{a}{W} \geq 0.2$$

ΔP is the applied cyclic load range and $Y(a/W)$ is a dimensionless specimen shape function. The solution of $Y(a/W)$ for conventional fracture geometries can be found in fracture mechanics textbooks. The general equation for determination of ΔK for the C(T) specimens becomes:

$$\Delta K = \frac{\Delta P}{B\sqrt{W}} * \frac{(2+\alpha)}{(1-\alpha)^{1.5}} * (0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4) \quad (4-4)$$

The ASTM standard suggested the use of fitted crack length, a_i corresponding to N_i in calculating the ΔK associated with da/dN (see Refs [172][177]) instead of the experimentally observed crack, a to compensate for human error in a measurement and material anomalies. This method tends to produce a fine arrangement of the da/dN vs ΔK data points along the power law curve. In their recommendation, the rate of crack growth, da/dN at a particular number of cycles, N_i and crack length a_i was calculated starting from the quadratic function as recommended by [172][177]:

$$\hat{a}_i = b_2 X^2 + b_1 X + b_0 \quad \text{for } -1 \leq X \leq 1 \quad (4-5)$$

b_0 , b_1 , and b_2 are the coefficients of the quadratic polynomial that are determined by using the least squares method. The least squares method involves minimizing the square of the deviations between the experimental crack lengths observed, a and the fitted crack length values $\hat{a}$ on the curve. In other words, a is the observed value and $\hat{a}_i$ is the fitted value of the crack length at N_i . This least-squares technique is illustrated in **Figure 4-10 (a)**.

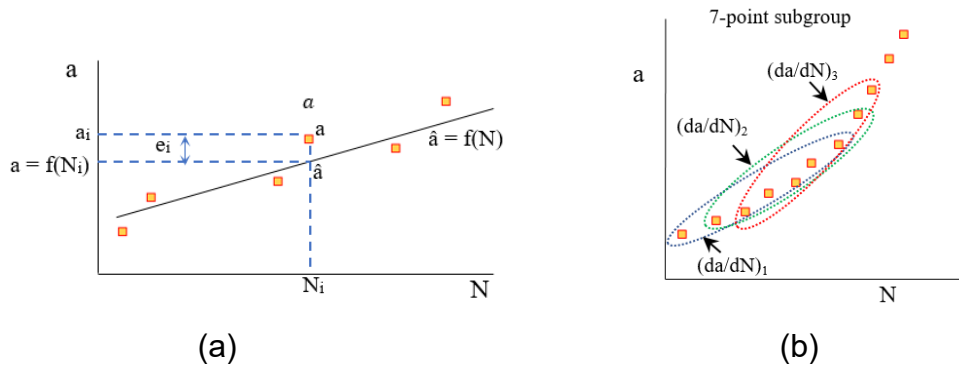


Figure 4-10 (a) Least-squares technique with error associated with fitted curve (b) the incremental polynomial

The curve, $\hat{a} = f(N)$ is fitted to a set of data points (a_1, N_1) , (a_2, N_2) , (a_3, N_3) , ... (a_m, N_m) . The e_i is the error or difference between the experimental a_i and $\hat{a}_i$ read from the fitted curve. Thus, the task in the least-squares method is to minimize the sum of the square errors $e_1^2 + e_2^2 + e_3^2 \dots + e_m^2$ where e is the error

associated with a certain set of points. In the quadratic polynomial, the values of the coefficients b_0 , b_1 , and b_2 are obtained by solving a system of equations. This operation nowadays can easily be implemented in *Excel*® (commercial software) and the error evaluated by the software.

In **Eq. 4-5**, X is given as $\left(\frac{N_i - C_1}{C_2}\right)$ by [172][177] so that **Eq. 4-5** becomes:

$$\hat{a}_i = b_2 \left(\frac{N_i - C_1}{C_2}\right)^2 + b_1 \left(\frac{N_i - C_1}{C_2}\right) + b_0 \quad \text{for } -1 \leq \left(\frac{N_i - C_1}{C_2}\right) \leq 1 \quad (4-6)$$

where

$$C_1 = \frac{1}{2}(N_{i+n} + N_{i-n}); \quad C_2 = \frac{1}{2}(N_{i+n} - N_{i-n})$$

$$\text{or simply } C_1 = \frac{1}{2}(N_{last} + N_{first}); \quad C_2 = \frac{1}{2}(N_{last} - N_{first})$$

The values of C_1 and C_2 are said to be used to scale the input data in order to avoid computational difficulty in determining the quadratic coefficients. The quadratic equation is fitted to sets of $(2n+1)$ successive data points. For the 7-point incremental polynomial method used in this study, $n = 3$, i.e. 7 successive data points, so that the local range is given by $a_{i-3} \leq a \leq a_{i+3}$. That is, 7 successive data points where used to obtain the parameters C_1 and C_2 which is used in turn to generate one fitted crack length from **Eq. 4-6**, and the process is repeated to generate all the fitted crack lengths with corresponding number of cycles. The local range or subgroup or the 7 successive data points in the 7-point incremental polynomial procedure is illustrated in **Figure 4-10 (b)**. Therefore, in this method you need many experimental data points to obtain satisfactory sizeable data for the da/dN vs. ΔK plot.

Note that it is expected that the value of the fitted crack length $\hat{a}_i$ corresponding to N_i should be used in calculating the ΔK associated with da/dN instead of the experimentally observed points a . The reason put forward to explain this is entirely to reduce human error in a observation and measurement, and minimize

the effects of local material anomalies [186]. The curve fitting is supposed to produce best estimates of the true a . This implies that in a situation where the a measurement is highly efficient with very little or no error, i.e., a situation where $\hat{a}_i$ is almost equal to a the experimental data points can be used straight away. The experimental N_i is used unaltered because it is read off from the computer counter and it is supposed to be free from human error. The rate of crack growth at N_i is obtained by differentiating **Eq. 4-6**, resulting to:

$$\left[\frac{da}{dN} \right]_{\hat{a}_i} = 2b_2(N_i - C_1) \frac{1}{C_2^2} + \frac{b_1}{C_2} \quad (4-7)$$

The value of the da/dN was then plotted against the corresponding ΔK on a log-log graph. When da/dN is plotted against ΔK in log-log axes, there are generally three regions observed in a *FCG* trend; (i) Region I (threshold) where ΔK_{th} , a minimum value of ΔK below which no fatigue crack propagation can be detected, is observed, (ii) Region II where a power-law relationship between da/dN and ΔK can be inferred, and the Region III of accelerated growth to failure. This study focuses on region II which is commonly used in engineering design. The equation describing this region is commonly referred to as the Paris Law. The Paris law is obtained from a straight line drawn through the linear part of the curve and the equation is given as:

$$\frac{da}{dN} = C \Delta K^m \quad (4-8)$$

where C and m are empirical power-law constants found by fitting a regression line to the test data as illustrated in **Figure 4-11**.

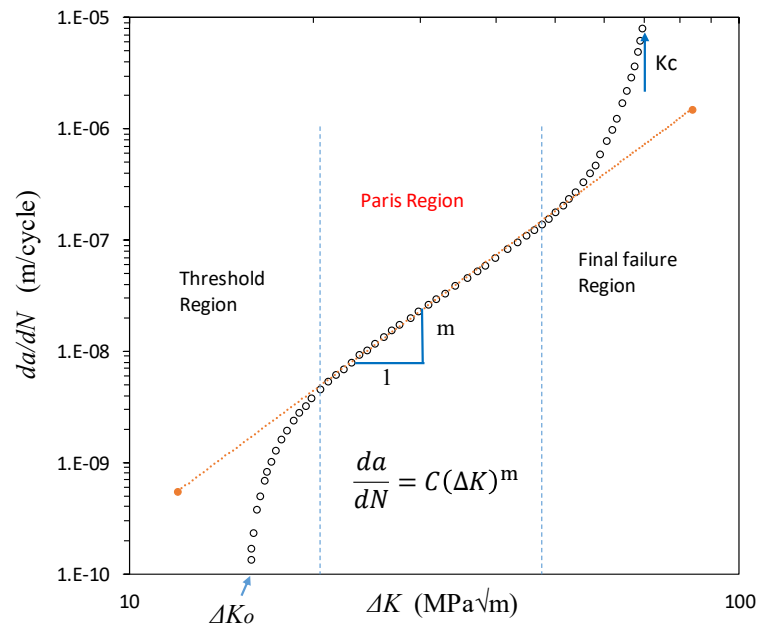


Figure 4-11 Schematic of variation in FCGR, da/dN with alternating stress intensity factor range (SIFR), ΔK in steels. (See [160][187][188]).

5 Materials selection for XL wind turbine support structures: A corrosion-fatigue perspective

5.1 Abstract

The continued growth of the offshore wind industry will depend essentially on reductions in wind energy production cost. Large cost reductions can be achieved through efficient, economic and optimised wind turbine support structures. To achieve maximum offshore wind adoption beyond 2020, significant industrial and research efforts are being made in optimised material selection and application. Fatigue and corrosion damage are the greatest challenges today in design and life estimation of wind turbine support structures. S355 steel is currently used in fabrication of most wind turbine monopile support structures. Clear understanding of their corrosion-fatigue properties and accurate steel selection will support the optimisation and economic design of extra-large wind turbines. This paper presents the *FCG* test results of advanced S355 TMCP steel in air and *SW* and compares the results with studies on commonly available S355 steel. The results show that S355 TMCP steels generally offer higher fatigue damage tolerance than normalised S355 steels in air and the factor decreases and tends towards a common value with increase in stress intensity factor range. However, in *SW* there is no significant difference in fatigue crack growth rates for all the S355 ferritic steels considered in this study.

5.2 Introduction

This study focused on FCGR test results in air and *SW* (under 0.3Hz) and comparative analysis with other studies in the literature.

The growth of offshore wind industry will depend essentially on reductions in wind energy cost. Large cost reductions can be achieved through efficient, economic and optimized WTSS. To achieve maximum offshore wind potential towards

reduction of green electricity cost beyond 2020, an important step is efficient material selection. Fatigue and corrosion damage are the greatest challenges today in design and life estimation of wind turbine support structures (WTSSs). S355 steel is currently used in fabrication of most of WT monopile support structures. Clear understanding of their fatigue properties and accurate steel selection will support the optimization and economic design of extra-large WTs. This section presents the *FCG* test results of advanced S355 TMCP steel in air and *SW* and compares the result with studies on some common S355 steel. The results show that S355 TMCP steels generally offer high fatigue damage tolerance than normalised S355 steels in air and the factor decreases and tends to a common value with increase in stress intensity factor range. However, in *SW* there is no significant difference in fatigue crack growth rates for all the S355 ferritic steels considered at this preliminary stage.

5.3 Experimental loading conditions

Tests were conducted in air and in free-corrosion (no cathodic protection) conditions using constant amplitude sinusoidal waveform as shown in **Figure 4-2 (c)**, at $R = 0.1$ and loading frequencies of 4Hz and 5Hz and at 0.3Hz in the substituted *SW*. In this preliminary test, the general procedures were as described in Chapter 3, particularly in Section 3-3.

5.4 Air tests results and comparative analysis

In all the tests and comparative analysis in this section, the units m/cycle for da/dN and $\text{MPa}\sqrt{\text{m}}$ for ΔK were used. The experimental results obtained for FCGR tests in air in the present study (PS) are shown in **Figure 5-1**. The result shows that in the Paris Region, load level and frequency appeared to have little or no effect on the FCGR in S355G10+M steel in air within the range of frequencies considered. An interesting observation is that, regardless of the starting a within the range used in this programme or ΔK , similar curve trends were obtained in the Paris Region.

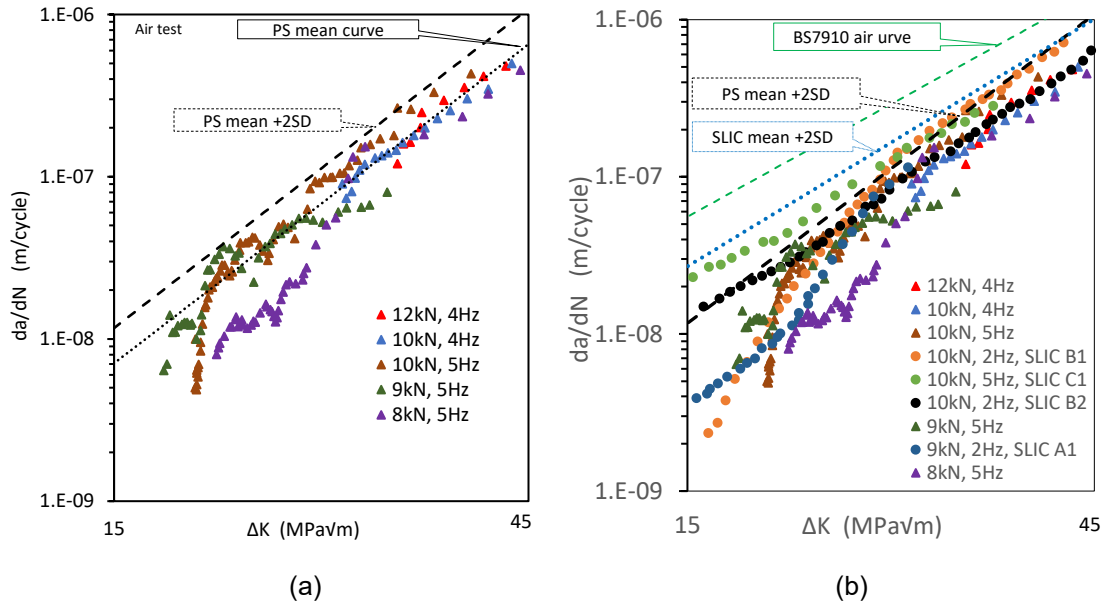


Figure 5-1 FCGR in air (a) for S355G10+M in the present study (b) Comparison of S355G10+M with S355G8+M in the SLIC study

In **Figure 5-1 (a)**, 10kN air test was performed twice using different frequencies and the result shows similar trend in the Paris Region. A mean curve is fitted to the air test results by removing the threshold converging points and treating the whole linear or Paris-region data points as if they were obtained from one test. The non-linear data or data in the Stage I removed are those below 19.13 MPa√m for 10kN, 5Hz; 27.48 MPa√m for 10kN, 4Hz; and 17.11 MPa√m for 9kN. Note that the whole data points are presented in **Figure 5-1 (a)** with the mean line from linear data points inserted. Using simplified Paris-law and shape function solutions as contained in ASTM E647 [172] and BS ISO 12108: 2012 [189], the mean curve is defined by the equation $da/dN = 1.01 \times 10^{-13} \Delta K^{4.12}$ while the power law equation for the mean plus 2 standard deviation (mean +2SD) in air is obtained as $da/dN = 1.66 \times 10^{-13} \Delta K^{4.12}$.

In **Figure 5-1 (b)** the result of the present study (PS) is compared with SLIC joint industry project's [175][190] fatigue data for S355G8+M in air. The C(T) specimen for SLIC project was machined from 90-mm thick EN-10225:01 S355G8+M steel

plate, with inspection certificate of 3.2 as per EN 10204:2004 and certified by Germanischer Lloyd. The chemical composition of the steel subgrade is given in **Table 5-1**.

Table 5-1 Composition (wt%) of S355 steels used in this study

Steel	C	Mn	Ni	Si	Cu	Cr	NB	P	Mo	Ti	V	S
Present study	0.06	1.57	0.33	0.27	0.24	0.034	0.022	0.013	0.006	0.003	0.001	0.001
SLIC	0.053	1.52	0.32	0.27	0.24	0.028	0.020	0.012	0.007	0.001	0.001	0.001
Scott	0.18	1.38	-	0.37	-	-	0.034	0.032	-	-	0.01	0.024
Thorpe	0.17	1.35	0.07	0.35	0.17	0.09	0.03	0.037	0.01	0.01	0.01	0.023
Jesus	0.10	0.64	0.10	0.15	0.38	0.08	-	0.022	0.014	-	0.003	0.041
Dhinakaran	0.09	1.45	0.072	0.38	-	0.07	-	0.004	-	-	-	0.004
Tsay	0.13	1.32	0.03	0.31	-	0.03	-	0.02	-	-	-	0.01
Fe – balance, N = 0.04 (PS & SLIC), Al = 0.05 (Thorpe)												

With regards to the composition, S355G8+M and S355G10+M are fundamentally the same. The SLIC mean plus 2SD curve for air test data is described by the equation $da/dN = 3.54 \times 10^{-12} \Delta K^{3.3}$ and this curve is re-produced in **Figure 5-1 (b)** while the SLIC mean curve equation for the air data is given as $da/dN = 2.83 \times 10^{-12} \Delta K^{3.23}$. Notice that SLIC tests were done at different test centres for the same loading condition, 10kN, 2Hz. It can be seen that the FCG gradient in the present study (PS) on S355G10+M is more inclined and showing more resistance to FGC than the SLIC S355G8+M especially at low ΔK . At higher values of ΔK , the crack growth rates for the two subgrades appears to converge and fall close to each other. Studies have shown that little scatter is observed for FCGR of steel in air. Considering this, the SLIC mean + 2SD curve appears to capture the upper scatter bound for the S355 TMCP steels. BS7910:2013+A1:2015 recommended designing steel structures using FCGR of $\frac{da}{dN} = 1.65 \times 10^{-11} \Delta K^3$ in air for steels with yield strengths less than 700 MPa. This curve has been reproduced in **Figure 5-1 (b)**. It is obvious that the design curve is much more conservative for the air data.

An interesting study on S355 steel was carried out by Thorpe, et al. [24]. They investigated the *FCG* properties of steel manufactured to BS4360 Grade 50D steel in air. This steel is equivalent to BS4360:1990 S355J2+N. They extracted

their C(T) specimen from 38 mm thick plate. The composition of this steel is given in **Table 5-1**. The quoted mechanical properties of the steel are: $\sigma_y = 370\text{MPa}$, $\sigma_{UTS} = 538\text{MPa}$, Elongation (%) = 32. The test was performed in laboratory air condition with frequency from 1 – 0.85Hz. The result from this work stated that the stress ratio, R, had little or no effect on FCGR in air and that FCGR in air does not depend on the specimen thickness. The 2-stage curve, as recommended in [174], was defined using the mean curve line. Their FCGR for Stage A was given as $da/dN = 3.26 \times 10^{-12} \Delta K^{3.27}$ as shown in **Figure 5-2** while that of Stage B was $da/dN = 2.44 \times 10^{-11} \Delta K^{2.67}$.

Scott, et al. in 1983 [154] also studied S355J2+N (BS4360 Grade 50D) steel. The C(T) for this experiment has thickness of 37.5mm extract from the steel plate in L-T (Longitudinal-Transverse) orientation. The plate was hot-rolled and normalized, with composition as shown in **Table 5-1**. **Table 5-1** assumes that the C(T) was extracted from 76mm plate. The quoted mechanical properties of the steel are: $\sigma_y = 360\text{MPa}$, $\sigma_{UTS} = 545\text{MPa}$, Elongation (%) = 26 and Charpy impact value (J) at $-30^\circ\text{C} = 117$. The test was performed in laboratory air condition. In Scott's work [154], cyclic frequency and stress ratio (in the range 0.85-1.0) had little or no influence on the FCGR in air. 2-stage Paris-law trends was used to describe the FCGR of the whole test in air. The FCGR for Stage A is given as $da/dN = 5.89 \times 10^{-12} \Delta K^{3.15}$, while that of Stage B is $da/dN = 2.80 \times 10^{-11} \Delta K^{2.65}$. The FCGRs are essentially the same to that of Thorpe, et al. and so not reproduced.

A.M.P. de Jesus, et al. [167] also studied fatigue growth in S355 steel. The subgrade name and the production history were not stated. The mechanical properties of the actual steel studied was quoted to be $\sigma_y = 419\text{MPa}$ and $\sigma_{UTS} = 732\text{MPa}$. It should be noted that tensile property of S355 is usually given in yield stress range of 470 – 630 MPa for $t(\text{mm}) \leq 100\text{mm}$ and 460 – 620 MPa for $t(\text{mm}) > 100\text{mm}$. The chemical composition of the material is shown in **Table 5-1**.

Using a C(T) specimen manufactured to ASTM E647 standard, FCG test was performed. Four stress ratios were used by de Jesus et al: 0, 0.25, 0.5 and 0.75.

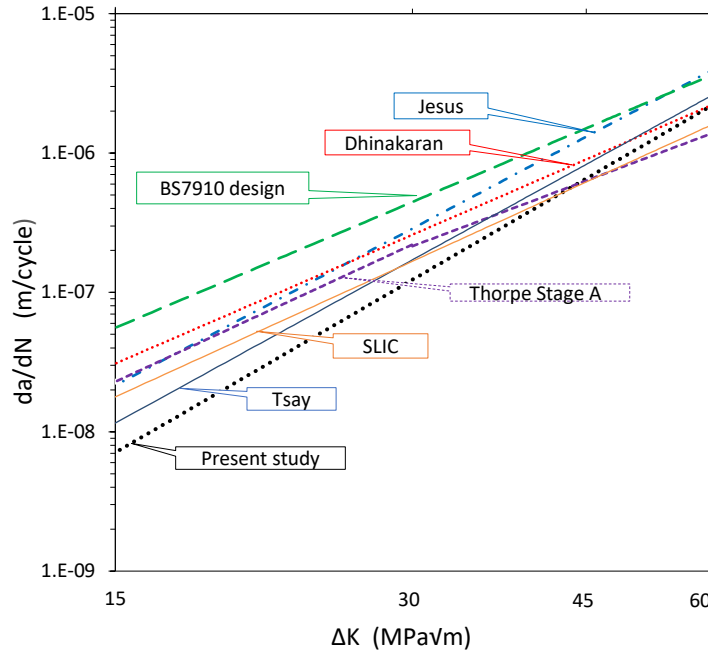


Figure 5-2 Comparison of FCGR in air for S355 steel subgrades

Table 5-2 FCGR power law constants for various S335 grades considered in this study

Study in air	FCGR (da/dN)
Present (TMCP)	$1.01 \times 10^{-13} \Delta K^{4.12}$
Tsay (TMCP)	$3.07 \times 10^{-13} \Delta K^{3.89}$
SLIC (TMCP)	$2.83 \times 10^{-12} \Delta K^{3.23}$
Jesus (Unknown)	$8.58 \times 10^{-13} \Delta K^{3.74}$
Thorpe* (Normalized)	$3.26 \times 10^{-12} \Delta K^{3.27}$
Scott* (Normalized)	$5.89 \times 10^{-12} \Delta K^{3.15}$
Dhinakaran (Unknown)	$7.56 \times 10^{-12} \Delta K^{3.07}$
BS7910 air design curve	$1.65 \times 10^{-11} \Delta K^{3.0}$
*Stage A only is included for Scott and Thorpe	

The crack growth was measured using a magnifying device with resolution of 1 μm . The frequency of test was reported to be 20 Hz, which was reduced at region of high FCGR, taken to mean about 0.3 mm per 1000 cycles. The results show that the stress ratios again have no effect on the FCGR in air. However, he pointed out that there was crack closure effect for tests done about zero (0) stress

ratios and this reduced the effective ΔK applied, hence decreased FCGR was obtained. In fact, the test they performed at $R = 0$ gave a decreased FCGR. The crack closure effect reduces with increase in R , so that the applied ΔK was fully effective. **Figure 5-2** shows the mean air Paris curve for the study. The FCGR was given as $8.58 \times 10^{-13} \Delta K^{3.74}$ where da/dN is in m/cycle and ΔK in $\text{MPa}\sqrt{\text{m}}$.

Tsay [191] studied the FGCR of a marine EH36 steel whose composition is also shown in Table 10 and the yield strength given as 420MPa. The steel was thermo-mechanically-control processed, commonly specified as fine-grained high tensile steel and generally used in the design of ship bodies. For normalized condition, the nearest equivalent of this steel is S355J2+N. The test result selected in this analysis is the test carried out such that the crack growth direction was normal to the rolling direction (LT specimen). The result of the FGR test is also included in **Figure 5-2**.

A standard C(T) manufactured as per ASTM E647 was used by Dhinakaran and Prakash [192] to study the corrosion-fatigue property of a steel with a composition as shown in **Table 5-1**. The subgrade name was not given. The mechanical property of the steel is given as $\sigma_y = 459\text{MPa}$, $\sigma_{UTS} = 530\text{MPa}$ and elongation 33%. The test performed under constant frequency amplitude and at constant ΔK is also included in **Figure 5-2**.

The mean FCGR power law curve constants for the various S355 grades considered in this study are shown in **Table 5-2**. The table shows increase in the Paris exponent in the TMCP steels, which implies that the TMCP curves are more inclined. In general, this study shows that S355 TMCP steels offer high fatigue damage tolerance than conventional S355 steels in air. In other words, FCGR is lowest in the TMCP steel compared with common steels studied.

Reliable factor of difference between common and TMCP S355 steel will be established by considering more tests and this was considered in the later

studies. But, clearly TMCP steels have superior resistance to fatigue crack propagation in air for S355 steel, and the factor decreases and tends to common value with increase in ΔK . Thus, in terms of structural application in air environment, where fatigue is the dominant issue selection of TMCP steels will be preferable. **Figure 5-2** also shows the inclusion of BS7910 curve (green curve) to the fatigue tests plots. It can be seen that the current BS7910 design curve in air is much more conservative for TMCP S355.

5.5 Seawater test results under 0.3Hz and comparative analysis

The CFCGR results obtained for freely corroding conditions in *SW* for the present study are shown in **Figure 5-3 (a)**. The mean curve is described by the equation $da/dN = 2.76 \times 10^{-12} \Delta K^{3.35}$ while the mean+2SD curve is $da/dN = 5.49 \times 10^{-12} \Delta K^{3.35}$. The mean+2SD satisfactorily enclosed all the data points in the Paris Region.

Figure 5-3 (b) compares the CFCGR of present study with SLIC project study [175] on S355G8+M. The SLIC tests were carried out at different centres, so tests under the same loading condition but in different centres have been included. The CFCGR for the mean+2SD curve, which serves as the upper scatter bound for the SLIC project is given as $\frac{da}{dN} = 1.25 \times 10^{-12} \Delta K^{3.86}$. What is apparent for the TMCP steels in **Figure 5-3 (b)** is that the difference in the mean load appears not to have a defined effect under the 0.3Hz. Generally, if material variability is considered, it can be concluded that CFCGR in the two steel subgrades are the same at frequency of 0.3Hz.

It is common observation that FCGR in *SW* shows much greater scatter than that in air, especially at low ΔK . The SLIC mean+2SD curve then appears to describe very well the upper limit scatter bound for the S355 in *SW*. For steel structures in marine environment with temperature up to 20°C, with or without cathodic protection, BS7910:2013+A1:2015 recommended using simplified Paris law

$da/dN = 7.27 \times 10^{-11} \Delta K^{3.0}$ as the CFCGR, where da/dN is in m/cycle and ΔK in $\text{MPa}\sqrt{\text{m}}$. The current BS7910:2013+A1:2015 seawater upper bound design curve has been included in **Figure 5-3 (b)**. It can be seen that the current simplified BS design recommendation curve for SW is very much over conservative for the thick S355 TMCP steel.

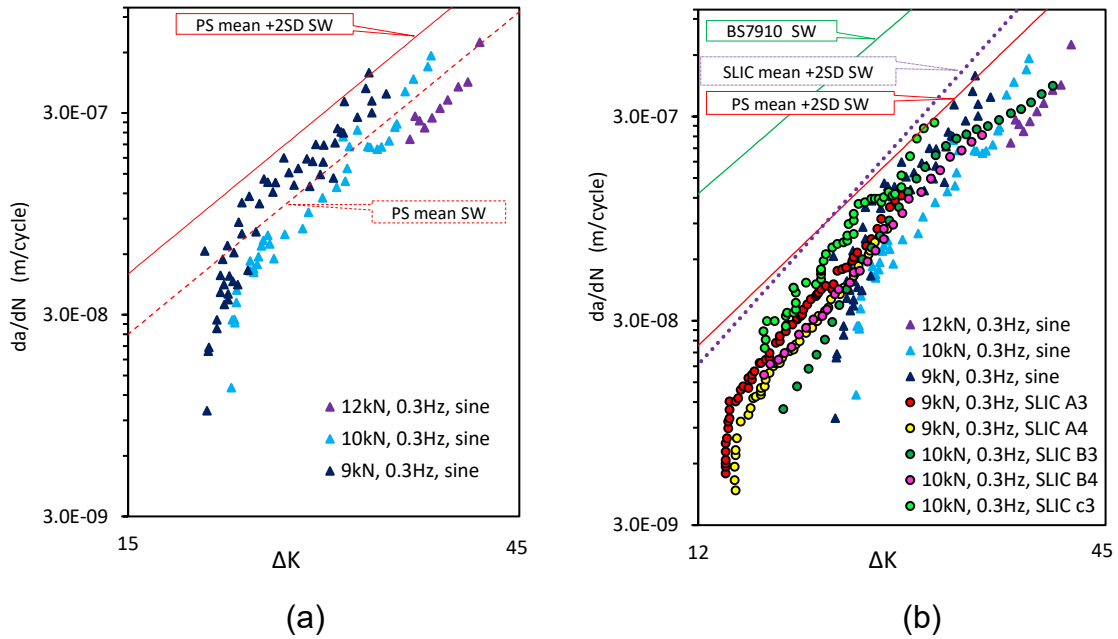


Figure 5-3 CFCGR of S355 in seawater (SW) under free-corroding conditions (a) CFCGR for S355G10+M in SW (b) comparing CFCGR of S355G10+M with SLIC study on S355G8+M in SW

Figure 5-4 compares the S355 TMCP data presented in **Figure 5-3 (b)** with works of Thorpe, et al. [24] and Scott, et al. [154]. It is pertinent to point out that Scott's SW test on S355J2+N was performed at higher stress ratio $R = 0.5$, and lower frequency 0.1Hz, temperature range 5 – 10 °C, and using triangular waveform at freely corroding potential (- 0.65V). A sinewave of 0.1Hz approximates to sea-wave frequency. The use of sea-wave frequency was an attempt to obtain fatigue data for application in offshore oil/gas production and exploration platforms. This would then enable extrapolation of the laboratory data to field structures. Severally studies have shown that triangular waveform test results are

comparable to those obtained from the sinewave tests [154]. The Thorpe's test was performed using $R = 0 - 0.1$, sinewave and under free corroding condition.

As seen in **Figure 5-4**, the data points obtained from different test programmes fall upon or close to each other with slight differences which can be attributed to inherent material variability. In general, the TMCP subgrades appear not to be superior to the S355 steels investigated by Scott and Thorpe.

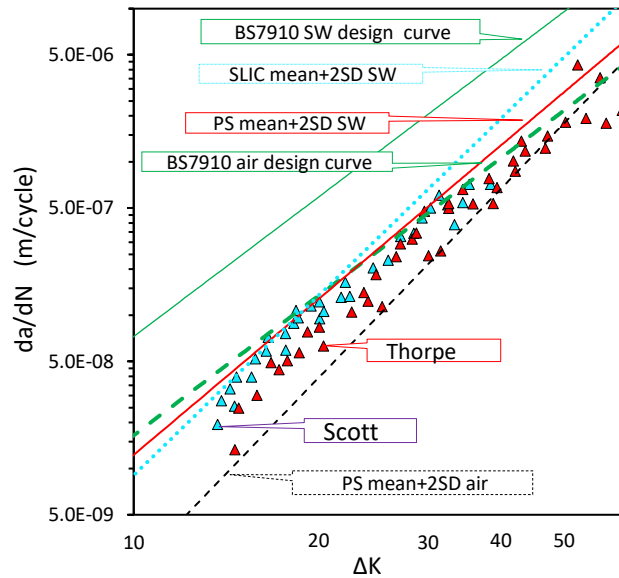


Figure 5-4 CFCGR of S355 in seawater (SW) under free-corroding conditions

Table 5-3 Upper bound FCGR power law constants in seawater

Study in SW	FCGR (da/dN)
PS mean+2SD	$5.49 \times 10^{-12} \Delta K^{3.35}$
SLIC mean+2SD	$1.25 \times 10^{-12} \Delta K^{3.86}$
BS7910 SW design curve	$7.27 \times 10^{-11} \Delta K^{3.00}$

Thus, in structural application in SW environment, where fatigue is the dominant issue, this study tends to suggest that there is no justification for the market price difference between thick S355 TMCP and normalized S355. This can only be concluded by considering many more data for normalised steels. **Table 5-3** lists

the upper bound (mean+2SD) FCGR power law constants in SW for the present study (PS), SLIC and that of BS7910 recommendation curve in SW.

5.6 Environmental reduction factor (ERF)

Figure 5-4 also shows the comparison of the mean+2SD curves of the present study with current recommendation curves in BS7910 for tests in air and SW. To explain the figure, the environmental reduction factor (ERF) at equal interval of ΔK for the present study is presented in **Table 5-4**. Design conservative factor (DCF) has been used here to make comparison with the BS7910 curves.

Table 5-4 Environmental reduction factor (ERF) and Design conservative factor (DCF)

Present study (ERF)		BS7910 air DCF (PS)		BS7910 SW DCF (PS)		BS7910 SW DCF (SLIC)	
ΔK	ERF	ΔK	DCF-air	ΔK	DCF -SW	ΔK	DCF -SW
20	3.34	20	3.46	20	4.68	20	4.54
30	2.35	30	2.22	30	4.10	30	3.13
40	1.92	40	1.57	40	3.62	40	2.42
50	1.62	50	1.25	50	3.44	50	2.02
Ave.	2.31	Ave.	2.13	Ave.	3.96	Ave.	3.03

By analysing the crack growth rates in air and SW using the fitted mean+2SD bounds in the ΔK , 20 – 50 MPa $\sqrt{m}$, ERF is seen to decrease with increase in ΔK and an average reduction factor of 2.31 is obtained. In other words, the SW generally accelerated the FCGR by a factor of 2.31. The DCF between air test and BS7910 curve for fatigue design in air environment is 2.13. Since scatter in fatigue data in air is minimal compared to test in aggressive medium, the current BS7910 air curve satisfactorily and conservatively defines the upper FCGR bound for all the S355 steels. An average DCF of 3.96 for the present study and 3.03 for SLIC are obtained between SW test and BS7910 curves for design in SW. It can be seen that the BS7910 curve in SW is much more conservative for the S355 steels considered. This increase in conservativeness will reduce the true lifespan of modern WT and impact unfavourably the LCOE.

5.7 Conclusions

This section presents the FCGR performance of TMCP S355 steel in air and SW and compared the results with studies on some common S355 steel sub-grades. The tests in this study were performed in accordance with BS 7910:2013+A1:2015 and ASTM E647-15. Constant amplitude sinusoidal waveform was used at load ratio, R of 0.1 and loading frequencies of 4Hz and 5Hz in air, and at 0.3Hz in artificial SW. Air tests were carried out at laboratory ambient temperature that averaged about 20°C, while the SW environment temperature varied from 7.5 to 8.2 °C. Crack lengths were measured in air using digital cameras and, in some cases, travelling microscope. Back Face Strain (BFS) method was used to measure crack growth in SW. The following conclusions have been drawn from this study:

1. The FCGR in advanced S355, particularly the subgrades S355G8+M and S355G10+M steels are essentially similar in SW. However, S355G10+M examined in this study generally shows more resistance to FCG in air compared to S355G8+M particularly at lower values of ΔK .
2. The general upper bound Paris law describing the FCGR in thick S355G10+M steel in air can be stated as $\frac{da}{dN} = 1.66 \times 10^{-13} \Delta K^{4.12}$ while that of SW is found to be $\frac{da}{dN} = 5.49 \times 10^{-12} \Delta K^{3.35}$.
3. With respect to the mean+2SD upper bound curves, the FCGR in S355G10+M steel in SW is higher than that of the air by an average factor of 2.31 throughout the Paris Region in the range of ΔK : 20 - 50 MPa $\sqrt{m}$.
4. In general, this study shows that S355 TMCP steels offer high fatigue damage tolerance than normalized S355 steels in air and the factor decreases and tends to common value with increase in ΔK . In terms of structural application in air environment, where fatigue is the dominant issue selection of TMCP steels will be preferable.

6 Critical assessment of the FCGR sensitivity to material microstructure in ferrite-pearlite steels in air and marine environment

6.1 Abstract

This paper presents a critical assessment of the effect of microstructure on the fatigue crack growth rate (FCGR) in ferrite-pearlite (α -P) steels in air and marine environment. This was done by evaluating the resistance of three subgrades of S355 steel, produced by normalized-rolling, NR (S355J2+N) and thermo-mechanical control process, TMCP (S355G8+M and S355G10+M), to FCG in air and SW. The data from the tests were then compared with results of several studies on α -P steels of similar compositions and microconstituents in the literatures. It was found that microstructure strongly affected the Paris region of the da/dN vs. ΔK sigmoidal curve in both media. This is contrary to the current believe that microstructure has little or negligible effect in the linear region of the FCGR curve in air. TMCP and quenched-and-tempered (QT) steels offered better resistance to FCG than NR steel in both media. TMCP steel offered the highest resistance to FCG. The crack path was non-planar with complex crack front. The tendency for crack deviation, branching, wedging action by metal crumbs and perhaps interlocking majorly explained why the FCGRs of the TMCP and QT steels are lower than that of NR. The result also suggests that ductility and yield stress do not have strong retarding effect on FCG in α -P steels at low ΔK . This finding has a fundamental implication on analytical and numerical models that wholly depend on mechanical properties of the steel and assumes plane horizontal crack propagation front in predicting FCGR in steels. Such models may likely fail in predicting FCGR even in a subgrade of the same grade of steel.

6.2 Introduction

There are many studies that claimed that microstructure has little or negligible effect at the intermediate fatigue crack growth rate (FCGR) in metallic materials provided that the material is not in a corrosive medium [29][193][194][195][196][197][198]. It appears no study has proved beyond doubts this claim for steel materials. Even information on the exact mode variation of microstructural phases affect the FCGR in the Paris Region in corrosive environment is scarce. What is often reported is that fatigue damage depends strongly on the local microstructural features such as the grain size and geometry, the crystallographic orientation relationship and the grain-boundary structure in the threshold region or early stages of fatigue crack development. These local features are reported to play little or no part when the crack is sufficiently large in air. The regions of microstructural and other factors influence are commoly presented as in **Figure 6-1**.

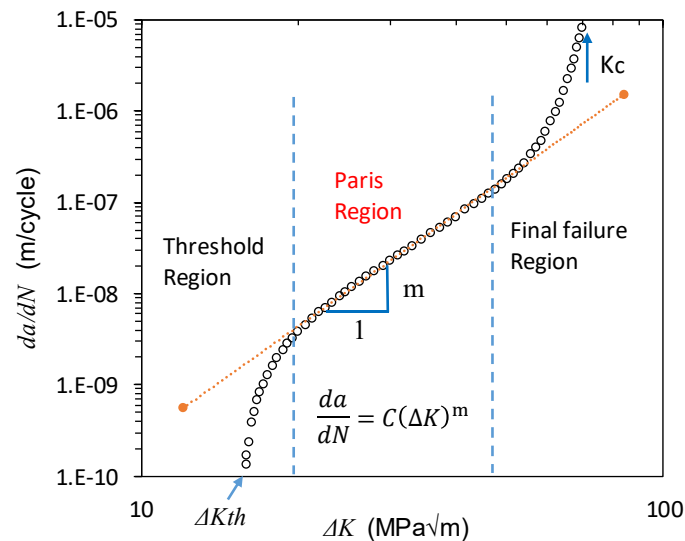


Figure 6-1 Schematic of FCGR, da/dN vs. cyclic stress intensity factor range (SIFR), ΔK in steels. (See [160][187][188]).

Mechanical properties have also been found to strongly influence the threshold and region of accelerated failure. Stonesifer [199] in his study concluded that

grain size affected sub-critical crack growth in A533 B steel only to the extent to which it affected fracture toughness, K_{Ic} and the onset of rapid fracture and that grain size did not affect the FCG threshold, ΔK_{th} . In **Figure 6-1**, microstructure, environment and mean stress are reported to have a large influence on the near threshold behaviour of metallic materials [29][187][197][199][200][201]. For the final failure region, microstructure, mean stress and thickness have a large influence while the environmental influence is little.

The crack path in fatigue is generally taken to be across the grains (transcrystalline or transgranular), although they may also propagate along the grain boundaries (intercrystalline or intergranular) depending on the material properties, loading and environmental conditions [202]. It is often reported also that at low strains, at least 90% of the fatigue lives of unnotched components is spent in crack initiation and about 10% for the rest of the stages [203][204]. This statement is not strictly correct as steel materials, no matter how well they are produced contain micro or subcritical equilibrium defects. Fatigue intrusion and extrusion start from these points of microdefects at the material surface or subsurface. If the steel is welded, then the scale of defect increases. Hence, the useful life of these structures is determined by the propagation of subcritical cracks to critical dimensions to cause failure [205]. Consequently, an understanding of FCGR in these steels provides information for material selection, prediction of the service life and planning of inspection intervals for steel structures operating in marine environment.

An important factor that affects the mechanism and fracture properties of steels is the size and arrangement of the grains in the microstructure. The strength and toughness also depend on the size, texture and shape of grains in the ferrite-pearlite (α -P) steel. Decreasing the grain size has the capacity to increase both the strength and the toughness of a steel simultaneously [105]. It has been reported that a fine-grained microstructure confers high mechanical strength, formability and toughness [104][203][206][207]. Because of these improved mechanical properties, grain refinement through thermomechanical control

process (TMCP) has been used in the design of advanced marine steels - for example the S355 M-series grades, for offshore applications. The steels listed in **Table 6-1** are regarded as the modern material choice for marine environment [42]. They have been manufactured specifically for offshore pipelines, platforms, pressure vessels, and modern WT installations [42].

Table 6-1 Mechanical properties of advanced M series S355 steel subgrades for offshore application [144]

Offshore Structural Steel Plates	
European	Typical offshore application
EN10225: Plates & tubulars	
S355G7+N	Primary Structure
S355G7+M	Primary Structure
S355G8+N	Critical Joints
S355G8+M	Critical Joints
S355G9+N	Primary Structure
S355G9+M	Primary Structure
S355G10+N	Critical Joints
S355G10+M	Critical Joints

The fatigue crack growth rate (FCGR) data for these new materials are very scarce, if not non-existent. In the recent study [144], the authors found that S355 TMCP steels generally offered higher fatigue damage tolerance than normalised S355 conventional steels in air, but in *SW* there is no significant difference in FCGRs for all the S355 ferritic steels considered in the study. This conclusion was made by considering the whole da/dN vs ΔK curve between ΔK of 10 - 50 MPa $\sqrt{m}$. Since the steels under consideration have similar microconstituents and the experiment performed under similar conditions, the differing values in the FCGR in air suggests that microstructure may have affected the crack propagation in the Paris Region. This present study attempts to validate this conclusion by expanding and critically analyzing this trend in a great detail and to examine the role of microstructure on the FCGR behavior in α -*P* steels both in air and *SW*, particularly in the Paris Region. This is done by comparing the authors' work with the FCGR results on α -*P* TMCP, quenched & tempered (QT) and

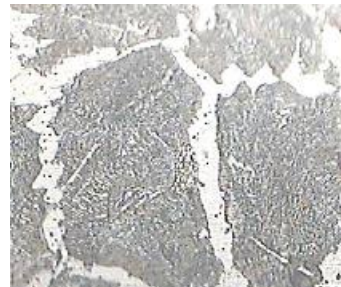
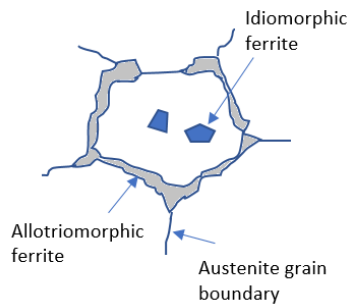
normalized-rolled (NR) steels obtained by various researchers which are available in the literatures. This empirical analysis is extremely important because it enables the design, selection and use of steels for offshore applications. It also supports realistic modelling and corrosion-fatigue life prediction of material employed in marine structures.

6.3 Chemical and phase composition of ferrite-pearlite steels studied

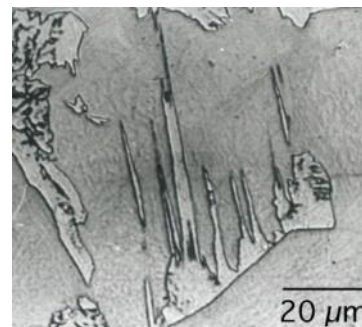
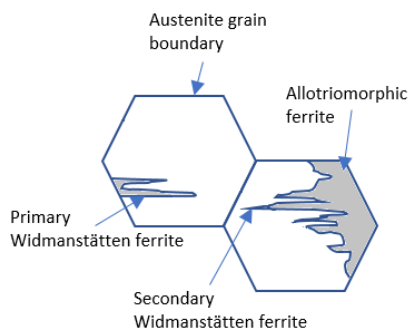
Fundamentally, a steel containing about or less than 0.10% carbon is commonly referred to as a ferritic steel. Ideally, it has a microstructure consisting of very small volume fraction of pearlite in a ferrite matrix. It may be called α - P steels where volume fraction of pearlite phase is significant. In this study, we are using the term ' α - P steel' because we are comparing large FCGR data from steel microstructures containing varying amounts of ferrite and pearlite phases. The ferrite phase is soft and ductile, while the pearlite is hard and brittle. As the carbon content increases, the volume fraction of pearlite increases while that of ferrite decreases. As the volume fraction of pearlite increases the bulk strength increases and ductility decreases. The morphology of the α - P phases depends on the processing route. Austenitization is the first operation in many of the most important deformation and heat treatment processes. After austenitization, upon cooling the austenite decomposes into various morphologies. The first is the ferrite morphology. Other microstructures that form as the transformation temperature is lowered are pearlite, bainite, and martensite. Ferrite-pearlite phases are of interest in this study. Any other phase that might be present for the purposes of clarity is neglected.

The common ferrite morphologies are grain boundary allotriomorphic ferrite, idiomorphic ferrite, Widmanstätten ferrite, and intragranular ferrite or acicular ferrite [208]. An allotriomorph nucleates at the austenite grain surfaces, forming layers which follow the grain boundary contours and the Idiomorphs nucleate without contact with the austenite grain surfaces as shown in **Figure 6-2 (a)**.

Another morphology of ferrite is Widmanstätten. Primary Widmanstätten ferrite appears to grow directly from austenite grain surfaces while secondary Widmanstätten ferrite develops from the existing allotriomorphic ferrite in the microstructure as shown in **Figure 6-2 (b)**.



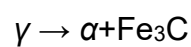
(a)



(b)

Figure 6-2 Ferrite morphologies: (a) allotriomorphic & idiomorphic ferrite (Note: the micrograph is showing only the allotriomorphic ferrite) (b) primary and secondary Widmanstätten ferrite [208][209]. The white etching is ferrite.

The Pearlite is a product of eutectoid reaction, which occurs in an austenite with composition exactly 0.8 wt% carbon cooled below the 723°C and it is often depicted as:



where γ is the austenite and α the ferrite. Pearlite appears as colonies of an interpenetrating bi-crystal of ferrite and cementite ($\alpha + \text{Fe}_3\text{C}$). It was formally thought that pearlite consists of many colonies of a thin lamellar, alternating mixture of ferrite (iron) and cementite (iron carbide) or ($\alpha + \text{Fe}_3\text{C}$) [118]) as shown in **Figure 6-3**. However, this idea is still in use today to describe pearlite.

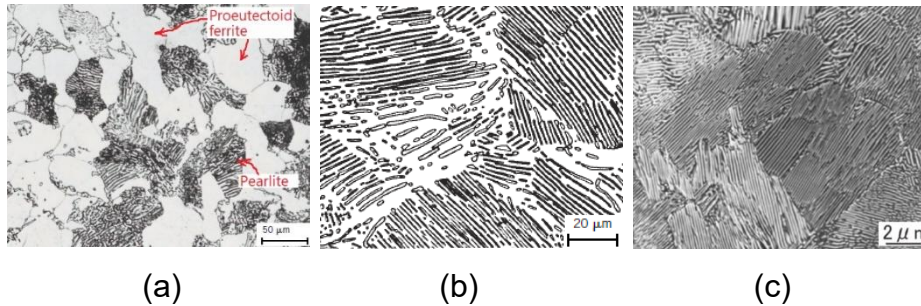


Figure 6-3 Photomicrographs of pearlite. (a) pearlite colonies and pro-eutectoid ferrite in an Fe–C–Mn steel (b) coarse pearlite (c) fine pearlite [208].

Figure 6-3 (a) shows a typical photomicrograph of a pearlite in α -*P* steel (Fe–C–Mn steel). Several distinct colonies can be seen (dark etching). The pearlite can be coarse or fine, as shown in **Figure 6-3 (b)** and **(c)** respectively, depending on composition and the temperature of transformation. This present study seeks to discover how variations in these phase morphologies affect FCGR in air and in SW.

6.4 Composition, mechanical properties and microstructure of ferrite-pearlite steels studied

The chemical composition in wt% (approximated to two decimal places) of the steels studied and compared in this study is given in **Table 6-2**. Where composition of S355 is not given in the literature, the general composition has been assumed to be the same as that given in **Table 6-2**. The composition ranges for ferritic steels in **Table 6-2** is shown in **Table 6-3**. **Figure 6-4** is a schematic phase diagram showing the range of carbon compositions of steel studied and phases expected in the microstructure of the steels. Generally, the phases in these steels are ferrite and pearlite ($\alpha + \text{Fe}_3\text{C}$).

Some typical microstructures in a hot normalized-rolled (NR), TMCP-UFGS and quenched-tempered (QT) steels are shown in **Figure 6-5**. The grade of steel and the reference sources are given in **Table 6-2**.

Table 6-2 Chemical compositions of the ferritic steels studied

Study	Steel Grade	C	Mn	Ni	Si	Cu	Cr	Mo	E %	YS (MPa)	UTS (MPa)
Present Study [144]	S355G10+M (TMCP)	0.06	1.57	0.33	0.27	0.24	0.03	0.01	36	435	545
Mehmanparast [175][190]	S355G8+M (TMCP)	0.05	1.52	0.32	0.27	0.24	0.03	0.01	35	447	549
Scott [154]	BS4360-50D (NR)	0.18	1.38	-	0.37	-	-	-	32	370	538
Thorpe [24]	BS4360-50D (NR)	0.17	1.35	0.07	0.35	0.17	0.09	0.01	32	370	538
Jesus [167]	EN 10025 S355 (NR)	0.10	0.64	0.10	0.15	0.38	0.08	0.01	22	419	766
Dhinakaran [192]	C-Mn steel (NR)	0.09	1.45	0.07	0.38	-	0.07	-	32	459	530
Tsay [191]	EH36 (TMCP)	0.13	1.32	0.03	0.31	-	0.03	-	21	420	-
Atkinson [150]	BSI501-213-32A (NR)	0.22	1.60	0.30	0.55	0.20	0.25	0.10	28	320	490
*Bhuyan [152]	CSA G40.21M 350 WT (NR)	0.22	1.50	-	0.40	-	-	-	29	405	513
Appleton [157]	BS4360-50D (NR)	0.13	1.22	-	0.32	-	-	-	32	368	508
Jones [155][156]	Naval (QIN) steel (QT)	0.17	0.34	2.36	0.21	0.08	1.23	0.34	27	625	735
Xiong [210]	Q345 steel (NR)	0.15	1.54	-	0.34	-	-	-	26	370	585
Laurito [211]	RD480 steel. (isothermally grain-modified)	0.09	1.17	0.01	0.05	0.01	0.02	0.01	32	459	541
Cheng [159]	ABS grade EH36 (NR)	0.12	1.39	0.03	0.38	0.05	0.05	0.01	21	350	-
Callister [212]	Kontroll-50-FSM (TMCP)	0.05	1.13	0.29	0.23	0.24	0.02	0.20	-	470	-
Callister [212]	BS 4360 G55F (QT)	0.12	1.18	0.03	0.35	0.02	0.02	0.14	-	580	-
Callister [212]	N-A-Xtra (QT)	0.15	1.01	0.09	0.62	0.06	0.78	0.38	-	690	-
Adedipe [149]	EN10025-2 S355J2+N (NR)	0.22	1.60	0.50	0.55	0.55	0.03	0.10	20	345	-
Chapetti [213]	TMCP-Ultra-Fine-Grained Steel (UFGS)	0.15	1.80		0.20	-	-	-	14	696	1183
Musuva, [168]	BS4360-50D (NR)	0.18	1.40	0.10	0.36	0.16	0.11	0.02	30	386	560
Thompson [214]	BS4360-50D (NR)	0.15	1.38	0.10	0.41	0.02	0.11	0.01	30	386	536
Achilles [153]	SA508 cl III forged RPV steel (grain-refined) (QT)	0.16	1.38	0.77	0.25	0.06	0.26	0.52	24	468	611
Gill [215]	HSLA (MF-80) (NR)	0.07	1.46	-	0.30	-	-	-	-	653	742
Barsom A [205]	A537 -A (NR)	0.16	1.26	0.23	0.32	-	0.12	0.05	34	407	572
*Barsom B [205]	A302 – B (NR)	0.19	1.40	0.03	0.16	-	0.07	0.49	26	386	607
	ABS – C (NR)	0.20	0.60	0.02	0.23	-	0.07	0.01	36	269	434
	A36 (NR)	0.24	1.66	0.03	0.02	-	0.06	0.01	32	248	517
R = rolled normalized TMCP = thermomechanical control process *Barsom B is the combination of A302 – B, ABS – C and A36 data points *Bhuyan steel CSA G40.21M 350 WT is equivalent to EN 10250-2: S355J2G3 hot rolled normalized steel. Note that BS4360 Grade 50D is equivalent to S355J2+N hot rolled normalized steel. †The mechanical properties of SA508 cl III in Achilles study is obtained from [216] Average %wt of other minor elements for all the steels: 0.01Ti, 0.02V, 0.01N, 0.02S, 0.02P, 0.03Al, 0.04NB *Gill steel HSLA (MF-80), contains 0.10% Columbium											

Microstructural examination of S355J2+N shows that, 80% of the microstructure is allotriomorphic ferrite and others pearlite, with traces of acicular or

Widmanstätten ferrite [119]. Hardness was measured to be around 168HV_{0.1} with equiaxed grains of average size of 15µm, measured in the through-thickness direction.

Table 6-3 Composition Ranges of ferrite-pearlite steels in this study

C	0.05 - 0.24
Mn	0.34 - 1.80
Ni	0.01 – 4.66
Si	0.02 – 0.62
Cu	0.01 – 0.55
Cr	0.02 – 1.23
Mo	0.01 - 0.52

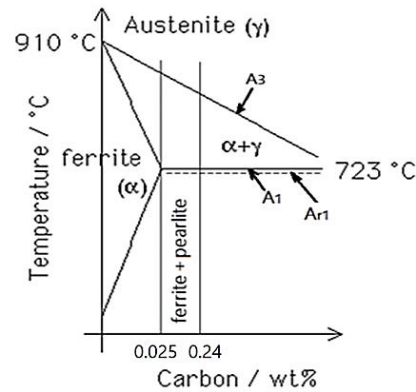


Figure 6-4: A schematic phase diagram of carbon composition range and phases expected in the microstructure of a ferritic steel Photomicrographs

It is reported that the rolled S355 steel had solute segregations and the pearlite and ferrite in this zone were harder and extends up to 400 µm [119]. Note that BS4360 Grade 50D is equivalent to S355J2+N hot rolled normalized steel. **Figure 6-5 (a to i)** confirm similar nature of this type of microstructure. **Figure 6-5 (i)** is MF-80 showing large volume fraction of ferrite. **Figure 6-5 (j)** is the QT microstructure of SA508 class 3 steel obtained under the cooling rate of 10.1+ 11.4°C/min (double-quench and tempered) [217].

Figure 6-5 (k, l & m) are the microstructures of α -P TMCP steel at the same magnification. These TMCP microstructures were taking from the short-transverse sections and are very much alike. **Figure 6-5 (k)** and **(l)** are the microstructures of S355G8+M and S355G10+M (TMCP) subgrades respectively.

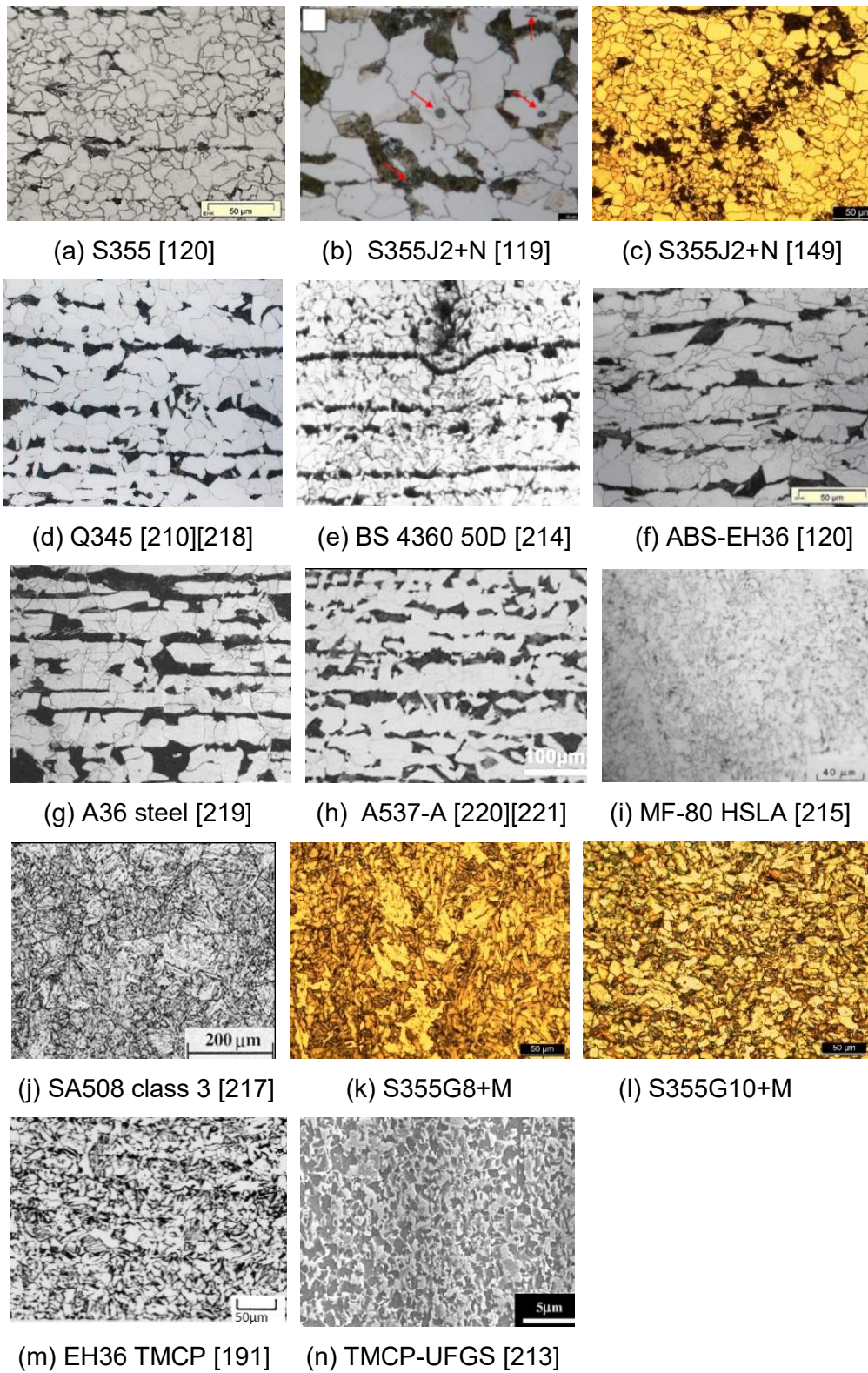


Figure 6-5: Typical microstructures of low alloy α -P steels

These microstructures, including that of **Figure 6-5 (c)** are part of ongoing project by the authors to ascertain the integrity of S355 thick plate subgrades. **Figure 6-5 (n)** is the scanning electron micrograph of the ultra-fine grains of steel produced through TMCP route [213]. The ferrite grain size was reported to be as low 1 μm . Low-carbon/low-alloy steels are prone to banding. Banding is the alignments of pearlite in the rolling plane. The banding can also be attributed to segregation of solutes in the last regions of the steel melt to solidify during the cooling of steel from the molten state. In low-carbon/low-alloy steels, solidification starts as delta δ – ferrite. As the volume fraction of this delta phase increases, solutes such as Mn, Si, P and S partition into the interdendritic regions, which then solidify later with a higher than average concentration of these solutes [123]. Successive deformation during the hot rolling causes these regions to spread out as bands. Since other alloying elements are relatively very low in ferritic steels, the segregation of concern will be mainly that of Mn – a substitutional element, although other solutes such as Ni, Cr, Si can be found in the pearlite band phase [122]. The arrows in **Figure 6-5 (b)** are pointing at some inclusions such as MnS [120]. The banding is common in hot-rolled normalized steels as can be seen in **Figure 6-5 (d to h)**. It is clearly seen in the longitudinal-short transverse (LS) section of rolled product [122][121]. For extra-large (XL) steel plates, this banding may possibly be seen on the outer region while the inner material is essentially non-banded. In fact, due to production technique and sample thickness, the banding phenomenon has been eliminated in the 90mm TMCP steel plates and grain sizes of the microstructures are well refined as can be seen in **Figure 6-5 (k, l & m)**.

In hot-rolled normalized ferritic steels, the separation of ferrite and pearlite phases into bands creates different local microstructural properties. The presence of a hard phase in a soft matrix is expected to influence FCG behaviour in steel. The difference in the local properties due to two different phases with dissimilar stiffnesses is likely to affect the rate at which a crack grows in these steels [124]. A study reported that banding contributes to lower FCGR and higher threshold values [124]. The orientation of the pearlite banding in the studied steel was seen

to offer higher resistance to FCG [124]. The pearlite bands were reported to be responsible in arresting the crack propagation as shown in **Figure 6-6 (a)**. The arrow in **Figure 6-6 (a)** shows where the crack tip was said to be arrested. The crack deviations and branching mode is shown in **Figure 6-6 (b)**. The crack branching was reported to occur after passage of the crack tip through the pearlite band [124].

It was also reported that the extensive crack branching reduced the fatigue crack propagation driving force at the local crack tip. The next sections will look at how these microstructural features affected FCGR both in air and in SW.

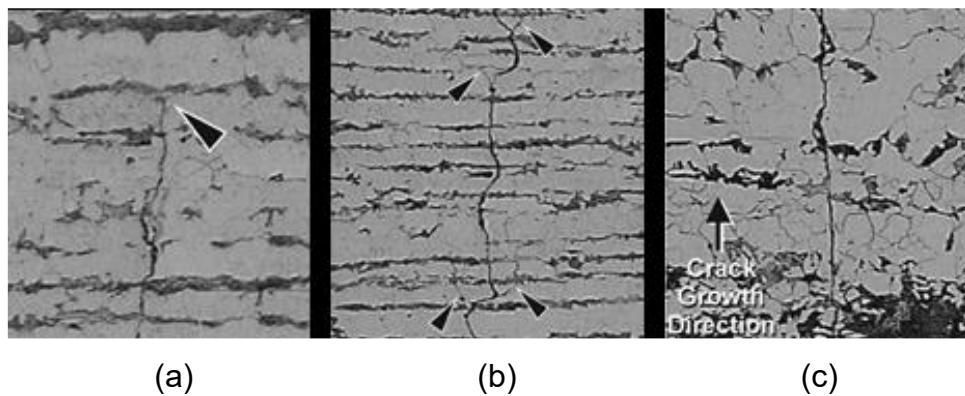


Figure 6-6: Fatigue crack growth path in ferrite-pearlite banding [124]: (a) Crack tip arrest by pearlite (b) nature of crack paths through banded microstructures (L-S) and (c) crack paths through non-banded microstructures.

6.5 Effect of microstructure on fatigue crack growth rate in air

Many of the existing FCGR data on steels are many decades old. The majority of the tests were performed on NR and QT steels and they form the basis for explaining the FCG behaviour and corrosion-fatigue mechanism in steels. Today, new sets of modern TMCP S355 steels are increasingly being used in the design of offshore support structures (see **Table 6-1**). The effect of grain refinement on FCGR in SW data for these new materials and comparative study with NR and QT steels are very scarce, if not non-existent. In order to fill in the gap in the knowledge, the authors have extensively studied the FCG behavior of

S355G8+M and S355G10+M subgrades in air and SW using the state-of-the-art equipment. In this paper, these results are presented and compared with similar studies in the literature with the intention of finding out the effect of grain size and phase morphology on FCGR in air and SW.

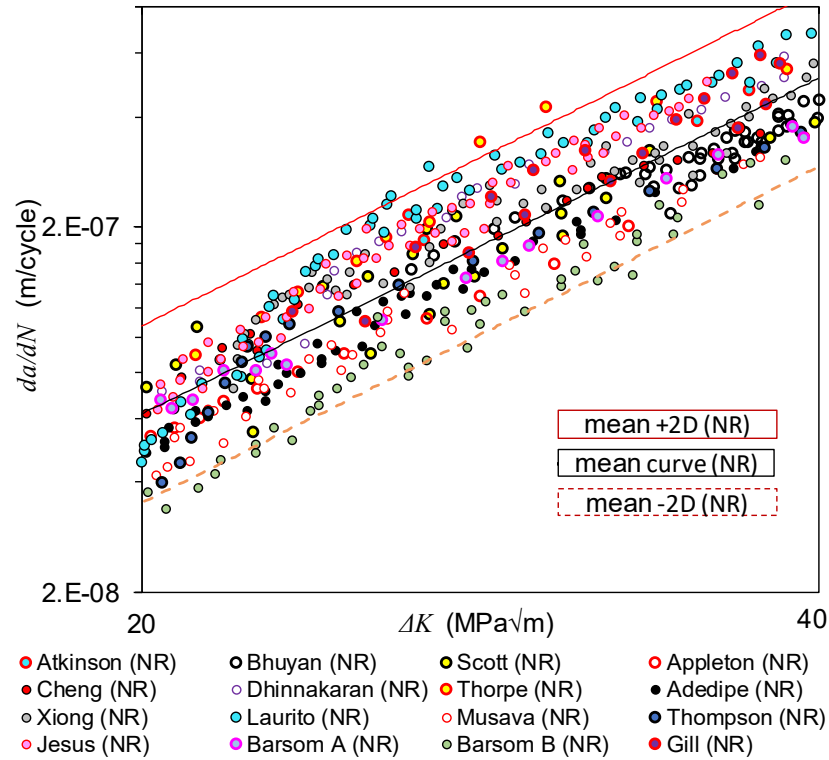


Figure 6-7: FCGR in air for normalized α -P steels (Note: Barsom B is the combination of A302 – B, ABS – C and A36 data points [205]) (NR denotes ‘Normalized-rolled’ and SD – standard deviation.)

Table 6-4 Paris law constants for the data in Figure 6-7

Paris Curve (NR) air	C	m
Mean	6.41×10^{-12}	3.06
Mean +2SD	1.12×10^{-11}	3.06
Mean -2SD	3.67×10^{-12}	3.06

Many researchers have studied FCGR in α -P steels. **Figure 6-7** presents FCGR studies on NR α -P steels only and **Table 6-5** lists all the studies and test conditions considered in these studies.

Table 6-5 Referenced studies for data plotted in Figure 6-7

Study by	Steel grade	Wave form	Freq. (Hz)	R	%EL	YS (MPa)	UTS (Mpa)
Atkinson [150]	BSI501-213-32A	h	1 - 50	-	28	320	490
Bhuyan [152]	CSA G40.21M 350 WT	s	0.005 - 2	0.05 - 0.1	29	405	513
Scott [154]	BS4360-50D	s	1 - 10	0 - 0.85	32	370	538
Appleton [157]	BS4360-50D	s	-	-	32	368	508
Cheng [159]	ABS grade EH36	r	10	0.1 & 0.5	21	350	-
Dhinakaran [192]	C-Mn steel	-	5	0.1	32	459	530
Thorpe [24]	BS4360-50D	s	1 - 10	0.5	32	370	538
Adedipe [149]	S355J2+N	s	2	0.1	20	345	-
Xiong [210]	Q345	s	0.5	0.1 - 0.5	26	370	585
Laurito [211]	RD480	s	10	0.1	32	459	541
Musuva [168]	BS4360-50D	s	30	0.08, 0.5, 0.7	30	386	560
Thompson [214]	BS4360-50D	s	0.5	0.1	30	386	536
Jesus [167]	S355	s	≤ 20	0.25, 0.5 & 0.75	22	419	766
Barsom A [205]	A537 -A	s	1	0 <R<0.7	34	407	572
Barsom B [205]	A302 - B	s	1	0 <R<0.7	26	386	607
	ABS - C	s	1	0 <R<0.7	36	269	434
	A36	s	1	0 <R<0.7	32	248	517
Gill [215]	HSLA (MF-80)	-	5	0.1, 0.67	-	653	742

s = sine, h = holdtime, r = random, c and m = Paris law constants

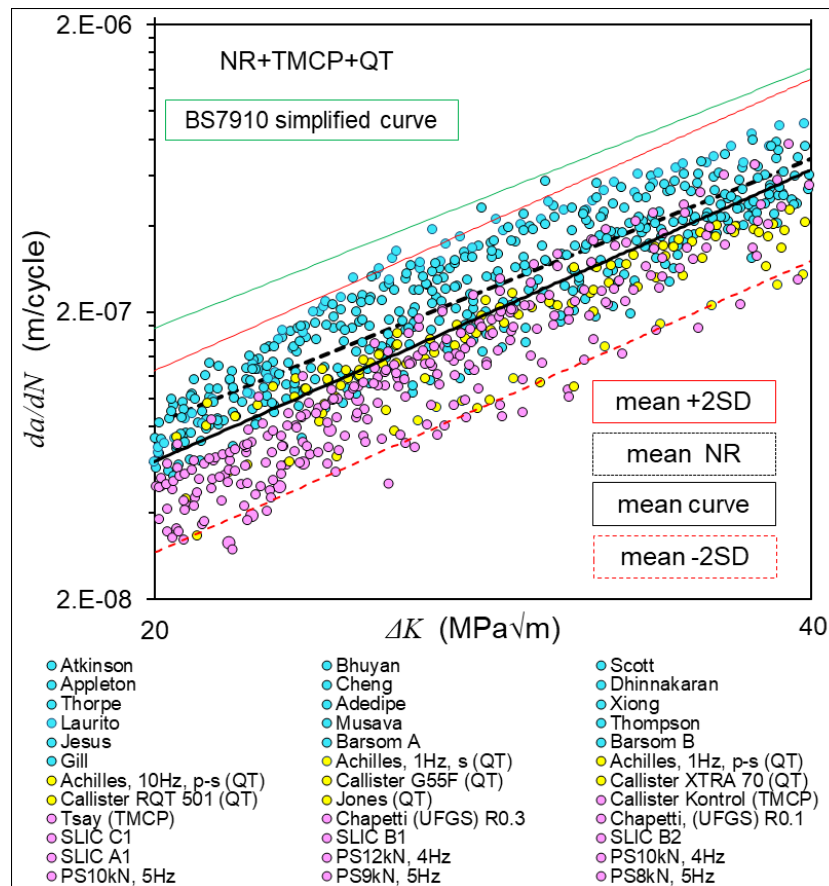


Figure 6-8: FCGR in air for normalized-rolled NR, TMCP and QT steels

Xiong studied Q345 hot-rolled steel commonly used in the nuclear industry and petrochemical plants. Laurito studied FCGR in intercritical-isothermally-quenched grain-modified steel designated as RD480. Cheng [159] studied ABS grade EH36 and this steel is used in shipbuilding.

The pertinent test conditions for all the other studies are as noted in **Table 6-5**. Also, presented in **Table 6-5** are the steel grades. Note also that in many of the studies, stress ratio (R) in the range 0.1- 0.85, frequency in the range 0.005Hz – 50Hz and waveforms have little or no effect on the FCGR in the α -P steels in air [150][153][161][167][168][169][170]. All the steels are designed to be used in both onshore and offshore structures. And all the tests were performed under constant-amplitude waveforms, except Cheng [159] who performed his test under random waveform. The general range for temperature of the studies is between 0°C and 27 °C. All the steels in **Table 6-5**. were conventionally rolled and normalized. From **Figure 6-7**, the Paris law constants for the mean curve, mean +2SD and mean -2SD of the normalized-rolled steels in air for frequency range 0.005Hz – 50Hz and stress ratio 0.05 – 0.85 are obtained as $da/dN = 6.41 \times 10^{-12} \Delta K^{3.06}$, $1.12 \times 10^{-11} \Delta K^{3.06}$ and $3.67 \times 10^{-12} \Delta K^{3.06}$ respectively. The Paris law constants are listed in **Table 6-4**. The Paris Region is taken in the range $20 \leq \Delta K \leq 40 \text{ MPa}\sqrt{\text{m}}$. An interesting note is that even though the steels have similar composition and micro-constituents, the FCGRs are significantly apart. Often, this variation in the Paris Region is associated with material variability or crack closure (R-factor), but we could see that the margin of separation is very much significant – up to a factor of about 3. Barsom B test was performed at high R where crack closure is not expected and has relatively low yield strength with normal ductility of 26% yet performed better than all the steels in resisting crack growth.

Six steels appear to perform relatively worse in terms of FCG resistance. These are steels studied by Laurito, Gill, Jesus, Dhinakaran, Thorpe and Scott. They generally have relatively lower carbon and manganese contents. This implies that the volume fraction of pearlite may be low. The steel employed in Laurito's study

has very low carbon content, better ductility, higher yield strengths, but is the worst in resisting crack growth. It should be noted that the production process for Laurito's steel is somewhat different, which is intercritical-isothermally-quenched grain-modified, which would likely produce more ferrite and may account why the FCGR is the lowest.

Table 6-6 Paris law constants of Figure 5-8

Paris curve (NR+TMCP+QT)	C	m
Mean	1.87×10^{-12}	3.37
Mean +2SD	3.87×10^{-12}	3.37
Mean -2SD	9.02×10^{-13}	3.37
BS7910 simplified law	1.65×10^{-11}	3.00

Table 6-7 Referenced studies for the refined-grains data plotted in Figure 6-8

Study by	Steel	Wave form	Freq. (Hz)	R	%EL	YS (MPa)	UTS (Mpa)
Achilles	SA508 cl III forged (QT)	s, p-s	1 - 10.	0.2	24	468	611
Callister [212]	Kontroll-50-FSM (TMCP)	s	0.5	0.6	-	470	-
Callister [212]	BS 4360 G55F (QT)	s	0.5	0.6	-	580	-
Callister [212]	N-A-Xtra (QT)	s	0.5	0.6	-	690	-
Tsay [191]	EH36 (TMCP)	s	5	0.1	21	420	-
Jones [155][156]	Naval (QIN) steel (QT)	h	0.008	0.0/0.5	27	625	735
Chapetti	TMCP-UFGS	s	30	0.3	14	696	1183
PS/SLIC	S355G8+M	s	2, 4, 5	0.1	36, 35	435/447	545/549
S = sine, h = holdtime, r = random, PS = present study							

Gill studied MF-80 and the carbon content is so low that the microstructure consisted of very fine-grained equiaxed ferrite [215]. It can be seen from **Figure 6-5 (i)** that MF-80 is essentially a ferritic steel with very little amount of pearlite. It was reported that the steel was hot rolled between 900 to 925°C, cooled at rates between 8 and 75°C/s, and coiled at 650°C after which the coil is air cooled to room temperature. In his comparative study, Gill [215] reported that this steel also exhibited the lowest FCG resistance than HS and HY-100 it was compared with. In between and including these two FCGR upper bound and lower bound fits made to the experimental data, the FCGRs appear to be independent of the mechanical properties, stress ratio and frequency. We may be tempted therefore,

to assert that this considerable difference is likely to be caused by features of the phases in the steel microstructural. To understand the role of the microstructural features, these FCGRs data points for the NR steels in air were expanded to include studies on α -*P* steel subgrades with grain-refined microstructures. The two types considered are TMCP and QT steels. The additional studies considered have been listed in **Table 6-7**.

PS/SLIC [175] studies were on thermomechanical control process (TMCP) steels. TMCP steels generally have a fine-grained microstructure that confers exceptional strength-toughness combinations. The experimental frequencies and stress ratios are also given in **Table 6-7**. Callister [212] and Tsay [191] also studied TMCP steels. Jones [155][156] studied a QIN steel that is quenched and tempered, producing a fine-grained microstructure. The QIN steel is a naval steel containing high Ni and Cr compared with conventional steel. It was produced by austenitization at 920°C for 2 h, water quenching and tempering for 1hr 30 min at 655 °C (below A_{c1}) and air cooled. A_{c1} temperature depicts the start point of the transformation between the α -ferrite and the austenite γ upon heating. Theoretically, fine pearlite is obtained between 700°C and 550°C. Since the steel is tempered at 655°C, fine distribution of non-lamellar ferrite and cementite is expected. The Jones' air test was carried out under hold-time (Trapezoid) and this result has been included in the analysis because, as mentioned earlier, waveforms have little or no effect on the FCGR in the α -*P* steels in air. Achilles [153] used sine and positive sawtooth in the study of FCGR in SA508 cl III forged RPV steel – a material used for reactor pressure vessel. Because this steel is forged, quenched and tempered, and to be used in nuclear plants, it is expected to have a refined microstructure with high level of purity and reduced macro-segregation. Note that the FCGRs for the two waveforms (sine and positive sawtooth) in Achilles' study were found to be similar. The mechanical properties of the forged steel is given in **Table 6-2** and **Table 6-7** and the data were extracted from elsewhere [222].

Chapetti [213] studied FCGR in a steel with ultra-fine grain structure (UFGS) of the order of 1 μ m (see **Figure 6-5** (n)), using sinewave frequency of 30Hz in air. In their study, they reported that stress ratio affected the FCGR in the range 0.1 – 0.7 in air. As aforementioned, the range of stress ratio considered in this study is 0.1 and 0.85. Chapetti's result contradicts the conclusions made by several researchers that stress ratio (R) within the range of his study has little or no effect on the FCGR in α -P steel in air. This tends to suggest that the finer the grains are, the more susceptible the material may become to damages due to high tensile stress fluctuations. However, no other studies known to the authors have collaborated this finding on UFGS. Their FCGRs under stress ratios of 0.1 and 0.3 have been included in Fig. 8. Another note from their study is that the ultra-fine-grained steel had a lower FCGR than a SM490 steel or a HT80 steel with which it was compared with originally. All these data points on fine-grained steels have been included in the FCGR data in **Figure 6-7** and re-presented in **Figure 6-8**. All the data analyzed here are within the temperature range of 0 °C to 27 °C. The chemical compositions and mechanical properties are given in **Table 6-2** and **Table 6-7**. The data points for the TMCP have been plotted using squares, the QT with triangle and that of conventional steels with monochromatic circles for clarity. Note that the blue colour data points are same as in **Figure 6-7**.

As previously noted, **Figure 6-8** clearly shows that despite the fact that these steels are composed essentially of two phases – ferrite and pearlite, their FCGR differ widely within the Paris Region considered, this time up to a factor of 4. It is interesting to know that some of the steels have received some level of grain refinement. A302 – B contains high amount of Mo. Molybdenum forms stable carbides and remains in ferrite, contributing to solid solution strengthening. Since this material is used in the design of nuclear plant and pressure vessels, it is expected to be very clean and fine grained to increase strength and it is only NR steel whose FCGR is somewhat below the mean curve. In general, **Table 6-5** contains the list of the studies whose FCGR data points are generally along and above the mean curve, while **Table 6-7** contains the list of studies whose FCGR data points are generally along and below the mean curve. **Figure 6-8** shows a

very clear demarcation between the FCG resistance of the two sets of microstructures of the same steel grade. Observation shows that the mean curve (NR) of the conventional steel (the dashed black line) is shifted down to the mean curve (NR+TMCP+QT) (the solid black line) after the FCGR data for the refined steel grain structure were added. The downward shift is 1.36 at 20 MPa√m, 1.28 at 24 MPa√m, 1.18 at 29 MPa√m and 1.05 at 40 MPa√m and this gives an average value of 1.22 in the Paris Region considered.

In **Figure 6-8**, the effect of the new data, which are from steels with refined grain structures can also be seen from the Paris law coefficients and exponentials of the mean curves. For the normalized steels it is obtained as $6.41 \times 10^{-12} \Delta K^{3.06}$ (see **Table 5-4**). After the grain-refined steel data were added, the coefficient value dropped, while the exponential value increased thus: $1.87 \times 10^{-12} \Delta K^{3.37}$ (see **Table 6-6**). This simply suggests that the effect of refining ferritic steel microstructure is tilting of the FCGR curve in the anti-clockwise direction rather than shifting straight downward. This tends to imply that SIFR threshold (ΔK_{th}) for the refined grain structure might be higher than that of the normalized steel (with larger grain size) within the temperature range of analysis. However, as soon as the crack has started propagating and grown to a certain size, a little change in the ΔK or load level may have a higher accelerating effect on the refined grain structure. Though, the FCGR of the refined structure for the most part of the Paris Region will still be substantially lower than that of normalized steels. It also suggests that as the ΔK is increasing, a point will be reached when the rate at which the crack is growing in the refined steel will outpace that of normalized steel. Considering less scatter in air test, the BS7910 recommendation curve (green line) still provides a good estimate of the FCG behaviour for these grades of steel. But clearly it is over-conservative for the fine-grained steels.

From the above analyses, TMCP and QT generally occupied the lower half, while the NR steels generally occupied the upper half of the mean curve (NR+TMCP+QT) regardless of the mechanical properties, frequency or stress ratio in the range considered. The steels with refined grain structure offered better

fatigue damage tolerance in air. In other words, FCGR is lowest in the refined steel compared with conventional steel in the Paris Region. What this result is suggesting is that the difference in the FCGRs in these steels could only be explained by the variations in the ferrite-pearlite morphology. Therefore, a microstructural study was carried out to examine the fatigue crack path in the test samples.

Figure 6-9 shows the microstructure of the subgrades of S355 steel obtained by the authors using 2% nital. **Figure 6-9 (a - c)** is the microstructures of S355J2+N normalized-rolled steel at a scale of 50 μ m **(a)** and 20 μ m **(b, c)**. It typically consists of ferrite (light etching) and pearlite (dark etching) using 2% nital.

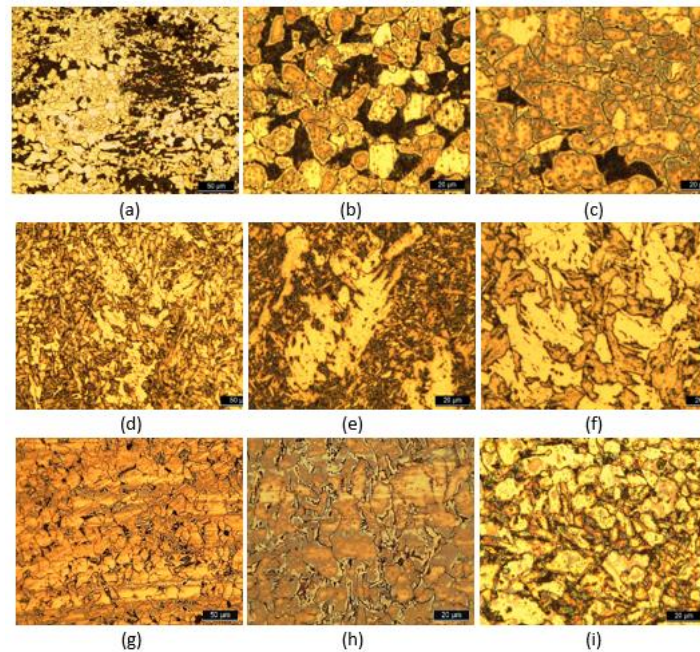


Figure 6-9: Microstructures of ferritic-pearlitic steels: (a - c) S355J2+N normalized-rolled; (d - f) S355G8+M TMCP; and (g - i) S355G10+M TMCP. ferrite (light etching) and pearlite (dark etching) using 2% nital.

The pearlites formed in colonies and in blocky form. The behaviour of this microstructure is taken as a representative of all the conventional ferritic steels, especially for thick plates as the sample was extracted from a 90mm steel plate.

Variations in the ferrite grain size is seen and this can be attributed to availability of nucleating sites and the driving force of transformation from austenite. **Figure 6-9 (d - f)** show the microstructures of S355G8+M steel, at a scale of 50 μ m (**d**) and 20 μ m (**e, f**). Note that the light etching is generally ferrite and the variation in the light shade is due to supersaturation of some of the ferrite phases with carbon. The pearlite is dark and the colonies in two dimensions appear thin. Due to low carbon content of the steel and extensive deformation process during manufacture, the pearlite has formed on the ferrite grain surface as an elongated phase located mainly along the boundaries of the ferrite grains.

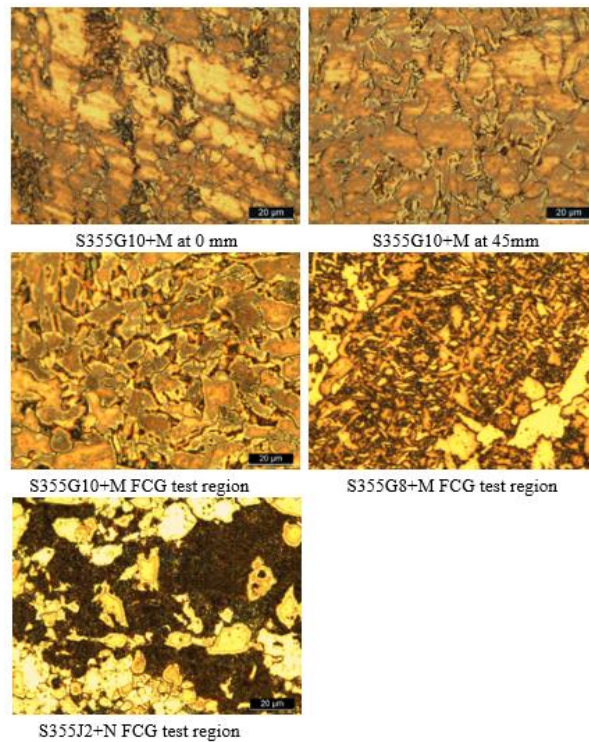


Figure 6-10: Variation of microstructure of S355 steel plate with thickness. Also indicated are the FCGR test regions for the three subgrades in this study.

Figure 6-9 (g - i) is the microstructures of S355G10+M at a scale of 50 μ m (**g**) and 20 μ m (**h, i**). The same phenomenon as in S355G8+M is observed, but the decoration or covering of the ferrite grain surface along the boundaries by thin pearlites is more pronounced and discrete. This thin shape of the pearlite tracing

ferrite grain boundaries is generally very small or absent in the conventional steels, which has pearlite colonies mainly in blocky form. Note that all the samples examined are from 90 mm steel plate. Thus, it was observed that the dimension of the pearlite colony thickens at the plate surface (0 mm) and this may be due to tempering effect from the dissipation of heat from the plate interior. **Figure 6-10** shows the variations in the microstructure of S355G10+M at the surface (0mm) and at the interior of about 45mm. The transverse dimension of the pearlite decoration of the ferrite grains appears to decrease with depth. The **Figure 6-10** also presents the microstructures of the regions of FCGR tests under consideration.

This variation in the dimensions of the pearlite colony shows that the property of extra-large steel plates is not homogenous throughout its thickness. Generally, the pearlite morphology is found to have profound implications on the fatigue crack path in α -P steels. To the knowledge of the authors, this implication has not been studied in detail by any researchers. As already noted, it appears there is no correlation in frequency, stress ratio and mechanical properties that explains the variation in the FCGR of α -P steels in air. Consequently, the authors assert that FCGR in α -P steels is strongly dependent on the nature of the pearlite morphology throughout the da/dN vs. ΔK sigmoidal curve.

6.6 Fatigue crack path in ferrite-pearlite steels in air

Potential ways microstructure could influence FCGR is by crack deviation and branching. Understanding the crack path is important in providing the complete solution of a fatigue problem, as well as for the selection of appropriate non-destructive testing procedures [223]. **Figure 6-11** shows the crack paths in the microstructures of the steels under consideration in air.

What is apparent in **Figure 6-11** is that the mode of fatigue crack propagation is not the same for these subgrades of S355. In the figures, the white arrows point to deviation of crack front by pearlite whose orientation offers least resistance for

the crack propagation. Interlocking can occur in these areas of crack deviation. The red arrows show points where the pearlite caused the splitting of the crack front, resulting to crack branching. The white triangle used in S355G10+M of **Figure 6-11** to show metal crumbs in-between branched and re-joined crack fronts. It is interesting to note that Korda A. A., at el. [124] studied a TMCP steel and also observed that pearlite colony deviates crack in a banded microstructure and this phenomenon lowers the FCGR. They also reported that the extensive crack branching is due to the banded pearlite orientation, which reduced the crack growth driving force.

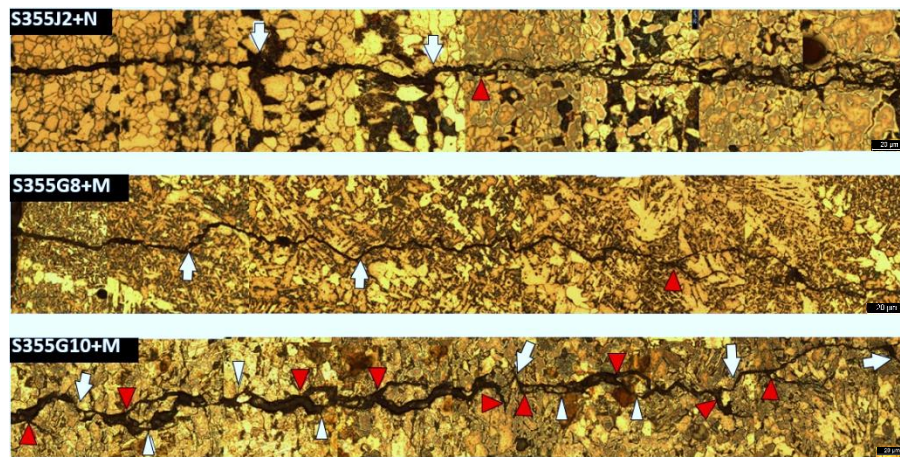


Figure 6-11: Crack path in subgrades of S355 steel in air (a) S355J2+N normalized-rolled steel (b) S355G8+M TMCP and (c) S355G10+M TMCP at a scale of 20µm.

The tendency for crack branching by the pearlite tends to increase with increase in the driving force or ΔK . This is more evident in S355J2+N of **Figure 6-11**. The branched crack fronts in the normalized steel generally appears to grow in a parallel formation to one another, forming what may be described as a fibrous crack front. It was observed that the fibrous crack front sometimes merged to form one main crack front which continued the propagation. Or a particular crack front overtakes all others. The angle of deviation is very low, and this appears not to contribute much to the FCG retardation. The angle of deviation and number of crack branching is by far more in the TMCP steels. The crack deviation can cause

interlocking due to plasticity-induced mismatch/ crack closure or crack closure due to surface roughness-induced closure [197]. In Korda [124] study, he said that when a crack traversed from ferrite grain into banded pearlite that a large deviation is obtained and that this large deflection induced roughness closure which retarded the crack growth. But, since the deviation mainly occurs in the pearlite colony which has a hard and brittle cementite, crack closure phenomenon which depends on ductile stretching or tearing of the fractured surface to form asperities is expected to be minimal in these local areas, unless there is mismatch of the fractured surfaces. The possible crack interlocking points along the crack path are indicated by the white arrows in **Figure 6-11**. Among the three samples in **Figure 6-11**, crack deviation and branching due to pearlite morphology and orientation is more for the TMCP steels. Additional feature in S355G10+M is that several branched cracks re-joined the main crack without been arrested, leaving some metal crumbs in-between. These metal crumbs if re-oriented can cause reduction of the effective ΔK at the local crack tip through wedging action, thereby retarding the crack growth. The distance between the main and branched crack fronts was up to three ferrite grains, and the parallel distance measuring about $62\mu\text{m}$.

Similar crack path was seen by Callister [212] in his study of quenched and tempered steel (N-A-XTRA 70) and TMCP steel (Kontroll 50 ESM) as shown in **Figure 6-12**.

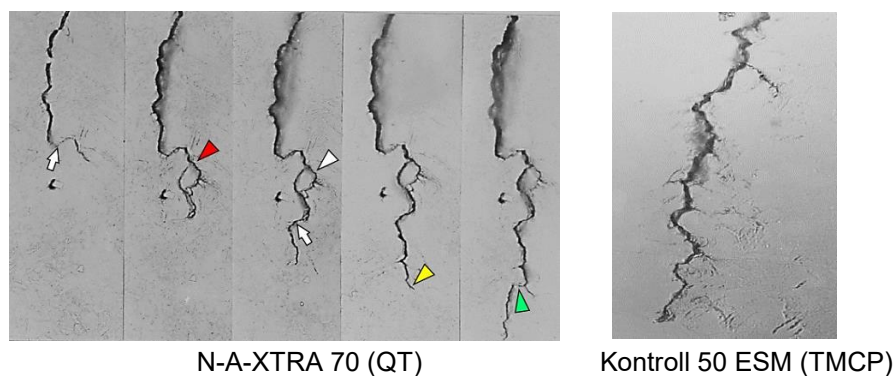


Figure 6-12: Crack paths in Callister's study [212]

Figure 6-12 shows non-planar crack path with high degree of crack deviation as shown by the white arrows. The red triangle shows a branched crack that was not arrested but re-joined the main crack and this re-joining created a metal crumb in between the re-joined crack fronts as shown by the white triangle. The yellow arrow points at the crack front at a particular ΔK . This crack front appears to have encountered high crack growth resistant phase which arrested it thereby causing another crack front to develop and follow path of least resistance as shown by the green arrow. The tendency for this crack branching appears to increase with ΔK . Secondary cracking seen on fractographs is formed by these branched cracks that have been arrested. Chapetti [213] also found that crack branching in UFGS increases with increase in the ΔK level. In general, TMCP microstructures showed high degree of crack deviation, branching and with complex multiplanar crack front. This form of crack was not usually found in NR steels in this study (see **Figure 6-11 (a)**).

From the foregoing, four phenomena which could affect FCGR in refined-steels are crack deviation, crack closure/interlocking, crack branching and wedging action of metal crumbs and all these factors can be promoted by the nature of the pearlite morphology. The four factors enumerated can cause stress shielding or reduction in the effective stress at the local crack tips resulting in retardation of FCGR. Thus, the degree of occurrence of these crack retarding factors will vary with variation in steel microstructure. This phenomenon explains why FCGR is lowest in the fine-grained structures (see **Figure 6-8**), especially in the TMCP steels that have their ferrite grain boundaries partially or completely covered by the thin pearlite. This assertion appears to be consistent with the findings of Mutoha, Y. et al. [224] who reported that in TMCP steel, the FCG resistance of the steel having a distributed pearlite structure was more than that of a networked pearlite structure. They attributed this performance to a tortuous crack path created by the distributed pearlite which caused crack closure and interlocking of crack surfaces. Coincidentally, this report is similar to what the authors observed in the present study. The networked and the distributed microstructures for the steels used in their study are shown in **Figure 6-13**. The distributed structure was

produced through thermomechanical process route. The networked structure is comparable with normalized rolled microstructure of S355J2+N at the same magnification of 50 μ m, as can be seen in **Figure 6-9 (a)**.

In general, from **Table 6-2**, **Table 6-5** and **Table 6-7** there is no obvious trend in the yield strength, tensile strength, ductility, frequency and stress ratio that could possibly explain the wide variation in FCGR data, up to a factor of about 4 in the Paris Region in **Figure 6-8**. In other words, we may assert that there is little or no correlation among the frequency, stress ratio (in the range 0.1 – 0.7) R and mechanical properties of the material with the FCGRs. The obvious explanation is that the FCGR in α - P steels is strongly dependent on the nature of the microstructure or grain refinement throughout the length of the da/dN vs. ΔK sigmodal curve in air. Mustapa, S. M., et al [225] in their study also reported that the pearlite particle size and spacing affected FCGR of ferrite–pearlite steel in the Paris Region even at ΔK as low as 5MPa $\sqrt{m}$. However, they attributed high FCGR resistance to higher volume fraction of pearlite, large spacing and size of pearlite particle. Their use of the words, ‘pearlite particle and ‘spacing’ can be understood by considering the microstructures studied as shown in **Figure 6-14**.

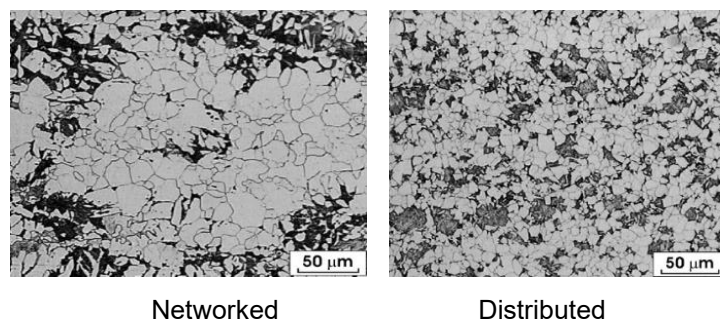


Figure 6-13: Steel microstructures showing the networked and distributed pearlite morphologies in Mutoha, Y. et al. [224] study.

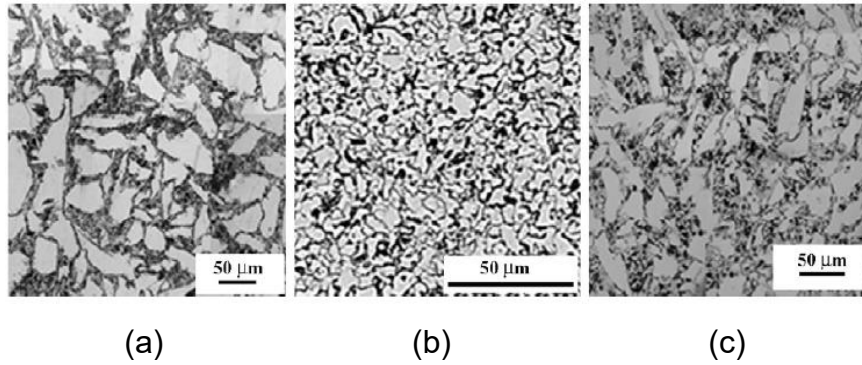


Figure 6-14: Steel microstructures in Mustapa, S. M., et al [225] study.

In interpreting the phases, it can be seen that **Figure 6-14 (a)** and **(b)** had the same volume fraction of pearlite but different particle size and spacing. It is very clear that **Figure 6-14 (b)** is finer, thus have smaller 'particle size'. **Figure 6-10 (a)** and **(c)** were reported to have different volume fractions of pearlite but the same particle size. What is not clear here is the contextual meaning of pearlite size and spacing. Pearlite forms in colonies. If we assumed that the particle size is same as size of colonies, then it is not known if the pearlite spacing is a measure of distance between colonies or between the lamellar size of ferrite and cementite in the pearlite. Thus, the study did not consider the significance of pearlite orientation or the directional alignment of the ferrite-cementite lamellae. Again, the use of the phrase 'pearlite particle' is vague in the sense that pearlite forms in colonies and appear in three dimensions as interpenetrating bi-crystal of ferrite and cementite ($\alpha + \text{Fe}_3\text{C}$) whose interlamellar spacing varies with temperature of formation. Barsom [205] in his study noted that α -*P* steels are more resistant to FCG than martensitic steels and attributed this to severe secondary cracking or to the lamellar microstructure of α -*P* phase or both. The Barsom's conclusion is in line with findings reported in this paper. We have already asserted that the morphology and orientation of pearlite in α -*P* steels strongly influenced the FCGR throughout the da/dN vs ΔK (including Paris Region). This trend as attributed to the pearlite morphology due to microstructural refinement, was also investigated for tests in *SW*, and the result is discussed in the next section.

6.7 Effect of microstructure on FCGR in seawater

Following the analyses in air, a comparative study on the *CFCGR* in α -*P* steels in SW was also carried out. **Figure 6-15** shows plots of CFCGR data obtained in the literatures for the conventional and grain-refined steels (TMCP+QT) (including those from the present study conducted by the authors).

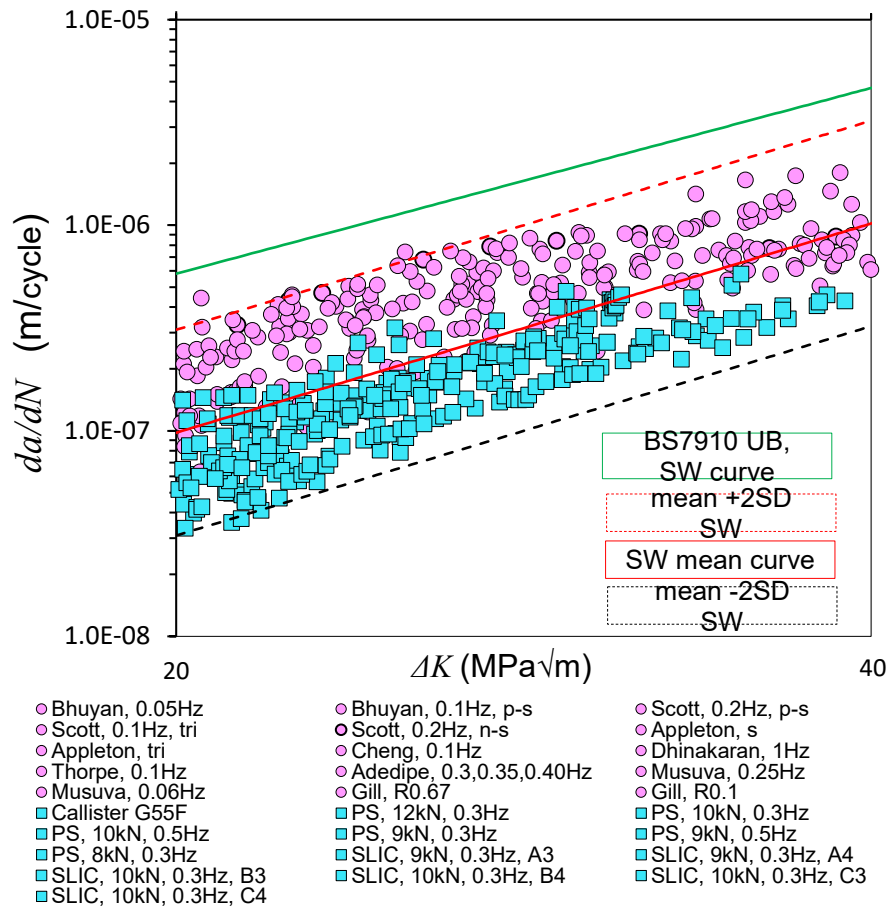


Figure 6-15: Comparison of CFCGR for different mean load, frequency and waveform

(SW -seawater, tri-triangular, n-s: negative sawtooth, p-s: positive sawtooth)

Table 6-8 Reference studies and Paris law constants of Figure 6-15

Paris curve SW	C	m
BS7910 SW	7.27×10^{-11}	3.00
Mean +2SD (SW)	1.24×10^{-11}	3.38
Mean -2SD (SW)	1.24×10^{-12}	3.38
All data Mean (SW)	3.92×10^{-12}	3.38

The circles data points are those of the conventional steels, while the square data points are for grain-refined steels - either by TMCP or tempering. The test waveform is mainly sine, but those of triangular, positive sawtooth (p-s) and negative sawtooth (n-s) have been included. For *SW* environment, studies [24][154][161][169] show that CFCGR under sine, triangular and positive sawtooth waveforms were similar. They also reported that sinewave was only slightly higher than triangular waveform. However, all the waveforms data were analyzed and included in this study.

All the data analyzed here are within the temperature range of 0°C to room temperature. Notice that the stress-ratio-increase effect of curve shifting to the left is not generally observed in *SW* between 0.0 to 0.7 within the data set considered in this study. The tests were performed in *SW*, 3% or 3.5% NaCl. Note also that the effect of the shape of the experimental specimens is largely ignored here, though the results within the limits under consideration show little or no effect in the Paris Region. This appears to be supported by Musuva [168] who found that thickness has considerable effect on smaller specimens only in the threshold region. As stress ratio and sample thickness increase, thickness effect diminishes and disappears. The experimental conditions for the data points in **Figure 6-15** are as given in **Table 6-9**. It is worthy to note that the BS7910 still provides a satisfactory design estimate of the FCG behaviour of the NR materials considered in free corrosion condition, but excessively over conservative for the TMCP steels. Generally, all the data points are within the +2SD and -2SD bounds. **Table 6-8** shows the Paris law constants for the mean, -2SD and +2SD curves of all data considered in *SW*. BS7910 recommendation curve is also given. It can be seen from **Figure 6-15** that the BS7910 curve is very much over conservative for the TMCP steels.

In **Figure 6-15**, it appears that there is a separation and general boundary, given roughly by the mean curve, between the normalized and grain-refined steels in terms of rate of CFCG in *SW*. Again, observation of the individual CFCGR curve tends to show no trend in chemical or mechanical properties that suggests this

demarcation. Crack closure mechanisms such as the plasticity-induced, the roughness-induced and the oxide-induced can cause retardation of the FCGR. However, in SW, the plasticity and the roughness/asperities of the fractured surfaces are rapidly eliminated by the corrosion process and these two mechanisms cannot therefore be said to contribute in the crack growth retardation.

Table 6-9 Reference studies and Paris law constants of Figure 6-15

Study by	Wave form	Env.	Freq. (Hz)	R	%EL	YS (MPa)	UTS (Mpa)
Bhuyan	s	SW	0.05	0.1	29	405	513
	p-s	SW	0.1	0.1	29	405	513
Scott	p-s	SW	0.2	0.5	32	370	538
	t	SW	0.1	0.5	32	370	538
	n-s	SW	0.2	0.5	32	370	538
Appleton	s	SW	0.17	0.5	32	368	508
	t	SW	0.17	0.5	32	368	508
Cheng	s	SW	0.1	0.1, 0.5	21	350	-
Dhinakaran	s	NaCl	1.0	0.5	32	459	530
Thorpe	s	SW	0.1	0.0, 0.1	32	370	538
Adedipe	s	SW	0.3, 0.35, 0.4	0.1	20	345	-
Musuva	s	3.5%NaCl	0.25	0.08	30	386	560
	s	3.5%NaCl	0.06	0.08	30	386	560
Gill	s	3.5%NaCl	0.5	0.67	-	653	742
	s	3.5%NaCl	0.5	0.1	-	653	742
Callister	s	SW	0.5	0.6	-	580	-
PS/SLIC (8, 9, 10, 12kN)	s	SW	0.3 – 0.5	0.1	36, 35	435/447	545/549

s = sine, t = triangular, p-s = positive sawtooth, n-s = negative sawtooth, SW = seawater

A plausible reason remains that under stress ratio range of 0.1 – 0.85, temperature range of 0°C to room temperature and frequency range of 0 – 0.7Hz, the microstructure influences the rate at which a crack grows in α -P steel in air and in SW, especially at low ΔK . This study therefore asserts that, the common believe that the microstructure does not significantly affect the Paris Region appears to be incorrect. This tends to be so, when the same grade of steel, but with different phase morphologies are put into considerations.

The major factor responsible for this difference appears to be the ferrite and pearlite morphologies. To give credence to this assertion, a microstructural study was carried out to see how the crack traversed in the steel material in SW. This is discussed in the next section.

6.8 Fatigue crack path in ferrite-pearlite steels in seawater

Figure 6-16 shows the crack paths in the ferritic microstructure in SW obtained at 20 μm , and 50 μm . The crack paths are similar to that seen in the air test. **Figure 6-16 (a)** and **(b)** are for tests performed under sinewave, while **(c)** is crack path seen under holdtime waveform. It is evident that TMCP is generally characterised by high degree of crack path deviation and branching both in air and SW. The fractography of the fractured surface for this sample will likely show many occurrences of secondary cracks. The nature of crack propagation appears to depend on the local orientation of the pearlite at the vicinity of the crack tip. When the crack encounters unfavourably aligned pearlite colony, it will propagate in transgranular mode across the ferrite grains ahead of it as shown in **Figure 6-17**. In other words, the angle the crack front makes with the pearlite colony or any least resistant path determines if the crack will propagate in transgranular or intergranular mode.

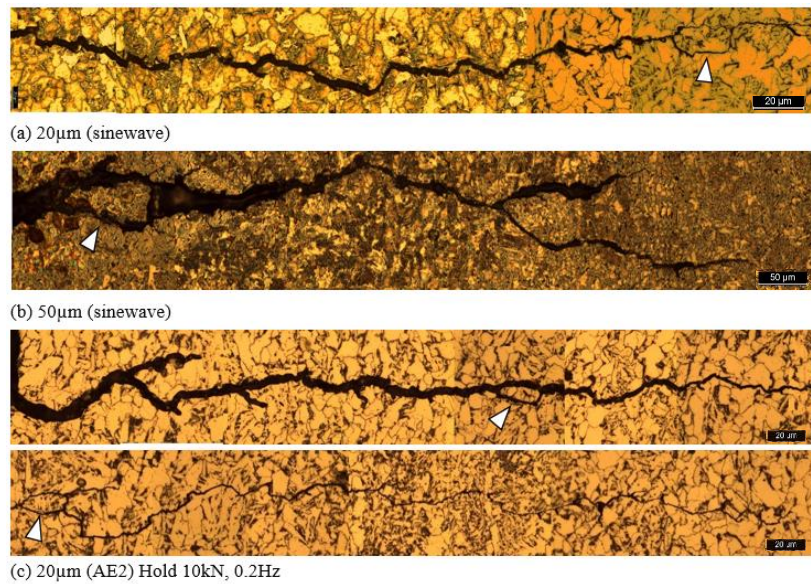


Figure 6-16: Crack path in S355G10+M in seawater (a) At magnification of 20 μm , (b) at magnification of 50 μm , (c) At magnification of 20 μm . Note: (a) and (b) were obtained under sinewave while (c) is under holdtime.

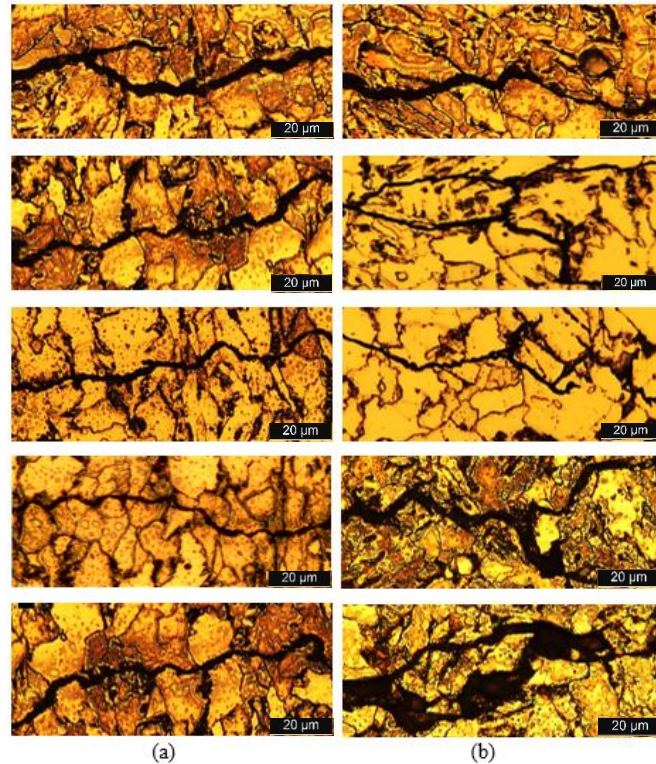


Figure 6-17: Crack path in S355G10: (a) Pearlite unfavourably oriented, (b) Pearlite favourably oriented, locally to the crack front.

Figure 6-17 (a) shows the pearlite colony and ferrite grain boundaries at about right angle to the crack front. At any instance the crack is moving horizontally, the tendency for crack to deviate or branch out appears to increase if the angle of pearlite colony orientation or the ferrite grain boundaries is acute to the crack front. In other words, if the orientation of the least resistant paths is in vertical position, the phases loses the potential to deviate the crack, making the crack to pass right through the grain in a transgranular mode. This appears to explain why there is very little deviation and generally no crack branching when the pearlite and ferrite grain boundaries are oriented at about 90° to the crack front. On the other hand, when the angle of the crack front is acute to the least resistant paths or favourably oriented, the crack front deviates or splits leading to fatigue crack retardation. Note that what appears to be a branched crack in this orientation could actually be a situation where a crack front that has been arrested by the surrounding phase caused another crack front to develop somewhere else. This phenomenon was observed by Callister [212] (see **Figure 6-12**). **Figure 6-17 (b)** shows the condition where the pearlite and ferrite grain boundaries align in almost

in the direction of the crack front – making an angle that is less than 90°. In the case of the intergranular mode, the crack simply deviates and follow the pearlite formation along the ferrite grain surfaces, or the crack follows least resistant ferrite path. The same phenomenon can occur if the ferrite to ferrite grain boundary is embrittled by hydrogen. This appears to be the mechanism that enables the TMCP steel to retard crack growth rate. The pearlite colony or embrittled ferrite grain boundary tendency to deviate and branch out propagating crack appears to be responsible for the considerable scatter in fatigue test, especially corrosion-fatigue test. Thus, high order of scatter is expected in fatigue test of TMCP or QT steels than NR steels. This finding may also have some consequences in the use of S-N curve concept in design and prediction of fatigue life of materials that are supposedly assumed to contain no cracks. Since the tendency for crack deviation and branching is a local phenomenon, the findings in this study have a fundamental implication on analytical and numerical models that specifically depend on the mechanical properties of the material (e.g. the use of plastic zone size in the vicinity of the crack tip) and assumes plane horizontal crack front in predicting FCGR in air and *SW*. It will also impact on any model that treated steel as a continuum. Such models are most likely to fail in predicting FCGR in TMCP and QT steels. In fact, such models may fail in predicting FCGR even in the same class of steel. This is another important study the authors will investigate in future work.

6.9 Conclusions

This study found that microstructure has a strong effect on the FCGR of α -*P* steels in the Paris Region, both in air and *SW*. This assertion is in contrary to common believe that microstructure has little or no effect in the Paris Region of the da/dN vs ΔK sigmoidal curve. It is also found that TMCP and QT steels resisted FCG more than NR steels both in air and *SW*, especially at low ΔK . The resistance was increased if the pearlite colonies or low resistant ferrite phase form as a thin continuous network around the high resistant ferrite grains. Three phenomena were found to contribute primarily in decreasing FCGR (in air),

namely crack deviation, crack branching and wedging action of metal crumbs in-between branched and re-joined crack fronts. The three factors retarded the crack growth by reducing the effective stress or driving force at the local crack tips and their effectiveness varied with variation in the phase morphology in the steel microstructure. The crack path was always non-planar with complex crack front. The mechanical properties – such as ductility and yield stress appeared to have no strong retarding effect on the FCGR in α - P steel. This finding has a fundamental implication on analytical and numerical models that specifically depend on mechanical properties of the steel and assumes plane horizontal crack propagation front in predicting FCGR in air and SW. Such models are likely to fail in predicting FCGR even in the same grade of steel. Hence, many models that attempted to predict FCGR in steel, can only be valid for the material data that is used in validating them. The analyses also indicated that the effect of refining α - P steel microstructure is tilting of the FCGR curve in the anti-clockwise direction rather than shifting linearly straight downward. This tends to imply that SIFR threshold (ΔK_{th}) for the refined-grain structure will be higher than that of the normalized steel. The anti-clockwise tilting of the FGCR curve suggests also that the FCGR in the refined-grain steels will outpace that of NR at very high ΔK .

7 Waveform and Frequency Effects on CFCG Behaviour in Modern Marine Steels

7.1 Abstract

The primary focus of this work is to investigate the sensitivity of cyclic waveform, frequency, load level and microstructure on the *CFCGR* in modern normalised-rolled and thermomechanical control process ferrite-pearlite steels, especially in the Paris Region of the da/dN vs. ΔK plot. Constant amplitude sinewave and trapezoid waveform (generally referred to here as hold-time) were used under frequencies of 0.2Hz, 0.3Hz and 0.5Hz and stress ratio of 0.1. Comparison is also made between the crack path in the S355 TMCP steel under sine and hold-time waveforms in *SW*. The role of microstructure in retarding or accelerating *FCG* in *SW* is discussed. This study was carried out with the aid of optical microscope (OM), scanning electron microscope (SEM) and Energy Dispersive X-Ray Spectroscopy (EDX). Experimental results show that the *CFCGR* of the sine waveform is higher than that of the hold-time for all the load levels and frequencies examined. Thus, the sinewave is the more damaging waveform and the consequence of hold-time is the retardation of crack growth rate in *SW* especially at high stress intensity factor range and low frequency. Generally, frequency in the range 0.2 - 0.5Hz has little effect; and for a given frequency an increase in load led to a reduction in the *CFCGR*, in the Paris Region (PR) under both sine and hold-time waveforms in *SW*. Under both sine and hold-time waveforms, the *CFCGR* in the TMCP steels (e.g. S355G8+M, S355G10+M) is lower than that of the normalised steels (e.g. S355J2+N). Thus, the microstructure of the TMCP steel is superior to that of J2+N in resisting crack growth in *SW* – a consequence of microstructural phase morphologies. Metallographic analyses on the fractured surface of corrosion-fatigue specimens show that the main active crack tip blunting process is the primary factor controlling the *CFCGR* of steel at high ΔK and low frequency in *SW*. Other fundamental factors affecting *CFCGR* are crack angle diversion, branching of crack front and formation of metal crumbs along the crack path.

7.2 Introduction

Steel structures in dynamic service environment degrade with time due to fatigue. Offshore engineering structures, such as wind turbines (WTs) are subjected to cyclic loading conditions due to constant exertion of wind, wave and current forces. The initiation and propagation of cracks in these structures could lead to catastrophic failure at loads far less than the yield stress of the material. The fatigue crack initiation and *FCG* are accelerated by the *SW* environment. That is, the combination of a corrosive environment and fatigue phenomenon is found to be more detrimental than each of them acting separately. This simultaneous action of cyclic load and chemical attack is commonly referred to as corrosion-fatigue (CF) and it is found to be an important failure mechanism in offshore structures [136][137]. The structural integrity concern for such structures is to ensure that they carry the applied loads during operation without any macro-scale failure throughout the life span. For this reason, it becomes a common engineering practice to measure the tendency of cracks to propagate under a given stress condition.

Corrosion is both time and temperature dependent [138][139]. Two factors that are known to control the extent of corrosion damage contribution to the *FCG* behaviour of steel are the rate at which the corrosion is occurring and the time available for the corrosion damage to occur per cycle. In a corrosion-fatigue study [24], these two factors have been used to explain the acceleration of crack propagation rates with decreasing test frequency under constant temperature. Although studies have been conducted before now by researchers to characterise the *CFCG* behaviour of materials used in offshore structures [24][151][154][156][159][175][226][227][228][229], the *CFCG* behaviour in modern steels used in the design of offshore wind turbine (OWT) foundation structures under various waveforms and frequencies is largely unknown. The sensitivity of the *CFCGR* to frequency, load level and waveforms is pertinent in predicting correctly the lifespan of these support structures under forcing conditions. Hence, accurate characterisation of *CFCGR* in these structures in

direct contact with the SW (i.e. in the absence of corrosion protection) is vital for inspection, maintenance and potential life extension.

7.3 Operational frequency range of offshore wind turbine support structures

For OWTs the external environmental forces such as wind and wave forces result in development of stresses in the support structures, which are dominantly made of monopoles [140]. The magnitude of these cyclic stresses directly determines the functionality and lifespan of the OWT, thus presents a major fatigue problem. The assessment of the first natural frequency of the OWT and frequency distribution is a very important part of structural analysis. The frequencies of the forces that are commonly taken into considerations are that of the wind and wave spectra, and operational intervals of the rotor. The primary excitation frequency due to the rotational speed of the rotor is commonly called the $1P$. For modern OWTs, it is reported that the rotor speed under normal operation is typically about 7–12 revolutions per minute (rpm), which corresponds to $1P$ frequency of 0.12 - 0.20Hz [142]. The second excitation frequency of OWT occurs due to creation of wind shade by the OWT tower. Anytime a single blade of the rotor passes across the wind shade of the vertical tower there is reduction of the wind loading on the tower causing additional cyclic load on the support structure at the frequency equal to $1P$ times the number of blades. Thus, $3P$ is the blade passing frequency for a 3-bladed WT. This implies that if the rotor completes 6 rpm, the $1P$ frequency is 0.1Hz while the $3P$ is 0.3Hz.

For wave loading, which is often produced by wind, the range of possible frequencies typical of North Sea offshore waves is given in the range 0.04 and 0.10 Hz [27]. Many of the studies on CFCGR in marine environment used a frequency of 0.1Hz and temperature range of about 4 - 10 °C. The intention of these studies has been to simulate offshore conditions and obtain fatigue data for application in offshore Oil & Gas production and exploration platforms. This would then enable extrapolation of the laboratory data to field structures. Generally, in designing commercial OWTs, frequency of the support structure is such that it lies between $1P$ and $3P$ [144]. This frequency condition of the OWT

is referred to as soft-stiff design [23]. This implies that fatigue experiments of practical significance to many of the commercial WTs in soft-stiff condition should be carried out in the range 0.16 – 0.54Hz [144]. However, there is no study in the literature to examine the frequency effects on CFCG behaviour of structural steels employed in fabrication of monopiles.

7.4 Previous studies on frequency and waveform effects on corrosion-fatigue

Corrosion is a time dependent material degradation mechanism. In corrosion-fatigue process, the loading frequency and the waveform of the load cycle are expected to have some potential effects on the crack growth behaviour of steel in corrosive environment. Many studies have shown that the FCGR is sensitive to the waveforms in seawater environment but not in air. It has been shown in previous studies found in the literatures that for the R in the range 0.1- 0.85, frequency in the range 0.005Hz – 50Hz and waveforms have little or no effect on the FCGR in the ferrite-pearlite steels in air [150][153][161][167][168][169][170]. However, these studies have reported that the CFCGR in SW is sensitive to the loading waveform due to the difference in the straining mode at the crack tip (crack tip geometrical difference). These studies have shown that under anodic dissolution, the amount of material dissolved at the vicinity of the crack tip within one cycle of loading was the same for sinusoidal, positive sawtooth, negative sawtooth and square waveforms. According to these studies, the CFCGR of the positive sawtooth was found to be higher than that of the sinusoidal waveform and the rates for the two waveforms were higher than that of negative sawtooth and square waveforms in free corrosion test condition. They also found that for the square waveform, holding for 0.1 s and 0.9 s at maximum load had little influence on the CFCGR. Consequently, they attributed the waveform crack growth rate difference to the continuous deformation of the crack tip material.

Atkinson, et al. [150] studied FCGR in Fe-C-Si-Mn structural steel designated BSI501-213-32A under triangular waveform in air, distilled water and natural lake

water at temperatures between 25 and 30°C and pH 6.0. They concluded that FCGR in aqueous environment was 7 times higher than air and that this increment is dependent on the frequency. The magnitude of environmental effect was found to depend on rates of loading. They also reported that negative sawtooth and square wave, which have very rapid rise-time (i.e. period of increasing tensile load) of the fatigue cycle, did not increase the CFCGR at frequencies as low as 0.01Hz. Also reported in this reference is that holding at maximum load in the SW for up to 10 minutes did not affect the FCGR at room temperature and at high cyclic frequencies the FCGR in the SW was similar to that in air.

Austen [164] showed that for En24 steel in distilled water, under positive sawtooth (p-s) waveform at $R = 0.25$, the CFCGR increased with increase in tensile load rise time (from 0.1s to 10 s). For the test under square waveform (sq) at $R = 0.25$, the result showed that CFCGR increased gradually with increase in hold-time from 0.01 – 1 s. The effect was more pronounced at lower ΔK . In the distilled water and at a frequency of 5Hz, $R = 0.25$ and under different waveforms, the CFCGR in the En24 steel increased thus; positive sawtooth (p-s)>triangular (tr)>square (sq)>negative sawtooth (n-s). The rise time for the p-s was 0.2s while that of tr was 0.1s. It is pertinent to note that the frequencies for these tests are not the same due to different hold-times, hence the independent effect of frequency reduction or hold-time increase on the waveforms cannot be delineated in this study. They also showed that for 300m steel in distilled water at frequency of 0.45Hz and $R = 0.5$, the CFCGR under sinewave is higher than triangular waveform and the lowest was the square waveform.

Nakasa [151] studied CFCGR in JIS SNCM439 steel quenched and tempered at 200°C and 500°C. He reported that CFCGR was higher in square load waveform than positive sawtooth for steel with high sensitivity to delayed failure. For the sample tempered at 500°C, which they reported has delayed failure under static stress, the CFCGR was higher under positive sawtooth compared with square waveform. They suggested that stress rise time and hydrogen concentration

control the crack growth in corrosion-fatigue. This implies that the interaction between the crack tip and hydrogen atoms controlled the crack growth mechanism.

Bhuyan [152] studied the effect of waveform on CFCGR in a steel manufactured to CSA G40.21M 350 in air and *SW* under constant amplitude loading. The pH of the *SW* was in the range 7.25-7.55 while the temperature was in the range 0°C - 21°C in *SW* and -15 to 4°C for the tests in air. The tests were carried out using sinusoidal and sawtooth waveforms. They reported that CFCGR in *SW* was on average 2.7 times higher than that in air for $R = 0.1$, frequency of 0.05Hz and in temperature range of 0 to 4°C. Their work showed that the CFCGR became almost similar to the value in air as the frequency was increased from 0.05Hz to 0.5Hz at $R = 0.1$. They reported that CFCGR increased by a factor of 2.3 as frequency was decreased from 0.5 to 0.05Hz. For the air test, the FCGR was almost unaffected in the temperature range of -15 to 4°C and was generally not influenced by the load ratio variation in the range of 0.05 – 0.3.

In a study conducted by Thorpe [24], the R and frequency were found to have no effect on FCGR in air also. Similar results were obtained from the test with triangular, positive sawtooth and sinusoidal waveforms. They also showed that in the corrosion-fatigue tests the CFCGR increased as test frequency is decreased, and R increased. Appleton [157] studied the CFCGR in BS4360 grade 50D steel in 3% NaCl at room temperature and found that there was no difference in the CFCGR for the tests done at 0.05Hz, 0.17Hz and 1Hz up to about ΔK of 18MPa $\sqrt{m}$. However, above 18MPa $\sqrt{m}$, CFCGR at 0.17Hz increased steadily above that of 1Hz. A large decrease in the CFCGR was seen at the test performed at 5Hz. In general, the effect of *SW* started decreasing steadily from 1Hz upwards towards that of air test value. Moreover, the results showed that for 0.17 Hz tests, there was no significant difference in the CFCGR curve in the R range of 0.1 - 0.7. Appleton [157] also performed tests at other different waveforms. The additional waveforms were triangular and square shape at 0.17

Hz. No difference in CFCGR was observed between sinusoidal and triangular waveforms.

Barsom [161] studied waveform effect on the FCGR in a 12Ni-5Cr-3Mo maraging steel and the test was performed in air and 3%NaCl and he compared the effect of sine, triangular, and square waveforms loadings. Frequency in the range 0.1 - 10Hz was used. The result showed that similar FCGR results were obtained in air for the three waveforms. In other words, waveform has no effect on crack growth in the air. In the 3%NaCl, the CFCGRs of the square waveform and negative sawtooth were the same and slightly higher than the air data, by a factor of 1.1. For the sinusoidal, triangular and positive-sawtooth waveforms the CFCGRs were similar and higher than FCGR in air by a factor of 3. Barsom [162] also showed that there was no significant difference in FCGR for test under random-sequence and ordered sequence load fluctuations. Wand, Z. F. [166] studied the influence of waveform on CFCGR in iron in 3.5%NaCl. The yield stress, tensile strength and elongation of iron are 150MPa, 295MPa and 42% respectively. The test frequency was 0.1 or 1Hz and $R = 0$. The waveforms investigated were sinusoidal, positive sawtooth, negative sawtooth and square waveform. They found that the positive sawtooth and square loading waveforms have very little or no effect on the FCGR in air. This result tends to confirm that, waveform generally has no effect on the rate at which a fatigue crack grows in steel in air.

Knop [163] studied the effect of cycle frequency rise-time on CFCGR in a high-strength steels. They also presented the effects of fall-times and hold-times at maximum and minimum loads on the CFCGRs. They found that FCG at the Paris Region is generally by plastic blunting and that the a increased during the rising of the load towards the cycle peak for each load cycle. They reported that the difference between crack growth in inert and embrittling environments was that less crack tip blunting occurred in the embrittling environment due to localisation of deformation slip and this resulted to increased crack growth rate. They suggested that CFCGR depended on the concentration of hydrogen adsorbed at

crack tips and at tips of nanovoids ahead of crack fronts. They concluded that crack growth is enhanced under increasing rise time and that longer fall times and longer hold-time at maximum and minimum loads may enhance crack growth when the rise time is short. In their study, the interval of increased effect of rise time was from 1 to about 10 s and the growth rate within this range depended on the ΔK .

7.5 Research motivation for the present study

A comprehensive review of the existing studies on frequency and waveform effects on corrosion-fatigue has revealed that CFCGR is sensitive to waveform and frequency. This sensitivity to waveform underscores the need to perform CFCGR tests based on the operational loading conditions that a structure is subjected to. Hence, in fatigue analysis of OWT foundations, a more practical analysis can be achieved under realistic representation of service loads, actual frequencies and waveforms. As much as perfect cyclic waveform of wind loading is difficult to obtain and simulate in a fatigue test, a near match waveform is necessary to generate realistic data for assessment of crack growth rates in OWT foundations. This will increase confidence in prediction of fatigue life and inspection planning of OWTs.

In all the studies mentioned above, there is no explicit comparison of waveform effect under the same frequency and load ratio for subgrades of a particular grade of steel with differing microstructures. There is also limited evidence of the cracking micrographs and in-depth investigation of contributions of the microstructural phases under these waveforms in corrosion-fatigue tests. The influence of waveforms other than sinusoidal on CFCGR for the steels used for OWT foundation design is non-existent, at least to the extent of information available to the authors. Therefore, the intention of the present study is to determine the effect of two different types of waveforms, sinewave and trapezoid waveform with hold-time at the maximum load, as well as various frequencies on CFCGR in *SW* to identify the most damaging waveform, hence the most

conservative analysis that can be employed in the life assessment of OWT monopile foundations.

In general, this paper presents new sets of experimental results on the effect of frequency, waveform and load level/ $K_{\max}$ on the corrosion-fatigue behavior of normalised-rolled (NR) and thermomechanical control process (TMCP) ferrite-pearlite steel subgrades of S355 which are widely used in the fabrication of modern offshore structures, including OWT monopile foundations. Sinewave and hold-time waveforms were considered and the experiments were carried out using frequencies of 0.2Hz, 0.3Hz, 0.5Hz at fixed stress ratio of $R=0.1$. The experiments performed in this study are expected to assist towards more reliable assessment of corrosion-fatigue in OWT foundations in free-corrosion condition.

7.6 Material Selection and Specimen Preparation

Traditionally, BS4360 Grade 50D was widely used in the design of offshore structures. In recent years, European Standard EN10025 S355 has become the main structural steel typically in use for OWT structures [33][34][230] and different sub-grades of S355 are selected for use in offshore applications, including monopiles. In fact, the equivalent grade of BS4360 Grade 50D is S355J2+N. The modern material choice for offshore application and the conventional structural steels are shown in **Table 7-1**. The materials for corrosion-fatigue tests in this study are S335G8+M, S355G10+M and S355J2+N, the first two of which are thermo-mechanical control processed (TMCP) steels while S355J2+N is normalised-rolled steel (NR). **Table 7-2** gives the chemical and mechanical properties of these sub-grades of S355 steel.

The FCG behaviour of S355G10+M in air has been previously characterised by the authors and the results are available in reference [144]. The intention of this study is to characterise the FCG behaviour of these subgrades in free-corroding environment and to examine the effects of hold-time and frequency on the FCGRs. In this study, compact tension C(T) specimens were extracted from the mid-thickness of a 90mm thick plates shown in **Figure 4-1 (a)** and **(b)**. The

specimens were extracted in such a way that they were far from previously hot rolled surfaces.

Table 7-1 S355 structural steels for marine applications, (a) advanced S355 subgrades (b) Conventional S355 structural steel grades

Offshore Structural Steel Plates			
(a)		(b)	
European	Typical offshore application	European	British
EN10225		EN10025 Part 2: 2004	BS4360:1990
S355G7+N	Primary Structure	S355	50A
S355G7+M	Primary Structure	S355JR	50B
S355G8+N	Critical Joints	S355JO	50C
S355G8+M	Critical Joints	S355J2	50D
S355G9+N	Primary Structure	S355J2+N	50D
S355G9+M	Primary Structure	S355K2	50DD
S355G10+N	Critical Joints	S355K2+N	50DD
S355G10+M	Critical Joints	S355NL	50EE

Table 7-2 Chemical compositions and mechanical properties of the examined materials in this study

Steel Grade	Notation	C	Mn	Ni	Si	Cu	Cr	Mo	E %	YS (MPa)	UTS (MPa)
S355G10+M (TMCP)	G10+M (G10)	0.06	1.57	0.33	0.27	0.24	0.03	0.01	38	435	545
S355G8+M (TMCP)	G8+M (G8)	0.05	1.52	0.32	0.27	0.24	0.03	0.01	35	447	549
S355J2+N (NR) [171]	J2+N (J2N)	0.17	1.54	0.04	0.04	0.08	0.02	-	20	385	531

Figure 4-1 (c) shows the 3D schematic of the experimental specimen manufactured following ASTM E647–15 standard [172]. The C(T) specimen dimensions are given in **Table 7-3**.

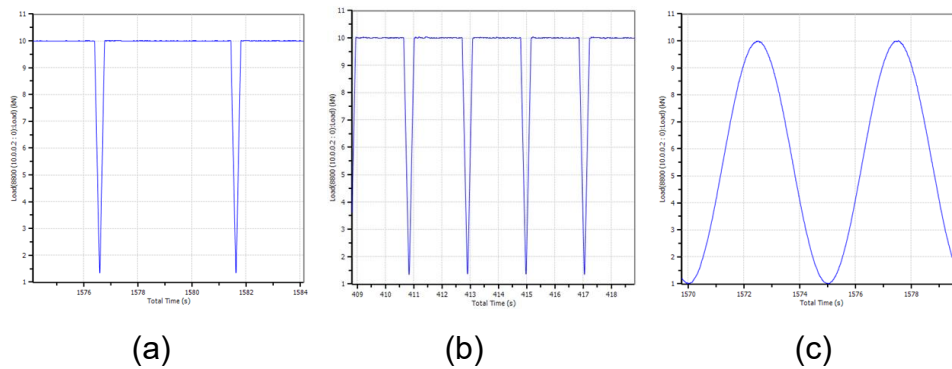
Table 7-3 Specimen dimensions and loading conditions in seawater tests:

#	Test ID.	W (mm)	B (mm)	a_o (mm)	a_f (mm)	f (Hz)	P_{max} (kN)
1	J2N-S10F0.2	50.00	16.00	18.32	28.02	0.2	10
2	J2N-H10F0.2	50.00	16.00	18.55	31.07	0.2	10
3	G8-H10F0.2	50.01	16.02	17.03	33.83	0.2	10
4	G10-S10F0.2	50.0	16.00	21.26	29.11	0.2	10
5	G10-H10F0.2	50.05	16.08	22.37	29.38	0.2	10
6	G10-S10F0.3	50.03	16.05	21.57	29.44	0.3	10
7	G10-H10F0.3	50.03	16.07	21.87	29.73	0.3	10
8	G10-S10F0.5	50.03	16.04	21.20	29.61	0.5	10
9	G10-H10F0.5	50.03	16.04	21.26	29.33	0.5	10
10	G10-S9F0.2	50.00	16.00	22.68	30.16	0.2	9
11	G10-H9F0.2	50.00	16.00	22.98	29.81	0.2	9
12	G10-S9F0.3	50.01	16.16	21.69	29.30	0.3	9
13	G10-H9F0.3	50.01	16.16	20.22	29.51	0.3	9
14	G10-S9F0.5	50.05	16.06	21.95	30.21	0.5	9
15	G10-H9F0.5	50.05	16.08	22.09	30.30	0.5	9

15 corrosion-fatigue tests were conducted in free-corroding condition. As seen in this table, all tests were performed on standard C(T) specimens with the width of $W=50$ mm and average thickness of $B=16$ mm. Also included in this table are the initial crack length, a_o , (i.e. after pre-fatigue cracking) and final crack length, a_f , for each test. Note that in test IDs, each samples is denoted with material type (e.g. J2N, G8 and G10), followed by the environmental/waveform condition (e.g. S which stands for corrosion-fatigue test with sinewave and H which stands for corrosion-fatigue test with hold-time), followed by the maximum load applied, P_{max} , on the sample (e.g. 9, 10 and 12 kN), followed by the frequency (e.g. F of 0.2, 0.3, 0.5, 2, 4 and 5 Hz). Also included in this table are the loading conditions for each test. All tests were conducted using the same load ratio of $R = 0.1$. All the fatigue tests were conducted in accordance with ASTM E647-15 [172] and BS 7910 [174]. The simulated SW for corrosion-fatigue tests was prepared in accordance with ASTM D1141-98 [144][173]. The test Set-up, crack growth monitoring in SW, a estimations using back face strain measurements and FCGR calculations are comprehensively described in Chapter 3.

7.7 Experimental waveforms in corrosion-fatigue tests

Two waveforms have been employed in this study; sinusoidal waveform and square wave form with hold-time at the maximum applied load. A commercial Instron software was used to programme the two waveforms in corrosion-fatigue tests. The load ratio of $R=0.1$ was applied in all tests and the effects of different frequencies of 0.2, 0.3Hz and 0.5 Hz on corrosion-fatigue crack growth behaviour were examined. The loading and unloading rates for the hold-time waveform were kept constant for the three frequencies and different load levels while the hold-time was varied depending on the test frequency. As an example, for 9kN corrosion-fatigue tests with hold-time waveform the test started with 0.9 kN (i.e. P_{min}) and then the load was ramped up to 9 kN (i.e. P_{max}) at a rate of 48.6kN/s and upon completion of the pre-defined hold-time period the sample was unloaded to 0.9 kN with the same rate of 48.6kN/s. This implies that the rise time from the P_{min} to P_{max} was about 0.17, 0.19 and 0.22 seconds in tests with the maximum applied load of 9, 10 and 12 kN, respectively. The frequencies 0.2, 0.3 and 0.5 Hz were generated by varying the hold-time at P_{max} while the loading and unloading rates were kept at 48.6kN/s in all tests.



applied frequency (e.g. 4.62s and 1.62s for the tests with applied frequency of 0.2 and 0.5 Hz, respectively).

7.8 Corrosion-Fatigue Crack Growth Rate Results and Discussion

7.8.1 Test results on S355G10+M

The CFCGR results obtained on S355G10+M for freely corroding conditions in SW for the sinewave and hold-time are shown in **Figure 7-2**.

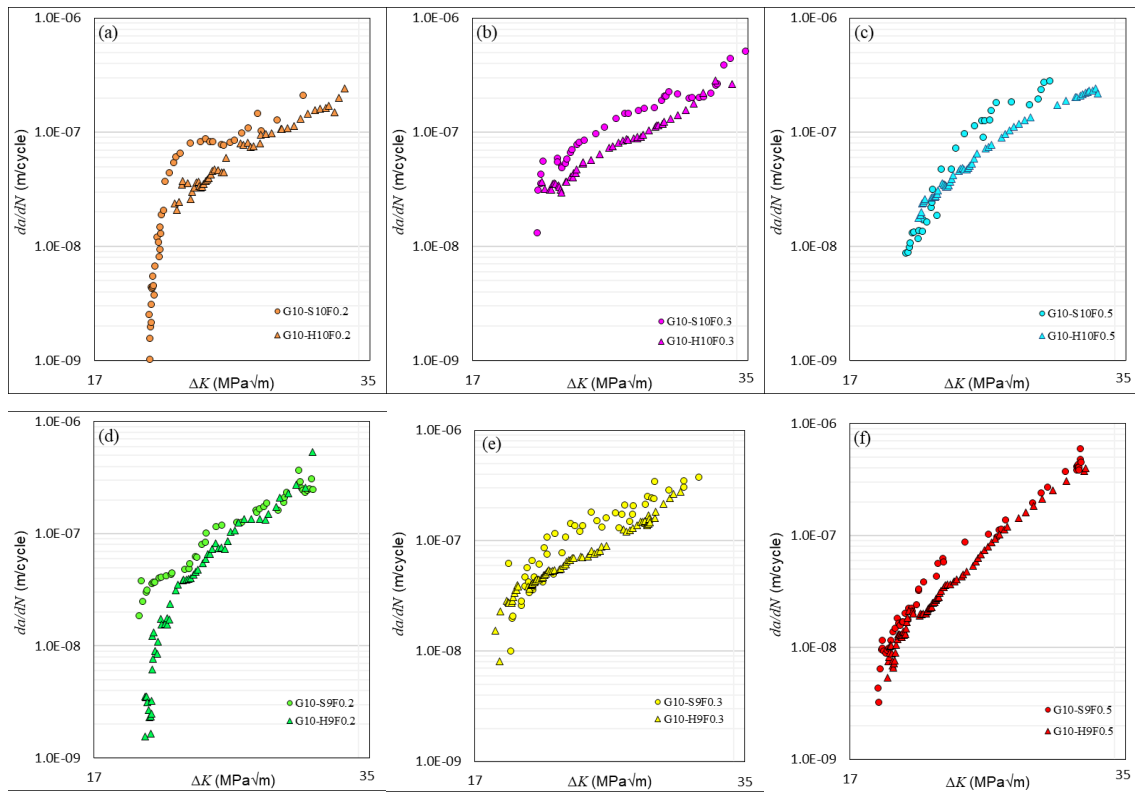


Figure 7-2: The effect of waveforms on CFCGRs in S355G10+M (a) $P_{max} = 10\text{kN}$, $f = 0.2\text{Hz}$, (b) $P_{max} = 10\text{kN}$, $f = 0.3\text{Hz}$, (c) $P_{max} = 10\text{kN}$, $f = 0.5\text{Hz}$, (d) $P_{max} = 9\text{kN}$, $f = 0.2\text{Hz}$, (e) $P_{max} = 9\text{kN}$, $f = 0.3\text{Hz}$, and (f) $P_{max} = 9\text{kN}$, $f = 0.5\text{Hz}$

The results from these 12 tests (see **Table 7-3**) are shown in pairs in **Figure 7-2 (a – f)** in order to compare the CFCGR in S355G10+M (denoted G10 in specimen

IDs) steel subgrade for the two waveforms, under frequencies of 0.2Hz, 0.3Hz and 0.5Hz.

In **Figure 7-2 (a – f)** the sine waveform and hold-time data points are shown with circles and triangles, respectively. **Figure 7-2 (a – c)** show the results from the tests performed at 10kN under various frequencies while **Figure 7-2 (d – f)** present the data for 9kN maximum applied load. It can be seen in **Figure 7-2 (a – c)** that under $P_{max} = 10\text{kN}$, the sine waveform results have exhibited higher CFCGR in the Paris Region, as compared with the hold-time waveform data, in all tests performed at various frequencies. Also, seen in **Figure 7-2 (a – c)** is that the separation in the crack growth rates between sinewave and hold-time waveform is more pronounced at lower frequencies of 0.2Hz and 0.3Hz in the Paris Region. Moreover, the results show that as the frequency increases to 0.5Hz, this separation appears to decrease and the data points from both waveforms fall upon each other in the lower ΔK region.

The results in **Figure 7-2 (d – f)** for the tests performed under $P_{max} = 9\text{kN}$ show the same trend as in 10kN tests. Similar to the observed trend at the higher load level in **Figure 7-2 (a – c)**, the results in **Figure 7-2 (d – f)** suggest that increasing the frequency appears to converge the CFCGRs of the two waveforms. In general, the test data presented in **Figure 7-2** show that for a given value of ΔK , the CFCGR appears to be higher in sinewave compared to the hold-time waveform in the Paris Region and the separation between the two waveform data sets is more pronounced at lower frequencies and higher loads. This indicates that characterising the CFCG behaviour of S355G10+M using sine waveforms can result in more conservative life predictions for offshore structures.

In order to examine the influence of load level on the CFCGR using different waveforms, the results on S355G10+M have been replotted in **Figure 7-3** for direct comparison between 9kN and 10kN data sets obtained using sine and hold-time waveforms at various frequencies. The results in **Figure 7-3 (a – c)** show that when corrosion-fatigue tests on S355G10+M were performed using sinewave,

the change in P_{max} from 10kN to 9kN resulted to slight increase in the CFCGR of the 9kN above that of 10kN in the Paris Region for lower frequencies 0.2Hz and 0.3Hz. This difference appeared to diminish as the frequency was increased to 0.5Hz.

Also, the results in **Figure 7-3 (d – f)** show that when the hold-time waveform is employed in the test higher CFCGRs were observed in the 9kN data sets, as compared to 10kN data, across the entire range of frequencies examined in this study. However, as seen in **Figure 7-3 (d – f)**, increasing the frequency to 0.5Hz causes the data points from 9kN and 10kN data sets to fall upon each other at lower ΔK values. But, the curves started to diverge at higher ΔK values of about 25.53 MPa $\sqrt{m}$. The interesting observation is that generally, the CFCGRs at 9kN are higher than that of 10kN data sets in **Figure 7-3**. This tends to suggest that the size of the local plasticity ahead of the crack tip plays a role in decelerating or accelerating the CFCGR in seawater – since the plastic zone size depends on the fatigue load level.

To further investigate the effect of frequency and applied load level in the presence of different waveforms, the results obtained from all tests performed on S355G10+M using sinewave and hold-time waveform are collectively presented in **Figure 7-4 (a)** and **Figure 7-4 (b)** respectively. **Figure 7-4 (a)** shows that in all corrosion-fatigue tests performed using the sinewave, there is no obvious difference in the CFCG trends within the Paris region obtained from the tests performed at different load levels (i.e. P_{max} of 9kN and 10kN) and frequencies (i.e. f of 0.2Hz, 0.3Hz and 0.5Hz), with the exception of the increase seen in 9kN, 0.3Hz test at lower ΔK . In other words, the sensitivity of CFCG to the frequency under sinewave is little within the Paris Region, in the range of 0.2Hz – 0.5Hz, and maximum applied load level in the range of 9kN – 10kN.

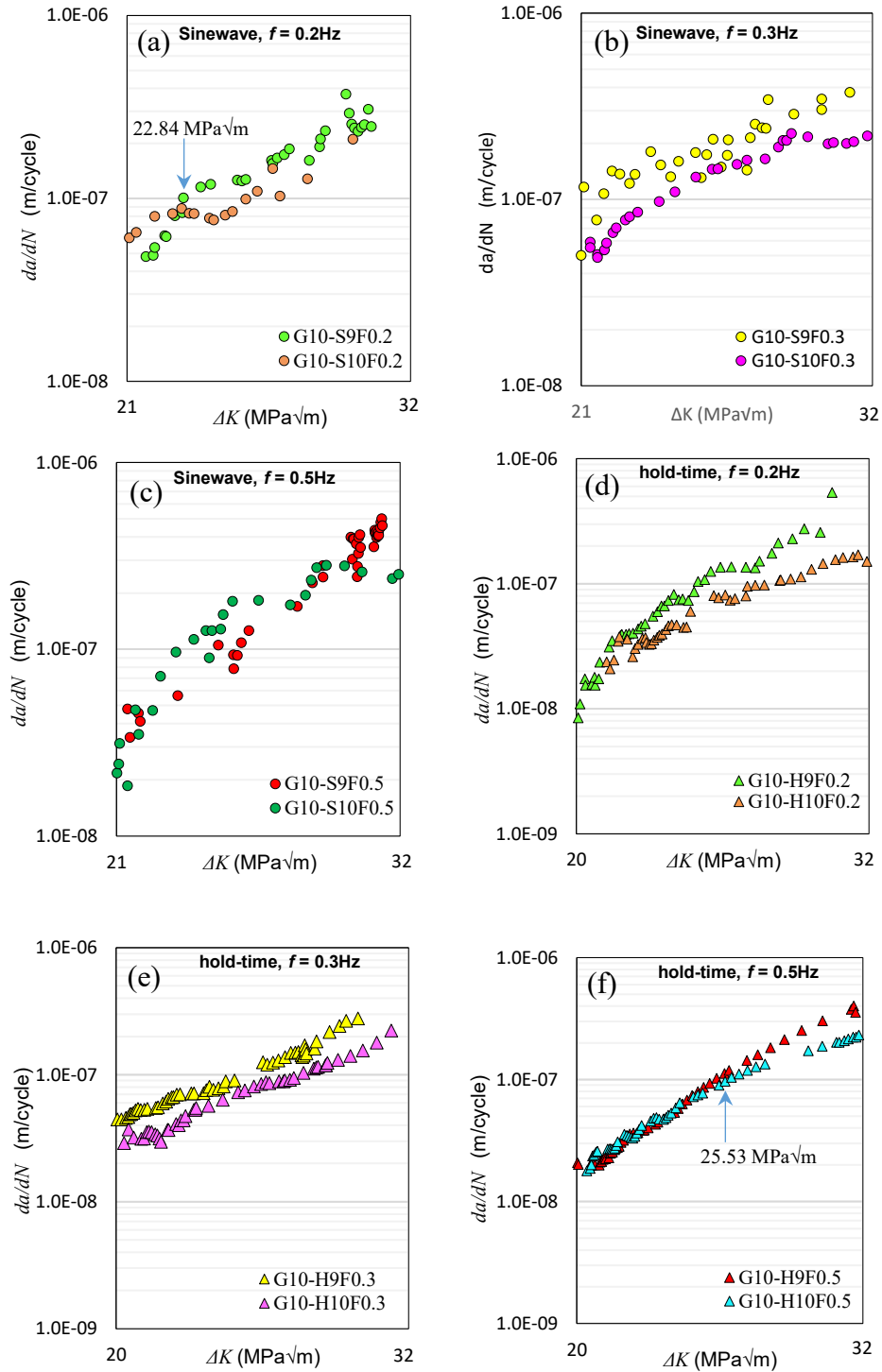


Figure 7-3: The effect of fatigue load level on the CFCGR using different waveforms in S355G10+M (a) sinewave with $f = 0.2\text{Hz}$, (b) sinewave with $f = 0.3\text{Hz}$, (c) sinewave with $f = 0.5\text{Hz}$, (d) hold-time waveform with $f = 0.2\text{Hz}$, (e) hold-time waveform with $f = 0.3\text{Hz}$, and (f) hold-time waveform with $f = 0.5\text{Hz}$

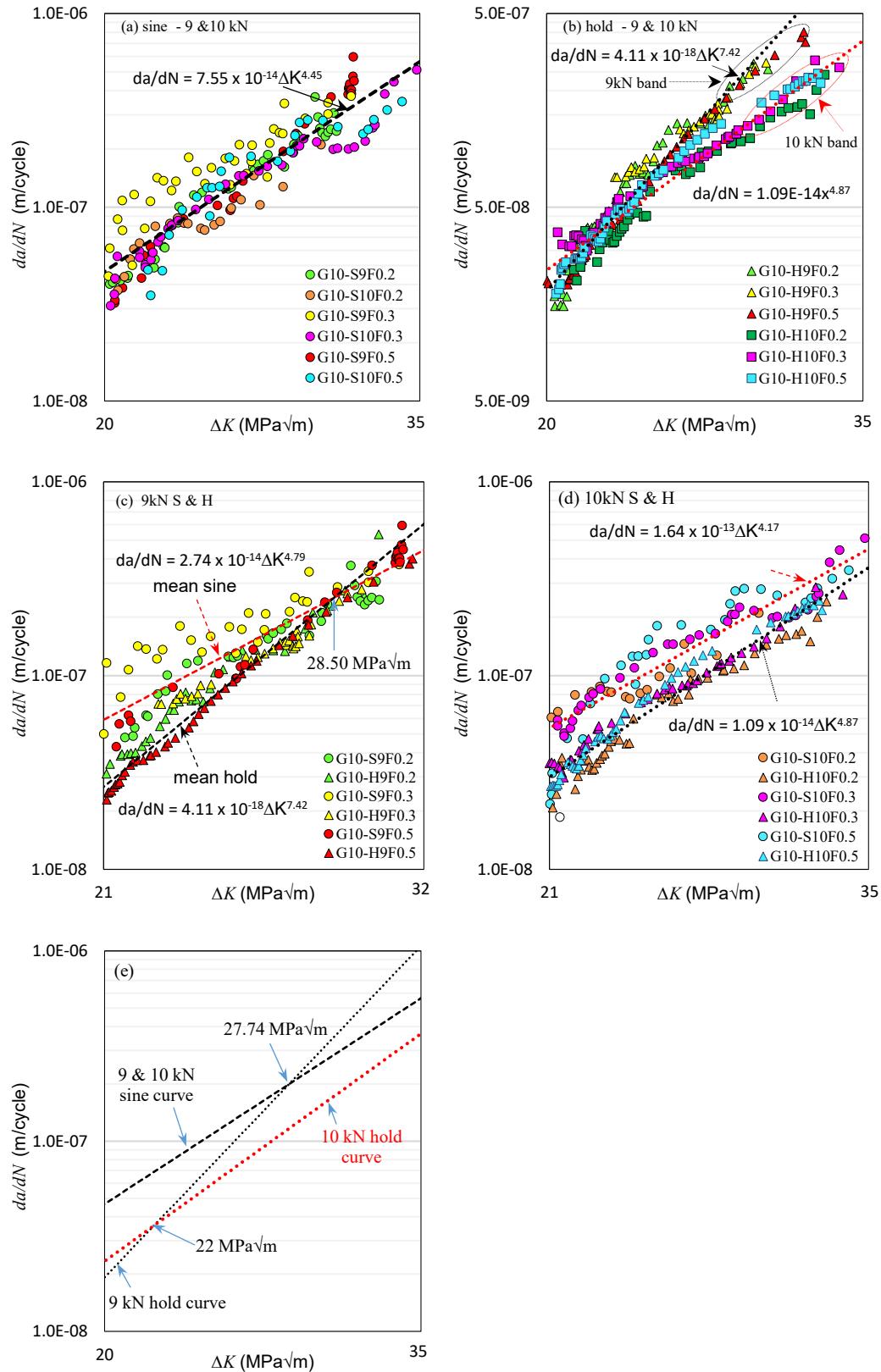


Figure 7-4: The effect of waveform, frequency and load level on CFCGR in S355G10+M under (a) sine waveform and (b) hold-time waveform (c) 9kN (d) 10kN (e) combination of mean curves of (a) and (b)

A line of best fit has been made to the entire set of data points in **Figure 7-4 (a)** and the Paris law constants for the sinewave tests are identified for $20.19 \leq \Delta K \leq 35 \text{ MPa}\sqrt{\text{m}}$ and summarized in **Table 7-4** for the G10+M steel.

Table 7-4 Paris law constants for CFCG tests performed on S355G10+M using sine and hold-time waveforms

Waveform type	Load level (kN)	C	m	R^2
Sine waveform	9 & 10	7.55×10^{-14}	4.45	0.859
Hold-time waveform	9	4.11×10^{-18}	7.42	0.960
	10	1.09×10^{-14}	4.87	0.965

Figure 7-4 (b) shows that for the corrosion-fatigue tests performed with the hold-time waveform there is generally no obvious trend observed in the CFCGR below a ΔK of $24 \text{ MPa}\sqrt{\text{m}}$. However, in the region with ΔK of greater than $24 \text{ MPa}\sqrt{\text{m}}$ the CFCGR for 9kN tests (with the frequencies of 0.2Hz, 0.3Hz and 0.5Hz) are found to be higher than the 10kN tests (with the frequencies of 0.2Hz, 0.3Hz and 0.5Hz). Thus, the results in the high ΔK region are sensitive to the loading condition. This may be associated with larger plastic zones ahead of the crack tip in the higher ΔK region in the tests performed with the higher load, which has led to some level of reduction in CFCGRs due to crack tip blunting by the corrosion process. Due to the clear separation of the CFCG trends in the high ΔK region, two separate lines of best fit have been made to the 9kN and 10kN data points and the inferred Paris law constants are included in **Table 7-4**.

Figure 7-4 (c) is the comparison of the waveform tests under 9kN. As seen previously, the CFCGR under sinewave is higher under 9kN and the rate decreased with increase in ΔK . The mean curves for the sine and hold-time waveforms are included. At about $28.50 \text{ MPa}\sqrt{\text{m}}$, the CFCGR of the hold-time appears to overtake that of sinewave. This tends to suggest that at very high ΔK , the CFCGR of the hold-time will be higher. This trend can also be seen from the Paris law constants of the curves shown in the figure and listed **Table 7-5**. The hold-time exponent is very high which gives the steeply nature of the curve.

Table 7-5 Paris law constants for CFCG tests under sine and hold-time waveforms for all 9kN and 10kN data for the S355G10+M

9kN sine (SW)	C	m	R ²
Mean	2.74×10^{-14}	4.79	0.866
Mean +2SD	4.67×10^{-14}	4.79	
Mean -2SD	1.61×10^{-14}	4.79	
9kN hold (SW)			
Mean	4.11×10^{-18}	7.42	0.960
Mean +2SD	5.73×10^{-18}	7.42	
Mean -2SD	2.95×10^{-18}	7.42	
10kN sine (SW)			
Mean	1.64×10^{-13}	4.17	0.912
Mean +2SD	2.46×10^{-13}	4.17	
Mean -2SD	1.10×10^{-13}	4.17	
10kN hold (SW)			
Mean	1.09×10^{-14}	4.87	0.965
Mean +2SD	1.41×10^{-14}	4.87	
Mean -2SD	8.41×10^{-15}	4.87	

Similar comparison of the waveform tests under 10kN is show in **Figure 7-4 (d)**. Under an increased load, the CFCGR of the sinewave is clearly and consistently higher than that of the hold-time in the Paris Region, though the factor of difference tends to gradually decrease with increasing ΔK . The exponential values of the Paris law, as listed in **Table 7-5** show also that the difference in the growth rates is gradually decreasing.

Figure 7-4 (e) is the combination of the mean Paris curves of **Figure 7-4 (a)** and **Figure 7-4 (b)**. Below ΔK of about $27.74 \text{ MPa}\sqrt{\text{m}}$, the CFCGR under sinewave is clearly higher than that of the hold time tests. Above this value, the CFCGR of the sinewave continues to be higher than that of 10kN hold-time test by an average, almost constant, factor of 1.55, but lower than that of 9kN hold-time tests and the factor of difference increases with increase in ΔK . Generally, this indicates that within the frequency range of 0.2Hz to 0.5Hz, the life assessment of offshore wind turbine monopiles made of S355G10+M will be more conservative when Paris law constants from sinewave corrosion-fatigue tests are employed in the analysis.

Another important observation in **Figure 7-4 (e)** is that below $22 \text{ MPa}\sqrt{\text{m}}$, the CFCGR of the 9kN hold-time is as may be expected lower than that of the 10kN hold-time test, but above this $22 \text{ MPa}\sqrt{\text{m}}$ and at a crack growth rate of $3.60 \times 10^{-8} \text{ m/cycle}$, its CFCGR became higher than that of the 10kN test. And, above $27.74 \text{ MPa}\sqrt{\text{m}}$ and at a crack growth rate of $1.42 \times 10^{-7} \text{ m/cycle}$, the CFCGR of the 9kN became higher than even that of the sinewave. This phenomenon of transitions may be explained using what we hereby referred to as *microplastic* zone geometry or size interaction with the corrosion process. For the hold-time test, below $22 \text{ MPa}\sqrt{\text{m}}$, this *microplastic* zone size in the 10kN is considerably small so that the interaction of the corrosion process with the small *microplastic* zone does not lead to rate retarding blunting process. In this domain, the higher driving force provided by the 10kN naturally increases the CFCGR more than that of lower driving force of 9kN. As the ΔK increases, the *microplastic* zone size gradually increases, and the rate retarding blunting process gradually develops in significance. In this domain, the blunting of the main active crack tip due to increase in this *microplastic* zone size appears to be the CFCGR controlling factor rather than the load level for the hold-time in seawater. The same can be said of the sinewave, but the effect of the *microplastic* zone size blunting process appears to be less severe than that of the hold-time up to about ΔK of $27.74 \text{ MPa}\sqrt{\text{m}}$. At high ΔK , the blunting process in the sinewave was significant to the point that its CFCGR became lower than that of the 9kN hold-time test. This transition at $27.74 \text{ MPa}\sqrt{\text{m}}$, appears to support the assertion that the main crack tip blunting process is a factor controlling the CFCGR retardation at high ΔK and low frequency in SW. And difference in the frequencies of 0.2Hz, 0.3Hz and 0.5Hz have little or no effect on the CFCGR process.

Error! Reference source not found. **(a)** shows the comparison of all the sinewave and hold-time CFCGRs data at all frequencies and load levels with the FCGR in air.

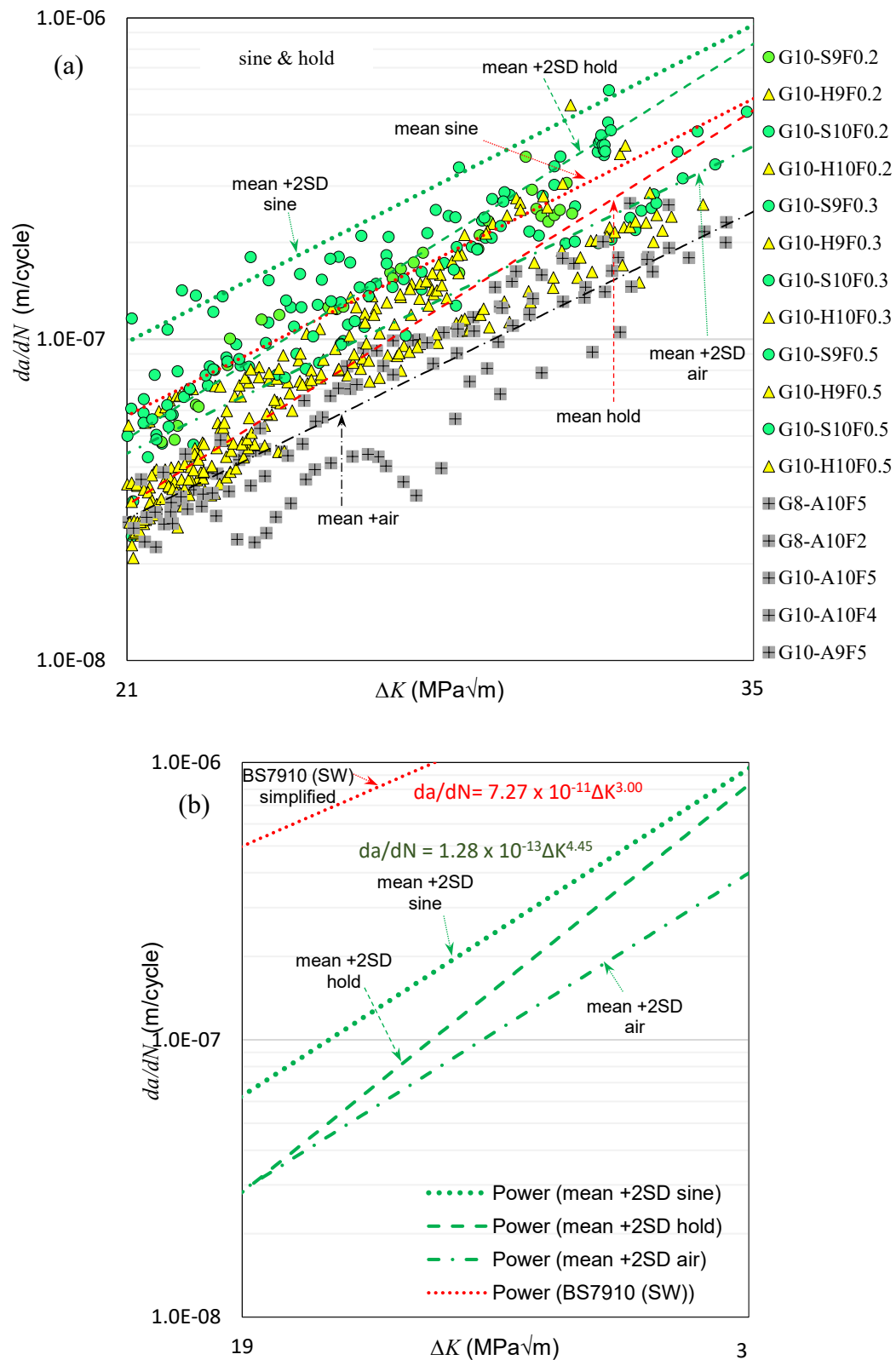


Figure 7-5: Comparison of all CFCGRs under sinewave and hold-time with FCGR in air for TMCP steels

Table 7-6 Paris law constants for all CFCG tests conditions for S355G10+M

Paris constants – All data points:	C	m	R ²
Air only			
Mean	5.33×10^{-14}	4.31	0.909
Mean +2SD	8.83×10^{-14}	4.31	
SW (sinewave) only			
Mean	7.55×10^{-14}	4.45	0.862
Mean +2SD	1.28×10^{-13}	4.45	
SW (hold-time) only			
Mean	1.44×10^{-15}	5.54	0.898
Mean +2SD	2.32×10^{-15}	5.54	

Table 7-7 Factor of difference between FCGR in air and the CFCGRs under hold-time and sinewave

(a)

ΔK	21	24	27	32	35	Ave.
Mean curve						
Air only	1.00	1.00	1.00	1.00	1.00	1.00
Hold-time (SW) only	1.10	1.27	1.51	1.72	1.82	1.48
Sinewave (SW) only	2.09	2.13	2.16	2.21	2.25	2.17

(b)

ΔK	19	21	24	27	31	35	Ave.
+2SD Mean curve							
Air only	1.00	1.00	1.00	1.00	1.00	1.00	
Hold-time (SW) only	1.00	1.15	1.31	1.52	1.81	2.07	1.48
Sinewave (SW) only	2.19	2.21	2.26	2.29	2.34	2.53	2.30
BS7910 simplified (SW)	7.86	6.66	5.65	-	-	-	-

Included in Error! Reference source not found. (a) are the mean and plus 2 standard deviation (+2SD) curves of the data points and the Paris Law constants are listed in **Table 7-9**. The air data are coloured grey, the hold-time data yellow and the sinewave data green. It is very clear that there is enhancement of the FCGR in seawater by a factor that depends on the waveform and load level in the Paris Region. It is also obvious that the sinewave is the most damaging waveform at all experimental conditions. However, the CFCGR of the hold-time

is a strong function of the load level. Error! Reference source not found. **(a)** also shows the mean Paris curve for (a) all the datapoints (SW & Air) on the graph, (b) air only, (c) sine (SW) only and (d) hold (SW) only. With reference to the mean air curve, **Table 7-7 (a)** shows that the factor of difference in SW for the growth rates of the sine and hold-time. The factors show that CFCGRs of the two waveforms increased with increase in the ΔK with respect to the FCGR in air. The mean CFCGR of the hold-time is generally higher than that of air by an average factor of 1.48 while that of sine is about 2.17.

The green curves are the corresponding mean plus 2 standard deviation (+2SD) curves. Since the mean +2SD curves are commonly used for design purposes, these curves have been compared separately, for clarity, in Error! Reference source not found. **(b)**. Error! Reference source not found. **(b)** shows the mean +2SD curves for crack growth rates in air and in SW under the two waveforms. The CFCGR of the hold-time tends to increase gradually to a very high growth rate at high ΔK . For the sinewave, the CFCGR was highest which decreased gradually across the Paris Region with respect to the hold-time. **Table 7-7 (b)** lists the factor of difference for the mean +2SD curves. The factors again show that the CFCGRs of the two waveforms increased with increase in the ΔK with respect to the FCGR in air. The average factor of crack growth rate with respect to air is the same as in mean curve and that of sinewave is 2.3. In general, the sinewave was consistently higher within the range of ΔK considered, with average factor of about 1.5 above the hold-time value and 2.24 above the air crack growth rate. The 2.24 was obtained by taking the average value of mean and mean +2SD. The BS7910 curve included in Error! Reference source not found. **(b)** and the factor of difference with respect to the sinewave shows the extent of conservativeness in the recommended design curve in the Standard.

7.8.2 Test results on S355G8+M and S355J2+N

To verify the trends observed in **Figure 7-2**, **Figure 7-3** and **Figure 7-4** for the S355G10+M steel, one additional test on S355G8+M (denoted G8-H10F0.2 in

Table 7-3) and two additional tests on S355J2+N (denoted J2N-S10F0.2 and J2N-H10F0.2 in **Table 7-3**) were conducted under $P_{max} = 10\text{kN}$, $f = 0.2\text{Hz}$ and $R = 0.1$. The results from these tests are presented in **Figure 7-6 (a)** and **Figure 7-6 (b)** for S355G8+M and S355J2+N, respectively, and compared with the sinewave CFCGR test data from previous studies on the same batches of material conducted by Mehmanparast [175] (denoted G8-S9F0.3A3, G8-S9F0.3A4, G8-S10F0.3B3 and G8-S10F0.3B4) and Adedipe [231] (denoted J2N-S9F0.4, J2N-S9F0.3 and J2N-S12F0.3).

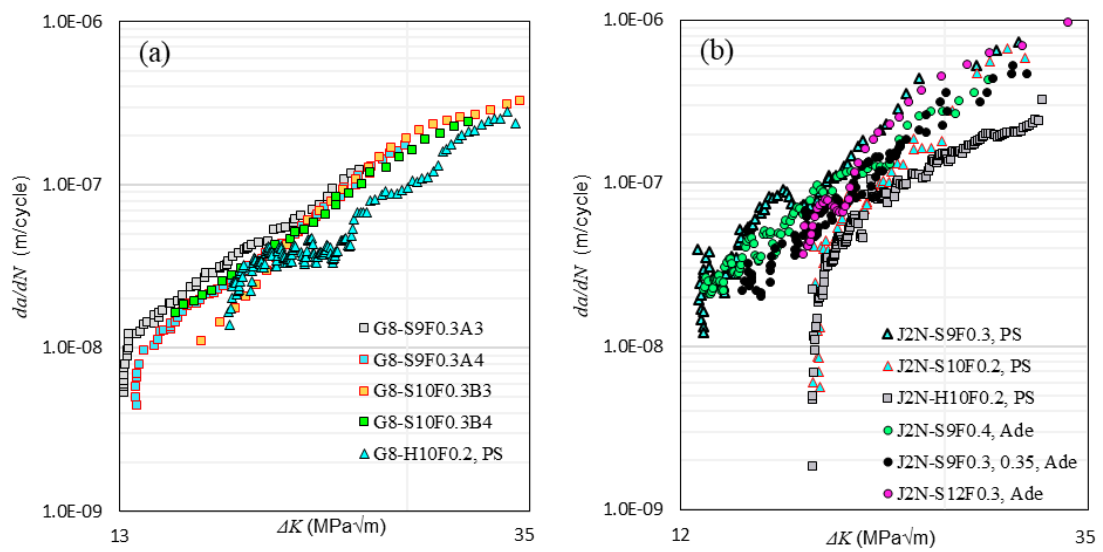


Figure 7-6: Comparison of new test results on (a) S355G8+M (blue triangular data points) and (b) S355J2+N (grey rectangular data points), with the literature data. (PS – present study, Ade – Adedipe)

It can be seen in **Figure 7-6 (a)** that the CFCGR in the test with the hold-time waveform is generally lower than those obtained from sinewave tests. This observation tends to confirm that similar to S355G10+M steel, the CFCG behaviour of S355G8+M is also sensitive to the cyclic waveform. Lower CFCGR is obtained when the tests were performed with hold-time. Also seen in **Figure 7-6 (a)** is that, similar to S355G10+M, the tests with sinewave on S355G8+M exhibited little or no sensitivity to the maximum applied fatigue load, in the range of 9kN – 10kN, within the Paris Region.

The results in **Figure 7-6 (b)** show similar results to S355G10+M and S355G8+M. The hold-time fatigue cycles have resulted in reduced CFCGRs in S355J2+N steel compared to the data points obtained from sinewave tests. It can be seen in this figure that for test under $P_{max} = 10\text{kN}$ and $f = 0.2\text{Hz}$, a lower CFCG trend is obtained from the hold-time test compared to the sinewave, and the separation between the two data sets increased with increase in ΔK . Also seen in **Figure 7-6 (b)** is that CFCGRs of the 9kN sinewave tests at 0.3Hz and 0.4hz are clearly higher than that of 10kN, 0.2Hz tests. The curves of the 12kN, 10kN and 9kN, tends to converge at high ΔK region in the S355J2+N steel. Another important observation is that increasing the load level and decreasing the frequency (e.g. for say J2N-S10F0.2, PS) had a crack growth retarding effect in SW – as compared for example with the J2N-S9F0.3, PS curve. Finally, it can be seen in this figure that P_{max} (in the range of 9kN to 12kN) is found to have a profound influence on CFCGR than the cyclic frequency (in the range of 0.2Hz to 0.4Hz).

7.8.3 Comparison of the corrosion-fatigue crack growth behaviour in normalised and TMCP steels

The results from all the corrosion-fatigue tests performed in this study on S355G10+M, S355G8+M and S355J2+N steels using hold-time waveform cycles are collectively presented in **Figure 7-7** and compared to each other. As seen in this figure, under the same loading conditions (i.e. $P_{max} = 10\text{kN}$ and $f = 0.2\text{Hz}$) the CFCG trend in S355J2+N is generally much higher than those of obtained from S355G10+M and S355G8+M, however the trends cross over each other in the high ΔK region due to the lower slope observed in S355J2+N data set.

Also seen in this figure is that although similar trends have been observed in S355G10+M and S355G8+M tests under similar loading conditions, the CFCGR in S355G8+M could be up to twice higher compared to S355G10+M. This indicates that the generalization of the CFCG behaviour in all the ferritic steels is invalid and each grade of normalised or TMCP steel must be characterized separately. **Figure 7-8** presents the plot of the CFCG data for all the ferritic steels

(including normalised and TMCP steels) under sine waveform only at different frequencies within the range 0.008Hz to 0.5Hz.

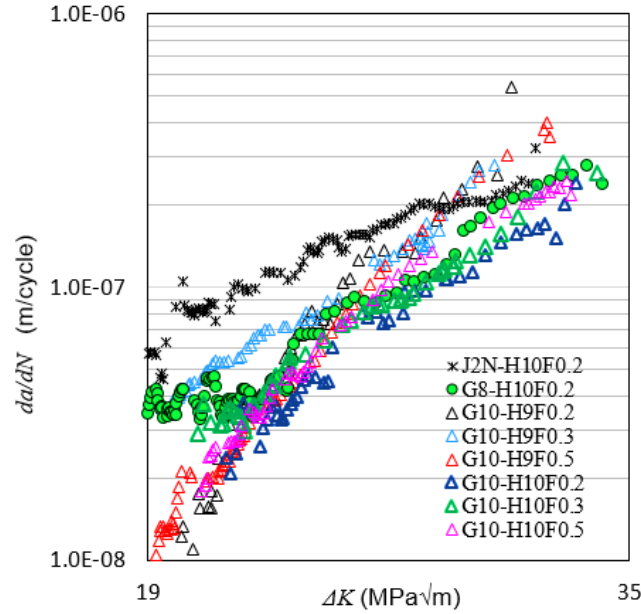


Figure 7-7: Comparison of CFCG results for S355G10+M, S355G8+M and S355J2+N steels under hold-time waveform

The figure includes all the test results obtained from sinewave corrosion-fatigue tests conducted in the present study as well as the literature data on other ferrite-pearlite steels, the details of which can be found in **Table 7-8**. The tests on S355J2+N steel is represented with triangles and coloured yellow and the replotted data of CFCGR of NR steels from the literature is represented with squares and coloured turquoise. The CFCGR data for the TMCP steels is coloured red. The general observation from comparison of such a wide range of ferritic steels tested under different loading conditions suggests that the plots, irrespective of the frequency in the range 0.05Hz – 0.5Hz and stress ratio in the range of 0.0 – 0.67, can be separated into two domains; the upper domain for the normalised steel (NR) and the lower domain for the TMCP steels.

In **Figure 7-8**, bi-linear CFCG trends can be observed for the NR steels which is similar to the 2-stage recommendations made in BS7910 Standard [174]. Two-

stage (A and B) lines – the solid black line is the mean value and the dash black line the mean+2SD curve.

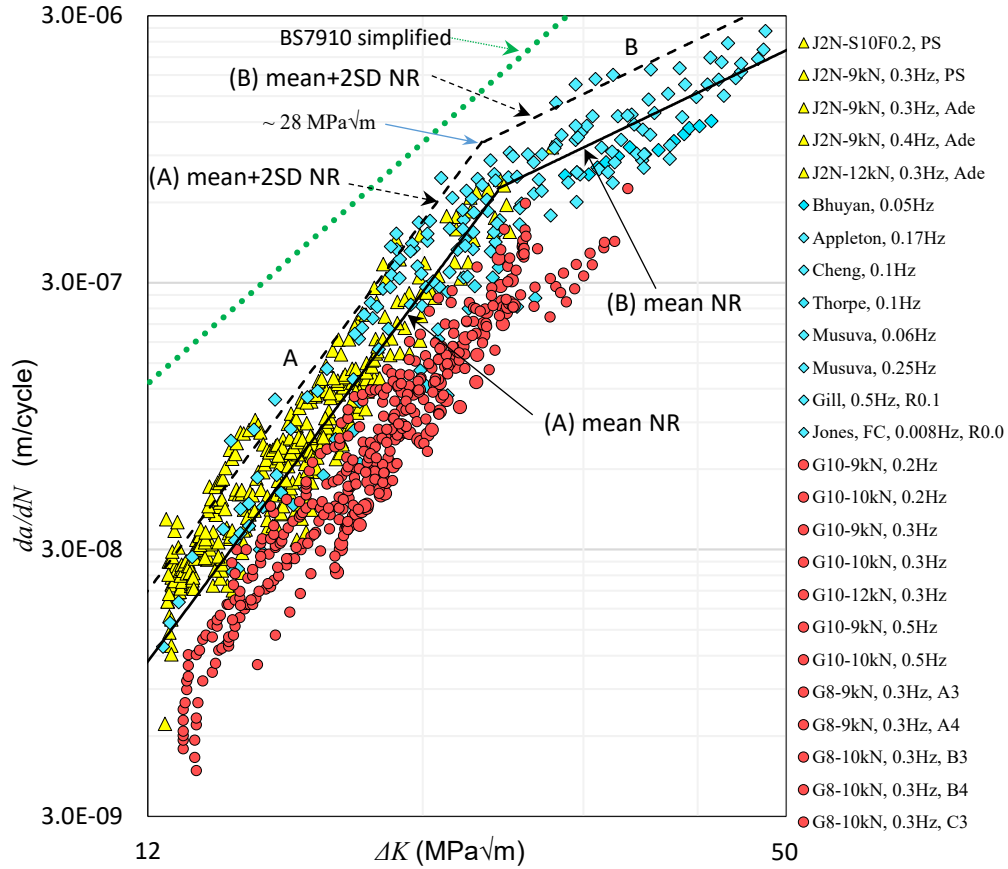


Figure 7-8: Comparison of the CFCG data for all the ferrite-pearlite steels considered in this study under sine waveform

Table 7-8: Loading conditions for the CFCG tests conducted by other researchers

Study by	Steel grade	Wave form	Env.	f (Hz)	R
Bhuyan [152]	CSA G40.21M 350	s	SW	0.05	0.1
		p-s	SW	0.1	0.1
Scott [154]	BS4360-50D	p-s	SW	0.2	0.5
		t	SW	0.1	0.5
		n-s	SW	0.2	0.5
		s	SW	0.17	0.5
Appleton [157]	BS4360-50D	t	SW	0.17	0.5
		s	SW	0.17	0.5
Cheng [159]	ABS grade EH36	s	SW	0.1	0.1
Dhinakaran [192]	C-Mn steel	s	NaCl	1.0	0.5
Thorpe [24]	BS4360-50D	s	SW	0.1	0.0, 0.1
Adedipe [149]	S355J2N	s	SW	0.3, 0.35, 0.4	0.1
Musuva [168]	BS4360-50D	s	3.5%NaCl	0.25	0.08
		s	3.5%NaCl	0.06	0.08
Gill [215]	HSLA (MF-80)	s	3.5%NaCl	0.5	0.67
		s	3.5%NaCl	0.5	0.1
Barsom [161]		s & tri	NaCl	0.1	0.1
Present Study and	S355G8+M	s	SW	0.3	0.1

Mehmanparast [175]					
Present Study	S355G10+M	s	SW	0.2, 0.3, 0.5	0.1
Present Study	S355J2N	s	SW	0.2, 0.3	0.1
Adedipe [149]	S355J2N	s	SW	0.3, 0.35	0.1
s = sine, t = triangular, p-s = positive sawtooth, n-s = negative sawtooth, SW = seawater					

The mean solid line appears to denote the possible boundary (diffused) between the data cloud for the normalised and TMCP steels in the range 13.33 to 50 MPa√m as all the data points of the TMCP are below this line. The Stage A mean curve of the NR steel is described by the Paris equation $1.24 \times 10^{-13} \Delta K^{4.60}$ while that of its B is described by $2.70 \times 10^{-9} \Delta K^{1.64}$. The Stage A mean+2SD curve for the NR steels is described by the Paris law $2.28 \times 10^{-13} \Delta K^{4.60}$ while the Stage B is described by $4.28 \times 10^{-9} \Delta K^{1.67}$. The Paris law constants are also presented in **Table 7-9**. It can be seen in **Figure 7-8** that within the wide range of load levels and frequencies considered in these studies, the CFCGR is generally found to be lowest in S355G10+M TMCP steel, followed very closely by S355G8+M TMCP steel used in the SLIC study [175] and lastly the normalised steels.

Table 7-9: Two-stage Paris law constants for upper bound (UB) of the NR and TMCP respectively in Figure 7-7

Material	C (Stage A)	m (Stage A)	C (Stage B)	m (Stage B)	ΔK transition from A to B
Mean NR	1.24×10^{-13}	4.60	2.70×10^{-9}	1.64	~29.0
Mean+2SD - NR	2.28×10^{-13}	4.60	4.28×10^{-9}	1.64	~28.0

The results presented in **Figure 7-8** suggest that the level of scatter in normalised steels can be higher than TMCP steels, though more tests on a wider range of subgrades of TMCP steels must be performed to confirm this observation. This scatter appears to be a function of the morphologies of the microconstituents of the steel which influenced the extent of the crack bifurcation, diversion and crumb formation found during metallographic study of the crack path (see section 7.9). The green dotted line is the simplified BS7910 recommendation curve which still provides a satisfactory design estimate of the *FCG* behaviour for these grades of steel in *SW*. But clearly it is over-conservative for the TMCP steels. Finally, seen in this figure is that the difference between the FCGRs at the lower bound and upper bound data can be up to around an order of magnitude. This suggests that the corrosion-fatigue behaviour of each subgrade of normalised and TMCP steels

must be carefully characterized for a given loading condition and the CFCG behaviour in the entire range of ferrite-pearlite steels cannot be defined using a single trend.

7.9 Metallography and Fractography Analyses

7.9.1 Analysis of untested samples

To understand the disparity in the CFCGR between the waveforms under the test conditions discussed above, a metallographic study was carried out. First step was to determine the phases present in these steel subgrades. This involved the use of optical and scanning electron microscopy (SEM) techniques. The microstructures of the three S355 steel subgrades examined in this study are shown in **Figure 7-9**.

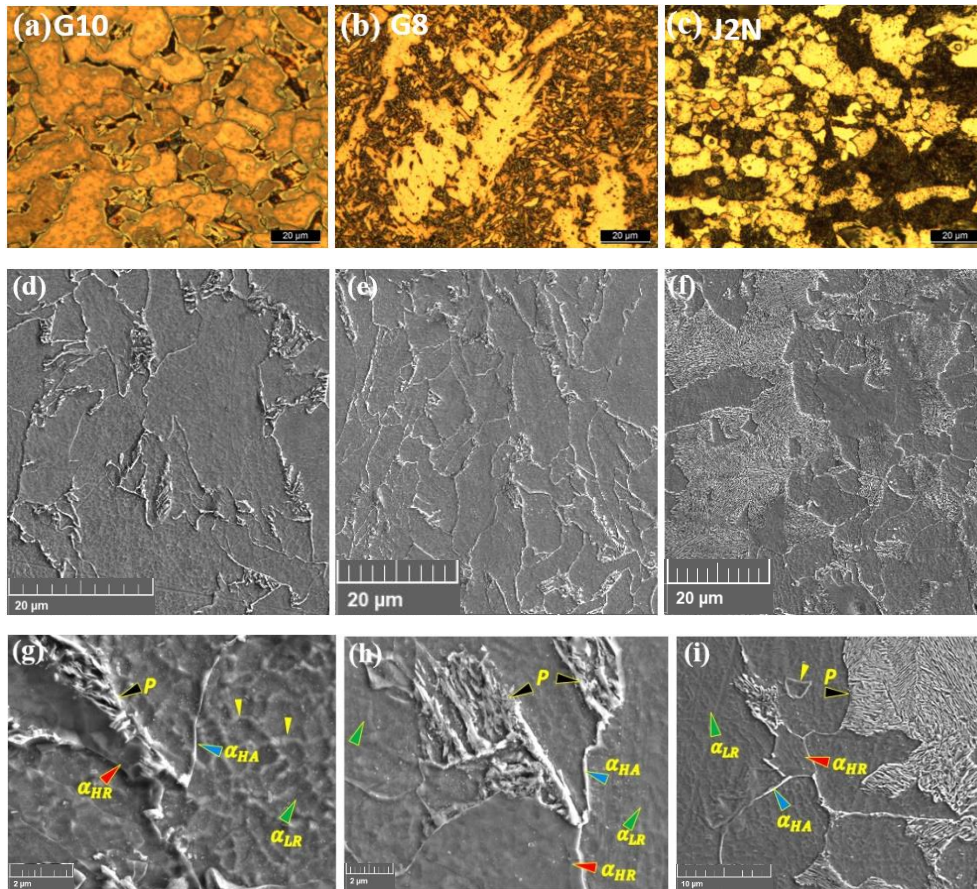


Figure 7-9: The optical and SEM micrographs of the S355 steel samples examined in this study (a, d, g) S355G10+M (b, e, h) S355G8+M (c, f, i) S355J2+N

In **Figure 7-9 (a, b and c)** the optical micrographs with magnification scale of 20 μ m are presented while **Figure 7-9 (d, e and f)** are the corresponding SEM micrographs obtained at 50 μ m. **Figure 7-9 (g and h)** were obtained at 2 μ m while **Figure 7-9 (i)** was at 10 μ m. As seen in **Figure 7-9**, the S355 steel subgrades examined in this study have microstructures consisting essentially of ferrite (α) matrix with varying volume fraction of pearlite (P). The microstructures of S355G10+M and S355G8+M are almost similar, but different from S355J2+N. Three types of ferrite were observed with the aid of EDX compositional analysis as noted in **Figure 7-9 (g, h and i)**. For clarity, these will be referred to as High relief ferrite (α_{HR}) indicated by the red arrows, low relief ferrite (α_{LR}) indicated by the green arrows and high alloy ferrite (α_{HA}) noted by blue arrows. Note that the difference in carbon content between the α_{HR} and α_{LR} is relatively small, if not negligible. The high alloy ferrite (α_{HA}) phase formed from the austenite as a thin ribbon completely covering the α_{LR} grain surfaces. From EDX analysis, this thin ferrite ribbon actually has carbon content approaching that of pearlite with segregation of alloying elements such as Al, Mg, Si and ‘impurities’ such as O, S and P. The detailed analysis of the alloying constituents of this phase is ongoing. The relative abundance of the major alloying elements in the phases is given in wt%. The EDX composition simply shows the relative abundance of the alloying elements in the phase and has no direct relationship with the steel overall composition shown in **Table 7-2**. Note that the low and high relief α phases could not be distinguished under optical micrographs.

The pearlite in **Figure 7-9** is indicated by black arrows. Another observation is the formation of sub-grains in the α_{HC} ferrite phase as shown by the smaller yellow arrows in **Figure 7-9 (g and i)**. Note that the parent ferrite grain (shown by the green arrow) is found to contain relatively higher carbon content. The formation of these sub-grains may be due to heating and re-heating processes inherent in TMCP steel production technique [232]. With prolonged heating carbon and other

alloying elements may have diffused further, producing these sub-grains which appear to be surrounded by the α_{HA} phase along the sub-grain surfaces. The pearlite colonies in the TMCP steels are somewhat slender and elongated. In the TMCP steels, the *Pearlite* appears to grow in two forms – small-blocky colonies and the thin, elongated or needle-like morphology colony that perhaps developed from the α_{HA} phase. In the S355J2+N steel the *Pearlite* colonies are large, blocky and dense as can be seen in **Figure 7-9 (f and i)**. Clearly, the phase morphologies in the S355J2+N and S355G8+M/G10+M TMCP steels are different. These different phase morphologies are expected to create dissimilar local microstructural properties and corrosion process variation. The presence of a hard phase (*P*) in a soft matrix (α) with dissimilar stiffnesses is expected to influence local fatigue behaviour in the steel [124] and may influence the waveform effects.

7.9.2 Post-mortem analysis of corrosion-fatigue tested specimens

Some examples of the post-mortem analysis on corrosion-fatigue tested samples for S355G10+M and S355J2+N are shown in **Figure 7-10**.

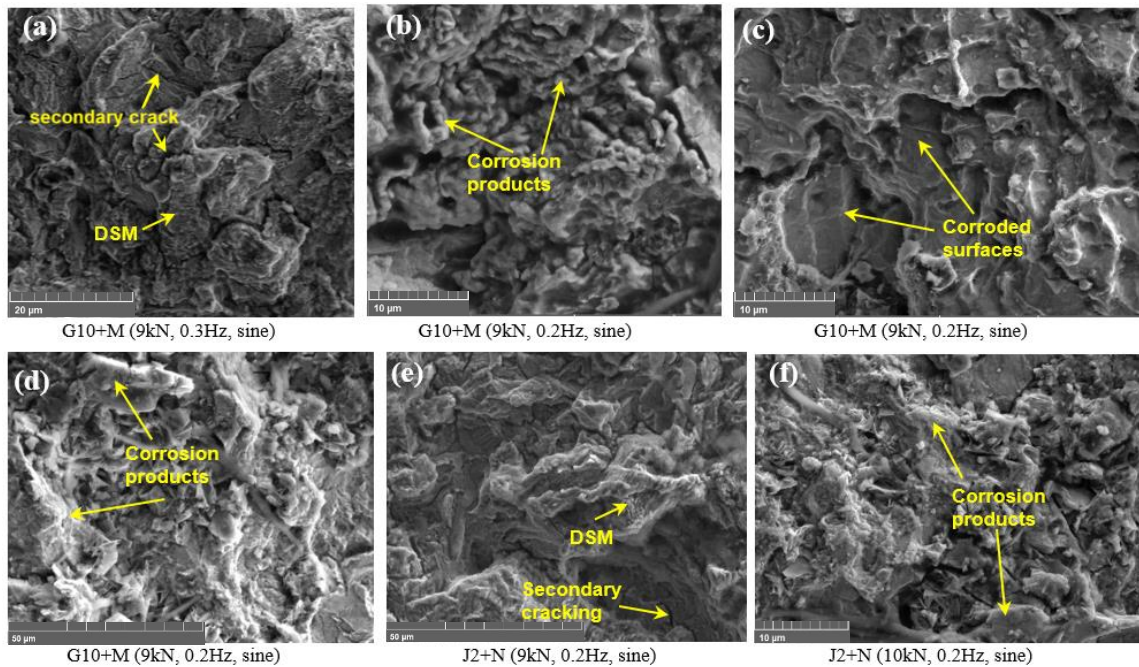


Figure 7-10: Fractographs of fracture surfaces of corrosion-fatigue testes specimens (a to d) for S355G10+M, (e & f) for S355J2+N

Post-failure examination of the fracture surface of corrosion-fatigue specimens shows that the striations in the samples tested in *SW* appeared to be washed out by the corrosion process. This washing out process progressed with time for the S355G10+M steel as can be seen in **Figure 7-10 (a) to (d)**. **Figure 7-10 (e and f)** show the fracture surface of S355J2+N samples in *SW* and the characteristics are similar to that observed in the TMCP steels (i.e. S355G10+M). Generally, cleavage failure was not observed in *SW* tests. The ductile striation mechanism (DSM) as seen in the fractograph (see **Figure 7-10 (a and e)**) is typical of the failure mechanism towards the end of the Paris Region for low carbon steel. These fractographs do not give any clue for the difference in the two waveforms under the test conditions discussed previously. An alternative approach is to examine the crack path and the features influencing crack growth in the materials.

7.9.3 Post-mortem analysis of corrosion-fatigue tested specimens

The crack path study was carried out on G10-S10F0.5 specimen, which was tested using sinewave, and the results are presented in **Figure 7-11**.

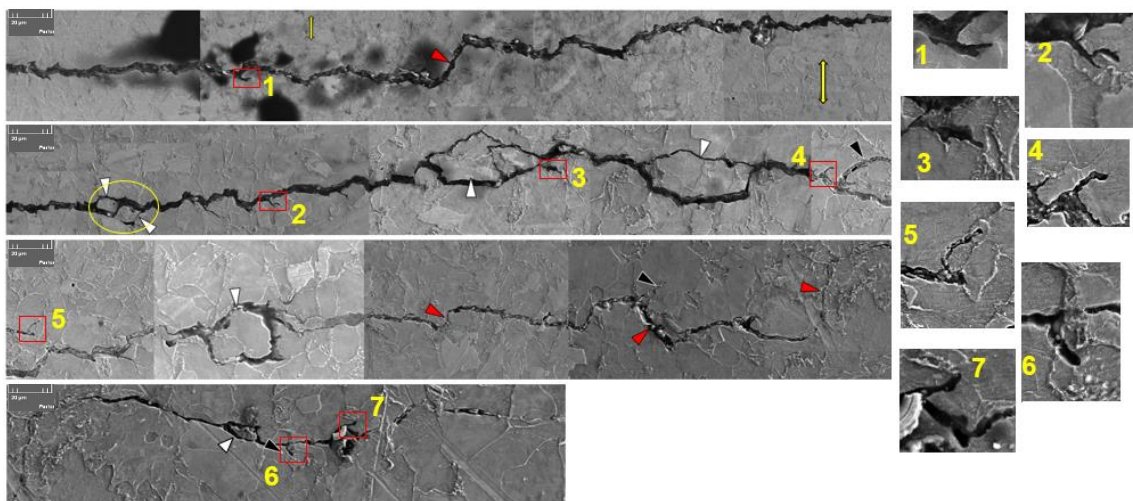


Figure 7-11: Crack path analysis on G10-S10F0.5 (scale: 20 μ m)

This test was performed on a S355G10+M specimen with $P_{max} = 10\text{kN}$ and $f = 0.5\text{Hz}$. The direction of the crack growth in this study was from left to right and the direction of the applied fatigue force was normal to the crack path (as shown by a yellow double-sided arrow). High angle diversion (as noted by the red arrows), branched crack front (indicated by black arrow) and formation of metal crumbs (indicated by white arrows) were found in **Figure 7-11**. Note that there were very few high angle crack tip diversions and branching compared to the observations made in the air tests on the same material. There were also few formations of metal crumbs in the range of approximately $18 - 22 \text{ MPa}\sqrt{\text{m}}$ SIFR. There was also the widening of the gap between the fatigued surfaces in comparison with air tests (not shown here) by the corrosion process and the branched crack tips maintained some sharpness. The magnified micrographs on the right of **Figure 7-11** show branched crack tips with some degree of tip blunting.

There is considerable crack blunting at number 5, 6 and 7 identified features which occurred at high ΔK . Number 2 and 5 features show severe blunting to the extent that another crack front budded or nucleated from its side. This suggests that increasing the fatigue load to a level where the crack opening displacement is significant in *SW* can result in some degree of crack tip blunting which may retard the crack growth.

The crack path analysis was also carried out on G10-H10F0.2 specimen, which was tested using hold-time waveform, and the results are presented in **Figure 7-12**.

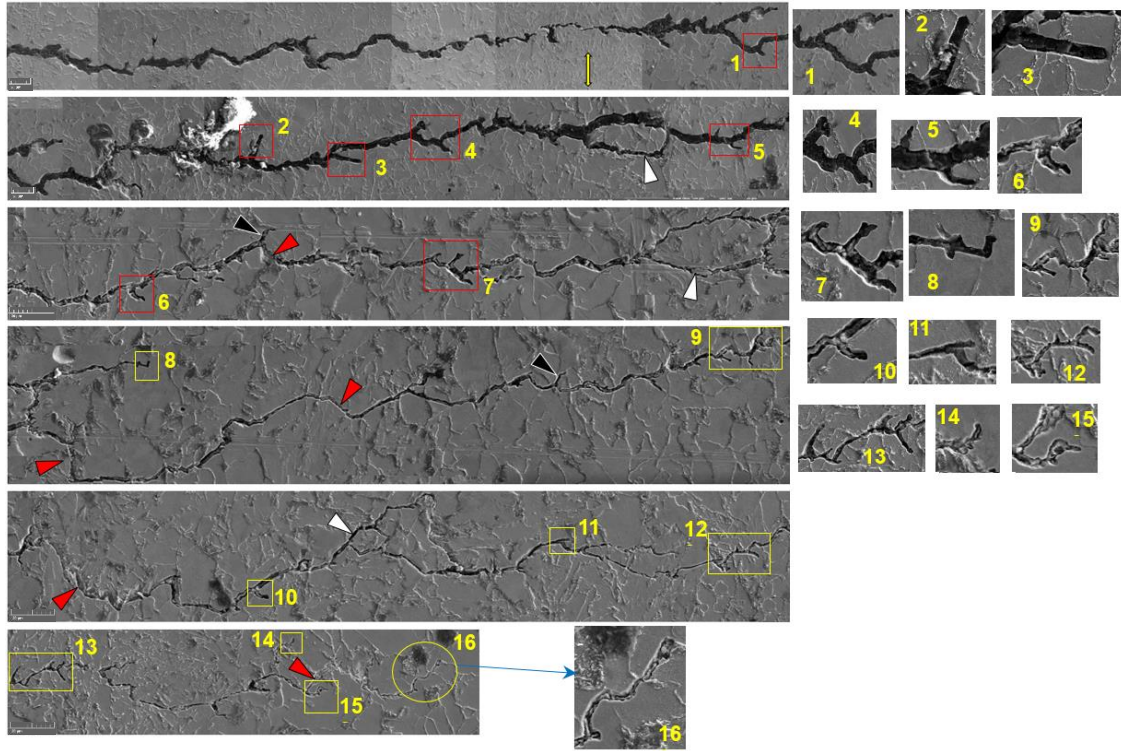


Figure 7-12: Crack path analysis on G10-H10F0.2 (scale: 20 μ m)

This test was performed on a S355G10+M specimen with $P_{max} = 10\text{kN}$ and $f = 0.2\text{Hz}$. The direction of the crack growth in this study was from left to right and the direction of the applied fatigue force was normal to the crack path. Initially, a low angle crack diversion and branching similar to that seen in **Figure 7-11** was observed. Similar features of high angle diversion (as noted by the red arrows), branched crack front (indicated by black arrow) and formation of metal crumbs (indicated by white arrows) were also found on this specimen. The basic difference between this specimen, which was tested with hold-time waveform, with the test conducted under sinewave is that the branched crack tips are extensively blunted as shown on the top right of **Figure 7-11** for sinewave and **Figure 7-12** for hold-time. Number 1 to 7 features clearly show the severe blunting that occurred at lower ΔK under 10kN in the low frequency test ($f = 0.2\text{Hz}$) performed with hold-time. The yellow boxes (features 8 to 15) show that the severe blunting process occurred also at high ΔK in this specimen. The point of test stoppage is encircled (feature number 16) and it can be seen that the tip of

the main propagating crack front is considerably blunted as shown by the magnified micrograph on the bottom right. This clearly explains why lower CFCGR trends were observed in the samples tested using hold-time waveform compared to sinewave (see **Figure 7-2**).

7.10 Summary of conclusions from previous work

The effect of waveform in steel has been reported in several studies. Atkinson, et al. [150] in his study found that holdtime did not increase the CFCGR at frequencies as low as 0.01Hz after holding up to 10 minutes. Nakasa, et al. [151] suggested that stress rise time and the interaction between the crack tip and hydrogen atoms controlled the crack growth mechanism in corrosion-fatigue. Scott, et al. [154] found that the CFCGR of square waveform was the least among sinusoidal, triangular and positive sawtooth waveforms. Thorpe, et al. [24] got the same result as Scott, et al. [154], and added that CFCGR increased as test frequency is decreased, and R increased. Barsom J. M. [161] reported that the CFCGR of square waveform is lower than that of sine, triangular and positive-sawtooth. He informed that corrosive medium enhances FCG when the tensile stress around the crack tip is increasing and not during holding at maximum stress or during decreasing stress in the cycle period. Knop, M., et al. [163] in their study informed that in the Paris Region, FCG is by plastic blunting and that the crack grows during the rising of the load towards the cycle peak for each load cycle. They also said that less crack tip blunting occurs in an embrittling environment due to localisation of deformation slip and this enhances the rate at which the fatigue crack grows. They attributed higher CFCGR to concentration of adsorbed hydrogen at the crack tip and at the tips of nanovoids ahead of the advancing crack front. Knop, M., et al. [163] also informed that crack growth is enhanced during rise time and that holding at maximum load and longer fall times can enhance crack growth when the rise time is short. However, their study was mainly on cathodically protected steel. Austen, I. M. [164] reported that in distilled water the CFCGR of square waveform was the lowest to sawtooth and triangular waveforms for $R=0.25$ and 5Hz. The CFCGR of the square waveform was also lowest to sinewave and triangular for 300m steel in distilled water at frequency of

0.45Hz and $R = 0.5$. Kawai, S., et al [165] in their work reported that FCG was enhanced during stress rise time in 3%NaCl. They attributed the high CFCGR to dissolution of newly exposed surfaces formed during the rise time by the corrosion process. In their study also, CFCGR was decreased by holding at maximum load. Wand, Z. F. [166] in his study of fatigue in iron said that under 0.1Hz, $R = 0$ and in 3.5%NaCl, the CFCGR in the iron is lowest under square waveform in comparison with sinewave and positive sawtooth. They also reported that straining mode at the crack tip or the difference in the geometry of the crack tip was responsible for the difference in CFCGRs of the waveform and not because of anodic dissolutions at the crack tip. They equally found that for the square waveform, holding for 0.1 s and 0.9 s at maximum load had no significant effect on the CFCGR. Consequently, they attributed the waveform crack growth rate difference to the continuous deformation of the crack tip material. BS EN ISO 11782 [177] informed that waveform determines the extent of rupturing of the film covering the corroding surface at the crack tip and this process may determine the rate of crack growth in corrosive environment. It also associated reduction of CFCGR under holdtime to crack tip blunting due to corrosion of the plastically deformed region around the crack tip.

In all the studies above, the difference in CFCGR of iron and steel under different waveforms had been attributed generally to increasing tensile stress at the crack tip during loading - necessary to dislodge the protective film, dissolution of newly exposed surfaces at the crack tip and the activity of hydrogen at the crack tip region and ahead of the advancing crack. The retardation of crack growth with square waveform has been attributed to crack tip blunting due to the dissolution reaction. These studies did not present the comparison of waveform effect under the same frequency and constant load ratio for steel subgrades, with a view of determining the effect of variation of microstructure within the same class of steel. There was also very limited micrographs showing clearly the mode of crack growth under different waveforms. Such study is expected to show clearly the damage mechanism and specific roles of microstructural phases under these waveforms in the Paris region. Within the knowledge of the authors, the influence

of waveform order than sinusoidal on CFCGR for modern S355 steel subgrades used in the fabrication of modern offshore structures, including WT monopile foundations is non-existent.

7.11 Conclusions

This paper presents the experimental investigation of CFCGR sensitivity to waveform, frequency and load level in *SW* tests on S355 steel subgrades; S355G8+M/S355G10+M TMCP steels and S355J2+N normalised-rolled steel. The waveforms considered in this study are sinewave and trapezoid waveform (generally referred to here as hold-time) and the experiments were conducted under frequencies of 0.2Hz, 0.3Hz, 0.5Hz. The stress ratio in all tests was 0.1 and the maximum applied loads were 9kN and 10kN. The results on S355G10+M have shown that the CFCGRs are higher in sinewave tests compared to the hold-time waveform and this effect appears to be more pronounced at lower frequencies (i.e. $f = 0.2\text{Hz}$) and higher applied loads (i.e. $P_{max} = 10\text{kN}$). Moreover, it has been shown that in the tests performed using hold-time waveform, the CFCGR in the Paris region increases slightly with decrease in the load level, while no noticeable difference was observed in sinewave tests with different load levels and frequencies. The results from additional tests on S355G8+M and S355J2+N have shown that these two subgrades of S355 are also sensitive to the cyclic waveform and up to three times lower CFCG trends can be obtained when the tests are performed with hold-time waveform. Comparison of the experimental results on S355G8+M, S355G10+M and S355J2+N has shown that under both sinewave and hold-time waveforms, the CFCG trend in normalised steels (e.g. S355J2+N) is higher than TMCP steels (e.g. S355G8+M, S355G10+M). Post-mortem analysis on the fractured surface of corrosion-fatigue specimens has shown that the CFCG mechanism is by ductile striations which appeared to be washed out by the corrosion process. Also observed in the metallurgical analysis of the crack paths is that in the samples tested under the same load level (i.e. $P_{max} = 10\text{kN}$), the specimen tested with hold-time waveform has exhibited extensively blunted branched and main crack tips which explains why the CFCG

trends are lower in hold-time waveform tests compared to sinewave. It is found that the main crack tip blunting process is the primary factor controlling the CFCGR of steel at high ΔK and low frequency in *SW*. Other fundamental factors are crack angle diversion, branching of crack front and formation of metal crumbs along the crack path. The results from this study are expected to improve the understanding of the CFCG behaviour in offshore wind turbine monopile foundations and enhance the structural integrity assessment of these assets by considering the influence of realistic loading conditions on the CFCG behaviour of the material.

8 The effect of microstructure on the CFCGR in ferrite-pearlite steels in the Paris Region

8.1 Introduction

Understanding how a crack propagates in a given material is fundamental to all forms of theoretical postulation, modelling, analytical and numerical analysis. If the premise of crack growth path is wrong, no matter how elegant the mathematical or numerical expression may appear, the result will not represent reality.

Although experimental investigations have been previously conducted by researchers to characterise the CFCG behaviour of various materials used in offshore structures [24][151][154][156][159][175][226][227][228][229], studies on the specific role of microstructural phases in retarding or accelerating FCGR in the Paris Region is non-existent for modern S355 ferrite-pearlite steel subgrades. Many of the corrosion-fatigue data of steels existing today are many decades old. Majority of the tests were performed on normalized, quenched and tempered steels and they form the basis for explaining corrosion-fatigue mechanism in ferrite-pearlite steels. Today, new sets of modern normalised-rolled (NR) and thermomechanical control process (TMCP) S355 ferritic steels are increasingly being used in the design of offshore support structures, especially WTs. The corrosion-fatigue crack growth data and the mechanism of CFCG for these new materials are very scarce, if not non-existent.

The knowledge of corrosion-fatigue and its mechanism is of immense value in fatigue study. Understanding the influence of microstructure on fatigue mechanism is fundamental because it supports: realistic fatigue modelling, fatigue life prediction of structures and design of fatigue resistant materials. Fatigue is a complex problem and there have been several publications on the problem of fatigue, each presenting similar or different ideas or trying to modify existing theories – using from simple to complex analytical and numerical

approaches. Generally, *FCG* in metallic materials is divided into three regions. Each region has been reported to exhibit different mechanisms and characteristics. These regions are the threshold region (or early stages of fatigue crack development), the Paris Region (or the linear and steady crack growth stage) and region of final failure (or unstable/accelerated crack growth stage). The Paris Region is of interest in this study because it is the part commonly used and recommended in ASTM E647-15 [172] and BS EN ISO 11782-2:2008 [177] for engineering design.

In many reports [29][187][197][199][200][201], microstructure, mean stress, mechanical properties and initial a were said have large influence on the threshold and final failure regions. In addition, the accelerated region is said to be influenced by sample thickness. These factors were reported to have little or negligible effects when the propagating crack has grown to a considerable length or beyond few grains, usually in the Paris Region [29][188][193][194][195][196][233]. In the threshold region, non-continuum or single shear mechanism was said to operate and the nature of the fractured surface was seen to be faceted. The crack tip was under both tensile and shear forces. The crack closure phenomenon was reported to be high and the plastic zone size is equal or less than the microstructural grain diameter. In the final or accelerated failure region, fatigue and additional static loading modes were said to operate. Microvoid coalescence, intergranular or additional cleavage failure mechanism have been found on the fractured surface and the crack tip was under tensile loading. The plastic zone size in this final region was reported to be very much bigger than the grain diameter.

Many researchers have reported that fatigue failure in the Paris Region is generally by transgranular ductile striation mechanism [233][234][235]. The crack path is generally taken to be across the grains (or transcrystalline), although they may also propagate along the grain boundaries or intergranular depending on the material properties, loading and environmental conditions [202]. In this linear crack growth region, the crack tip is under tensile loading and the plastic zone

size is reported to be greater than a grain diameter. The crack closure is low and mechanism of growth is by striation – alternating or simultaneous shear on two slip systems [235]. The conclusion [29][187][197][198] that microstructure has little or no influence on the FCGR of metallic materials is often based on the frequent observation of ductile striation mechanism in the Paris Region in fractographs. One thing that must be noted is that most of the fatigue theories existing today – including that of crack extension [235], crack path [234] and crack closure [236][237] were propounded from experiments performed on non-ferrous materials. For example, two other phenomena which are associated with crack retardation across metallic materials in the literature are crack closure and interlocking. It is pertinent to note that crack closure effect may be more important for non-ferrous ductile metals. The concept of crack closure and the use of effective ΔK was proposed by Elber [236][237]. He argued that there is a premature contact of the crack faces during unloading from tension in a fatigue test. This effect reduces the effective stress at the crack tip. Elber proposed the use of effective ΔK instead of the conventional ΔK in plotting fatigue curve. This theory is based on the fact that plastically deformed surface wake is left behind as the crack propagates. The mechanisms such as the plasticity-induced and the roughness-induced crack closure can cause retardation of the FCGR. However, it is pertinent to note that Elber's theory is based on his study of 2024-T3 aerospace aluminium alloy of a moderate ductility. The validity of this theory to steel will vary since; (a) steel can have high ductility in annealed or fine-grained condition to 'near-brittle' ductility in martensitic or very low temperature condition, and (b) the thickness effect in steel material. The existence of plan stress condition at the outer surface and plain strain condition in the interior for relatively thick steel sample may further weaken the theory of crack closure for thick steel samples. As sample thickness of steel is increased, the plain stress region tends to be eliminated leaving only plain strain condition where plastic zone becomes little or almost negligible. Hence, the concept of crack closure cannot be rigorously applied to steel at all conditions, coupled with fatigue test condition under plain strain condition – i.e. with little or negligible plastic zone at the crack tip. In other words, when the plastic zone is little or the material is in corrosive

medium, the effect of crack closure can then be ignored, or its consequences will be insignificant.

None of these studies in practical terms showed vividly how the crack propagated through the phases in the microstructure of the materials studied. The validity of these theories to steel generally is in doubt since steel property is strongly influenced by its microstructure. There are many variables, e.g. alloying elements and their concentrations, forming process and the mode of deformation, heat treatment - involving temperature range, heating time, cooling time, etc. which can be combined in so many ways to obtain varieties of steel properties, e.g., ranging from high ductility to almost completely brittle steel. The focal point of this section therefore, is primarily to disprove this age long conclusion of microstructural effect on Paris Region as it relates to ferrite-pearlite (α - P) steels by presenting factors responsible for such deviation. It also evaluated the resistance of the modern ferrite-pearlite steel produced by Normalized-rolled (NR) and Thermo-mechanical control process (TMCP) to fatigue crack growth (FCG) in air and *SW* and assesses the contribution of microstructural phases to the fatigue phenomenon.

8.2 Metallurgy and characterisation of phase constituents of experimental steels

The increasing demand for economic steels with higher toughness, better weldability and less expensive alloying elements has resulted to modern thermomechanical control process (TMCP) steels that are increasingly being used in the design of offshore support structures. Some of the advance structural steels are shown in **Table 8-1 (a)** and conventional steels in **Table 8-1 (b)**. The TMCP steel grades have been manufactured specifically for offshore pipelines, platforms, pressure vessels, and modern WT installations [42] and currently replacing the more traditional ones in **Table 8-1 (b)**. The BS4360 Grade 50D in **Table 8-1 (b)** is equivalent to S355J2+N hot rolled normalized steel.

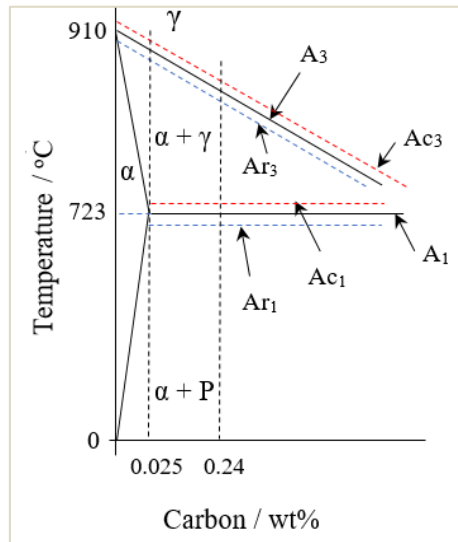
Table 8-1: (a) Advanced S355 steel subgrades for marine applications [144] (b) Conventional S355 structural steel grades [33][34][39][40]

Offshore Structural Steel Plates			
(a)		(b)	
European		European	British
EN10225	Typical offshore application	EN10025 Part 2: 2004	BS4360:1990
S355G7+N	Primary Structure	S355	50A
S355G7+M	Primary Structure	S355JR	50B
S355G8+N	Critical Joints	S355JO	50C
S355G8+M	Critical Joints	S355J2	50D
S355G9+N	Primary Structure	S355J2+N	50D
S355G9+M	Primary Structure	S355K2	50DD
S355G10+N	Critical Joints	S355K2+N	50DD
S355G10+M	Critical Joints	S355NL	50EE

Figure 8-1(a) is a schematic of common Fe-C phase diagram showing important regions in the heat treatment of marine steels. γ denotes austenite, α ferrite and P pearlite. A_1 is an equilibrium lower transformation temperature. Eutectoid reaction occurs at A_1 which is 723°C in the binary diagram. The Ar_1 temperature depicts the lower transformation temperature upon cooling. Ac_1 temperature depicts the start point of the transformation between the α -ferrite and the austenite γ upon heating. A_3 is an equilibrium upper transformation temperature where alpha iron ($\alpha\text{-Fe}$) transforms to gamma iron ($\gamma\text{-Fe}$) and it is 910°C for pure iron and it is lowered with addition of carbon. Ar_3 is the temperature at which austenite γ starts transforming to ferrite α during cooling. Ac_3 temperature depicts the upper transformation temperatures upon heating. Above Ac_3 , the austenite strength is lower, plastic and malleable [238]. The line in **Figure 8-1 (a)** are straight - in fact in the case above they are drawn arbitrarily. In reality, the shape and position of the lines – straight or curved, depend on the steel chemistry, cooling rate and rolling or deformation variables.

Austenitization is the first operation in many of the most important deformation and heat treatment processes. After austenitization, upon cooling the austenite decomposes into various morphologies. From the austenite region, the bcc α -

ferrite is the first to nucleate at the austenite grain surfaces. Other microstructures that form as the transformation temperature is lowered are pearlite (P), bainite, and martensite.



(a)

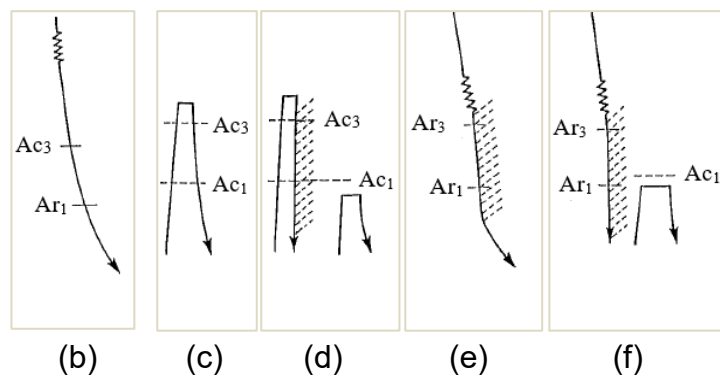


Figure 8-1: Heat treatments of modern steel. (a) a schematic phase diagram of carbon composition range and phases expected in the microstructure of marine steel, (b & c) rolling and normalization (d) TMCP route involving reheating, accelerated cooling and self-temper or tempering at temperature below A_{c1} (e) rolling, accelerated cooling and (f) rolling, direct-quenching and tempering at temperature below A_{c1} [238][232].

The nature of ferrite-pearlite microstructure due to heat treatment can be explained by the type of heat treatment process – commonly referred to as as-rolled (AR), normalized (N), normalized-rolled (NR), quenched and tempered

(Q&T) and thermomechanical control process (TMCP). The common phases in these steels are ferrite and pearlite ($\alpha + \text{Fe}_3\text{C}$).

The heat treatments of interest in this study are AR, N, NR and TMCP. *Normalizing* involves cooling of austenitized steel, usually in air, to a temperature substantially below the transformation range [239]. The austenitizing temperature for normalizing is typically 30–80°C above the A_{c3} transformation temperature (the upper critical temperature, 800 – 925) for hypoeutectoid steels. As-rolled is obtained by austenitizing at say about 1250 °C, holding for specified time and then deforming at roughing (R) and finishing (F) mills within the austenite range. The finishing rolling may, for example, be at some temperature above A_{c3} and the steel is allowed to cool naturally to room temperature as shown in **Figure 8-1 (b)**. Normalized steel (N-steel) is obtained when the as-rolled steel is reheated to about 900°C or very little above A_{c3} , held for a desired period and allowed to cool naturally back to room temperature as shown in **Figure 8-1 (c)**. When the finishing rolling operation is done at a temperature above about 900°C before cooling naturally to room temperature, a NR-steel is obtained. Here no reheating is necessary. Normalized and NR steels have practically the same properties and so are designated 'N'. The strength or other properties of the steel can be increased by addition of small amount of alloying elements and then quenched and tempered as shown in **Figure 8-1 (d)**. The tempering is then done below A_{c1} .

It has been known for over 50 years now that when deformation is carried out at lower temperature, the rate of recrystallization becomes slower and the finer is the recrystallized grains [240]. Hence, to obtain the desired fine grain size, the final hot-working (finishing) is done by lowering the temperature. In fact, the rolling can be extended into the $\alpha + \gamma$ phase region. Thermo-mechanical controlled process (TMCP) is a microstructural control technique combining controlled rolling (CR) and accelerated cooling to obtain exceptional strength-toughness combinations in low-alloy steels by grain refinement [232][240][241][242]. In TMCP, the nucleation sites for ferrite phase during cooling is increased by CR, where the austenite grain is refined and strained. After this, the strained austenite

is rapidly cooled to lower temperature where diffusion of atoms is slower while the driving force for transformation is high. This results to production of fine-grained microstructure that confers high mechanical properties and better toughness. In general, controlled-rolling, controlled-cooling, accelerated cooling and direct-quenching are typical examples of thermomechanical processing [232]. Controlled rolling involves carrying out of the finishing deformation at temperature sufficiently low to prevent grain growth of the recrystallized austenite during cooling. The property obtained may not be different from those obtained after normalization [232] [238]. The TMCP route is illustrated in **Figure 8-1 (e and f)** [238]. Traditional marine steels shown in **Table 8-1 (b)** are produced by AR, N or NR. The TMCP steels production route is proprietary and so processing conditions and number of deformations are slightly different, producing what may appear as variation in the ferrite and pearlite sizes and morphology.

Ideally, most common marine steels have microstructures consisting of very small volume fraction of pearlite in a ferrite matrix. It may be called ferrite-pearlite steels where volume fraction of pearlite phase is significant. The morphologies of ferrite are often referred to as grain boundary allotriomorphic ferrite, idiomorphic ferrite, Widmanstätten ferrite, and intragranular ferrite or acicular ferrite [208]. The common ferrite morphologies are grain boundary allotriomorphic ferrite, idiomorphic ferrite, Widmanstätten ferrite, and intragranular ferrite or acicular ferrite [208]. The ferrite morphology commonly encountered or that dominates in marine steels is the allotriomorphic and idiomorphic ferrites. In this study they will collectively be referred to as ferrite, α . The microstructure shown in **Figure 8-2 (a)** is air-cooled of medium carbon steel and the relatively rapid method of cooling, limited the ferrite grain growth. The ferrite then appears to form layers which followed the austenite grain boundary contours. The prior austenite grain boundaries are completely covered by the ferrite allotriomorphs and the residual austenite has transformed into pearlite upon reaching the temperature, 723 °C, of eutectoid reaction. The ferrite phase is soft and ductile, while the pearlite is hard and brittle. Pearlite grows in colonies and each colony consists of a thin lamellar, alternating mixture of ferrite, α (iron) and cementite, θ (iron carbide) or

($\alpha + \text{Fe}_3\text{C}$). **Figure 8-2 (b)** shows a situation where the Fe - 0.4C steel is slowly cooled to room temperature, producing equiaxed grains. Several distinct pearlite colonies can be seen (dark etching).

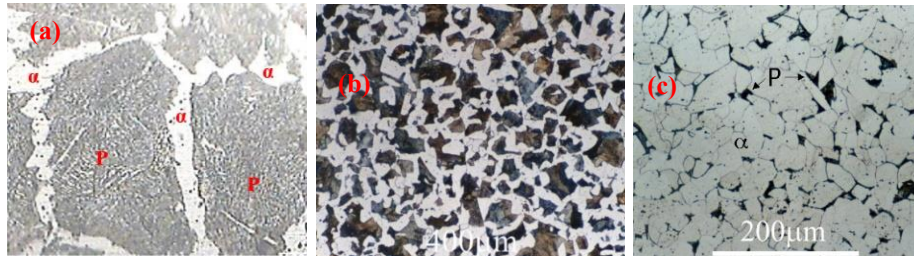


Figure 8-2: Ferrite and pearlite morphologies: (a) ferrite in Fe-0.46 steel that forms layer traced the austenite grain boundary contours, [209] (b) equiaxed ferrite in a Fe-0.4C steel [208], (c) ferrite in low carbon steel [208] (α = ferrite, P = pearlite). The white etching is ferrite while the dark-etching is pearlite.

The ferrite phase (light etching) is soft and ductile, while the pearlite is hard and brittle. As the volume fraction of pearlite increases the bulk strength increases and ductility decreases. If the carbon content of the steel is decreased, the volume fraction of pearlite in the microstructure will decrease while that of ferrite increases as can be seen in **Figure 8-2 (c)**. The morphology or shape of the ferrite-pearlite phases depends on the processing route discussed above and these two phases are of major interest in this study. Any other phase that might be present for the purposes of clarity is neglected.

The S355 steel subgrades for this study are S355J2+N (J2+N), S355G10+M (G10+M) and S355G8+M (G8+M) with compositions are given in **Table 8-2**.

Table 8-2: Chemical compositions and mechanical properties of test materials (see [230])

Steel Grade	Notation	C	Mn	Ni	Si	Cu	Cr	Mo	CE	E %	YS (MPa)	UTS (MPa)
S355G10+M (TMCP)	G10	0.06	1.57	0.33	0.27	0.24	0.03	0.01	0.37	38	435	545
S355G8+M (TMCP)	G8	0.05	1.52	0.32	0.27	0.24	0.03	0.01	0.35	35	447	549
S355J2+N (NR)	J2N	0.17	1.54	0.04	0.04	0.08	0.02	0.10	0.57	20	385	531

The composition of J2+N was extracted from [171]. They are marine steels having low carbon contents with the microstructures consisting of very small volume fraction of pearlite (P) in a ferrite (α) matrix [230]. **Figure 8-3** shows the phase morphologies of the microstructure of a NR – steel S355J2+N obtained for the S355 steel subgrades studied in this paper.

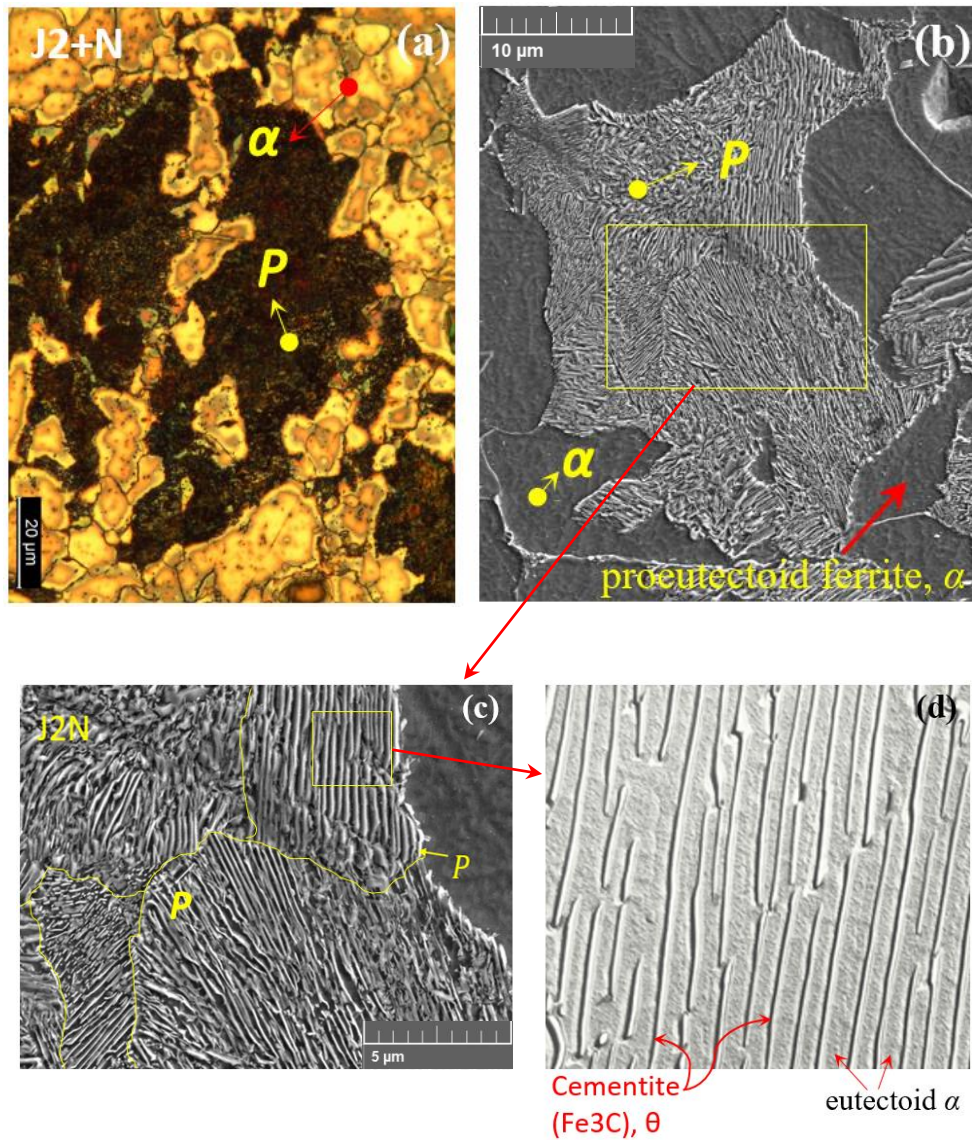


Figure 8-3: Photomicrographs of ferrite-pearlite steels - NR - S355J2+N

In **Figure 8-3 (a)**, the optical micrograph shows that the microstructure is composed of pearlite (dark etching) and the rest ferrite (light etching). Microstructural examination of S355J2+N shows that, 80% of the microstructure

is allotriomorphic ferrite and others pearlite, with traces of acicular or Widmanstätten ferrite [119]. The pearlite colonies in the J2+N- steel are large, blocky and dense. This is a consequence of its chemical composition, heat-treatment and deformation or manufacturing method. **Figure 8-3 (b)** is the SEM micrograph of the microstructure. The blocky pearlite (*P*) and the proeutectoid ferrite (α), are noted accordingly. The proeutectoid ferrite is the ferrite phase that formed before the formation of pearlite at the eutectoid temperature (723°C). **Figure 8-3(c)** is the part denoted with a yellow box in **Figure 8-3 (b)** and the micrograph shows differently oriented pearlite colonies as demarcated by the yellow lines. **Figure 8-3 (d)** shows what would have been the image of the part in the yellow box in **Figure 8-3 (c)** on a typical transmission electron microscopy (TEM) micrograph. The structure that appears as parallel rods in that colony is cementite. Note that the cementite (θ) plates are not generally continuous. The ferrite that formed in between the cementite rods are referred to as eutectoid ferrite – because they are produced during the pearlite formation at the eutectoid temperature. **Figure 8-3 (d)**, is extracted from Ref [208].

Figure 8-4 shows the microstructures of G8+M, G10+M. These microstructures are found to be similar, but it is evident that they are distinct from that of J2+N in **Figure 8-3**. **Figure 8-4 (a)** is the optical image of G10+M while **Figure 8-4 (b)** is the SEM image. Similarly, **Figure 8-4 (c)** is the optical image of G8+M while **Figure 8-4(d)** is the SEM image. It is evident that the SEM image shows more details of the microconstituents. In both microstructures, the pearlite nodules are very small compared to that of J2+N and the volume fraction of the pearlite is also less than that of J2+N. **Figure 8-5** shows the Energy-dispersive X-ray (EDX) analysis of the chemical composition analyses on the phases present in S355J2+N steel at positions 1 to 31 and the results are shown in **Table 8-3**. The relative abundance of the major alloying elements in the phases is given in wt%. Note that the EDX composition simply shows the relative abundance of the alloying elements in the phase and has no direct relationship with the steel overall composition shown in **Table 8-2**. The ferrite phases in **Figure 8-5** tends to have three forms – high relief (HR), low relief (LR) and ribbon like structure (RB).

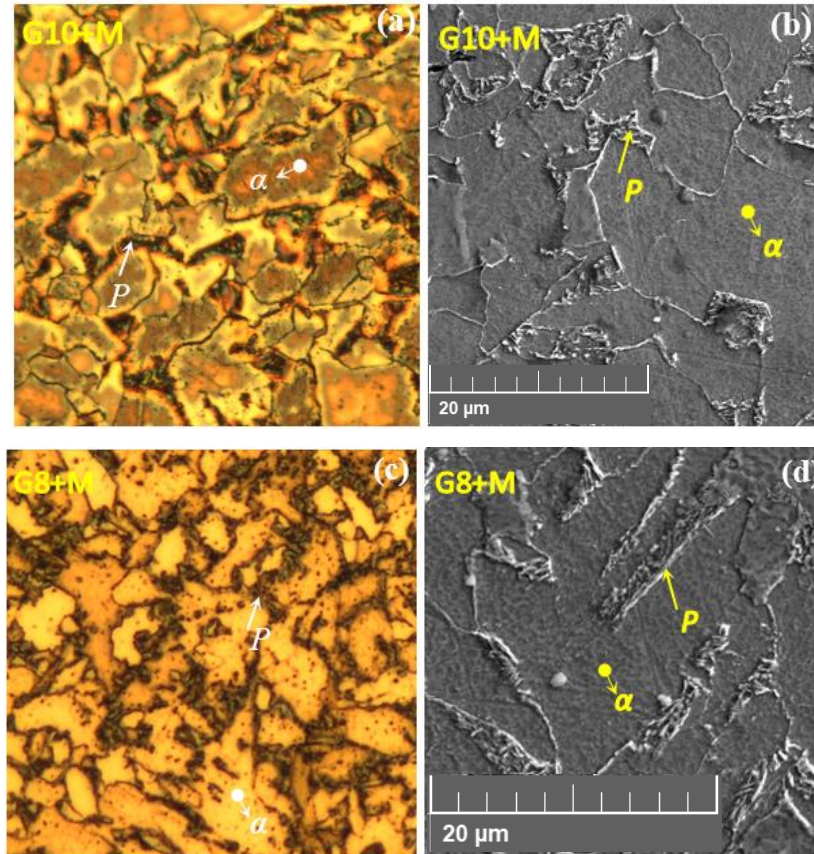


Figure 8-4: Optical and SEM micrographs of ferrite-pearlite TMCP steels (a) TMCP - S355G10+M

Note that the low and high relief α phases could not be distinguished under optical micrographs. Numbers 1, 3, 4, 16, 21, 30 etc., are examples of ferrite phase with high relief, while 2, 5, 17, 25, 28, etc. are examples of low relief ferrite phase. The light etching is the pearlite (P), e.g. numbers 6, 7, 8, 13, 26 etc. For the purposes of description, α_{HR} denotes α with high relief and α_{LR} is low relief while the ferrite ribbon is denoted α_{RB} .

A simple observation of the EDX result shows that the alloying compositions of LR and HR ferrite phase is generally the same. But for the RB, all the alloy compositions are the same with that of LR and HR except that the amount of carbon C, oxygen O and silicon Si are higher as can be seen from **Table 8-3**.

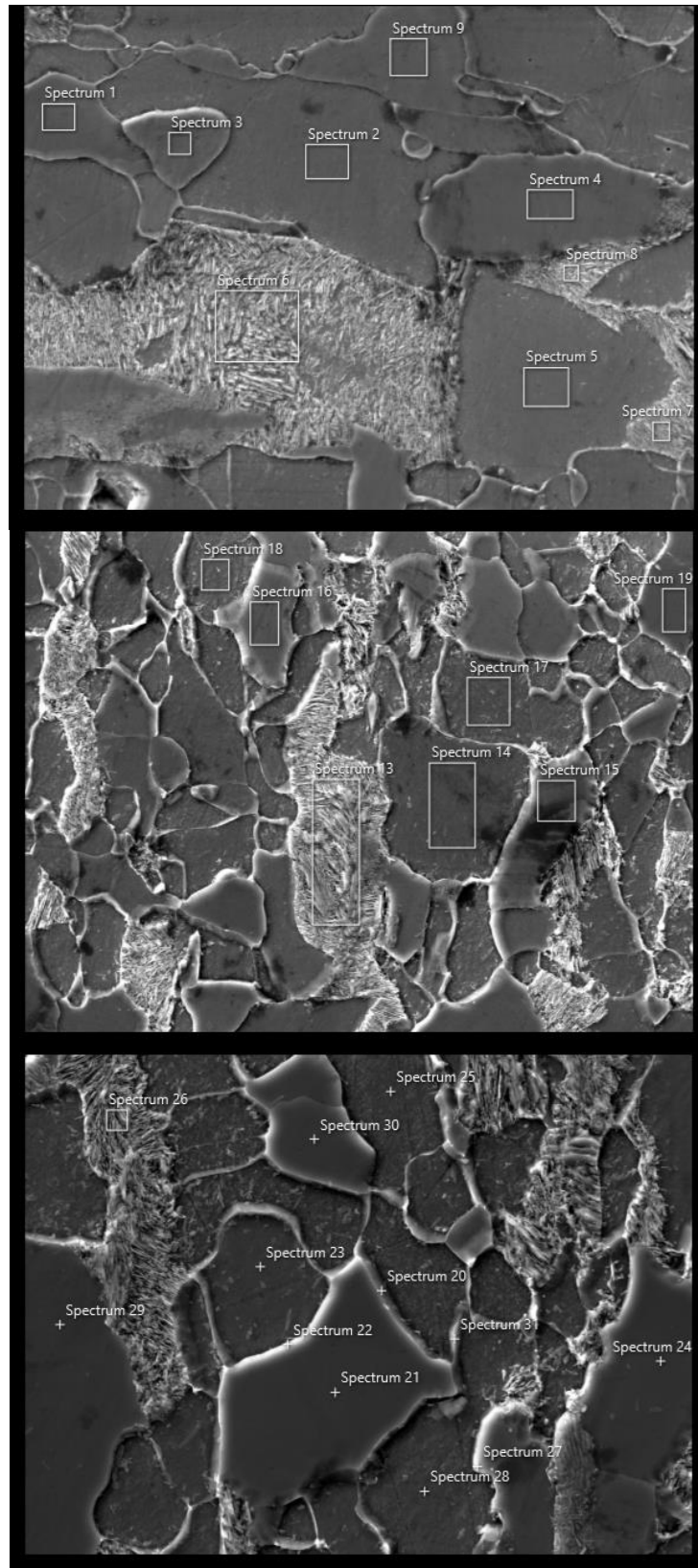


Figure 8-5: EDX analysis of phase composition in S355J2+N steel

Table 8-3: EDX compositional results of ferrite and pearlite phases in Figure 8-5
(HR = high relief α , LR = low relief α , RB = α ribbon, EDX= Energy-dispersive X-ray (EDX) analysis)

Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
LR 2	4.77	1.27	0.12	0.35	0.01	0.03	0.04	0.13	0.42	0.01	0.00	0.00	92.32
LR 5	4.87	1.58	0.08	0.38	0.01	0.03	0.00	0.13	0.41	0.02	0.01	0.16	92.25
LR 14	6.14	1.40	0.11	0.35	0.04	0.00	0.07	0.10	0.41	0.01	0.00	0.07	91.26
LR 17	5.82	1.39	0.11	0.34	0.02	0.03	0.00	0.11	0.41	0.03	0.00	0.09	91.61
LR 18	5.86	1.42	0.08	0.39	0.05	0.02	0.00	0.13	0.43	0.02	0.03	0.03	91.49
LR 23	5.65	1.38	0.11	0.36	0.01	0.01	0.00	0.09	0.38	0.02	0.00	0.13	91.63
LR 25	5.34	1.43	0.08	0.39	0.02	0.02	0.00	0.12	0.42	0.02	0.01	0.16	91.91
LR 28	5.55	1.40	0.06	0.37	0.00	0.03	0.00	0.11	0.37	0.03	0.01	0.13	91.82
Average	5.50	1.41	0.09	0.37	0.02	0.02	0.01	0.12	0.41	0.02	0.01	0.10	91.79

Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
HR 1	4.81	1.29	0.05	0.36	0.01	0.02	0.00	0.12	0.39	0.01	0.02	0.00	92.18
HR 3	5.79	1.29	0.07	0.37	0.00	0.02	0.04	0.12	0.39	0.01	0.00	0.21	91.66
HR 4	4.73	1.34	0.07	0.37	0.03	0.01	0.00	0.12	0.40	0.02	0.03	0.08	92.74
HR 9	4.78	1.26	0.09	0.36	0.03	0.02	0.03	0.12	0.42	0.01	0.00	0.11	92.50
HR 15	8.55	1.29	0.06	0.34	0.03	0.02	0.02	0.11	0.38	0.02	0.00	0.11	89.01
HR 16	6.35	1.44	0.07	0.39	0.03	0.01	0.00	0.13	0.45	0.03	0.01	0.00	91.05
HR 19	5.71	1.29	0.11	0.35	0.02	0.03	0.00	0.11	0.41	0.02	0.03	0.05	91.82
HR 21	5.15	1.37	0.08	0.38	0.00	0.02	0.00	0.12	0.36	0.02	0.01	0.07	92.33
HR 24	4.95	1.66	0.09	0.44	0.02	0.01	0.00	0.11	0.43	0.02	0.00	0.02	92.18
HR 29	5.47	1.37	0.09	0.38	0.00	0.00	0.03	0.12	0.40	0.00	0.00	0.06	92.01
HR 30	5.15	1.36	0.11	0.36	0.00	0.02	0.00	0.13	0.39	0.03	0.02	0.08	92.22
Average	5.59	1.36	0.08	0.37	0.02	0.02	0.01	0.12	0.40	0.02	0.01	0.07	91.79

Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
RB 20	9.97	1.32	0.09	0.51	0.06	0.03	0.00	0.21	0.38	0.02	0.01	0.34	87.03
RB 22	10.03	1.33	0.07	0.47	0.03	0.02	0.02	0.19	0.37	0.03	0.00	0.78	86.38
RB 27	9.77	1.53	0.10	0.39	0.01	0.02	0.00	0.12	0.40	0.04	0.01	0.33	91.12
RB 31	9.65	1.44	0.09	0.48	0.07	0.02	0.00	0.15	0.41	0.03	0.02	0.36	87.96
Average	9.86	1.41	0.09	0.46	0.04	0.02	0.01	0.17	0.39	0.03	0.01	0.45	88.12

Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
P 6	7.97	1.41	0.07	0.42	0.01	0.02	0.05	0.12	0.39	0.02	0.00	0.92	88.51
P 7	8.23	1.67	0.07	0.48	0.05	0.03	0.00	0.10	0.32	0.03	0.02	1.46	87.39
P 8	9.58	1.40	0.08	0.41	0.04	0.02	0.00	0.11	0.41	0.01	0.03	1.72	86.10
P 13	10.92	1.51	0.08	0.34	0.07	0.01	0.00	0.12	0.42	0.02	0.03	1.31	85.05
P 26	9.94	1.49	0.10	0.34	0.13	0.02	0.00	0.09	0.37	0.02	0.03	1.51	85.91
Average	9.33	1.50	0.08	0.40	0.06	0.02	0.01	0.11	0.38	0.02	0.02	1.38	86.59

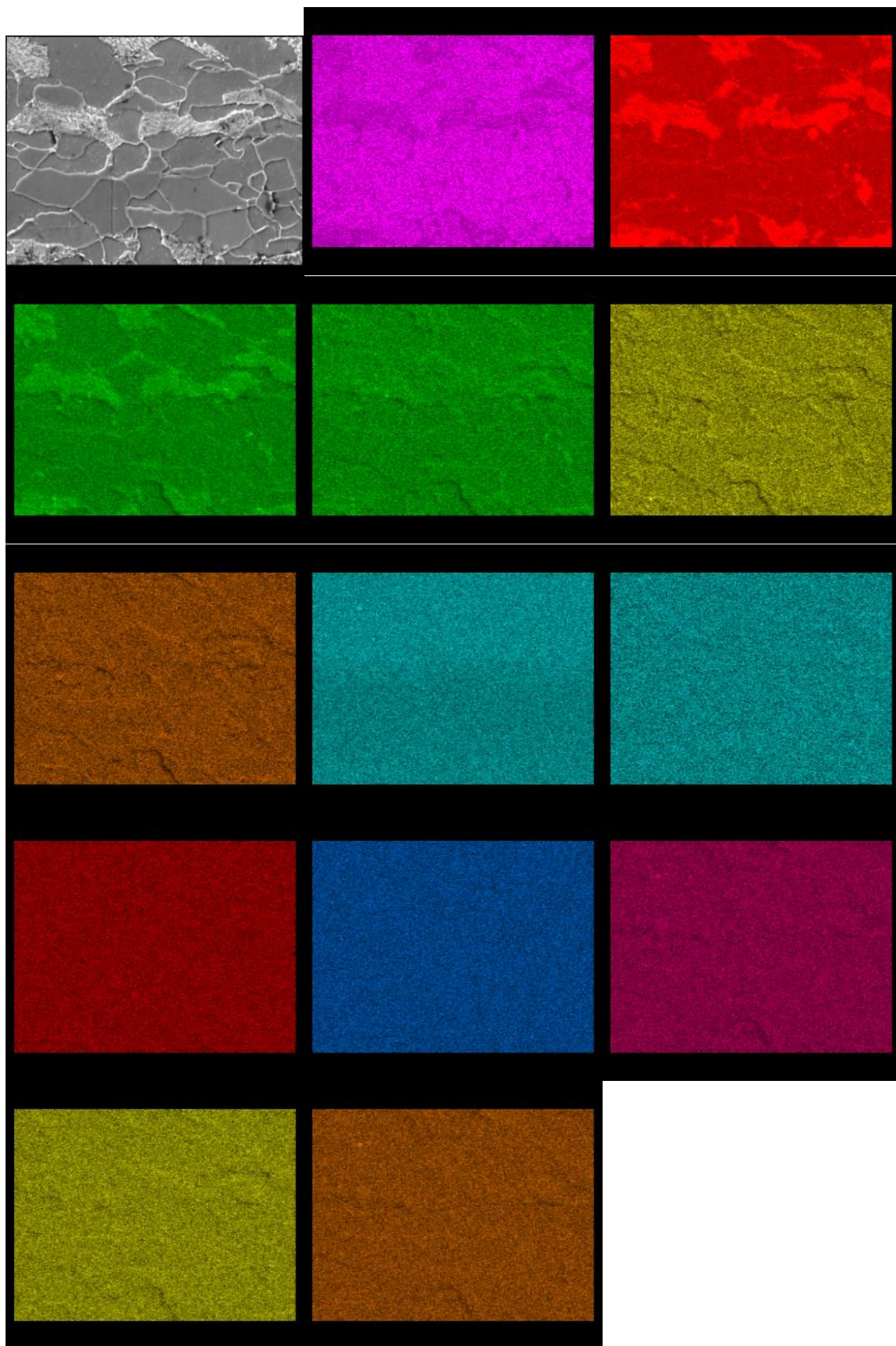


Figure 8-6: EDX mapping of phase composition in S355J2 – sample 1

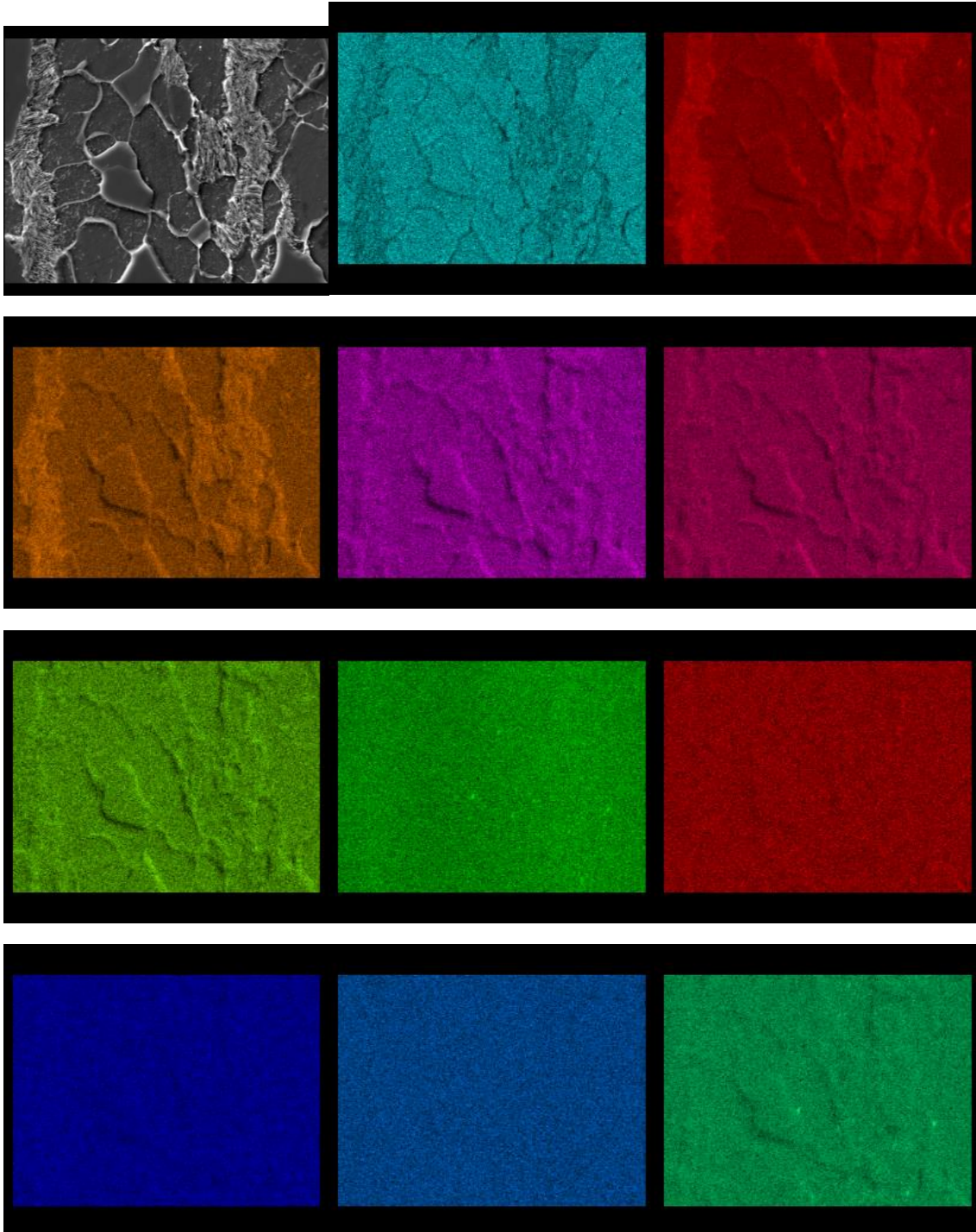


Figure 8-7: EDX mapping of phase composition in S355J2 – sample 2

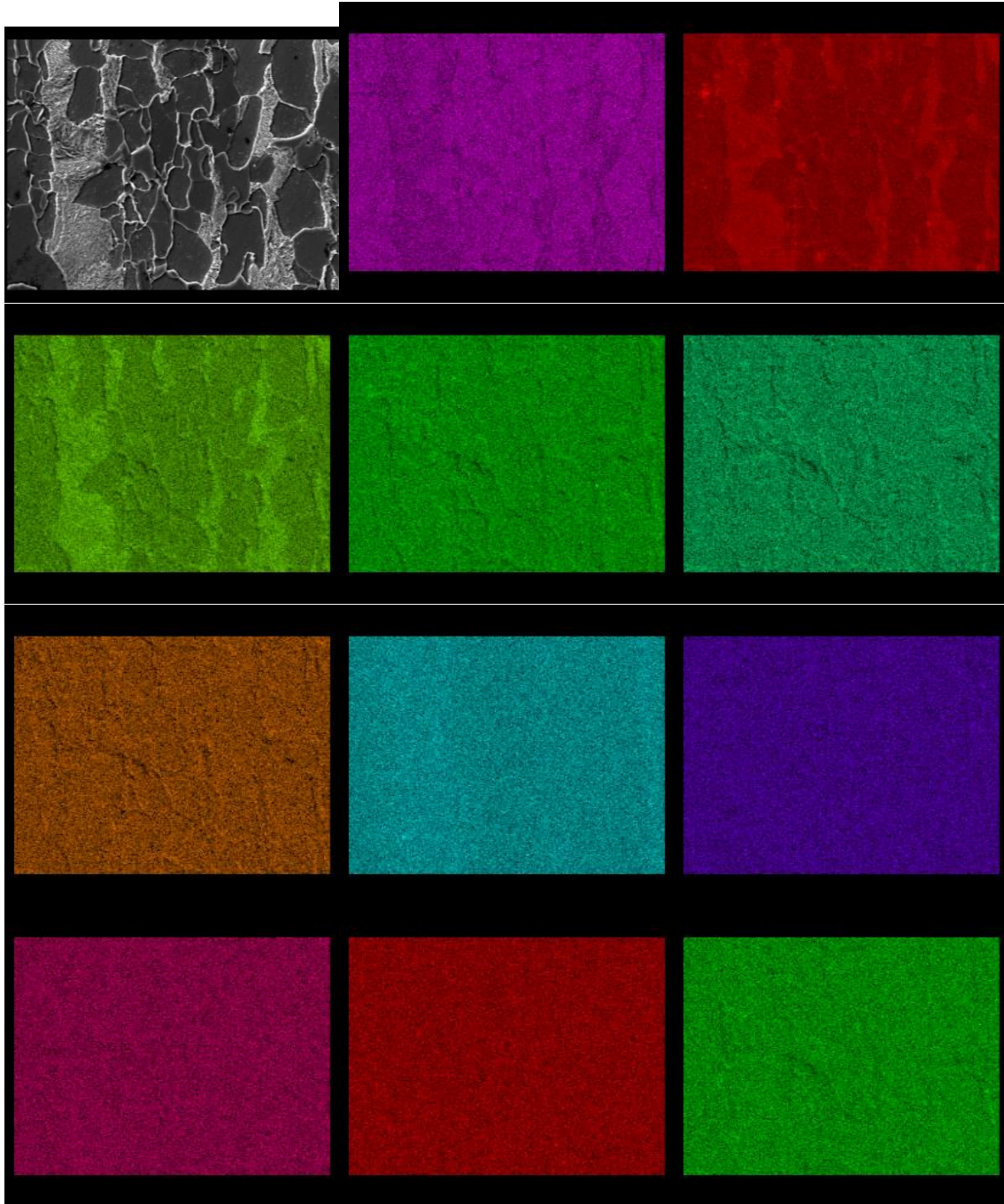


Figure 8-8: EDX mapping of phase composition in S355J2 – sample 3

The pearlite is found to have generally the same composition with the RB, but with higher oxygen content. The Fe contents of RB and P are less than that of LR and HR. EDX mapping was carried out to understand the distribution of these

alloying elements and the results are presented in **Figure 8-6**, **Figure 8-7** and **Figure 8-8**. The light area shows where a particular element has partitioned.

The Fe map confirms that Fe is less in P and RB and the phases appeared as dark areas. The C map shows that the carbon is partitioned majorly into the P and RB as indicated by the light areas. Other elements that can be seen to partition into the RB are Si, Al and Mg. This implies that RB is a ferrite phase with high concentration of solutes. An important observation here is that the concentration of these solutes may weaken this phase, making it an easy path for fatigue crack growth. A microhardness test was carried out on the NR and TMCP steels to determine the hardness of the ferrite and pearlite phases across the length of the samples. The test was done using a Vickers hardness tester Zwick/Roell Indentec with 100g test load (HV0.1). A pyramidal indenter was used. The length of the diagonal of the square indent made by a pyramidal was measured with a built-in optical microscope and result displayed digitally. Several measurements were taken, some which are presented in **Figure 8-9**.

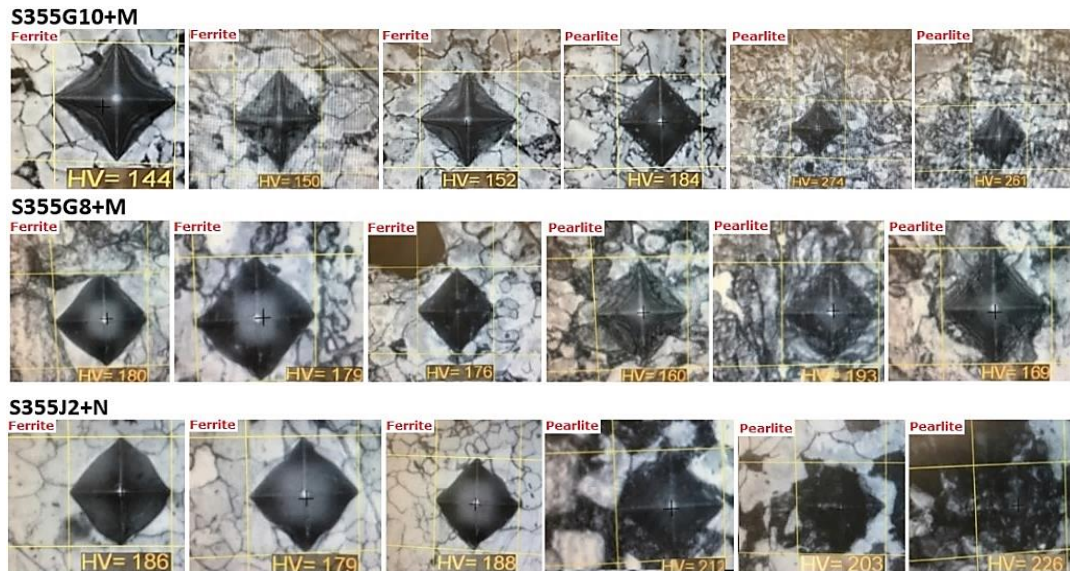


Figure 8-9: Hardness measurement of phases in G10, G8 and J2N steels

The average Vickers hardness (HV) values of ferrite and pearlite phases are given in **Table 8-4**.

Table 8-4: Microhardness measurement of ferrite and pearlite phases in G10+M, G8+M and J2+N (in wt%)

(a)	Ferrite (HV)		
Phase no.	G10	G8	J2N
1	152	180	186
2	144	171	179
3	166	179	188
4	128	176	193
5	150	159	184
Average	148	173	186

(b)	Pearlite (HV)		
Phase no.	G10	G8	J2N
1	184	193	184
2	177	160	226
3	274	180	212
4	261	176	196
5	169	185	203
6	176	169	227
Average	207	177	208

Table 8-4 (a) shows that the ferrite phase in J2+N is the hardest, followed by G8+M and lastly G10+M. From **Table 8-4 (b)**, the pearlite is generally harder than the ferrite, but this hardness varies with the pearlite morphology or nodule. Despite that the overall carbon content of NR (0.17wt%C) is higher than that of G10 (0.06wt%C), the hardness of pearlite in G10 appears to be of the same value with that of J2+N. The hardness of the pearlite in G8+M is the least. This difference in pearlite hardness may be attributed to the pearlite interlamellar spacing. It has been reported that the flow stress of pearlite increases with decrease in the interlamellar spacing [243][244][245][246][247]. Similarly, the hardness of the pearlite increases with decrease in interlamellar spacing [248][249]. This tends to imply that the accelerated cooling associated with TMCP technique was such that fine pearlite spacing was produced.

Figure 8-10 shows phase features that are important in analysing the work presented in this paper. **Figure 8-10 (a & b)** are microstructures of J2+N, **Figure 8-10 (c & d)** are those of G8+M and **Figure 8-10 (e & i)** are for G10+M. In **Figure 8-10**, the yellow arrow with a dot indicates the LR ferrite α_{LR} and the red triangle shows the HR ferrite α_{HR} .

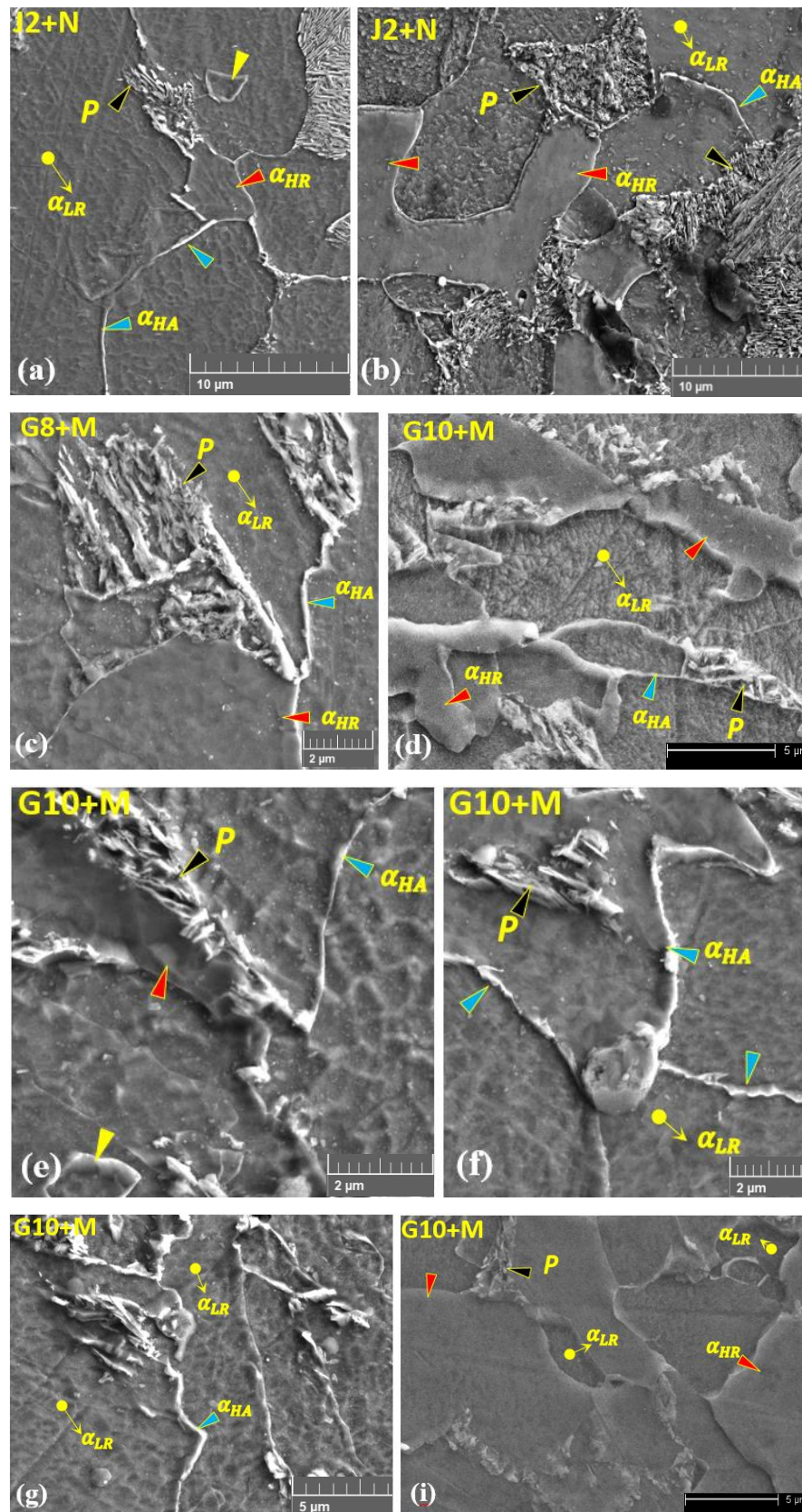


Figure 8-10: Phase properties in modern ferrite-pearlite steels

The RB, α_{HA} is denoted with a blue triangular arrow and the HR with red arrow. The pearlite, P , denote black barrow, appears to form from the ferrite ribbons and the width is relatively very small in the TMCP steels. In the NR steel, P forms as a blocky phase in the α matrix.

8.3 Microstructural effect on the da/dN vs. ΔK curve

8.3.1 The Effect of using fitted crack length $\hat{a}_i$ in determination of crack growth rate

To study the true effect of microstructure, waveform and the environment, the effect of da/dN curve smoothening must be accounted for. It is pertinent to note that there are two main disadvantages in using the fitted a in calculating FCGR. One of these is drawing and locating the best fit or determining which degree of polynomial gives best fit – hence is subjective. This is so because it will influence the Paris law constants. Though nowadays, *Excel*® can perfectly fit such curves. The ASTM recommends the use of 2nd order polynomial or a quadratic equation (Eq. (4-11)) and the associated scaling parameters. The second problem is smoothening of the a vs. N data [186][250], which removes or blurs important effects of the microstructure and environment on FCG mechanism. Phenomena such as waveform effect, crack deviation, crack closure/interlocking, crack branching and wedging action of metal crumbs in-between branched and re-joined crack fronts will cause varying levels of deviations in data points in the a vs. N plot. The morphology/grain orientation of steel material can cause local acceleration and retardation of crack propagation. The establishment of the best-fit curve using least square method and using the fitted crack length in calculating crack growth rate is like interpreting these effects as human errors and material anomalies.

To demonstrate the effect of using fitted crack length, **Figure 8-11** shows the a vs. N raw data plot for air tests used in calibration process for G10-A9F5, G8-A10F5 and G10-A10F5. The trends are typical of crack growth behaviour in steel. We can see that the slope increased rapidly towards the final failure. **Figure 8-11 (a)**

shows the fitting of 2nd, 3rd (**Figure 8-11 (b)**) and 4th (**Figure 8-11 (c)**) degree polynomials (dotted lines) to the experimental fatigue data. The $a = F(N)$ fitting equations for the two load levels are shown in **Table 8-5**.

For the polynomials generated in **Table 8-5**, generally we can express them as:

$$\hat{a} = f(N) = b_0 N^n + b_1 N^{n-1} + \dots + b_{n+1} \quad (8-1)$$

The first derivative of Eq. 8-1, which is the FCGR, da/dN becomes:

$$\frac{d\hat{a}_i}{dN} = nb_0 N_i^{n-1} + (n-1)b_1 N_i^{n-2} + \dots + b_n \quad (8-2)$$

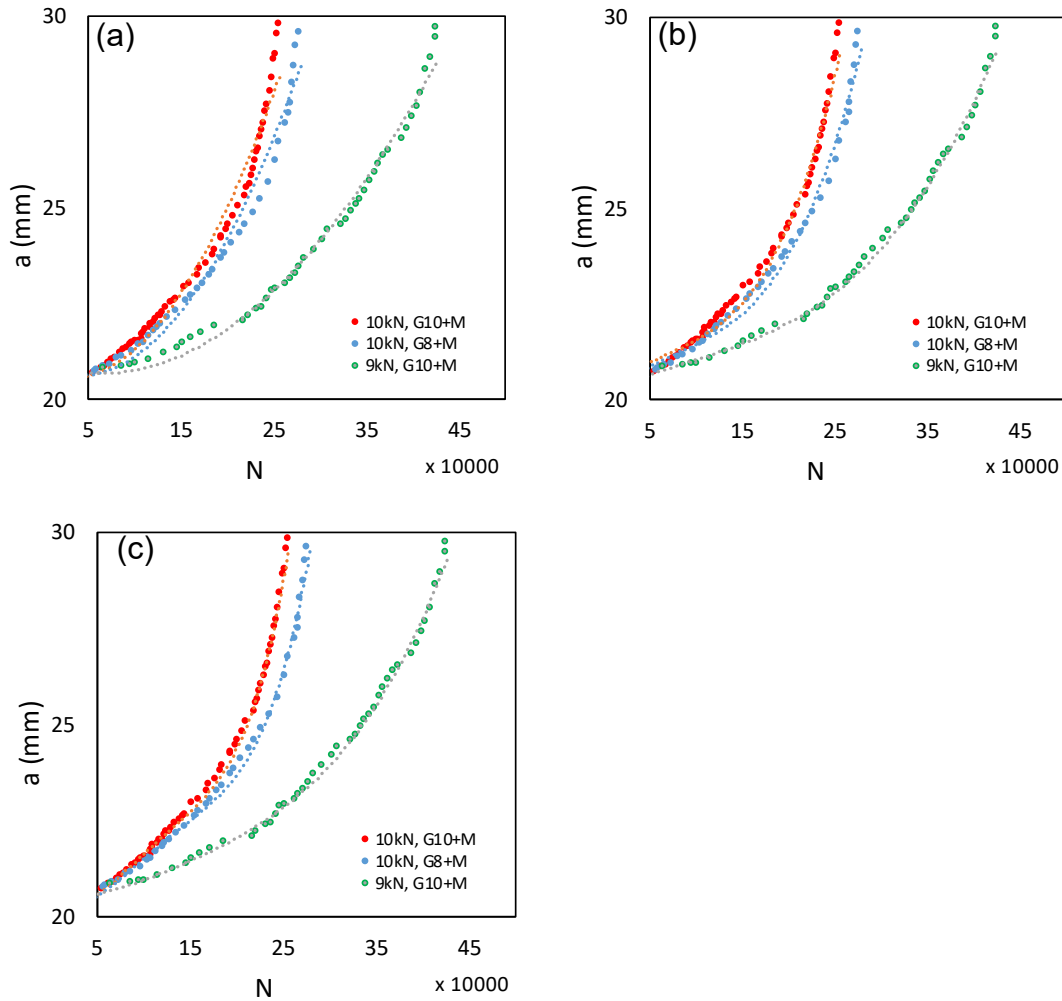


Figure 8-11: Plot of a vs. N in air with different smoothing curves (a) 2nd (b) 3rd and (c) 4th degree polynomials

An attempt to fit all the fatigue data points in **Figure 8-11** increased the polynomial order. The derivatives of the $\hat{a}_i = f(N)$ is given in **Table 8-6**. **Figure 8-12** shows the use of the experimentally measured crack length data and the fitted crack length $\hat{a}_i$ data from *Excel*® at different degrees of polynomial function in calculating FCGR using the two load levels for S355G10+M steel subgrade.

In **Figure 8-12 (a)** and **(b)**, the 2nd order polynomial equation (**Table 8-5**) for determination of the fitted crack length $\hat{a}$ and $d\hat{a}/dN$ (derived in **Table 8-6**) where compared with the $d\hat{a}/dN$ obtained using the fitted crack length $\hat{a}$ in the 7-point polynomial method (used in the present study). Note that in the former case, only number of cycle N is needed to obtain the $d\hat{a}/dN$, but in the latter case, as implemented in the *Excel*®, both $\hat{a}$ and N are needed. The results show that the two methods gave the same curve, and this will enable us to observe reliably the variation between the raw data and fitted data.

Table 8-5: Fitted crack length polynomials

Test	$\hat{a}$ vs. N curve	Fitting
	(a) 2nd order	
10kNG10	$\hat{a} = 1.55 \times 10^{-10}N^2 - 9.54 \times 10^{-6}N + 20.7$	$R^2 = 0.977$
10kNG8	$\hat{a} = 1.44 \times 10^{-10}N^2 - 1.22 \times 10^{-5}N + 20.9$	$R^2 = 0.975$
9kNG10	$\hat{a} = 6.17 \times 10^{-11}N^2 - 7.77 \times 10^{-6}N + 20.9$	$R^2 = 0.989$
	(b) 3rd order	
10kNG10	$\hat{a} = 9.54 \times 10^{-16}N^3 - 2.13 \times 10^{-10}N^2 + 2.71 \times 10^{-5}N + 20.0$	$R^2 = 0.990$
10kNG8	$\hat{a} = 7.47 \times 10^{-16}N^3 - 1.85 \times 10^{-10}N^2 + 2.65 \times 10^{-5}N + 19.9$	$R^2 = 0.988$
9kNG10	$\hat{a} = 1.55 \times 10^{-16}N^3 - 4.43 \times 10^{-11}N^2 + 1.20 \times 10^{-5}N + 20.1$	$R^2 = 0.996$
	(c) 4th order	
10kNG10	$\hat{a} = 1.07 \times 10^{-20}N^4 - 4.68 \times 10^{-15}N^3 + 7.39 \times 10^{-10}N^2 - 2.78 \times 10^{-5}N + 20.6$	$R^2 = 0.997$
10kNG8	$\hat{a} = 7.61 \times 10^{-21}N^4 - 3.60 \times 10^{-15}N^3 + 6.11 \times 10^{-10}N^2 + 2.38 \times 10^{-5}N + 20.6$	$R^2 = 0.995$
9kNG10	$\hat{a} = 3.72 \times 10^{-22}N^4 - 1.79 \times 10^{-16}N^3 + 5.29 \times 10^{-11}N^2 + 2.01 \times 10^{-6}N + 20.4$	$R^2 = 0.996$

Table 8-6: Derivatives of polynomials in Table 8-5

Test	$d\hat{a}/dN$
	(a) 2nd order
10kNG10	$3.10 \times 10^{-10}N - 9.54 \times 10^{-6}$
10kNG8	$2.88 \times 10^{-10}N - 1.22 \times 10^{-5}$
9kNG10	$1.23 \times 10^{-10}N - 7.77 \times 10^{-6}$
	(b) 3rd order
10kNG10	$2.86 \times 10^{-15}N^2 - 4.26 \times 10^{-10}N + 2.71 \times 10^{-5}$
10kNG8	$2.24 \times 10^{-15}N^2 - 3.70 \times 10^{-10}N + 2.65 \times 10^{-5}$
9kNG10	$4.65 \times 10^{-16}N^2 - 8.86 \times 10^{-11}N + 1.20 \times 10^{-5}$
	(c) 4th order
10kNG10	$4.28 \times 10^{-20}N^3 - 1.40 \times 10^{-14}N^2 + 1.48 \times 10^{-9}N - 2.78 \times 10^{-5}$
10kNG8	$3.04 \times 10^{-20}N^3 - 1.08 \times 10^{-14}N^2 + 1.22 \times 10^{-9}N + 2.38 \times 10^{-5}$
9kNG10	$1.49 \times 10^{-21}N^3 - 5.37 \times 10^{-16}N^2 + 1.06 \times 10^{-10}N + 2.01 \times 10^{-6}$

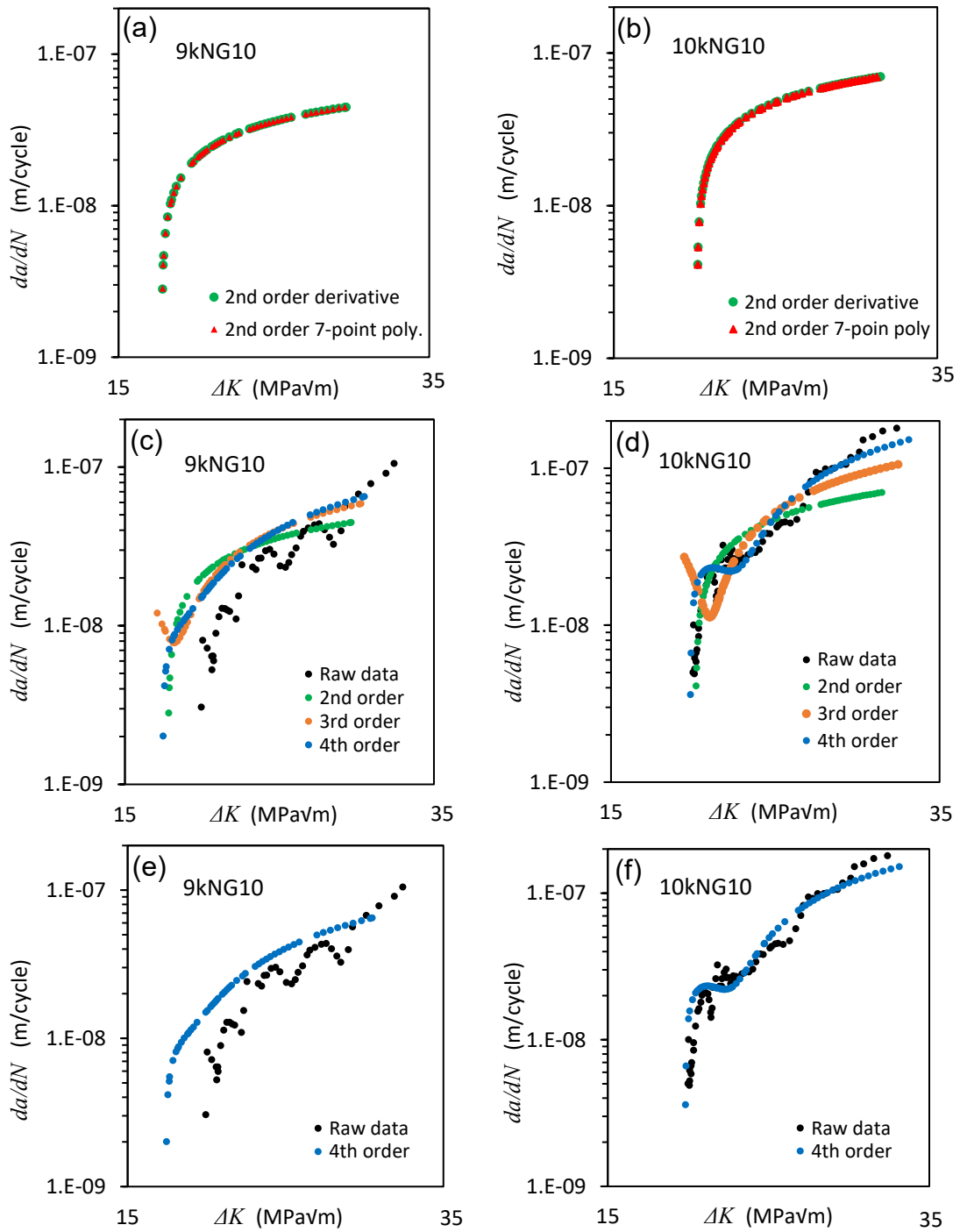


Figure 8-12: Smoothing effect on FCGR curve

Figure 8-12 (c) and (d) show the plotting of the raw unfitted data for the two load levels 9kN and 10kN using 7-point polynomial technique implemented in *Excel*® (black data points) and fitting the raw data using 2nd, 3rd & 4th order polynomial

equations to obtain the fitted crack lengths $\hat{a}$, using the fitted crack lengths to calculate ΔK . The $d\hat{a}/dN$ was obtained from the $\hat{a}_i = f(N)$ derivative in **Table 8-6** and finally plotting the $d\hat{a}/dN$ against ΔK for all the cases as shown. The results show that the degree of polynomial equation used affects the nature and the position of the curve.

That is, despite using the same data set, different curve-shifts and orientations were obtained. The variation is so much that considerable number of errors will be made in conclusions regarding, for example, the apparent threshold converging point and Paris law constants. This effect also varies with load level. The 4th degree polynomial appears to give the best match to the curve of the raw data as shown in **Figure 8-12 (f)**, but there is significant disparity for the 9kN as shown in **Figure 8-12 (e)**. Also note that the curves are smooth.

The increase in the order, especially the 4th order, improves the fitting of the polynomial function to the data set, but the undulating effect in the raw data set has been smoothened out. It can be seen that all the fitted data removed the wavy nature of the 9kNG10 raw data curve. If this rippling nature of the curve is caused by the microstructure or the environment, then such effect would no longer be observed in the processed information. Thus, to study microstructural effect on FCGR, the use of raw data is most desirable and indispensable as smoothed curve tends to remove the local crack growth acceleration or retardation effects on the ΔK vs. da/dN curve. Also note that as the order increases, the inflexions increase which yielded negative crack growth rate. Since there is no such thing as negative crack growth, the information in the raw data has been profoundly undermined.

Before now, some methods of crack measurement were such that enough scatter was obtained in the fatigue data. It appears to be the subjectivity of the fitting curve that necessitated the use of fitted curves in Standards. In this modern time, we now have powerful and dedicated optical cameras and associated software that can pinpoint almost exactly where the crack tip is at any instance and the

image projected on the computer screen. With modern measurement software installed in the computer, the a per pre-defined fatigue cycles can be measured with higher accuracy. The a measurement and data acquisition can even be made remotely using simple remote desktop app. The facilities can allow for continuous acquisition of fatigue data. In this case, highly reliable fatigue data can be obtained, and use of fitted crack length may no longer be required because whatever scatter that may be seen in the a vs. N curve will be due to the effect of the materials property, loading condition or/and the environment. We can see from **Figure 8-11** that there are many data points and there is also little or no scatter in the fatigue data obtained in air. Thus, with the aid of modern optical cameras, associated software, development of sensitive and reliable strain gauges, crack growth measurements can be made with very minimal human errors. Such modern crack measurement facilities were used in obtaining the fatigue data reported here and 7-points incremental polynomial technique was implemented on the raw data.

8.4 The influence of microstructure on the Fatigue crack growth rate in air

The formation of these phases into different morphologies created different local microstructural properties. The presence of a hard phase (pearlite) in a soft matrix (ferrite) is expected to influence local FCG behaviour in steel [124]. The crack paths in these microstructures are also expected to vary. Steel microstructure influences fatigue crack propagation primarily by changing the mode of crack propagation. The mode can change from ductile striation to cleavage mode and the combination of the two modes is reported to increase crack propagation rates.

There are many studies that claimed that microstructure has little or negligible effect at the intermediate fatigue crack growth rates (Paris Region) [29][187][193][194][195][196][197][198]. Some other studies concluded that grain size affected sub-critical crack growth to the extent to which it affected fracture toughness, K_{Ic} and the onset of rapid fracture and that grain size did not affect

the FCG threshold, ΔK_{th} [199]. Zerst, U in his recent publication [197] reported, I quote, ‘... *Perhaps the most astonishing observation is that the curve, in the Paris Region, is only slightly influenced by the microstructure of the material within whole material classes such as steels or aluminium alloys.*’ In fact, it is commonly reported or almost a concluded study that microstructure has little or no influence in the Paris Region of the da/dN sigmoidal curve. This section is essentially to show that this conclusion is not true for ferrite-pearlite steels. Therefore, this work is focused towards understanding how ferrite-pearlite grain refinement affects the fatigue performance of these marine steels, with emphasis on the Paris region of the da/dN vs. ΔK curve.

In a recent study, the authors evaluated the resistance of the Normalized-rolled (NR) and Thermo-mechanical control process (TMCP) steels given in **Table 8-2** to fatigue crack growth (FCG). The details of the experimental methods have been discussed in Refs [144][230]. The results of that experimental work - for FCGR tests in air are shown in **Figure 8-13**.

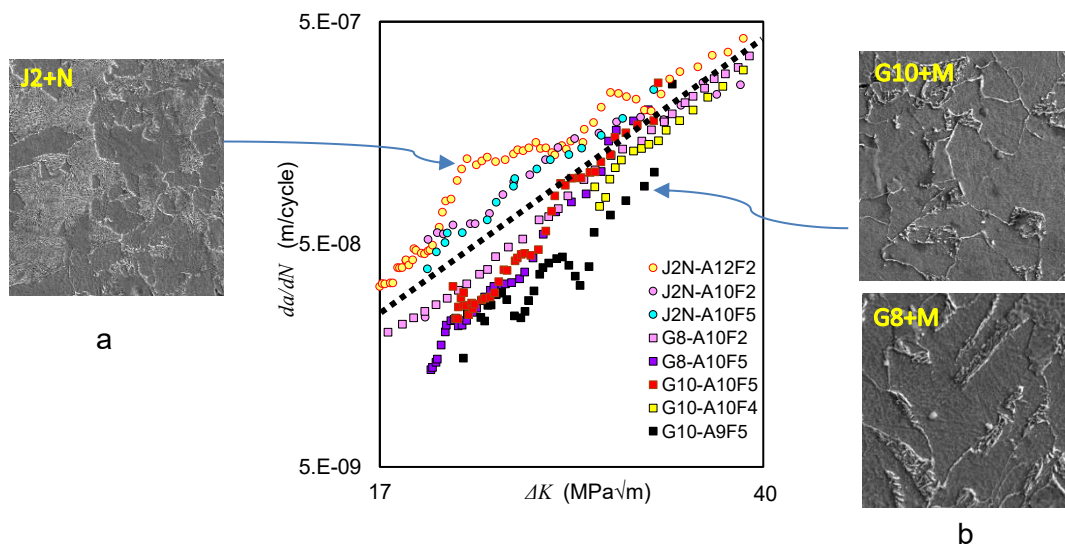


Figure 8-13: Plot of FCGR in air for NR (J2+N) and TMCP steel subgrades (G8+M & G10+M)

Shown also are the microstructures of the NR steel (J2+N) and TMCP steels (G10+M & (G8+M) examined. In **Figure 8-13**, it appears there is a demarcation between the FCGR of NR and that of TMCP in the Paris Region. The arbitrary line drawn as shown appears to be the demarcation line. The FCGR in G8+M and G10+M steels are essentially the same and significantly lower than that of the J2+N steels. The result showed that the FCGR in the NR is significantly higher than that of the TMCP steel in the Paris Region under the same experimental conditions. Generally, the FCGR increased with increase in load level from 10kN to 12kN in J2+N and also from 9kN to 10kN in TMCP. This is the expected trend in air, where increased in fatigue load level increases the FCGR. We can also observe that the FCGR in J2+N for 10kN under 2Hz and 5Hz are the same. This tends to confirm that frequency has no effect on the FCGR of steel in air under the same loading condition.

To validate this finding, a comparative study of FCGR in NR and TMCP was carried out. The steel grade, chemical composition in wt% (approximated to two decimal places), carbon equivalent (CE) and mechanical properties of the steels studied and compared with the data in this study are included in **Table 8-7**. The composition range for the steels in **Table 8-7** is shown in **Table 8-8**. It is pertinent to note that BS4360 Grade 50D, Q345, RD480 and EH36 are equivalent to S355J2+N hot-rolled normalized steel. Q345 is common structural steel in China and used in the nuclear industry and petrochemical plants. Laurito studied FCGR in intercritical-isothermally-quenched grain-modified steel designated as RD480. Though the chemical composition is comparable to other steels, the properties of the microstructure generated may differ significantly due to this heat treatment. Two microstructures of the same ABS grade EH36 steel produced by NR and TMCP are presented. This steel is often used in shipbuilding. Comparing these two microstructures of the same alloy will clearly support this study. **Table 8-9** lists all the studies and test conditions under considerations in this paper.

All the steels are designed to be used in both onshore and offshore structures. They are fundamentally the same. The only difference perhaps is that the steels were produced under different proprietary methods by different steel plants.

Table 8-7: Chemical compositions of the ferritic steels studied

Study	Steel Grade	C	Mn	Ni	Si	Cu	Cr	Mo	CE	E %	YS (Mpa)	UTS (Mpa)
Scott [154]	BS4360-50D (NR)	0.18	1.38	-	0.37	-	-	-	0.44	32	370	538
Appleton [157]	BS4360-50D (NR)	0.13	1.22	-	0.32	-	-	-	0.39	32	368	508
Thorpe [24]	BS4360-50D (NR)	0.17	1.35	0.07	0.35	0.17	0.09	0.01	0.44	32	370	538
Musuva, [168]	BS4360-50D (NR)	0.18	1.40	0.10	0.36	0.16	0.11	0.02	0.46	30	386	560
Thompson [214]	BS4360-50D (NR)	0.15	1.38	0.10	0.41	0.02	0.11	0.01	0.42	30	386	536
Correia [251]	P355NL1 (NR)	0.13	1.38	0.12	0.35	0.14	0.03	0.00	0.39	-	418	568
Jesus [167]	EN 10025 S355 (NR)	0.10	0.64	0.10	0.15	0.38	0.08	0.01	0.26	22	419	766
Adedipe [149]	S355J2+N (NR)	0.22	1.60	0.50	0.55	0.55	0.03	0.10	0.57	20	345	-
Xiong [210]	Q345 (NR)	0.15	1.54	-	0.34	-	-	-	0.44	26	370	585
Laurito [211]	RD480 (IGM)	0.09	1.17	0.01	0.05	0.01	0.02	0.01	0.30	32	459	541
Cheng [159]	EH36 (NR)	0.12	1.39	0.03	0.38	0.05	0.05	0.01	0.37	21	350	-
Callister [212]	K-50 (TMCP)	0.05	1.13	0.29	0.23	0.24	0.02	0.20	0.32	-	470	-
Tsay [191]	EH36 (TMCP)	0.13	1.32	0.03	0.31	-	0.03	-	0.37	21	420	-
Chapetti [213]	UFG – steel (TMCP)	0.15	1.80	-	0.20	-	-	-	0.47	14	696	1183
PS (T)	S355G10+M (TMCP)	0.06	1.57	0.33	0.27	0.24	0.03	0.01	0.37	38	435	545
PS & Ali [175][190]	S355G8+M (TMCP)	0.05	1.52	0.32	0.27	0.24	0.03	0.01	0.35	35	447	549

PS = present study

NR = rolled normalized

TMCP = thermomechanical control process

IGM = isothermally grain-modified

UFGS = Ultra-Fine-Grained Steel

Note that BS4360 Grade 50D, Q345, RD480 and EH36 are equivalent to S355J2+N hot rolled normalized steel.

Average %wt of other minor elements for all the steels: 0.01Ti, 0.02V, 0.01N, 0.02S, 0.02P, 0.03Al, 0.04Nb

$$C_{IW} (CE) = C + \frac{Mn}{6} + \left(\frac{Cr + Mo + V}{5} \right) + \left(\frac{Ni + Cu}{15} \right)$$

C_{IW} , CE = carbon equivalent

In calculating CE, Mo = 0.02, Cr = 0.02, Cu = 0.3, Ni = 0.3 have been assumed where the elements are missing in the published paper.

Table 8-8: Composition Ranges of ferritic steels in this study

C	0.05 - 0.22
Mn	0.60 - 1.80
Ni	0.01 – 0.50
Si	0.05 – 0.94
Cu	0.01 – 0.55
Cr	0.02 – 1.23
Mo	0.01 - 0.52

Table 8-9: Experimental conditions for the studied steels in

Study by	Steel grade	Wave form	Freq. (Hz)	R	%EL	YS (MPa)	UTS (Mpa)
Scott [154]	BS4360-50D	s	1 - 10	0 - 0.85	32	370	538
Appleton [157]	BS4360-50D	s	-	-	32	368	508
Thorpe [24]	BS4360-50D	s	1 - 10	0.5	32	370	538
Musuva [168]	BS4360-50D	s	30	0.08, 0.5, 0.7	30	386	560
Thompson [214]	BS4360-50D	s	0.5	0.1	30	386	536
Correia [251]	P355NL1 (NR)	-	-	0.0	-	418	568
Jesus [167]	S355	s	≤ 20	0.25, 0.5 & 0.75	22	419	766
Adedipe [149]	S355J2+N	s	2	0.1	20	345	-
Xiong [210]	Q345	s	0.5	0.1 - 0.5	26	370	585
Laurito [211]	RD480	s	10	0.1	32	459	541
Cheng [159]	ABS grade EH36	r	10	0.1 & 0.5	21	350	-
Callister [212]	Kontroll-50 (TMCP)	s	0.5	0.6	-	470	-
Tsay [191]	EH36 (TMCP)	s	5	0.1	21	420	-
PS (T)	S355G10+M (TMCP)	s	4&5	0.1	38	435	545
PS & Ali [175][190]	S355G8+M (TMCP)	s	2&5	0.1	35	447	549
Chapetti [213]	UFGS (TMCP)	-	30	0.3	14	696	1183

s = sine, r = random, PS = present study

All the tests were done under constant-amplitude sinewave, except Cheng [159] who performed his test under random waveform. The general range for temperature of the studies is about 27 °C to 0°C. From the conclusions in the literatures [150][153][161][167][168][169][170], stress ratio (R) in the range 0.1-0.85, frequency in the range 0.005Hz – 50Hz and waveforms appear to have little or no effect on the FCGR in the ferrite-pearlite steels in air. Thus, the inferences here are within these limits. The marine steels have microstructures consisting essentially of small volume fraction of pearlite in a ferrite matrix. The morphology of the phases varied from banded to unbanded depending on the production process and sample thickness.

The general microstructures of the steels in **Table 8-7** are shown in **Figure 8-14**. The grade of steel and the reference sources are also given. **Figure 8-14 (a, b, c, d)** are microstructures of NR steels.

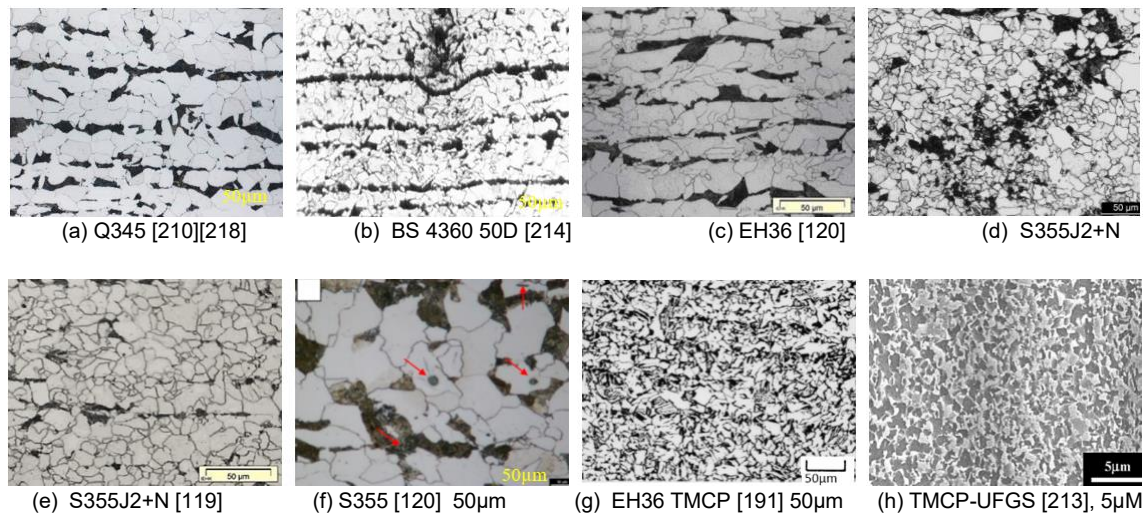


Figure 8-14: Typical microstructures of low alloy ferrite-pearlite steels

The black etching is a banded pearlite forming in the rolling direction and the light etching is the ferrite. **Figure 8-14 (e & f)** are microstructures of the same NR steels, but taken from directions that are un-banded [119].

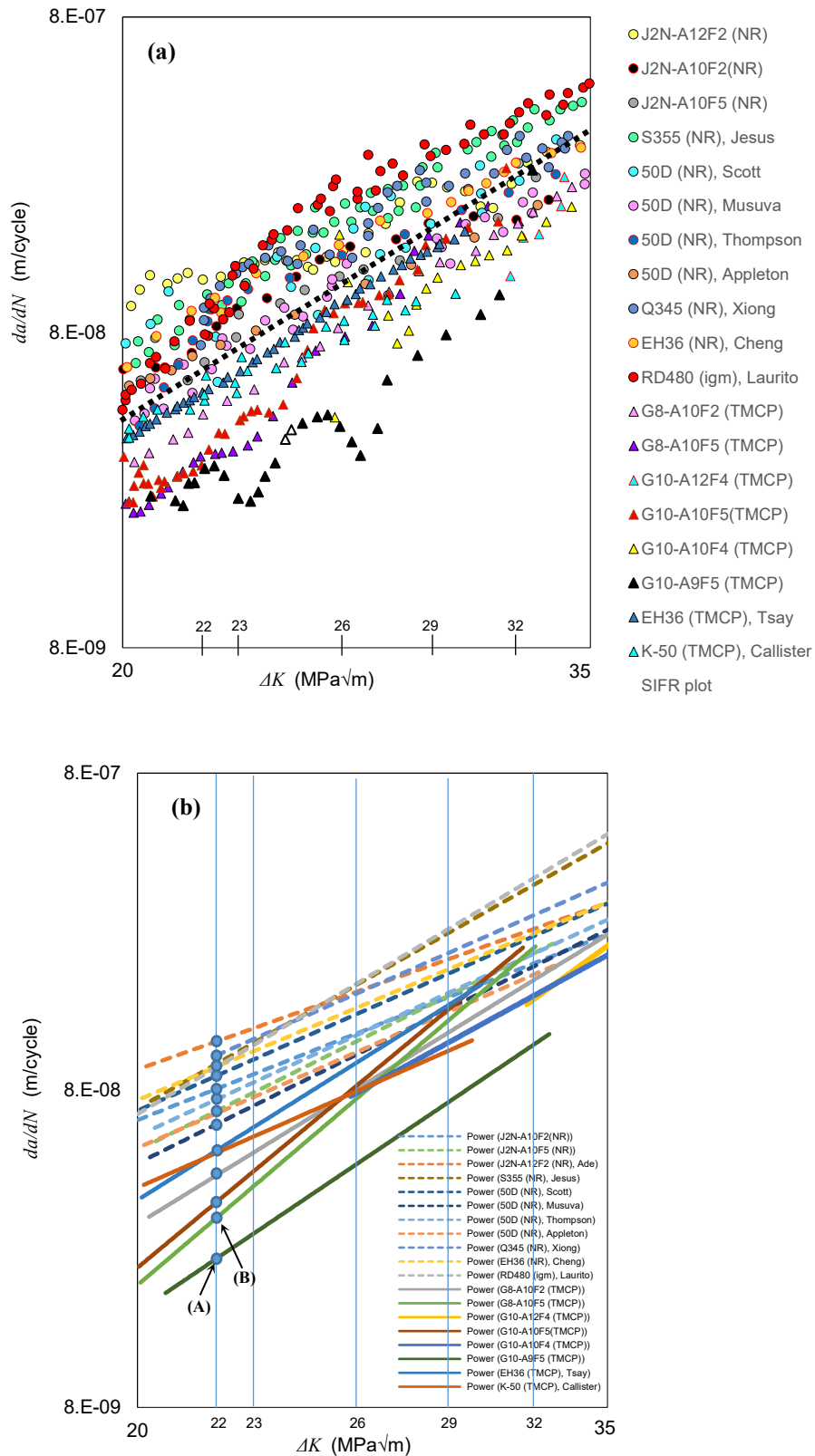


Figure 8-15: The FCGR in NR and TMCP in air (a) Paris Region with demarcation line (b) placing of vertical lines to determine effect of mechanical factors influencing da/dN

Figure 8-14 (g & h) are TMCP microstructures of the steels studied by Tsay and Chapetti. The Chapetti's [213] steel had an ultra-fine grains and the ferrite grain size was reported to be as low as 1 μm .

The microstructures of TMCP steels are greatly refined as compares to those of conventional or normalized rolled steels, resulting in a significant improvement in strength and toughness. Microstructural examination of S355J2+N [119] shows that, 80% of the microstructure is allotriomorphic ferrite and others pearlite, with traces of acicular or Widmanstätten ferrite. Hardness was measured to be around 168HV_{0.1} with equiaxed grains of average size of 15 μm , measured in the through-thickness direction. **Figure 8-15 (a)** presents the results of the FCGR studies on the NR and TMCP steels listed in **Table 8-9**, including the results of the present study presented in **Figure 8-13**. The same trend observed in **Figure 8-13** is also seen **Figure 8-15 (a)**.

The FCGR in NR are generally higher than that of TMCP in air. It can be observed that in the Paris Region taken in the interval $20 \leq \Delta K \leq 35$, the FCGR for the two microstructures separated into two domains, but with diffused inter-boundaries. The dash line is drawn arbitrarily to show possible demarcation line. Occupying upper half is the fatigue data points of NR steels and the lower half the FCGR data points for the TMCP.

One interesting note here is that even though the steels have similar micro-constituents, the FCGRs are significantly apart, especially at low ΔK . Often, many researchers associate this variation in the Paris Region with material variability or crack closure (R-factor), but we could see that the margin of separation is very much significant – up to a factor of about 3. Note that within this region the effect of crack closure is insignificant. This suggests that microstructure affected the FCGR in air. This is contrary to the current believe that microstructure has little or no effect on the FCGR in air.

8.5 Assessment of mechanical factors affecting FCGR in steel

8.5.1 Ductility (%EL) and Yield Strength (YS)

From **Figure 8-13** and **Figure 8-15**, there is no doubt that the TMCP steels with smaller grain sizes are superior in slowing down crack growth. It is pertinent to note that the microstructure of ferrite-pearlite steel controls the values of the YS, *TS* and %*EL*. Since the steels under consideration have similar microconstituents, the differing values of these parameters may be because of phase: distributions or texture, volume fraction, composition and general phase morphology. To examine the effect of microstructure on the FCGR of the steels, other factors that are believed to influence da/dN – such as frequency, stress ratio, yield stress (YS) and ductility (%*EL*), need to be assessed.

The effect of tensile strength (*TS*) on da/dN is not discussed, but it is believed to have no effect on the FCGR in air. Stress ratio (*R*) in the range 0.1- 0.85 and frequency in the range 0.005Hz – 30Hz has been found to have little or no effect on the FCGR in the ferrite-pearlite steels in air [150][153][161][167][168][169][170]. The stress ratio used in the studies presented in **Table 8-7** or **Table 8-9** is generally 0.1, and other few ratios included had similar da/dN with *R* = 0.1. This is to exclude the effect of the so-called crack closure, so that if such effect existed at all it would affect the steels equally. In fact, it has been reported that crack closure has no effect in the Paris Region. To study the effects of the remaining parameters, the mean Paris curve for the individual test in **Figure 8-15 (a)** was drawn as shown in **Figure 8-15 (b)**. And vertical lines were placed arbitrarily at about 22, 23, 26, 29 and 32 MPa√m to cut across the mean curves in **Figure 8-15 (b)**. The corresponding values of YS, %*EL* and da/dN at the point where the vertical ΔK line meets each mean curve (the dots) were determined. For example, at point (A) in **Figure 8-15 (b)** the values of YS, %*EL* and da/dN were obtained at that point, so also at point B, etc. This process was repeated for the lines at 22, 23, 26, 29 and 32 MPa√m and tabulated. Since it is known that ductility improves

the toughness of steel, the relationships da/dN vs. $\%EL$ and da/dN vs. YS were plotted as shown in **Figure 8-16 (a)** and **(b)** respectively.

In **Figure 8-16 (a)**, though we have considerable scatter, what can be seen from the curve is that retardation of crack propagation rate increases from about 26 $\%EL$ towards what appears to be an upper limit with increase in ductility.

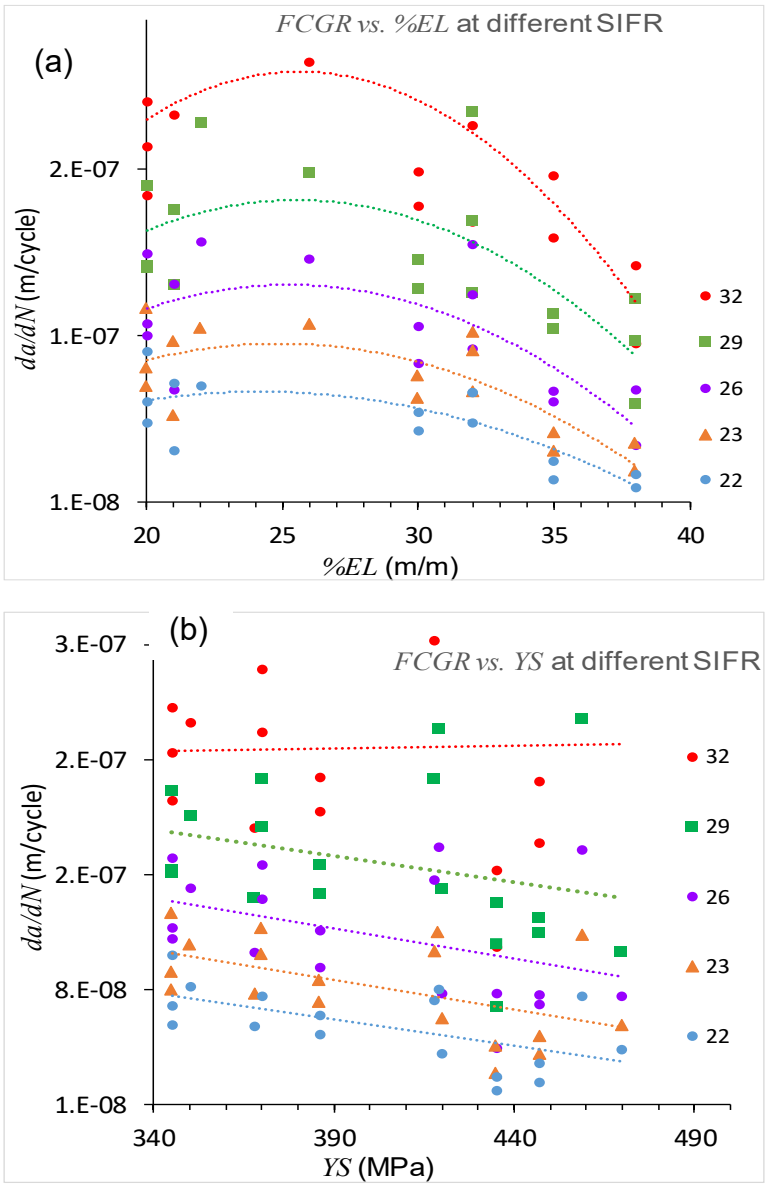


Figure 8-16: Relationship between FCGR and materials mechanical properties.

The equation describing the relationship between FCGR and ductility in ferrite-pearlite steel for a specific ΔK is of the form:

$$\frac{da}{dN} = -A\Delta^2 + B\Delta - C \quad (8-3)$$

where A, B and C are coefficients of the quadratic function and Δ is the %EL. The coefficients vary with change in the ΔK .

The crack retardation effect of ductility appears to decrease with decreasing ΔK , as shown by the legend at the right-hand side of the plot. This tends to suggest that at very low ΔK , ductility will have little or no effect on crack propagation rate. Since most of the FCGR difference between NR and TMCP steels in **Figure 8-13** and **Figure 8-15** is found at lower ΔK , ductility effect will then be very small or insignificant and cannot explain the huge difference in FCGR at low ΔK observed.

Figure 8-16 (b) shows the plot of da/dN vs. YS. Though the points are relatively scattered, we could still see a downward trend. The plot appears to show that increase in YS has retardation effect with decreasing ΔK and this effect disappeared at high ΔK as shown by the legend on the right-hand side of the plot. At 32 MPa $\sqrt{m}$, the FCGRs of the TMCP steels are still lower than that of NR. Thus, the large difference in FCGR between NR and TMCP, especially at lower ΔK can only be contributed by another factor. This factor is proposed to be a consequence of phase morphology in these steels. In other words, this considerable difference is likely to be caused by microstructural features of the ferrite-pearlite phases. Note that there is no correlation at all between FCGR and the carbon equivalent (CE).

8.6 The grain size effect – as reported in the literatures

It has been reported that grain size has no effect on the intermediate FCGRs [187][193][252][253][254]. Stonesifer [199] in his study reported that sub-critical

crack growth rate in A533 B steel is influenced by grain size through its influence on fracture toughness, K_{Ic} and the onset of rapid fracture. The value of fracture toughness K_c is usually increased by grain size reduction, other variables being constant [113]. The grain reduction increases ductility by reducing dislocation pile up stress which can initiate cleavage cracking. Some other studies on Al-Mg alloy reported that grain size affected fatigue crack growth rate (FCGR) only when the a is of the order of the microstructure [255]. In NaCl, the grain size has no effect in Al-alloy while it affected FCGR in Ti-alloy [194]. Little grain size effect was observed in Hoepfner [195] study. Weber [256] studied 70/30 brass and 305 stainless steel and reported that thermomechanical processing did not affect FCGR significantly. In fact in his work, Thompson [193] concluded unequivocally that grain size did not affect FCGR in a truly polycrystalline material. Whatever grain size effect that was seen in some other studies was attributed to slip character of the materials [253], increasing stress intensity [257] or plastic strain amplitude. It is important to note that most of the grain size effect studies were on non-ferrous materials.

To the knowledge of the authors, it appears that there is very limited study on the effect of grain size on FCGR in low alloy steels in the Paris Region. A good study on the effect of grain size on FCGR in low alloy steel at room temperature was carried out by Gejza [258] and **Figure 8-17 (a)** shows a plot of the data extracted from his study. The graph shows that FCGR decreases with increase in grain size and the effect is more pronounced in the threshold region. As the ΔK increases to say about $15 \text{ MPa}\sqrt{\text{m}}$, the grain size effect apparently diminished considerably. This tends to agree with the common conclusion that grain size effect is found mainly in the threshold region while crack propagation is independent of grain size in the Paris region. He also reported that coarse grained microstructures offered highest resistance to FCG [258].

The theory of grain size effect is commonly explained on the basis of the plastic zone ahead of the crack tip. It is often reported that at stage I of the FCGR the influence of grain size occurs when the grain boundaries act as obstacles for the

dislocations emitted from a plastic zone, with dimensions smaller than or equal to the grain size [254].

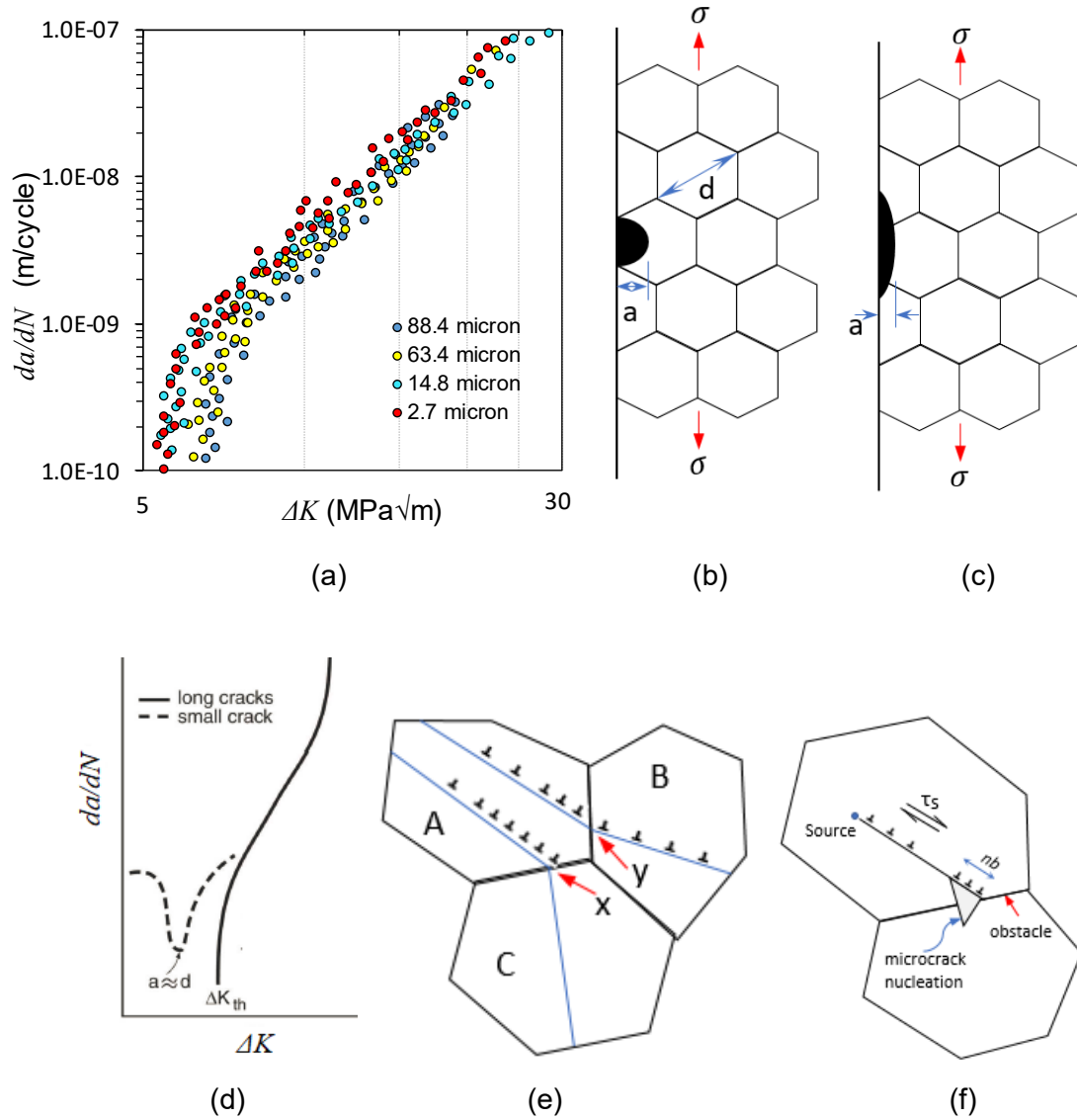


Figure 8-17: Grain size effect (a) FCGR in low alloy steel with different grain diameter (data extracted from [258]), (b) microstructural short crack, (c) physically short crack, (d) short and long cracks curves (e) dislocation movement along slip planes and pile up, [198] (f) Model of microcrack formation at a pile-up of edge dislocations (adapted from [107]) [259]

In other words, the growth of a newly nucleated crack will be strongly influenced by the grain boundary, phase boundaries of say pearlite (P) lamellae of ferrite (α) and cementite (θ) plates, precipitates and pores [260][261][262]. Because the dimension of the crack is not bigger than a grain diameter, it is often called a

microstructural short crack [198]. **Figure 8-17 (b)** illustrates the dimensions of the microstructural short crack relative to the material grain size commonly presented in the literatures.

Figure 8-17 (c) can be regarded as a physically small crack but with one dimension large compared with the grain size. The fatigue scatter in the case of **Figure 8-17 (b)** is reported to be higher than that of **Figure 8-17 (c)**. **Figure 8-17 (d)** shows the nature of curves commonly presented for short cracks in comparison with long crack. The short crack growth rate is expected to decrease within the grain and then increase rapidly when the a exceeds the size of the grain as illustrated in **Figure 8-17 (d)** – the dash line. It is commonly reported also that once the a grows more than several grain diameters, a mechanical short crack is obtained and the influence of the microstructure on FCGR disappears [198]. The crack growth then depends on the nature of plastic zone ahead of the crack tip. A physically short cracks is said to exist if the plastic zone at the crack tip is negligible and linear elastic concept applies. Note that the order - microstructurally short, physically short and mechanically short cracks are often not followed by many researchers. According to ASTM E647 [172], a short crack has length smaller than the microstructural features and the plastic zone at the crack tip is small. The microstructural effect on mechanical and physically short cracks are reported to be little and tensile crack growth mode is said to apply. The minimum rate at which the short crack will grow would now be determined by the degree of plasticity induced in a grain ahead of the advancing crack.

The degree of plasticity transmitted across the grain boundary is reported to depend on the crystallographic misorientation/misalignment between the slip planes (blue lines) on the preceding grain (A) and the grain ahead of it (B & C) as shown in **Figure 8-17 (e)**. The theory says that plastic deformation by dislocation movement occurs when critical resolved shear stress on a slip plane is exceeded in a favourably oriented grain. The emitted dislocation on reaching the grain boundary or second phase or other microstructural obstacles is constrained. If the slip plane (slip systems) misorientation (or misalignment) angle in the grain

ahead is high, as shown by blue line in A and C grains in **Figure 8-17 (e)**, the grain boundary then forms an effective obstacle to FCG and the emitted dislocations will pile up at a location indicated by arrow X and the crack growth is retarded or halted [107][198][263]. Consequently, the piled-up dislocations cause increase in stress concentration at the leading dislocation. The increase in stress is due to combination of the externally applied stress and that due to the interaction of forces in the pile up dislocations. The continued piling up of dislocations will cause the stress level at the head of the pile up to activate dislocation sources in the grain C- i.e. initiate yielding on the other side of the obstacle [264] or nucleate a crack at the barrier as shown in **Figure 8-17 (f)** [107]. That is, the shear stress acting on the slip plane pushes the dislocation together to the point that wedge crack of height nb is formed – where nb is number of Burgers vector. In a situation where the slip systems in the grain ahead of the advancing crack, are lying almost in the same plane, as it is the case for small-angle grain boundaries between grains A and B in **Figure 8-17 (e)**, the slip or dislocation is transmitted unhindered and the FCGR remains almost the same [265]. A long crack is regarded as a fully developed crack with attendant plastic-induced crack closure effect and of length usually greater than 0.5 mm [198], the plastic zone is larger than one-fiftieth of the crack length ($r > 1/50a$) and the rate of crack propagation is described by the Paris law. Short cracks and long cracks tend to describe similar da/dN vs. ΔK curve except at low ΔK .

Another important aspect of fatigue theory pertinent to the present study is on the propagation of the crack through the material crystal structures – in the Paris region. **Figure 8-18** illustrates a well-known concept of a crack traversing coarse- and fine-grained metallic structure. The macroscopic crack path in the metallic materials is commonly represented as a zig-zag line which fluctuates about the line AB for coarse-grained and A'B' for fine-grained structures as shown in **Figure 8-18 (a & b)**. The amplitude of the crack line fluctuation is presented to be smaller in fine-grained structure than in coarse grain. The SIF threshold is reported to be higher for coarse grained material than material with refined grain because of the height of the amplitude fluctuation [254][266]. Because of the longer crack path

(ACDEF) in coarse grained material, it is commonly reported that a higher ΔK is needed to traverse the path as compared with the same length of crack path in fine-grained material (i.e. A'C'T'D'E'F'G'). Thus, the threshold in coarse-grained material is expected to be higher. The theory of grain size effect on the Paris region is commonly explained on the basis that the plastic zone ahead of the crack tip is much larger than the grain size in this stage [254] as shown in **Figure 8-18 (c)**.

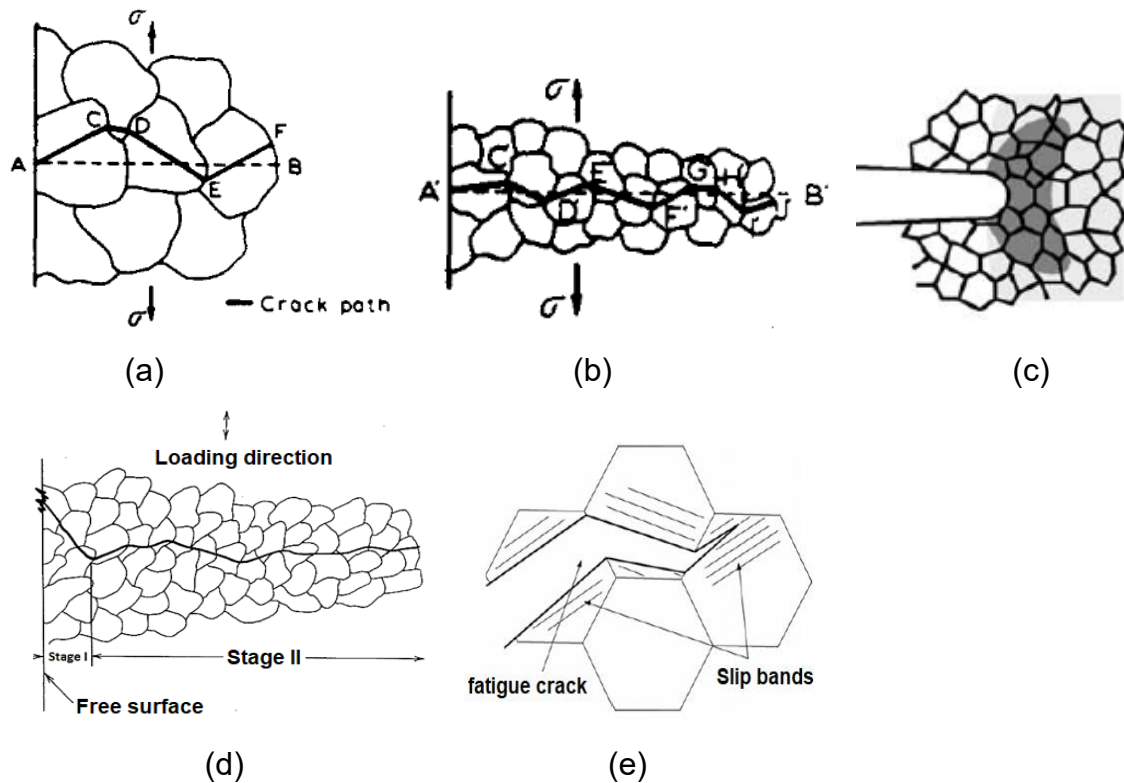


Figure 8-18: Fatigue crack path (c) coarse-grained ferrite, (d) fine-grained ferrite, (e) plastic zone extending across many grains [254]

The cyclic plastic zone is often presented as traversing several grains and the maximum crack tip opening displacement is more than few grain diameters as shown in **Figure 8-18 (c)** [213], consequently the microstructural effect is concluded to be little or negligible. This appears to be what prompted Zerbst, U. [197] in his recent review publication to maintain that microstructure has only a slight effect in the Paris region in all classes of steels. **Figure 8-18 (d & e)** present

common illustrations of the Stage I and Stage II fatigue crack growth processes from nucleation at persistent slip bands points at the free surface – often at points of microstructural inhomogeneities. The crack path in Stage II is generally taken to be across the grains (transgranular) [202]. But the question must be asked, since most of the studies used in establishing fatigue theories were carried out on non-ferrous materials, how well do these theories expressed above apply to modern marine steels? Present study have shown that contrary to this conclusion, in modern ferrite-pearlite steels, the FCGR is strongly influenced by the microstructures in the Paris Region as seen in **Figure 8-13** and **Figure 8-15**. This influence is a consequence of microstructural phase morphologies as discussed in the next sub-section.

8.7 Crack path analysis in air

Post-failure examination of the fatigued surfaces in air revealed features shown in **Figure 8-19**. The fractographs **Figure 8-19 (a & b)** were obtained for NR steel (J2+N) in the current study, while that of **Figure 8-19 (c & d)** were obtained by Adedipe [231] for the same steel grade. **Figure 8-19 (e)** is the fractograph obtained in air for the TMCP steel. The part in a red box in **Figure 8-19 (e)** was magnified as shown in **Figure 8-19 (f)** and the area in the red box in **Figure 8-19 (g)** was magnified to that shown in **Figure 8-19 (h)**.

From all the fractographs, it is very obvious that the failure in air had occurred by ductile striation mechanism with secondary crackings. Few secondary crackings in the normalised steel, J2N were observed. The occurrence of the secondary cracking increased with increase in ΔK . Striations were seen even inside the secondary cracks. This shows that the groove was made by a moving crack front and not due to brittle cracking or cleavage. We will see later that these secondary cracking seen on fractographs was formed by part of branched cracks that stopped growing or have been arrested.

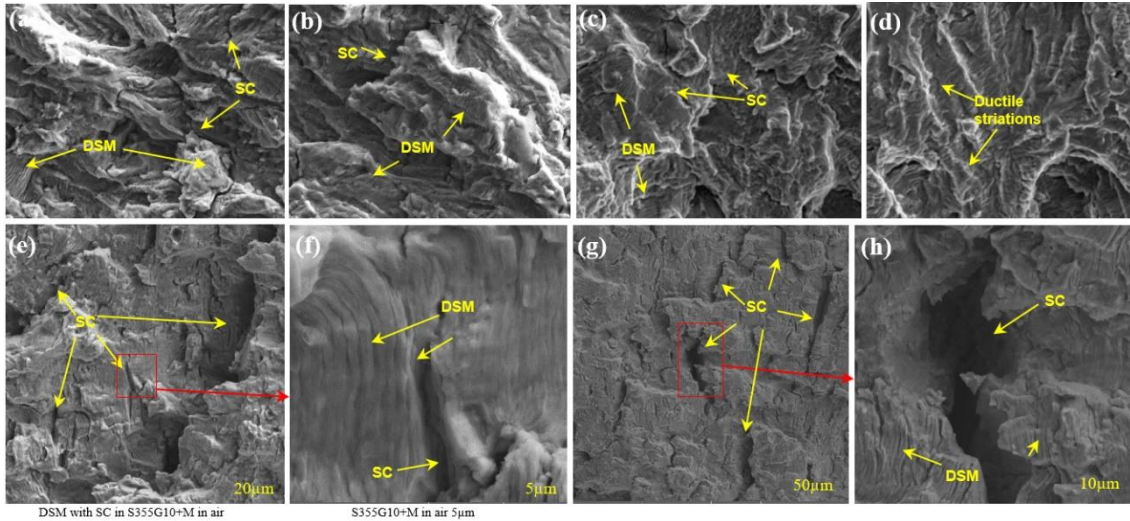
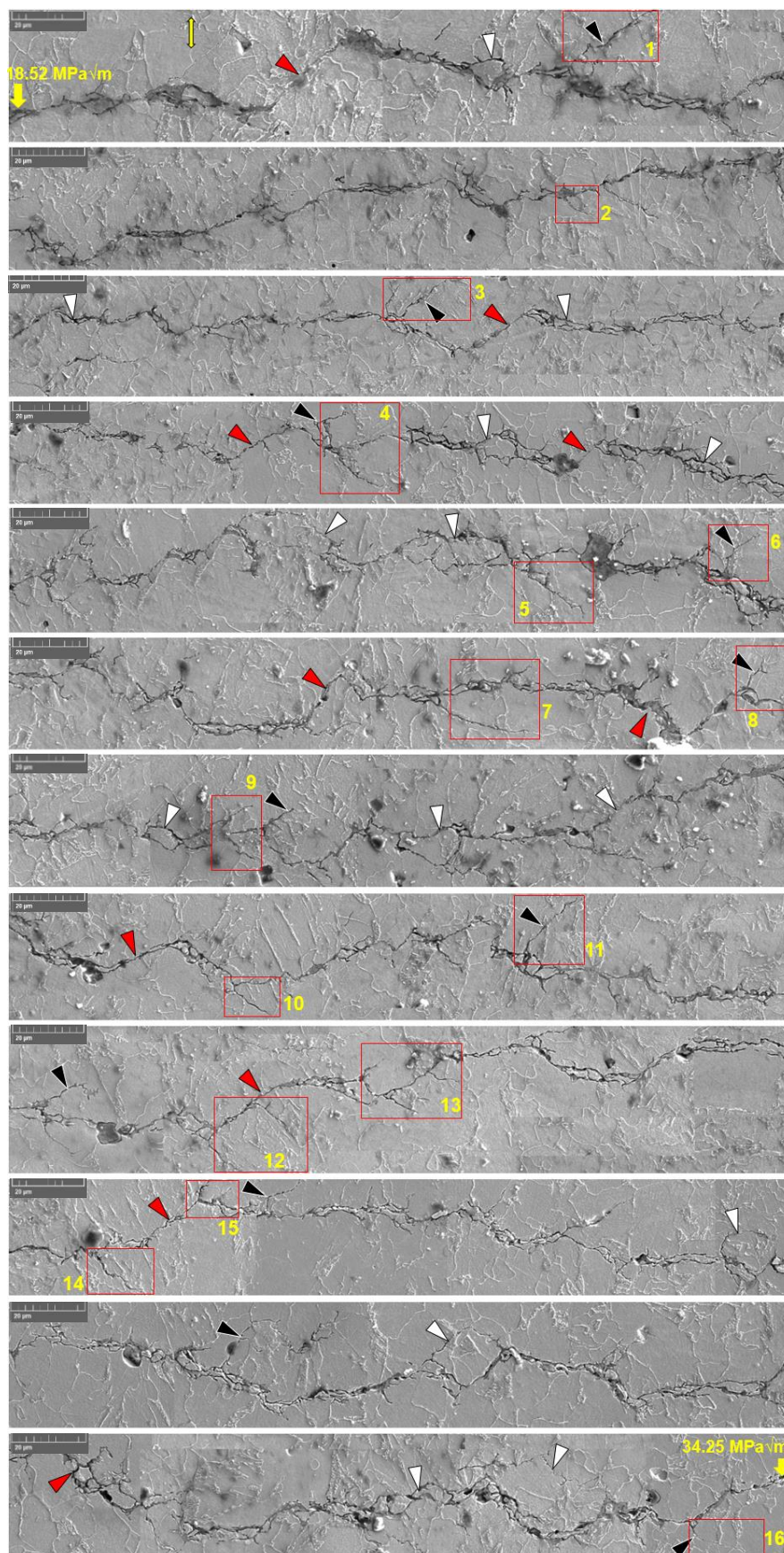


Figure 8-19: Fractographs of fatigued surfaces (a-d) for J2+N (e-h) for the G10+M in air (Ductile striation mechanism (DSM) with secondary crack (SC))

The ductile striation mechanism as seen in the fractograph is typical of the failure mechanism in the Paris Region for low carbon steel. These fractographs do not give significant clue to the observed difference in the FCGR in **Figure 8-13** and **Figure 8-15**. An alternative approach is to examine the crack path and the features influencing crack growth in the materials.

Figure 8-20 (a) shows the general crack paths seen in the TMCP steels used in this study in air for 10kN, 5Hz, stress ratio of 0.1, under sinewave. The yellow double line arrow in all the figures shows the direction of the applied fatigue load. The long crack path was obtained for ΔK of about 18.52 to 34.25 MPa $\sqrt{m}$, i.e. fully in the Paris region. It is very clear that the crack path is non-planar, complex with extensive high degree of large angle crack diversion as shown by the red arrows and bifurcation as shown by black arrows. The arm of the branched crack is considerably long, and it was found to generally follow the thin ribbon ferrite phase. Another phenomenon that is excessively prominent is the formation of metal crumbs in between branched and re-joined crack fronts as shown for some instances by the white arrows in **Figure 8-20 (a)**.



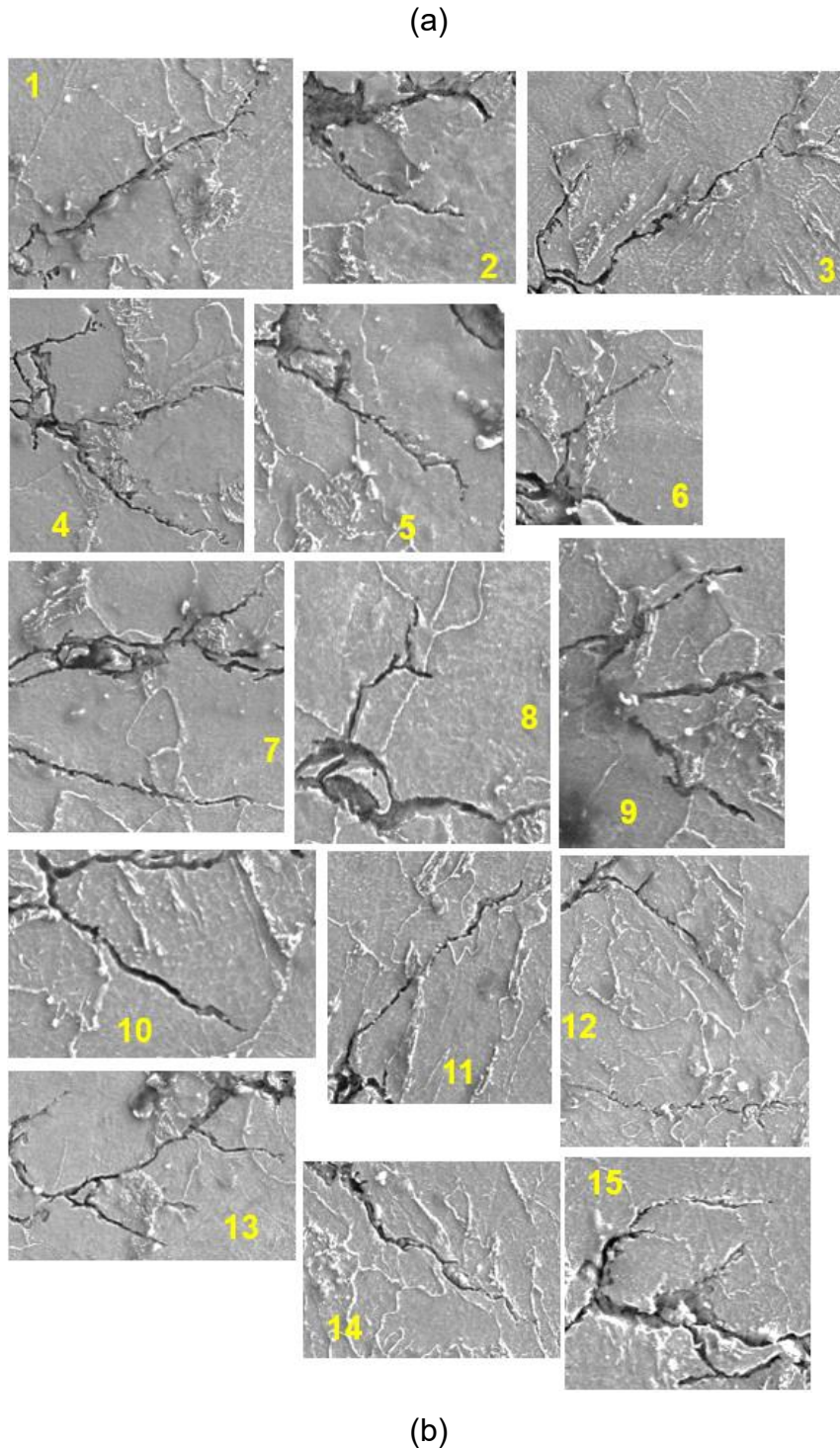


Figure 8-20: Crack path in TMCP steels in air: G8, 10kN, 5Hz at a scale of 20 μm , crack length between about 20.33 mm (18.52 $\text{MPa}\sqrt{\text{m}}$) to 30.09 mm (34.25 $\text{MPa}\sqrt{\text{m}}$) (direction: from right to left) (scale = 20 μm)

It can also be observed that severally small metal crumbs were formed as the crack propagated in air and the occurrence increased with increase in ΔK . In other

words, the tendency for crack branching and re-joining tend to increase with increase in the driving force for crack propagation. An important feature of the crack growth is that generally the tip of the propagating crack was always sharp as can be seen from the tip of the branched crack front as shown in the magnified micrographs in **Figure 8-20 (b)**. The crack tip maintained its sharpness throughout the Paris region in the range 18.52 to 34.25 MPa $\sqrt{m}$.

The fractographs of fatigued surfaces as shown in **Figure 8-19** can only show mechanism of the growth but cannot show these phenomena found in **Figure 8-20** - specially the formation of the metal crumbs, which has not been reported in any literature. Due to many branched and arrested cracks, it is expected that the fractography of the fractured surface for this sample will likely show many occurrences of what is referred to as secondary cracks. This is evident in **Figure 8-19 (a to c)** where many secondary cracks are present.

Figure 8-19 shows the crack path observed in NR steel in this study in air. A low-angle crack diversion and branching angles were seen. The red arrows show the low angle crack diversion, the black arrows the low angle crack branching and the white arrows show some metal bits or crumbs. The formation of several small metal crumbs which is a prominent feature of the crack growth in the TMCP in air is almost absent in the J2N – only very few crumbs were found. The amount of the branched crack front is very small, and the extension of the branched arm is very short compared to that of air. Clearly, the crack branching (with long arms) and deviations are severe in the TMCP steel.

The metal crumb has been formed in-between the branched and re-joined crack fronts. The crack deviation, branching and creation of metal crumbs can reduce the effective stress at the local crack tips resulting in retardation of the FCGR. The longer the branched arm, the more the crack growth is retarded due to re-distribution and reduction of the driving force at the main crack tip. Where high angle crack bifurcation and diversion are absent, or the two phenomena and formation of metal crumbs are few, one would expect a high FCGR.

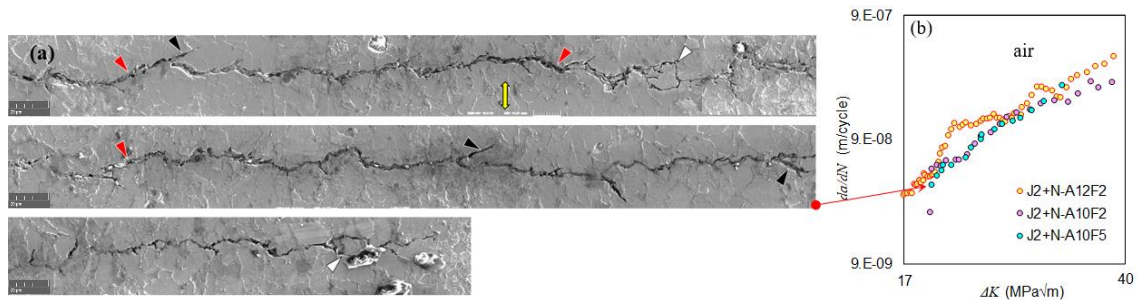


Figure 8-21: FCG path in J2+N steel in air (a) Crack path in 10kN, 5Hz at a scale of 20 μ m (b) FCGR in the J2+N steels (scale = 20 μ m)

Fundamentally, the degree of crack tip stress shielding will vary with the angle of crack deviation and the nature of metal crumbs created. The metal bits will be effective in retarding the crack growth if it re-orientes and causes mismatch or wedging action. The tendency to re-orient will increase if the crumbs are smaller in rough-round shape rather than in plate-elongated form. These two natures of metal crumbs will be a function of the steel microstructure. The long red arrow points to the specific curve that has the crack path shown in **Figure 8-21 (b)**. **Figure 8-21 (b)** shows that the FCGRs at 10kN, 2Hz (J2+N-A10F2) and that of 10kN, 5Hz (J2+N-A10F5) are generally the same. These tests show that frequency has no effect on the FCGR in air. The FCGR of the 12kN, 2Hz (J2+N-A12F2) is higher, of course due to increase in the driving force.

In comparison with the crack growth path in J2+N, the crack path in the TMCP is more complex with extensive high degree of large angle crack diversion indicated by the red arrows and high angle crack branching as shown by black arrows. The lengths of the branched cracks are considerably longer than that of J2+N and it was found to generally follow the thin ferrite ribbon, RB , α_{HA} . There was severe formation of relatively large metal crumbs – i.e. metal bits in between branched and re-joined crack fronts in air as shown by the white arrows in **Figure 8-20** and **Figure 8-21**. Severally small metal crumbs were also seen as the ΔK increased. An important feature of the crack growth is that generally the tips of the

propagating cracks in the TMCP steel in air are always sharp. The crack tips were found to maintain their sharpness throughout the Paris region in the range 18.52 to 34.25 MPa $\sqrt{m}$ (see **Figure 8-20**).

A plot of the FCGRs in air of the two microstructures of the S355 steel subgrades is given in **Figure 8-13**. As noted, the test in air were performed at 12kN, 10kN, 9kN, and 2Hz, 3Hz and 5Hz for the three materials - J2N, G8 and G10. Note that the microstructures of G8 and G10 are generally the same and that the fatigue tests were carried out under the same mechanical conditions. As already noted, High angle crack bifurcation and diversion are generally absent, and the formation of metal crumbs are few in the NR steel. It can be seen from this figure that the crack path having low angle deviation, low angle crack branching, short branched arm and less complex crack front (observed in J2N steels) had the highest FCGRs. In fact, it appears that there is a demarcation between the two microstructures as represented by the arbitrary dash line. This implies that the longer the branched arms, the more the deviation and branching angles the more the crack growth is retarded due to re-distribution and reduction of the driving force at the main crack tip.

Fatigued surface features in form of fractographs are commonly presented when mechanism of FCG is discussed. And the mechanism of FCG in air for low alloy steels is found to be the same - ductile striation. Hence, it is commonly reported that microstructure has little or no influence on the FCGR curve in the Paris region [29][188][197] . Contrary to this conclusion, the current study clearly shows that the nature of the microstructures of the ferrite-pearlite steels affected strongly the FCGR in the Paris region of the da/dN vs. ΔK sigmoidal curve in air.

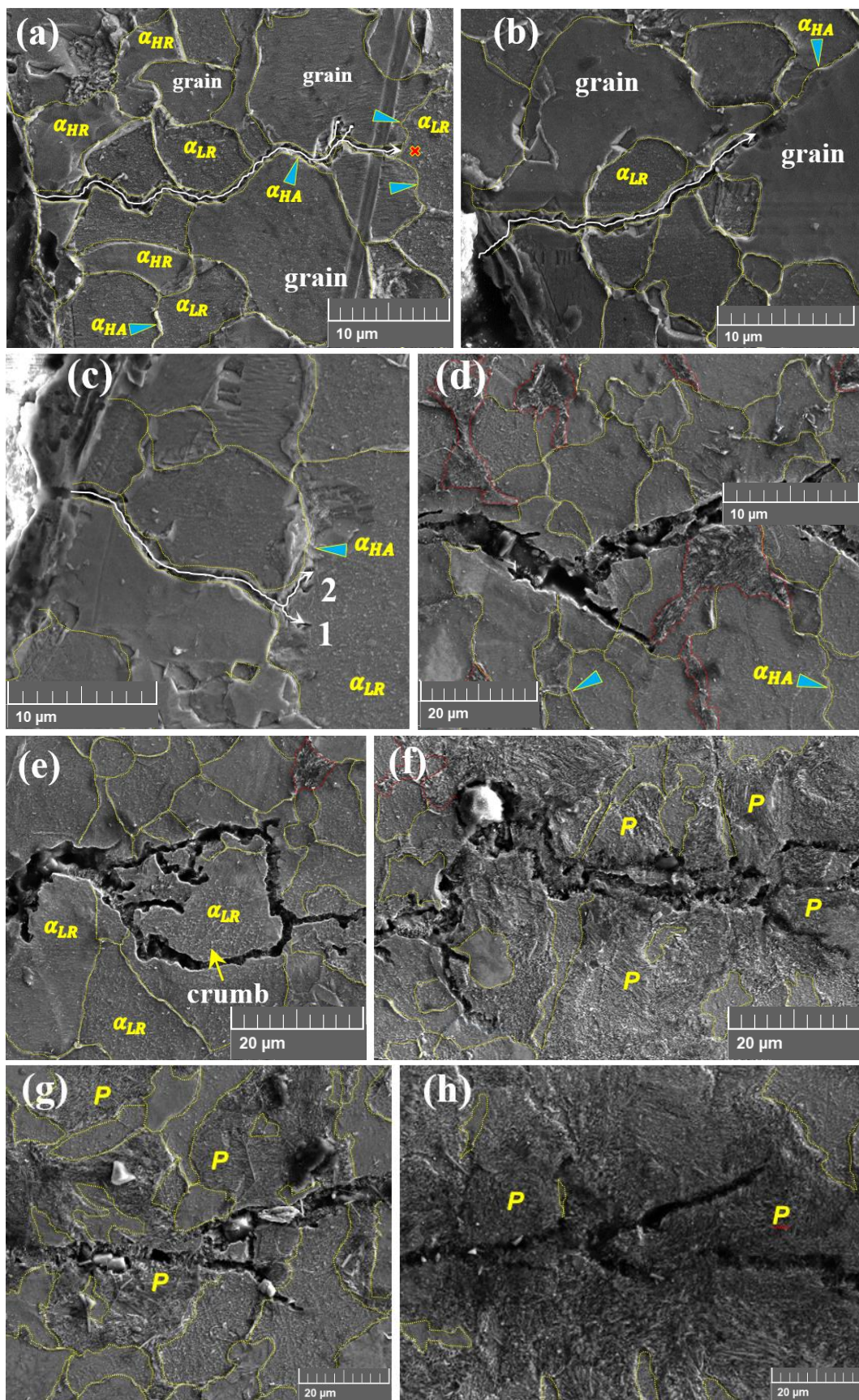
8.8 Microstructural phases influencing FCG in air

From the previous discussions, the microstructure influenced the FCGR by crack path diversion, branching and formation of metal crumbs in-between branched and re-joined crack fronts. High FCG retardation was obtained when large angle

diversion, large angle branching and change in the orientation of metal crumbs occur. This implies that the material microstructure influenced the crack growth in steel in the Paris Region.

The phases responsible for this disparity in air were examined by scanning along the entire length of the fatigue crack from the initiation point to point of test stoppage. Some of the distinct micrographs showing phases and morphologies influencing FCGR are presented in **Figure 8-22** for the NR steel and **Figure 8-23** for the TMCP steel. As already presented, ferrite with HR is denoted α_{HR} while low relief is α_{LR} . The P denotes pearlite. **Figure 8-22 (a), (b) & (c)** shows instances of the propagation of the crack that initiated from the CT specimen notch of length 20.50 mm. In current theory, a long crack is regarded as a fully developed crack with attendant plastic-induced crack closure effect and of length usually greater than 0.5 mm [198], the plastic zone is larger than one-fiftieth of the crack length ($r > 1/50a$) and the rate of crack propagation is described by the Paris law. At a length of 20.50 mm, the crack shown in **Figure 8-22** is a very long one.

The dotted yellow line shows the boundaries of the grains. The encircled spaces by the yellow dotted lines are the ferrite grains while the red dotted lines encircled the pearlite nodules. The white solid line shows the movement of the fatigue crack through the microstructural phases. If the pointed end is a dot, the crack has stopped growing while an 'arrow indicates the mean active crack front. In **Figure 8-22 (a, b & c)**, the crack appears to have initiated from and then propagated along what appears as the grain boundaries or 'intergranularly'. A closer observation shows that the crack propagated through the thin ribbon of the high alloy ferrite phase, α_{HA} , that surrounds the α_{LR} grain surfaces. This mode therefore is properly referred here as quasi-intergranular cracking mode. At the point marked X in **Figure 8-22 (a)**, the thin ferrite phase ribbons, denoted by the blue triangular arrows is not favourably aligned to the crack front – i.e., they are aligned at a very high angle, towards the vertical, to the crack front. In this case, the crack will have to advance transgranularly through the grain α_{LR} in front of it.



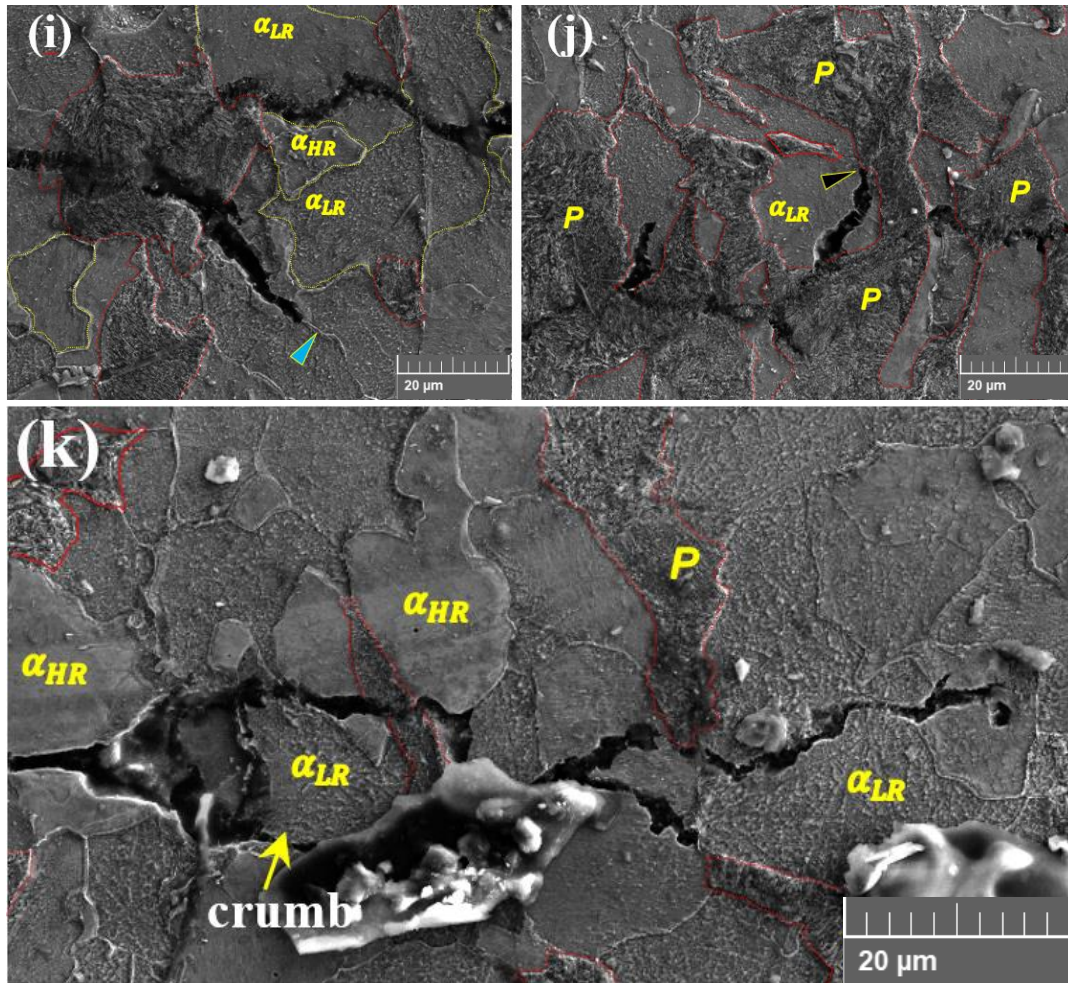


Figure 8-22: Effect of microstructure on FCGR in NR steel in air, for 10kN, 5Hz

In **Figure 8-22 (b)**, the phase ribbon is aligned favourably or lies at low angle to the crack front, hence, the crack advanced following this ferrite ribbon. The situation similar to **Figure 8-22 (a)** is seen in **Figure 8-22 (c)**. Here, it appears that the grain of α_{LR} in front of the advancing crack front resisted transgranular propagation of the crack in the direction 1 and this caused the sprouting of another crack front that attempted to follow the α_{HA}/α_{LR} boundary shown by the blue arrow in direction 2. This resulted to crack bifurcation and tends to suggest that the α_{LR}/α_{HR} boundary is the preferable path for the crack growth, especially when the boundary is lying at low angle to the crack plane. Note that in **Figure 8-22 (a, b & c)**, the ferrite phase dominated the area with no pearlite phase.

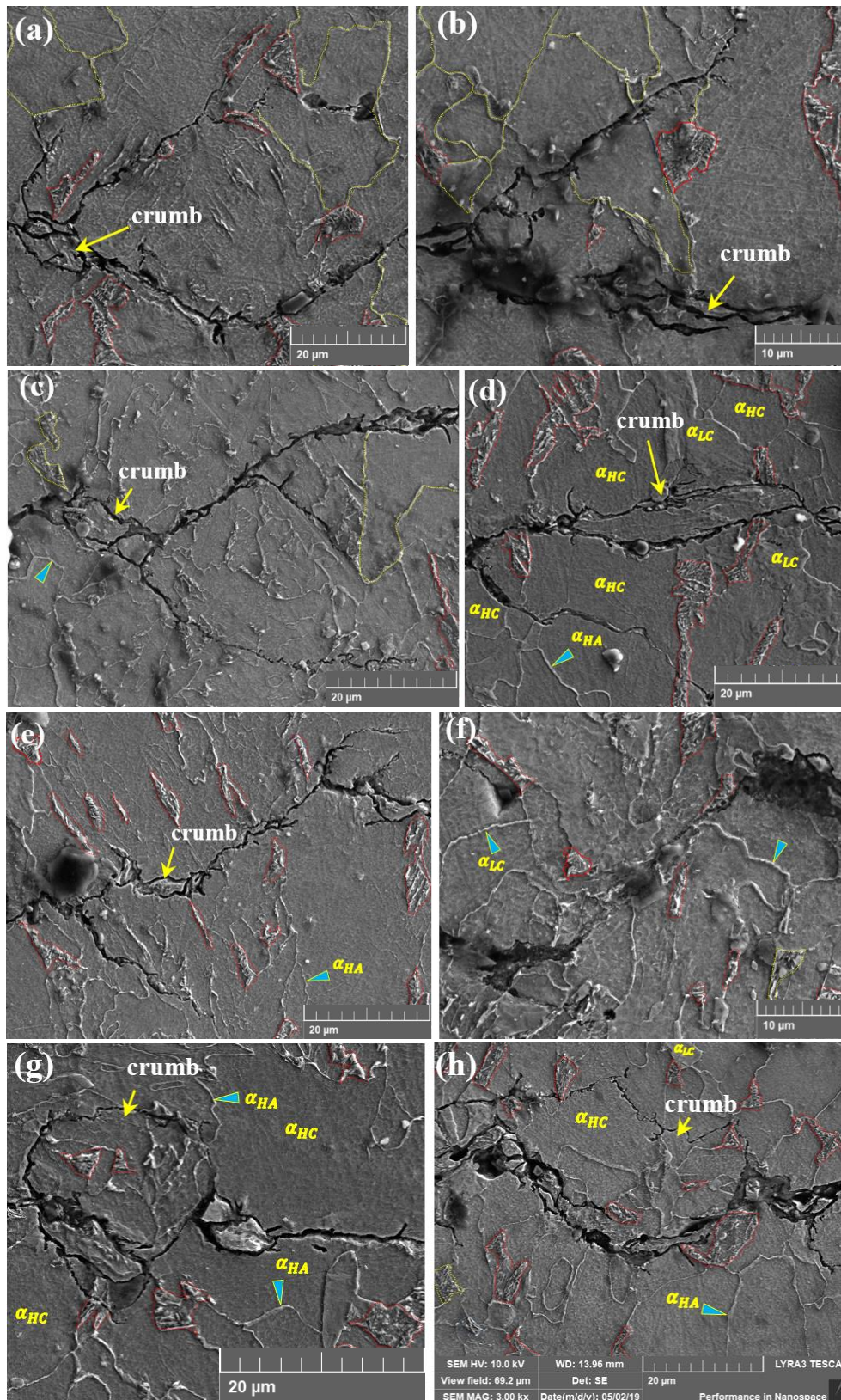
As the ΔK increased and the crack moved into areas of mixed ferrite and pearlite phases as in **Figure 8-22 (d to k)**, it appears that the phases are equally likely to cause crack deviation, crumb formation and crack branching. **Figure 8-22 (d)** shows intragranular crack growth and bifurcations. This region is dominated by the α_{LR} phase and some few pearlite phases. The α_{HR} phase and the α_{HA} are almost absent in this region. This may explain why the crack propagated transgranularly with diversion which appears perhaps as a consequence of the crack growth mechanism. **Figure 8-22 (e)** shows the formation of metal crumbs in between branched and re-joined crack fronts. This microstructural region is also dominated by the α_{LR} . In **Figure 8-22 (f to J)** show multiple crack branching in the pearlite phase, especially in **Figure 8-22 (f)**. The tendency for crack branching in the pearlite phase may be attributed to the orientation of the pearlite colonies and α_{HA} boundaries. In the first case, this tendency may be due to the fact that pearlite has two alternating phases – ferrite, α (iron, Fe) and cementite, θ (Fe_3C) [267][118] with different physical and chemical properties. Daeubler [247] studied the influence of microstructure on the surface FCG behaviour of pearlitic steels and presented a classical case where the pearlite colony orientation influenced the crack growth, especially when the lamellae are favourably oriented to the load axis. In the second case, it was observed in this study that α_{LR}/α_{HA} or and α_{HA}/P boundaries are preferable paths of fatigue crack growth. The blue arrow in **Figure 8-22 (i)** shows a branched crack that followed the thin ferrite ribbon, α_{HA} and the black arrow in **Figure 8-22 (j)** shows what appears to be a branched crack front following α_{LR}/P boundary. **Figure 8-22 (k)** shows another metal crumb that formed along the a in the vicinity of mainly α_{HR} and α_{LR} phases. The formation of the metal crumb appears to occur in the α_{HR} and α_{LR} phase fields as shown in **Figure 8-22 (e & k)**.

To understand the FCGR difference between NR and TMCP steels, a closer examination of the crack path was made as presented in **Figure 8-23** for G8+M, test in air, under 10kN, 5Hz stress ratio of 0.1 and sinewave. The a measured about 10mm obtained under ΔK of about 18.50 to 34.25 MPa $\sqrt{m}$. The number of high angle deviation and crack branching is by far more in the TMCP steel as

compared with low angle crack deviation and crack branching that was consistently found in NR as presented in **Figure 8-22**. Measurements on the angle of crack branching in NR steel show that they are generally below 45° . The only time high angle deviation is found in NR steel to be a little bit more than 45° is when the crack is avoiding α_{LR} phase or following the ferrite ribbon or α/P boundaries that is favourably oriented in its path. The ferrite ribbon is shown in **Figure 8-23** by the red triangular arrow. The space enclosed by the red dotted line is the pearlite phase while the yellow dotted line is the ferrite phase. Multiple formations of metal crumbs are also shown. These micrographs show that the α_{LR} phases dominated the microstructure in this region with extensive formation of the ferrite ribbons, α_{HA} . Many of the ferrite ribbons appear as sub-grain phases inside the α_{LR} phase grains. The presence of α_{HR} is very small. The pearlite nodules are very small also, even some nodules are purely single colonies. The crack path in the TMCP steel is completely non-planar and more complex than that of NR steel.

It can be seen that the crack deviation and branching in **Figure 8-23 (a to i)** are caused by the crack tip following path of least resistant created by the α_{HA} or the α_{LR}/α_{HR} or α/P boundaries - but mostly by the α_{HA} ribbons. The ability of the α_{HA} to deviate crack front appears to lie in its carbon or alloying composition relative to the surrounding phases. It must be mentioned that the ability of the α/P interface to divert the crack front is dependent on the morphology of the pearlite.

The crack bifurcation is quite extensive with long arms as can be seen in **Figure 8-23 (a to e)** and the deviation of the main active crack front as shown in **Figure 8-23 (f)** was measured to be about 43° to the horizontal plane. The branched arms are measured to be between 38° and 46° . It is interesting to see that the path of least resistant caused many of the branched arms to deviate almost at an angle within this range. This simply implies that both tensile and shear stresses acted on the crack as it propagated in the TMCP.



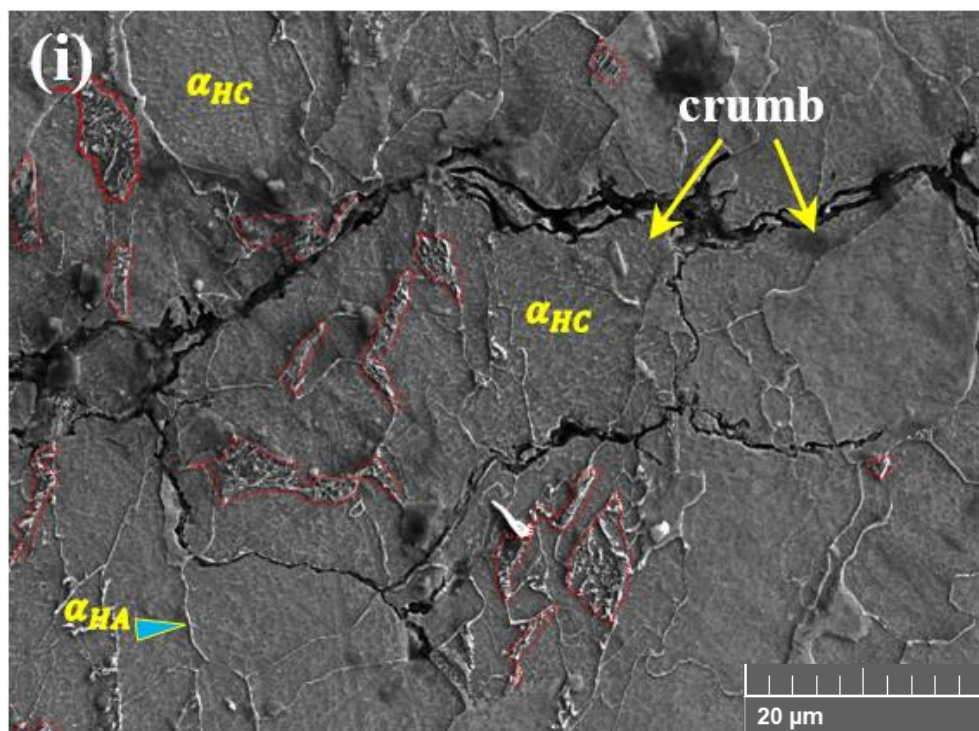


Figure 8-23: Fatigue crack growth in TMCP, G8+M, 10kN, 5Hz in air

The metal crumbs as noted in **Figure 8-23 (d, e, g, h & i)** are by far more in the TMCP steels. It was observed also that the α_{HA} does not only have the potency to divert and branch out crack front, it can as well facilitate re-joining of branched crack fronts into one main crack front resulting to the crumb formation. *This crumb is found to be a tore-off of the steel grain or block of grains. In other words, the fatigue crack front in ferrite pearlite steel does not have a horizontal plane edge, but rather series of fluctuating wavelike crack front, whose nature depends on the steel grains and the phases surrounding the grain surfaces.* It is important to note that this metal crumb along the crack path in the Paris Region has not been reported elsewhere in the literatures or considered in any theoretical formation or treatment.

The pearlite colonies in the NR steel are large, blocky and dense. In the TMCP steels, the pearlite appears to grow in two forms – small-blocky colony and a thin, elongated or needle-like morphology colony along what may be described as the ferrite grain boundaries as shown by the space enclosed by the red dotted line in

Figure 8-23. The effect of thermomechanical control rolling process appears to be the disruption of the pearlite banding and blocky morphology into randomly distributed thin pearlite needles. Thus, the interface between the ferrite and the pearlite becomes an additional effective preferable path for the crack growth if the pearlite nodule is elongated and aligns favourably to the crack front. In **Figure 8-23**, the branched cracks did not just move about linearly but tends to trace the ferrite ribbons or the α/P interface. This implies that the energy needed to propagate through the α_{LR}/α_{HR} or α/P interface is much less than that for traversing the α_{LR} grain. The influence of the phase morphology in the Paris Region persisted to a length of 9mm and at 32 MPa $\sqrt{m}$.

The current theory says that the crack path in the Paris Region or Stage II in the da/dN vs. ΔK sigmoidal curve is generally taken to be across the grains transgranular [202] and the material microstructure has little effect in this region [29][187][197][198]. The current study found in the contrary that microstructure of a ferrite-pearlite steel has a strong influence on the rate at which fatigue crack grows in the material as evidenced in the above discussions. Again, both transgranular and intergranular modes of propagation were observed in air, and what appears as intergranular cracking mode predominated in the TMCP steel. Though what appears as an intergranular mode is actually the growth of the crack through a thin layer of α_{HA} adjacent to the α_{LR} or the P . Hence this mode is identified here as the quasi-intergranular mode. The morphology and chemistry of the material phases and load level appear to determine which mode the crack growth will adopt. Another factor is that the angle the crack front makes with the phases ahead of it tends to determine if it would propagate in transgranular or quasi-intergranular mode. However, in **Figure 8-19 (b)** ductile striations was found inside the secondary crack – which actually appears as a branched crack when viewed from the crack line. This clearly shows that the mode is not strictly intergranular, but the crack moved in the α_{HR} phase ribbon immediately adjacent to the α_{LR} or the pearlite interface, justifying the use of word - quasi-intergranular. In general, TMCP steel is more resistant to crack propagation than NR steel of the same steel grade. This is made possible because the refined-microstructure

of the TMCP branched (with long arms) and diverted crack front at large angles as compared to the small angle path diversion found in the NR steel.

8.9 FCGR in ferrite-pearlite steels in seawater

A similar study described in section 8.7 above was carried out in SW. The authors assessed the FCG resistance of the NR and the TMCP steels given in **Table 8-2** in SW. The details of the experimental procedures have been discussed in Refs [230][144] and in Chapter 4. The experimental results are shown in **Figure 8-24 (a)**.

Three CFCGR curves, obtained at 12kN, 9kN and 0.3Hz, for the S355J2+N studied by Adedipe [149] (abbrev. 'Ade') were plotted from the data extracted directly from a vs. N curves given in Ref [149]. In the current study, one test on the S355J2+N was performed at 9kN, 0.3Hz, $R = 0.1$ to validate the results obtained in the Adedipe's study. The Adedipe's study [149] proved to be highly reliable. Mehmanparast [175] studied CFCG in G8+M TMCP steel under the same loading conditions and at three different locations noted as (A3, A4), (B3, B4) and C3.

In identifying the experimental samples in this study, the prefix gives the steel subgrade – J2N (or J2+N), G8 (or G8+M), G10 (or G10+M), followed by S (S = sinewave), P_{max} (maximum load), F (frequency, followed by the frequency value) and the last is the test centre for the Mehmanparast [175] study. All the tests were performed in SW and under sinewave. **Figure 8-24 (a)** suggests that there is a general boundary for the FCGR in SW between NR and TMCP in the Paris Region. The dash line is an imaginary line placed to show the possible line of demarcation. The CFCGR for the NR steel is higher than that of the TMCP. The result of the NR steel tends to show that decreasing the load level has more accelerating effect than increasing the fatigue load as shown by the 9kN and 12kN under 0.3Hz.

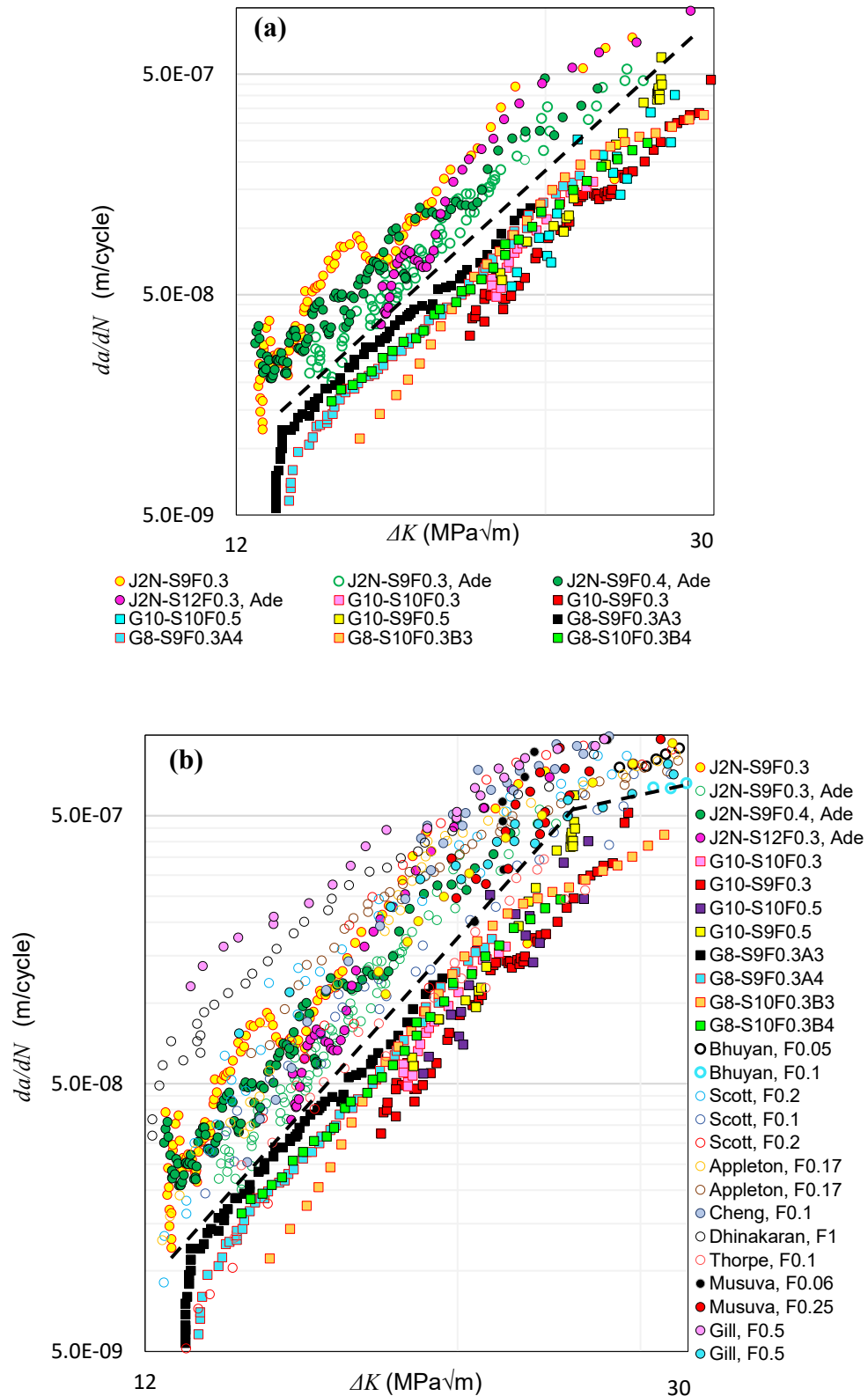


Figure 8-24: Plot of FCGR in seawater (a) NR and TMCP steels in the present study (b) CFCGR curves of ferrite-pearlite steels reported in the literatures

Selected results for the G8+M steel studied by Mehmanparast [175] show that increasing the fatigue load level to 10kN at 0.3Hz did not necessarily increase the CFCGR above that of 9kN. In fact, below about 18.50 MPa√m, the CFCGR of the 10kN decreased below that of 9kN. For the G10+M steels, two load levels – 9kN, 10kN and two frequencies 0.3Hz, 0.5Hz were used. At about 20 MPa√m and above, there is no obvious difference in the CFCGR for the test conditions. In general, the CFCGRs of G10+M for all the test conditions were slightly lower than that of G8+M and the CFCGRs of all the TMCP subgrades were lower than that of J2+N steel. In fact, the dash line is an imaginary line placed to show the possible line of demarcation between the CFCGRs of the NR and TMCP S355 steel subgrades.

In a similar way to air study, the trend found in **Figure 8-24 (a)** is validated by carrying out a comparative study of CFCGRs between NR and TMCP ferrite-pearlite steels reported in the literatures. The steel grades, chemical composition in wt% (approximated to two decimal places) and mechanical properties of the steels compared with the CFCGR data in this study are shown in **Table 8-10**. It must be mentioned that there are limited studies on the corrosion-fatigue of marine steels under the conditions listed in **Table 8-10**. **Figure 8-24 (b)** presents the CFCGR studies on the NR and TMCP steels listed in **Table 8-10**.

A large difference in FGCR between the NR and TMCP steels is once again observed for the tests performed in *SW*. It can be observed from **Table 8-10** and **Figure 8-24 (b)** that even some studies carried out at very low stress ratios such as 0.08 by Musuva did not reduced CFCGR in **Figure 8-24 (b)**. It is often reported in the literatures that plasticity-induced, roughness-induced and oxide-induced crack closure can cause crack growth retardation. However, in *SW*, it is important to note that the plasticity and the roughness/asperities of the fractured surfaces are areas of high energy. Hence, if these areas existed, they would corrode very fast if in contact with corrosive environment and would not contribute to crack retardation. The only factor that may cause retardation is oxide-induced crack closure and the extent will depend on the flow rate of the testing medium.

Table 8-10: E Experimental conditions for the CFCGR data used in this study

Study by	Steel grade	Wave form	Env.	Freq. (Hz)	R	%EL	YS (MPa)	UTS (Mpa)
Bhuyan	CSA G40.21M 350 WT	s	SW	0.05	0.1	29	405	513
Scott	BS4360-50D	p-s	SW	0.1	0.1	29	405	513
		p-s	SW	0.2	0.5	32	370	538
		t	SW	0.1	0.5	32	370	538
		n-s	SW	0.2	0.5	32	370	538
Appleton	BS4360-50D	s	SW	0.17	0.5	32	368	508
		t	SW	0.17	0.5	32	368	508
Cheng	ABS grade EH36	s	SW	0.1	0.1	21	350	-
Dhinakaran	C-Mn steel	s	NaCl	1.0	0.5	32	459	530
Thorpe	BS4360-50D	s	SW	0.1	0.0, 0.1	32	370	538
Adedipe	S355J2N	s	SW	0.3, 0.4	0.1	20	385	531
Musuva	BS4360-50D	s	3.5%NaCl	0.25	0.08	30	386	560
		s	3.5%NaCl	0.06	0.08	30	386	560
Gill	HSLA (MF-80)	s	3.5%NaCl	0.5	0.67	-	653	742
		s	3.5%NaCl	0.5	0.1	-	653	742
Mehmanparast	S355G8+M	s	SW	0.3	0.1	35	447	549
PS	S355G10+M	s	SW	0.3 & 0.5	0.1	36	447	545
	S355J2N	s	SW	0.3	0.1	20	345	531

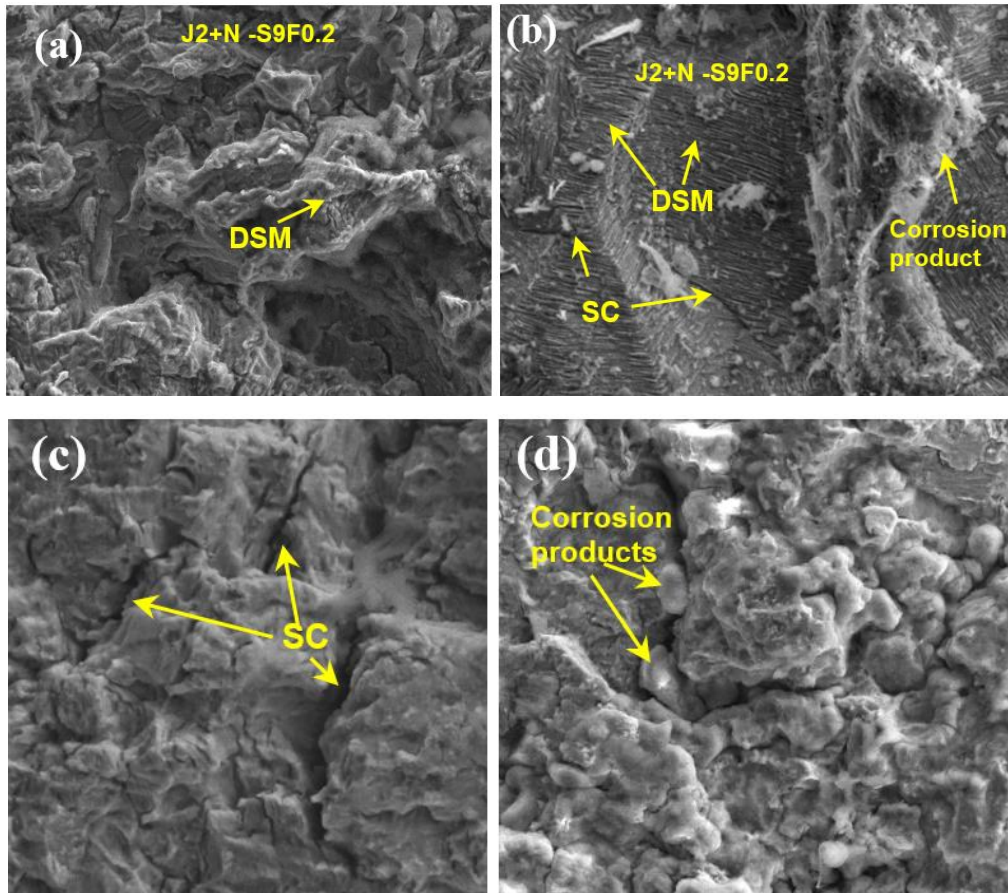
s = sine, t = triangular, p-s = positive sawtooth, n-s = negative sawtooth, SW = seawater

8.10 Crack path analysis in TMCP steel in seawater under sinewave

A metallographic study was carried out to further assess the cause of the disparity in **Figure 8-24 (a)**. The fractographs for the NR and TMCP steels in SW are shown in **Figure 8-25**. **Figure 8-25 (a)** is the fractograph obtained in the present study for J2+N and **Figure 8-25 (b)** is for the same steel obtained by Adedipe [231]. **Figure 8-25 (c)** is the corrosion-fatigued surface of the G10+M obtained at about 20MPa√m and **Figure 8-25 (d)** is the common feature of the corrosion-fatigued surface of the same material above about 23 MPa√m.

Generally, cleavage cracking was not observed in SW. The failure mode remains by ductile striation is typical of the failure mechanism in the Paris region for low carbon steel. The striations in SW simply appeared to be washed out by the corrosion process and the surfaces were covered by the corrosion products. This washing out progressed with time as can be seen from (c) to (d) in **Figure 8-25**. The secondary cracks in the TMCP steel were larger than that of NR and the

corrosion products appear to collect into these secondary cracks over time as seen in **Figure 8-25 (d)**.



DSM and no cleavage in S355G10+M in SW

Figure 8-25: Fractographs of fatigued surfaces in seawater (a & b) for J2+N (c & d) for the G10+M (Ductile striation mechanism (DSM) with secondary crack (SC) and no cleavage mechanism)

The corrosion products into the cracks has the potential of reducing the crack driving force, hence, the rate at which the crack is moving. However, the flow of test environment and the pumping of the SW in and out of the crack tip as a result of the cyclic action – especially under sinewave, has the capacity of removing some corrosion products at the crack tip, minimizing the oxide-induced retardation. Since there was no cleavage cracking in the J2+N, retardation due to the corrosion product or the oxide-induced crack closure is not sufficient to account for the large difference in CFCGR in **Figure 8-24 (a)**.

Note that we have already mentioned that roughness or plastically deformed areas of fatigued surface, which leads to crack closure – especially for ductile materials cannot contribute to crack reduction in aggressive environment. The surface roughness is caused by ductile stretching or tearing of the fractured surface to form asperities. As already pointed out, roughness/asperities and plastically deformed parts are areas of high energy. In corrosive environment, these areas will be eliminated rapidly if they existed leaving a smooth surface or deep contour in the case of plastically deformed areas. Hence, they would not contribute to crack retardation. An alternative approach again is to examine the crack path and the microstructural features influencing crack growth in the SW.

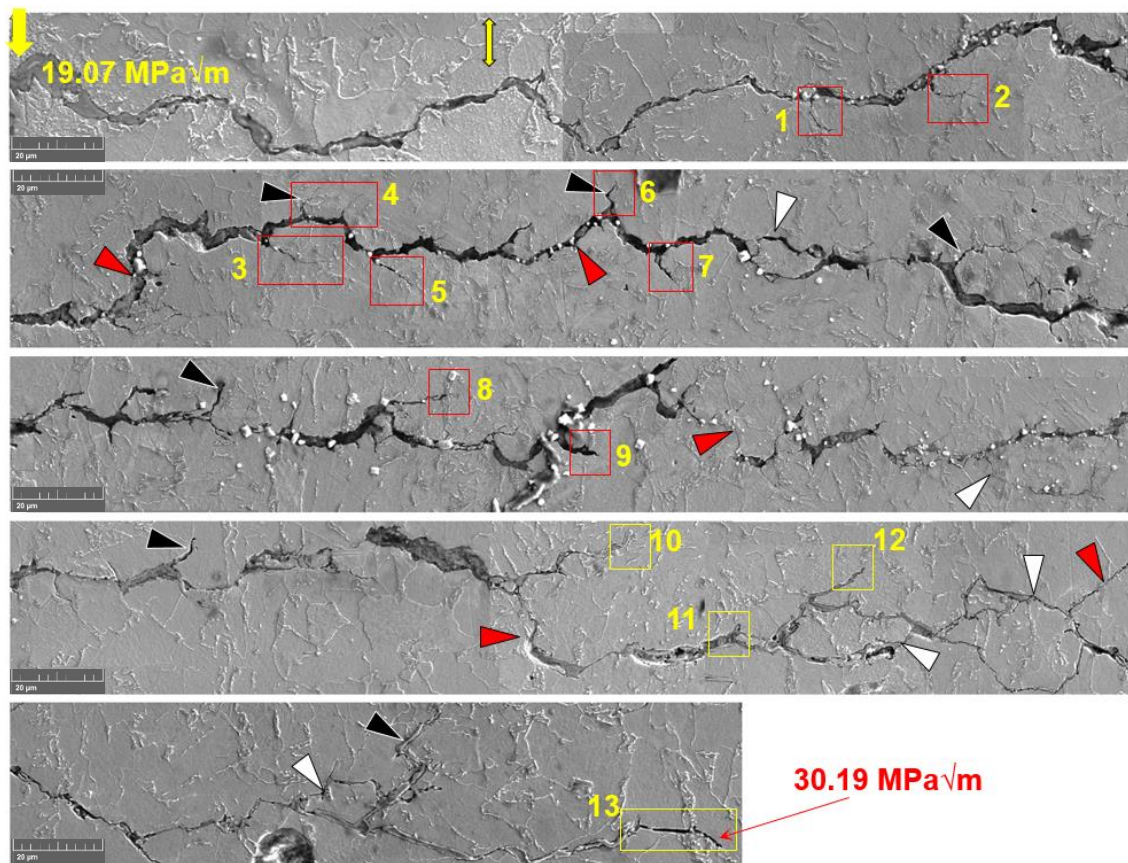
Figure 8-26, Figure 8-27 and Figure 8-28 show crack path in the materials in SW. The direction of all the crack growth in this study was from left to right and the direction of applied fatigue force is as given by the yellow double edge arrow. The crack path in **Figure 8-26** was observed for the G10 under sinewave, 9kN and 0.2Hz between the ΔK of 19.07 and 30.19 MPa $\sqrt{m}$. The same high angle crack path diversion as denoted by the red arrow, branching (black arrow) and metal crumb (white arrow) phenomena characteristic of these TMCP steels in air were also observed in SW. But, the number of the branched crack fronts is small, and the extension or length of the branched arm is short compared to that of air. The corrosion process had limited the extent or length of the arm of the branched crack. Metal crumbs were also observed, but the extent is small compared to air test.

Another feature that can be seen is the corrosion attack of the separated surfaces which resulted to widening of the fatigued surfaces as compared with the test in air (see **Figure 8-20**). This widening process may be due to elimination of the small metal crumbs, plastically deformed wake and the *microplastic*-zone areas by the corrosion process. Another observation of high importance is that the crack tips of the branched crack fronts from number 1 to before number 10 are reasonably sharp as shown by the magnified micrographs on the right of **Figure 8-26 (b)**. From branched crack tip number 10, to the tip of the propagating crack

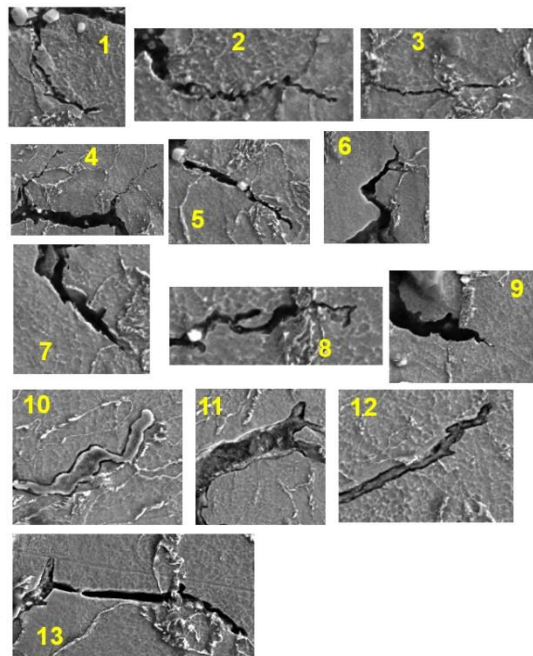
front (no. 13), all denoted by yellow boxes, appeared to have some degree of crack tip blunting. It can be reasoned that at high ΔK , the crack-tip appeared to be sufficiently open to allow some degree of blunting by the corrosion process. This generally suggests that the crack tip under 9kN, 0.2Hz, sinewave was always sharp up to a particular ΔK , before it started experiencing some significant blunting effect of the corrosion process. The sharpness of crack tip ensured that the crack growth was not retarded. The blunting effect at high ΔK may not have much retarding effect because of the high crack driving force (mechanical fatigue is likely to predominate) at high ΔK .

Figure 8-27 shows the crack path in the G10 under sinewave, 10kN and 0.5Hz. Similar features of high angle diversion as noted by the red arrow, branched crack front (black arrow) and formation of metal crumbs (white arrow) were found. Note that there were very few high angle crack tip diversions and branching as compared with the air test. There were few also formation of metal crumbs in the range 18 – 22 MPa $\sqrt{m}$ of the ΔK . There was also the widening of the gap between the fatigued surfaces in comparison with air test by the corrosion process and the branched crack tips still maintained some reasonable sharpness. However, there is considerable crack blunting at number 5, 6 and 7 and these occurred at high ΔK . Number 2 and 5 blunted to the extent that another crack front budded from its side.

The presence of these sharp crack tips under sinewave suggests two things: 1. the sinewave maintains sharp crack tip at almost every stage of the crack growth, except at high ΔK , and 2.



(a)



(b)

Figure 8-26: Crack path in G10 TMCP steels in seawater: sinewave, 9kN and 0.2Hz.
(scale = 20μm)

The widening of the fatigued surfaces under sinewave is an event that happened majorly near and immediately behind the sharp crack tip as it propagates and not because of exposure to the corrosive environment for a long time. This appears to be so when one examines some of the magnified crack tips shown in **Figure 8-26 (b)**, especially number 6, 7, 8 and 9.

One could observe a re-orientation of the metal pieces as encircled in **Figure 8-27**. The metal bits have the potential of reducing the driving stress at the crack tip thereby retarding the crack growth at that portion of formation. Observed also is that the branched crack tip at the low ΔK was not as sharp as those of 9kN. The magnified micrographs at the bottom of **Figure 8-27** shows a branched crack tip with some degree of tip blunting. This suggests that increasing the fatigue load to a level where the crack opening displacement is significant in *SW* can result to some degree of crack tip blunting which may retard the crack growth. An important observation also is that high angle crack deviation and branching are considerably reduced in *SW* and this is a consequence of repeated crack tip blunting by the chemical dissolution process.

Ordinarily, one would expect that the test under 10kN (G10-S10F0.5) would give a higher CFCGR than that of 9kN (G10-S9F0.2) due to higher driving force, but that is not the case. The plot of the da/dN vs. ΔK curve in *SW* between the two test conditions at the bottom right of **Figure 8-27** shows that above about $21\text{MPa}\sqrt{\text{m}}$, the growth rates are similar, but below this value, the growth rate of the 9kN test became higher than that of 10kN).

Since there was similar formation of metal crumbs in the two tests, this reduction may be attributed primarily to the blunting effect by the increased corrosion process due to sufficient crack tip opening at low ΔK . The advancing crack tip under 9kN (see **Figure 8-26**), is sharper than the crack tip under 10kN and crack tip blunting retards crack growth.

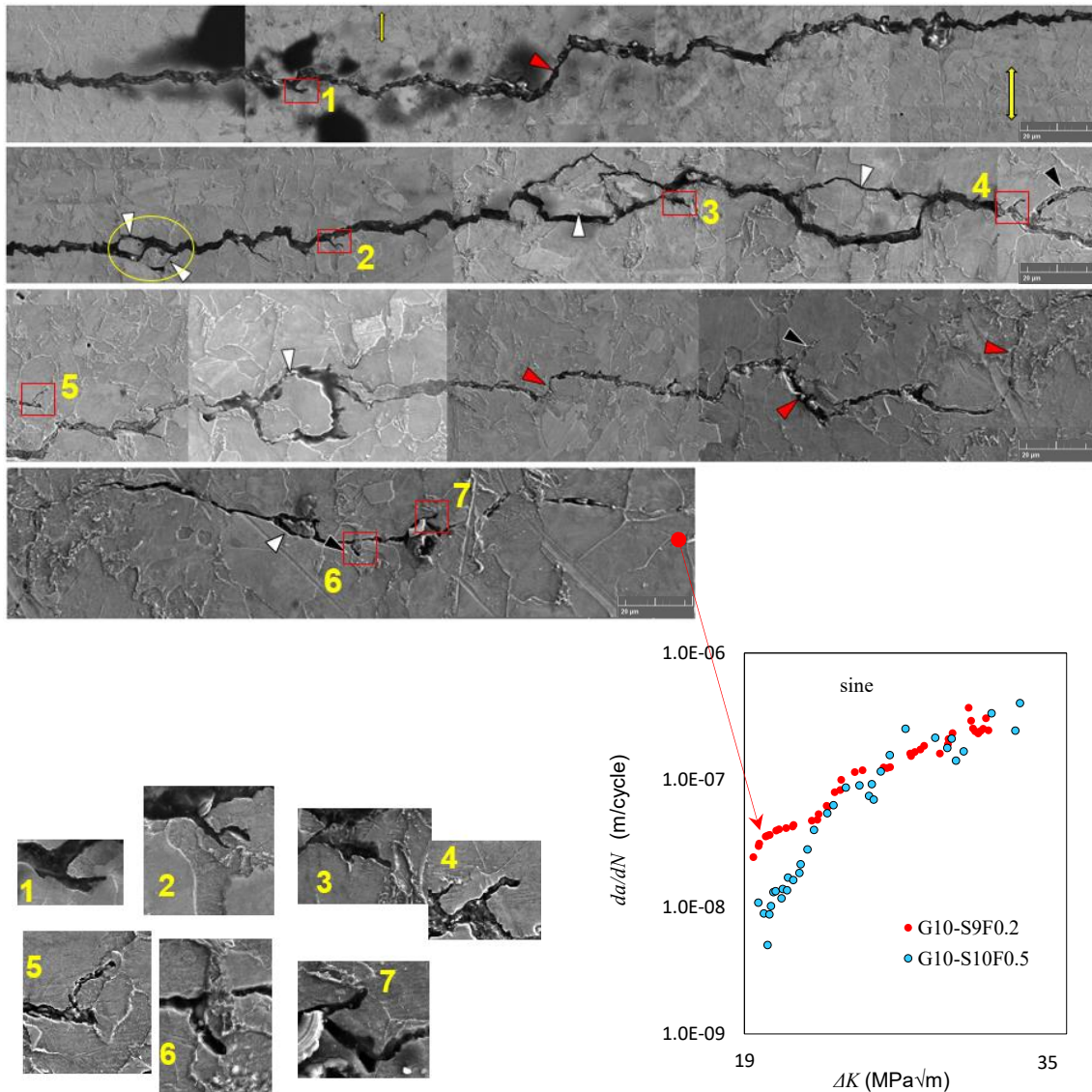


Figure 8-27: Crack path in G10 in seawater: sinewave, 10kN and 0.5Hz
(scale = 20 μ m)

As previously noted, this reduction is primarily attributed to the crack tip blunting at lower ΔK under relatively high load or loading condition that exposed the plastically deformed areas ahead or behind the propagating crack to more damaging chemical species. The chemical dissolution effect or the sharpness of the crack tip appears to be the primary factor influencing crack growth in SW than frequency.

8.11 Crack path analysis in the J2N steel in seawater under sinewave

Figure 8-28 shows a typical nature of the crack path in the J2N steel in SW. The crack propagated with very low angle of diversion as shown by the red arrow and very limited low angle crack branching with the shortest arm (shown by the black arrow) in comparison with the TMCP steels.

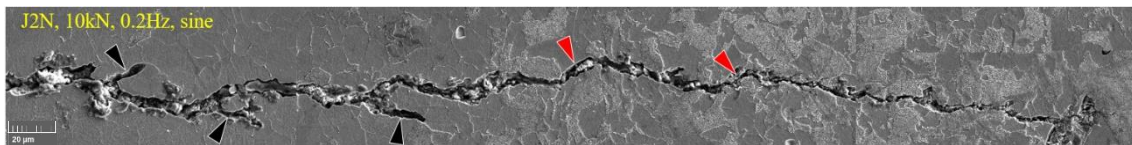


Figure 8-28: Crack path in J2N in seawater under sinewave (scale = 20μm)

The crack moved on a horizontal plane with minimal zig-zag motion. The crack growth pattern is generally the same with that in air (**Figure 8-21**) except that the branched arms are shorter and there is widening of the crack gap as a result of the corrosion process. No crumb formation was seen, and crack diversion and bifurcation are less than that in air, hence, one would expect a very high CFCGR for this crack growth pattern if blunting of the main active crack tip did not occur.

In **Figure 8-28**, it appears that there is no predominate features of the phases influencing the crack growth, except that the crack diverted in a zig-zag manner which is a consequence of the crack propagation mechanism which is currently under study. However, we could see in that the branched crack tips appeared to be following the ferrite ribbon or the α/α and α/P interfaces. In other words, the preferred crack paths appeared to be the α/α and α/P boundaries, and these boundaries are decorated by the ferrite ribbon of high alloy content. The black arrow is the α/α while the blue arrow is the α/P . These distinct phases and morphologies influencing the FCGR in SW for the J2+N are noted in **Figure 8-29**. We can also note the widening activity near and immediately behind the crack tip as it propagates. This is likely to be as a result of intense corrosion activity around the *microplastic* zone region.

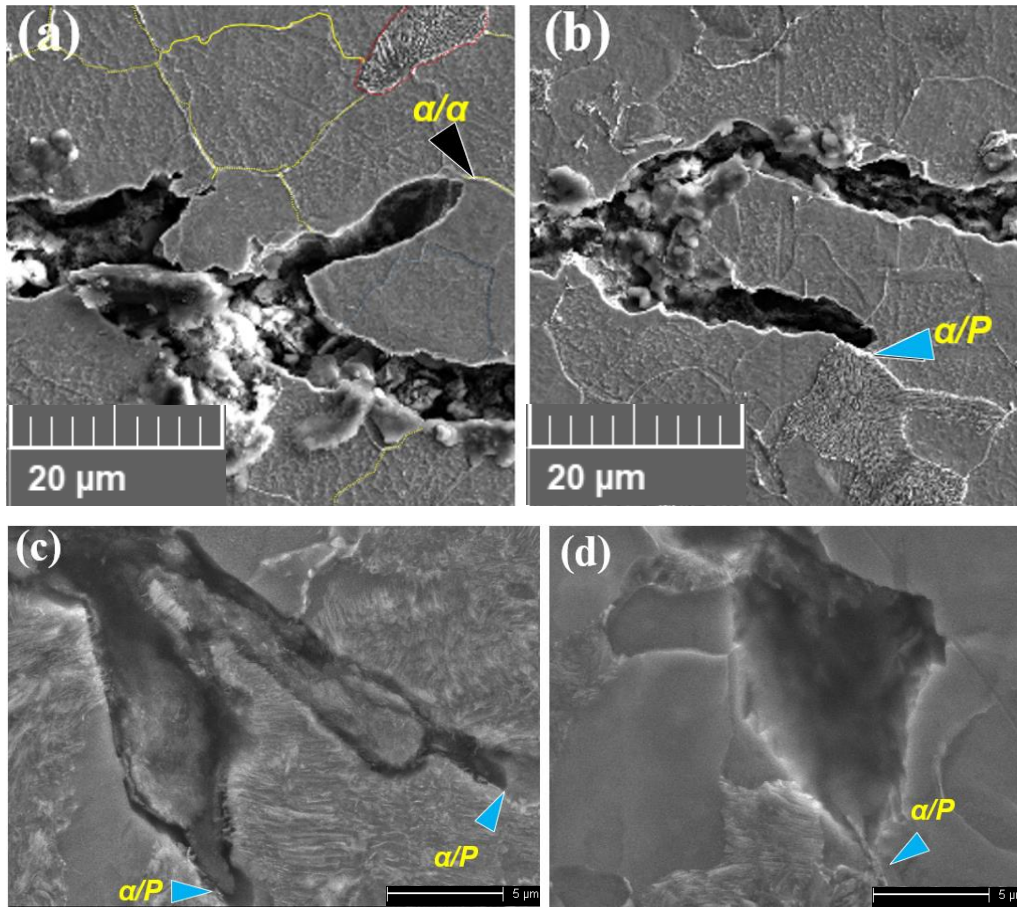


Figure 8-29: Phase influence on FCGR in NR steel in seawater

Figure 8-30 shows the comparison of the FCGR curves for the air and *SW* tests under sinewave denoted (see **Table 4-3**): J2N-A10F5Hz; G10-A10F5Hz; J2N-S10F0.2; G10-S9F0.2; G10-S10F0.2 and G10-S10F0.5. **Figure 8-30 (a)** shows that CFCGR of the 0.2Hz is higher than that of air by an average factor of 2.5 in the Paris region for the J2N steel.

Above about 24 MPa√m, the rate increased up to an average factor of 4.0. These values are typical of normalised ferrite-pearlite steels of the same microstructure. The sudden jump in the CFCGR to a factor of about 4 can be explained by examining the crack path in air and *SW*. In air (**Figure 8-21**), bifurcation, crumb formation and some zig-zag crack propagation were sustained to high ΔK .

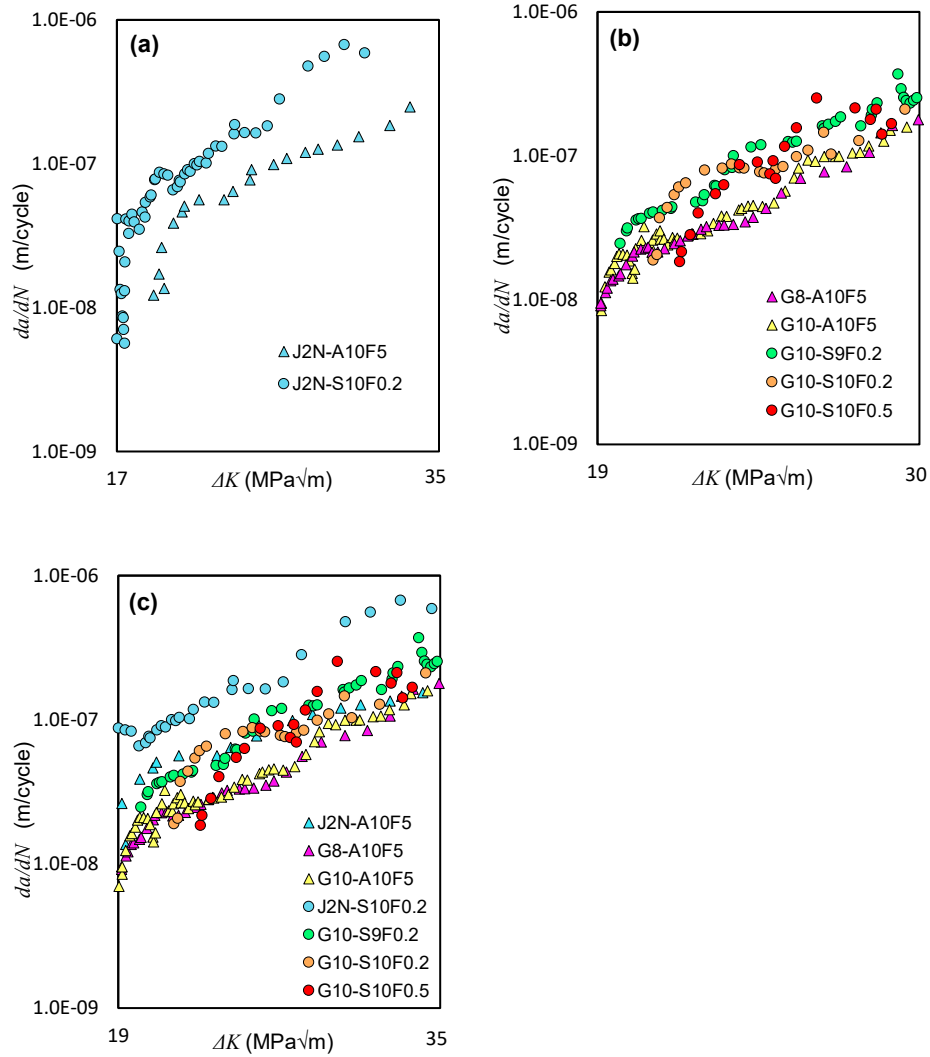


Figure 8-30: CFCGR in TMCP and J2N steels at different frequencies under sinewave

In the SW, at ΔK above the 24 MPa $\sqrt{m}$, the crumb formation and crack branching were almost absent (**Figure 8-28**) and this resulted to a sudden increase in the FCGR to the factor of 4. In **Figure 8-30 (b)**, the crack growth rates of the G8 and G10, denoted by triangular data points, at 10kN, 5Hz are the same, confirming that the microstructures are generally the same. In the Paris region taken from 19.50 MPa $\sqrt{m}$, the FCGR in SW is just higher than that of air by a factor of about 1.5 measured from their respective mean lines. In **Figure 8-30 (b)**, there is no obvious trend in the CFCGRs of the three tests in SW under sinewave. We can

reasonably say that the CFCGRs are the same above about 21 MPa $\sqrt{m}$, but below this value the 9kN, 0.2Hz test is somewhat higher.

Figure 8-30 (c) shows the crack growth rates for the two microstructures – J2N and TMCP. It is very clear that the crack path with the least crack diversion, branching and crumb formation gave the highest FCGR in air and SW. *The extent of these three phenomena is a function of the nature of the material microstructure and crack tip condition. If the microstructure, environment and loading condition are such that the three phenomena are extensive then crack growth retardation occurs - otherwise the rate will increase.* High loading condition in SW tends to limit the extent of crack diversion and branching. This happens as a result of generation of relatively large plastically deformed areas around and ahead of the crack tip by the high loading condition and subsequent dissolution of these areas by the electrochemical process resulting to various degrees of crack tip blunting. An incessant blunting of the crack front will limit the length of the diversion and branched arm, forcing the crack to propagate with a small zig zag amplitude along the horizontal plane, and this will increase the crack growth rate. This explains why the crack path found in J2N-A10F5 (**Figure 8-21**) and J2N-S10F0.2 (**Figure 8-28**) produced the highest crack propagation rate in air and SW respectively as shown in **Figure 8-30 (c)**. In general, the crack path having considerable diversion, bifurcation and crumbs will lower the rate at of the crack propagation. In **Figure 8-30 (c)**, the CFCGR of the J2N is higher than that of TMCP steel by a factor of about 2.0 and this implies that the microstructure of the TMCP steel is superior to that of J2N in resisting crack growth in ferrite-pearlite steel. Technologically, this implies that the microstructure of steel can be engineered to retard FCG.

A general stress ratio, $R = 0.1$ has been used in this comparative study. This is an attempt to exclude the effect of the so-called crack closure, as if such effect existed it would affect the steels equally. In fact, the elimination of the asperities or rough edges makes roughness induced crack closure in corrosive environment unlikely. In the case of interlocking of deviated cracks, in SW the corrosion

process will be rapid in these areas too because the mechanical rubbing action will expose fresh surfaces for fast attack. Thus, the superiority of TMCP steels over the NR in SW, even in air cannot be attributed to crack closure and is unlikely for interlocking.

A plausible reason remains that under stress ratio range of 0.1 – 0.67, temperature range of 0°C to room temperature and frequency range of 0.05 – 0.7Hz, the microstructure strongly affected the rate at which a crack grows in ferrite-pearlite steel in air and in SW in the Paris Region – as compare with a more refined TMCP microstructure. This study therefore asserts that, the common believe that the microstructure does not significantly affect the Paris Region appears to be incorrect. This tends to be so, when the same grade of steel, but with different phase morphologies are put into considerations. The major factor responsible for this difference appears to be the ferrite and pearlite phase morphologies and properties, which are function of the processing method and chemical composition of the steel.

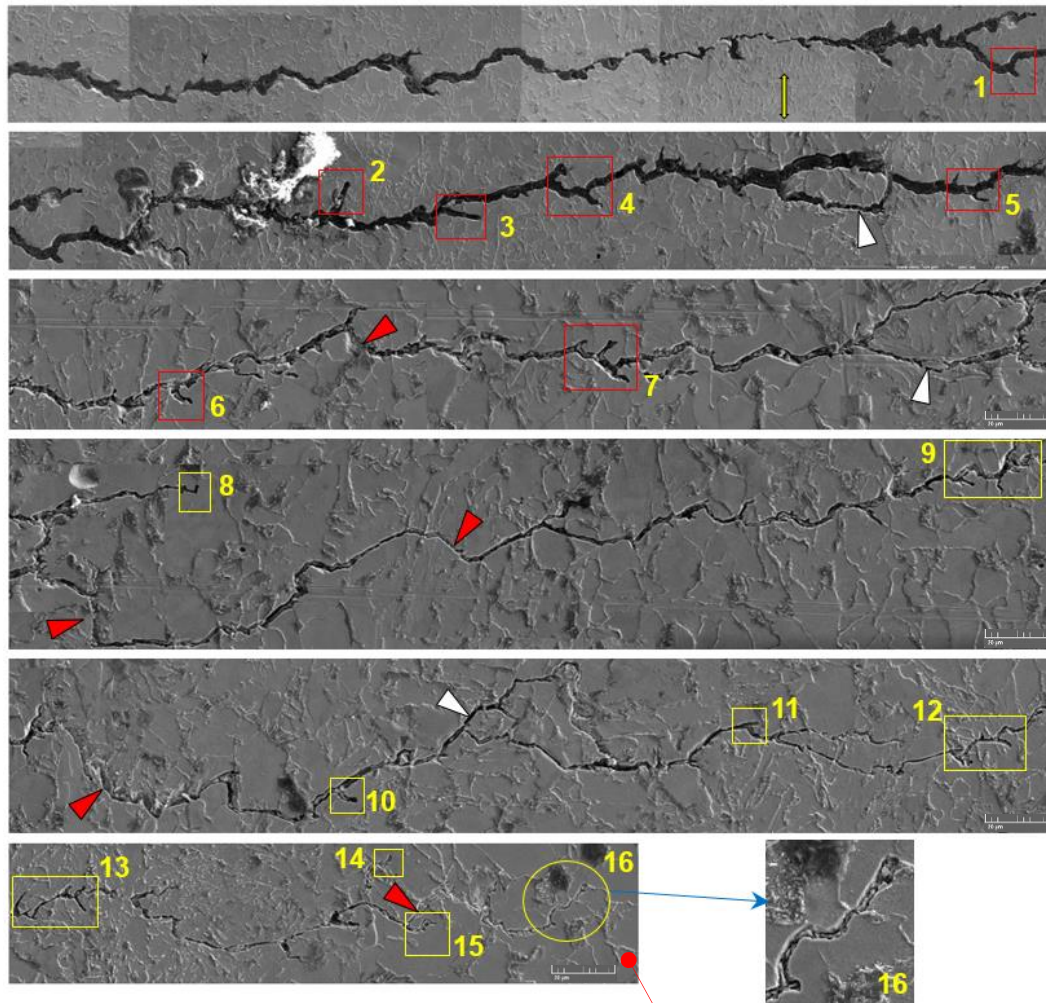
8.12 Crack path analysis in seawater under holdtime

The crack path in SW for the G10 under holdtime, 10kN and 0.2Hz is shown in **Figure 8-31**. Initially, a low angle crack diversion and branching similar to that of **Figure 8-27** was observed. Similar features of high angle diversion as noted by the red arrow, branched crack front and formation of metal crumbs were also found. The basic difference with test conducted under sinewave is that the branched crack tips are extensively blunted as shown in **Figure 8-31 (b)**. Number 1 to 7 show the severe blunting that occurred at lower ΔK under 10kN. The yellow boxes – from 8 to 15 show that the severe blunting process occurred also at high ΔK . The point of test stoppage is encircled (no 16) and it can be seen that the tip of the propagating crack is considerably blunted as shown by the magnified micrograph on the bottom right. We expect that the effect of this severe blunting at the tip of the main propagating crack front should be retardation of the crack growth rate, even crack arrest can occur.

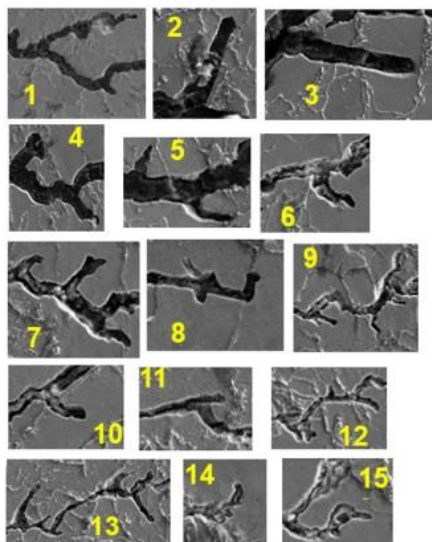
Figure 8-31 (c) shows the CFCGRs under sinewave and holdtime at 9kN and 10kN. The red long arrow shows the specific curve obtained from the crack path shown. The figure shows that the growth rate under 10kN holdtime test is the lowest. The lower the frequency of the holdtime at 10kN, the more the crack retardation process. This retardation effect can be attributed to the general crack tip blunting as seen in **Figure 8-31 (b)**. Note that under free corroding condition, the mechanism of crack tip growth should be by anodic dissolution (AD).

As the fatigue load increases, the amount of lattice distortion or *microplastic* zone around and ahead of the crack tip increases in size. The distorted or plastically deformed regions are areas of high energy and will be regions of high corrosion activity and hydrogen transport path. It is expected that crack blunting effect due to rapid corrosion process of these increased areas of high atomic energy would result. When the distortion or *microplastic* zone is small, as it is the case under low ΔK or load, the blunting effect is reduced. In fact, under this condition, the corrosion process may help make the crack tip sharp, hence, the CFCGR increases. Thus, increase in CFCGR for 9kN with respect to that of 10kN occurred due to the reduction in blunting effect. This appears to be the rate controlling factor since change in frequency at 9kN under waveforms used in this study has little effect on the CFCGR as shown in **Figure 8-32 (a & b)**.

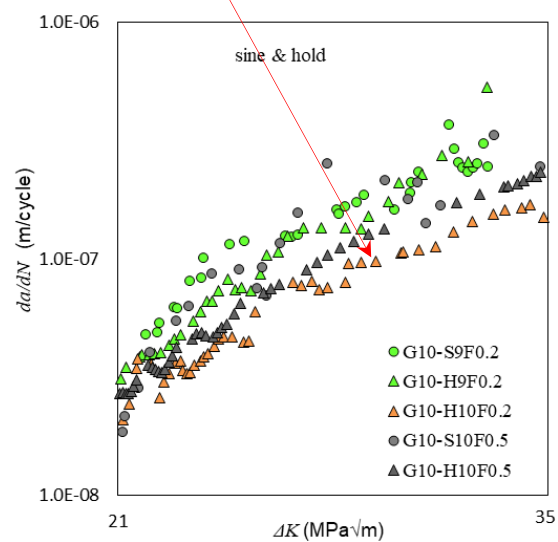
Figure 8-32 (a) shows that at low load (9kN) under sinewave, the frequency had no clear effect on the CFCGR. Though the CFCGR for 0.3Hz is slightly higher, but the CFCGRs of 0.2Hz and 0.5Hz are the same. For the hold-time (**Figure 8-32 (b)**), the frequency had a slight effect below about 25.50 MPa $\sqrt{m}$. **Figure 8-32 (a & b)** clearly shows that lowering the load level diminishes the effect of frequency. **Figure 8-32 (c)** shows some corrosion-fatigue tests conducted on G10 steel subgrades. About 85% of the sinewave data points lie above the imaginary dash line while the rest is below the dash line. About the same value is observed for the holdtime data points.



(a) (scale = 20 μ m)



(b)



(c)

Figure 8-31: Crack path in G10 in seawater: holdtime, 10kN and 0.2Hz (scale = 20 μ m)

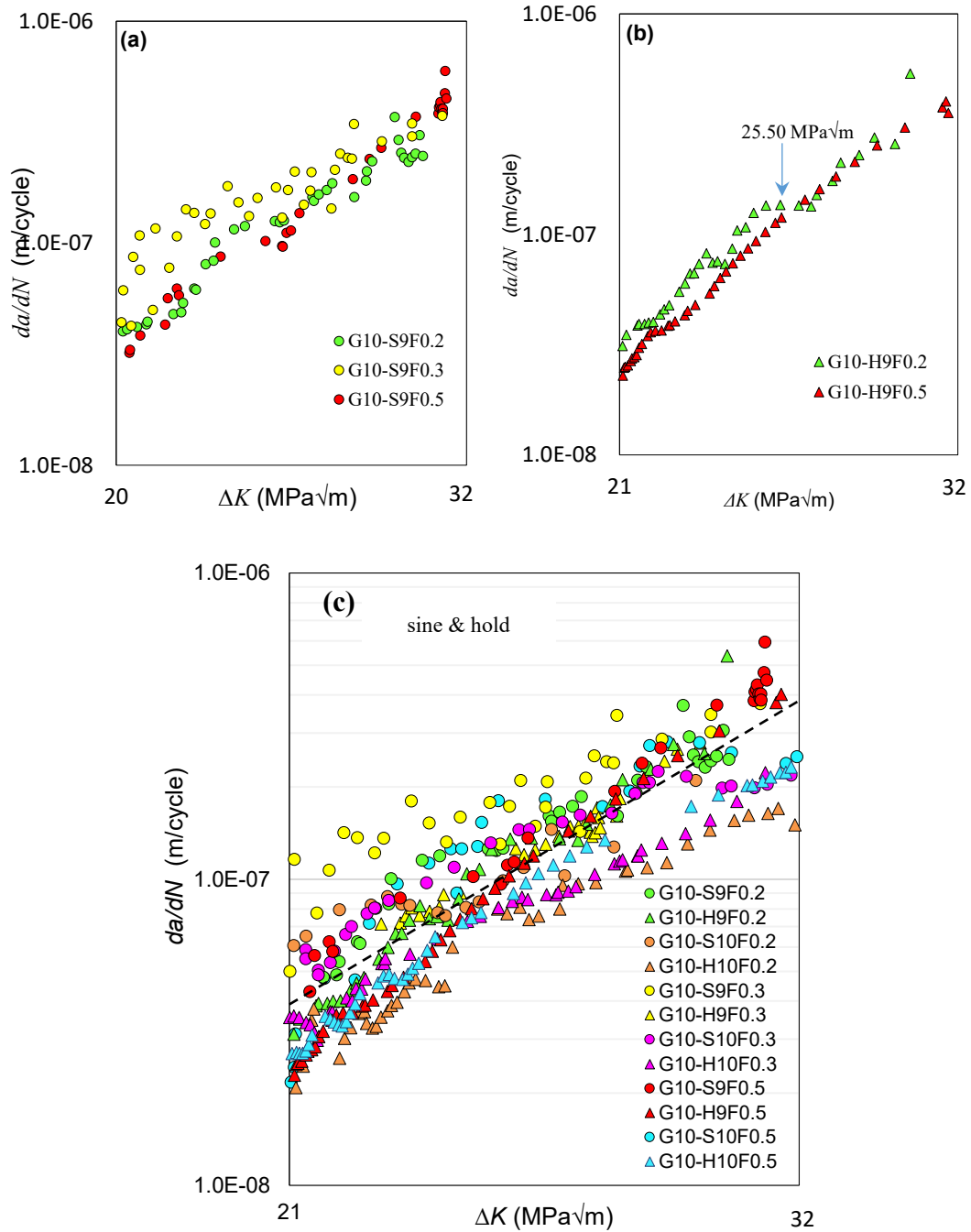


Figure 8-32: Effect of frequency and fatigue load on the CFCGR in G10 under sine and holdtime waveforms

This implies that CFCGR under sinewave is higher than that of holdtime from about 21 MPa $\sqrt{m}$ for all the load levels and frequencies used in this study. The 10kN, 0.2Hz gave the lowest CFCGR. For 10kN, 0.2Hz the crack tip is expected to be well open and reasonable time allowed for the corrosion process to take place. This produced severe crack tip blunting and a considerable reduction in

CFCGR as seen in **Figure 8-31 (b)**. However, as the *microplastic* zone size increases it cuts across many phases making it possible for the growing crack to overcome the blunting process by diverting or branching into phases of least resistance or nucleating a new crack front that follows the path of least resistance. But, generally, the effect of holdtime in the present study is retardation of crack growth rate and sinewave is found to be the more damaging waveform.

8.13 Conclusions

In this study, frequency is found to have no observable effect on the FCGR of steel in air. Generally, three phenomena were identified that primarily retard crack growth in ferrite-pearlite steels in air. These are crack diversion, crack bifurcation (branching) and metal crumbs. If the angles of the diversion and branching are high, crack retardation occurs. If the metal crumbs experience re-orientation, resulting to a mismatch or wedging action, retardation occurs. That is, a crack path containing considerable high angle crack diversion, bifurcation and crumbs will offer strong resistance to crack growth. The three factors retarded the crack growth by reducing or re-distributing the effective stress or driving force at the main active crack tips. The degree of reduction in the driving force at the main active crack tip appears to be a function of the variation in the angle of crack diversion, branching and the nature of metal crumbs created. Decrease in the crack diversion angle, crack branch length and number of metal crumbs is found to increase FCGR both in air and SW. In other words, the crack path with the least crack diversion, branching and crumb formation will give the highest FCGR both in air and SW. This, perhaps, explains why FCGR is lowest in the TMCP steels as compared with the normalized steel in both environments. The extent of formation of the three phenomena appears to be a function of the nature of the material microstructure and crack tip condition. If the microstructure, environment and loading condition are such that the three phenomena are extensive then crack growth retardation occurs - otherwise the rate will increase.

In *SW*, increasing the fatigue load and decreasing the frequency, especially for holdtime has the highest retarding effect under corrosion dominated regime. This is probably because as the fatigue load increases, the amount of lattice distortion or plastically deformed regions around and ahead of the propagating crack tip increases in size. The high load exposes the crack tip region and allows the entry of more damaging chemical species into the plastically deformed region. These areas of high atomic energy or non-equilibrium surface atoms are rapidly attacked by the time-dependent electrochemical process resulting to rapid crack tip dissolution and blunting. When the main active crack tip is blunted the crack propagation process is considerably reduced and or even arrested for a very long time as observed especially for the holdtime. The effect of holdtime in the present study is retardation of FCGR and sinewave is found to be the more damaging waveform.

An important observation also is that high angle crack deviation and branching are considerably reduced in *SW* and this is a consequence of repeated crack tip blunting by the chemical dissolution process. If the branched crack tips are blunted such that the branched length is short or limited and the main active crack tip is sharp, CFCGR can increase up to about a factor of 4. If the main active crack tip is blunted as well, the crack propagation speed is considerably reduced. However, as the *microplastic* zone size increases it cuts across many phases increasing the tendency for the blunted crack front to nucleate a new crack front that follows the path of least resistance. In this study also, corrosion products are found to build up in and around the blunted crack tip leading to further retardation of the crack growth.

This study has been conducted to exclude the crack-closure effect. The elimination of the asperities or rough edges makes roughness induced crack closure in corrosive environment unlikely. In the case of interlocking of deviated cracks, the mechanical rubbing action will expose fresh surfaces for fast corrosion attack. Thus, making crack retardation by interlocking unlikely or very minimal in *SW*. The chemical dissolution of the *microplastic* zone or crack tip blunting,

accumulation of corrosion product in and around the crack tip and the sharpening of the crack tip at low fatigue load level appears to be the primary factors influencing crack growth in ferrite-pearlite steel in SW. In other words, since ΔK is a function of a and stress, given a sharp elastic crack in an infinitely wide plate thus: $\Delta K = \Delta\sigma(\pi a)^{0.5}Y$; *at the same ΔK , combination of low $\Delta\sigma$ and long a will be more damaging than combination of high $\Delta\sigma$ and short a in SW at low frequency.* It is found also that reduction of the load level tends to reduce the effect of frequency in SW and the difference in CFCGR for waveforms diminishes with increasing frequency and ΔK .

Finally, this study supports the assertion that the nature of the microstructures of the ferrite-pearlite steels has a strong influence on the FCGR in the Paris region of the da/dN vs. ΔK sigmoidal curve in air. This is contrary to the current belief that microstructure has little or no influence on the FCGR curve in the intermediate FCG region. Also, current theory says that the crack path in the Paris Region or Stage II in the da/dN vs. ΔK sigmoidal curve is generally by transgranular. This research observed both transgranular and intergranular modes of propagation in both environments. Though what appears as an intergranular mode is actually the growth of the crack through a thin layer of high solute ferrite ribbon, α_{HA} adjacent to the low relief, α_{LR} or the pearlite, P phases, thus properly referred to as quasi-intergranular. The morphology and chemistry of the material phases locally to the main crack front and load level appear to determine which mode the crack growth will adopt. Another factor is that the angle the crack front makes with the phases ahead of it tends to determine if it would propagate in transgranular or by quasi-intergranular mode.

9 Overall conclusions and future work

9.1 Overall conclusions

All indications show that the global energy from wind power will rise due to rapid reduction in the cost of wind energy. This is driven by the advances in turbine technology, efficient and optimized turbine support structure. WT's in the offshore environment are subjected to constant fatigue forces and the foundations are in direct contact with SW. There are many areas of the WT that are unprotected. Cracks could nucleate and grow in these areas, even from protected areas. At critical length, these cracks could cause catastrophic structural failure at loads far less than the maximum design load and the problem is worsened for cyclic loading in offshore environment. Linear Elastic Fracture mechanics (LEFM) approach is commonly employed in predicting the load at which a crack of given length will grow and cause failure. This practice is important to aid economic steel selection decisions, prediction of operating lifetime of offshore structures, effective preventive maintenance and generally safe structural design.

S355 TMCP steels are reported to be well suited for the support structures of modern extra-large (XL) wind turbine support structures (WTSSs). At present, the corrosion-fatigue performance of these advanced sub-grades of S355 steels is scarce. This research is an experimental investigation of the defect tolerance behaviour of these steels - particularly S355 steel subgrades of 90 mm thickness (as purchased) designated as S355G10+M, S355G8+M and S355J2+N under the influence of the fatigue waveforms, change in fatigue load level and the material microstructure in SW corrosion. The frequencies of interest are 0.2Hz, 0.3Hz and 0.5Hz which are frequencies of commercial WT's in the North Sea. Sine and holdtime waveforms of constant amplitude and stress ratio of 0.1 were used. The maximum applied fatigue loads were 9kN and 10kN. The temperature is ambient for the air test and varied from 7.5 to 8.2 °C for the SW environment. The Linear Elastic Fracture Mechanics (LEFM) method was employed. All the tests and investigations were limited to the Stage II of the da/dN vs. ΔK plot,

where LEFM applies. All the tests have been performed in accordance with BS 7910:2013+A1:2015 and ASTM E647-15.

This study found that frequency has no obvious effect on the FCGR of the steels in air. Under a test frequency of 0.3Hz and sinewave, the FCGR in S355 TMCP steel subgrades S355G8+M and S355G10+M steels are generally the same in air *and* SW. The upper bound FCGR in S355G10+M steel in air is about $\frac{da}{dN} = 1.66 \times 10^{-13} \Delta K^{4.12}$ while that of SW is about $\frac{da}{dN} = 5.49 \times 10^{-12} \Delta K^{3.35}$. The SW enhanced the FCGR by a factor of 2.31 throughout the Paris Region in the range of 20 - 50 MPa $\sqrt{m}$.

Generally, three phenomena were identified that primarily retarded crack growth in ferrite-pearlite steels in air. These are crack diversion, crack bifurcation and metal crumbs. A crack path with extensive high angle crack diversion, bifurcation and crumbs will offer strong resistance to crack growth, hence low FCGR. The three factors retarded the crack growth by reducing or re-distributing the effective stress or driving force at the main active crack tips. Decrease in the crack diversion angle, crack branch length and number of metal crumbs is found to increase FCGR both in air and SW. The factors explain why FCGR is highest in the normalized steel as compared with the TMCP steels in both environments. The extent of formation of the three phenomena is a strong function of the steel microstructure. This implies that microstructure has a strong effect on the FCGR of α -P steels in the Paris Region, both in air and SW. This conclusion is in contrary to current theory that microstructure has little or no effect in the Paris Region of the da/dN vs ΔK sigmoidal curve. The crack path was always non-planar with complex crack front.

This study shows that the CFCGR of the sine waveform is higher than that of the holdtime for all the load levels and frequencies used. This effect is more noticeable at combination of high applied fatigue load and low frequency. In other words, the sinewave is the more damaging waveform and the consequence of

holdtime is retardation of crack growth rate in *SW*. This implies that in the frequency range of 0.2Hz to 0.5Hz, the use of Paris law constants from sinewave corrosion-fatigue tests for life assessment of OWT monopiles made of S355G10+M will be a conservative approach.

Change in load level and frequency did not affect the CFCGR of the TMCP steels in *SW* under sinewave, but slight change was observed with decrease in the load level for the holdtime. The P_{max} (in the range of 9kN to 12kN) is found to have a profound influence on CFCGR than the cyclic frequency (in the range of 0.2Hz to 0.4Hz). Decreasing the load level reduces the effect of frequency in *SW* and the difference in CFCGR for waveforms diminishes with increasing frequency and ΔK . Comparison of the experimental results of S355G8+M, S355G10+M and S355J2+N for both sine and holdtime waveforms showed that FCGR in normalised steel (e.g. S355J2+N) is consistently higher than that of TMCP steels (e.g. S355G8+M, S355G10+M) in both air and sea environments.

In *SW*, increasing the fatigue load and decreasing the frequency, especially for holdtime has the highest retarding effect under corrosion dominated regime. This is a consequence of the size of the *microplastic* zone. High load level (1) increases the *microplastic* zone size – i.e. the degree of lattice distortion or plastically deformed regions closest and ahead of the propagating crack tip and (2) exposes the crack tip region and allows the entry of more damaging chemical species into the plastically deformed regions. Because these areas are regions of high atomic energy or non-equilibrium surface atoms, they are rapidly attacked by the time-dependent corrosion process resulting to rapid crack tip dissolution and blunting. When the main active crack tip is blunted the crack propagation process is considerably reduced or even arrested for a very long time as observed especially for the hold time test. This appears to suggest that for same ΔK value, combination of low stress range and long crack length, a will be more damaging than combination of high stress and short a in *SW*.

High angle crack deviation and bifurcation are significantly reduced in *SW* and this is a consequence of repeated crack tip blunting by the chemical dissolution process. If the branched crack tips are blunted such that the branched length is short and the main active crack tip is sharp, CFCGR can increase up to about a factor of 4. If the main active crack tip is blunted as well, the crack propagation speed is considerably reduced or even arrested. In the corrosion dominated region, increase in load level increases crack tip blunting process. However, as the *microplastic* zone size increases it cuts across many phases increasing the tendency for the blunted crack front to nucleate a new crack front that follows the path of least resistance. In other words, there is continuous blunting and re-initiation of crack front in *SW*. Corrosion products are also found to build up in and around the blunted crack tip leading to further retardation of the crack growth. Extensive blunting of the crack front due to availability of time explains why the CFCGR is generally lower in holdtime as compared to sinewave.

This study found that high loading condition in *SW* tends to limit the extent of crack diversion and branching. An incessant blunting of the crack front appears to limit the length of the diversion and branched arm, forcing the crack to propagate in a small zig zag amplitude along the horizontal plane, and this propagation mode increased the crack growth rate. This explains why the crack path found in the normalise steel (J2N-A10F5 and J2N-S10F0.2) resulted to the highest crack propagation rate in air and *SW* respectively.

In this research, the chemical dissolution of the *microplastic* zone or the crack tip blunting process, accumulation of corrosion product in and around the crack tip and the sharpening of the crack tip at low fatigue load level appears to be the primary factors influencing crack growth in ferrite-pearlite steel in *SW*. Other fundamental factors are the crack angle diversion, branching of crack front and formation of metal crumbs along the crack path.

Also, current theory says that the crack path in the Paris Region or Stage II in the da/dN vs. ΔK sigmoidal curve is generally by transgranular. This is supported by

the ductile striation mechanism often seen on the fatigued surfaces of samples. In this study, transgranular and quasi-intergranular modes of propagation in both environments were observed. The quasi-intergranular mode is a situation where the fatigue crack propagates through a thin layer of high solute ferrite ribbon, α_{HA} adjacent to the low relief, α_{LR} or the pearlite, P phases. The morphology and chemistry of the phases local to the main crack front and load level appear to determine which mode the crack growth will adopt. The angle the crack front makes with the least resistant path ahead of it tends to determine if it would propagate by transgranular or quasi-intergranular mode.

9.2 Future work

1. There is need to carry out similar experiments under the same conditions in this research, but with investigation into the effect of variations in the fatigue loading rate – i.e. effect of longer and shorter rise times.
2. As a modification to the holdtime test, fatigue turbulence can be introduced at the maximum load and the effect on the CFCGR examined.
3. The effect of mean stress (increase in R ratio) in the light of the current research conclusions can be investigated for complete understanding of the FCG phenomenon.
4. Theoretical models need to be developed to account for the effect of microstructural effect in the Paris Region for ferrite-pearlite steels.
5. Accelerated tests may be conducted to understand more the effects of the experimental factors employed in this research.
6. There is need to study in detail the best combination of chemical composition, heat treatment and deformation procedures that will offer the highest resistance to FCG in the experimental steels. This can lead to patentable outcome. Modelling may be employed to guide the experimental process. This is purely a steel design (microstructural engineering) and fatigue testing endeavours. It is the intention of the researcher to apply for a research grant from willing organisations to continue this work in this direction.

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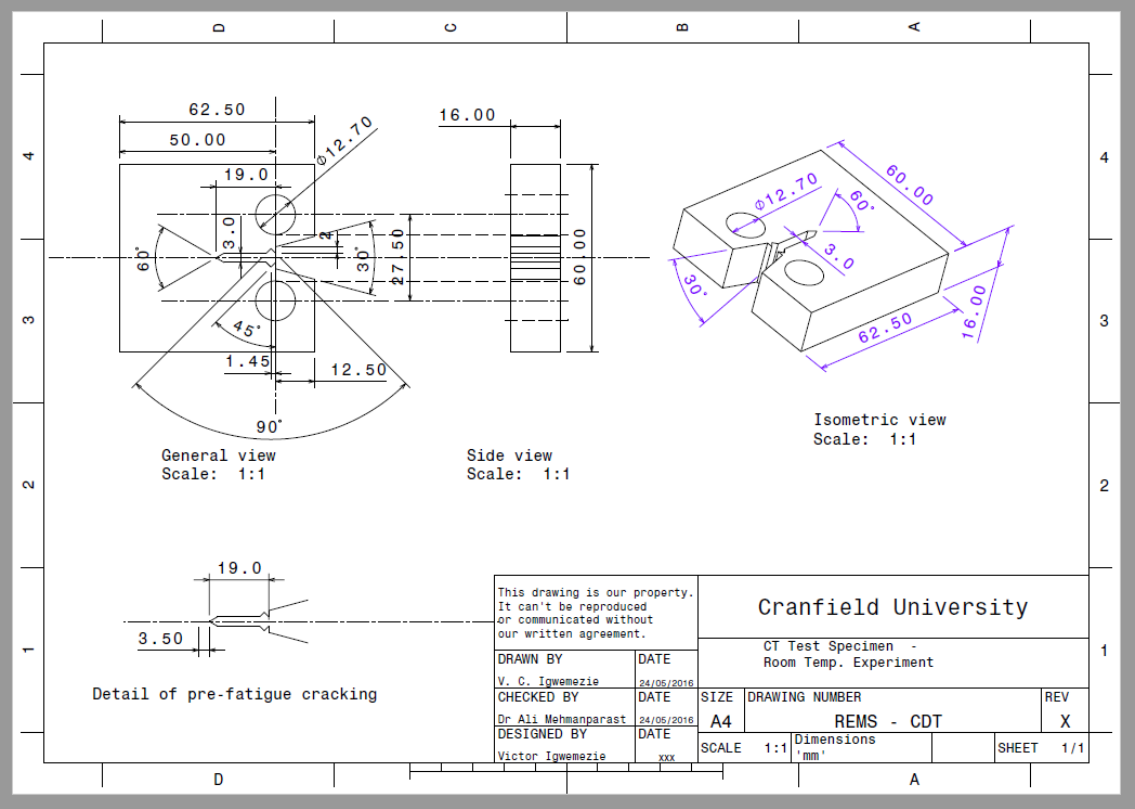
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APPENDICES

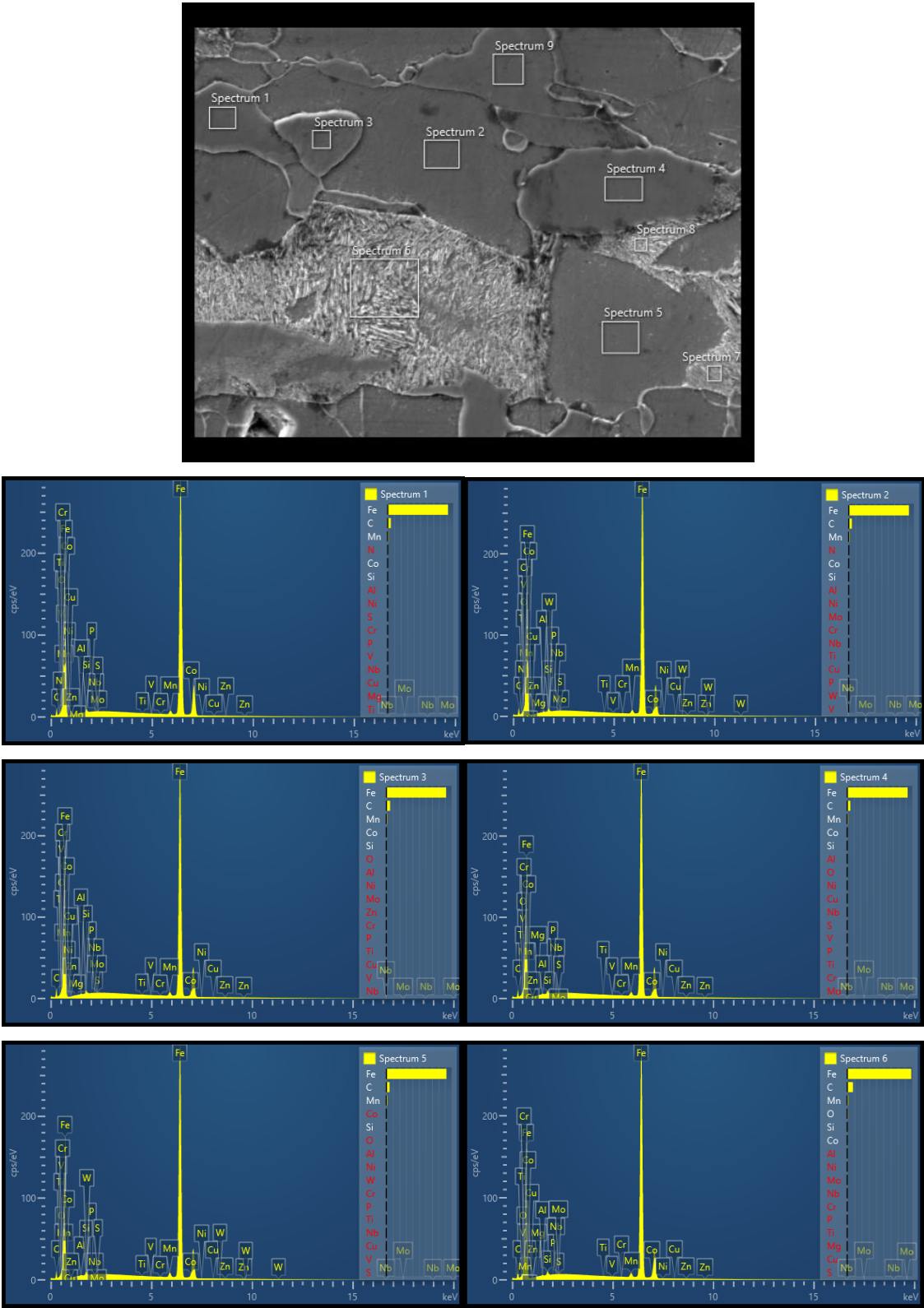
Appendix A : Test material certificate

QM-System: Certification as per ISO 9001										 DILLINGER HÜTTE																							
Erläuterungen siehe Rückseite/Explanations voir au verso/See reverse for explanations (www.dillinger.de/certificate)										A10 Advice of dispatch No./ Date of dispatch 432227-05.09.12										A09 Manufacturer's order/ A03 Certificate No. 376191-013 Sheet 1/...													
A02 INSPECTION CERTIFICATE 3.2 AS PER EN 10204:2004 INSPECTION REPORT 3.2 AS PER EN 10204:1991+A1:1995 + AS PER ISO 10474:1991										A06 Purchaser EEW SPC, ROSTOCK A07.1 No. 49012-01 A07.2 No. 49012-01										A01 Product HOT ROLLED PLATES													
A05 Established inspecting body NV										A08 Final receiver EEW SPC, ROSTOCK																							
B02/ Steel design B03 Any suppl. requirements										TYPE-III-8355G10+M EN-10225:09/EN-10164:04 0628-SPE-01-012_01:REV.01-23.12.11-PAG27-28																							
B01-B99 Description of the product																																	
B14 Item No.		B08 Number of pieces		B09 Thickness		B10 Width		B11 Length		B12 Theoretical mass		B04 Product delivery condition		B07.2 Heat No.		B07.1 Rolled plate No./ Test No.		A09 Purchaser article number															
27		1		90,00		x 2610		x 8500		15674		TM		376469		44026-01		49012															
***		1								15674		Actual mass: 15986 KG																					
B06 Marking of the product																																	
ITEM NO. : 27 STEEL DESIGNATION TYPE III 8355G10+M HEAT NO. / TRADEMARK / ROLLED PLATE NO.-TEST NO. / INSPECTOR'S STAMP																																	
C10-C29 Tensile test																																	
B14 Item No.		B07.2 Heat No.		B07.1 Roll plate/ Test No.		B05 Reference (heat) treatment		C01		C02/ C03 Temp. GR.C		C10 C11 MPA RP02		C12		C13		A % L0=50		C14-C15 RP02/RM		Z %											
27		376469		44026						K2 SV RT		523		524		522				0,82		59,0		62,2		47,4							
										K4 Q RT 460		561		23																			
C40-C49 Impact test																																	
B14 Item No.		B07.2 Heat No.		B07.1 Roll plate/ Test No.		B05 Reference (heat) treatment		C01		C02/ C03 Temp. GR.C		C41 Width of test piece		C40 Type of test piece		C44 Testing method		C46 Energy Joule		C45 AV-J		C42 Individual values AV-J		C43 Mean value									
27		376469		44026						K4 QO -40		CHP-V		3,5%-250C-1H		750		AV 251		316		315		294									
										K4 QM -40		CHP-V				750		AV 87		217		306		203									
										K4 QO -40		CHP-V				750		AV 245		273		251		256									
										K4 QV -40		CHP-V				750		AV 308		292		285		295									
C66-C68 Supplementary tests on test samples																																	
ITEM NO. : 27 DEEP ETCH TESTING AS PER DH-STANDARD <= 2B: SATISFACTORY.																																	
C70-C99 Chemical composition % - Heat analysis																																	
B07.2 Heat		C70		C		SI		MN		P		S		N		CU		MO		NI		CR		V		NB		AS		SN			
376469		Y		0,061		0,280		1,58		0,013		0,0007		0,0041		0,254		0,006		0,342		0,034		0,001		0,022		0,003		0,001			
B07.2 Heat		C70		TI		PB		B		SB		CA		BI		AL-T																	
376469		Y		0,003		0,000		0,0003		0,0010		0,0028		0,0001		0,032																	
C94 Heat analysis Carbon equivalent / Alloying restrictions																																	
B07.2 Heat		FO-02=		0,37		FO-31=		0,17		FO-51=		0,02		FO-52=		0,03		FO-A1=		7,80													
376469																																	
C95 Ladle treatment																																	
ITEM NO. : 27 HEAT OF THE INDICATED ITEM: VACUUM DEGASSED																																	
C70-C99 Chemical composition % - Product analysis																																	
B07.2 Heat		B07.1 Test No.		C01		C		SI		MN		P		S		N		CU		MO		NI		CR		V		NB		TI		B	
376469		44026		K4		0,060		0,271		1,57		0,013		0,0009		0,0039		0,241		0,006		0,333		0,034		0,001		0,022		0,003		0,0000	
376469		44026		K4		0,059		0,272		1,57		0,013		0,0009		0,0040		0,242		0,006		0,335		0,034		0,001		0,022		0,003		0,0000	
B07.2 Heat		B07.1 Test No.		C01		AL-T																											
376469		44026		K4		0,030																											
376469		44026		K4		0,030																											
C94 Product analysis Carbon equivalent / Alloying restrictions																																	
B07.2 Heat		B07.1 Test No.		C01		FO-02=		0,37		FO-31=		0,17		FO-51=		0,02		FO-52=		0,03		FO-A1=		7,69									
376469		44026		K4		FO-02=		0,37		FO-31=		0,17		FO-51=		0,02		FO-52=		0,03		FO-A1=		7,50									
376469		44026		K4		FO-02=		0,37		FO-31=		0,17		FO-51=		0,02		FO-52=		0,03		FO-A1=		7,50									
C94 Carbon equivalent formula / Alloying restrictions																																	
FO-02 = C + (MN/6) + (CR+MO+V) / 5 + (NI+CU) / 15 FO-31 = C + SI / 30 + (MN+CR+CU) / 20 + MO / 15 + V / 10 + NI / 60 + 5B FO-51 = V + NB FO-52 = V + NB + TI FO-A1 = AT / N																																	
D01 Marking and identification, surface appearance, shape and dimensional properties																																	
ITEM NO. : 27 EXAMINATION OF MARKING, SURFACE, SHAPE AND DIMENSIONS: THE RESULTS MEET THE REQUIREMENTS. SURFACE AS PER EN-10163-B3 THICKNESS AS PER EN-10029-B10 LENGTH AND WIDTH AS PER EN-10029:10																																	
D01 Marking and identification, surface appearance, shape and dimensional properties																																	
FLATNESS AS PER EN-10029:10-T4L																																	
D02 Non-destructive tests - Ultrasonic testing																																	
ITEM NO. : 27 UT-SPECIFICATION : EN-10160 CLASS S1 AND E2 SCANNING PLAN BODY : LONGITUDINALLY SCAN LINES SPACING 100 MM EDCS : 100 MM PERSONNEL QUALIFICATION : LEVEL 2 IN ACCORDANCE TO EN 473 THE TEST RESULTS MEET THE REQUIREMENTS OF THE ORDER.																																	
A04				2012/02/20 We hereby certify, that the above mentioned materials have been delivered in accordance with the terms of order. DNV CERT ESN-12-33781-13				Digitally signed by: Jungbluth, Norbert Location: DNV Essen, Germany Signing date: 2012-09-05				B. BALDAUF Test House Manager				AG der Dillinger Hüttenwerke Postfach 1580, D-66748 Dillingen/Saar Inspection department		A01															
																		Date 05.09.12		EDI		BM		1									

Appendix B : C(T) specimen dimensions



Appendix C : EDX Analysis of S388J2+N steel



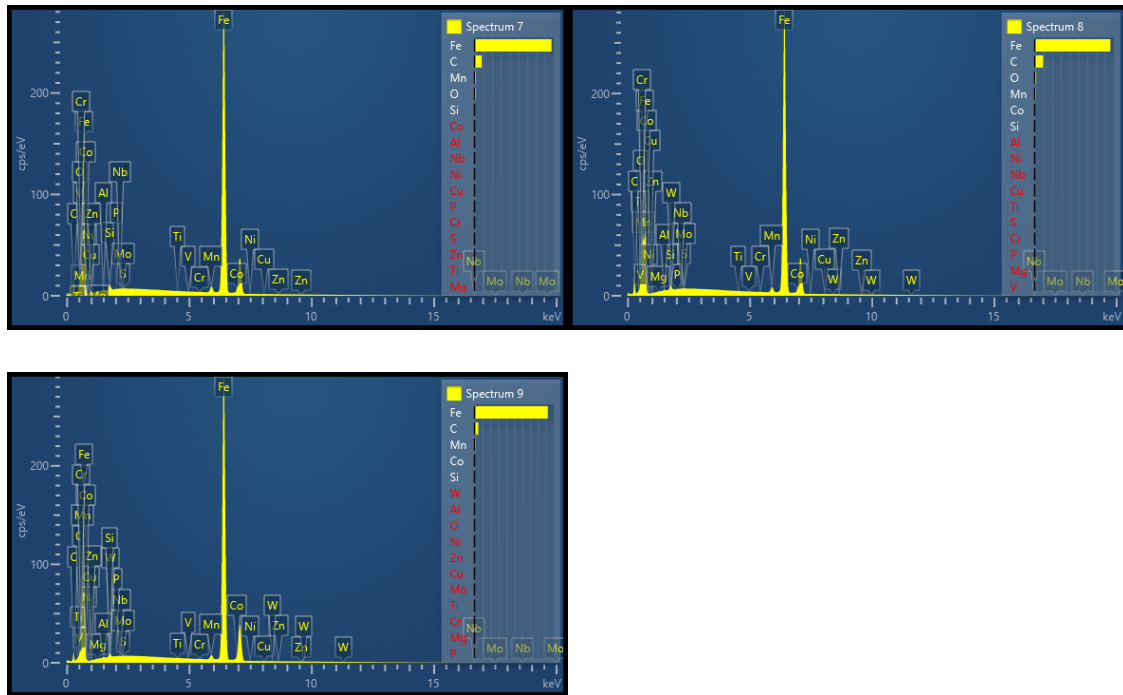


Figure C-1 Result of EDX compositional analysis on S355J2+N – sample 1

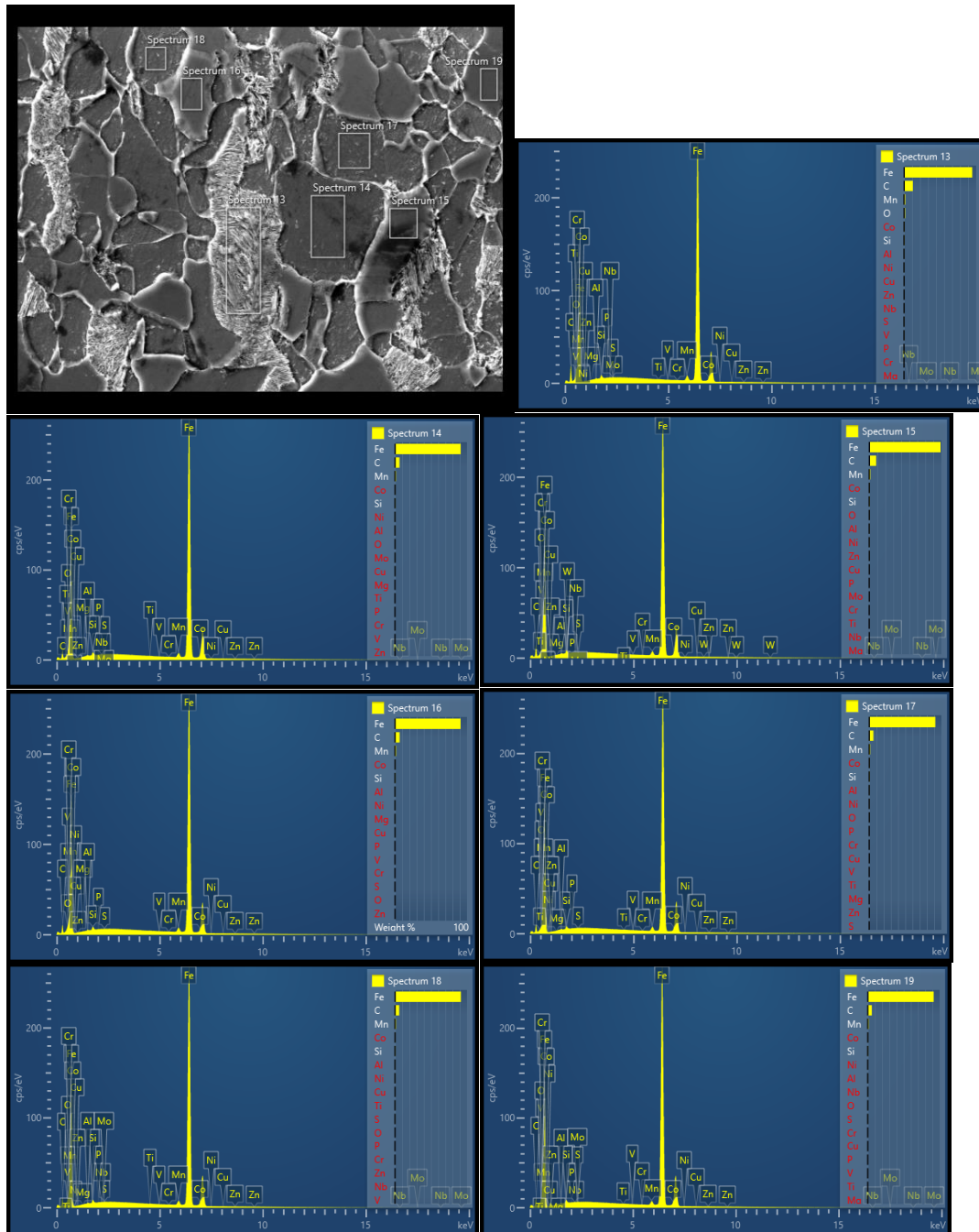
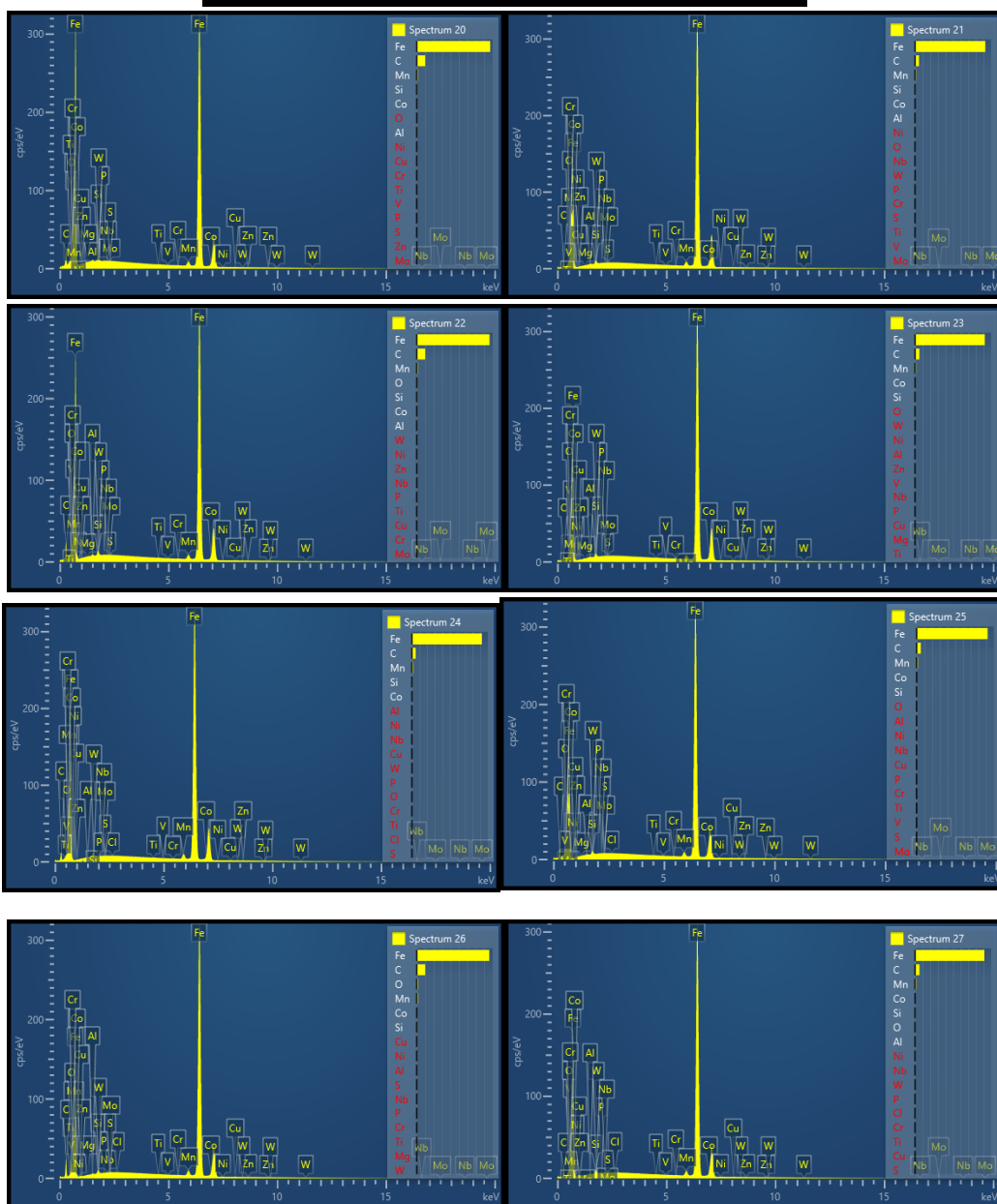
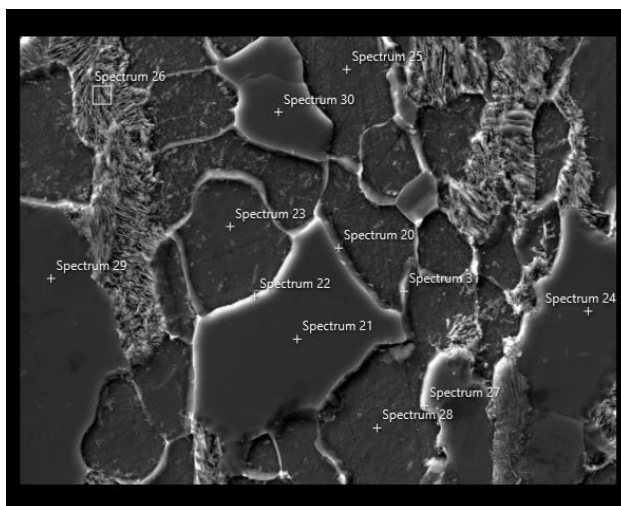


Figure C-2 Result of EDX compositional analysis on S355J2+N – sample 2



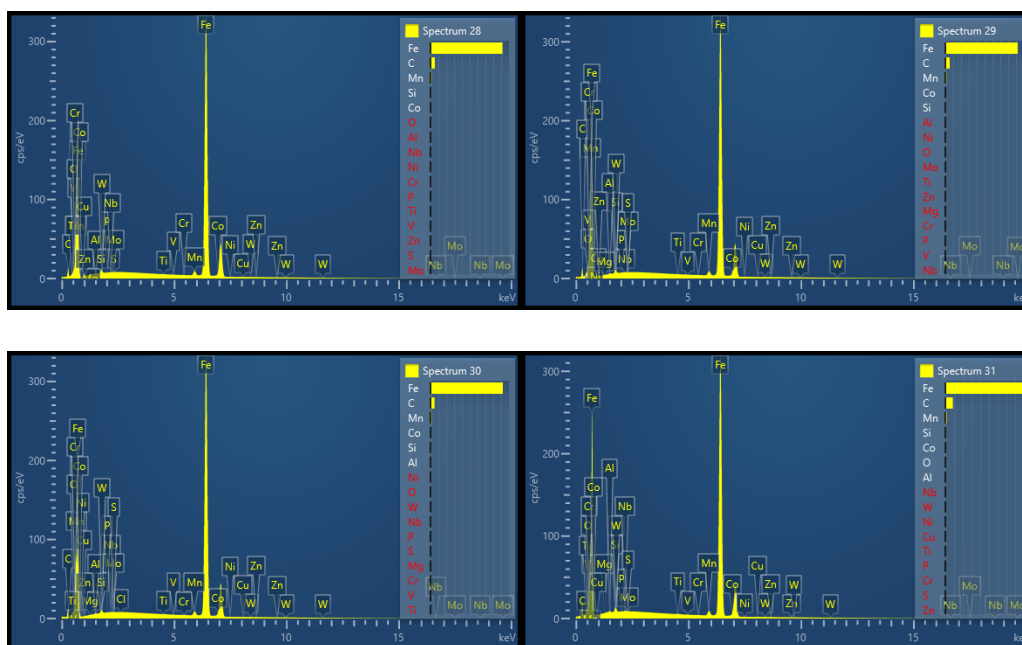


Figure C-3 Result of EDX compositional analysis on S355J2+N – sample 3

Table C-1 EDX compositional results of ferrite and pearlite phases for Figure C-1, Figure C-2 and Figure C-3

Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
Spectrum 1	4.81	1.29	0.05	0.36	0.01	0.02	0.00	0.12	0.39	0.01	0.02	0.00	92.18
Spectrum 2	4.77	1.27	0.12	0.35	0.01	0.03	0.04	0.13	0.42	0.01	0.00	0.00	92.32
Spectrum 3	5.79	1.29	0.07	0.37	0.00	0.02	0.04	0.12	0.39	0.01	0.00	0.21	91.66
Spectrum 4	4.73	1.34	0.07	0.37	0.03	0.01	0.00	0.12	0.40	0.02	0.03	0.08	92.74
Spectrum 5	4.87	1.58	0.08	0.38	0.01	0.03	0.00	0.13	0.41	0.02	0.01	0.16	92.25
Spectrum 6	7.97	1.41	0.07	0.42	0.01	0.02	0.05	0.12	0.39	0.02	0.00	0.92	88.51
Spectrum 7	8.23	1.67	0.07	0.48	0.05	0.03	0.00	0.10	0.32	0.03	0.02	1.46	87.39
Spectrum 8	9.58	1.40	0.08	0.41	0.04	0.02	0.00	0.11	0.41	0.01	0.03	1.72	86.10
Spectrum 9	4.78	1.26	0.09	0.36	0.03	0.02	0.03	0.12	0.42	0.01	0.00	0.11	92.50

Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
Spectrum 13	10.92	1.51	0.08	0.34	0.07	0.01	0.00	0.12	0.42	0.02	0.03	1.31	85.05
Spectrum 14	6.14	1.40	0.11	0.35	0.04	0.00	0.07	0.10	0.41	0.01	0.00	0.07	91.26
Spectrum 15	8.55	1.29	0.06	0.34	0.03	0.02	0.02	0.11	0.38	0.02	0.00	0.11	89.01
Spectrum 16	6.35	1.44	0.07	0.39	0.03	0.01	0.00	0.13	0.45	0.03	0.01	0.00	91.05
Spectrum 17	5.82	1.39	0.11	0.34	0.02	0.03	0.00	0.11	0.41	0.03	0.00	0.09	91.61
Spectrum 18	5.86	1.42	0.08	0.39	0.05	0.02	0.00	0.13	0.43	0.02	0.03	0.03	91.49
Spectrum 19	5.71	1.29	0.11	0.35	0.02	0.03	0.00	0.11	0.41	0.02	0.03	0.05	91.82

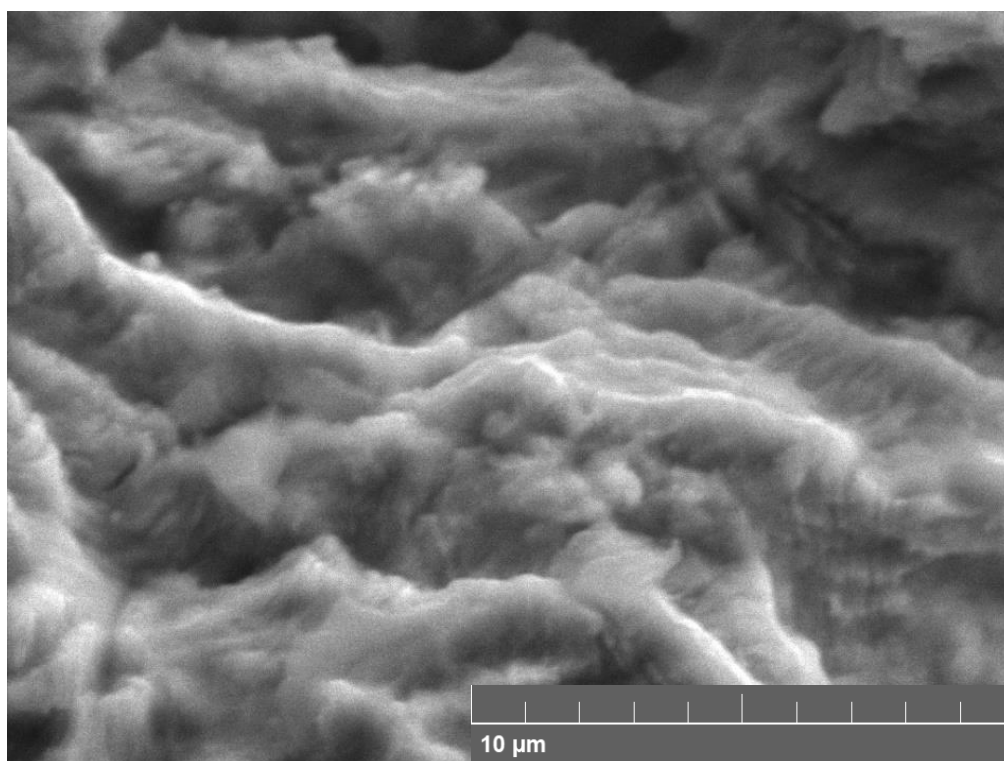
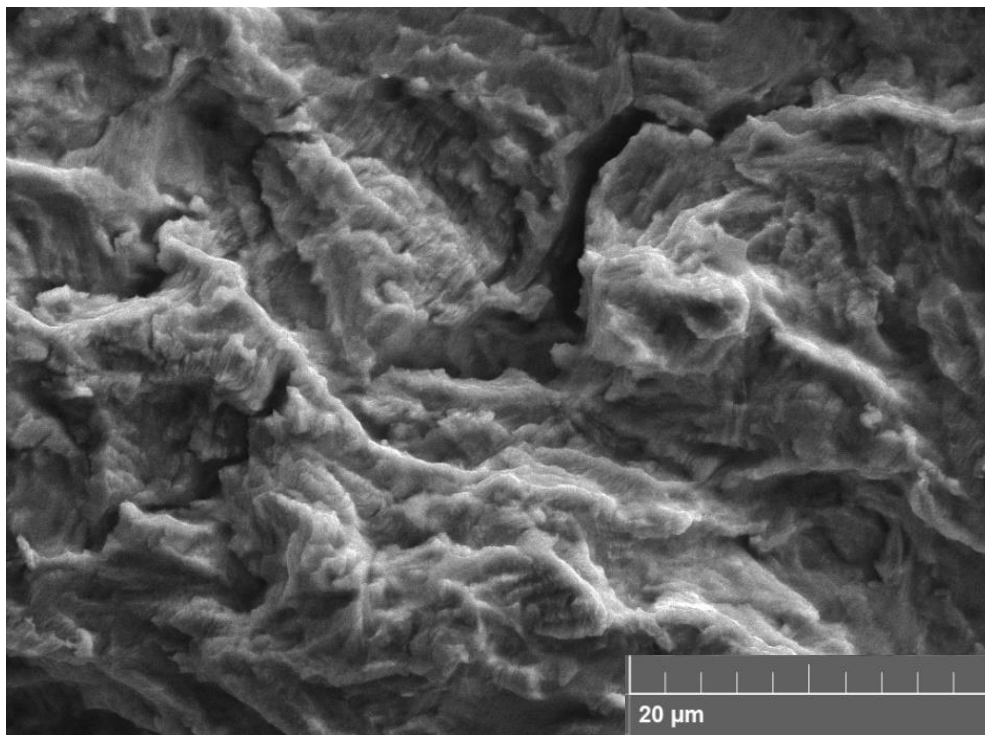
Spectrum Label	C	Mn	Ni	Si	Cu	Cr	Mo	Al	Co	P	S	O	Fe
Spectrum 20	9.97	1.32	0.09	0.51	0.06	0.03	0.00	0.21	0.38	0.02	0.01	0.34	87.03
Spectrum 21	5.15	1.37	0.08	0.38	0.00	0.02	0.00	0.12	0.36	0.02	0.01	0.07	92.33
Spectrum 22	10.03	1.33	0.07	0.47	0.03	0.02	0.02	0.19	0.37	0.03	0.00	0.78	86.38
Spectrum 23	5.65	1.38	0.11	0.36	0.01	0.01	0.00	0.09	0.38	0.02	0.00	0.13	91.63
Spectrum 24	4.95	1.66	0.09	0.44	0.02	0.01	0.00	0.11	0.43	0.02	0.00	0.02	92.18
Spectrum 25	5.34	1.43	0.08	0.39	0.02	0.02	0.00	0.12	0.42	0.02	0.01	0.16	91.91
Spectrum 26	9.94	1.49	0.10	0.34	0.13	0.02	0.00	0.09	0.37	0.02	0.03	1.51	85.91
Spectrum 27	9.77	1.53	0.10	0.39	0.01	0.02	0.00	0.12	0.40	0.04	0.01	0.33	91.12
Spectrum 28	5.55	1.40	0.06	0.37	0.00	0.03	0.00	0.11	0.37	0.03	0.01	0.13	91.82
Spectrum 29	5.47	1.37	0.09	0.38	0.00	0.00	0.03	0.12	0.40	0.00	0.00	0.06	92.01
Spectrum 30	5.15	1.36	0.11	0.36	0.00	0.02	0.00	0.13	0.39	0.03	0.02	0.08	92.22
Spectrum 31	9.65	1.44	0.09	0.48	0.07	0.02	0.00	0.15	0.41	0.03	0.02	0.36	87.96

Appendix D : Unit conversion and equivalent used in this study.

Unit Conversion in this study			
To convert B to A, multiply by	To convert A to B, multiply by	A	B
39.3701	0.0254	in	m
0.101972	9.80665	Kgf/mm ²	MN/m ²
0.14504	6.89476	ksi	MN/m ²
3.22462	0.31011	Kgf/mm ^{3/2}	MPa m ^{1/2} (MN/m ^{3/2})
0.91005	1.09884	ksi √ in	MPa m ^{1/2} (MN/m ^{3/2})
0.02877714	34.7498	ksi √ in	N/mm ^{3/2}
0.03162278	31.62278	MN/m ^{3/2} (MPa √ m)	N/mm ^{3/2}
1000	0.001	mm/cycle	m/cycle
Unit equivalent used in this study			
Pa = N/m ²	MPa = MN/m ²	MPa = N/mm ²	MPa √ m = MN/m ^{3/2}
1 inch = 25.40mm	1000mm = 10m	1 ksi = 6.896MPa	1 MPa = 10 ⁶ Pa

::

Appendix E : Ductile striation mechanism in the Paris Region



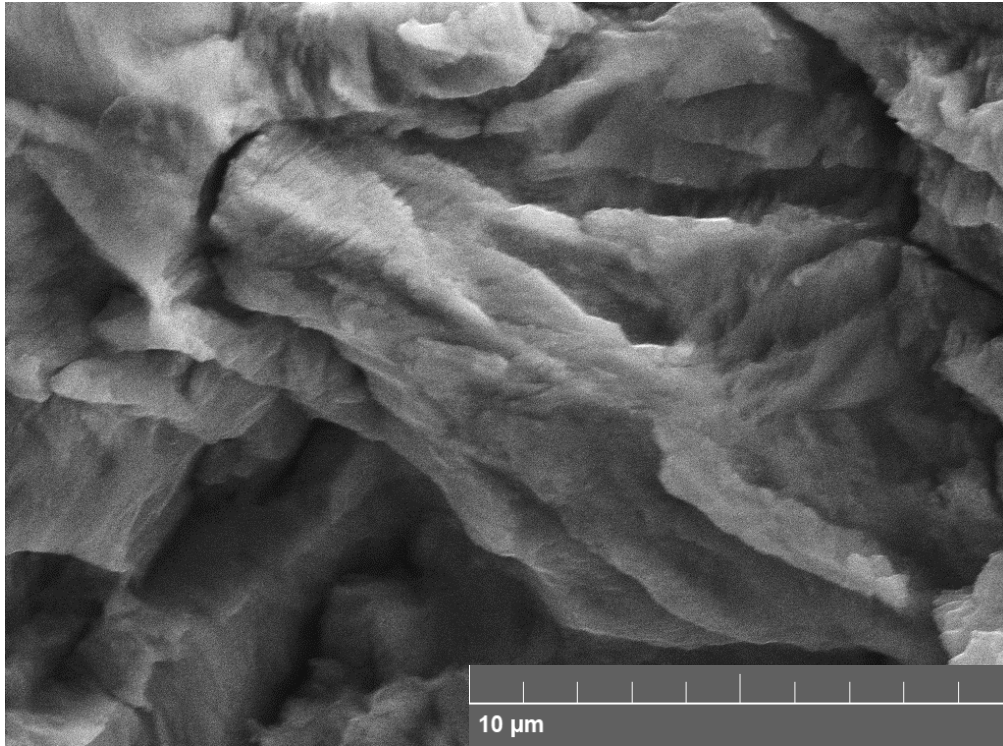
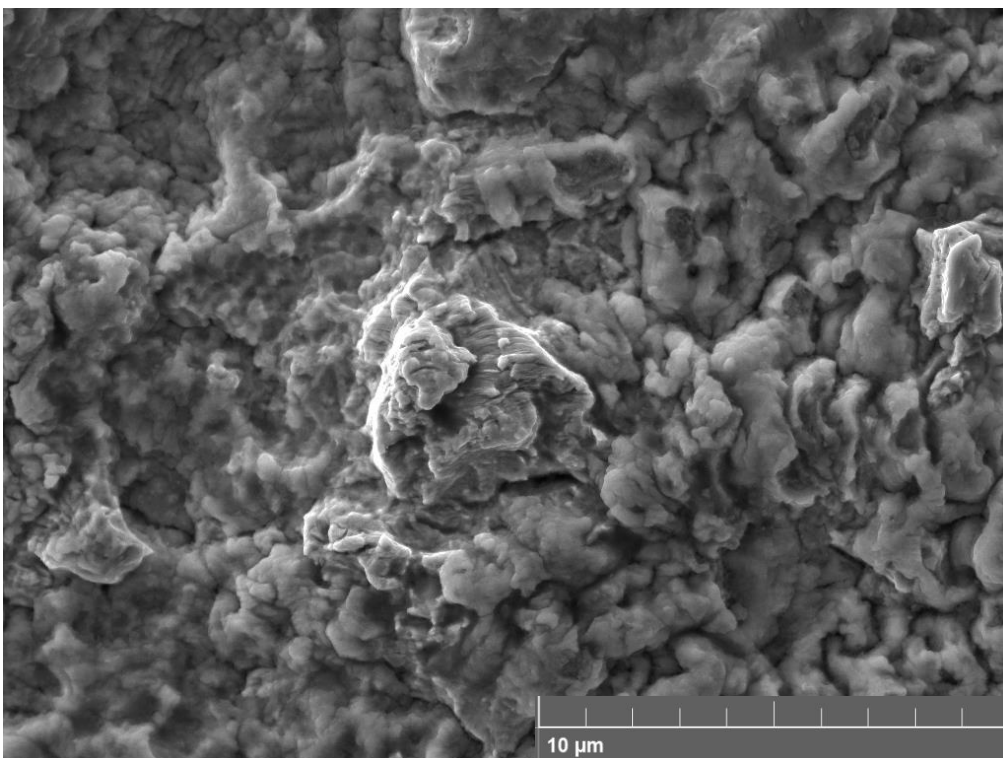


Figure F-1 For J2N, 10kN, 5Hz (air)



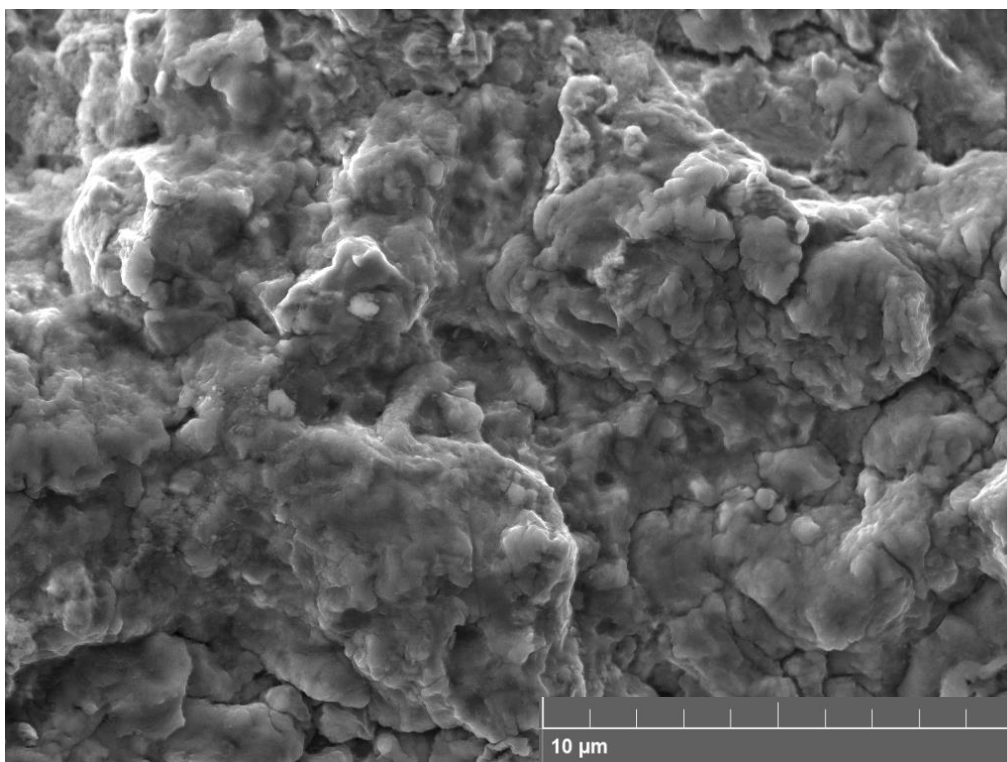
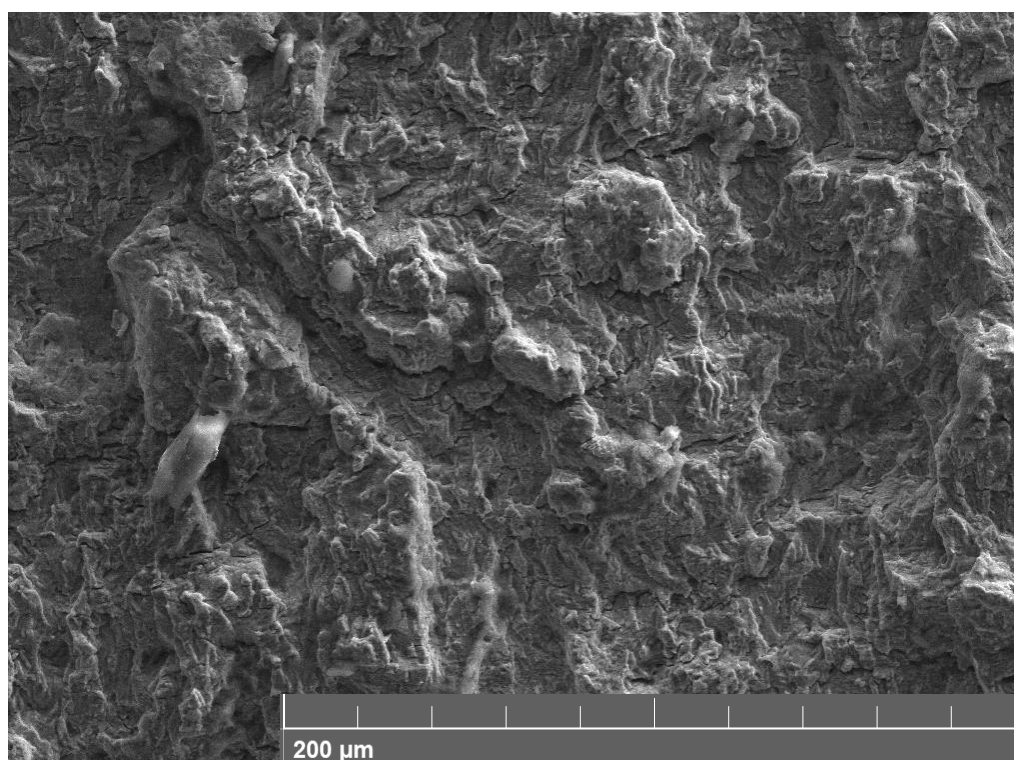


Figure F-2 For J2N, 10kN, 0.2Hz (seawater)



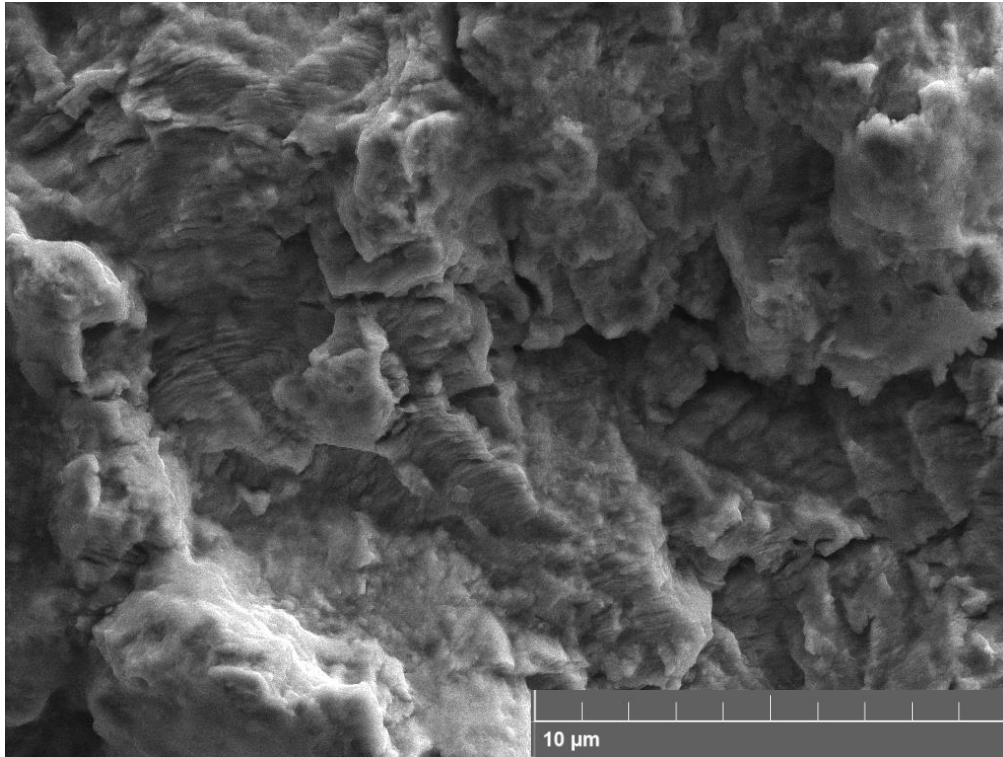


Figure F-3 For G8+M, 10kN, 5Hz (air)

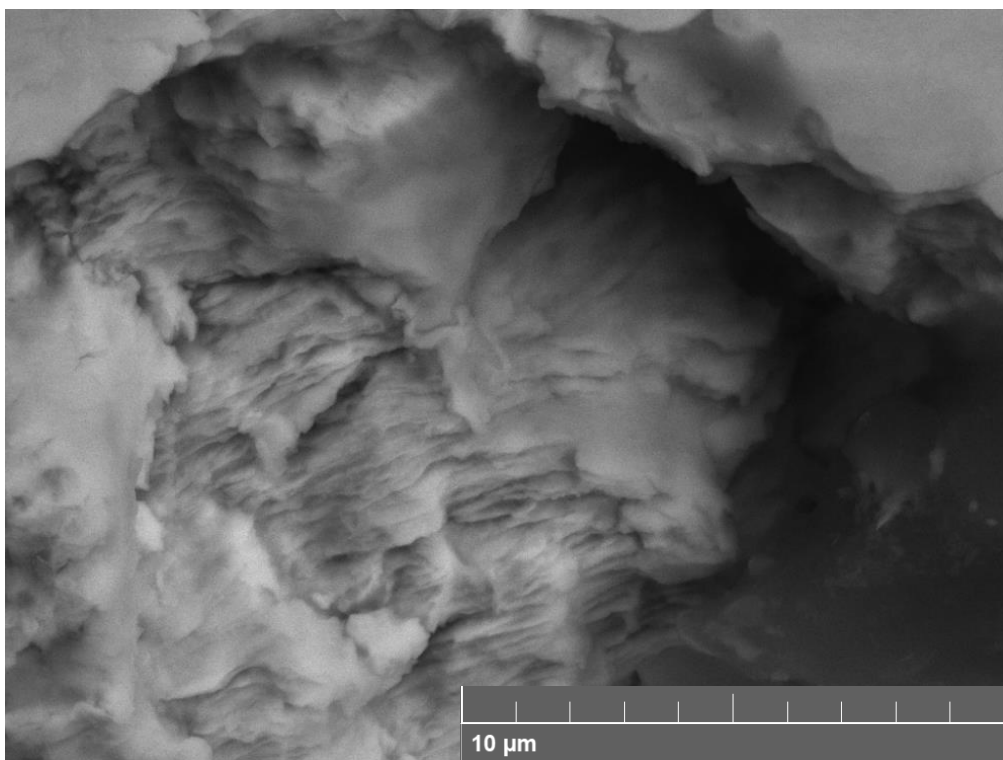
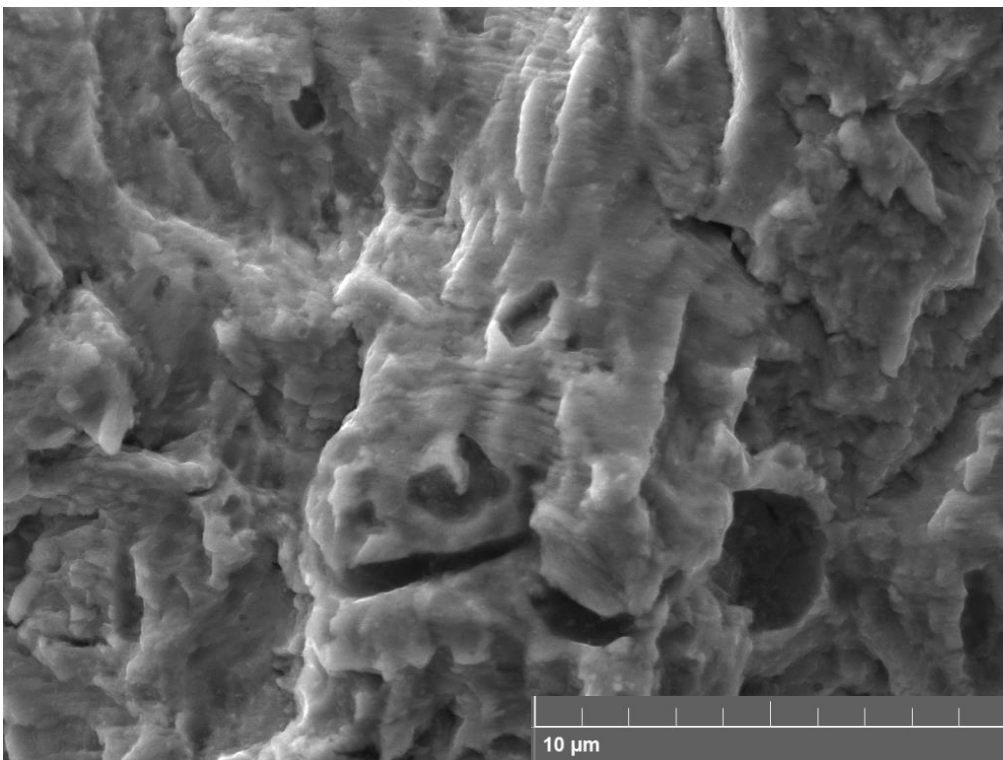
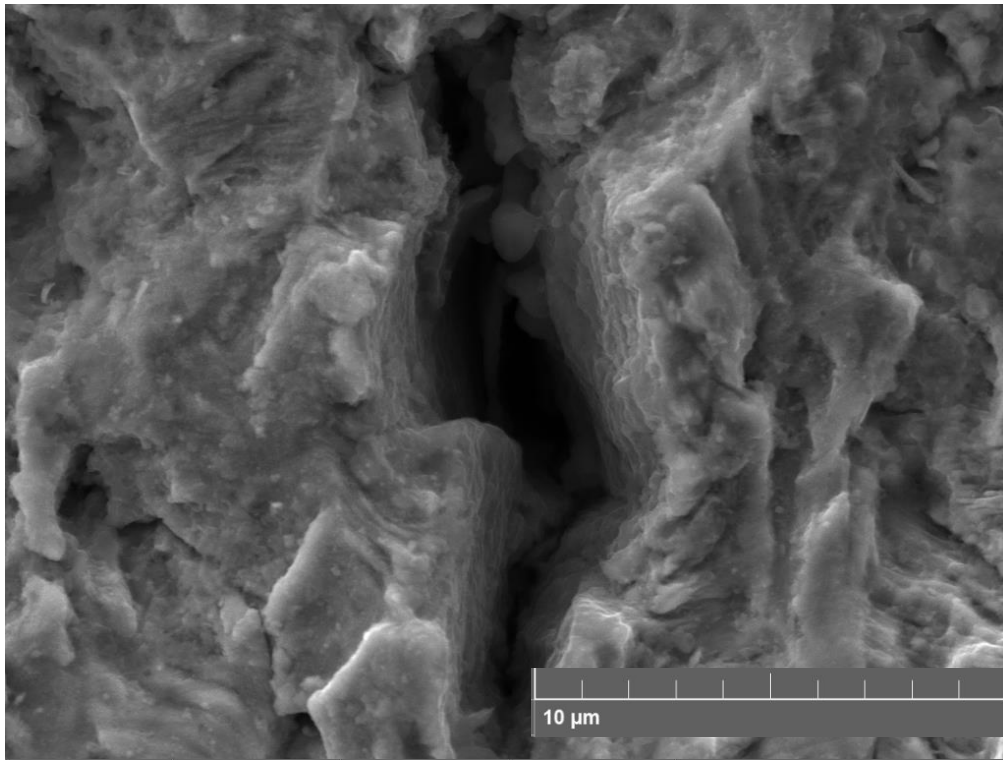


Figure F-4 For G8+M, 10kN, 0.2Hz (seawater)



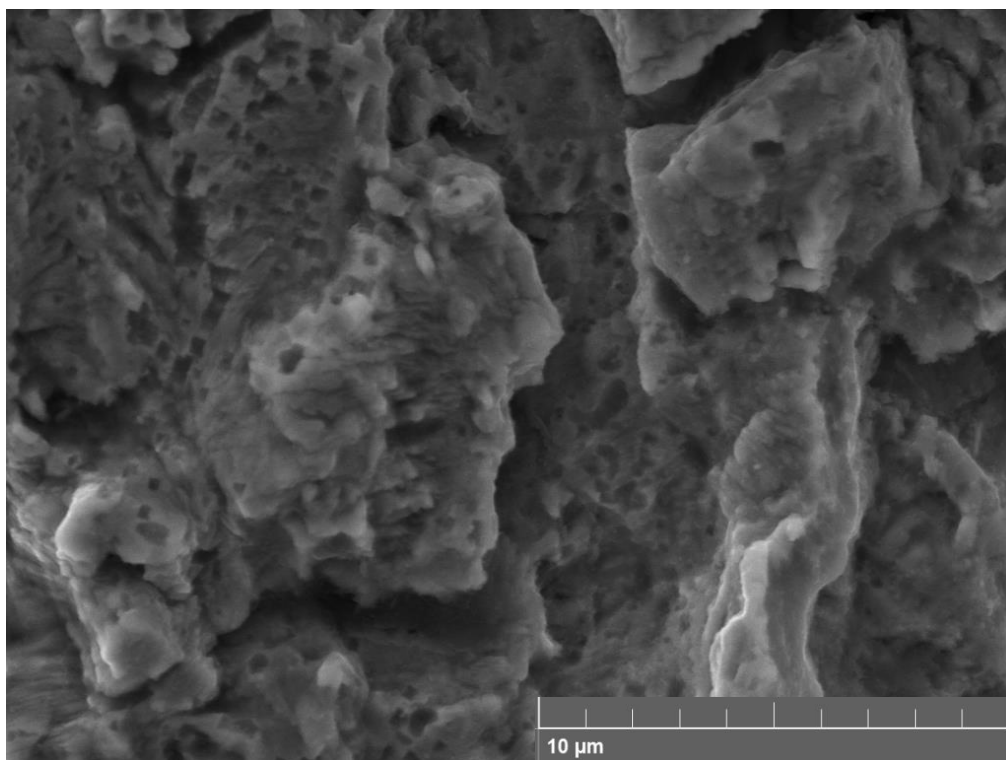
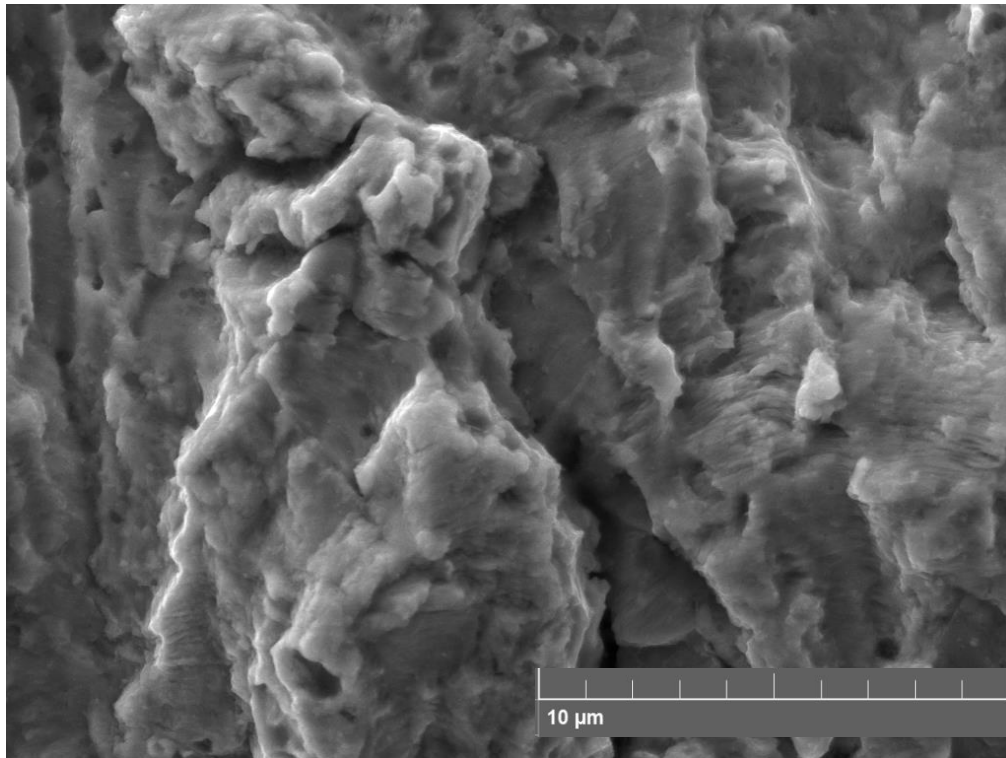


Figure F-5 For G10+M, 9kN, 0.2Hz sine (seawater)

Appendix F : Accumulation of corrosion product along a crack path

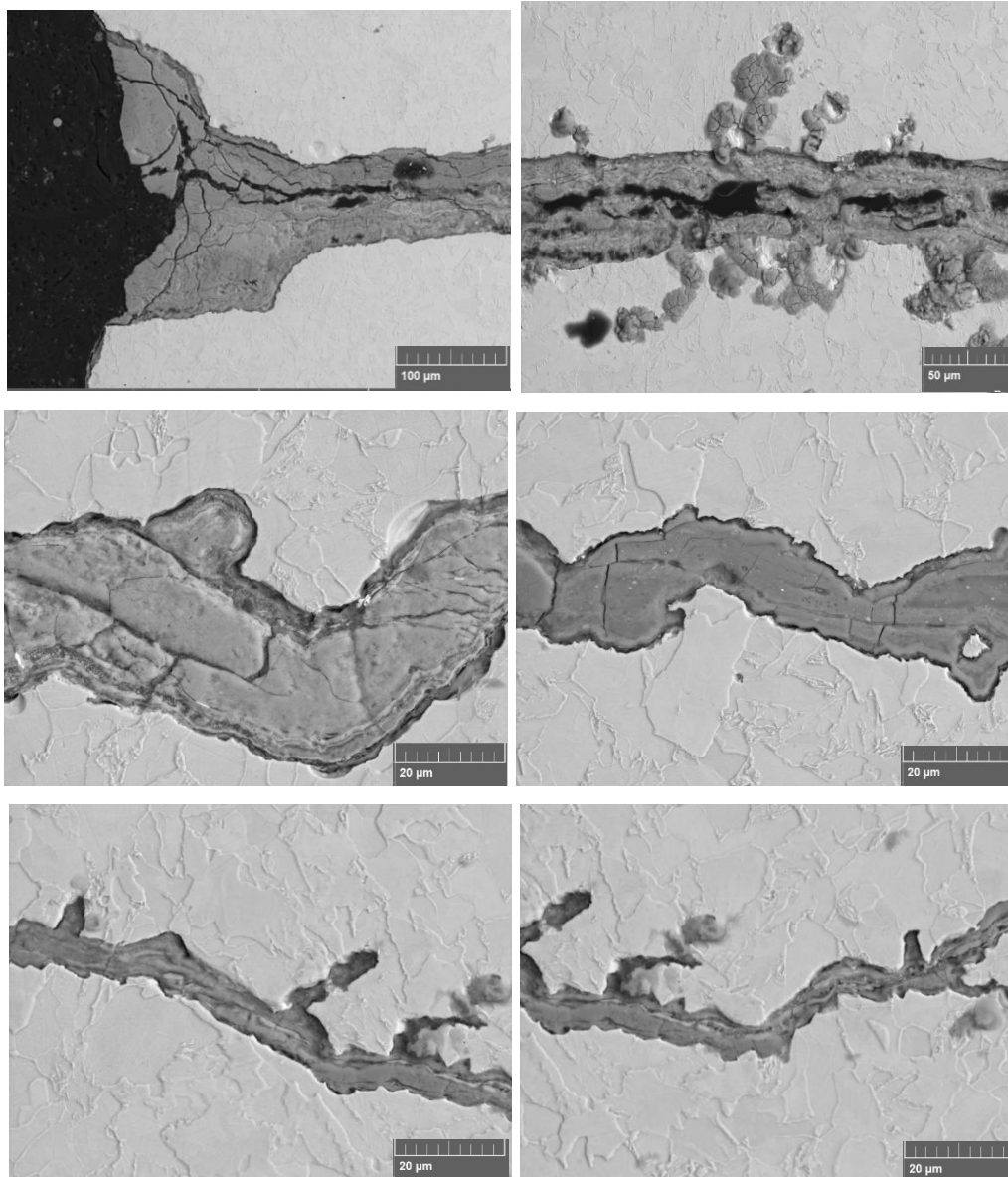
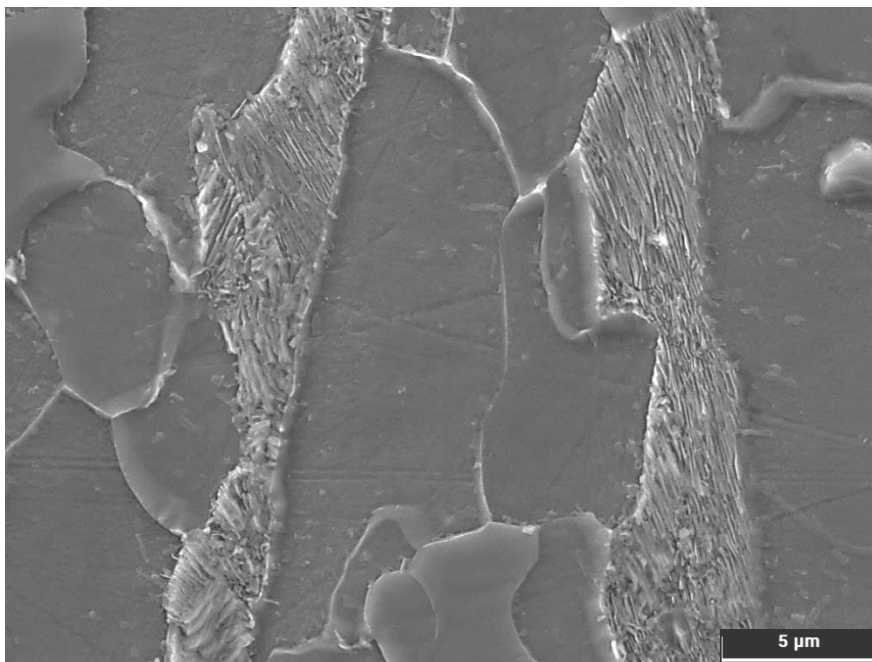
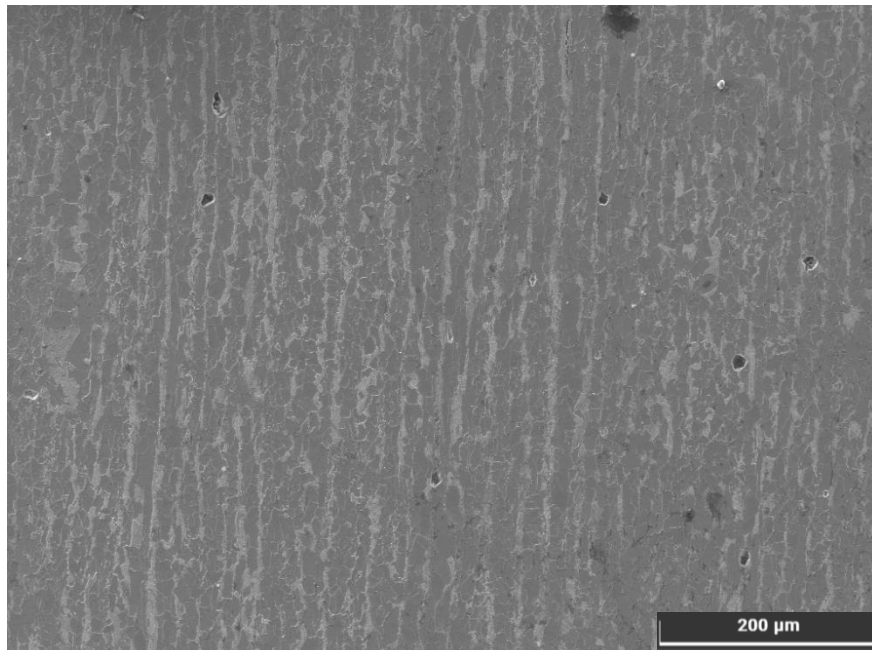
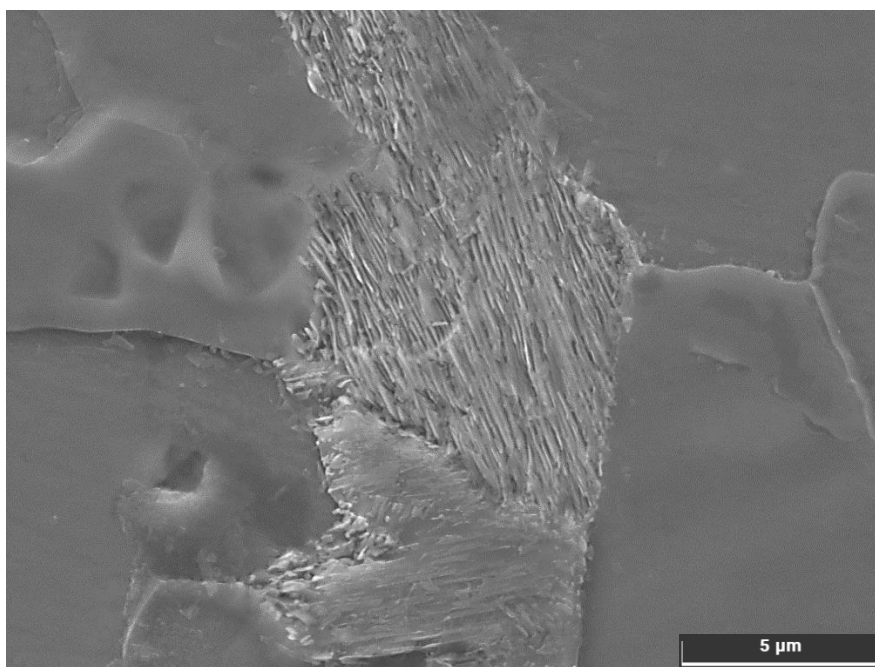
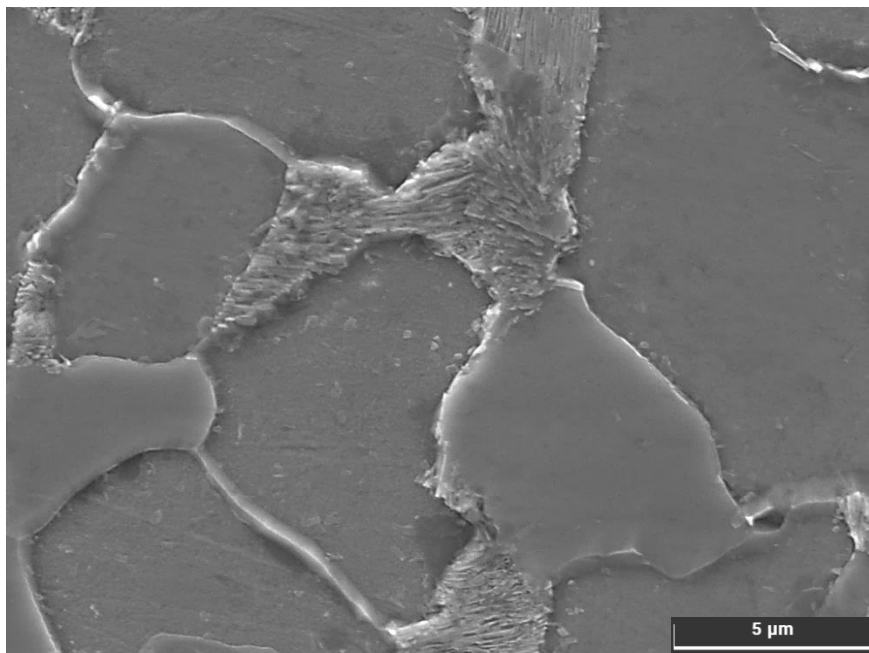


Figure F-1 For G10, 9kN, 0.2Hz - holdtime

Appendix G : Pearlite, P, low relief, high relief and ferrite ribbon in TMCP steel in the rolling direction





Victor's research philosophy:

- *Science is the discovery of how Nature behaves.*
- *Experiment remains the best method to study Nature.*
- *Making simple things complex is not intelligence*
- *Making complex things simple is the apex of academic intelligence.*