

CRANFIELD UNIVERSITY

PHILIP DIRISU

**DEVELOPMENT OF WIRE + ARC ADDITIVE MANUFACTURE
FOR LARGE SCALE APPLICATION FOR THE ENERGY
INDUSTRY**

SCHOOL OF AEROSPACE, TRANSPORT AND
MANUFACTURING

Welding Engineering And Laser Processing Centre

PhD in Manufacturing
Academic Year: 2019

Supervisor: Dr. Supriyo Ganguly
Associate Supervisor: Dr. Filomeno Martina

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ABSTRACT

Wire + arc additive manufacture (WAAM), as the most discussed tool in manufacturing, can complement the design of a fully optimised structure through multi-material deposition and incorporation of geometric design features not possible by traditional manufacturing technologies. Despite all the progress made in WAAM, detail studies on the implication of WAAM parameters on the structural integrity of steel-built components in both machined and as-built condition, and their fracture mechanics properties are largely unknown. To effectively and economically commercialise WAAM, the integrity and reliability of built components under specific loading conditions must be assured.

In these studies, the properties of steel components built with WAAM were determined, and the possible methods to improve strength, fatigue performance, and fracture toughness, by tailored deposition of different wire grades, through the modification of the build surface from rolling, and through the addition of cold wire during deposition were determined. The results were presented through mechanical test data (tensile properties, hardness, Charpy and fracture toughness, and fatigue crack growth rate and fatigue life results, along with microstructural analysis (electron microscopy of fractured surfaces, EBSD, XRD, and optical microscopy), and the surface topography of the WAAM build, as characterised by the surface waviness.

The studies showed that WAAM process parameters are responsible for the fatigue initiation of WAAM as-deposited (ASD) structures and that reduction of SW by process parameter optimisation and the use of side rolling will improve the fatigue life of ASD WAAM components. The possible cause of variability in fracture toughness in the WAAM steel deposits is the hardening and softening regions in the layer bands caused by the variation in peak temperature and cooling rates of the adjacent layer in the building direction. The addition of cold wire into the melt pool mop-up the excess heat, controlled the surface morphology and improved the mechanical properties and the deposition rate. In this study, a deposition rate of 6.27kg/hr was achieved.

In conclusion, low carbon and high strength steel wires will be suitable for making bespoke components with WAAM application for large scale energy manufacturing, and WAAM mild steel components could be used in the as-deposited condition for engineering application when rolling is applied to reduce its SW.

Keywords:

Wire assisted additive arc manufacturing (WAAAM)

Fracture mechanics

Low carbon and high strength steel

High strength low alloy (HSLA) steel

Microstructural characterisation

Mechanical characterisation

Graded structures

Fatigue life

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7. **Philip Dirisu**, Supriyo Ganguly, Xiaoyong Zhang, Victor Igwemezie, Ali Mehmanparast, The role of microstructural Anisotropy on the crack tip opening displacement of wire + arc additively manufactured HSLA steels, Additive Manufacturing (to be submitted in March 2020)
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LIST OF ABBREVIATIONS

WFS	Wire feed speed	HAZ	Heat affected zone
ASD	As deposited	CW	Cold wire
AM	Additive manufacture	WAAM	Wire + arc additive manufacture
ASTM	American Society of testing and material	WAAAM	Wire assisted arc additive manufacture
B	Bainite	FCC	Face centred cubic
CG	Coarse grain	BCC	Body centred cubic
CTOD	Crack tip opening displacement	PAD	Plasma arc deposition
DIC	Digital image correlations	CMT	Cold metal transfer
EBSD	Electron backscatter diffraction	TBL	Tempered bainitic lath
FG	Fine-grain	LCOE	Levelised cost of Energy
FL	Fatigue life	DED	Direct Energy Deposition
GBF	Grain boundary ferrite	CAPEX	Capital expenditure
HI	Heat input	OPEX	Operating expenditure
CMT	Cold metal transfer	GMAW	Gas metal arc welding
HSLA	High strength low alloy steel	LPBF	Laser powder bed fusion
LM	Lath martensite	EBM	Electron beam melting
M-A	Martensitic austenite	LMD	Laser metal deposition
MHP	Machine hammer peening	LENS	Laser Engineered Net Shaping
MIG/MAG	Metal inert /active gas	EBAM	Electron beam additive manufacturing
OEM	Original equipment manufacturers	mBAAM	Metal big area additive manufacturing
PAW	Plasma arc welding	SLS	Selective laser sintering
PGF	Polygonal ferrite	SLM	Selective laser melting
QPF	Quasi polygonal ferrite	UFGPF	Ultra-fine grain polygonal ferrite
R-A	Retained austenite	DBTT	Ductile to brittle transition temperature

List of Abbreviations

SR	Surface roughness	CWTSAW	Cold wire tandem submerge arc welding
TB	Tempered bainite	PWHT	Post weld heat treatment
TEM	Transmission electron microscopy	FN	Ferrite number
TIG	Tungsten inert gas welding	PGMAW	Pulsed gas metal arc welding
TM	Tempered martensite	CTWD	Contact tip to work distance
X-Dir	Deposition direction	TWFS	Total wire feed speed
Z-Dir	Built direction	EDS	Energy dispersive spectrometry
XRD	X-ray diffraction	TS	Travel speed
RF	Rear feeding	<i>WFS/TS</i>	Wire feed speed/travel speed

Nomenclature

a_o	Initial crack	a_l	Instantaneous crack length
σ_{ys}	Yield stress	δ	Crack tip displacement
σ_{UTS}	Ultimate tensile stress	a_w	Widmanstätten ferrite
σ_{max}	Maximum stress	θ	Cementite
$\Delta\sigma$	Cyclic stress amplitude	SCF	Stress concentration factor
Δ_k	Stress intensity factor range	DR	Deposition rate
K	Stress intensity factor-SIF	WD	Wire diameter
SIFR	Stress intensity factor range	K_Q	Candidate fracture toughness
Y	Geometrical correction factor	J_{1c}	J integral fracture toughness
a_d	Crack depth	r_{pc}	Cyclic plastic zone
da/dN	Fatigue crack growth (m/cycles)	AF	Acicular ferrite
P_{max}	Max load	QPF	Quasi polygonal ferrite
P_{min}	Min Load	EPFM	Elasto-plastic fracture mechanics
R	Stress or Load ratio	LLD	Load line displacement
f_{min}	Volume of other phases	FCGR	Fatigue crack growth rate
f	Volume of the ferrite phase	W	Width of specimen
E	Young modulus	B_o	Uncracked ligament
ε^*	Stress free strain	B_N	Net thickness
R	Radius of yield zone	A_{pl}	Plastic area under the curve
ν	Poisson ratio	B_e	Specimen effective thickness
γ_o	Yield strength of ferrite	m	Material constant
θ	Angle from X1-X2 plane	LEFM	Linear elastic fracture mechanics
σ_A	Yield stress of ferrite	B	Specimen thickness
r	Radii of a thin shell	C & m	Paris constant

Nomenclature

γ_h	Yield strength of the harder phase	OS	Oscillatory strategy
γ_s	Yield strength of soft phase	SP	Single-pass strategy
$\sigma \bar{A}$	Weighted average yield stress	J_Q	Energy resistance to fracture toughness
R_a	Mean surface roughness	H_h	Hard zone
R_v	Maximum valley depth	B_s	Layer band spacing
R_p	Maximum peak height	σ_{eR}	Fatigue limit
R_t	Maximum height of the profile	$\alpha^1\text{-}\gamma$	Martensitic austenite
R_q	Root mean square surface roughness	P	Penetration
SW	Surface waviness	BW	Bead width
T_{ww}	Total wall width	BH	Bead height
E_{ww}	Effective wall width	AR	Aspect ratio
a	Notch depth	RR	Re-melt ratio
$\dot{\rho}$	Notch radius	CA	Contact angle
LH	Layer height	α^1	Martensite
RR	Re-melting ratio	D_{Rate}	Deposition rate
$\beta 1 \ \& \ \beta 2$	Excluded angles in WAAM beads	FF	Front feeding
ℓ	Density of steel	SF	Side feeding
S_{el}	Elastic stress concentration factor	K_f	Fatigue strength reduction factor
S_{el}	Effective stress concentration factor	r_p	Monotonic zone
PCE	Percentage elongation	C_R	Cooling ratio
q	Notch sensitivity	S_f	Shape factor
N_f	Cycles to failure	γ_{SG}	Interfacial tension of the air to substrate interface
σ^f_{YS}	Material yield stress at fatigue pre-crack	γ_{SL}	Interfacial tension of the melt pool to substrate interface

Nomenclature

σ^T_{YS}	Material yield stress at test temperature	γ_{LG}	Interfacial tension of the air to melt pool interface
K_{max}	Maximum stress intensity	θ_c	Contact angle
P_m	Pre cracking force	LAGBs	Low angle grain boundaries
J_{el}	Elastic component of J	HAGBs	High angle grain boundaries
J_{pl}	Plastic component of J	α_{ac}	Acicular ferrite
J_i	Instantaneous fracture resistance	α_B	Bainitic ferrite

1 INTRODUCTION

1.1 Background

1.1.1 UK wind energy sector

The EU has recorded a substantial increase in offshore wind turbine projects in recent decades, coupled with the competitiveness of turbine manufacturers and low cost of oil and gas in the past two years. The falling capital cost and improvement in designs and material selection have also made way for the wind energy sector to compete with other renewable energy sectors, such as the fossil fuels and hydropower, which are faced with incessant peak times [1]. In the US, the annual electricity generation increased from 2.5GW to more than 35GW during the period from 2000 to 2009 [2].

Wind power is now the cheapest form of energy produced in the UK when compared to coal-powered stations, oil and gas. Onshore wind is currently producing 60% of the UK's wind power output and is far cheaper compared to offshore wind turbine energy in which the current government has a keen interest. 1.9GW was added to the UK renewable sector in the year 2013 while 648MW was added to the onshore wind energy sector in the year 2014/2015. This was a reduction of 452MW from the previous year, although the currently ongoing construction of onshore wind power project summed up to 2GW [3]. The metallic components of the turbine cost over 34% of the entire turbine construction. Harnessing improved methods of manufacturing this huge percentage of structural components will save a great deal of cost.

In a bid to meet the vision 2020 target by the UK government concerning the renewable energy industry, there has been a huge investment and development in the onshore and offshore wind energy sectors. However, development in the offshore wind energy sector is expensive due to its move to the deep waters with depths above 45m; here bigger foundations, bigger towers and their components are required for producing greener electricity [4][5] [6]. The future growth of offshore wind energy will depend on the construction cost of 1 MWh of energy. Based on the cost reduction monitoring framework for the offshore wind

programme board, the levelised Cost of Energy (LCOE) was projected at £100/MWh to be achieved by 2020 but was achieved four years earlier than the target date. 11% reduction of this anticipated value was achieved as at the end of the year 2014 [3], and a drastic reduction from £142/MWh to £97/MWh was achieved in 2016. Scottish renewables targeted £49.4/MWh for their 2018/19 auction in real 2017 terms [7]. This auction is expected to mitigate 8 million tonnes of carbon dioxide in the next 25 years of operation. Wind farms accounted for 49.2% of all renewable energy generated in the UK for the second quarter of 2018.

The current largest offshore wind farm commissioned in the UK is Walney extension, which has a capacity of 659 MW with 87 wind turbines that can power 600,000 homes. Other offshore wind energy projects are under construction; these include the Hornsea project (1,218MW capacity on completion) and East Anglia one (714MW capacity on completion). The world offshore wind energy generation at the end of 2013 is shown in Figure 1-1, depicting that over 50% is being generated in the UK. The 2030 vision of achieving 40GW of offshore wind energy in the UK to meet the National European renewable electricity target has led the UK government to support this sector by committing to increase subsidies and biannual contracts for variation in auctions to increase offshore wind capacity [8][9].

Despite the government's commitment, additional costs due to manufacturing processes (development of new materials that could meet the strength and toughness requirements of deep offshore application) are still a constraint to investors. Supply chain & logistics delays as a result of increasing sizes of offshore turbine components and installation costs of these heavy components in the deep offshore are still constraints to original equipment manufacturers (OEMs), constructors and operators [10][11][12]. Some of these turbine components might require topological optimised structures to reduce their weight to meet current loading conditions. This will further ease the installation process as the requirement for heavy lifting equipment will be bypassed.

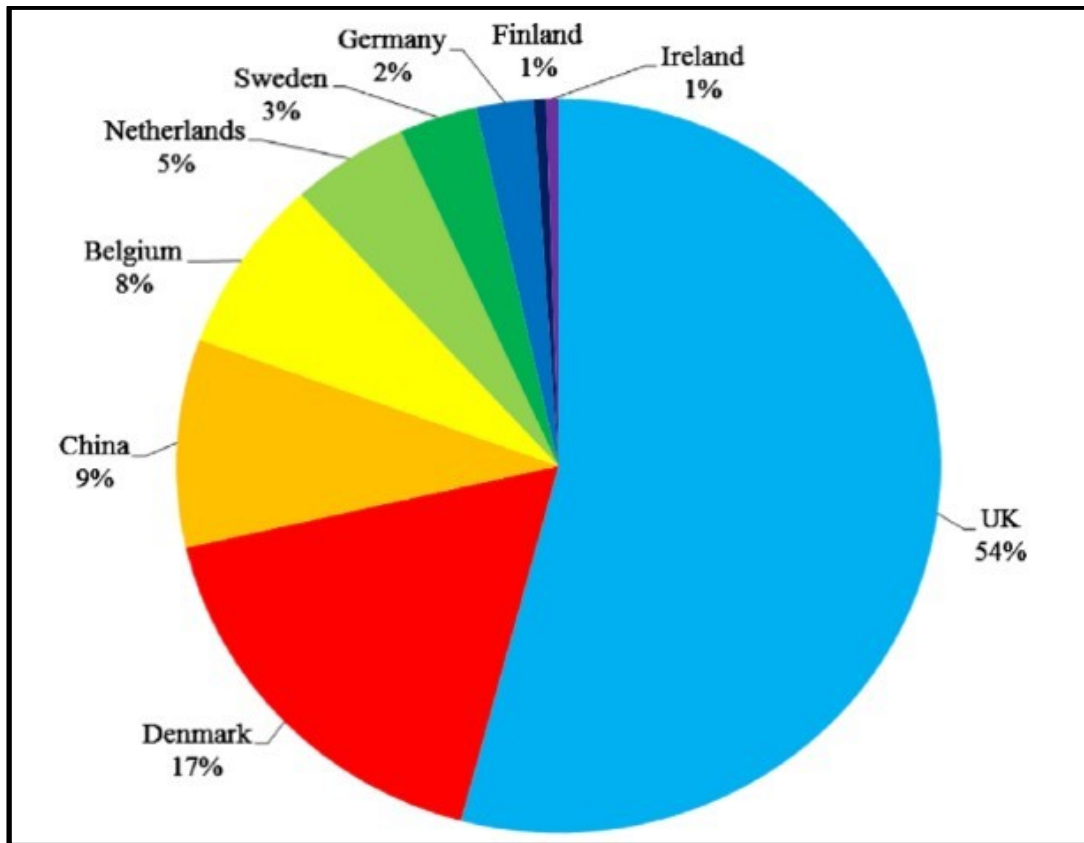


Figure 1-1 International breakdown of installed offshore wind capacity [8]

1.1.2 Current manufacturing technique for the wind turbine mainframe

The current manufacturing process which further requires a subtractive manufacturing process leaves close to 10% extra material in the component that must be machined off, as shown in Figure 1-2 and Figure 1-3 [13]. The flexibility for topological optimisation within the subtractive manufacturing process is limited; hence only single material is being used in the process, and this is the reason why about 40% of non-valuable material is contained in the wind turbine mainframe [13].

Sand casting technology might not be able to handle the trend of new material being proposed for the wind turbine mainframe due to the high cost implication. The cost of the sand-casting technique is relatively high, and it would not make any sense to adapt them for a cheap manufacturing process, such as additive manufacturing. Most of the manufacturing companies incur huge costs in moving

the 40% over-heavy mainframe on haulage and transport logistics because the manufacturing processes are done by different groups that operate far apart and taking the product from one group site to another is time-consuming and cost eroding. Thus, further reduction in the cost of a wind turbine project will be enhanced by improvement in manufacturing techniques for the most used turbine members, which are, the mainframe, hub, tower flanges and drive gear trains.

The total installation cost of a 5MW offshore wind turbine is about £6m [14], where the average cost of the nacelle, rotor, and tower installation are £2.5m, £1.3m and £1.2m, respectively; this adds up to over 83% of the installation cost. About 93% of the nacelle components comprising the gearbox, generator, and the mainframe, which is built from steel, account for 12.8% of the entire weight. Also, 100% of the rotor materials are currently built from steel, and this also accounts for 6% of the entire weight while 98% of the tower material is made from steel which accounts for 73.6% of the entire weight.

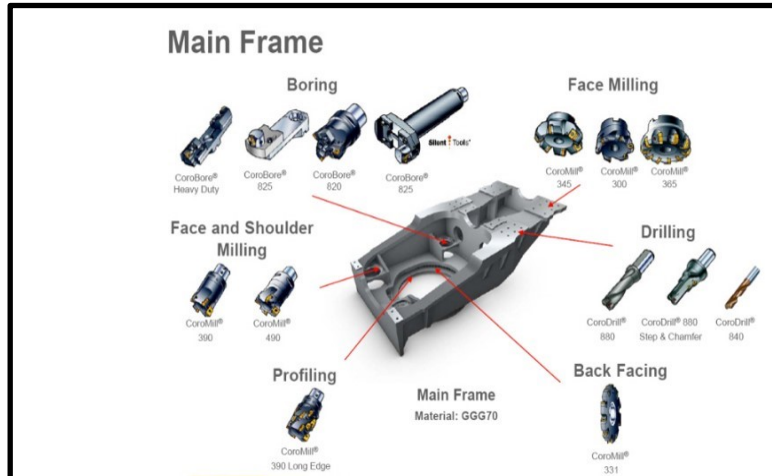


Figure 1-2 Manufacturing process of a wind turbine mainframe [15]

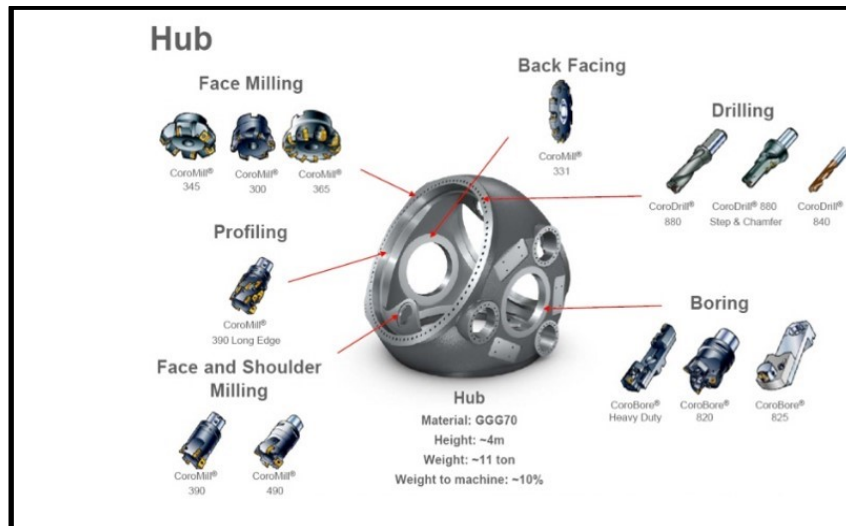


Figure 1-3 Manufacturing process of wind turbine hub [15]

1.1.3 Additive manufacturing

Additive manufacturing (AM), as defined by ASTM standard[16] is the process of making a near net-shaped structure using a layer by layer deposition sequence [17]. AM is a cost-saving approach for the fabrication of parts made from expensive materials such as titanium and nickel alloys etc. [18]. AM processes can be high deposition processes with much lower manufacturing cost compared to subtractive manufacture [19]. The traditional classification of different AM systems is demonstrated in Figure 1-4. They can be classified according to the energy sources, namely, beam or arc source delivery, and feed stock, namely wire or powder delivery. The beam sources delivery could be by laser beam or electron beam, which are currently being used by some AM equipment makers. The arc sources delivery can use the tungsten inert arc, plasma arc or MIG arc [20]. The powder feed system has a slightly higher build volume when compared to the powder bed systems; this system requires the conveying of powder through a nozzle into the build zone and the help of a laser source or electron beam source to melt the powder into shapes as desired. The wire feed system currently possesses higher build volume compared to other feed systems as presented in Figure 1-5.

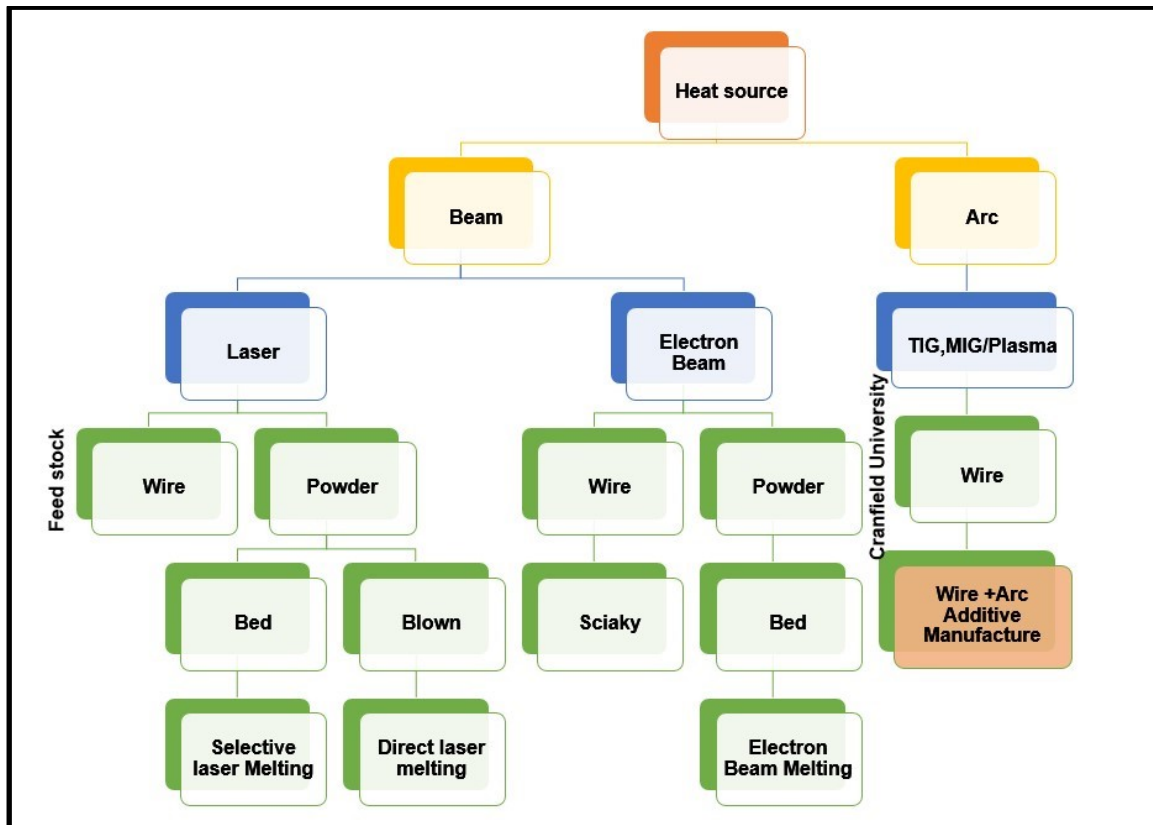


Figure 1-4 Classification of additive manufacturing systems

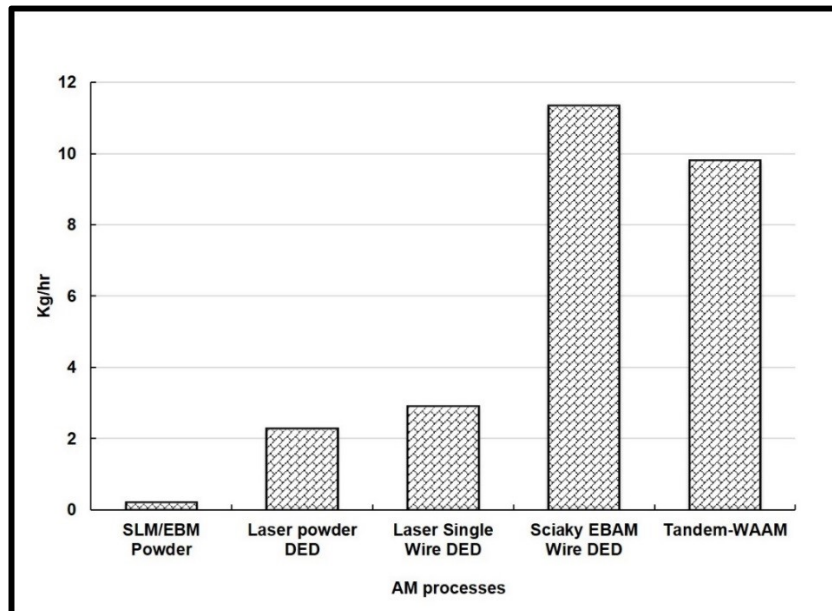


Figure 1-5 Metal additive manufacturing deposition rate [21] [22]

Table 1-1 Cost comparator for wire sizes and powder AM [21]

Material Feedstock	Titanium 6Al-4V	Tantalum	Inconel 625	Stainless Steel 316
Wire—0.035" Diameter (0.9 mm)	\$58/lb.	\$545.30/lb.	\$26.73/lb.	\$5.19/lb.
Wire—0.045" Diameter (1.1 mm)	\$54/lb.	\$545.30/lb.	\$23.30/lb.	\$4.63/lb.
Wire—0.062" Diameter (1.6 mm)	\$50/lb.	\$524.88/lb.	\$22.17/lb.	\$4.57/lb.
Wire—0.093" Diameter (2.4 mm)	\$48/lb.	\$502.30/lb.	\$21.43/lb.	\$4.81/lb.
Wire—0.125" Diameter (3.2 mm)	\$45/lb.	\$438.96/lb.	\$21.02/lb.	\$4.75/lb.
Wire—0.156" Diameter (4.0 mm)	\$44/lb.	\$438.96/lb.	N/A	\$4.69/lb.
Powder—AM Grade	\$120/lb.	\$522/lb.	\$48/lb.	\$10/lb.

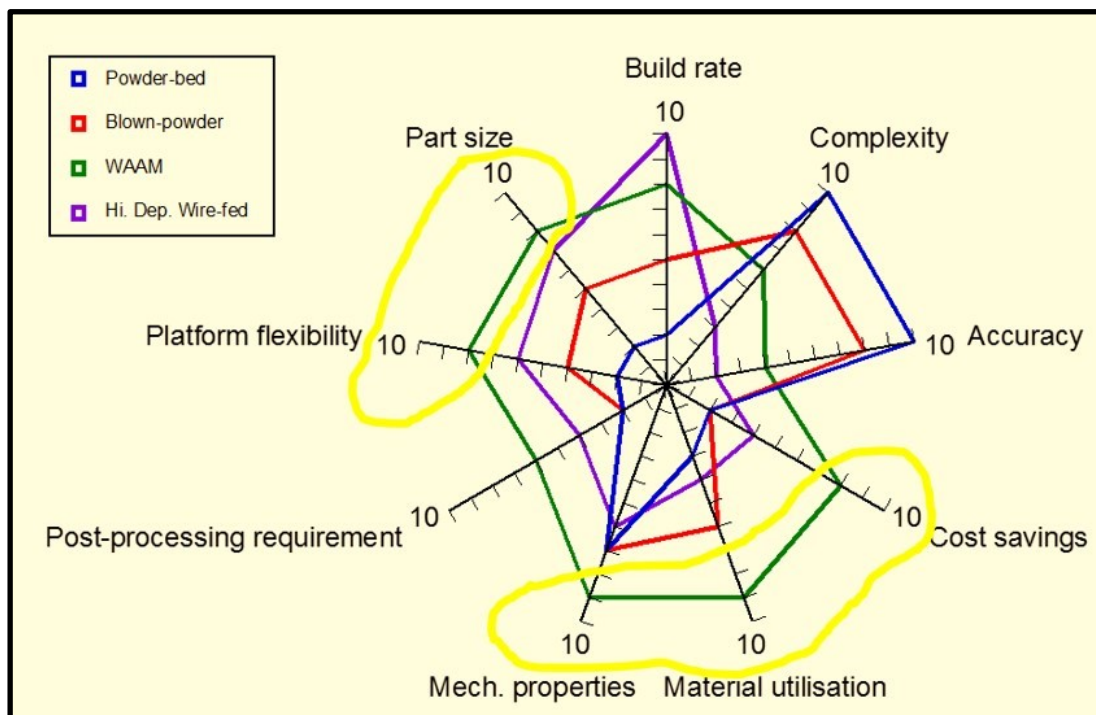


Figure 1-6 Assessment of AM systems [13]

Table 1-1 presents a cost comparator of wire and powder AM. Powders are more expensive than wire. Steel wire costs between £2/kg and £15/kg, depending on its diameter and alloy composition [23]. Powder technology could also possess some health hazards due to the possibility of inhalation and safety challenges due to its flammability and ease of ignition. It is obvious from the above; therefore, that wire AM is cheaper than powder AM. Figure 1-6 depicts the assessment of the various AM processes. WAAM promises to be the most viable AM process based on its unique advantages over the powder bed and blown powder.

1.1.4 Wire + arc additive manufacture

WAAM, as the most discussed tool in manufacturing, can complement the design of a fully optimised structure through multi-material deposition and incorporation of geometric design features not possible by traditional manufacturing technologies [24][25]. Another key unique advantage of WAAM in comparison to the powder bed/blown powder process is the high deposition rate potentials combined with its high material utilisation and lower cost. Thus, WAAM might be a game-changer for the renewable energy industry and suppliers to meet the current LCOE. Even with all these advantages of WAAM, its as-deposited condition is affected by surface waviness (SW), which might hinder its use in the as-deposited state.

Most of the current power sources used for WAAM application, such as TIG are slow deposition processes suitable for high-quality parts with a good surface finish, although affected by the likelihood of tungsten inclusions in a deposit. PAW is more suitable for small high-quality parts and better than TIG for WAAM due to the directionality of the arc and absence of tungsten inclusion possibilities. CMT, which is a cold deposition process, is also limited by its deposition rate, while conventional GMAW, although possessing the capability for high deposition rate, is faced with high arc energy that is not suitable for WAAM built [22].

Tandem GMAW on the other hand, is faced with three significant challenges to achieving high productivity, namely (1) the uncertainty in the mechanical properties and structural integrity of the final product due to heat accumulation, (2) slow cooling of the deposited material, and (3) the high cost of special jigs and

fixtures required with this process to cater for the enormous residual stress in-built as a result of heat accumulation [22]. The cost of the additional power source and power usage is another limiting factor to the use of tandem GMAW [26].

WAAM is proven to be the AM of choice when it comes to building large engineering components at a low cost in the aircraft and automotive industries. However, it has not been used in the production of structural steel for the renewable energy industry. The flexibility of controlling the effect of the thermal heat of the deposited layers by using cold wire addition into the melt pool of the high arc energy process, such as conventional GMAW, could improve the deposition rate, and make it an ideal manufacturing process for the renewable energy industry. The use of WAAM for making most of the integrity components such as the wind turbine mainframe, hub and driveshafts, which were previously made by the casting process with CNC machining, will be a viable option.

In total, 92.49% of the entire wind turbine structure is made of steel, so there could be a huge cost saving for manufacturers if the current manufacturing processes are harnessed to deploy layer by layer additive manufacturing technology, thus saving material wastage and the carbon footprint of the manufacturing industry. There is a great advantage in additive manufacturing of wind turbine components, such as the hub, driveshafts, bearings, mainframe, [27].

*The research project was coined out from Vestas's scope in the **Horizon 2020 project called the Large additive subtractive integrated modular machine-LASSIM-Proposal number; 72360**, this involves building a large dynamically loaded structural component (main Frame) which carries the loads from the wind turbine rotor and blades into the wind turbine tower and at the same time constituting the foundation for the wind turbines power train equipment, the gearbox and the generator. Vestas can sell close to 10,000 2 MW turbines per year if the costs are radically reduced and logistic challenges can be solved locally. The WAAM process is expected to enable Vestas in designing a main frame with at least 40 % reduced material weight fabricated on location, thus preventing handling and transportation costs and enable a higher energy capture*

of the turbine by taking advantage of the lower top mass weight and the high strength steels ability to carry larger rotor diameters [13].

1.2 Research motivation

The overall motivation of this research is to explore the manufacturing opportunities that WAAM offers. The primary research questions to be addressed here are:

- Can WAAM be used to manufacture bespoke steel components for the energy industry using plain carbon and low alloy structural steels and will the mechanical properties satisfy the requirements for such application?
- Can the WAAM as-deposited structures be deployed to service without machining? And if not, how can the as-deposited structures be made functional?
- Is it possible to deposit graded structures using plain carbon and low alloy structural steels and what will their mechanical behaviour be?
- What will the fracture mechanics properties of these WAAM structural steels be since they exhibit some form of ductility?
- Will the mechanical properties of this WAAM steels be better than ductile cast steels ?
- Can the deposition rate of WAAM be optimised to make the process more competitive and less expensive than other traditional manufacturing methods?
- How can the current challenges with heat distribution across intersections and WAAM deposition of complex structures as currently observed be addressed?
- Is it possible to control the cooling rate and surface waviness by adding cold wire? And can the deposition rate, mechanical and fracture toughness properties be optimised by this method?

1.3 Research Aim and Objectives

The aim of the research is to explore the optimised route of producing a multi-alloy and single alloy structural component and evaluate their suitability for application in the machined and as-welded condition, i.e. without any machining, thus proffering a cost-saving manufacturing solution to the ever-growing energy industry. The objectives are outlined below:

- To explore the application of WAAM for both single and multi-alloy components.
- To understand the influence of SW in an as-deposited structure on overall fatigue initiation performance and process a study on how to enhance the functionality of such an as-deposited structure.
- To understand crack microstructural interaction in WAAM produced parts through fracture mechanics concepts using J1c and how it can be modified to improve the properties.
- To explore how independently feed cold wire can be used to improve the manufacturing process and enhance the WAAM built and its properties.

1.4 Research outline and thesis structure

The methodological approach to the research is shown in Figure 1-7. The research started with a review of the literature on the Additive manufacturing process, WAAM strategies and WAAM applications relevant to the study and mechanical properties of WAAM steel deposits. Stage 1 of the research involves development of deposition parameters for different WAAM power sources and processes, which are entirely covered in Chapters 4 and 8. Stage 2 of the study, which includes building structures with the parameters developed in stage 1 and carrying out mechanical, fatigue and microstructural studies are fully covered in Chapters 4 and 7. While stage 3 of the research, which involves characterisation

of the factors controlling as-deposited response to fatigue initiation and cold work methodology to improve these conditions, are fully covered in Chapters 5 and 6. The overall structure of this PhD thesis is contained in nine chapters.

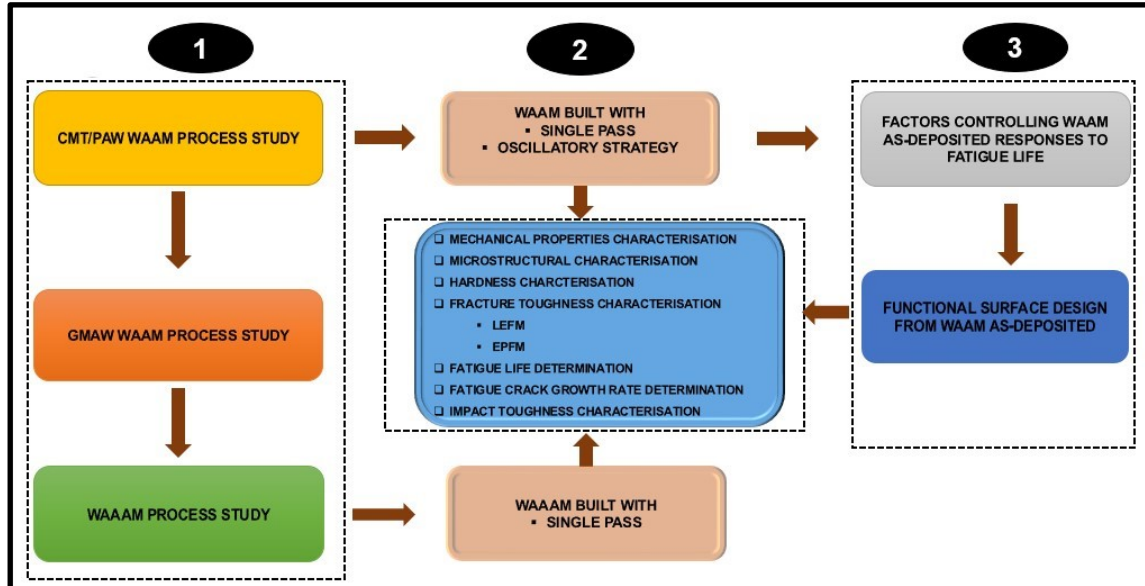


Figure 1-7 Outline of research methodology

Chapter 1 presents an overview of the research background, renewable energy background, challenges in modern manufacturing, research motivations and gaps, aims and objectives, and the thesis outline.

Chapter 2 presents a literature review which focuses on the main areas of this research; it also highlights additive manufacturing in the industry at a glance, current AM processes being used for large scale structure, WAAM deposition strategies, Microstructural variation in WAAM HSLA steel deposits, and the current welding process being used for WAAM. The mechanical properties of ductile cast iron used for main frame and hub fabrication were also reviewed. The effect of alloying elements on steel microstructure and effect of oxide inclusions on HSLA steel were reviewed and finally suggest improvement strategies for the WAAM process.

Chapter 3 details a pictorial and descriptive presentation of the experimental procedures; it also includes the base materials, power sources and welding

consumables used, the mechanical and metallography preparation of test samples, and the data acquisitions methods used.

Chapter 4 presents the detailed responses of WAAM structural steel to static and fatigue loading for both the single and graded structures. The microstructural variation in the single grades and graded structure are also studied.

Chapter 5 presents a process study for investigating the factors affecting the suitability of WAAM application in the as-deposited condition. The effects of the WAAM deposition parameter influence on the as-deposited response is also presented.

Chapter 6 presents the cold work process for improving the functionality of the WAAM as-deposited surface and the corresponding response of the functional surface to fatigue initiation.

Chapter 7 presents the investigation of the effects of the WAAM process parameters on the crack microstructural interaction in WAAM produced parts. Analysis with an elasto-plastic fracture mechanics concept using J1c & CTOD methodology is also presented in this chapter.

Chapter 8 presents the process development, and investigation into the wire assisted arc additive manufacturing process (WAAAM). This chapter covers the studies carried out on the various cold wire feeding positions and effects of mechanical and microstructural variation in the WAAAM built.

Chapter 9 summarises the research with the main conclusions drawn with the recommendations and scoping for further work.

2 LITERATURE REVIEW

2.1 Additive manufacturing

The AM system can be classified according to the energy sources deployed, namely beam or arc source delivery. The beam sources delivery could be Laser beam or Electron beam, which are currently being used by some AM equipment makers. The arc sources delivery are the tungsten inert gas arc, plasma arc and MIG/MAG arc [20]. Three common feedstock systems are currently used for the AM process, namely the powder bed systems, which utilise low build volumes less than $0.03m^3$; the energy sources used here could either be a laser beam or electron beam. This system is currently being practised by the Swedish company ARCAM; the singular advantage of this feeding system is its ability to produce high-resolution geometries, internal passages and have good dimensional accuracy compared to others [17].

Figure 2-1(a) shows the typical process channel for this system of laser powder bed fusion (LPBF) AM, which utilises a laser system, reservoir for powder and an inert atmosphere build chamber. One of the key advantages of this process is that the powder bed serves as an in-process support structure for overhangs and orientated complex features, hence allowing for components with high geometrical accuracy ($\pm 0.05mm$) to be manufactured.

Although faced with porosity defects, non-uniform shrinkages and poor dimensional accuracy, but with the better surface quality compared to other wire feed AM [28][29][30], other names for this process are selective laser sintering (SLS) and selective laser melting (SLM). The main difference between these processes and the electron beam melting, which is also the powder bed fusion process as shown in Figure 2-2(a), is that the EBM uses an electron beam instead of a laser and hence has the requirement of a vacuum.

The powder feed system has a slightly higher build volume when compared to the powder bed systems; this system requires the conveying of powder through a nozzle into the build zone and the help of a laser source or electron beam source to melt the powder into shapes as desired. And, because of the flexibility

with the deposition head both in the stationary and off stationary position, it could be used to repair worn-out spare parts [17]. Figure 2-1(b) shows a typical sketch of laser powder feed systems, which are also referred to as direct energy deposition (DED), where the powder is delivered into the substrate with the aid of a carrier gas and the melting is done by a high energy laser source, also shown in Figure 2-2(b).

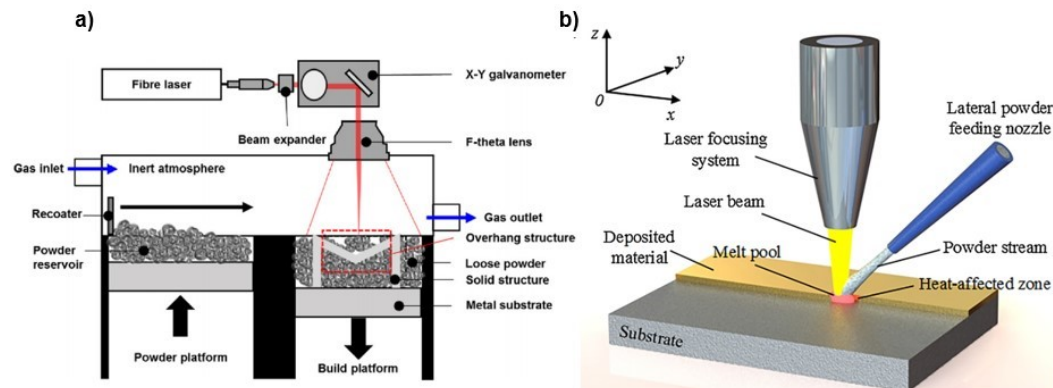


Figure 2-1 (a) Laser powder bed fusion (LPBF) [28]

(b) Laser powder feed (LPF-AM) [31]

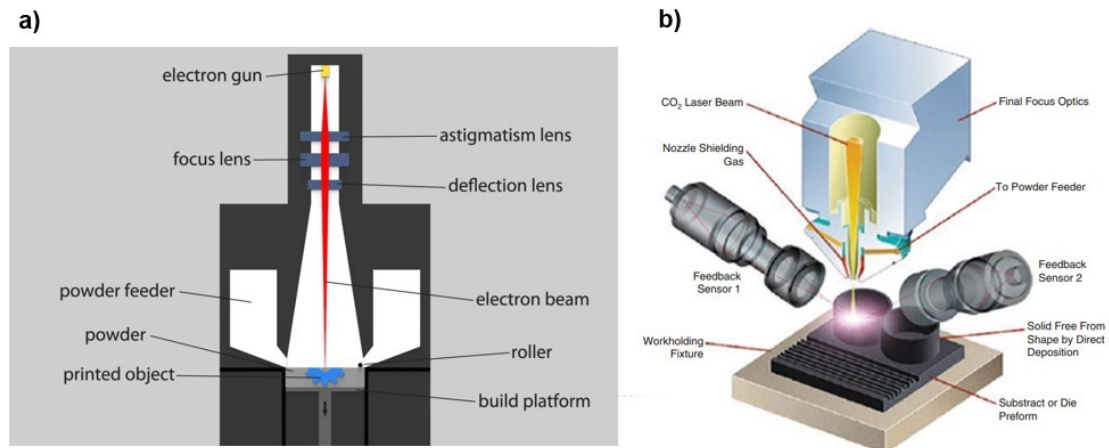


Figure 2-2 (a) (Electron beam melting-EBM) [32]

(b) Direct energy deposition (DED) [33]

According to Huang et al., the mechanical properties and microstructure of parts produced by these processes are strongly affected by the thermal characteristics

of the melt pool, solidification rate and thermal gradient [31]. Another form of DED is the direct laser fabrication by Wang et al. as shown in Figure 2-3, where both wire and powder are fed into the melt pool, and laser sources energy is used in the melting to fabricate the desired part. This AM process has been used successfully to build compositionally graded structures [34].

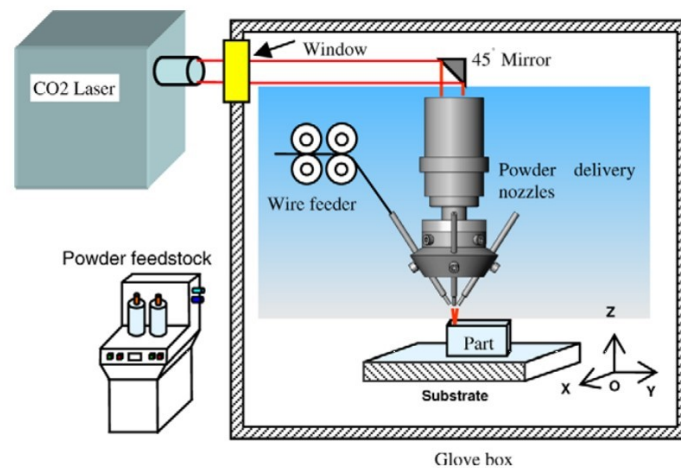


Figure 2-3 Direct laser fabrication using powder and wire [34]

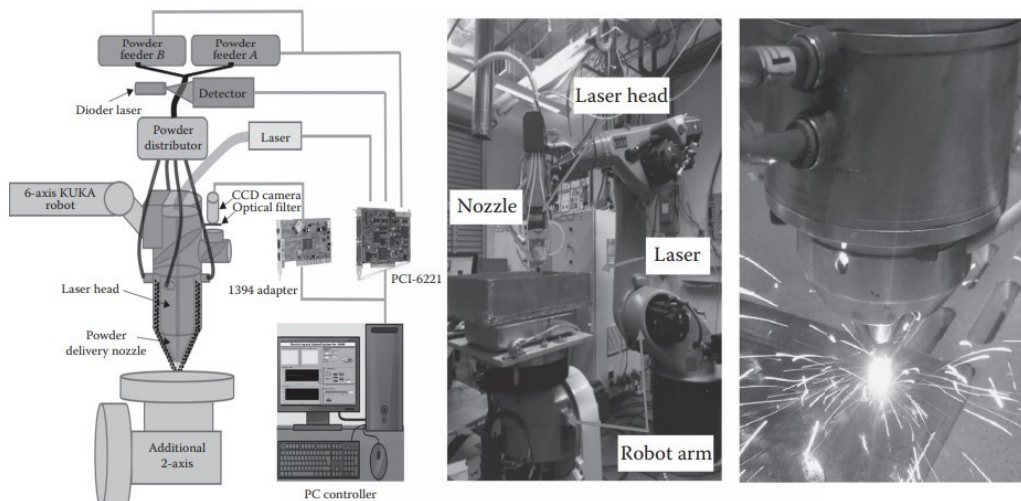


Figure 2-4 Robotized laser powder feed AM [35]

Another form of the DED is the robotised laser powder feeding AM utilising two powder feed nozzles into the melt pool, as shown in Figure 2-4; structures built with this process are shown in Figure 2-5. Although this AM process seems good for metamaterials with different Poisson's ratio, it possesses the challenges of pore formation [35].

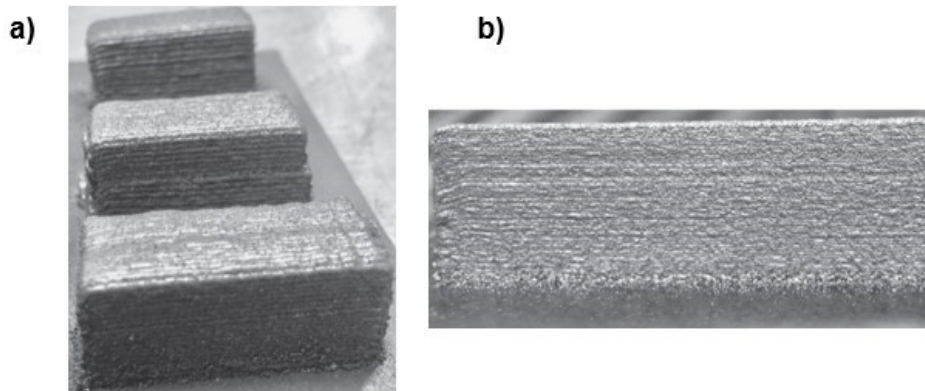


Figure 2-5 built with robotised laser powder feed AM [35]

(a) wall (b) Single bead wall

The powder feed system is also classified as laser metal deposition (LMD), and laser Engineered Net Shaping (LENS). Another DED system is the laser metal deposition with wire (LMD-w). The feedstock for this system is wire as shown; the energy sources could be a laser beam, electron beam and arc system. The electron beam sources are currently used by Sciaky and called electron beam additive manufacture (EBAM). The EBAM system uses an electron beam gun in a closed atmosphere to build solids parts from wires. Its built rate is dependent on the material type, path geometry and features.

EBAM has a build rate of 3.18 to 11.34 kg of metal deposition per hour. Common materials build by Sciaky EBAM systems are 70-30 nickel, copper, Zircalloy, 4340 steels, tungsten, tantalum, Inconel 718, 626, and titanium alloys. The huge cost of the closed system for this application is one of the major concerns of the EBAM system, despite being the fastest metal AM process that can produce high-quality metallic components. A typical EBAM gun deposition system is illustrated in Figure 2-6(a-e). A contrast of the cost implications of the EBAM system for different wire feed diameter ranges from 0.9 – 4.0mm on Titanium 6Al-4V,

tantalum, Inconel 625 and stainless steel, are enumerated in Table 1-1. This further shows that the cost decreases as the wire diameter increases for the highlighted materials, and the cost values are smaller than using powder as the feedstock. The single wire EBAM feed system is shown in Figure 2-7.

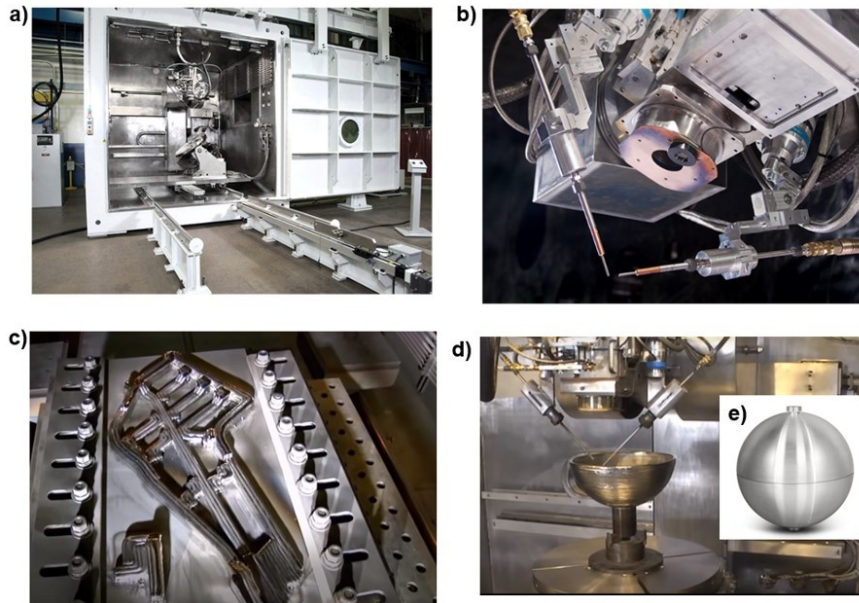


Figure 2-6 Sciaky EBAM systems [21]

(a) Deposition chamber (b) EBAM twin wire feed (c) as-deposited titanium part (d) titanium vessel deposition (e) final machined outlook (courtesy of Lockheed Martin)

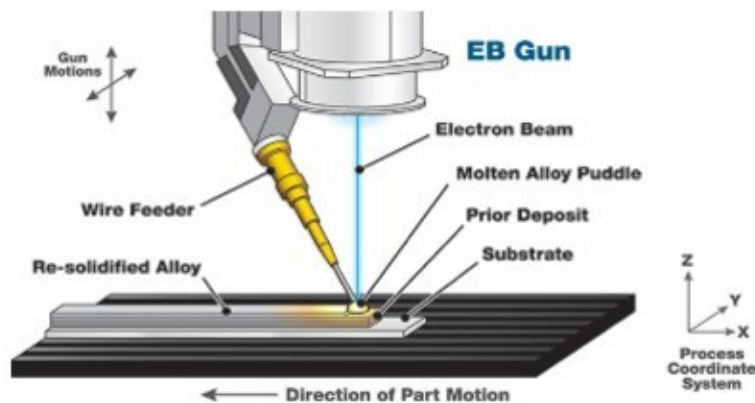


Figure 2-7 Single wire EBAM feed system [21]

The arc system could be any one of plasma arc, TIG arc and MIG/MAG arc, as shown in Figure 2-12; they are collectively called wire feed metal additive manufacturing. This system has a higher deposition rate over the powder bed and blown powder process, although limited by its poor resolution and surface roughness; it also does not require any enclosed chamber as in the EBAM systems. Greer et al. developed a new wire feed metal AM process called the metal big area additive manufacturing (mBAAM) using the MIG power source to build an excavator arm whose original length and weight was 2.1m and 133kg, as shown in Figure 2-8(a). Their structural integrity checks (Figure 2-8(b)) showed that the built arm could withstand the stock excavator arm in actual operation after machining [36].

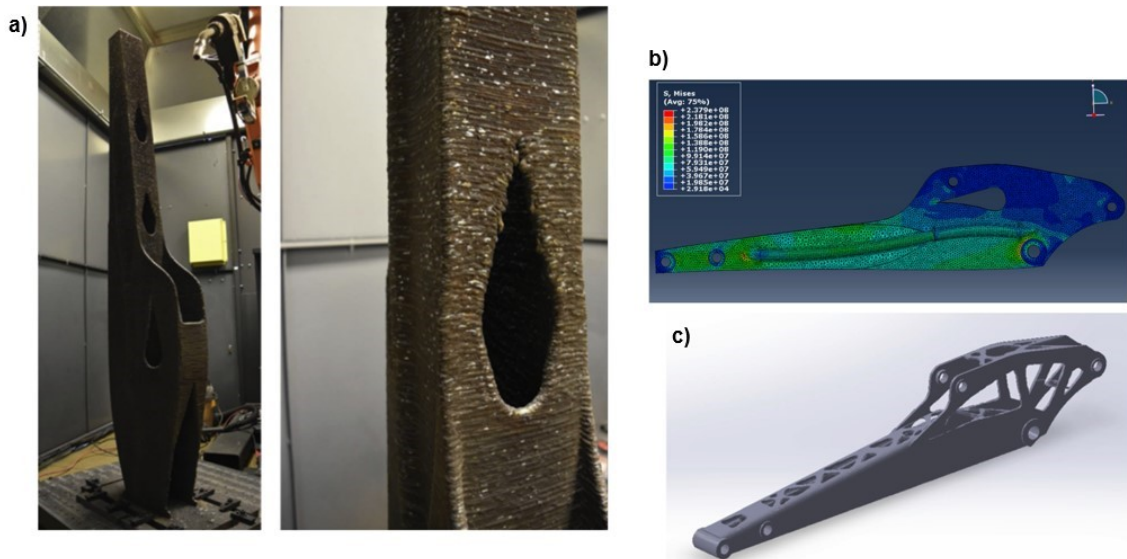


Figure 2-8 mBAAM system [36]

(a) Printed excavator arm (b) FEA system check (c) Final arm geometry

The pioneer wire feed system developed by Cranfield University called the wire plus arc additive manufacturing has gained considerable interest in the industries as a result of its low-cost involvement, superior mechanical properties, platform flexibility, and material utilisation efficiency, as depicted in Figure 1-6. Arc based WAAM consists of a robot providing the motion, coupled with a power source, wire feeder, and rotary table if required.

In the plasma arc deposition (PAD) process, the arc created between the non-consumable electrode and the copper anode in the presence of the shielding gas is forced through an orifice between the cathode and anode (arc constriction). This higher density energy is used in melting the filler metal. The process is more stable than any other WAAM arc source. Hence it is used for special materials, although it has the least deposition rate (1.8kg/hr) compared to others [37]. Figure 2-9 (a-d) shows some of the titanium components built in Cranfield University using the PAD process.

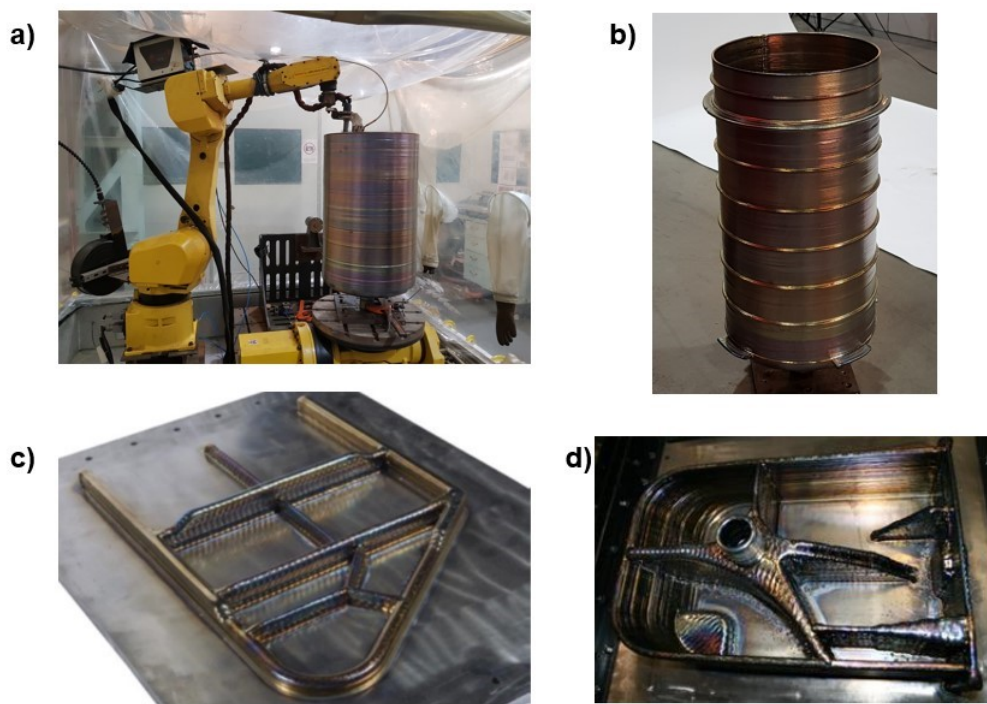


Figure 2-9 Plasma arc deposition (courtesy of Cranfield University)

(a) 6-axis robot, manufactured parts (b) Airspace tank (c & d) Aerospace parts

Another key WAAM power source is the *GMA* power source. This utilises the filler wire as the consumable electrode. Many GMAW power source variants are currently available for WAAM application, namely the conventional GMA, power source, CMT power sources and the Tandem GMA power source. The choice of a particular GMAW power source is a function of the desired outcome, as the conventional GMA gives a high deposition with its high heat input process. It has

four modes of metal transfer, namely the short-circuiting, spray, globular and deep transfer modes. In all the transfer modes, dip transfer, which is a combination of low current with intermittent short-circuiting between the melt pool and filler wire, is preferred because of its low heat input capability.

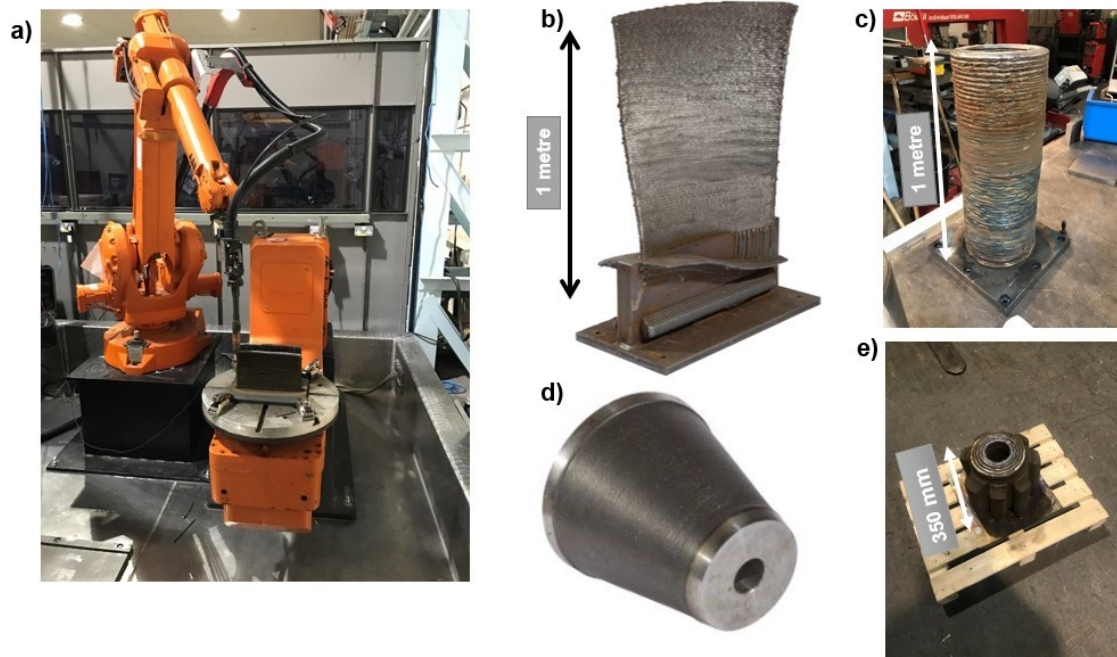


Figure 2-10 MIG arc deposition (courtesy of Cranfield University)

(a) 6-axis robot (b-e) WAAM parts deposited with steel wires



Figure 2-11 High built multidirectional HSLA steel WAA structure[38]

CMT another GMA variant has four operating synergic lines with different transfer modes and heat input ranges. CMT and GMA modes have been used in Cranfield University to build many components, as shown in Figure 2-10. A deposition rate of 9.5 kg/hr has been achieved at a wire feed speed of 9m/min and travel speed of 1.2m/min using the tandem GMAW lead and trial wires using oscillatory strategy [22]. The TIG arc source also uses a non-consumable tungsten electrode but with 1/3 of the arc energy produced by the PAW power source. Figure 2-11 shows the complexity of shapes build using high deposition WAAM[38]

Table 2-1 Different AM process characteristics

Feedstock	Process	Layer thickness(μm)	Surface roughness(μm)	Accuracy (mm)	Deposition rate(Kg/hr)
Powder	SLM [39]	20-100	9-10	± 0.04	N/A
	SLS [40]	75	14-16	± 0.05	0.006
	DLF [41]	200	~ 20	± 0.13	0.6
	EBM	65	9-10	± 0.05	0.1 to 0.2
	LC [42]	N/A	1-2	± 0.025 - ± 0.069	Up to 1
Wire	WAAM [22]	1500	200	± 0.2	Up to 9.5
	EBAM [43]	1500	~ 200	± 0.05	Up to 11.34
	LMD-w				2.9

A comparison of the various AM process and their corresponding characteristics is enumerated in Table 2-1. Other powder systems (SLS, SLM, LC) possess good surface finish with a low deposition rate when compared to the wire feed system such as the WAAM and EBAM. These and many more factors, for example, roughness, mechanical properties, deposition rate, and built volume, should be considered before selecting the desired AM process to deploy.

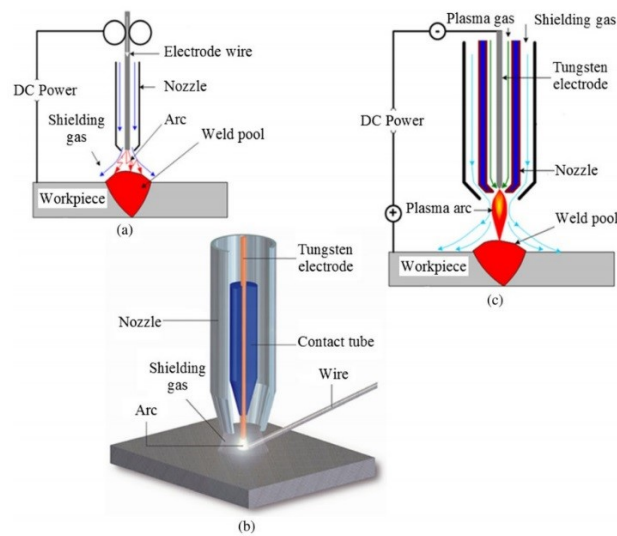


Figure 2-12 WAAM power sources [44]

(a) GMA (b) TIG (c) PA

2.2 Power sources used with additive manufacturing

The GMAW process is one of the oldest welding processes, used from the early 90s with its unique metal transfer modes such as globular, spray and pulse [45], it has also surpassed the SMAW process because of its reduction in cost, high productivity, good weld quality and low heat input. The GMAW process can weld thicker material and ensure weld consistency with a deep penetration weld. Some of the important features of the pulse transfer mode of GMAW are the production of low heat input and good bead appearance. It also gives the welder better directional control over the weld bead. GMAW is mostly used in the renewable energy sector because its automatic form will give faster productivity.

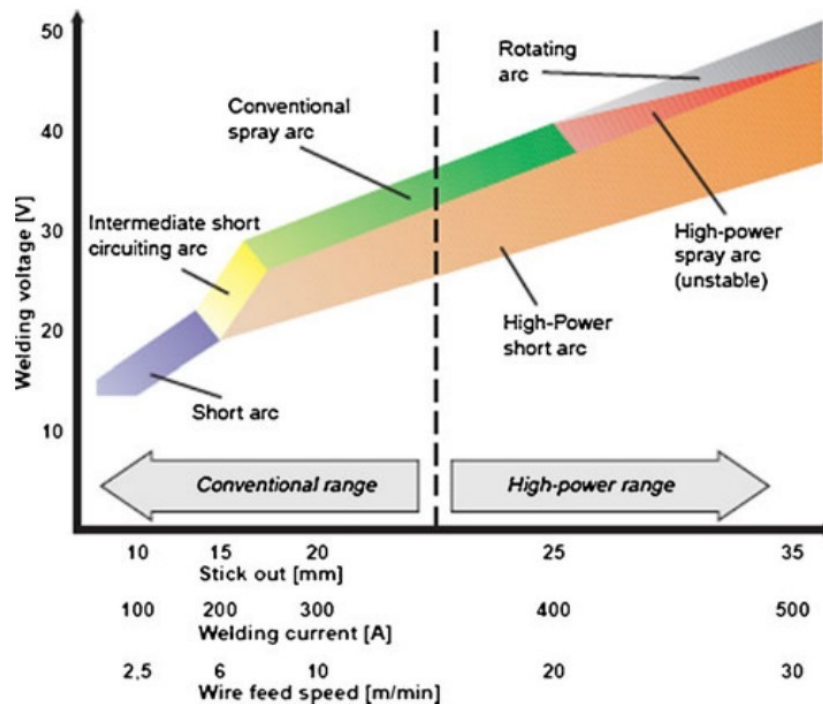


Figure 2-13 GMAW Arc types and their operating intervals [46]

The basic operating modes of the GMAW can be classified into conventional and high power ranges, as shown in Figure 2-13 with welding current and voltage being the dividing factor for the different type of arc process as these parameters influence the droplet transfer modes and stability of the arc [46]. Three major arc transfer modes are popularly used, namely the short-circuiting deep transfer mode, droplet or spray transfer mode and the pulsed arc transfer mode. The major issues with the conventional GMAW modes are the high volume of spatter as a result of the alternating current and voltage cycle waveform. Various welding equipment manufacturers in recent times came up with different designs of electronic and digital control in order to enhance the accuracy of the processes by reducing spatter and heat input at the same time. A close comparison of the modern GMAW arc control and the conventional GMAW arc control are shown below in Table 2-2; it reveals that the CMT, STT, Cold arc, CP and Wise fusion possess close to excellent properties in terms of spatter free weld, low heat input, deposition rate, arc stability and cost-saving.

Some of these current waveforms designed to control arc stability and reduce spatter are shown in Figure 2-14, Figure 2-15, Figure 2-16, Figure 2-17, Figure

2-18 (Merkle cold MIG, EWM cold arc, Lincoln STT & pulsed droplet). Another conventional method of reducing spatter is by re-introducing more stable welding hardware and by using the controlled bridge transfer method (CBT) integrated with computerised controlling technologies to monitor and predict the behaviour of the arc at ignition and re-ignition [47].

Table 2-2 Comparison of different GMAW arcs modes [45]

Properties	Short arc	Globular arc	Spray arc	Pulsed arc
Spatter	Medium	High	Low	Minimal
Quality of weld	Weak	Weak	Excellent	Excellent
Heat input	Low	High	High	Low
Current level	Low to medium	Low to medium	Medium	Low to medium
Deposition rate	High	High	High	Medium
Welding speed	Medium	High	High	High
Arc stability	Good	Weak	Excellent	Excellent
Productivity	Weak	Good	Excellent	Excellent
Saving cost	Weak	Weak	Excellent	Excellent
Penetration	Excellent	Weak	Excellent	Good
Need of welder skills	Low	Low	High	Low
All position welding	Yes	No	No	Yes

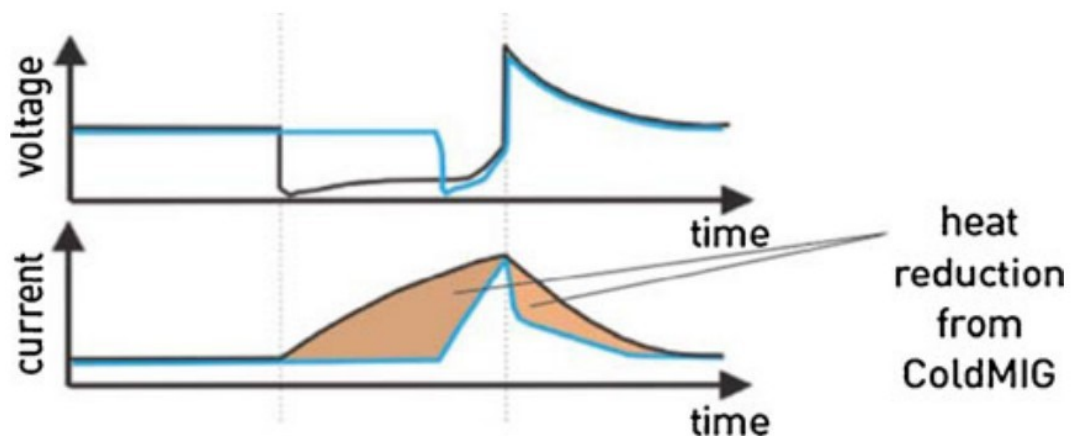


Figure 2-14 Merkle cold MIG [46]

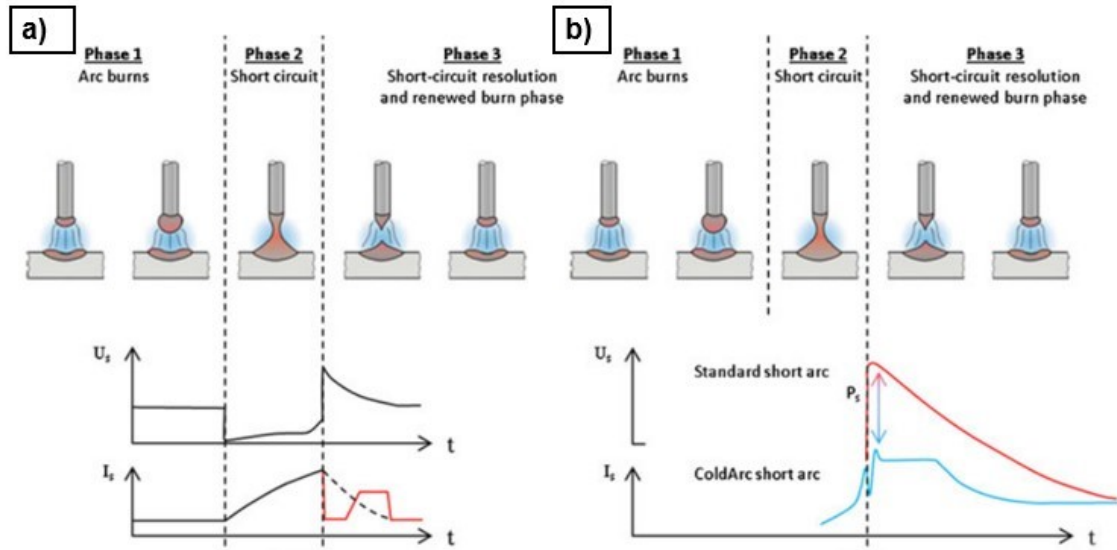


Figure 2-15 EWM cold Arc [48]

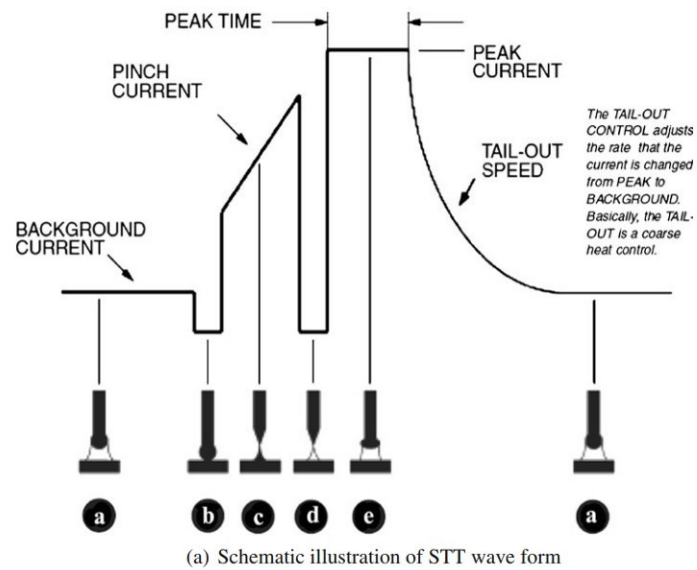


Figure 2-16 Lincoln STT [48]

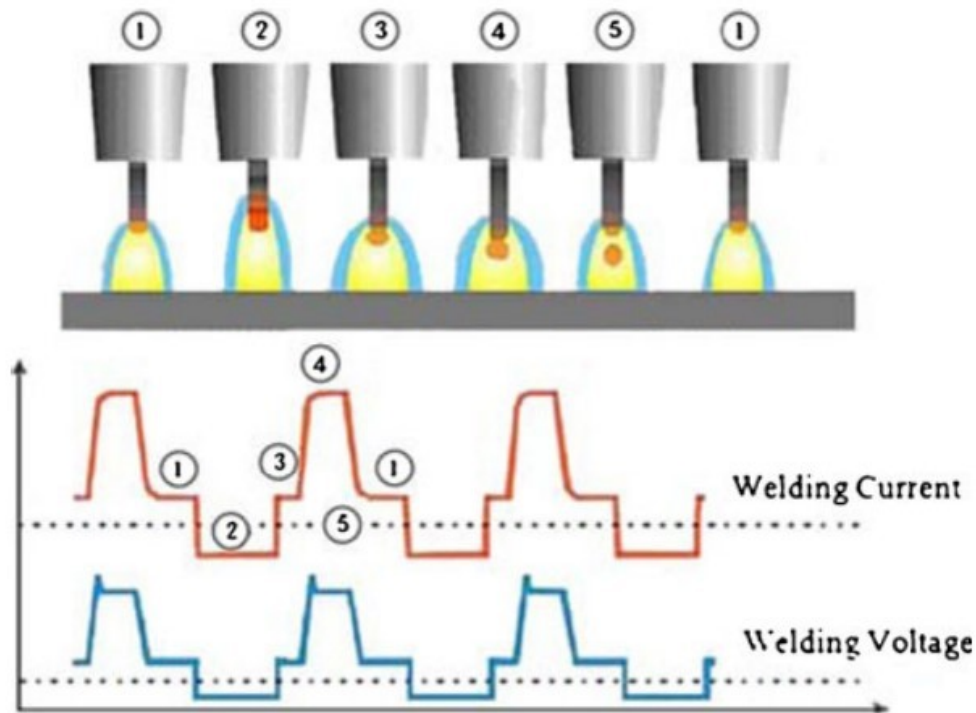


Figure 2-17 Pulsed droplet transition and current and voltage waveform of the cold welding process [46]

The popularly used GMAW process for WAAM is the Fronius cold metal transfer process (CMT). The inbuilt electronic control can detect short circuits and detach the droplets by retracting the wires more than 120 times per second. This retracting backward movement aids the droplet detachment during the short circuit which digitally controls the current by minimising it, thus resulting in spatter free metal transfer. During the welding phase, the wire moves forward and is pulled back again as soon as the short-circuiting happens, hence only a small amount of heat is dissipated. The arc length is always detected and adjusted mechanically by the system [49]. Figure 2-19 shows a typical standard CMT process described above and Figure 2-18(a) shows a typical waveform of the standard CMT process.

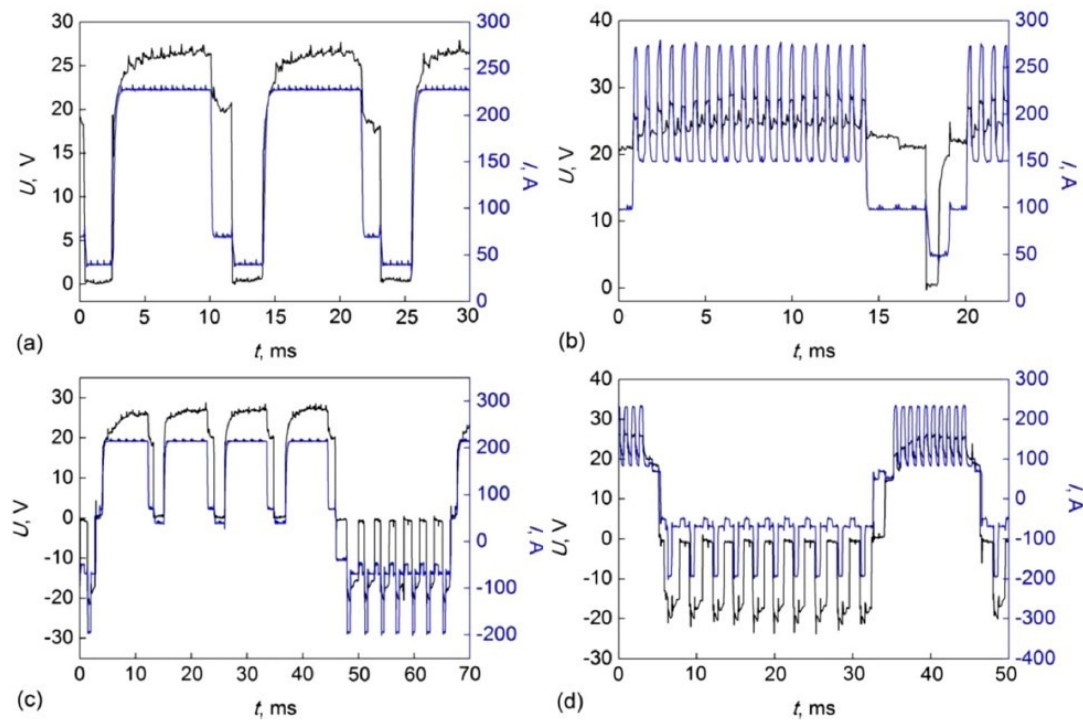


Figure 2-18 Voltage waveform and arc current [50]

(a) CMT, (b) CMT + P, (c) CMT ADV, (d) CMT Pulsed + ADV process at (WFS = 7.5m/min)



Figure 2-19 CMT process [48]

The added advantage of integrated wire motion, low spatter, low heat input and excellent stable arc make CMT a process of choice for wire plus arc additive manufacture. The CMT pulse + advance process (Figure 2-20) combines negative poled CMT cycles and positively poled pulsing cycles to give excellent arc stability with absolute precision and lowest heat input, as shown in Figure 2-18(d). CMT advance is shown in Figure 2-21; here, the polarity of the welding

current is made an integral part of the process control where the polarity reverses in the short circuit phase.

CMT advance possesses a highly controlled thermal process and gives a higher deposition rate than all CMT processes and higher heat input than standard CMT, as shown in Figure 2-18(c). The CMT + pulse cycle (Figure 2-22), combines the CMT cycle and the pulsed cycle hence dissipate more heat into the system, as shown in Figure 2-18(b).

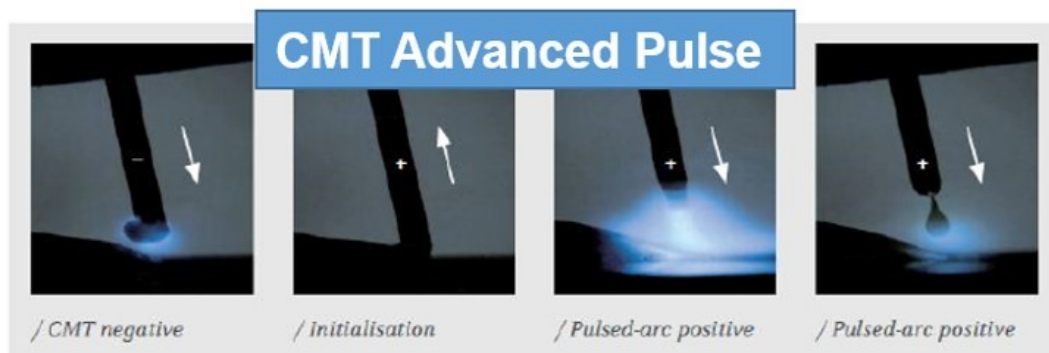


Figure 2-20 CMT negative and pulsing cycles[49]

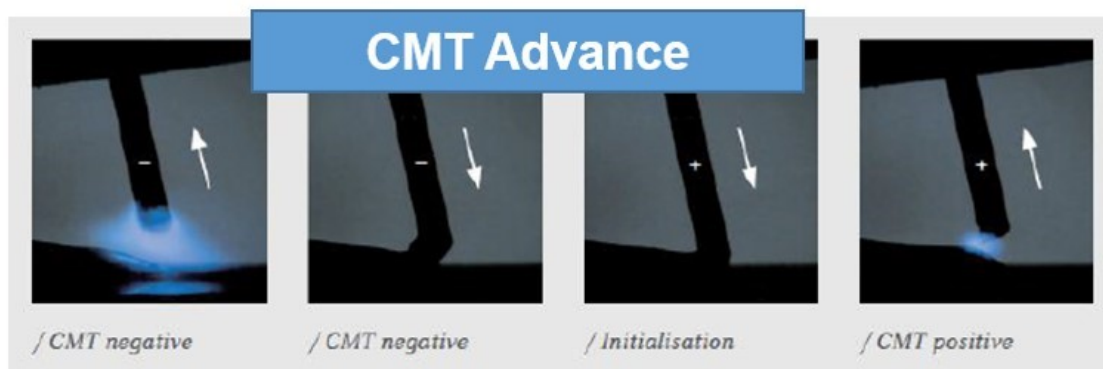


Figure 2-21 CMT cycle and pulsing cycles[49]

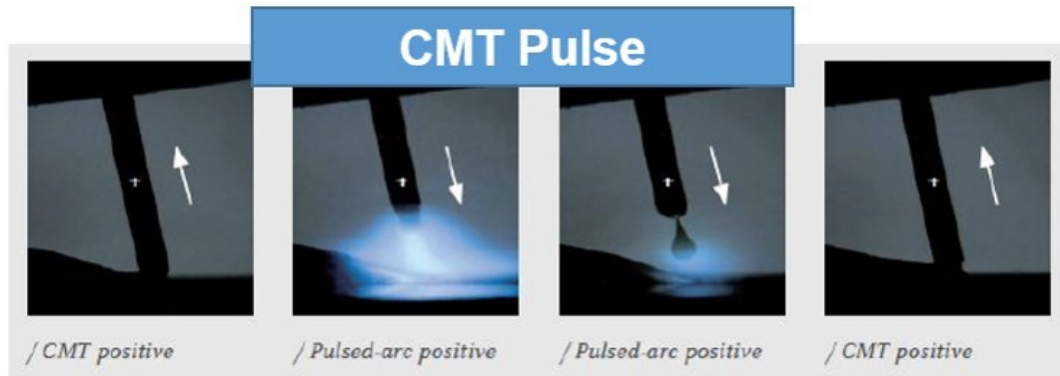


Figure 2-22 CMT cycle and pulsing cycles[49]

In a recent study to determine the suitability of these four CMT variants in additively manufacturing Al-6.3%Cu alloy, it was discovered that the level of porosity found in the weld microstructure depends on the CMT arc variant used, and it was concluded that CMP-PADV is the most suitable variant that gives less porosity while the standard CMT process is not suitable for the AM process [50].

A similar work by Cong et al. also concluded that CMT pulse plus advance is more suitable in controlling porosity. Further, it is more effective than the other three CMT variants due to its low heat input capability and the cleaning effect on the tip of the wire during the negative cycle, as attested by the microstructural findings. However, it is not suitable for building block structure due to the lack of inter-run fusion caused by the low heat input process to melt the interlayers [51] although this claim was not verified with low carbon low strength steel and high strength low alloy steel deposition.

2.3 Deposition of HSLA steel using WAAM

2.3.1 Deposition Strategy and Surface waviness

Waqas et al. used the GMAW based additive technique (WAAM) to deposit copper-coated HSLA steel wire (ER70S-6) using the single-pass deposition strategy while maintaining a constant profile of deposition at the arc striking, arc-extinguishing and steady stage areas [52]. They succeeded in depositing an 8.2mm thick structure using a wire diameter (WD) of 1.2mm, as shown in Figure 2-23(a). A similar strategy was used by Rodrigues et al. to deposit HSLA

ER110S-G with WD 1mm using the GMAW arc process [53]. The WAAM structure deposited here was $5.6 \pm 0.3 - 8.8 \pm 0.6$ mm in thickness, as shown in Figure 2-23(b). The SW in Waqas et al.'s work was controlled by parameter compensation (changing travel speed at constant current at the subsequent steady layers); hence the 92.1% deposition efficiency obtained despite using the single direction travelling, which resulted in arc striking and arc extinguishing area deterioration.

Sequeira Almeida and Xiong et al. studied the interaction of WAAM process parameters with SW and roughness (SR), both found that TS influences SW and roughness as shown in Figure 2-24(a) [54] and Figure 2-25(b) [55]. Almeida also established that smaller WD is better in any TS (Figure 2-24(c)) for establishing reduced SW, which is also dependent on wire composition (Figure 2-24(d)).

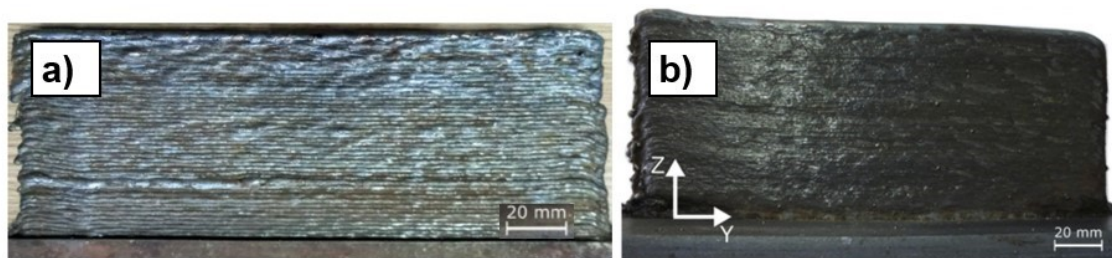


Figure 2-23 WAAM parts

(a) Waqas et al. [52] (b) Rodrigues et al. [53]

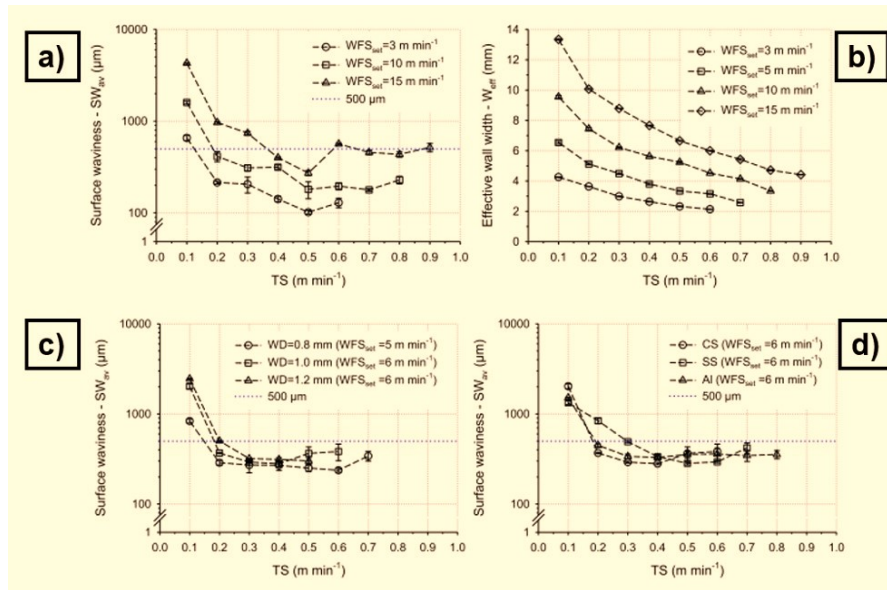


Figure 2-24 Effects of process parameters on surface waviness [54]

(a) SW vs. TS (b) EWW vs TS (c) SW vs WD (d) SW vs BM

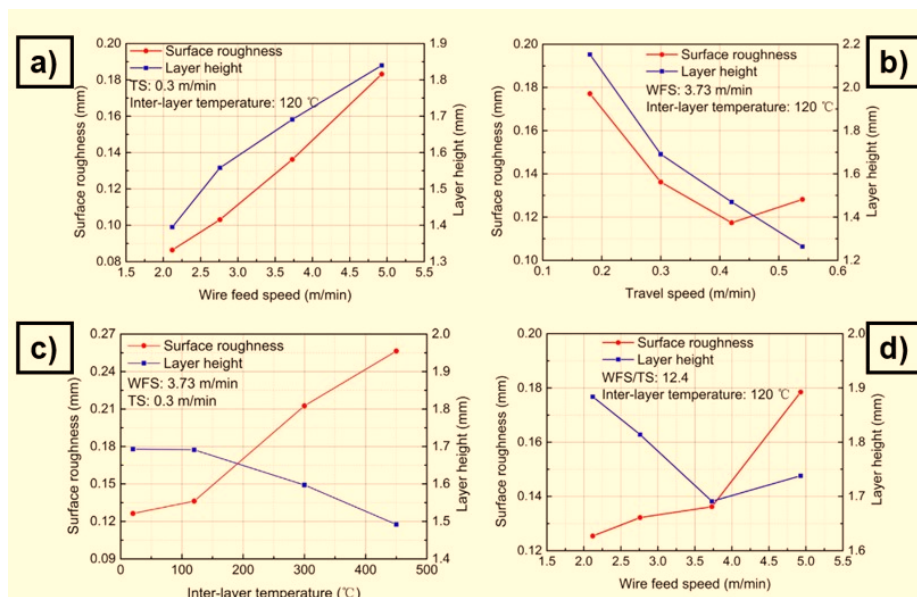


Figure 2-25 Effect of process parameters on surface roughness [55]

(a) SR vs WFS (b) SR vs TS (c) SR vs IT (d) SR vs LH

The work by Xiong et al. characterising the influence of WAAM process parameters on SR also points to the same findings, that SR increases as WFS increase and decreases as the TS increases to a limit of 0.4m/min after which it starts to increase again, as shown in Figure 2-25(a & b); this is comparable to

SW behaviour. Interpass cooling was also proved to affect the SR quality; SR decreases as the interlayer temperature increases, and it correspondingly decreases the layer height, as shown in Figure 2-25(c & d) [55]. In general, the layer by layer deposition with PGMAW is affected by the deposition parameters (WFS, current, voltage, TS, interpass cooling and wire alloy composition). Layer height, which is determined by the WFS and TS is also an SW/SR controlling parameter in WAAM. The work of Rodrigues et al. [51], which showed that WAAM parameters leading to high heat input and high layer height generate low SW in WAAM, is contrary to the work of Almeida and Xiong et al. although wire composition, which was not taken into consideration during their studies, might play an important role in this regard.

2.3.2 Microstructural variation

The elemental composition of HSLA steels plays a vital role in their microstructural and mechanical behaviour. Alloying with Cr, Mn, Mo and Cu gives the microstructure strength, toughness and weldability through hardenability by the transformation of austenite to a fine ferritic microstructure. The Cu rich particles through precipitation provide strength during aging in the absence of carbide containing micro constituents. The low carbon content of the alloys is responsible for the good toughness and weldability [56]. Unlike normal traditional welding, layer by layer deposition (WAAM) might show some form of microstructural variation as a result of differential cooling rates, thickness and build parameters (heat inputs). Waqas et al. used GMA power source with Ar + 20%CO₂ shielding gas to deposit ER70S-6, low alloyed steel wire containing (in weight percent) 0.06 C, 1.4 Mn, 0.8 Si (minimum values), 0.15 Ni, 0.15 Mo, 0.15 Cr, 0.5 Cu (maximum values). Dendritic structures were observed at the deposition direction as a result of the fast cooling occasioned by the large heat sink at the base [52] as shown in Figure 2-27(a-c). This same trend was observed by Xu et al. in the WAAM of maraging steel at the topmost layers that did not undergo thermal cycling [57]. Sridharan et al. deposited low alloyed steel wire containing (in weight percent) 0.09 C, 1.63 Mn, and 0.90 Si, using TIG power source in the X, Y & Z planes to obtain a 56mm, 64mm & 10mm thick structure.

Five distinctive zones were observed by them in the microstructure, the fine grain (FG), coarse grain, (CG), heat affected zone (HAZ), FGHAZ, and CGHAZ. A refined polygonal ferrite microstructure was observed in the FGHAZ region of the X-direction specimen, as shown in Figure 2-27(a). The X-direction specimen in Figure 2-27(e) presents three distinctive regions. The highlighted region (CG region) contains a higher fraction of displacive transformation product growing epitaxially through the solid-liquid interphase. This brittle zone region is the direct consequence of the fast differential cooling of the carbon-rich austenite to martensite (MA) and acicular ferrite (AF) [58]. The sample fabricated in the Y-direction showed some presence of AF, as shown in Figure 2-27(f).

The cooling rate ($t_{5/8}$) of the X, Y and Z was determined analytically using the three and two-dimensional heat conduction relationship in Eq. (2-1) & (2-2).

$$\Delta t_{8/5} = \frac{n}{2\pi\gamma} \cdot \frac{VI}{v} \frac{1}{d^2} \left(\left(\frac{1}{500 - T_0} \right)^2 - \left(\frac{1}{800 - T_0} \right)^2 \right) \quad (2-1)$$

$$\Delta t_{8/5} = \frac{n^2}{2\pi\gamma\rho c} \cdot \frac{VI^2}{v} \left(\frac{1}{500 - T_0} - \frac{1}{800 - T_0} \right) \quad (2-2)$$

n ; Thermal process efficiency, V ; arc voltage, I ; arc current, c ; specific energy, T_0 ; the working temperature in $^{\circ}\text{C}$, v ; travel speed, γ ; thermal conductivity, ρ ; specific density, d ; built thickness in cm.

In this instance, the cooling time from 800 $^{\circ}\text{C}$ to 500 $^{\circ}\text{C}$ ($\Delta t_{8/5}$) was 2.59 seconds for the X & Y samples and 60 seconds for the Z sample. The significant of the ($\Delta t_{8/5}$) cooling times plays a huge role in the microstructure of steels as it determines the solid-state transformation products formation in steels. As seen in Figure 2-26(a), welding parameters giving temperatures above the A_{c3} temperatures (in austenitic region) promotes grain growth, which is responsible for coarse grains formation and recrystallisation in the heat-affected zone, while temperature below A_{c3} and above A_{c1} leads to the formation austenite, polygonal

ferrite and widmanstätten ferrite. Temperatures below A_{c1} leads to the formation of acicular ferrite and grain boundary ferrite. Cooling rates between 800°C to 500°C ($\Delta_{t8/5}$) is consistent with proeutectoid ferrite formation (tempering of the microstructure occurs in this region), Figure 2-26(b) also shows hardness as a function cooling time between 800°C to 500°C ($\Delta_{t8/5}$), it is evident that hardness decrease rapidly with decreasing cooling rates and martensitic structures are formed at faster cooling rates.

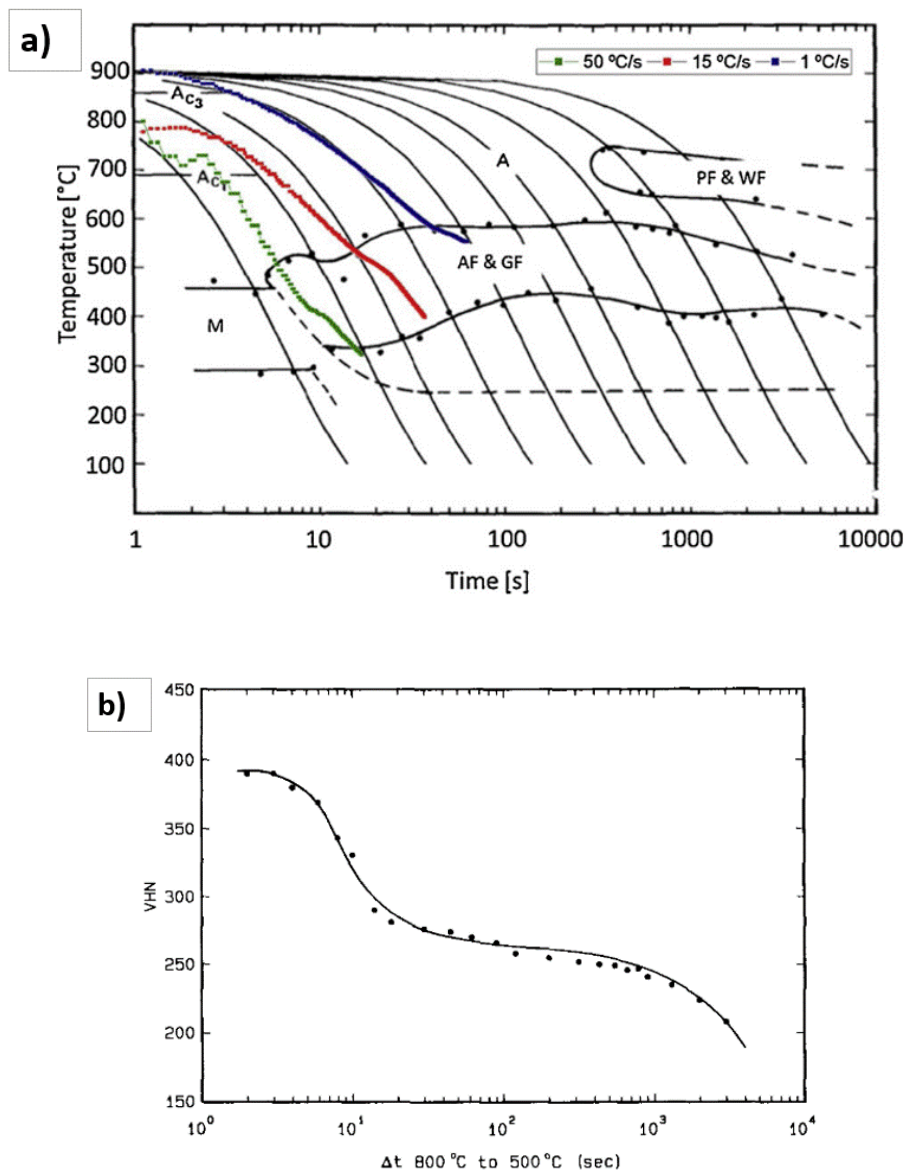


Figure 2-26 Low carbon HSLA steel[56]

(a) CCT Diagram[53] (b) Hardness variation at various cooling rates

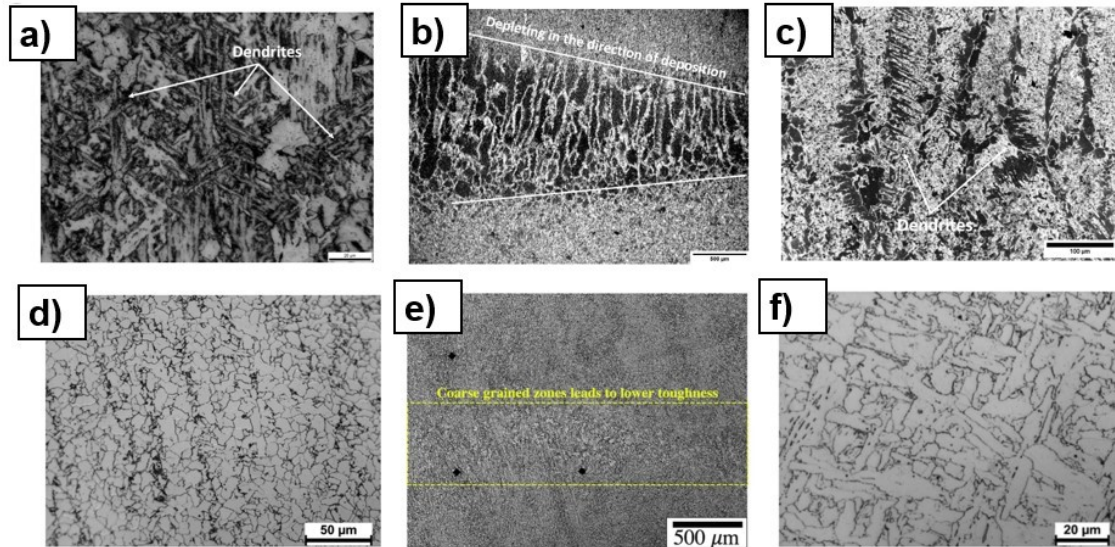


Figure 2-27 Microstructural variation in ER70S-6

(a-c) Dendritic structures in build direction [52], (d) Refined polygonal ferrite structures in fine grain HAZ [58], (e) Different regions in the X-direction [58] (f) Acicular ferrite microstructure in the Y deposition direction [58]

In a similar work by Rodrigues et al. [53] dendritic structures were found emanating from the base of the deposition, as seen in Figure 2-28(d), the hardness value is high in this region as shown in Figure 2-29(a & b) due to the huge thermal gradient associated with fast cooling brought about by the large heat sink at the base. The top region of the low heat input hardness plot showed higher hardness than the high heat input hardness plot. Due to the relative height of the test piece used here. It is not clear why the hardness variation from the different sector vary much. Acicular ferrite and granular morphologies were also observed along the build direction, as shown in Figure 2-28(a-c). Hence the relatively low hardness compared to the base region. Although the deposit was free of defects common with other AM processes.

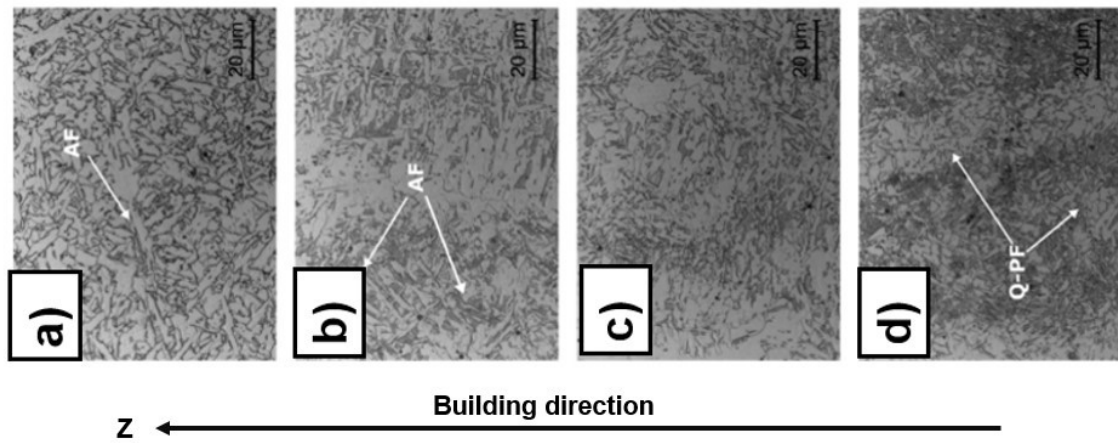
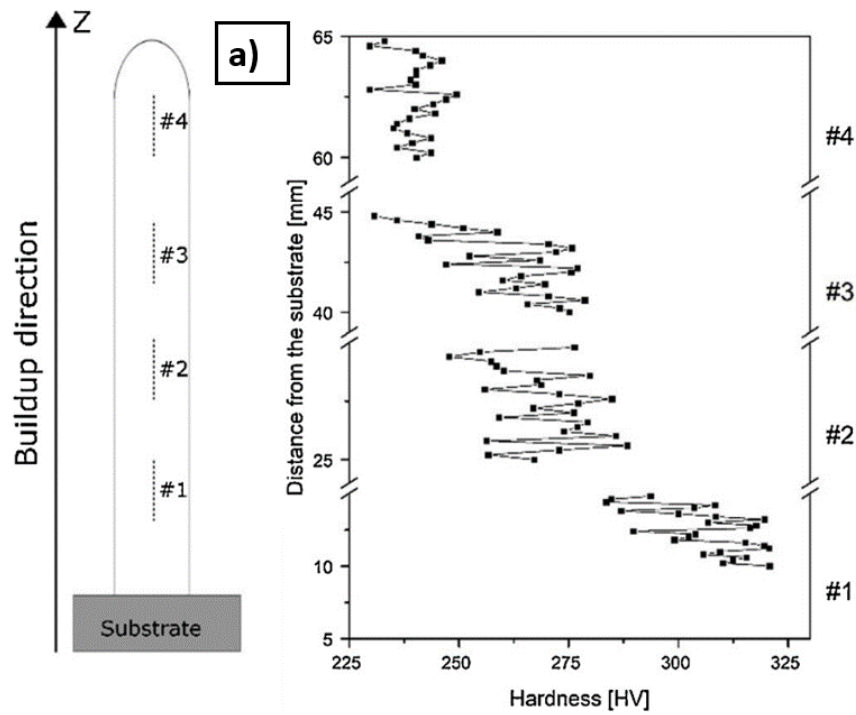


Figure 2-28 Microstructural variation in ER110S-G

(a-d) Top-bottom of deposition [53]



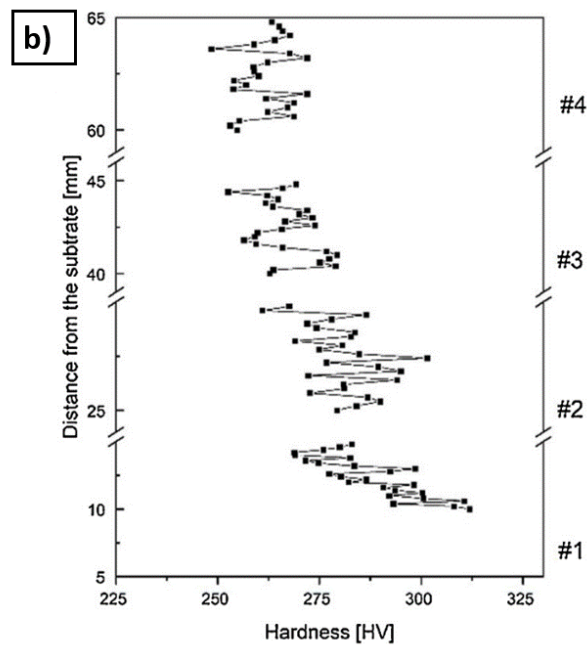


Figure 2-29 WAAM HSLA steel hardness variation along 65mm deposit[53]

(a) High heat input (b) Low heat input

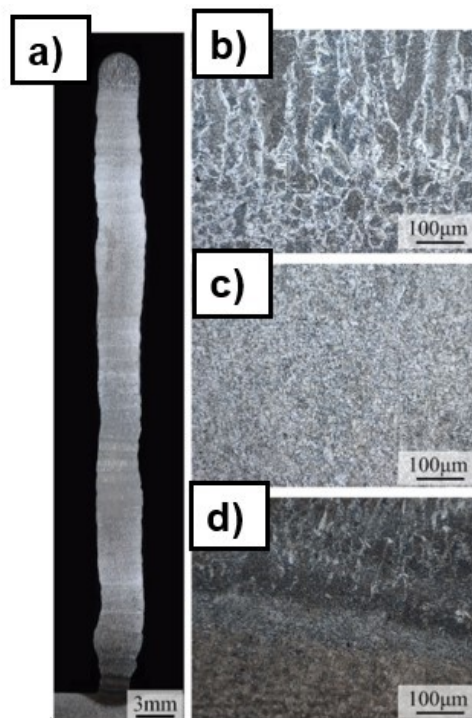


Figure 2-30 ER70S-6 [59]

(a) Microstructure of WAAM part (b) Top zone (c) Intermediate zone (d) Bottom zone

The work by Lu et al. uses GMA power source with 95%Ar + 5%CO₂ shielding gas to deposit low alloyed steel wire containing (in weight percent) 0.08 C, 1.5 Mn, 1.05 Si, 0.45 Cu to produce thin wall WAAM parts. The rapid cooling by conduction and convection at the topmost region shown in Figure 2-30(b) leads to the formation of primary austenite dendritic structures which transformed to a Widmanstätten ferrite structure consisting of ferrite and lamellar pearlite [59]. The thickness of the structure was estimated to be around 1.5 ± 0.5 mm. The Widmanstätten ferrite structure is zones of high-stress concentration prone to brittle failure in steels. The intermediate zone in Figure 2-30(c) exhibits a similar structure to that in Figure 2-27(e) with reheated and re-melting regions due to the effects of the molten pool on the subsequently deposited layers. These two regions play a vital role in the anisotropic behaviour of the WAAM part as the ferrite grain differs in size and length. Lu et al. recorded a 20% variation in grain size across these two regions. The bottom region in Figure 2-30(d) consists of two distinct zones, the martensitic microstructure close to the base metal, which will be machined off, and the tempered dendritic structures, as shown in Figure 2-30(d).

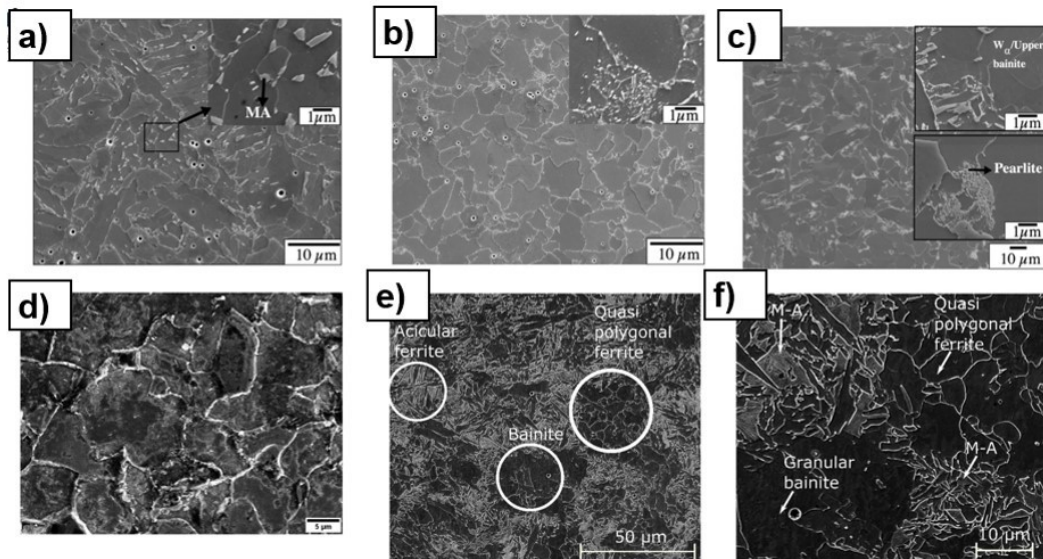


Figure 2-31 SEM microstructural Images of WAAM structures

ER70S-6 (a) X-direction build sample (CGHAZ), (b) X-direction build sample (FGHAZ), (c) Z-direction build sample (UFGPF) [58], (d) Equiaxed grains [52], ER110S-G (e) Overview (f) High magnification [53]

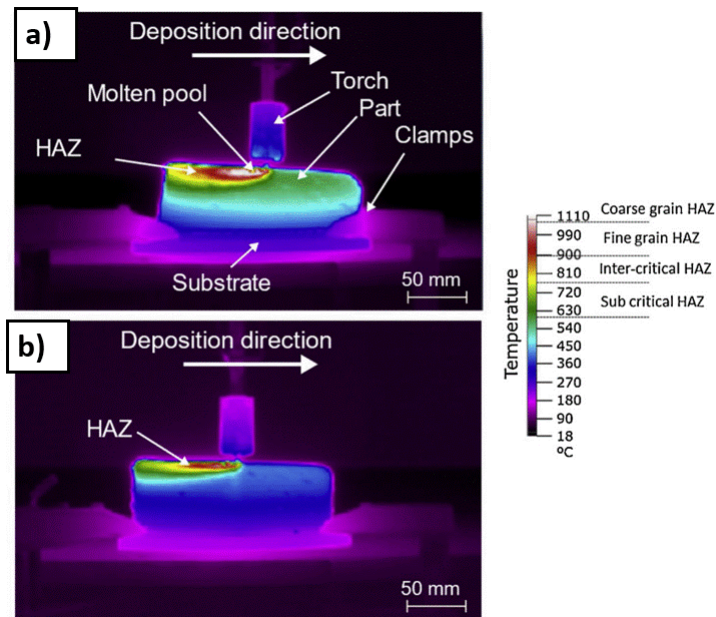


Figure 2-32 Thermographic analysis of the last deposited layers [53]

Due to the observed cooling rate ($\Delta_{t_{8/5}}$) in Sridharan et al. for the thin part (10mm), which is 60 seconds (slower cooling), predominantly uniform fine-grained polygonal ferrite, and lamellar pearlite were obtained in the microstructure, as shown in Figure 2-31(c). Lamellar cementite carbides (Fe_3C) were observed as well. The microstructure of the fast cooling rate ($\Delta_{t_{8/5}}$) (2.59secs) observed in the thick part (54 & 64mm) fabricated in the X- & Y-directions led to the formation of a higher volume of displacive transformation products. The CGHAZ and the FGHAZ regions of the X-microstructure were composed of upper bainite with pools of MA and polygonal ferrite with spheroid pearlite, as shown in Figure 2-31(a&b). The MA formed here is blocky with a stringer like morphology between the bainitic sheaves [58]. The stiffer blocky MA led to concentration and tri-axiality of stress at the point near the ferrite boundaries, hence the cleavage of the ferrite grains, which gives rise to micro-cracks, as seen in Figure 2-33(c). Other failure mechanisms observed were through microvoid coalescence, as shown in Figure 2-33(a & b) [58].

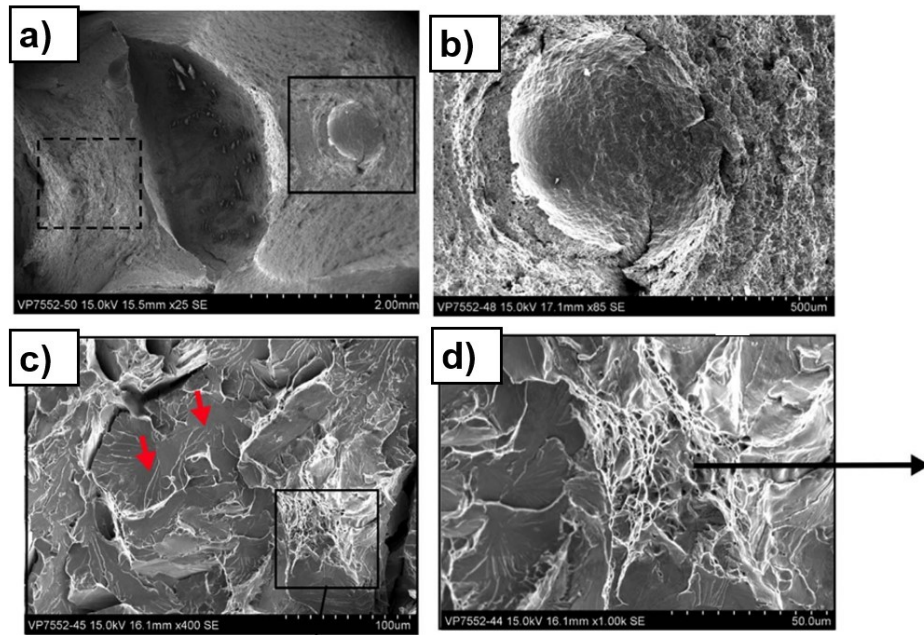


Figure 2-33 Fractographs of failure mechanism [58]

(a) Fracture initiation points in the Y direction (b) Failure initiated adjacent to a micro void (200µm) marked in (a) (c) Cleavage failure region (d) Higher magnification of marked point in (c).

SEM observed in Waqas et al., as shown in Figure 2-31(d), present equiaxed grains with some Cu inclusion precipitates on the grain boundaries. Rodrigues et al. [51] superimposed their obtained cooling curves on the CTT diagram proposed by Thompson et al. for low carbon and HSLA steel plates of similar composition to the WAAM wires adopted in their studies [56] as shown in Figure 2-26(a). Figure 2-32 also showed the thermographic analysis of the last deposits, predicting the temperature trends for the various regions of the deposit. All microstructural features as per the SEM images in Figure 2-31(e & f) were in agreement with features shown on the CCT diagram as per superimposing his cooling rate curves (50°C/s, 15°C/s & 1°C/s) [53]. Although the cooling rates obtained in Sridharan et al.'s study were 120°C/s & 5°C/s, no fully martensitic structures were obtained as claimed by Rodrigues et al. According to Avila et al., M-A and stinger type structures are obtained between 10 and 40°C/s and above [60].

According to Davis and King [59], four mechanisms were identified to initiate fracture in HSLA steels. The two distinct morphologies of M-A (blocky type with sizes between 3 and 5 μm and elongated stringer type that develops between bainite and martensitic lath with width ranging from 0.2 to 1 μm with appreciable length) are route ways to quick fracture failure, especially in the stringer type morphology (Figure 2-34(a)). The martensitic transformation induced tangential residual tensile stress as a result of volume expansion takes place in the ferrite matrix support cleavage fracture (Figure 2-34(b)). The strength mismatch resulting from the M-A and the surrounding ferrite matrix leads to concentration and tri-axiality of stress which assist cleavage fracture (Figure 2-34(c)). Carbon segregation has also been found to weaken the interface between M-A and the ferrite matrix, which leads to debonding as a result of increased matrix stress concentration. Debonding, on the other hand, results in micro-cracks at the MA and ferrite matrix interface (Figure 2-34(d)).

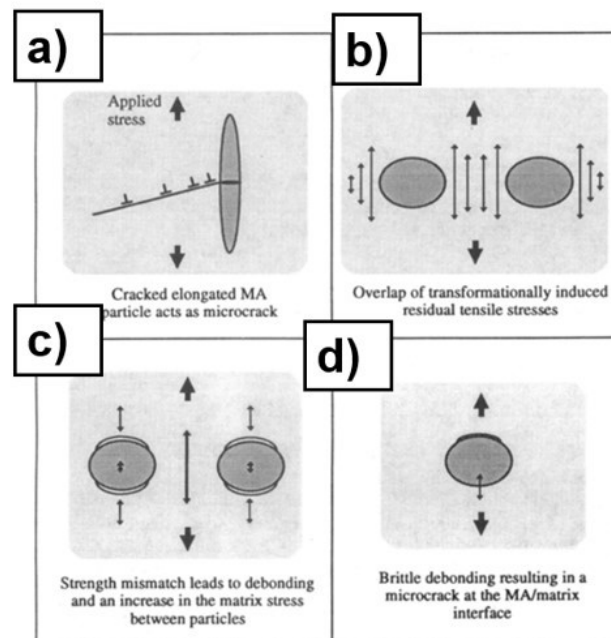


Figure 2-34 Fracture initiation mechanism in dual-phase steel [61].

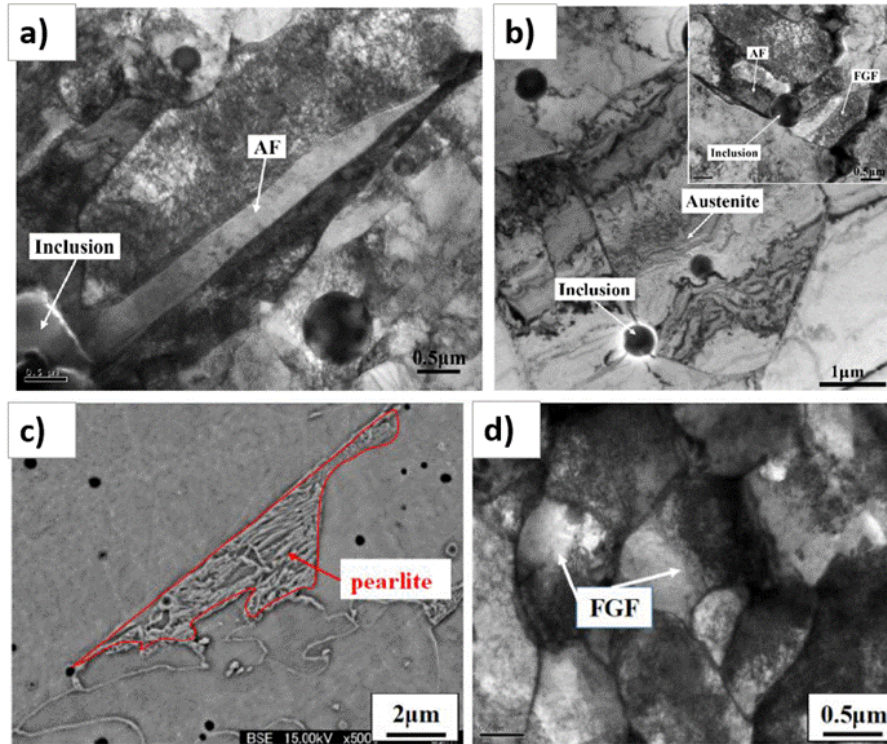


Figure 2-35 Microstructural variation in HSLA steel [38]

(a) Solidification zone (TEM) (b) Austenitizing zone (TEM) (c) Tempering zone (SEM) (d) Tempering zone (TEM)

Yu et al. determined three key regions of microstructural transformation in his studies of HSLA steels WAAM with chemical composition 0.08%C, 1.40%Mn, 0.78%Si, 0.30%Cr, 0.70%Ni and 0.50%Mo. They discovered that the solidification zone contains inclusions of Al_2O_3 , MnO , SiO_2 , MnS and inclusion induced AF as shown in Figure 2-35(a). The austenitising zone contains more inclusion induced transformation, while the tempering zone contains polygonal ferrite and pearlites of equiaxed grain sizes, as shown in Figure 2-35(c & d)[38].

2.3.3 Mechanical properties and impact toughness

The stress-strain curve, as presented by Sridharan et al., is shown in Figure 2-36(a-d). The strength in the sample fabricated in the X- and Y-directions are much better than the Z fabricated sample, due to much displacive transformation product displayed in the X and Y samples as a result of their high thermal heat

diffusivity factor. The Y-direction sample has more scatter in the strength properties, which could be due to process defects, such as gas porosity and lack of inter-run fusion of the layers. The X-direction samples indicate better strength and PCE value; this is inconsonant with Lu et al.'s [59] work, as shown in Figure 2-37(a), although built with minimal thickness. Variability in WAAM tensile properties, in general, is inconsonant with the cooling rate and volume fraction of displacive transformation products. The hardness distribution is also related to the cooling rate and volume fraction of the M-A constituent trapped in the layers. Hardness values are always higher close to the base of the substrate (if not machined off) and at the top of the final deposited layers, as shown in Figure 2-37(b). The UTS, YS and PCE in ER70S-6 as shown by Sridharan et al. [58] is in the range of UTS (480-520MPa), YS (380-410MPa), PCE (13-28%). The UTS, YS and PCE in WAAM ER110S-G as studied by Rodrigues et al. [53] is UTS (700-795MPa) PCE (17.1 -25.6%).

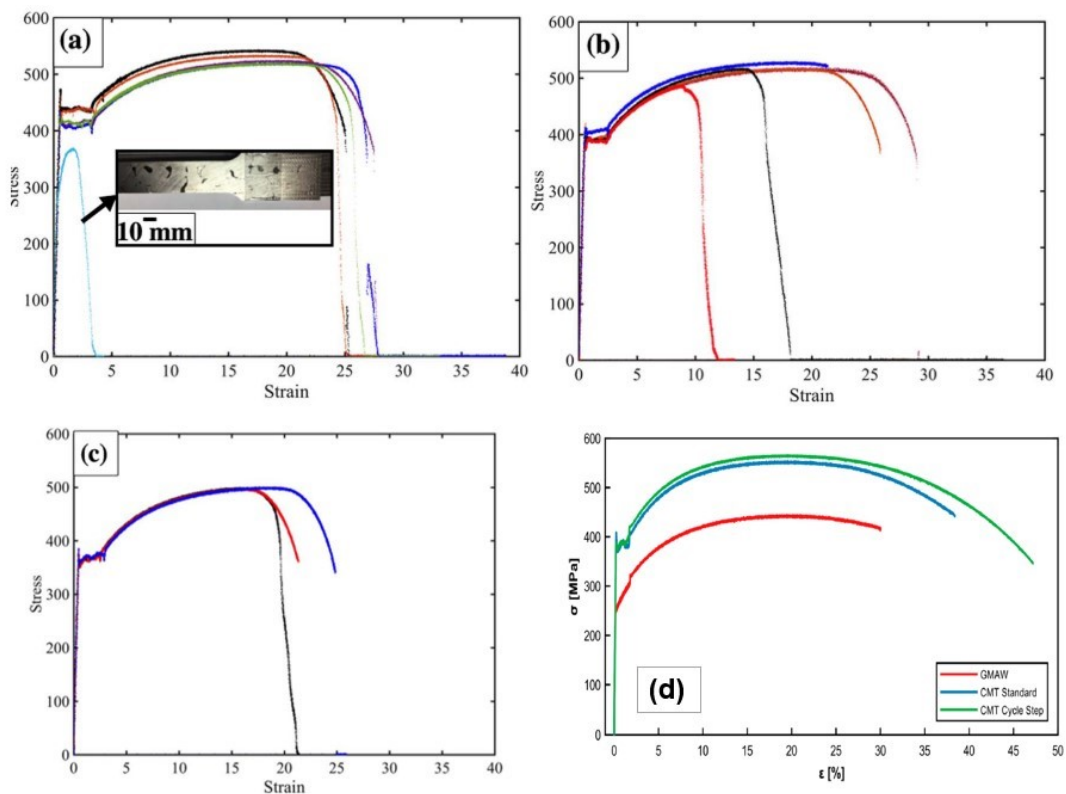


Figure 2-36 Stress-Strain curves of WAAM steel components

(a) X-direction (b) Y-direction (c) Z-direction [58] (d) 3-different power sources [62]

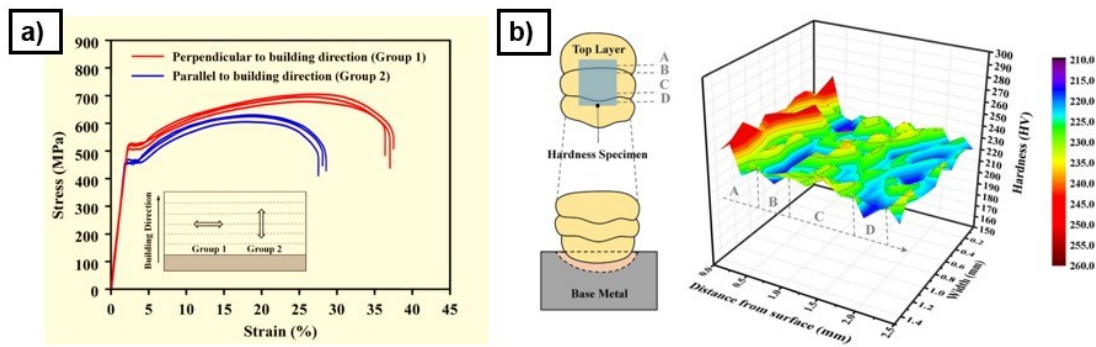


Figure 2-37 Stress-Strain and hardness profile

(a) Stress-strain curve (X vs Z) (b) Hardness profile [59]

There is huge scatter in the impact toughness properties of ER70S-6 fabricated by Sridharan et al., as shown in Figure 2-38. The X-direction build had poor impact toughness while the Y and Z builds had better impact toughness properties.

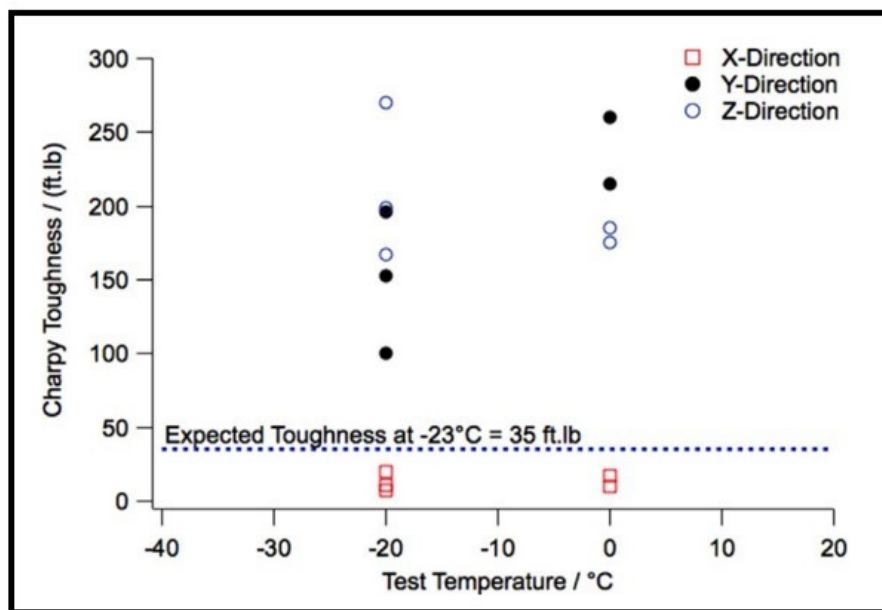


Figure 2-38 Charpy impact toughness (X vs Y vs Z) [58]

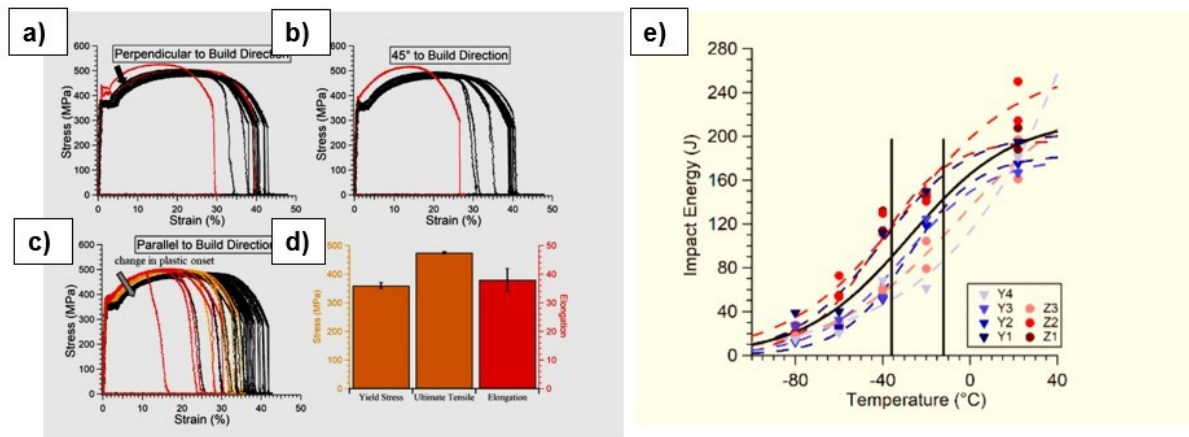


Figure 2-39 Mechanical strength and DBTT curve for as studied by [63]

(a-c) Stress-strain curve (X vs. Z vs. XY) (d) UTS vs. YS vs. PCE (e) DBTT

Shassere et al. deposited ER70S-6 using a surface tension transfer power mode of MAG process to build a wall structures of 12 mm thickness using two parallel beads with centre distance of 4.5 mm [63]. The obtained mechanical properties as shown in Figure 2-39 were in the same range with Ghaffari et al. who deposited a 150 mm height, and 135 mm long, and 22 mm thick low carbon low alloy steel walls with six passes laid parallel to each other. This strategy and overlap used gave room for some lack of fusion defects between the intern layers with some porosity defects. Although the strength values displayed in both studies were comparable to A36 wrought values but low ductility was recorded in the build direction as shown in Figure 2-40 which actually was supposed to give the best result [64]. The ductile to brittle transition temperature (DBTT) as displayed by this mild steel (Figure 2-39(e)) was from -17°C to -36°C.

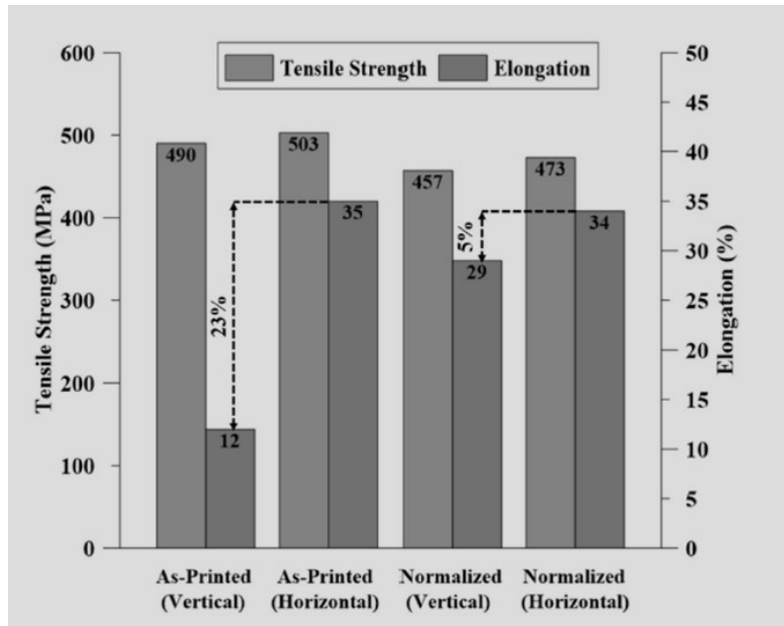


Figure 2-40 Strength variation in WAAM mild steel[64]

Plangger et al. deposited high strength of Böhler X90-IG with classification G89 6 M21 Mn4Ni2CrMo using CMT power source to build 18mm thick blocks. The impact toughness displayed by this built at room temperature were maximum and minimum of 94J and 84J with hardness in the range of 220-440 HV as shown in Figure 2-41(a & b)[65]. The impact toughness values were quite lower than expected, possibly due to the higher hardness observed in the process map. Hence, the cooling rate was a factor to be highlighted here.

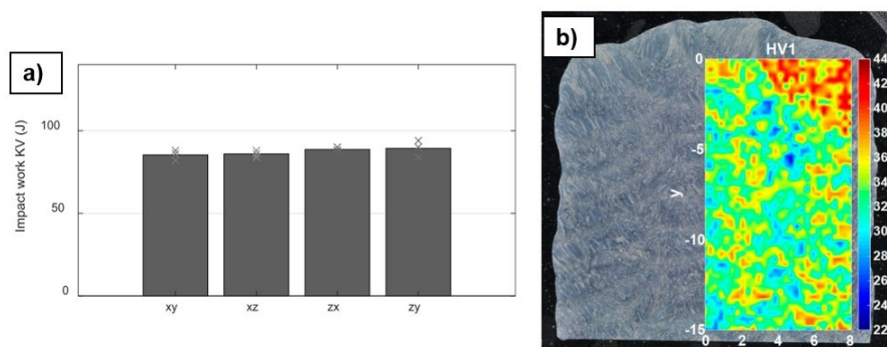


Figure 2-41 WAAM Böhler X90-IG as reported by[65]

(a) Impact toughness room temperature (b) Hardness map

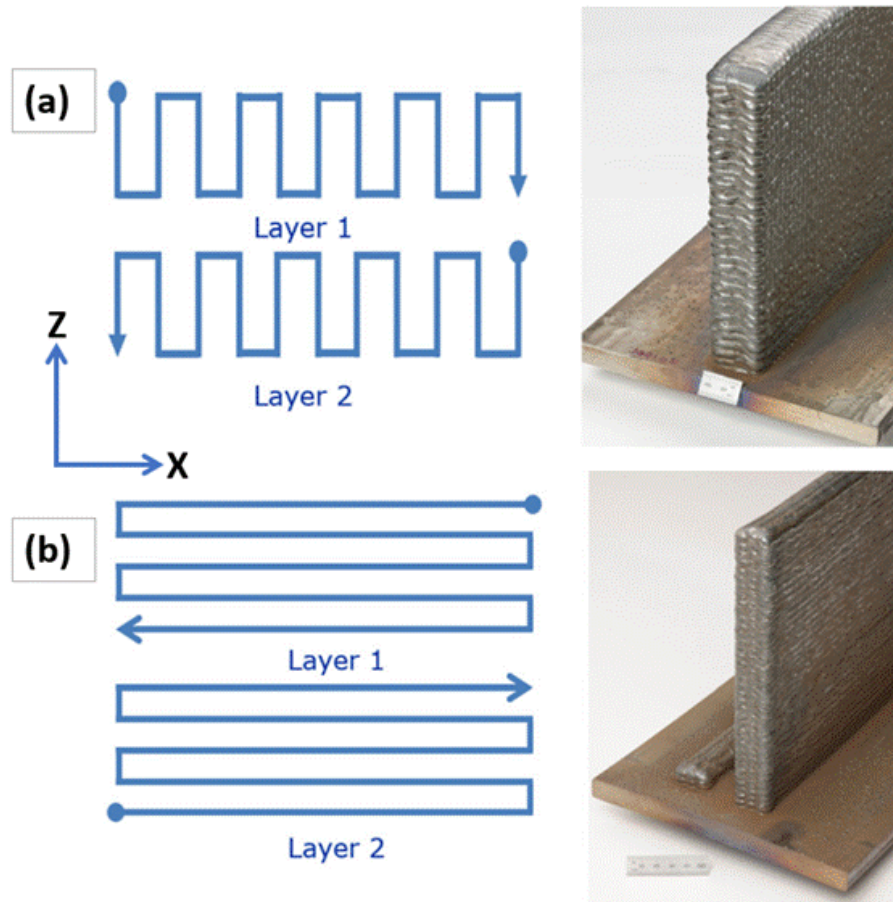
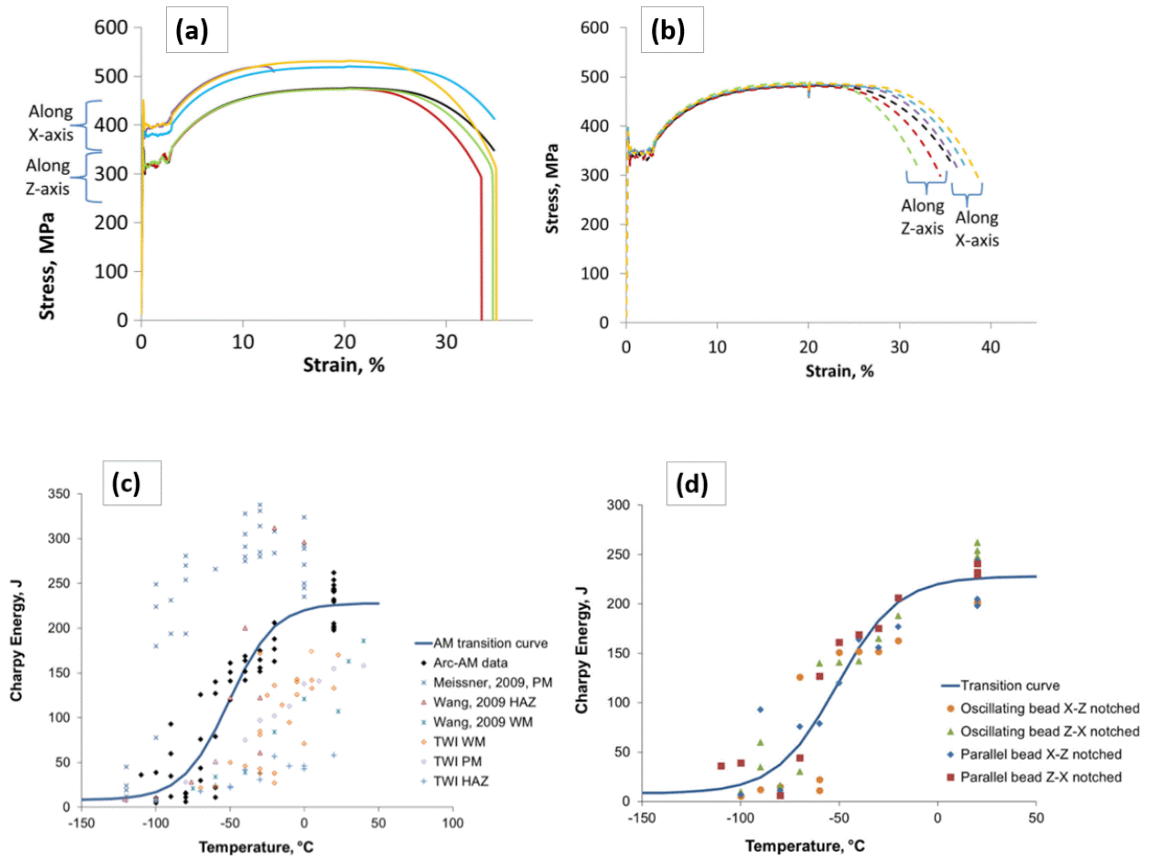


Figure 2-42 WAAM Deposition strategies[66]

(a) Oscillatory (b) Parallel

Moore et al. used oscillatory and parallel deposition strategy as presented in Figure 2-42 (a & b) to deposit ER70S, although the WAAM controlling parameters (layer height, current, voltage, interpass temperature, oscillation width, robot travel speed, wire diameter) and wire chemistry were not mentioned. Since the variation of %C, Ni and Al in this chemistry are responsible for the microstructural and mechanical variation, hence there was no basis to compare with the work carried out by other key authors using the same ER-70S wire. Their work showed that AM steel structure display higher mechanical properties when loaded in the X-direction as display in Figure 2-43 (a & b) and lower mechanical and toughness properties when the notch crack is located in the Z-direction Figure 2-43 (c & d).



Moore et al. deposition using oscillatory strategy gave close to isotropic properties in both direction for strength and Charpy impact toughness, Although the CTOD fracture toughness was performed at -20°C using SENB specimen, which is less conservative than CT specimen. BS 7448 Part 1 was used for the analysis. The CTOD results were currently the highest published for any WAAM steel current in the literature [66].

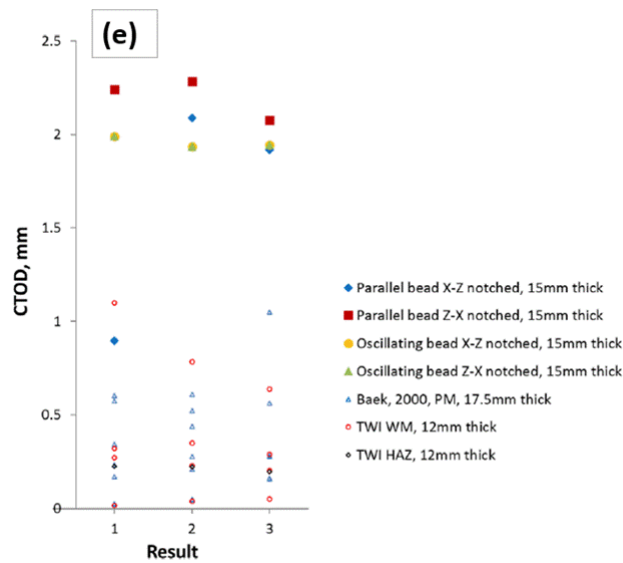


Figure 2-43 Mechanical properties of WAAM ER70S[66]

(a) stress-strain (Fig.2-42. b) (b) stress-strain (Fig.2-42. a), Charpy data (c) WAAM-AM, (d) ER-70s WAAM (e) CTOD for WAAM ER-70s & reference data(-20°C)

2.3.4 Texture, grain sizes and misorientation angles

Rodrigues et al. [53] in their EBSD crystallography characterisation of the WAAM produced sample, stated that there was no significant preferential orientation adopted by the grains, as shown in Figure 2-44(a), although the various build orientation EBSD profiles were not shown in their work to correlate this claim. More so, the misorientation profile in Figure 2-44(c) did not show the correlated misorientation angle. More misorientation measurements in different build orientations would have been ideal to conclude whether the high angle grain boundaries (HAGBs) or low angle grain boundaries (LAGBs) are the controlling structures in the build. Although his uncorrelated measurement showed that HAGs with misorientation above 40° are dominating, hence the good impact fracture results obtained. Ghaffari et al. presented the WAAM built EBSD IPF of ER70S-6, he further claimed that the non-textured structure as shown Figure 2-44(b) was printed, despite the various builds orientation EBSD profile was not shown in his work [64].

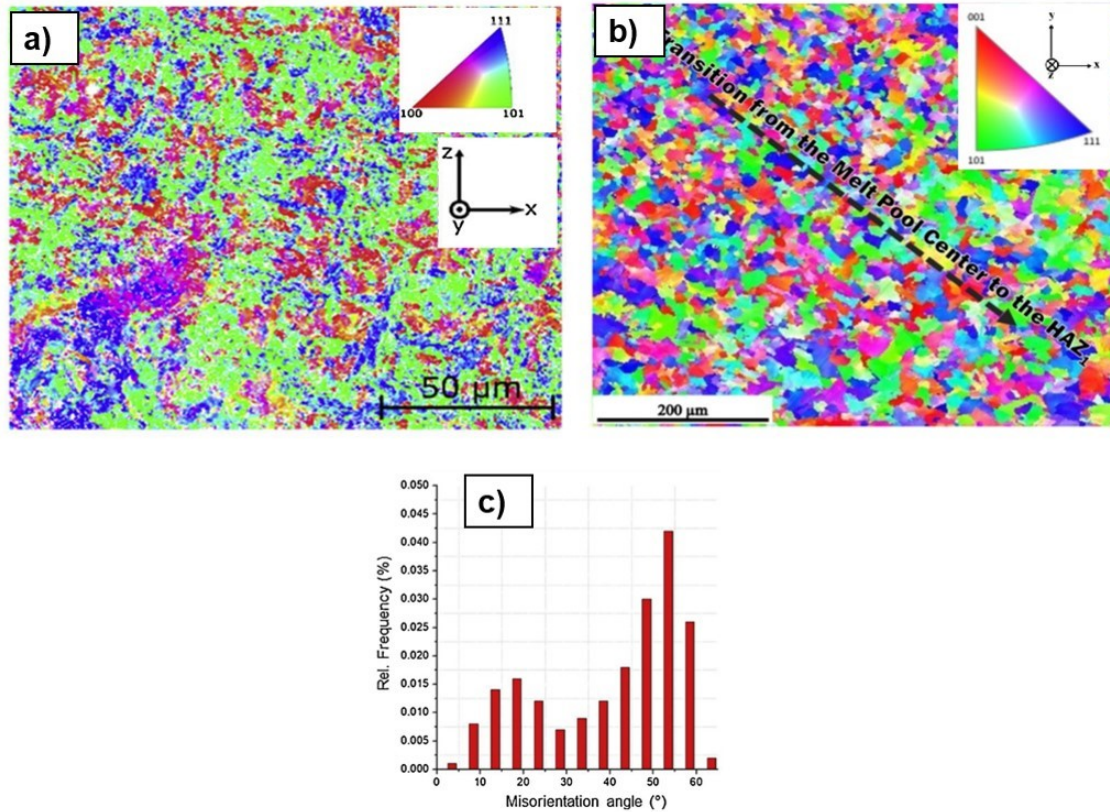


Figure 2-44 EBSD analysis, IPF as reported by

(a) ER110S-G (b) ER70S-6 (c) Misorientation angles of (a) [53] [64]

2.3.5 Gas metal arc welding flexibilities

The use of cold wire to improve bead geometry and deposition efficiency has been in use for some time now. This process has successfully been used to join structural plates of heavy thickness at a fast filling rate with improved properties [67]. Costa et al. feed CW into the GMAW process, as shown in Figure 2-45(a), to lower the amount of residual stress since the addition of CW lowers the amount of heat energy given to the base metal. This extra non-energised wire acts as a heat sink in the molten pool [68]. Mohammadjoo et al. coupled a CW feeder into a tandem SAW system, as shown in Figure 2-45(b), to improve the productivity of the system and deposited bead aspect ratio. The effect of the CW addition on the TSAW system at a heat input of 2.15kJ/mm is shown in Table 2-3. There was an increase in the reinforcement area and a decrease in the CGHAZ area, penetration area, dilution and CGHAZ hardness. The effect of the addition of CW

into the system was a considerable reduction of heat into the melt pool and consequently an increase in cooling rate. CW addition was responsible for the formation of finer prior austenite grains (PAGs) and reduction in the M-A constituent within the CGHAZ, as seen in Figure 2-46 and Figure 2-47. The M-A fraction obtained in both cases, using the SEM and the optical micrographs, revealed that CWTSAW had 2.2% and 1.9% reduction in the CGHAZ respectively [69]. The decrease in the PAGs size is responsible for the decrease in martensitic start temperature and consequently, the lower volume fraction of martensite [70].

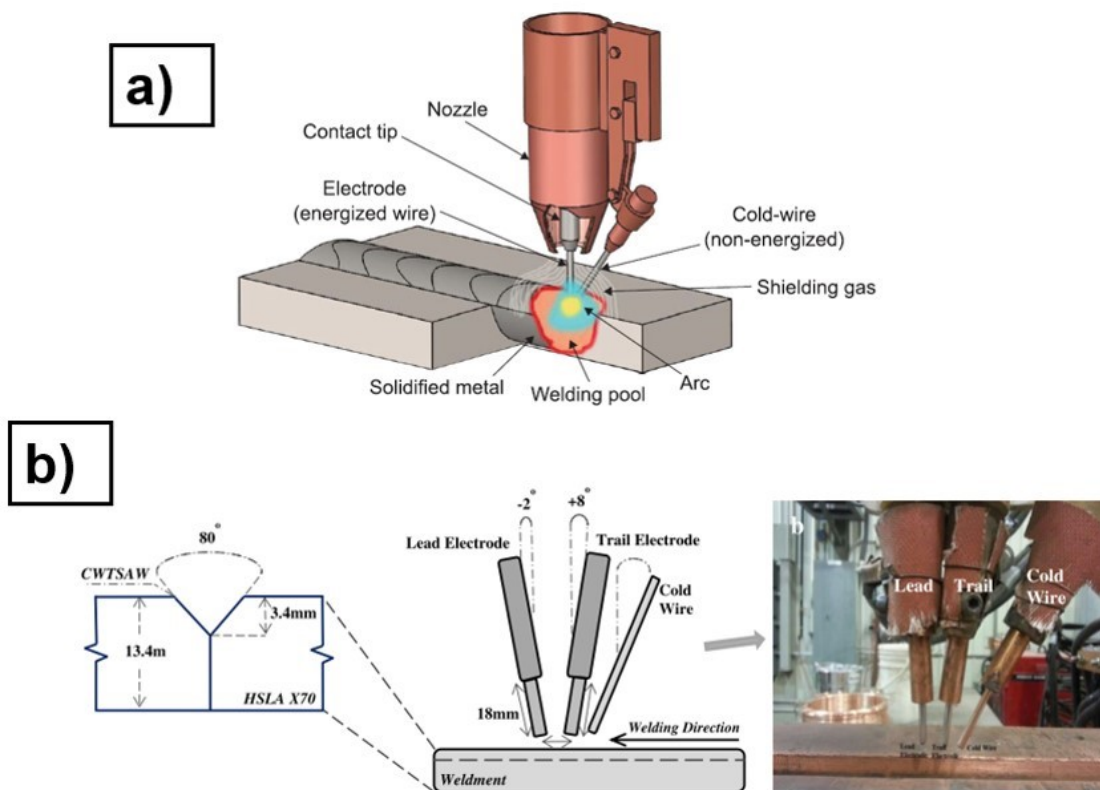


Figure 2-45 CW- GMAW process

(a) GMAW + CW [68] (b) Tandem + CW [69]

Table 2-3 Effect of CW addition on bead parameters [69]

Heat input: 2.15kJ/mm	CWTSAW(CWFS 8.5mm/s)	TSAW
CGHAZ area(mm^2)	19.10 ± 0.81	21.30 ± 0.63
Reinforcement area(mm^2)	24.90 ± 1.82	18.60 ± 1.50
Penetration area(mm^2)	86.50 ± 0.50	97.30 ± 1.04
Dilution	0.56 ± 0.01	0.65 ± 0.02
CGHAZ hardness(HV0.5)	222.93 ± 6.46	236.11 ± 7.02

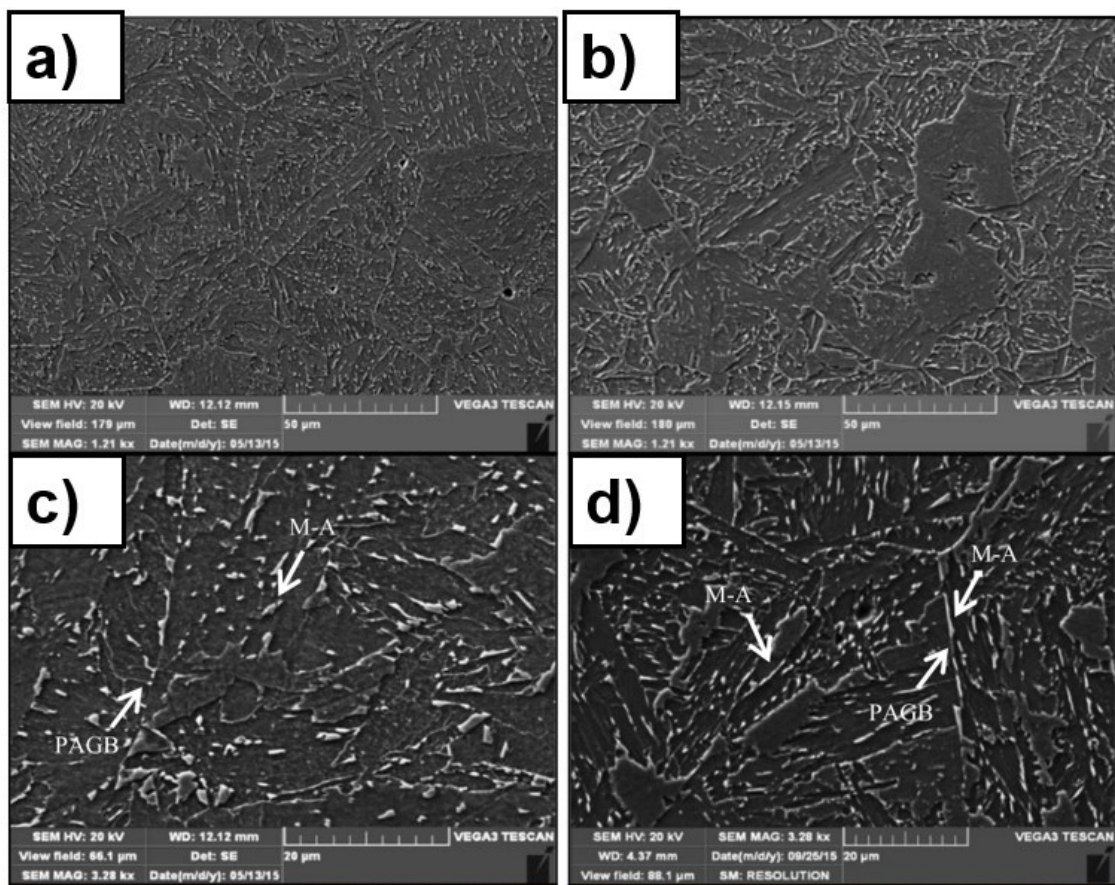


Figure 2-46 SEM images of CGHAZ [69]

(a & c) CWTSAW (b & d) TSAW

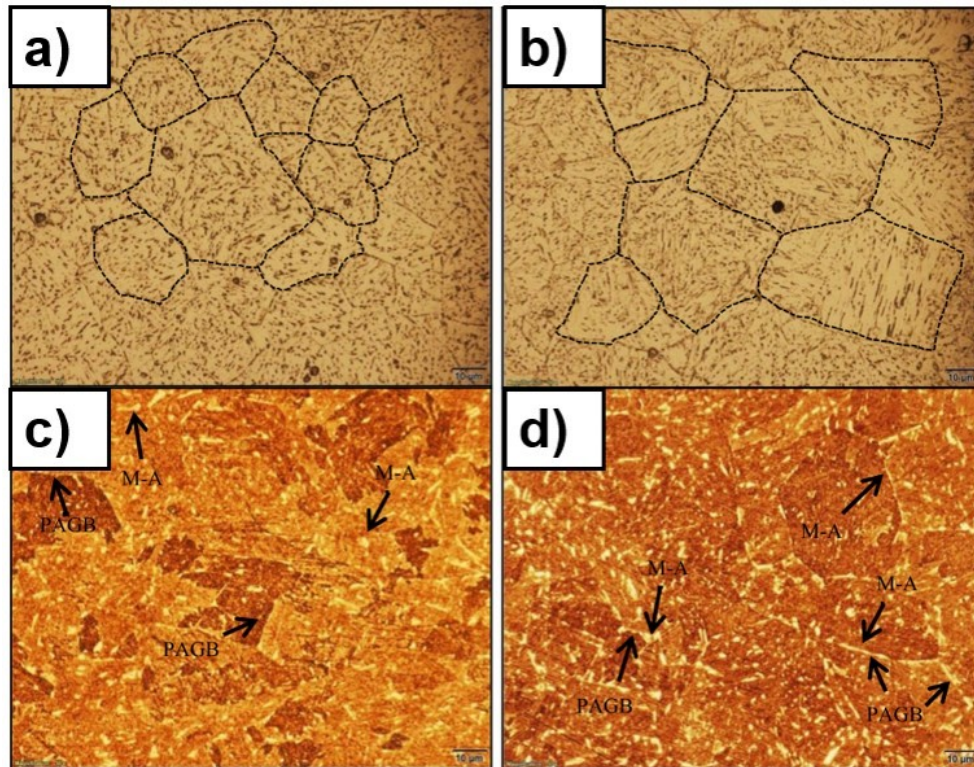


Figure 2-47 Optical micrographs of CGHAZ [69]

(a & c) CWTSAW (b & d) TSAW

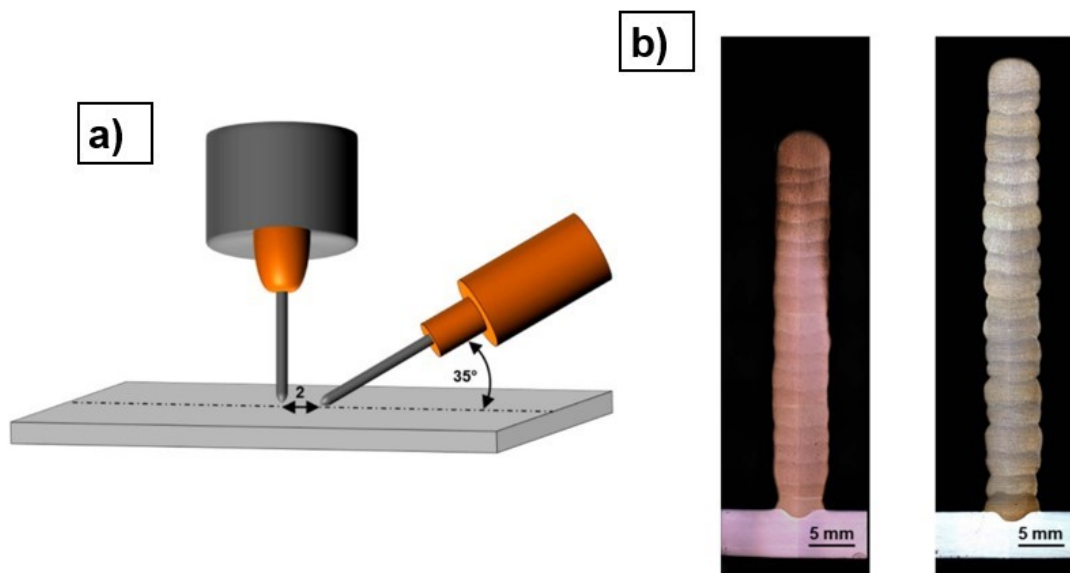


Figure 2-48 CW addition to WAAM [71]

(a) Feeding orientation (b) Geometry of deposited layers

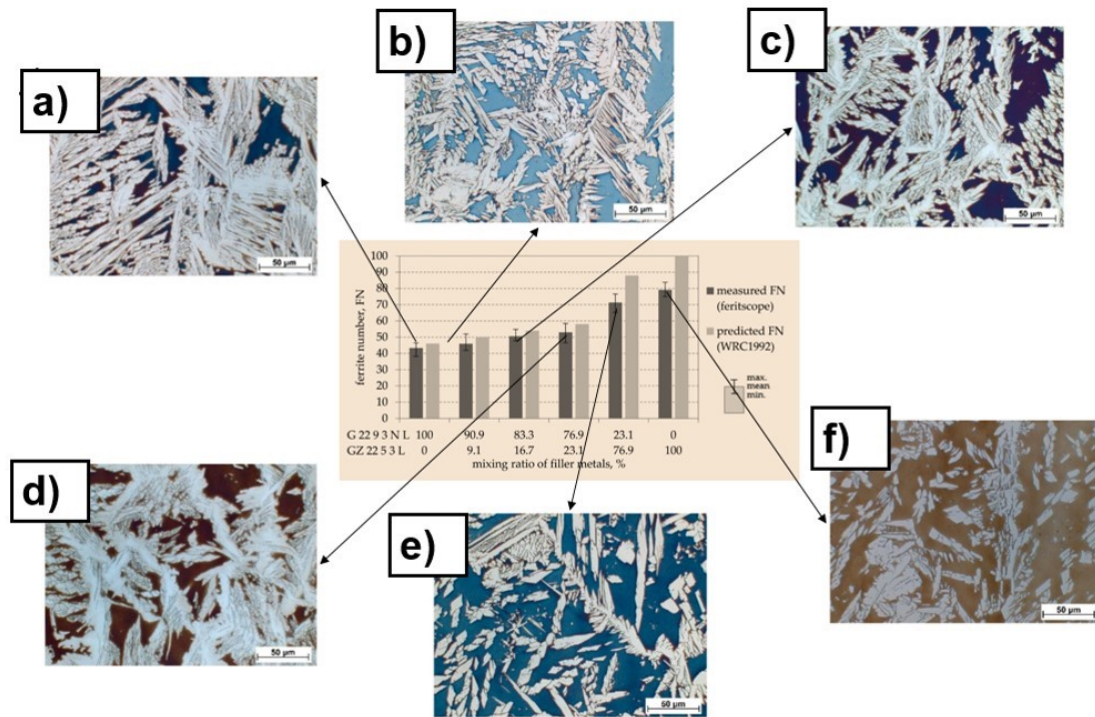


Figure 2-49 Relationship between measured ferrite and microstructural variation
[71]

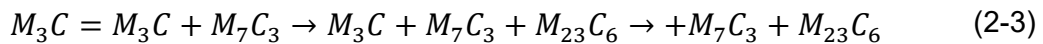
(a) 100% G 22 93 NL (b) 90.9% G 22 93 NL + 9.1% GZ 22 53 L (c) 83.3 / 16.7%
(d) 76.9 / 23.1% (e) 23.1 / 76.9 % (f) 100% G 22 53 NL

One of the key advantages of CW was deployed by Stützer et al. in adjusting the ferrite austenite ratio of wire and arc additive manufactured duplex stainless steel components. It is on record that the ferrite-austenite ratio depends on the alloy chemistry and cooling rate, and since WAAM parts are always subjected to a very low cooling rate, the final microstructure could be deleterious if the cooling rate is not controlled. It has also been reported that excess austenite formation in duplex stainless steel is as a result of the increased cooling time due to the repeated dwell time in the temperature ranges of 800-1200°C which is common with WAAM [71]. By the addition of CW into the GMAW melt pool, Stützer et al. have been able to reduce the size of the austenite plates formed during the reheating of subsequent layers due to the low heat dissipation in WAAM layering. The GMAW filler used here is G 22 93 NL, and its weld microstructure (all weld metal) as shown in Figure 2-49(a) is largely a Widmanstätten structure with the austenite subduing the ferrite. The ferrite number (FN) here is just less than 40 [71]. With

the addition of GZ 22 5 3 L CW (similar composition), the FN has been improved, from 39-72%, as shown in Figure 2-49(b-f). The desired microstructure can be engineered in WAAM. This is one of the unique features of the WAAM process.

2.3.6 HSLA steel (2.4 Cr-1Mo steel)

Low alloy 2.4 Cr -1Mo steels are used in the oil and gas sector, refinery and nuclear plants due to their high resistance to tempered embrittlement, hydrogen corrosion cracking and high temperature creep resistance [72]. This steel contains micro alloying elements such as niobium, vanadium and titanium, which aids grain refinement and strength increase [73]. The stable dispersion of carbides precipitates in these steels is needed for improved creep resistance. Carbides coarsening and precipitation of more stable carbides during in-service operation of this steel often soften the HAZ [74]. The as-welded and PWHT microstructure consist of lath bainite in a fine distribution of carbides (Figure 2-50(a & b)) and coarsened lath bainite structure and carbides. The HAZ of the as-deposited condition is composed of transformed austenite and tempered ferrite, as shown in Figure 2-50(c-e) [72]. The carbide transformation sequence during tempering and normalising of 2.4Cr-1Mo steel is shown in Figure 2-51. Changes in carbide transformation at the tempering temperature of 700°C are shown in Eq (2-3)



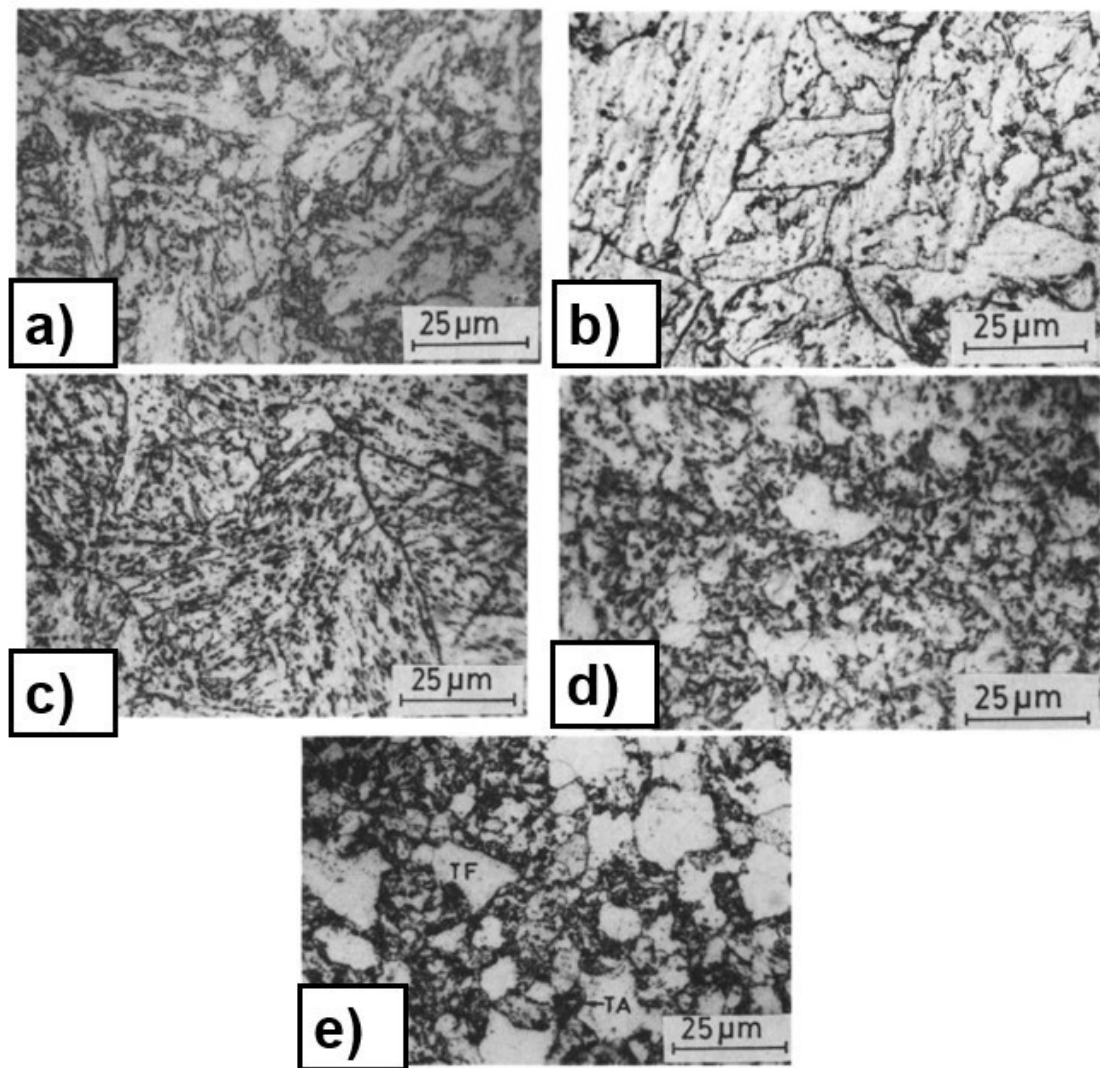


Figure 2-50 Microstructure of 2.4Cr-1Mo[72]

(a) Weld metal (b) Microstructure of PWHT, HAZ in as welded (c) CG region
(d) FG region (e) ICG region

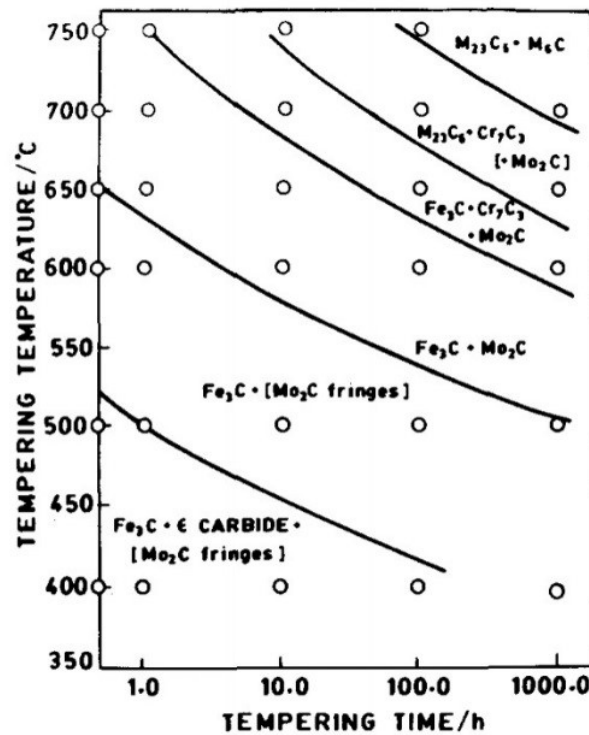


Figure 2-51 Carbide transformation sequence during PWHT (normalizing and tempering)

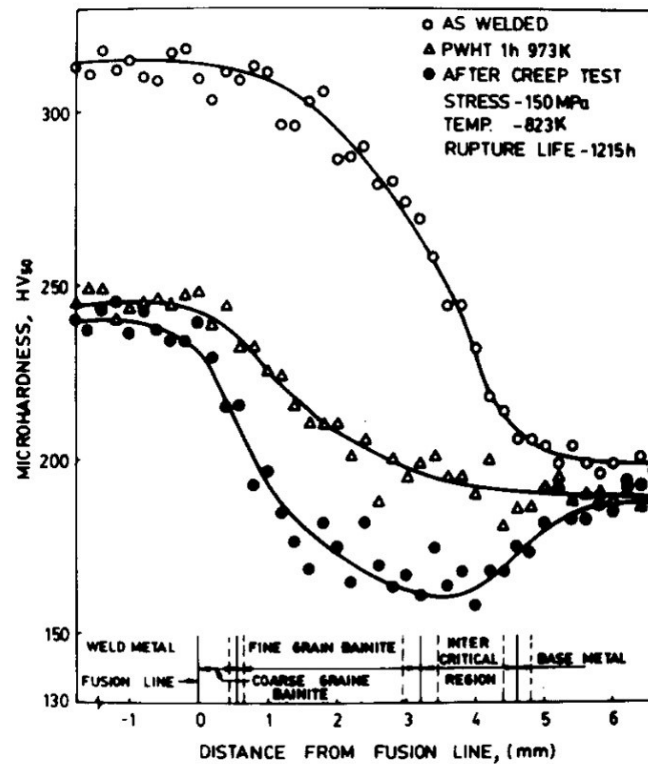


Figure 2-52 Micro-hardness variation of 2.4Cr-1Mo steel

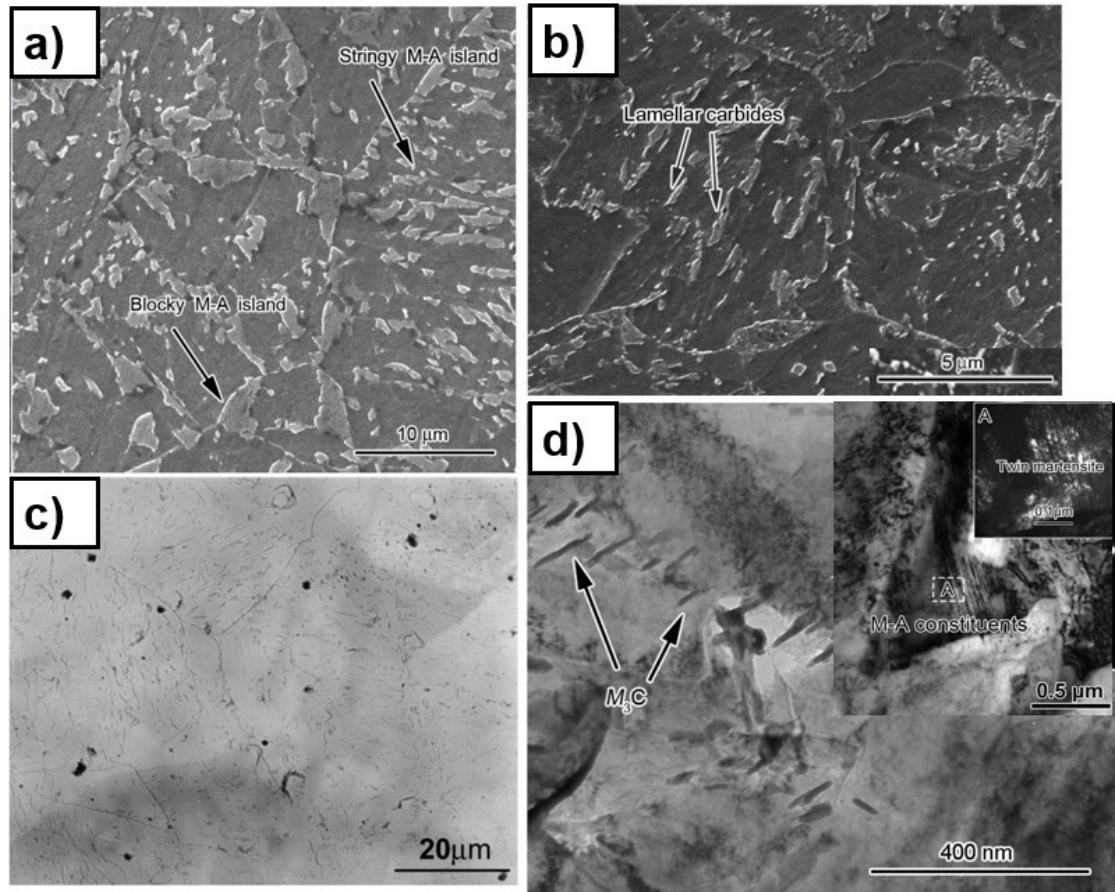


Figure 2-53 SEM and TEM images of 2.4Cr-1Mo steel

(a) Normalised 2.4Cr-1 Mo steel [75] (b) Tempered at 700°C [75] (c) Precipitate distribution in re-austenitised at 1200°C [73] (d) Normalised and tempered at 700°C [75]

According to Jiang et al., the degree of distribution of carbides (spherical, rod-shape, lumped shaped and lamellar (Figure 2-53(a-d)) in the granular bainite and the type and sizes of M-A island present (blocky or stringy (Figure 2-53(a)) which is a direct consequence of the bainite ferrite recrystallisation, determines the key impact toughness properties of the 2.4Cr-1Mo steel [75]. According to Tao et al., the weakening of the effect of precipitation in the granular bainite, which causes coarsening of the carbides leads to decrease in hardness [76]. The hardness profile distribution, as shown in Figure 2-52 reveals that the base metal hardness is in the range of 200HV due to a large amount of proeutectoid ferrite in the microstructure. The HAZs' hardness is generally lower than the weld metal region

due to the presence of coarse $M_{23}C_6$ and M_6C precipitates and absence of Fe_3C and Mo_2C precipitates.

2.3.7 The fatigue life of high strength steel

The ease of crack formation in a material is influenced by the provisions of discontinuities due to the manufacturing process, mechanical properties, composition of alloys and dislocation in the materials. It is estimated that half of all mechanical failures in the industry are due to fatigue loading with the cost of failures estimated at 4% of the total US gross national product [77]. In general, fatigue failures start at points of high-stress concentration such as non-metallic inclusions, pores and discontinuities (defects in a structure). The texture of a material due to the preferred crystallography orientation as a result of anisotropy and grain sizes is also a contributing factor to the fatigue strength of a material [78].

Two design approaches to fatigue life determination of engineering components are currently used, namely the stress-based approach and the strain-based approach. In the former, materials are only allowed to deform in an elastic manner, and all local plastic strains are neglected, while in the latter, both the elastic and plastic response to applied loading is utilised. Three different grades of steel with grain sizes, as shown in Figure 2-54(a-c), were examined by Chapetti et al. in line with Hall-Petch postulation as shown in equ (2-4) which predicts an increase in the fatigue strength and yield strength with a decrease in grain size. The smaller the grain size, the more resistance is posed by the grain boundary to deformation, as shown in Figure 2-55(a) [79]. A linear relationship was also presented by Chapetti et al., see Figure 2-55(b & c), showing the relationship between smooth fatigue limit with grain size and with UTS. The ultra-fine grade steel with grain sizes ranging from $1 \pm 0.5\mu m$ and ultimate tensile strength range of 1183MPa (UFGS2) performed better than the coarse grain steel with grain size ranging from $12.5 \pm 7\mu m$ and ultimate tensile strength of 691MPa (CGS). This is due to the strong microstructural bonds that need to be overcome as a result of its structure.

$$\sigma_y = \sigma_0 + Kd^{-1/2} \quad (2-4)$$

where σ_y & d are yield stress and grain size, σ_0 & K are constants independent of *grain size*. Other methods to improve fatigue life are through solid solution strengthening and precipitation hardening. Low carbon steel performs better under low cycle fatigue life ($<10^3$) with higher elongation property [80]. Compressive residual stresses that increase the fatigue life of a component are compared to tensile residual stresses. The composition of steel is vital in depicting its fatigue properties; steel with high carbon content and tempered martensitic microstructure has higher fatigue properties and performs better than the coarse pearlitic microstructure.

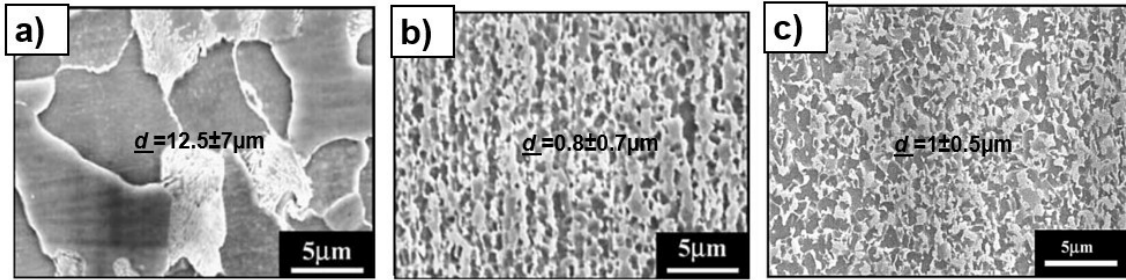


Figure 2-54 High strength steel microstructure [79]

(a) CGS (b) UFGS1 (c) UFGS2

The fatigue limit of notched (σ_e) component is related to the fatigue limit of the unnotched component (σ_{eR}) by the relationship given by Chapetti et al. [79]. in Eq (2-5)

$$K_f = \frac{\sigma_{eR}}{\sigma_e} \quad (2-5)$$

where K_f is the fatigue strength reduction factor. $K_f < K_t$ (theoretical stress concentration) for ductile material. The three different SCFs used for the ultra-fine grain steel responded differently to fatigue strength, as shown in Figure 2-55(d). Pang et al. classified fatigue damage mechanism into intrinsic defects (IDs) which are inherent in the materials that control the microstructure, static and dynamic properties of the steel. Processing defects (PDs) which are based on the processing technology, can be avoided. Low strength ductile steel only

nucleates at persistent slip bands and grain boundaries but not at process defects [81].

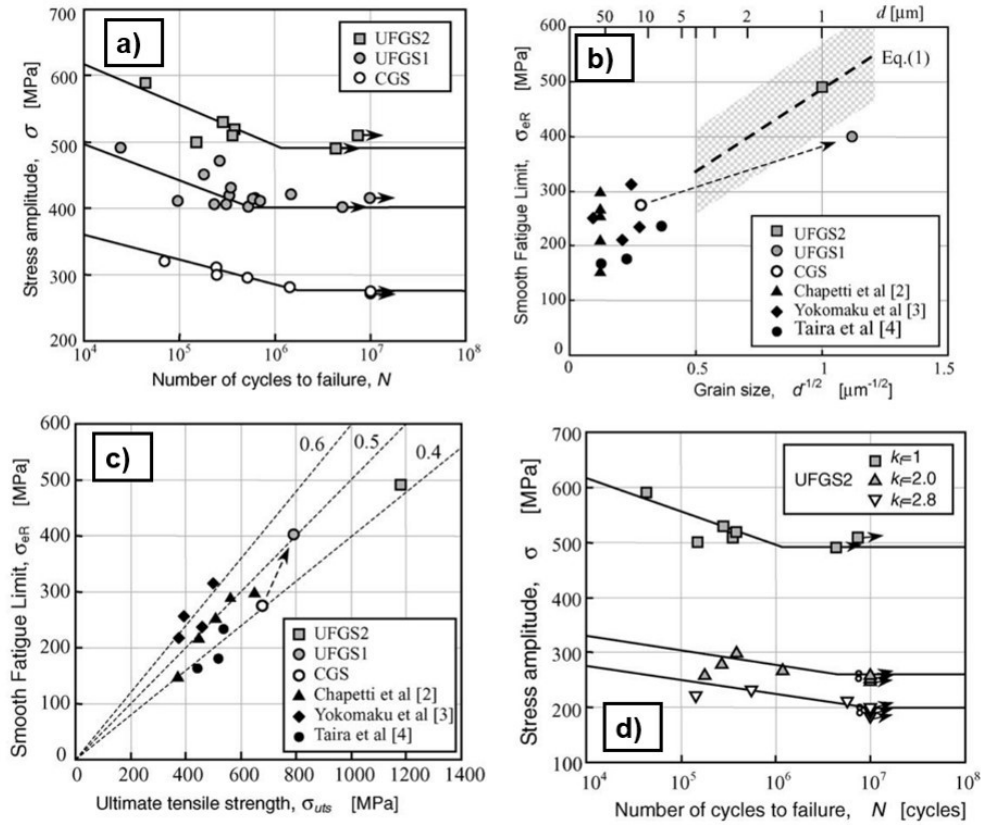


Figure 2-55 Fatigue strength relationship [79]

(a) S-N curve (b) Smooth fatigue limit and grain size (c) Smooth fatigue limit and strength (d) Stress concentration factor effect

2.3.8 Fatigue crack growth rate of mild steel

The effect of increasing stress ratio on the da/dN curve of low carbon steel was studied by Chapetti et al. [82]. Only the Paris region was used for the regression analysis (the most used region in engineering design). The monotonic (r_p) zone, cyclic plastic zones (r_{pc}) and crack tip opening displacement (δ_{max}) were evaluated using Eqs (2-6), (2-7) & (2-8) [83] and plotted as a function of stress intensity, as shown in Figure 2-56.

$$r_p = \frac{1}{3\pi} \left(\frac{K_{max}}{\sigma_y} \right)^2 \quad (2-6)$$

$$r_{pc} = \frac{1}{3\pi} \left(\frac{\Delta K}{2\sigma_y} \right)^2 \quad (2-7)$$

$$\delta_{max} = \frac{1 - \nu^2}{2\sigma_{ys}E} K_{max}^2 \quad (2-8)$$

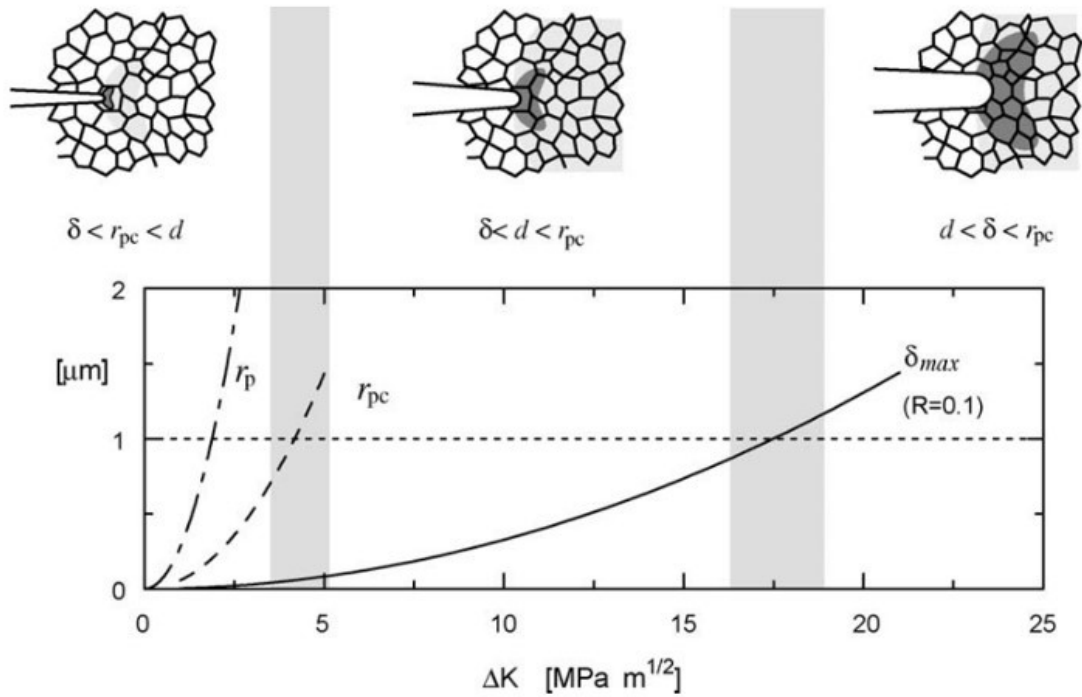


Figure 2-56 Effects of monotonic and plastic zone sizes [82]

K_{max} maximum applied stress intensity; σ_y yield strength; ΔK applied stress intensity range; ν Poisson's ratio; E Young's modulus. The key regions in the plot identified as applicable to fracture mechanics are the ($\delta_{max} < r_{pc} < d$) region which has been reported to be the order of grain size at threshold ($r_{pc} \sim 1\mu\text{m} = d$) [84][85]. In the second region $r_{pc} > d$ and $d > \delta_{max}$, hence, the process zone is smaller than the grain size, although propagation is ensured. In the last zone: $d < \delta_{max} < r_{pc}$ the process zone is far greater than the grain size, hence crack branching, and multiple cracking around crack tip is ensured. In general, crack

closure phenomenon, crack tip blunting phenomenon, crack tip stress field and crack tip environment are also influenced by K_{max} [82].

2.4 Structural steel use for the mainframe and Hub

The nacelle bedplate (mainframe) support the drives and the rest of the nacelle components and transfers the load from the rotor to tower. The mainframe is designed to withstand fatigue loading, extreme load conditions, and is strength and wear resistance. The hub is the connector between the blade and the main shaft; both are currently being manufactured by the expensive sand-casting process and followed by subtractive manufacture without the possibility for post-fabrication heat treatment. ENGJS-400-18U-LT SG iron or cast steel are used for the casting purpose. Ductile iron is often used for casting the mainframe, wind turbine drive shaft and hub.

These show graphite nodule shapes achieved by adding nodulising elements such as magnesium when viewed under the microscope, and this is different from the grey iron shape. This attribute helps the hub to resist the creation of cracks and make-up for improved ductility; typically used hubs are the tri cylindrical and spherical hub. Other processes of subtractive manufacturing, such as face milling, boring, back facing, and profiling, are carried out after the casting process. To achieve the required shape and configuration of the frame and hub desired, the disadvantages with this method of manufacturing is that It leads to wastage of materials by machining almost close to 10% of the unwanted material deposited by the sand

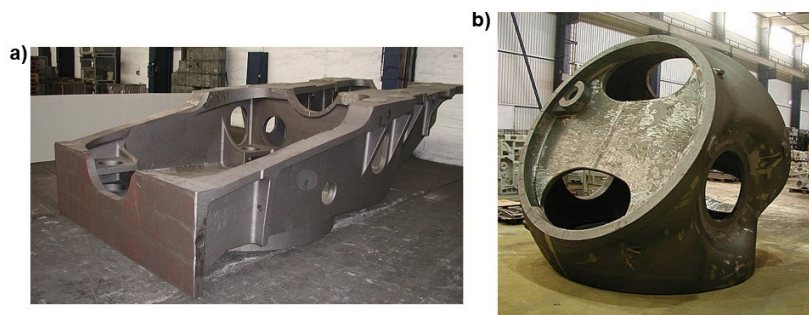


Figure 2-57 Wind turbine components

(a) Mainframe (b) Hub (Courtesy, MEUSELWITZ GUSS Eisengießerei GmbH Industriepark Nord)

Table 2-4 Properties of Cast alloys for casting (SN Cast Iron B.V, 2009) [87]

Item according to DIN EN 1563	Measure	Unit	EN-GJS-350- 22-LT	EN-GJS-400- 18-LT	EN-GJS-400- 18	EN-GJS-500-7	EN-GJS-600-3	EN-GJS-700-2	EN-GJS-800-2
			EN-JS 1015	EN-JS 1025	EN-JS 1020	EN-JS 1050	EN-JS 1060	EN-JS 1070	EN-JS 1080
Tensile Strength	R _m	min.M PA	350	400	400	500	600	700	800
0.2% Yield Strength	R _{p0,2}	min.M PA	220	240	250	320	370	420 ²⁾	480 ²⁾
Elongation Strength	A	%	22,0	18,0	18,0	7,0	3,0	2,0	2,0
Brinell hardness		HB	110-150	120-160	140-190	170-220	200-250	230-280	250-330
ISO-V impact test by -20 ± 2 °C				12,0					
ISO-V impact test by +23 ± 5 °C	K _v	min.J	17,0 ³⁾	14,0 ³⁾					
Modules of elasticity	E	GPa	170	170	170	175	175	175	175
Poisson number	ν	-	0,280	0,280	0,280	0,280	0,280	0,280	0,280
Compressive Strength	σ _{dB}	MPa	-	700	700	800	870	1000	1150
Fracture Toughness	K _{IC}	MPa · √m	31	30	30	25	20	15	14
Density		g/cm ³	7,1	7,1	7,1	7,1	7,2	7,2	7,2

2.5 Factors affecting mechanical properties of HSLA steels

The mechanical properties of steel and its alloys are vastly influenced by the chemical composition of the material, pre- and post-processing conditions (rolling and heat treatment condition) [87]. Carbon content has been reported to affect the hardenability of steels. Martensite, austenite and pearlite constituents are determined by the amount of carbon in the chemistry; while austenite and ferrite are being formed with manganese content above the 1.8-1.8 wt%, the toughness of metal is improved by nickel and manganese addition alongside molybdenum which can lower the transformation temperature of austenite ferrite. Titanium and aluminium addition is responsible for austenitic ferrite nucleation and must be added in a controlled amount because high levels of addition cause inclusions in weld metals as a result of its reaction with oxygen [88]. The effect of carbon content on the mechanical properties of dual-phase steels was investigated, and it was found that the mechanical properties are controlled by the fraction of martensites and ferrite in the phases [89]

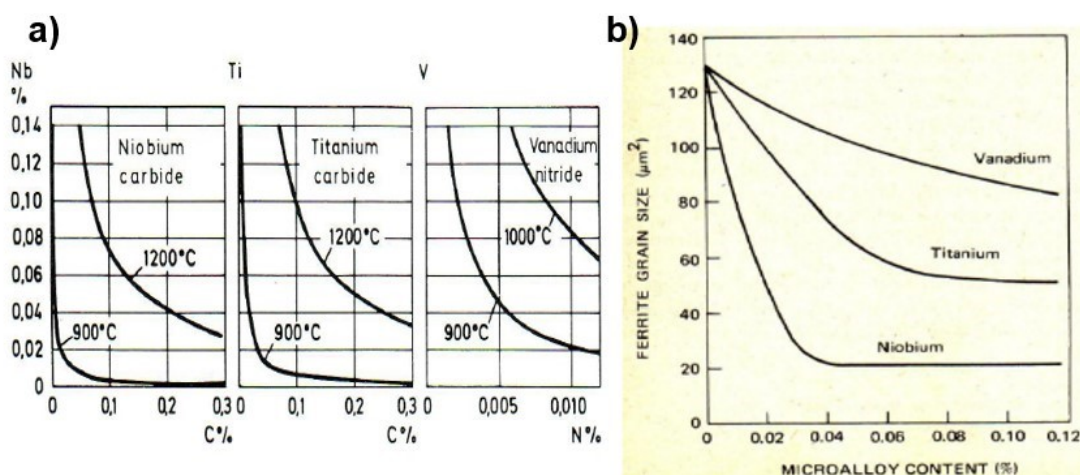


Figure 2-58 Effect of Nb, Ti & V content on the ferrite grain size of as hot rolled low carbon steel [90]

These elements, Al, Nb, Ti & V, in low fractions between 0.03 and 0.10% help to inhibit grain growth at the austenitising temperature because of their presence as dispersed carbides nitrides and only high temperatures above 1200°C, as seen in Figure 2-58, can make the phase inhibiting grain grow into solution, thereby increasing the grain size.

2.6 Effect of alloying elements on microstructure

The static and dynamic properties of weld metals are fully dependent on the microstructural variation of the as-deposited weldment. These microstructural variations depend on the following factors: maximum temperature of the thermal cycle, cooling thermal gradient of the weld pool (rate of cooling) and chemical composition of the weld metal, which also depends on the chemistry of wires being used, type of shielding gases, material being joined together and the welding process to some extent [91]. In meeting a compromise between strength and toughness in the weld metal, certain alloying additions to the composition must be selectively done to avoid brittle phase formation, inclusions and retained austenite, which leads to harder martensitic formation.

Molybdenum and chromium are good in promoting bainitic formation, while manganese and nickel increase the hardenability of the weld metal by helping in the formation of more acicular ferrite, and nickel on its own helps in solid solution strengthening. Some inclusions such as Al₂O₃, Ti₂O₃, SiO₂, MnO and MnS act as nuclei for acicular ferrite formation; they also help to pin down austenitic grain boundaries thereby reducing their sizes, which in turn, according to Hall-Petch postulation, supports that the reduction in the size of the grains leads to an increase in yield strength and improvement in fatigue limit [79] .

Another work by Bose Filho et al. also proved that increasing the Ti content of weld metals helps to increase hardenability alongside molybdenum, chromium and nickel, and the common inclusions in welds with low Ti contents are manganese and silicon. It was shown in their work that the hardness range for acicular ferrite is 267HV and that of bainite is 325HV, while that of martensite was 390HV, and a maximum hardness of 250HV was recorded for ferrite and acicular ferrite with 20% of martensite and austenite fraction [91], as shown in Figure 2-59.

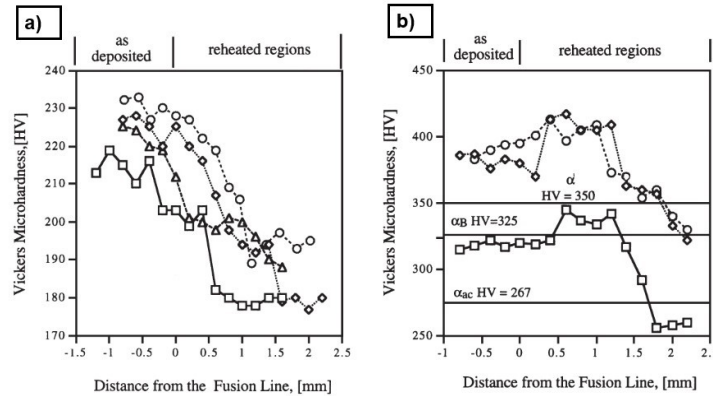


Figure 2-59 Hardness variation across the fusion line for α_{ac} , α_B , α^1 [91]

(a) Manual welding (b) Submerged arc welding

Aluminium and molybdenum in low quantities help to create fine grain structures and cause weldability issues similar to copper in high quantities. In general, carbon reduces ductility and toughness properties but in turn increases the strength, hardness and wear resistance as does chromium, which also improves resistance to corrosion in high quantities.

Tantalum and niobium increase toughness and tensile strength in steel. In general, micro-alloying elements such as Nb, Al and Ti help to increase the strength and toughness of weldments [92]. Martensites are formed at a high cooling rate in C-Mn steel, while at low cooling rates acicular ferrite or side plate ferrite (Widmanstätten ferrite) with some coarse grain acicular ferrite are formed. Further lowering the cooling rate will generate high linear welding energies that will cause the austenite to transform into polygonal or equiaxed ferrite and carbide ferrite aggregate [93]

2.7 The effect of oxide inclusion in HSLA steel welds

Inclusions in weld metals are possible through three modes, namely, the chemistry of the base metal, the chemical composition of the consumables, and the welding process and shielding gas composition in use [94]. Some inclusion has devastating effects on the overall mechanical properties of the entire weld while some inclusions such as oxygen content help to promote good fracture toughness of the weld metal. Since some inclusions have far-reaching benefits

for the weld metal, it will be important to devise means of improving the enhancement of these beneficial inclusions while suppressing the deleterious ones [95].

It was established that oxides of the Ti-Mn and Si-Mn-Al amorphous phase act as nuclei for intragranular nucleated BCC grains, also called acicular ferrite. If this type of ferrite is well aligned in the microstructure in terms of inclusion sizes and orientation to form an interlocking structure, they could be beneficial to the mechanical properties of the weld metal, such as ductility and toughness [95][96]. The interlocking nature of acicular ferrite also makes it resistant to crack propagation [97]. The optimisation of inclusion, size and geometry in a structure could be beneficial to the various steel strengthening mechanisms to improve toughness, ductility and strength.

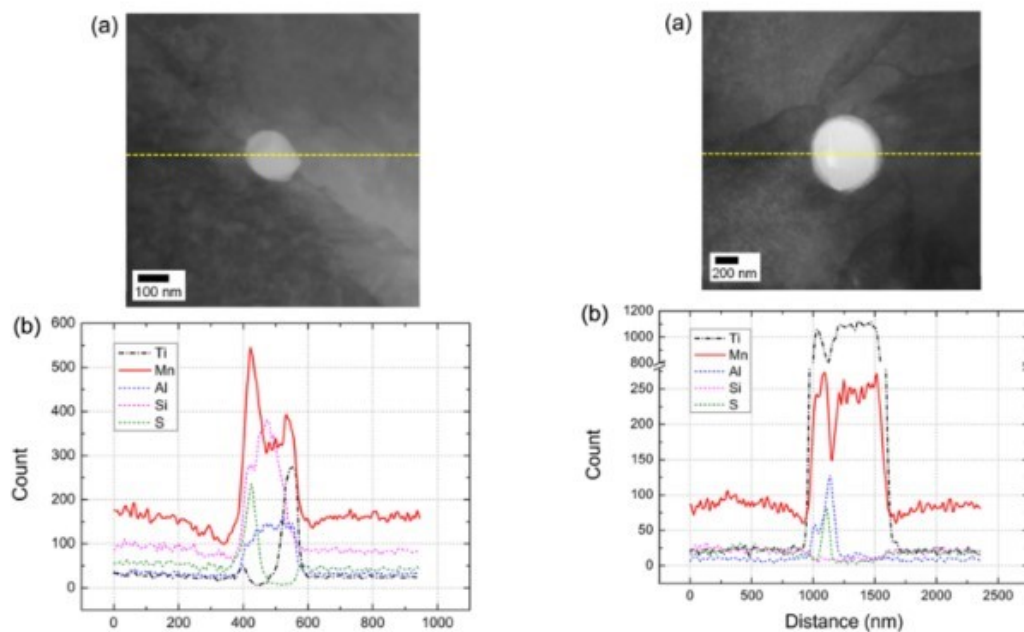


Figure 2-60 EDS Line scanning analysis across an inclusion in HSLA steel weld

(a) T70, (b) T1000[97]

The effects of Ti composition on non-metallic inclusion formation and acicular ferrite nucleation in HSLA steels weld metals were investigated, as shown in Figure 2-60. It was found that Ti depletion affected the strength of the inclusions in the microstructure with the formation of bainite in the columnar zone of the

higher Ti composition and interlocking acicular ferrite in the lower Ti composition, while the strength increases slightly in favour of the higher Ti composition.

The titanium-rich oxide in the inclusion was changed from MnTi₂O₄ rich spinel to the ilmenite phase, which consists of a solid solution of Ti₂O₃ and MnTiO₃ [97]. Another similar work carried out by Harrison et al. reveals that inclusions are necessary for grain boundary pinning of weld microstructures. They also stated that oxide rich inclusions in HSLA steels' weld microstructure are responsible for austenite to ferrite transformation with high inclusion content reducing the austenite grain size by favouring high-temperature transformation formation structures such as Widmanstätten, secondary side plates and polygonal ferrite. Oxide inclusions also aid the nucleation of fine intragranular ferrite [98].

Table 2-5 Non-metallic inclusion common found in HSLA steel weld metal [94]

Element	Carbide	Nitride	Nonmetallic Inclusions
Fe	Fe ₃ C	—	FeO, FeS
Si	—	Si ₃ N ₄	SiO ₂ , (Fe,Mn)O–SiO ₂ , Al ₂ O ₃ –SiO ₂ , (Fe,Mn)O–Al ₂ O ₃ –SiO ₂ , CaO–Al ₂ O ₃ –SiO ₂
Mn	—	—	(Mn,Fe)O, MnO–SiO ₂ , (Mn,Fe)S
P	—	—	—
S	—	—	Sulfides
Ni	—	—	—
Cr	Cr ₇ C ₃ Cr ₂₃ C ₆	Cr ₂ N	Cr ₂ O ₃ , (Fe,Mn)O–Cr ₂ O ₃
Mo	Mo ₂ C, Mo ₆ C Mo ₂₃ C ₆	Mo ₃ N	—
Ti	TiC	TiN	Ti ₂ O ₃ , (Fe,Mn)O–Ti ₂ O ₃ , TiO, TiS, Ti ₃ S ₄ , Ti ₄ (C,N) ₂ S ₂
Zr	ZrC	ZrN	ZrO ₂ , ZrOS, ZrS, Zr ₄ (C,N) ₂ S ₂
V	VC	VN	FeV ₂ O ₄ , V ₂ O ₅ , VO
Nb	NbC	NbN	FeNb ₂ O ₆ , NbO ₂
Al	—	AlN	Al ₂ O ₃ , (Fe,Mn)O–Al ₂ O ₃ , CaO–Al ₂ O ₃ –Al ₂ ON ₃
Ca	—	—	CaO–Al ₂ O ₃ , CaO–Al ₂ O ₃ –SiO ₂ , CaS

The effect of CO₂ content on 950MPa class steel weld microstructures was investigated by Gouda et al. using different percentages of CO₂ in the shielding gas mixture. It was revealed that the weld with pure argon as the shielding gas consists mainly of a fine lath structure, as shown in Figure 2-61(a). Further analysis of the fine lath structure using TEM also revealed a retained austenite phase being formed as dark images. The increasing CO₂ content in the shielding gas leads to a degradation of the microstructure to produce a granular structure,

as seen in Figure 2-61(b & c). The TEM images for this also revealed some phases of ferritic lath structure and some retained austenite in the non-lath structure that consist of martensitic and austenitic constituents [99]

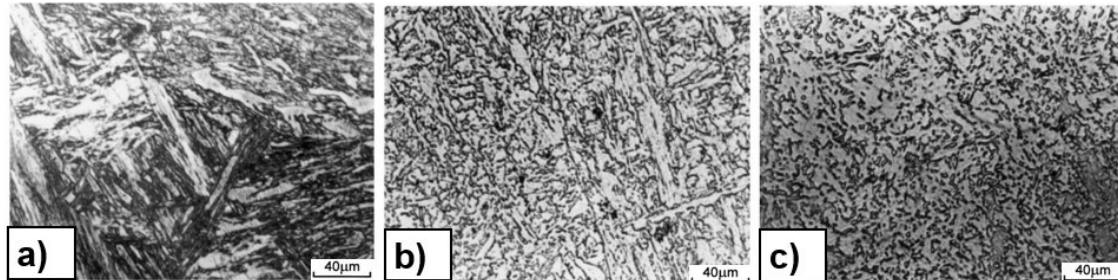


Figure 2-61 (a) pure Ar, (b) Ar + 15%CO₂, (c) Ar + 25%CO₂, [99]

The white and black areas shown in Figure 2-62(a) revealed a martensitic microstructure with a higher hardness which decreases with %CO₂ content and a bainitic microstructure with a lower hardness, which also decreases with increasing CO₂ content. Further microanalyses of the black and white areas revealed that the white contains more Mn, Ni, and Si than the black area, which supports the view that the white area has higher hardenability than the black [99].

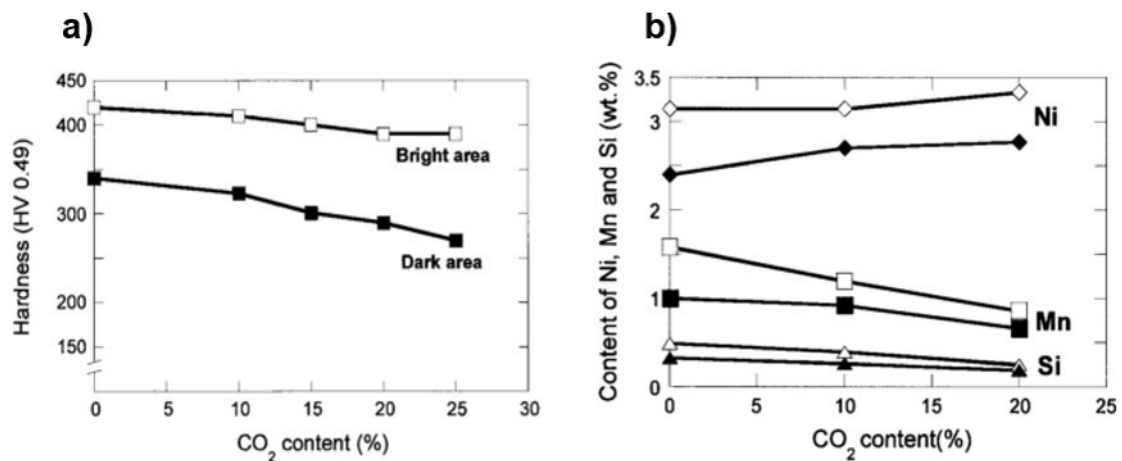


Figure 2-62 Effect of %CO₂ content in [99]

(a) Weld microstructure, (b) N, Mn & Si inclusion in the weld

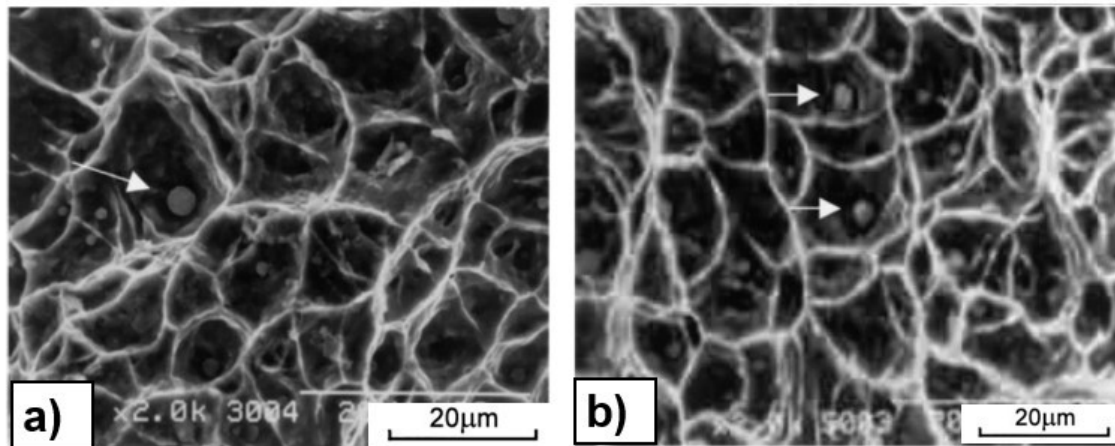


Figure 2-63 Fractography images of weld metal deposited [99]

(a) 15%CO₂ (b) 25%CO₂

Fractography images of weld metal deposition shown in Figure 2-63(a & b) revealed that the number of oxide inclusions (Mn, Si, Ti & Al) increased as the %CO₂ in the shielding gas increased from 15% to 25%. Hence a higher percentage of CO₂ content greater than 10% will lead to the formation of oxide inclusion in the welds and an increase in granular bainitic microstructure due to the depletion of Ti, Al, Mn and Si in the weld, which consists of predominantly martensite and bainite. [99]. In contrast, a similar work carried out by Olsen and Liu, reveals that coarse austenite grains with a large number inclusions greater than 0.2 microns are associated with high acicular ferrite microstructures and that the inclusion particle size is a function of the oxygen content [100]. Another work carried out by Ricks et al. reported that intragranular nucleated Widmanstätten ferrite is a component of acicular ferrite and that weld metal inclusions are specific nucleation sites for its formation [101]

2.8 Gaps in Literature

Based on the literature review, the following gaps were identified.

- Detailed studies of the relationship between different build strategies and how heat is being dissipated / effect of heat accumulation on microstructural evolution and mechanical properties have not been studied for WAAM low carbon and HSLA steel.

- The mechanism of the failure mode to explain the poor impact toughness results in the X-direction WAAM samples was not known, even with the detailed studies of Sridharan et al.[58].
- Rodrigues et al.[53] in his EBSD measurements of the WAAM produced sample stated that there was no significant preferential orientation adopted by the grains, although the various build orientation EBSD profile was not shown in their work to correlate this claim. More so the misorientation profile did not show the correlated misorientation angle. More misorientation measurement in different build orientation would have been ideal to conclude whether the HAGBs or LAGs are the controlling structures in the build.
- In all the WAAM studies to-date on structural steels, detailed relationships between the SW factors (notch radius and depth) and the deposition process parameters are lacking. The quantitative effect of SCF is lacking as well.
- Although it has been acknowledged that certain GMAW process is not suitable for WAAM as a result of its independent control of heat input effects, despite its capability for high productivity.
- Detailed studies on processes to optimise WAAM deposition rate for steel and at the same time enhance the mechanical properties are lacking
- Although a few mechanical studies of WAAM ER70S-6 & ER110S-G have been done, detailed studies on the dynamic and fracture mechanics properties of WAAM HSLA steels are lacking.
- Detailed knowledge on how to improve the functionality of WAAM as-deposited structures and its consequent on static, crack initiation, and microstructural properties are unknown
- In-depth knowledge on how to resolve the effect of heat input distribution across intersections of large WAAM structures is unknown.

3 METHODOLOGY

3.1 Introduction

This section enumerates the different aspects of the methodology followed during this research. A detailed discussion of each methodology is followed separately in the various subchapters.

3.2 WAAM manufacturing processes

3.2.1 CMT-WAAM deposition

Figure 3-1 shows a presentation of the set-up used for this study, as detailed in Chapters 4 – 7. An IRB 2400 ABB robot was used to deposit the materials on a fixed bed with a Fronius CMT (VR 7000 CMT) advanced power source.

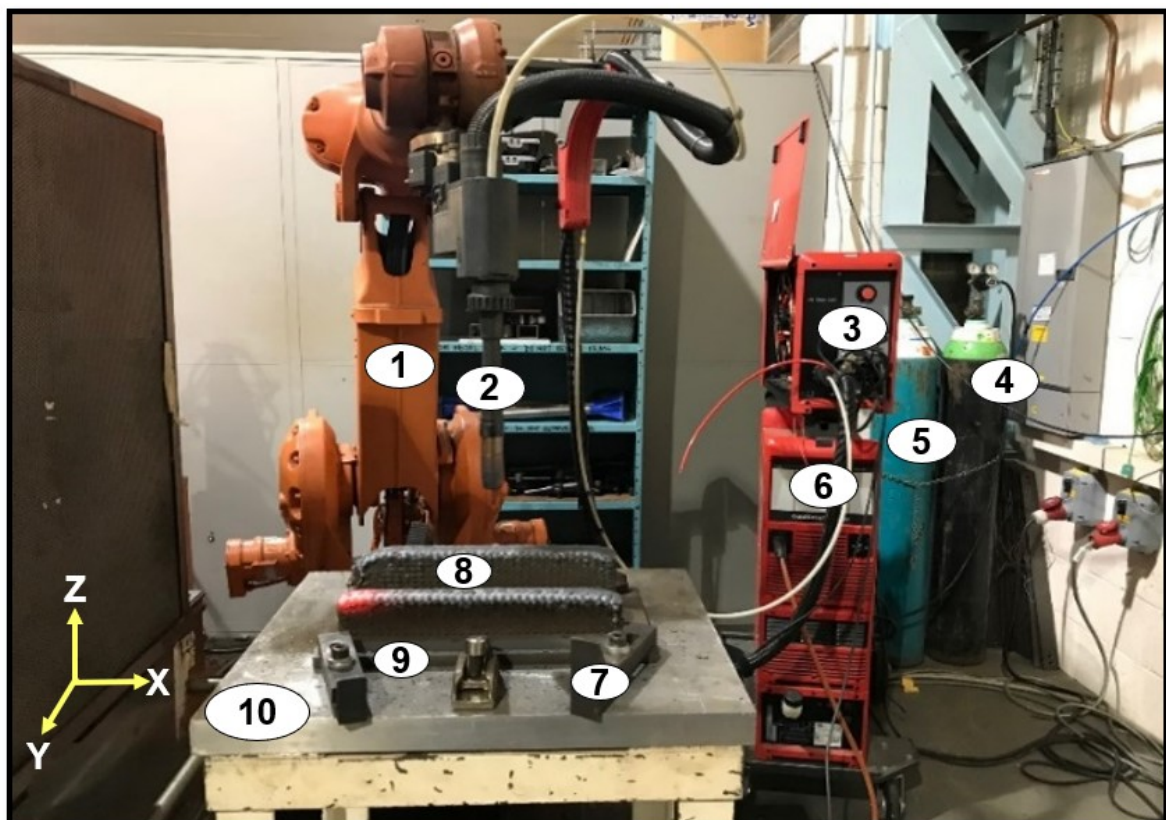


Figure 3-1 CMT WAAM Set-up for single and Oscillatory pass strategies



(1) ABB robot, (2) Torch, (3) CMT wire feeder, (4) Spec shield heavy, (5) Spec shield light, (6) CMT power Advance power source, (7) Clamps platform, (8) Oscillated Walls, (9) S275JR substrate, (10) Aluminium platform, (11) Fronius controller, (12) Oscillated wall, (13) AMV 4000 parameter monitor (14) Wire spool

The same power source was used to study both deposition strategies (single pass and oscillatory pass) in a single grade, and single pass in mixed grades. All studies were done in standard CMT mode. An AMV 4000 and Fronius computer monitor, as shown in Figure 3-1(11-14) were used to measure the instantaneous and average data (current; I, voltage; V, and wire feed speed; WFS).

3.2.2 Cold and interpass rolling

Cold rolling on the surface of the as-deposited structure was done on ER70S-6 structures using different loads. Interpass top rolling in-between each layer was carried out at room temperature on ER70S-6 samples alone using different loads. The rolling was mechanically aided by a hydraulic cylinder through a flat roller whose diameter and effective width were 100 & 20mm respectively. The roller

material was made of BS 4659 BH13. The rolling rig set-up is shown in Figure 3-2.

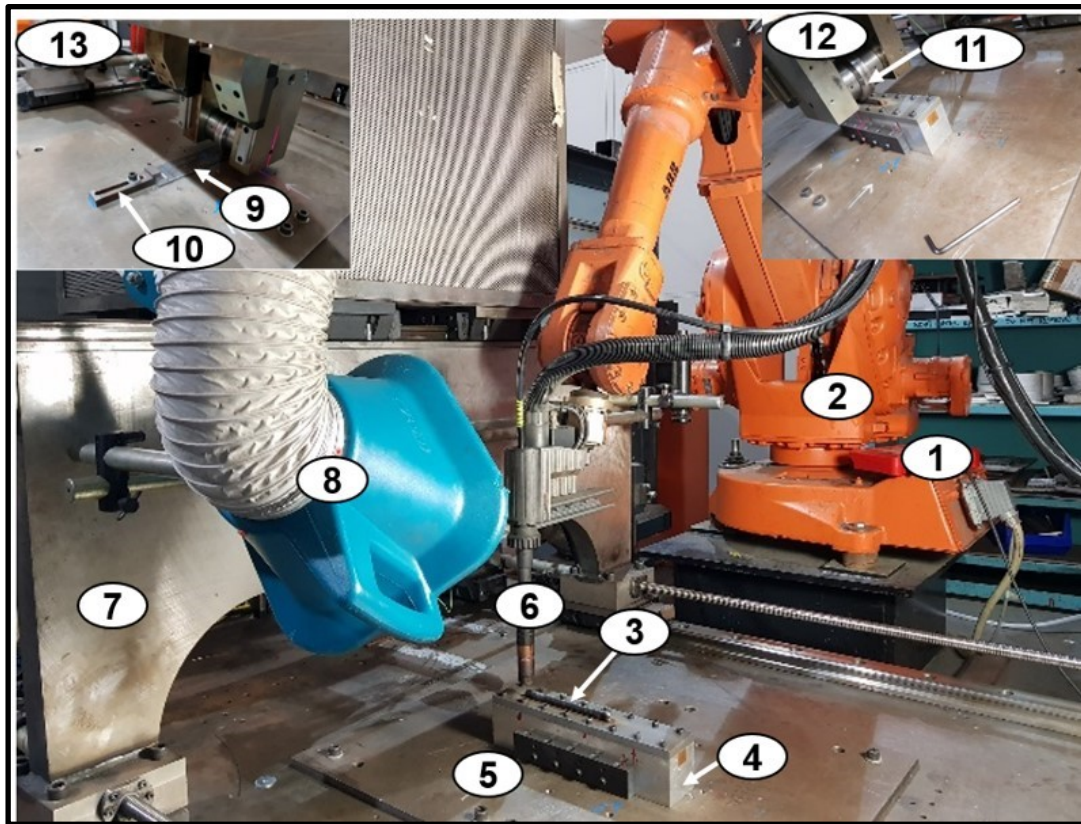


Figure 3-2 SW reduction set-up

(1) CMT PAM, (2) ABB robot, (3) WAAM built, (4) Copper base, (5) Steel base, (6) CMT torch, (7) Rolling rig, (8) Fume extractor, (9) WAAM sample, (10) Clamp, (11) Roller, (12) Top rolling, (13) Side rolling

3.2.3 Wire assisted arc additive manufacture (WAAAM) process

Figure 3-3 shows the presentation of the set-up used to deposit the bead on plate studies for the WAAAM process. The cold wire was fed through a Lincoln cold wire feeder. Before the experiment, the Lincoln cold wire feeder was calibrated to determine the relationship between the set WFS and the actual WFS, as shown in Figure 3-4. Furthermore, the set V, I and HI were monitored using the AMV4000. An IRB 2400 ABB robot was used to deposit the materials on a fixed bed with a Fronius CMT (VR 7000 CMT) advanced power source. Although GMA pulse synergy was used in this study, detailed set-up schematics and feed

orientation of the cold wire set-up used for the studies and for building the structures will be presented in Chapter 8.



Figure 3-3 Wire assisted arc additive manufacture

(1) AMV 4000, (2) Shielding gases, (3) Fronius CMT advance power source, (4) Cold wire feeding configuration, (5) Exhaust extractor, (6) CMT torch, (7) Processing monitoring camera, (8) Steel substrate, (9) Deposited beads, (10) Cold wire feeding set-up (feed angle and centre distance), (11) Cold wire feeder (12) CTWD set-up.

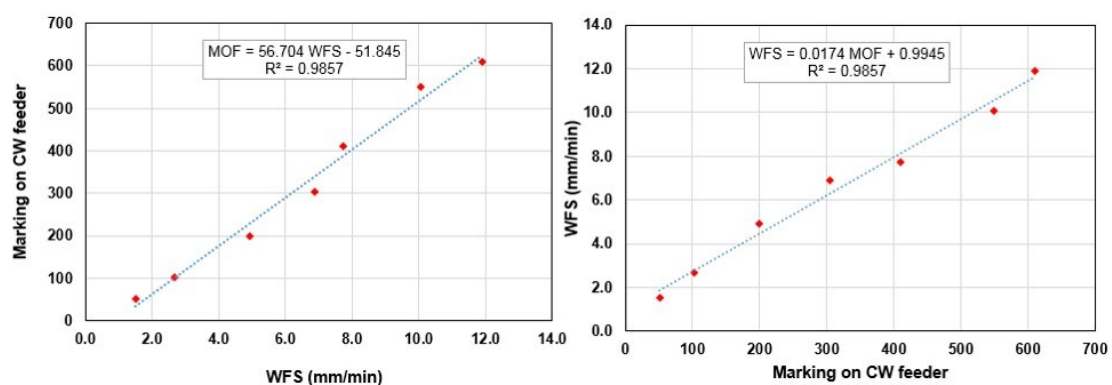


Figure 3-4 Cold wire calibration chart

3.3 Materials

3.3.1 Substrate and wires

The material substrate used in the experiment is S275 JR structural steel grade with a thickness of 12.5/15 mm and minimum yield strength of 275MPa. The wires used for the deposition were from Lincoln Electric (SupraMig Ultra; ISO 14341-A-G 50 5 M21/ 46 3 C1 4Si1; AWS A5.18 ER70S-6) and Böhler Union X 90 (EN ISO 16834-A-G 89 6 M21 Mn4Ni2CrMo; AWS A5.28: ER120S-G) and Böhler CM 2-IG (EN ISO 21952-A: G CrMo2Si; AWS A5.28: ER90S-B3 (mod) with nominal diameters of 1.0 mm and 1.2 mm. Their as-deposited chemical composition and mechanical properties are given in Table 3-1 and Table 3-2. Consumables were chosen based on the initial preliminary stress analysis using FE modelling done by the sponsors to depict their application requirements. The wire manufacturers test certificates are shown in appendices (Fig.9 & 10). Substrate material and thickness was chosen based on the readily available supplied steel material in the Welding Engineering research centre. The S355G10 + M low carbon steel material was used in conjunction with the WAAM steel to make a comparison on their fatigue crack growth rate and fracture toughness.

Table 3-1 Chemical composition (%W)

Alloys	C	Si	Mn	P	S	Cr	Mo	Ni	V	Cu	Nb	Ti	Al	Zr
ER70-S6	0.06	0.94	1.64	0.013	0.016	0.020	0.005	0.020		0.020		0.004		0.002
ER90S-B3	0.08	0.58	0.89	0.005	0.002	2.4	0.99	<0.1	<0.01	<0.1	<0.01			
ER120S-G	0.10	0.79	1.77	0.008	0.010	0.35	0.63	2.27	<0.01	0.07		0.07	<0.01	<0.01
S355G10+M	0.06	0.27	1.57	0.013	0.001	0.034	0.006	0.334	0.001	0.241	0.022	0.003		

Table 3-2 Mechanical properties

Alloys	UTS(MPa)	YS(MPa)	PCE (%)	CEV
ER70-S6	480.6	362.8	35.32	0.34
ER90S-B3	>500	>400	>22.0	0.96
ER120S-G	950-1180	>890	>15	0.76
S355G10+M	545	435	38.0	0.34

3.3.2 Gases

Shielding was achieved with two types of gases in these experimental studies: spec shield light with a composition of 80% Ar and 20% CO₂, and spec shield heavy with a composition of 98% Ar and 2% O₂.

3.4 Microstructural characterisation

3.4.1 Sample Preparation

All metallographic samples were prepared and examined in the Metallurgy laboratory at Cranfield University. Metallography samples were extracted from the deposited structures in a unique matrix. They were mounted using a hot mounting, followed by subsequent grinding using a Buehler Metaserve 2000 and a Motopol automatic grinding machine. 240, 1200 and 2500 grit papers were used for grinding, and 0.05µm colloidal silicon suspension with an emery cloth was used for final polishing. The etchant used was as follows:

- 5% Nital (5% nitric acid in ethyl alcohol) for macrographs
- 2% Nital (2% nitric acid in ethyl alcohol) for micrographs.

3.4.2 Micro Hardness Measurements

A Zwick/Roell automatic hardness tester (Auto - C.A.M.S machine), as shown in Figure 3-5, was used with a diamond indenter and two lenses of magnification by 10X and 40X. The measured diagonals of the indentations were checked for validation before acceptance. A load of 500/300g for 15 seconds was used.

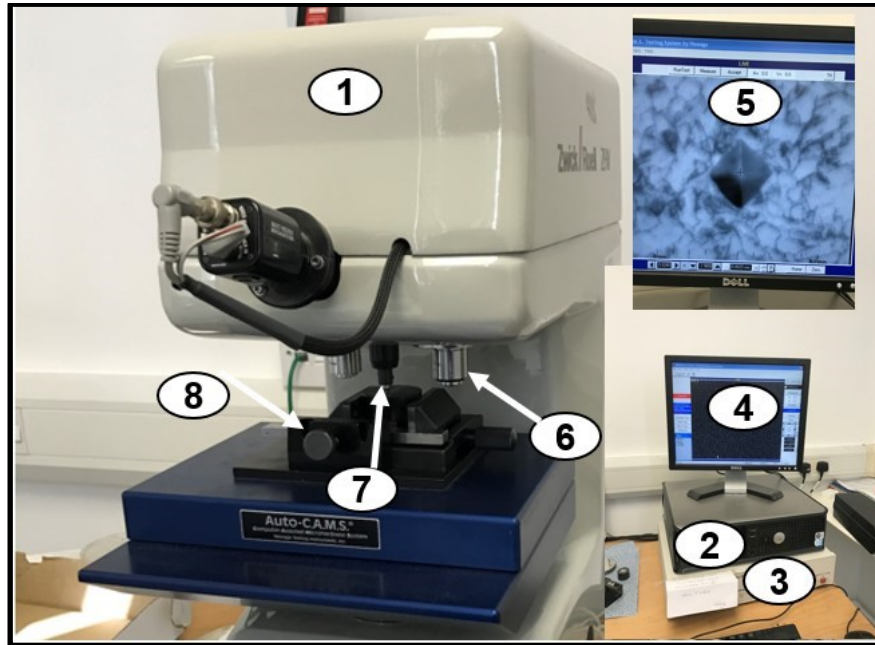


Figure 3-5 Hardness characterisation

- (1) Zwick/Roell hardness machine (2) CPU (3) 6 axis motion controller (4) Processing Monitor (5) Sample indentation (6) Lenses (7) Diamond indenter, (8) Sample anvil

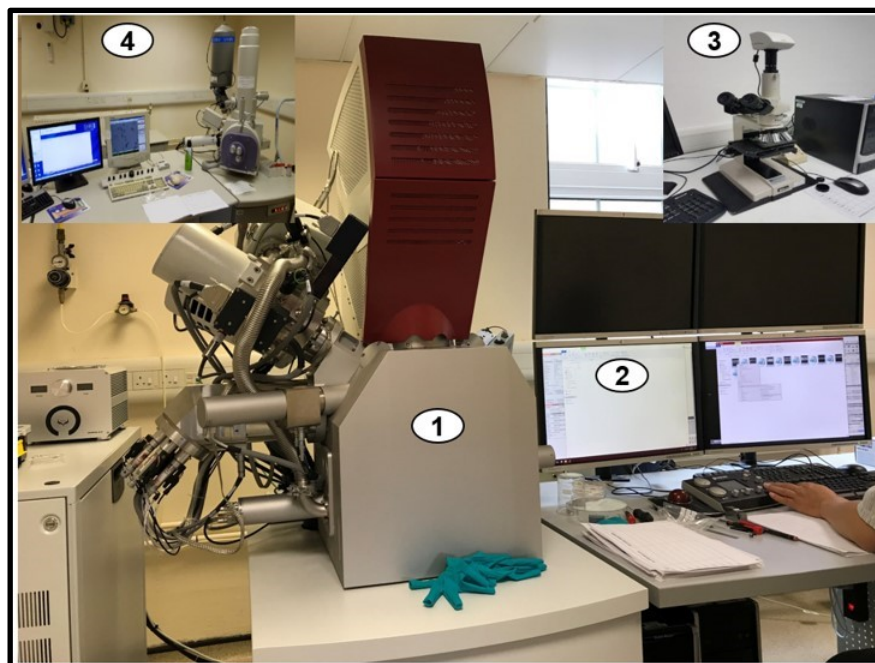


Figure 3-6 Microstructural characterisation

- (1) LYRA TESCAN for SEM & EDS, (2) Image processing monitor (3) Nikon microscope (4) Philips SFEG for SEM & EBSD

3.4.3 Optical microscopy and SEM

A Nikon Optiphot Microscope was used to observe the macro and microstructural features; after etching the samples, various magnifications were used on a case by case basis. An SEM FEI XL 30-SFEG equipped with energy dispersive spectrometry detector (EDS) and LYRA3 TESCAN, as shown in Figure 3-6, was used to characterise both the microstructure and tensile FCGR, and FT fractured surfaces.

3.4.4 Surface waviness and roughness characterisation

The macrographs from Section 3.4.3 were imported to solid works and Axio Vision software to measure specific geometrical attributes for SW and stress concentration factors. A Bruker Contour GT was used for 3D optical profilometry; it utilises vertical scanning/white light interferometry. A typical topography scan was taken of the X-Z plane with a single image area of $1.2 \times 0.95\text{mm}$. A total area of $5 \times 5\text{mm}$ was scanned for every sample to characterise the surface roughness. The results were analysed using VISION64 software to visualise the 3D contour and to calculate different surface roughness parameters.

3.4.5 EBSD and TEM

Electron backscattered diffraction (EBSD) was used to characterise the crystal orientation, texture, phases, grain statistics, grain boundaries and strain field using the same equipment presented in Figure 3-6. Transmission electron microscopy of TEM: (JEM-2100F, JEOL) operating at 200kV, was used to characterise the dislocation distribution and precipitates. The thin-foil TEM samples were prepared by focused ion beam (FIB).

3.4.6 XRD

The X-ray diffraction (XRD) pattern was measured by a D8 Advance X-ray diffractometer (Bruker AXS, USA) using a Lynxeye detector and Cu K α radiation, with wavelength $\lambda = 0.154\text{ nm}$, 30 kV, and 10 mA as shown in Figure 3-7 The scanning angle was between 10° and 90° , at a scanning speed of $10^\circ/\text{min}$, and a step size of $0.02^\circ/\text{s}$.

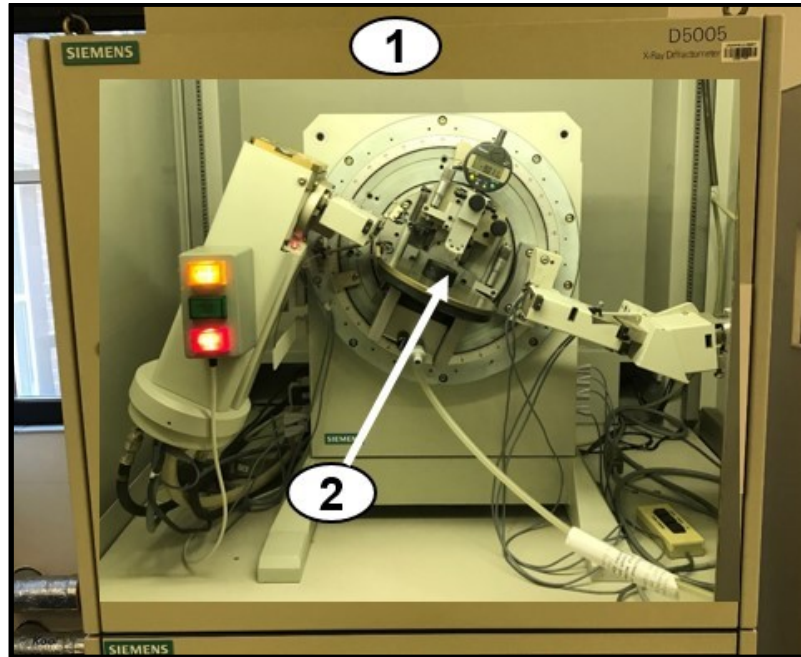


Figure 3-7 Phase characterisation

(1) XRD chamber, (2) Scanning sample,

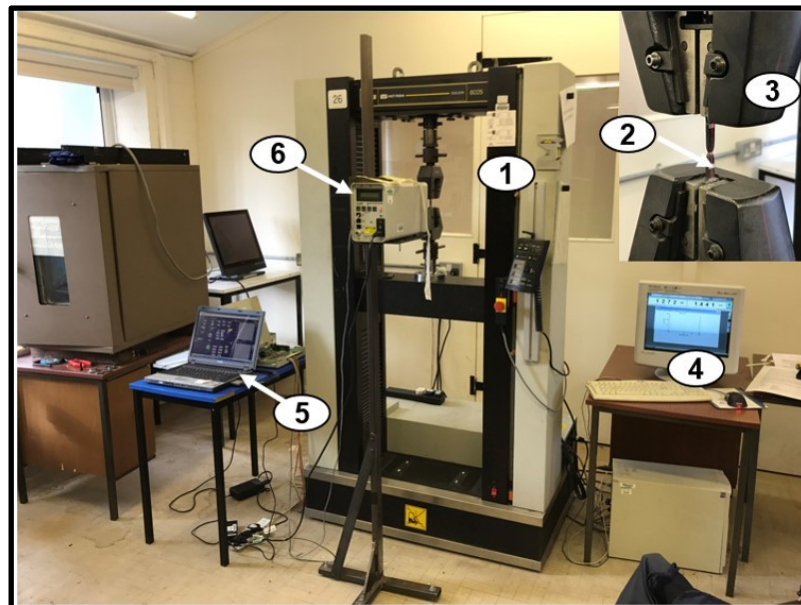


Figure 3-8 Tensile testing set-up

(1) Instron 8801 machine, (2) Tensile specimen, (3) Machine grips, (4) Monitor, (5) LabVIEW controller, (6) Extensometer

3.5 Tensile strength, fatigue and fracture toughness

3.5.1 Tensile test

The tensile test was carried out with an Instron 5500R electro-mechanical testing machine coupled to a laser extensometer for gauge length monitoring, as shown in Figure 3-8. A load cell of 30/100kN and a loading rate of 1mm/min was used in all testing. A unique tensile testing matrix was adopted in each case. Graphite coating was applied to the samples' gauge length to give a distinct visibility, and a reflective tape was placed on the gauge limits for extension monitoring during the testing. The samples were prepared per ASTM E8/E8M-16a standards. The txt file obtained from the LabVIEW monitor was uploaded to Excel software for further analysis in order to obtain the UTS, YS and PCE from the stress strain curves.

3.5.2 Digital Image Correlation (DIC)

Digital image correlation was used, together with the tensile machine, to study the interaction between the interphase of two graded structures to determine the effect of strain partitioning and the individual properties at a specific location on the graded structure, as shown in Figure 3-9. DIC tensile samples were sprayed with fluorescent substance to aid pattern generation during the testing. The calibration of the cameras was done by targeting the GI-06-WMB_{9x9} calibration chart at five images in a random orientation. Two high-speed cameras with a powerful lighting arrangement were used at a recording frequency of 2Hz and a resolution of 150 x 100 pixels. The basic principle of DIC involves comparing speckle patterns (speckle size and density) generated in a deformed sample with their corresponding location on a reference sample (undeformed sample). The data pattern processing is done by Istra 4D software, which generates an ASCII file that was used to produce a stress-strain curve, and the corresponding YS was determined at each point. A MATLAB generated code was used to extract the data for localised strain and Young's modulus.

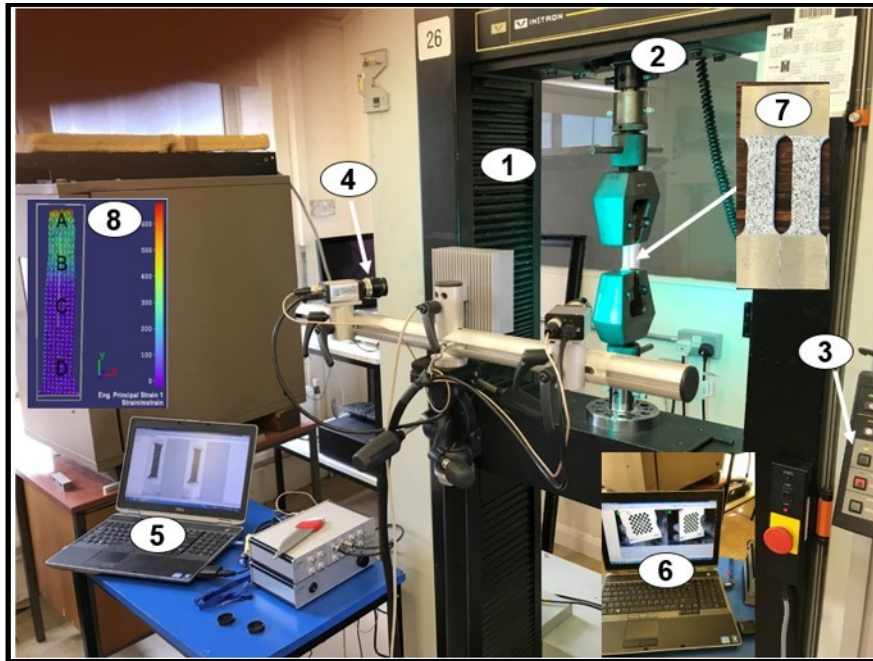


Figure 3-9 Characterisation of graded structure using DIC

(1) Instron tensile machine, (2) Pin end, (3) System control, (4) High-speed cameras, (5) LabVIEW Images control, (6) DIC images Calibration charts (7) Fluorescent painted tensile coupons (8) Strain response of tensile coupon

3.5.3 Fatigue life (FL)

A constant amplitude fatigue testing was carried out at room temperature with a frequency of 10Hz and stress ratio of ($R=0.1$) using an Instron machine equipped with a 250kN load cell, as shown in Figure 3-10. All fatigue samples were prepared as per ASTM E466-11 standards. Fatigue strength was derived using the staircase method as per ISO 12107 standard. A sinusoidal waveform and a load ratio of $R = (P_{min}/P_{max})$ was maintained at 0.1 in these tests.

3.5.4 Fatigue crack growth rate (FCGR)

The FCGR rate studies were carried out using an Instron 8801 servo-hydraulic test rig equipped with a 100kN load cell, as shown in Figure 3-11. All test samples were prepared as per ASTM E647 standards. A load ratio of $R = 0.1$ and frequency of 5Hz were utilised in all tests. The crack length was measured using a Baumer digital camera. All tests were done at room temperature and with a

compact tension C (T) specimen. Detailed testing procedures are covered in Chapter 7.

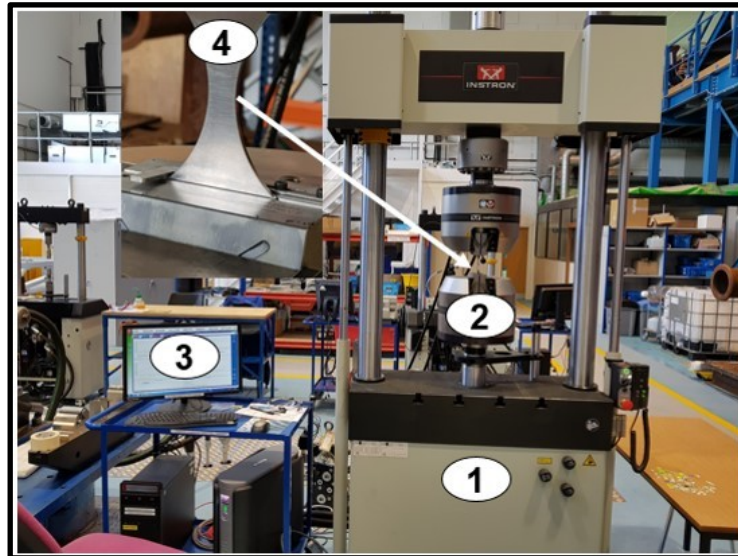


Figure 3-10 Fatigue response and strength characterisation

(1) Instron 250kN dynamic tester, (2) Machine grip, (3) Wave matrix monitor, (4) Fatigue sample

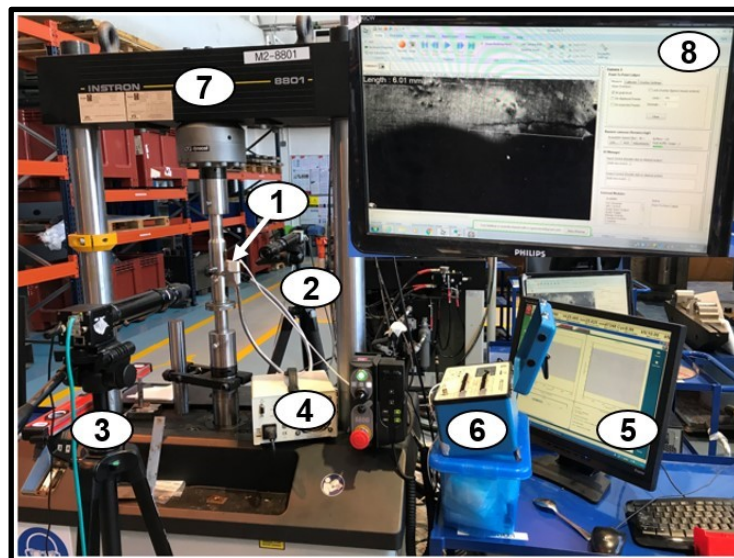


Figure 3-11 Fatigue crack growth characterisation

(1) C (T) sample, (2 & 3) Baumer camera, (4) White light for visibility, (5) Wave matrix monitor, (6) Strain gauge, (7) 100kN machine, (8) Crack monitoring screen

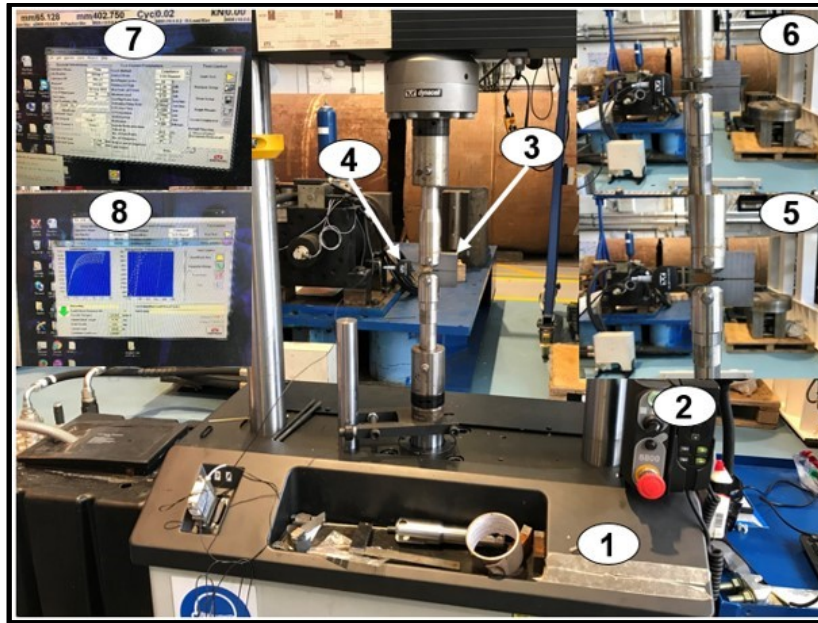


Figure 3-12 Characterisation of fracture toughness by J1C method

(1) 100kN machine, (2) Controller (3) C (T) specimen, (4) CTOD gauge, (5) Z-specimen, (6) X-specimen, (7) Input data, (8) Output J-R curve

3.5.5 Fracture toughness (FT)

Pre-cracking was carried out using an Instron 8801 servo-hydraulic test rig with a 50kN load cell, as shown in Figure 3-11. A load ratio of ($R=0.1$), loading rate of 1mm/min, and a test frequency of 5Hz were utilised. Fracture toughness, by J1c, was also conducted as per ASTM E1820 standard, as shown in Figure 3-12 using an Instron 8801 servo-hydraulic test rig equipped with a 100kN load cell. A stress ratio ($R=0.1$) and frequency of 5Hz was used. A crack compliance method and LLD channel control method was also utilised. All tests were done at room temperature and with a C (T) specimen. Detailed testing procedures are covered in Chapter 7. American standards were used for this testing because the sponsors insisted on their standards and codes to enable them to have a strong base for comparison with their previous work on other materials and manufacturing process.

4 TAILORED DEPOSITION OF WIRE GRADES

4.1 Introduction

The direction of weld deposition in relation to the loading direction is a criteria that determines the changes in the mechanical properties of the as-deposited structure [102]. In the WAAM process, the loading stresses acting on a component must be considered when building parts that will serve the purpose of their design life. Since the heat distribution and direction of grain growth plays a huge role in the anisotropy behaviour of the deposited structure. This is more obvious when the structure will be used in the as-deposited condition. This is even more critical for the building of mixed graded structures that requires different loading condition.

In this study, the first objective was to evaluate the mechanical responses of single pass WAAM low carbon and HSLA structural components in as deposited and machined condition. The second objective was to evaluate the microstructural variation in the as-deposited condition. The third and fourth objectives were to evaluate the mechanical behaviour of the graded structures in different loading directions and their microstructural variation across the graded structures.

4.2 Experimental procedure

The WAAM of the single and mixed grade structures was carried out with a standard CMT power source. The set-up shown in Figure 3-1 was used in the experimentation. Local shielding was achieved with spec shield light (Ar + 20%CO₂) and heavy (Ar + 2%O₂) with a flow rate of 15L/min. The base plates used here are as per Section 3.3.1 with dimensions of 450mm x 200mm x 15mm. The plates were secured by steel jigs tightened to an aluminium base plate. The filler wire used as feedstock in these experiments is as shown in Table 3-1 and Table 3-2. The wires were deposited layer by layer onto the substrate with a single pass strategy, as shown in Figure 4-3 and Figure 4-4, while keeping all welding parameters constant, except the first three layers. For each successive

layer, the deposition direction was reversed with a constant upward increment of 1.7mm of the torch distance for a new layer deposition to avoid stubbing of the wire in the as-deposited weld.

All deposition was done in the PA (1G/Down hand) position from the base of the plate with the torch and wire feeding in the vertical direction utilizing the defined process parameters as shown in Table 4-1. The walls were built with their first layer deposited at a wire feed speed (WFS) of 5m/min, the second to third layers were deposited at WFS of 4.5m/min and the rest of the layers deposited at a constant WFS of 4m/min, until a height of 200mm was attained. The welding parameters of the first two layers were selected to account for the high thermal heat sink encountered with the base metal before the actual welding parameters. The layer height increment used was established by obtaining the average of the initial layer height taken from the next four layers of depositions. The deposition parameters were monitored using a portable Fronius arc monitor, as shown in Figure 4-2. A time interval of 60 seconds was applied after the first layer to the subsequent layers. Six structures comprising two for each alloy grade ER70S-6, ER90S-B3, and ER120S-G were deposited with WAAM. Each structure has a dimension of 400mm length x 200mm height as shown in Figure 4-3 and Figure 4-4. The PAW structures were built in an enclosed charged chamber.

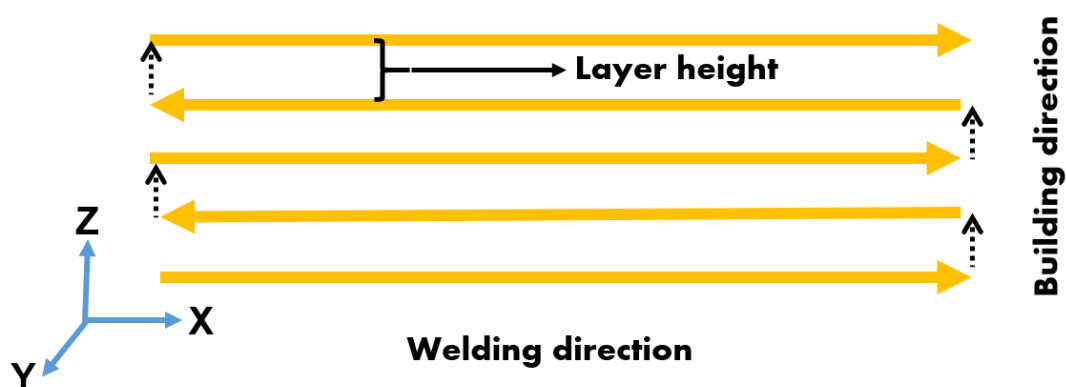


Figure 4-1 Single-pass strategy

Tailored Deposition of Wire Grades

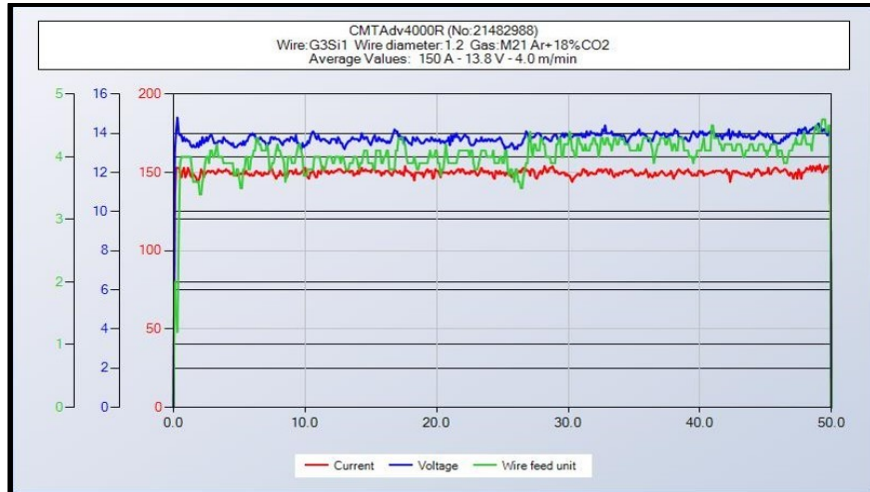


Figure 4-2 Average plot of WFS, current and voltage

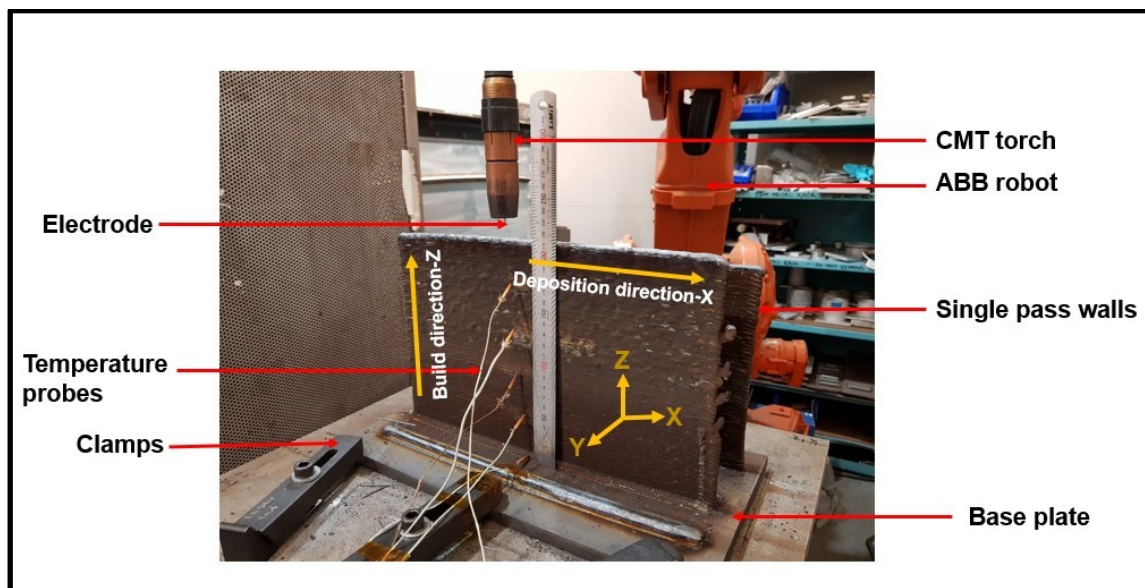


Figure 4-3 Single-pass single grade structure

More WAAM built single pass walls are shown in appendices (fig.7)

Tailored Deposition of Wire Grades

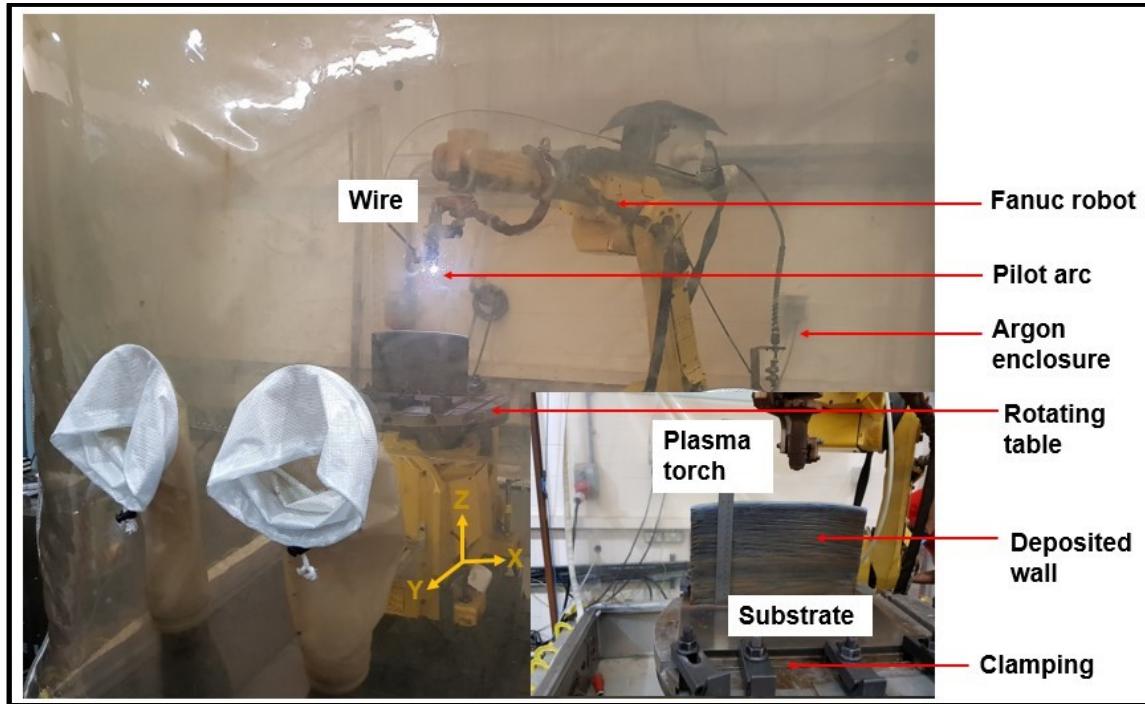


Figure 4-4 Set-up for WAAM deposition with a plasma power source

The mixed grades were also deposited using the single pass deposition strategy with the same parameters and process. Deposition of one grade was followed by the next grade as shown in Figure 4-5. Two structures were deposited each for the mixed grades.

Table 4-1 WAAM deposition parameters

Matrix	current (amps)	Voltage (Volts)	HI (kJ/mm)	WFS (m/min)	TS (mm/sec)	GPF (secs)	WFS/ TS	CTWD	LH (mm)
Single grade	150-165	13.9-16.4	0.313-0.382	4.0-6.5	6.67-10	1.4	10	11.5-13.0	1.7
Mixed grade	141.9 -159	12.38 -14.8	0.26 - 0.322	4.0-6.5	6.67-10	1.4	10	11.5-13.0	1.7
Plasma	200	23.5		1.5	3.8		6.5		0.9

Tailored Deposition of Wire Grades

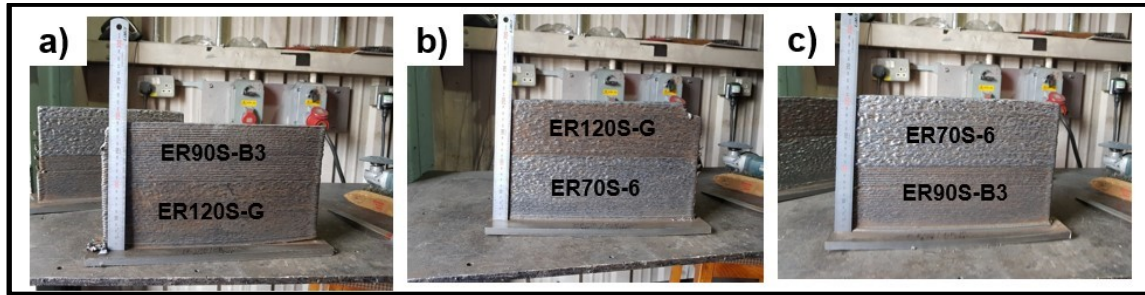


Figure 4-5 Single-pass mixed grades structures

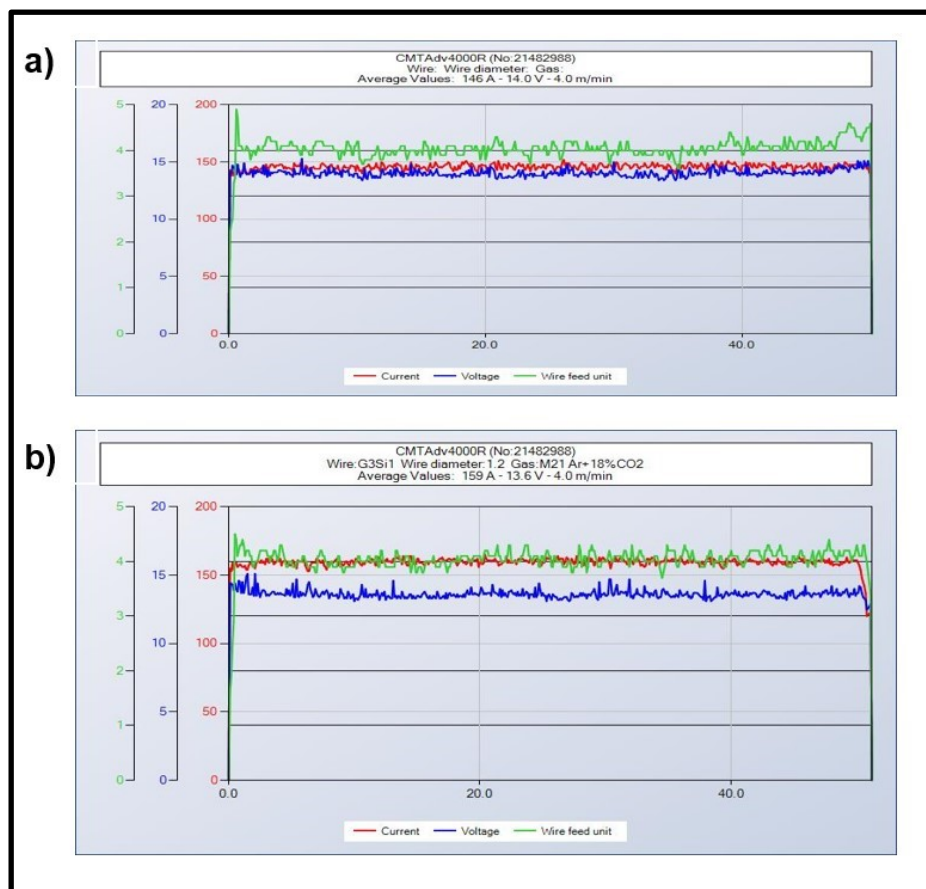


Figure 4-6 Deposition parameters

A tensile mechanical test was carried out on the extracted samples from the build in the single and mixed grade, as shown in Figure 4-7. A set of six tensile specimens each were cut from each orientation in the single grade. A set of six tensile specimens was cut from the mixed grade in the welding direction (X) and

built direction (Z) with each grade equally mating. A gauge length of 50 and 25mm was used for the studies.

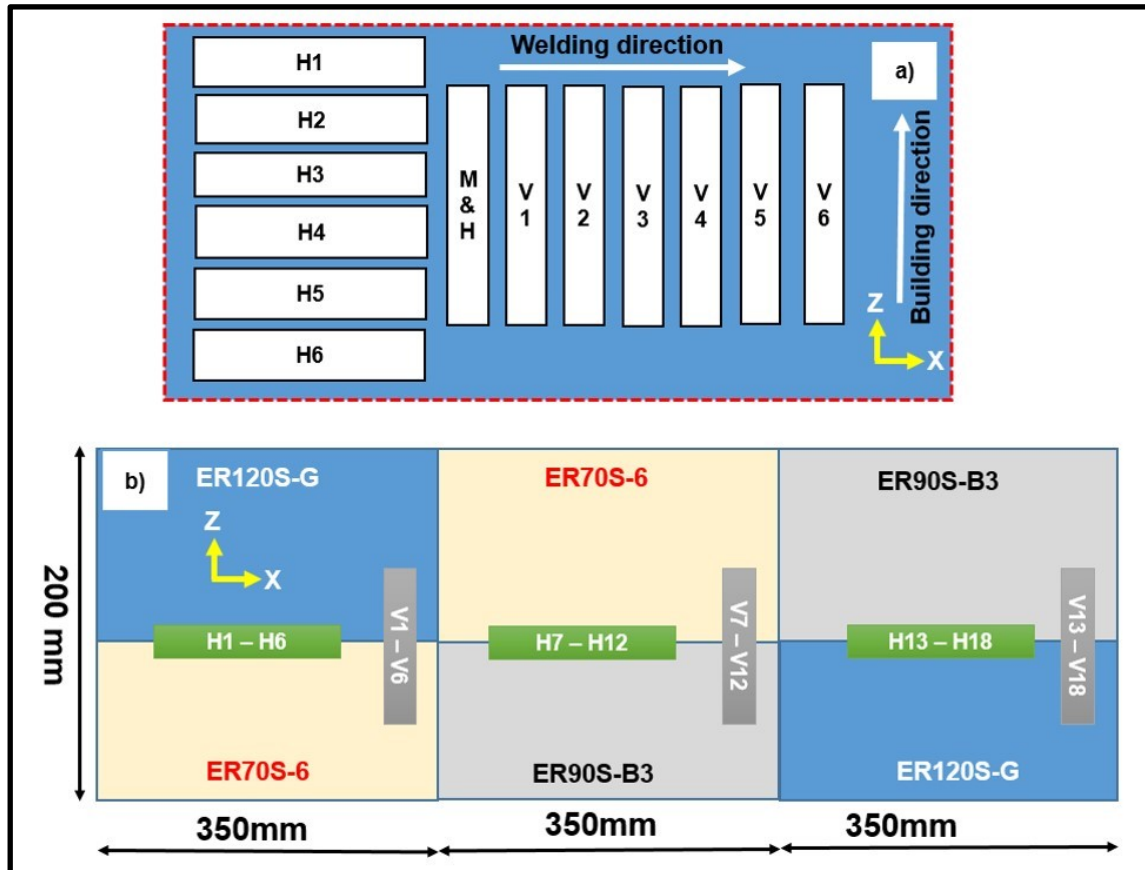


Figure 4-7 Machining and testing matrix

(a) Single grade (b) Mixed grades

All tensile and Charpy impact toughness specimens were machined and prepared in accordance with Figure 4-8 and tested as per Figure 3-8. DIC data of the mixed grades interactions were obtained as per Section 3.5.2, as shown in Figure 3-9. The metallography and Vickers hardness testing using a 300kg load at a spacing of 0.5mm across the single and mixed grades were studied as per Sections 3.4.2, and 3.4.3 as shown in Figure 3-5 and Figure 3-6. The phase content of the mixed grade interface, single grade, grain sizes, texture, and elemental composition was studied with EBSD and EDS count, as outlined in section 3.4.5 and shown in Figure 3-6.

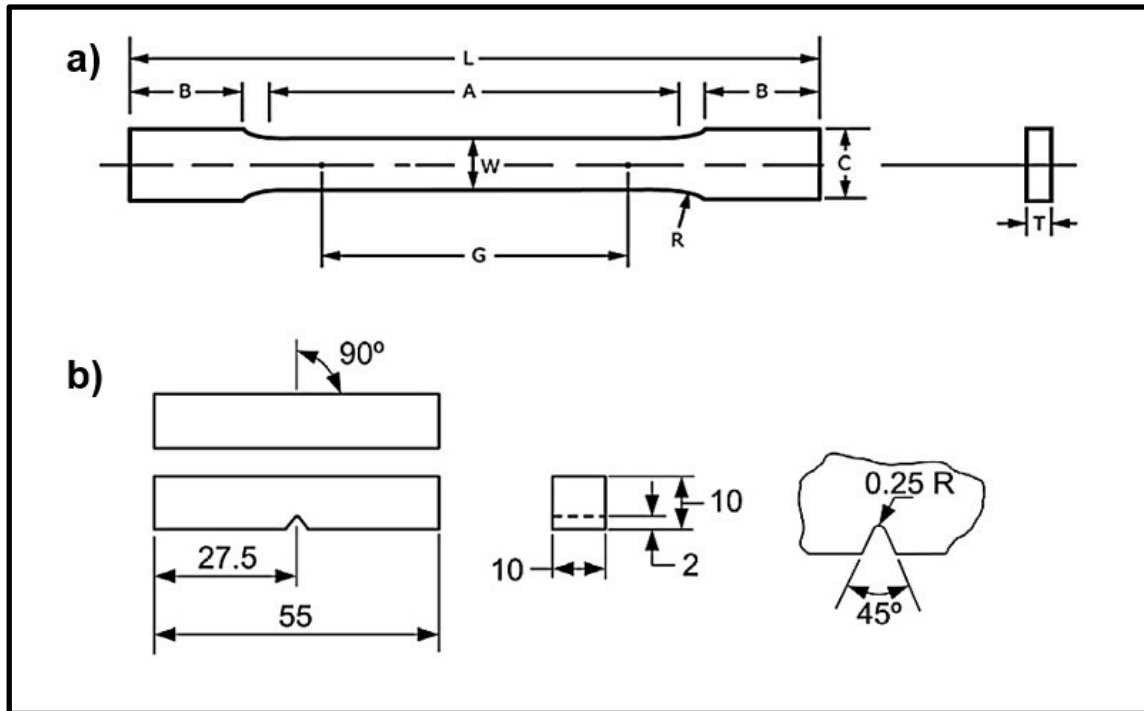


Figure 4-8 Tensile and Impact toughness schematic

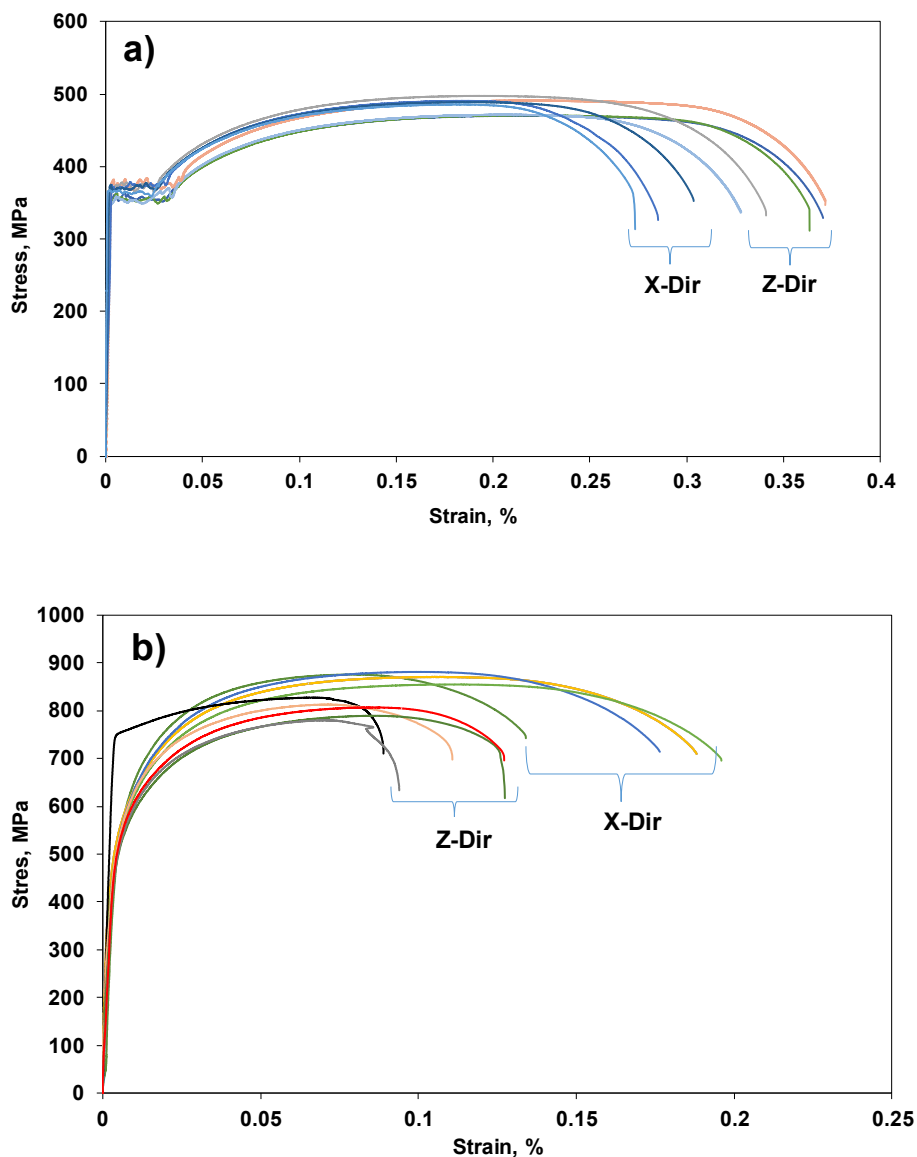
Tensile specimen dimensions were $A = 57\text{mm}$, $B = 35\text{mm}$, $C = 20\text{mm}$, $L = 130\text{mm}$, $G = 50 \pm 0.1 / 25 \pm 0.1\text{mm}$, $W = 12.5 \pm 0.2\text{mm}$, $T = 3.5$, $R = 12.5\text{mm}$

4.3 Results

4.3.1 Mechanical behaviour across the single-pass single grades

Figure 4-9 shows the variation in stress-strain curves obtained for the machined and as-deposited grades of the WAAM structures using the SP deposition strategy. The fractured tensile samples are shown in Figure 4-10. The SP strategy, which was deposited with layer by layer single bead molten steel, has one unique TS and cools very fast as a result of more heat conduction routes along the build, hence, its limited deposited thickness ($<10\text{mm}$). The machined condition (Figure 4-10(a-c)) performed better than the as-deposited condition in all the grades due to the effects of SW obtained from the layer by layer deposition. The same plot also shows the anisotropy in strength. It is also clear from the studies that the ASD condition fractured in the X-direction (Figure 4-10(d-f)) except ER70S-6 ASD condition performed better than the ASD

condition fractured in the Z-direction (Figure 4-10(d-f)). The fracture forces are acting in the perpendicular direction to the SW in any case, and the ASD loaded in the Z-direction will be more affected because the peaks and valleys of the SW are in line with the acting forces, unlike the ASD loaded in the X-direction, where the peaks and valleys of the SW are in the opposite direction to the acting forces, hence the display of higher strength and ductility before fracture. The SW causes stress concentration at the valley, and the severity of the stress concentration is dependent on the SW geometry observed. This severity is a function determined by the WAAM deposition parameters applied. Three distinctive stress-strain curves were exhibited by these WAAM structures.



Tailored Deposition of Wire Grades

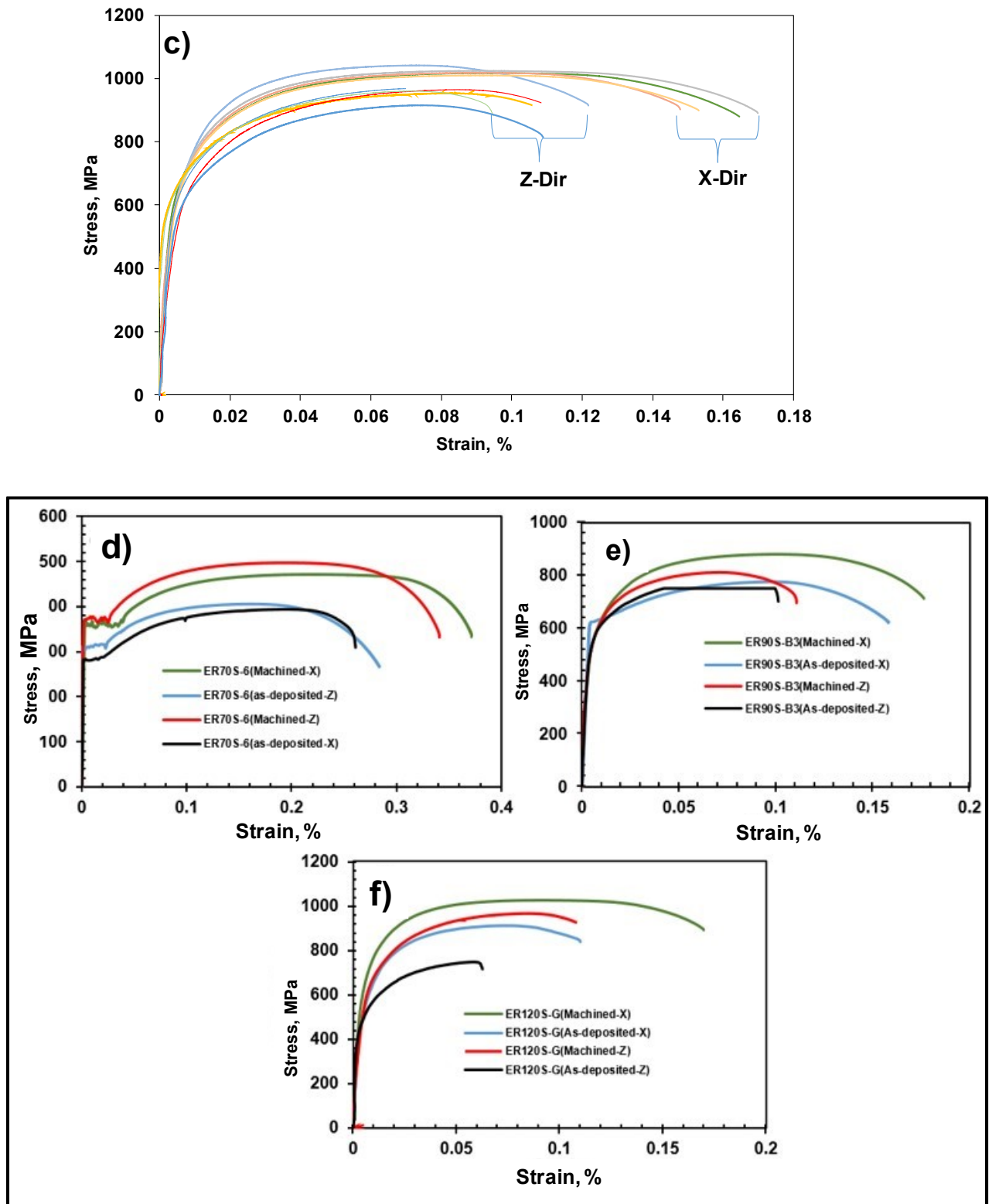


Figure 4-9 Stress-Strain plots for single pass single grades

Machined, X Vs Z (a) ER70S-6 (b) ER90S-B3 (c) ER120S-G, Machined Vs as-deposited (d) ER70S-6, (e) ER90S-B3, (f) ER120S-G

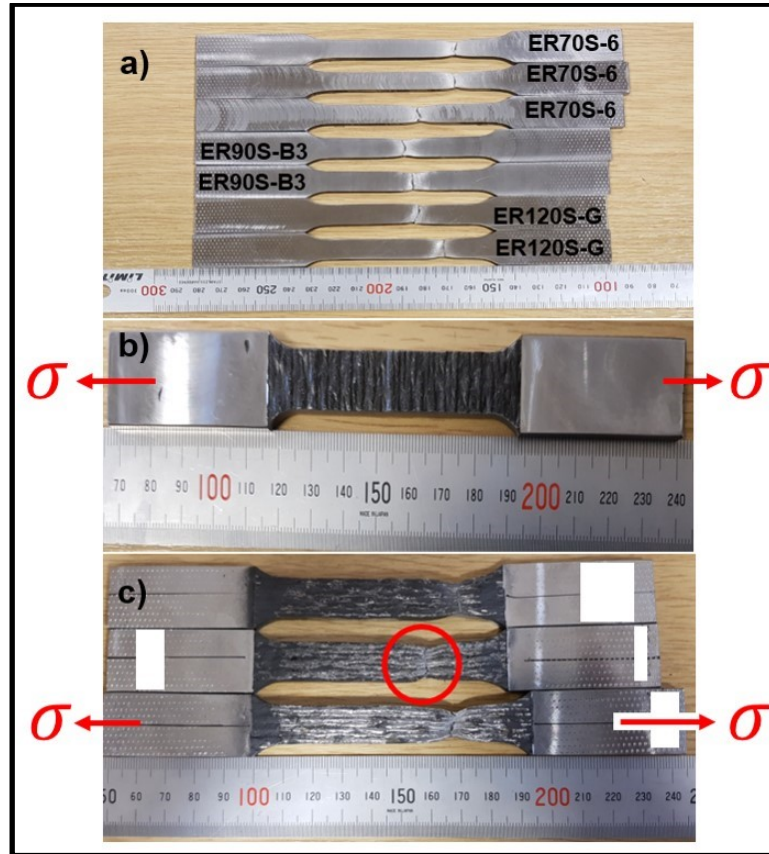


Figure 4-10 Tensile coupons

(a) Machined condition (b) As-deposited loaded in Z-direction (c) As-deposited loaded in X-direction, more tensile fractured coupons are shown in appendices (fig.1)

4.3.2 Mechanical behaviour across the single-pass mixed grades

Figure 4-11 and Figure 4-12 present the stress-strain curve for the mixed grades fractured in the Z-direction and the tensile samples of the mixed grades fractured in the Z-direction respectively. All the mixed grades (ER120S-G+ER90S-B3, ER120S-G + ER70S-6 and ER90S-B3+ER70S-6) fractured in the Z-direction failed outside the mixed grade interface towards the lower grade as shown in Figure 4-12(a-c), hence the interface was stronger than the strength of the lower grade (ER70S-6 or ER90S-B3) respectively. As seen from Figure 4-11, the mixed grades with ER70S-6 exhibited a non-continuous yielding phenomenon, which is a characteristic attribute of mild steel. Here the Lüders band range is shorter compared to that of the single grade of ER70S-6 tested in the same direction

because in the mixed grades of ER120S-G+ER70S-6 & ER90S-B3+ER70S-6, there exist more than two phases at the interface and the volume fraction of the ferrite phase is low compared to cementite and martensite (the stronger phase). Hence the reduced time taken for yielding to be completed and a jump to plastic deformation is quick, unlike the single grade of the ER70S-6 that has two distinct phases which exhibit a full Lüders strain because the volume fraction of the ferrite phase was more than the cementite; and possibly no or little martensitic phase, hence the delay in reaching the plastic deformation stage due to more time needed to exit from the soft ferrite phases. This seems to corroborate the findings of Mohsenzadeh and Mazinani [103]. Mixed grades fractured in the Z-direction (Figure 4-12(d)) exhibited higher UTS and YS as seen in (ER120S-G + ER70S-6 and ER90S-B3 + ER 70S-G, but with lower PCE compared to the corresponding lower grade in the mixture for the same reasons discussed above.

Figure 4-13 and Figure 4-14 present the stress-strain curve for the mixed grades fractured in the X-direction and the tensile samples of mixed grades fractured in the X-direction, respectively. All the mixed grades here (ER120S-G+ER90S-B3, ER120S-G + ER70S-6 and ER90S-B3+ER70S-6) fractured in the X-direction commenced crack initiation from the area of least weakness in the lower strength grade, as shown in Figure 4-14 (C1 & C2)

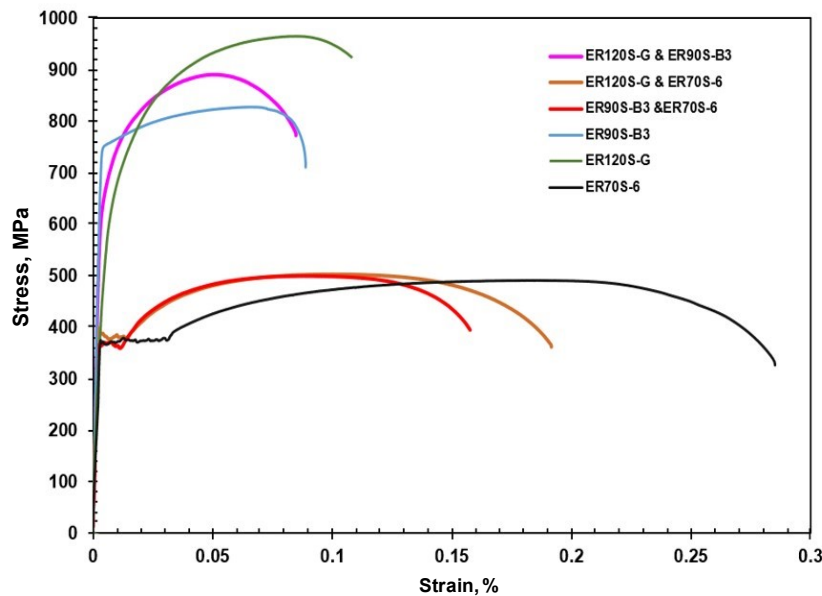


Figure 4-11 Stress-strain curves for mixed grades loaded vertically (Z)

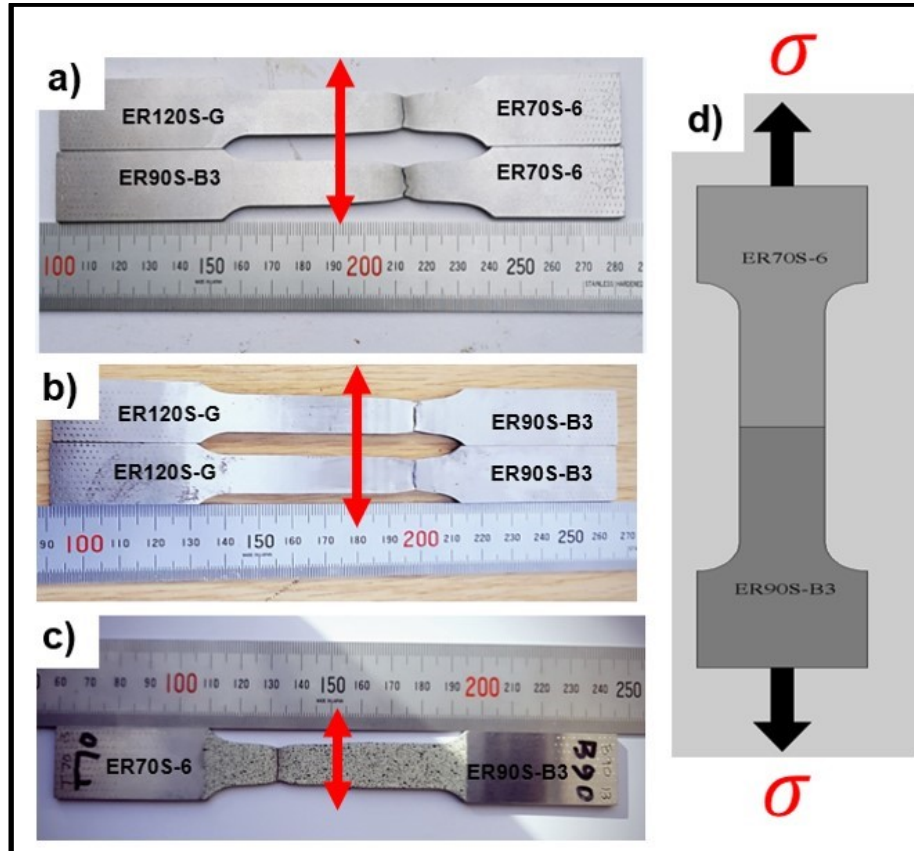


Figure 4-12 Tensile samples of mixed grades fractured in Z-direction

**(a) ER120S-G + ER70S-6, (b) ER120S-G + ER90S-B3, (c) ER70S-6 + ER90S-B3,
(d) Z- fracture orientation**

In this fracture orientation, the higher strength grade in the matrix governs the fracture properties in the X-direction as can be seen in the curves. The lower strength grade (ER70S-6) of the multi grade reached its plastic limits first and is pinned up in the matrix by the higher grade of the matrix (ER120S-G or ER90S-B3) which is yet to pass its elastic region. Hence more stress will be concentrated in the lower grade (ER70S-6) where fracture initiation will begin with microvoid coalescence. The collapse voids in the lower strength grade serve as the stress concentration points that initiate failure in the higher grade as well. This physical manifestation is further explained by digital image correlation images in subsequent pages.

Tailored Deposition of Wire Grades

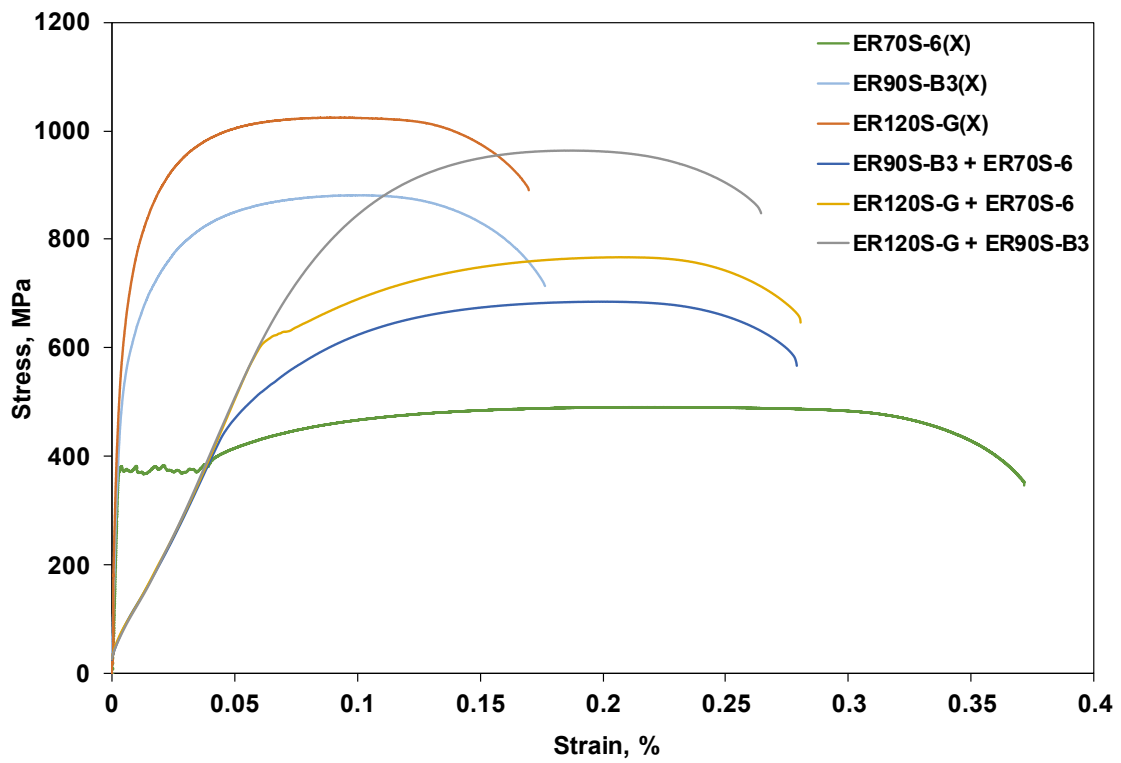


Figure 4-13 Stress-strain curves for mixed grades loaded horizontally (X)

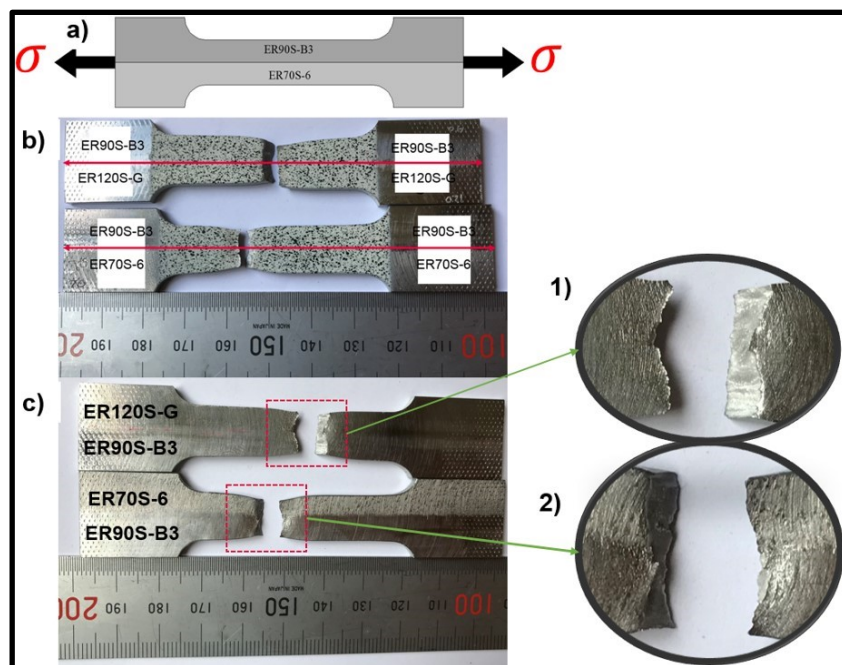


Figure 4-14 Loading mixed grades in X-direction

Mixed grades fractured in the X-direction (Figure 4-13) exhibited higher UTS, YS and PCE than the mixed grades fractured in the Z-direction, and their corresponding lower grade in the matrix. The Lüders band phenomenon disappears in the X-matrix loading as a result of the effect of multiple microstructural phases that displace the effect of soft ferrite.

4.3.3 Mechanical and microstructural variation in tailored grades

Figure 4-15 presents the mechanical anisotropy in single and mixed grades. And Figure 4-16(a & b) present the average mechanical properties of single, mixed grades fractured in both Z and X directions and manufacturers specification. ER120S-G exhibited higher UTS, YS, and PCE in the X-direction compared to the Z-direction with an average value of 1022MPa, 839MPa, and 21%. For ER70S-6, UTS and YS are higher in the Z-direction (491MPa and 367MPa) compared to the X-direction (471MPa and 351MPa). PCE is higher in the X-direction (38%) than in the Z-direction (33%). This work corroborates the findings of Lu et al. [104]. The minimum UTS and YS obtained in both directions for the ER70S-6 are higher than the values obtained from wrought ASTM 36 with UTS and YS of 434MPa and 261MPa [105] respectively. This anisotropic behaviour of WAAM deposit is due to the direction of heat flow, grain growth, and layer stacking during the deposition process

The wires used for this experiment, as shown in Table 3-1, have low carbon content for improved weldability. ER70S-6, which has <1% Ni, has lower UTS with higher elongation compared to ER120S-G & ER90S-B3 wires which exhibit lower ductility. Higher UTS was attained with ER120S-G & ER90S-B3 having more elements of Cr, Mo, and Ni; alloy elements also make the diffusion of carbon atoms more difficult and hence the formation of non-equilibrium phases responsible for increase hardenability, giving high strength phases martensite and bainite. The high strength in ER120S-G is achieved by a strengthening mechanism called solid solution strengthening. The Cr, Mo, and Mn distort the lattice, generating stress around it, which impedes the dislocation motion. Also, the higher strength observed with ER120S-G is due to the addition of micro-

Tailored Deposition of Wire Grades

alloying elements of V (<0.01%) and Ni (2.27%), hence classified as high strength low alloy steel (HSLA). This work corroborates the findings of Schnitzer et al. [92]. The carbon content of the HSLA steels is in the range of 0.08 – 0.10%, supporting their high strength level while a carbon content of 0.06% for ER70S-6 supports its low strength and high ductility. This mechanism is known for controlling yield strength. The single grades are comparable to the manufacturer's specification shown in Figure 4-16 (b), hence WAAM for HSLA steel promise to be a good manufacturing option using this steels.

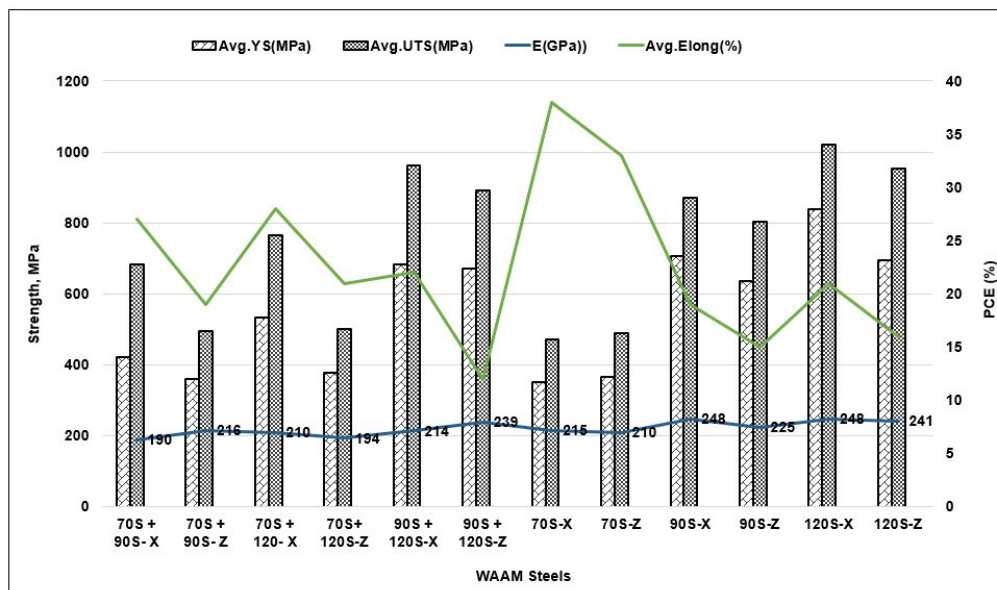
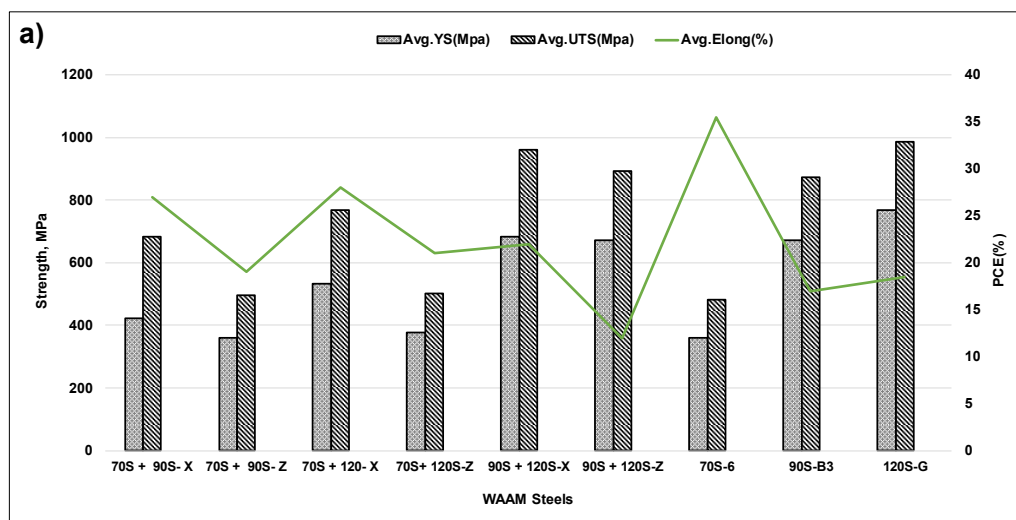


Figure 4-15 Mechanical anisotropy in single and mixed WAAM steels



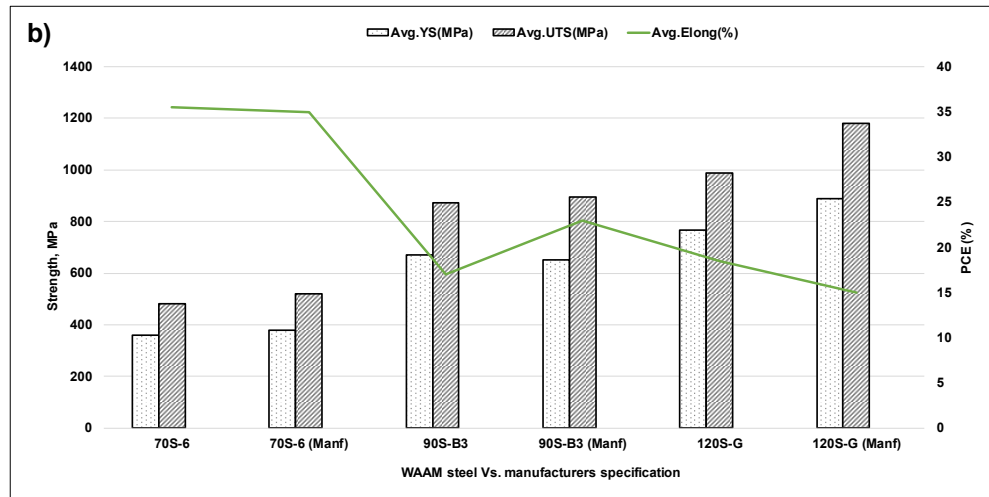


Figure 4-16 Average mechanical properties

(a) single grades vs mixed grades (b) Single grades vs manufacturers specification

Figure 4-17 present the hardness profile along the build section. Figure 4-18 presents the Charpy impact toughness variation in the various grades. Figure 4-19 also present the fracture surface morphology of the impact toughness specimen. Figure 4-20 presents the microstructural variation in the single grade. The ER120S-G microstructure, as revealed in the SP (Figure 4-20(e-f)), presents majorly martensitic lath (ML), quasi polygonal ferrite (QPF) formed as a result of fast cooling and some acicular ferrite (AF). The microstructure of the ER90S-B3 SP, as presented in Figure 4-20(c-d) clearly shows some martensitic bainite (MB), QPF, bainite (B) decorated with finely dispersed carbides of $\text{Fe}_3\text{C} + \text{M}_2\text{C}$ as a result of prior composition. The microstructure presented by the ER70S-6 SP in Figure 4-20(a-b), contains mostly polygonal ferrite (PF) and pearlite (P). The SEM images in Figure 4-20(b, d, & f) show clearly this structure within the grain boundaries.

It was observed that the highest hardness of the single grade deposition occurs at the beginning of the deposition near the substrate because of the high thermal cycle effects, which gradually reduces as the number of layers increases; subsequent layers near the deposition promoted annealing and softening of previous layers. In the same way, the last top layers were incomplete

austenitisation take place. They contain some entrap retained austenite which reveals high hardness as well. Hardness profile in Figure 4-17 was obtained from a cross-section of a built sample with the top layers omitted out of the plot. The average hardness obtained for the ER70S-6, ER90S-B3, and ER120S-G is 165.9HV, 269HV, and 327.5HV, respectively. The layer stacking and thermal cyclic heating, which cause the hardening and softening of adjacent layers, are responsible for the hardness variation along the built. Similar hardness trends were reported in the WAAM structure built with HSLA steel and maraging steel [53,57]. The average toughness of 109J, 93J & 37J was displayed by ER70, ER120 & ER90 respectively at -30°C, which is the minimum design mean temperature of the envisage project application.

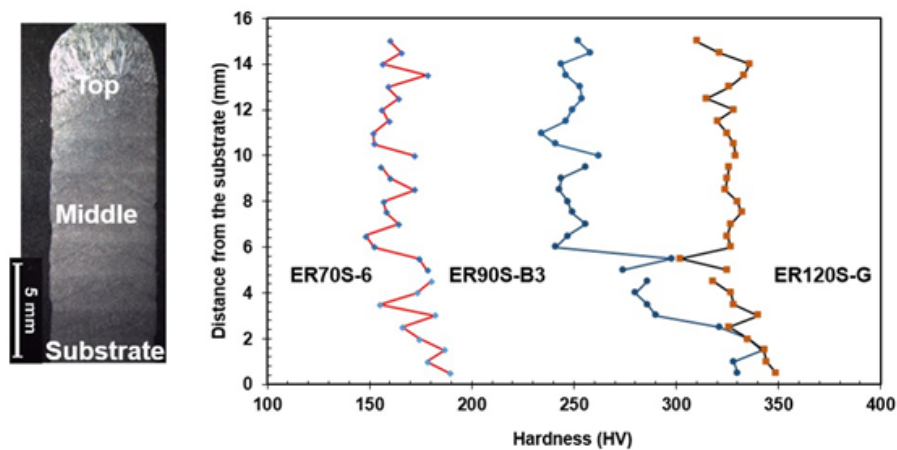


Figure 4-17 Hardness variation in the single pass grade

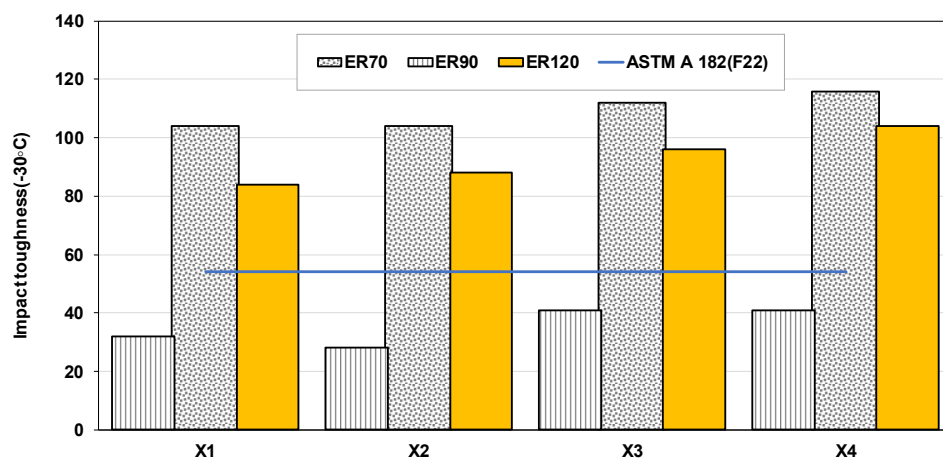


Figure 4-18 Charpy impact toughness test at -30°C

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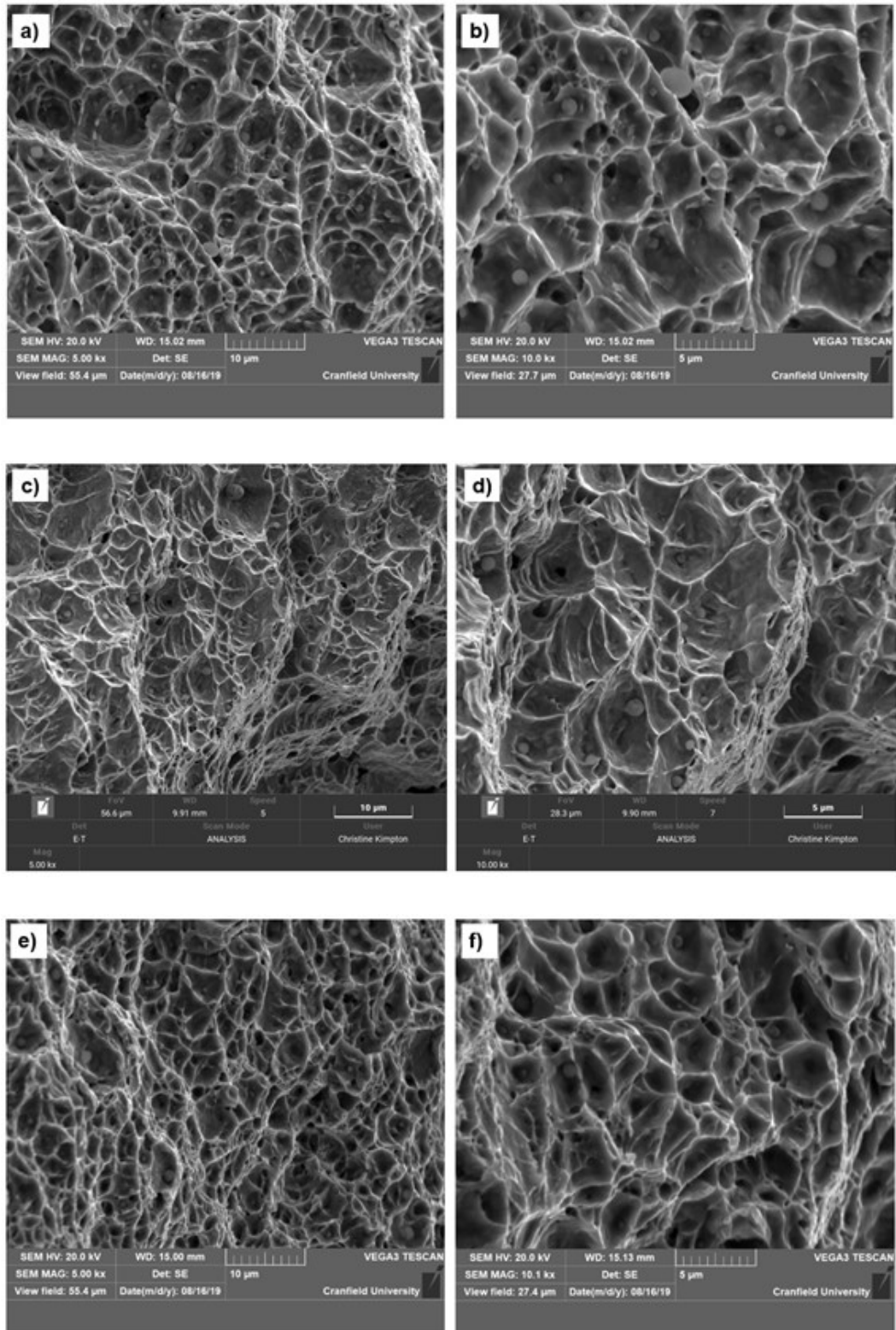
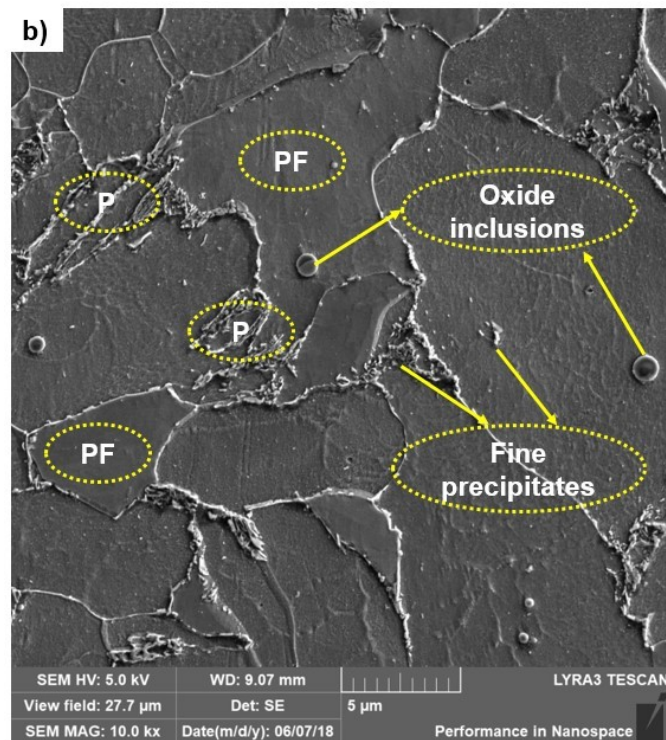
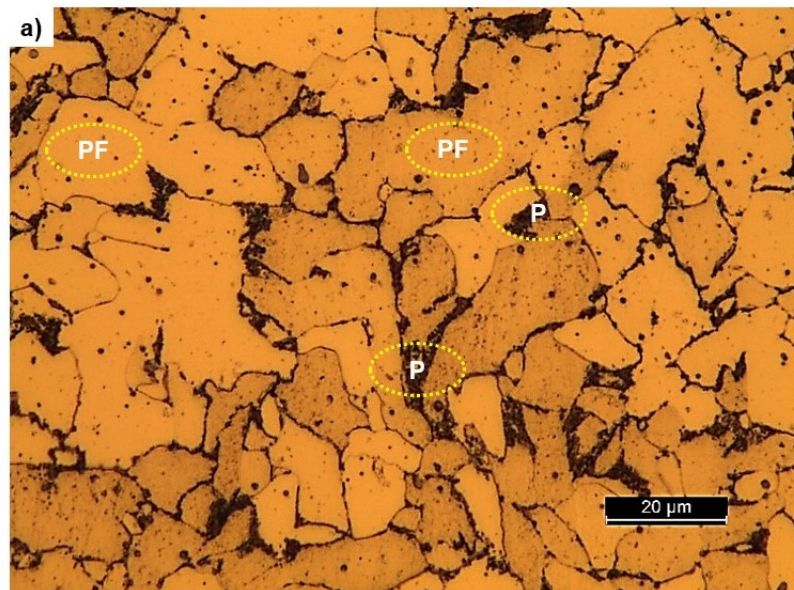
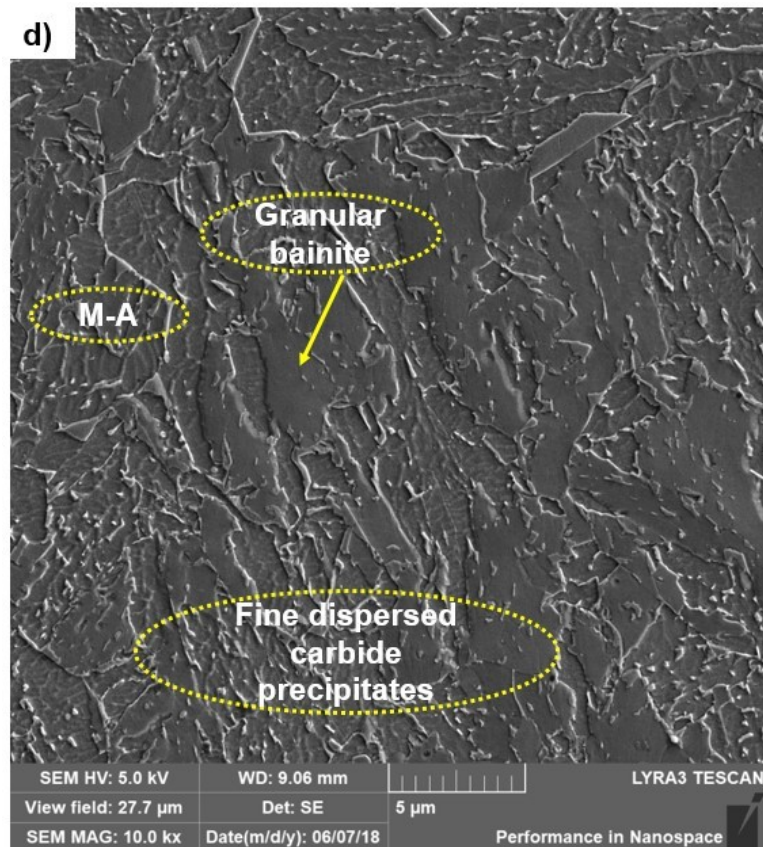
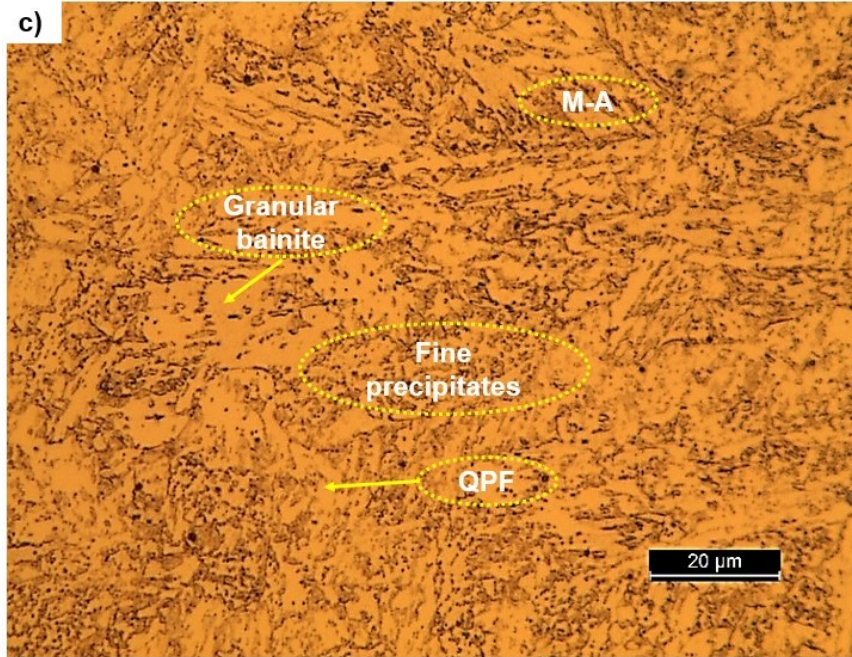


Figure 4-19 Impact toughness fractured surface -30°C

(a-b) ER70S-6 (c-d) ER90S-B3 (e-f) ER120S-G

The SEM images of the fracture morphologies of the impact toughness specimen of ER70S-6, ER90S-B3 and ER120S-G presented in Figure 4-19 reveals the occurrence of ductile fracture behaviour with equiaxed and shear dimples resulting from plastic fracture. However, the ER90S-B3 showed some form of cleavage and ductile region Figure 4-19 (c & d).





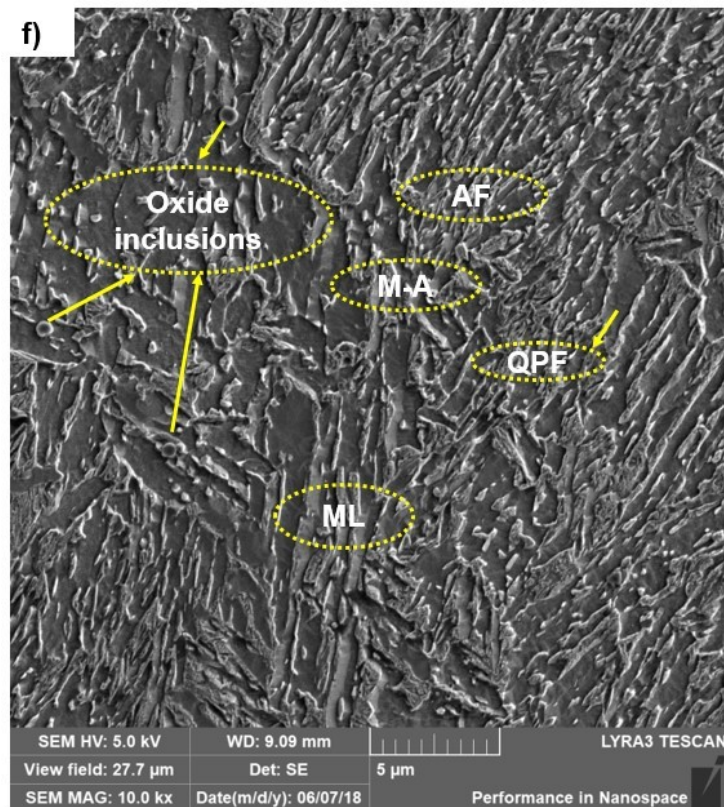
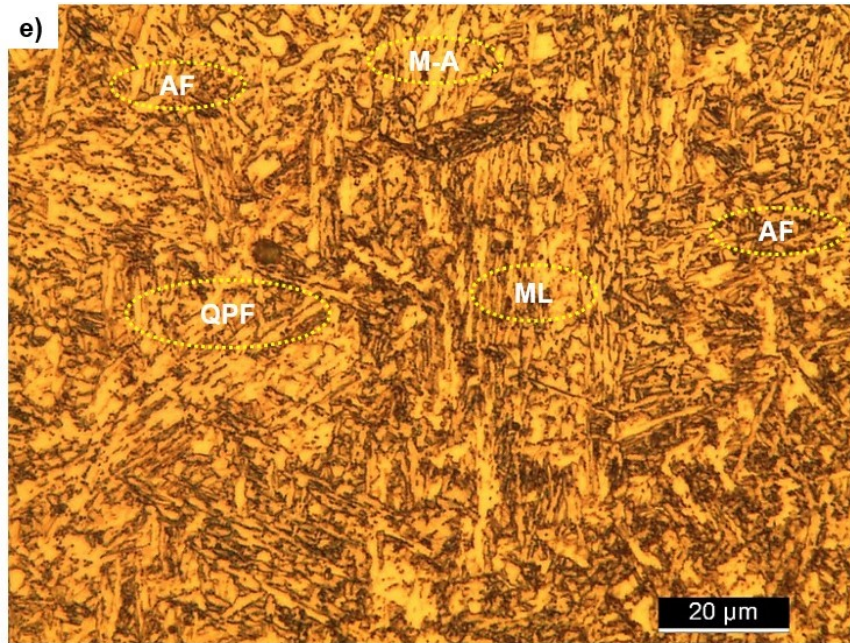
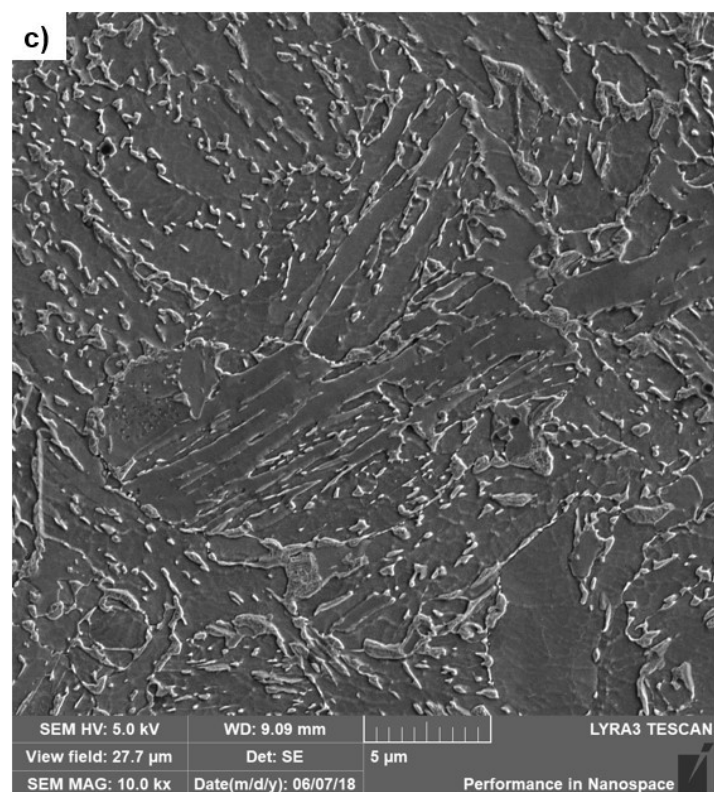
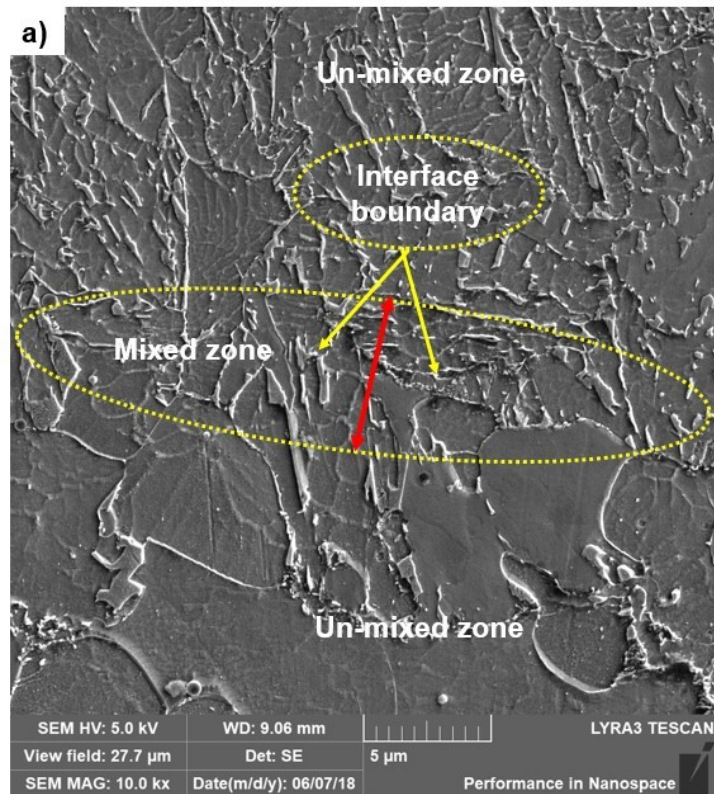


Figure 4-20 Microstructural variation in single pass grades

Microstructures (a) ER70-S6 (b) SEM ER70S-6 (c) ER 90S-B3 (d) SEM ER90S-B3
(e) ER120S-G (f) SEM ER120S-G



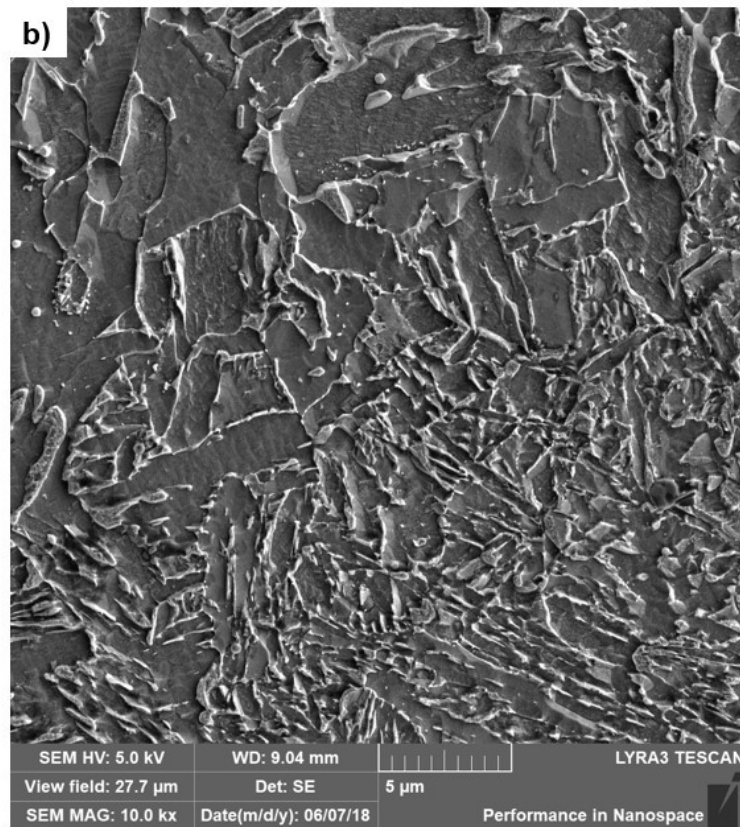


Figure 4-21 SEM analysis of single pass mixed grades

(a) ER90S-B3 + ER70S-6 (b) ER120S-G + ER90S-B3 (c) ER 70S-6+ER120S-G

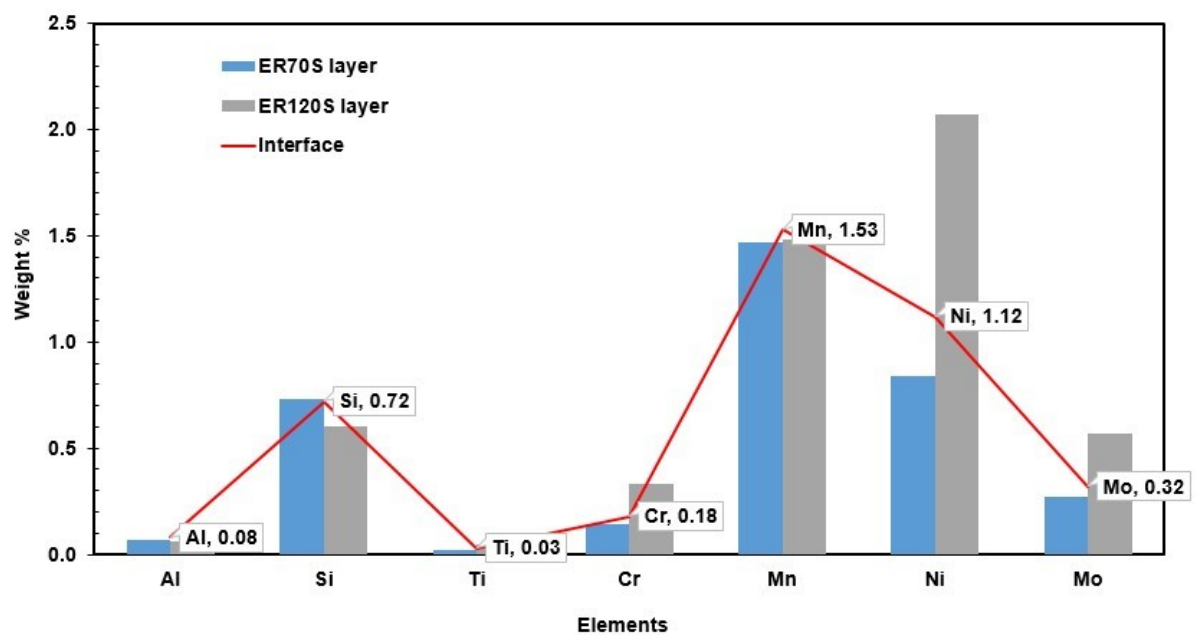


Figure 4-22 Elemental variation across ER70S-6 +ER120S-G

Figure 4-21 presents the microstructural variation in the mixed grades, and Figure 4-22 presents the elemental variation across the mixed grades. Figure 4-23 presents the hardness and EDS line scan across the mixed grades. In the mixed grades of ER120S-G + ER70S-6, the hardness at the interphase is about 260HV due to the slight increase and decrease in Ni & Mn content towards the ER120S-G, as shown in Figure 4-23(c). In the mixed grades of ER120S-G and ER90S-G, there is a slight increase in Ni & Mn and a decrease in Cr towards the ER120S-G, hence the slight hardness slope with the interphase value of 320HV. The ER90S-B3 + ER70S-6 had a variation of Cr and Mn at the interphase hence the sharp interface hardness slope with the value of 240HV. In all the mixed grades cases, the hardness variations were within the acceptable limits for low carbon low alloy and high alloy steels [53] for the various applications. It is important to know that the interface of ER70S-6 + ER120S-G (red line) in Figure 4-22 has better properties higher than the lower grade by the display of its analysed chemistry.

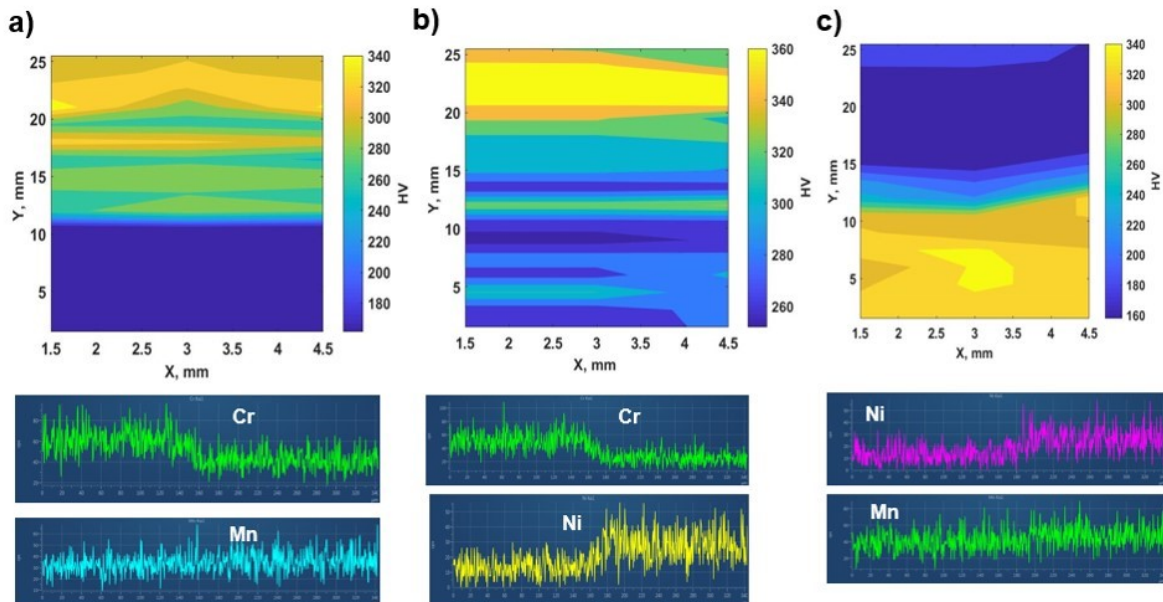


Figure 4-23 Hardness variation in mixed grades and EDS scan

(a) ER90S-B3 + ER 70S-6 (b) ER120S-G + ER90S-B3 (c) ER70S-6 + ER120S-G

4.3.4 The influence of anisotropy in the single and mixed grades

EBSD characterisation methodology was used to investigate the crystallographic orientation in the single-pass deposits to confirm the intrinsic relationship between crystal growth direction and the thermal history experienced by the built structure. The variation in grain sizes in both X and Z direction are presented in Figure 4-24. The EBSD inverse pole figure maps showing the ferritic, bainitic, and martensitic growth orientation as a function of the two machined direction X & Z orientation are depicted in Figure 4-25 respectively. Figure 4-26 presents the corresponding $\{111\}$, $\{101\}$, $\{100\}$ pole figures. The legend showing the relationship between the crystal orientation and colours is inserted in the top upper right corner of Figure 4-25 (a).

The microstructure in the single-pass ER70S-6 in both orientations consisted of ferrite and pearlite, which is consistent with the results shown in Figure 4-20 (a & b); they consist of equiaxed grains with an average grain size of $12.92\mu\text{m}$ in the Z-direction and $14.79\mu\text{m}$ in the X-direction. The microstructure and grain sizes obtained here are similar to that in Colegrove et al. (2013) [106]. The result shows a very different texture type and generally strong texture intensity within both directions, which indicates that an anisotropic ferrite microstructure in the random crystallographic texture was obtained in ER70S-6 deposited using the CMT process, which is due to the re-orientation by recrystallisation and grain growth during deposition. The microstructure presented in ER90S-B3 is mainly elongated bainitic lath and martensitic lath with an average length greater than $40\mu\text{m}$; they consist of an average grain size of $10.42\mu\text{m}$ and $11.32\mu\text{m}$ in the X and Z direction as shown in Figure 4-24.

The microstructure in the ER120S-G is mainly elongated martensitic lath with an average length of $30\mu\text{m}$ randomly distributed between different ferrite grains. The average grain sizes in the X and Z-direction are $4.98\mu\text{m}$ and $6.89\mu\text{m}$ respectively in WAAM CMT; the heat conduction is normally parallel to the build direction (Z-direction), hence the directional growth of the lath martensite in the Z-direction. This is in support of the work carried out by Ge et al. [107]. This result shows a

very similar texture type for both ER120S-G and ER90S-B3 and a generally weak texture intensity within both directions, which indicates that isotropic lath martensitic microstructure in random crystallographic texture was obtained in both ER120S-G and ER90S-B3 deposited using the CMT process. Hence no significant preferential orientation was formed as a result of the WAAM process. This work supports the claim by [53] and disproves the claim by [108], which claimed that the {100} orientation is the preferential grain growth direction during the solidification of carbon steel.

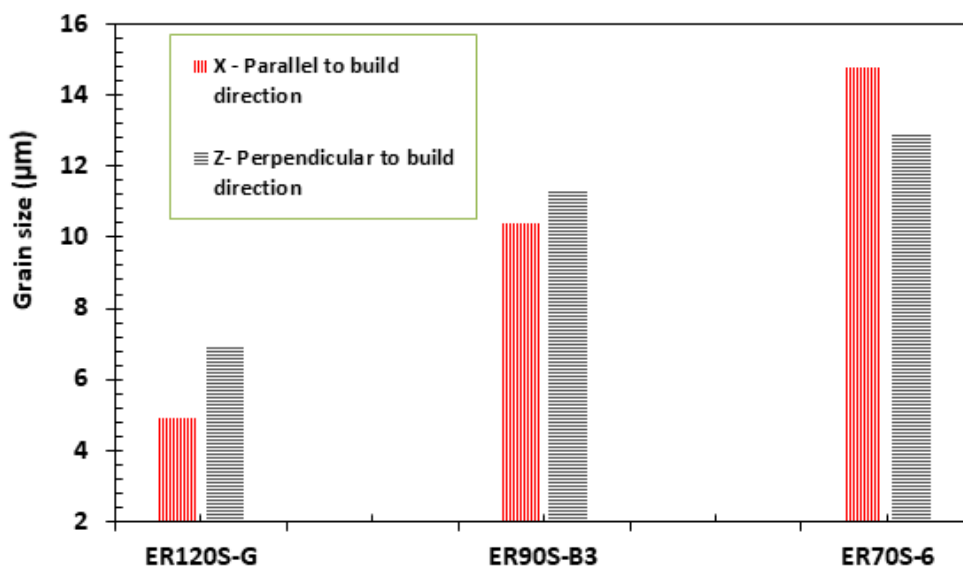


Figure 4-24 Variations in grain sizes in single pass deposition

The EBSD inverse pole figure maps showing the ferritic, bainitic and martensitic growth orientation in the mixed grade of ER120S-G + ER70S-6 are depicted in Figure 4-27. The microstructure in the mixed grade consisted of ferrite and lath martensite of non-equiaxed grains with an average grain size of 8.5 -12μm. The microstructure and grain sizes obtained here are similar to those obtained with the single grades. The microstructure in the S355 wrought comprised of ferrite and pearlite with an average grain size of 15.29μm. The WAAM steel generally presents fine microstructures compared to their wrought counterpart.

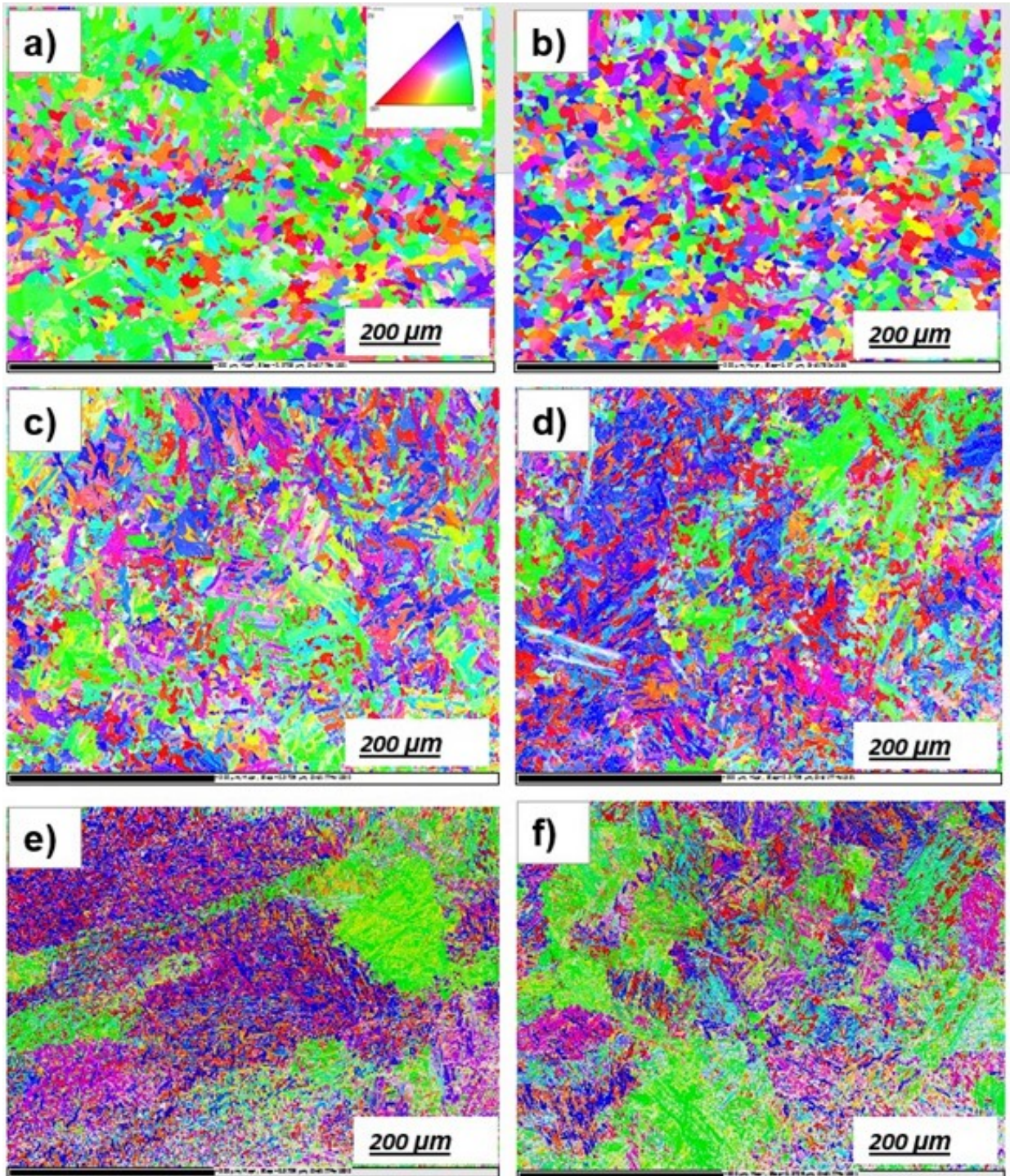


Figure 4-25 EBSD maps showing grain orientation in single pass deposition

(a) ER70S-6 (X-direction) (b) ER70S-6 (Z-direction) (c) ER90S-B3 (X-direction) (d) ER90S-B3 (Z-direction) (e) ER120S-G (X-direction) (f) ER120S-G (Z-direction)

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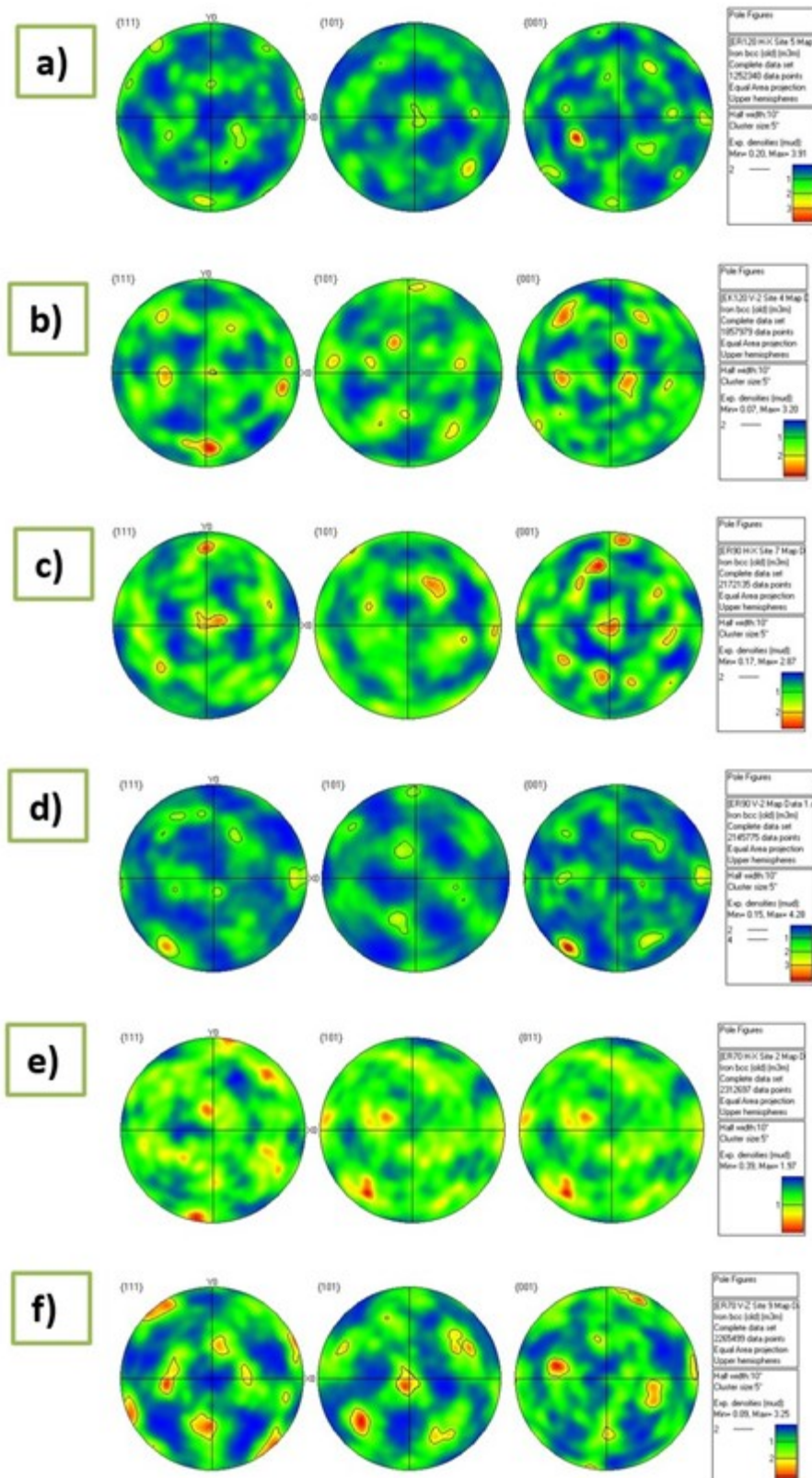


Figure 4-26 Pole figures of α phase in single pass grades

(a) ER120 (WD) (b) ER120 (BD) (c) ER90S-B3 (WD) (d) ER90S-B3 (BD) (e) ER70S-6(WD) (f) ER70S-6 (BD)

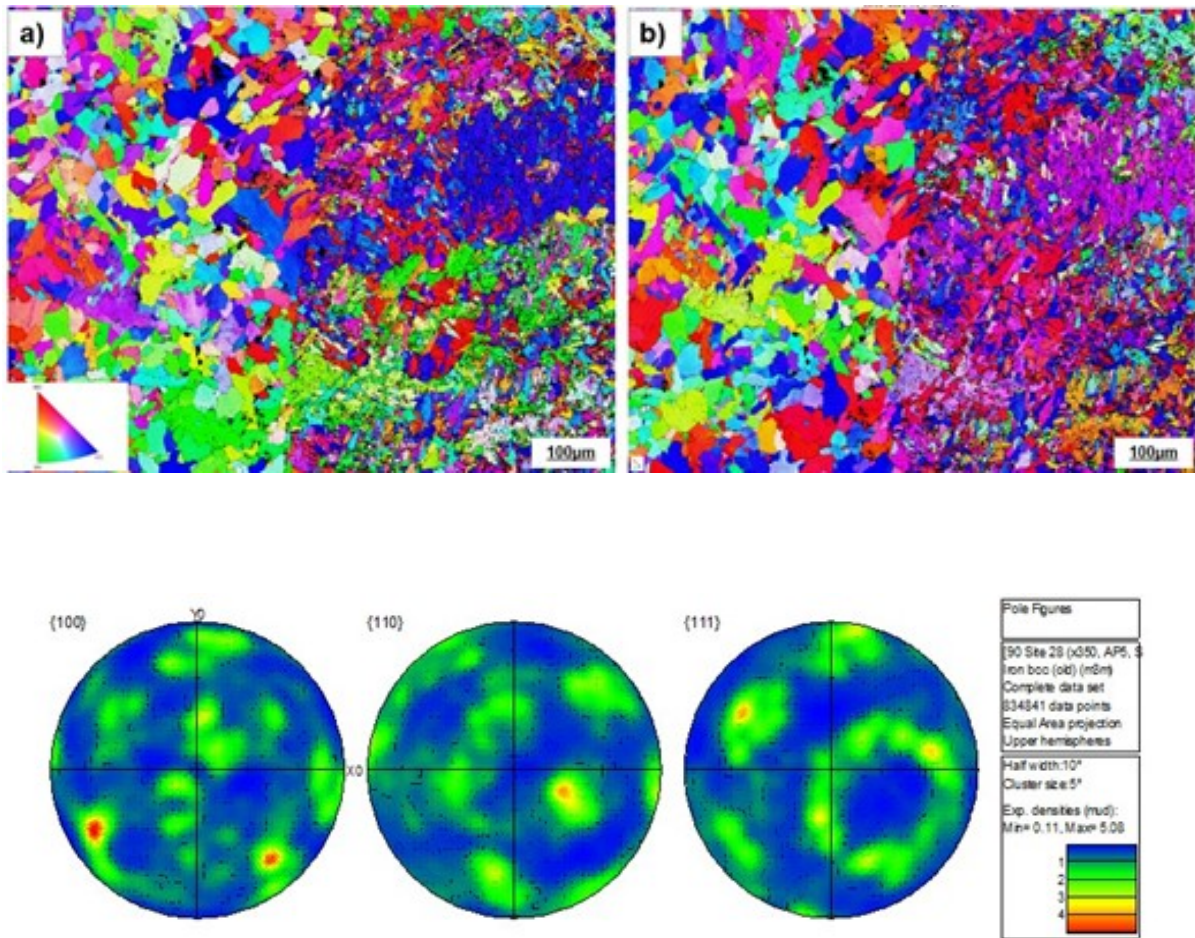


Figure 4-27 Mixed grade EBSD maps showing grain (ER120S-G + ER70S-6) & pole figure

4.4 Discussion of results

4.4.1 The influence of microstructural variation on the tailored grades

Figure 4-28 and Figure 4-29 present the bright TEM micrographs of WAAM single pass ER70S-6 and ER120S-G. The stress strain curves displayed by the ER70S-6 are typical of mild steel with the upper and lower yield points. This phenomenon is related to their microstructural influence on the properties. The single grades contain more colonies of ferrite forest, PF colonies, AF colonies, and cementites, as shown in Figure 4-28, hence the behaviour. ER90S-B3 displayed a stress-

strain curve with sharp yield points, unlike the ER120S-G that is not defined due to the transition from elastic to plastic. The single grade of ER120S-G contains AF and M-A colonies (hard phases), with some lath martensites colonies, as shown in Figure 4-29, hence the behaviour. Hence the 0.2% strain has to be applied.

The interaction between the interstitial C and N solute atoms with dislocation leads to the formation and propagation of Lüders bands in the HSLA steels [109]. The Lüders band range is shorter in the mixed grade with ER70S-6 fractured in the Z-direction (Figure 4-11), compared to that of the single grade of ER70S-6 fractured in any direction (Figure 4-11). The reduction of the yield point phenomenon in the Z-direction was due to the third phase (martensite) which serves as a mobile dislocation into the ferrite matrix of the lower grade as a result of volume expansion (dilution) accompanying austenite to martensite transformation; this speeds up the dislocation movement and a jump to plastic deformation. In contrast, the single grade of the ER70S-6 has two common phases that exhibit full Lüders band strains because the volume fraction of ferrite phases is greater ($f \geq f_{min}$) than the cementite phase, as evident in Figure 4-20(a & b). The minimum volume required by the second phase to cause total yielding of the ferrite all over in the microstructure is given in Eq (4-1). [110][103]

$$f_{min} = \left[\frac{a^3}{R^3} \right] \times 100 = \left[\frac{(1 - \mu)Y_0}{E\varepsilon^*} \right] \times 100 \quad (4-1)$$

where f_{min} is the volume of other phase, f is the volume of ferrite phase, a is the radius of the martensitic or cementite particle island, R is the radius of yield zone, ν is Poisson's ratio, Y_0 is the yield strength of ferrite at the temperature at which the internal stress induced yielding occurs, E is Young's modulus of ferrite and martensite, ε^* is a stress-free strain. The magnitude of f_{min} for this specimen is less than any other phase in it. The yield stress of the ferrite in the elastic deformation zones using the von Mises yield criterion is expressed by Eq (4-2)

$$\sigma_A = Y \left\{ \frac{3 \sin^2 \theta - 1}{2} \frac{Y_0}{Y} \cdot \frac{R^3}{r^3} + \sqrt{\left(\frac{3 \sin^2 \theta - 1}{2} \frac{Y_0}{Y} \cdot \frac{R^3}{r^3} \right)^2 - \left(\frac{Y_0}{Y} \cdot \frac{R^3}{r^3} \right)^2 + 1} \right\} \quad (4-2)$$

where σ_A is the yield stress of ferrite Y , is the yield strength of ferrite left in an elastic state, r is radii of a thin shell in the ferrite, θ is the angle from the $x_1 - x_2$ plane to the r direction. The delay in reaching the plastic deformation stage is due to more times needed to exit from the soft phases of ferrite in ER70S-6, due to higher dislocation density, as shown in Figure 4-28(c). This seems to corroborate the findings of [103].

In the case of the mixed grade of ER70S-6 + ER120S-G fractured in the X-direction, the yielding begins from the preferential yielding zones in the lower strength grade; these consume the ferrite phases with an increase in uni-axial tension stress to reach their plastic limit much earlier, but is pinned in the matrix of the higher strength grade. The interphase of the matrix, as shown in Figure 4-30, is characterised by twin martensite precipitates and retained austenite (RA), hence the failure outside this location in both orientations. The stronger phase martensite will yield after the ferrite has yielded [110]. The soft phase yields as soon as the applied stress exceeds the YS limit of the soft phase and the harder phase will cause a strain hardening effect and hence governs the failure stress. The strain hardening coefficient of 1.40 for mixed grades fractured in the X-direction could be responsible. The applied stress needed to yield the martensite phase and the corresponding mean plastic strain, which is average over the ferrite phase, and the martensite phase is given by Eq (4-3) and (4-4). below [111].

$$\sigma_0 = (1 - f)y_s + f y_h \quad (4-3)$$

$$\epsilon_0 = (1 - f)(y_h - y_s)/(kE) \quad (4-4)$$

The yield strength of the harder phase is y_h , and that of the softer phase is y_s , k is the strain hardening factor, and f is the volume fraction of the harder phase. Here, the Lüders band which is an undesirable problem for the industry disappeared [112]; the continuous yielding phenomenon is observed here since

the volume fraction of martensite and bainite in this mixed grade is much greater than in other phases. Hence internal stress and plastic zones are generated in the entire microstructure due to mobile dislocations. The weighted average yield stress on the transformation induced yielded zone will be reduced to Eq (4-5) [110]

$$\bar{\sigma A} = \int_0^{\pi/2} \sqrt{\left(\frac{3\sin^2 \theta - 1}{2} \cdot \frac{Y_0}{Y}\right)^2 - \left(\frac{Y_0}{Y}\right)^2 + 1} \cdot \cos \theta \, d\theta \quad (a < r \leq R) \quad (4-5)$$

In the plastic deformation zone, the $\bar{\sigma A}$ depends on Y , Y/Y_0 and θ . $\bar{\sigma A}$ is higher for the single grade of ER70S-6 fractured in either direction and the mixed grade fractured in the Z-direction. $\bar{\sigma A}$ is on the decrease in the mixed grade fractured in the X-direction due to the decreasing volume fraction of ferrite in the interphase [103]. The carbon content of ER70S-6 is lower compared ER120S-G and ER90S-B3, as shown in Table 3-1; this is also a contribution to the reduction of the size of the Lüders band [113][114]. Song et al. also showed that the increase in carbon content could reduce the Lüders band from 3.3 % to 2.6 % [114][115]. Mixed grade deposition with WAAM is an indispensable strategy in optimising functionality and strength.

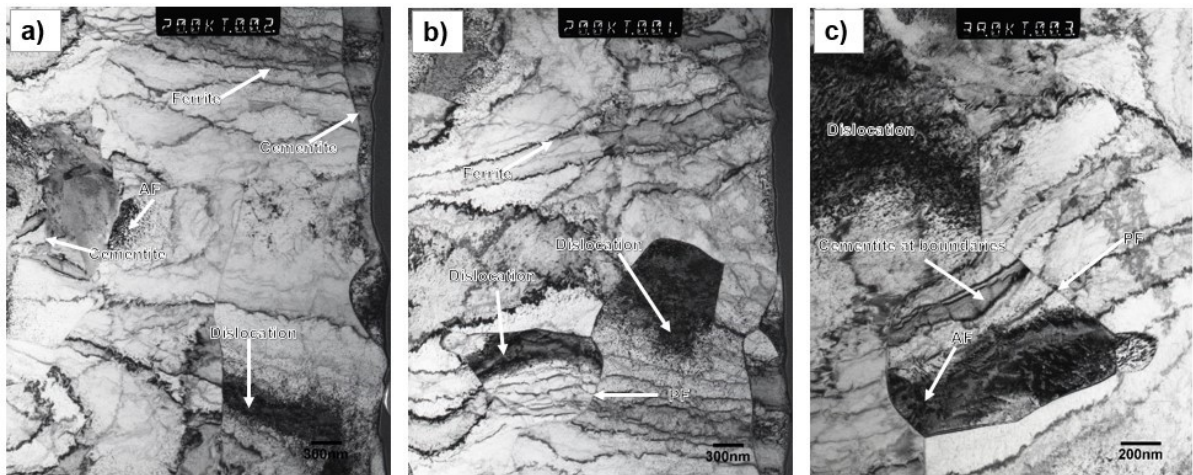


Figure 4-28 (a-c) Bright TEM micrographs of WAAM single pass ER70S-6

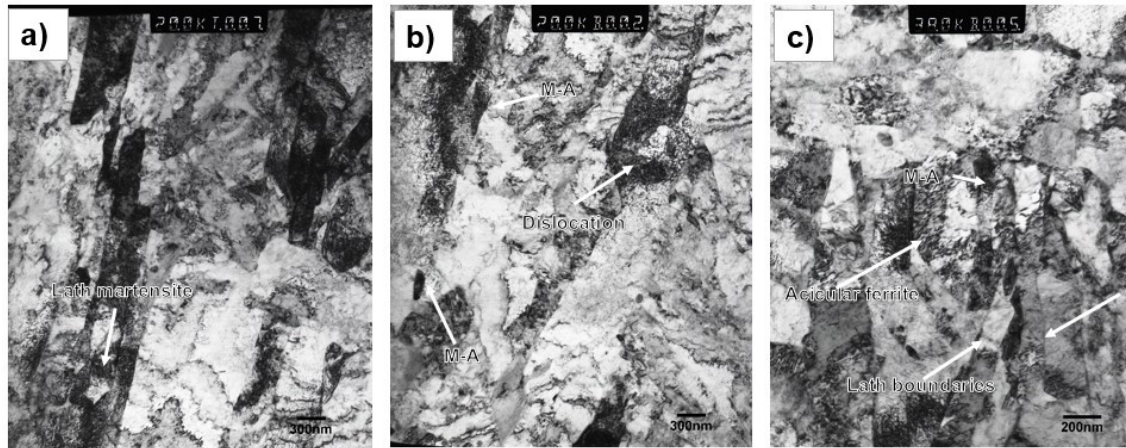


Figure 4-29 (a-c) Bright TEM micrographs of WAAM single pass ER120S-G

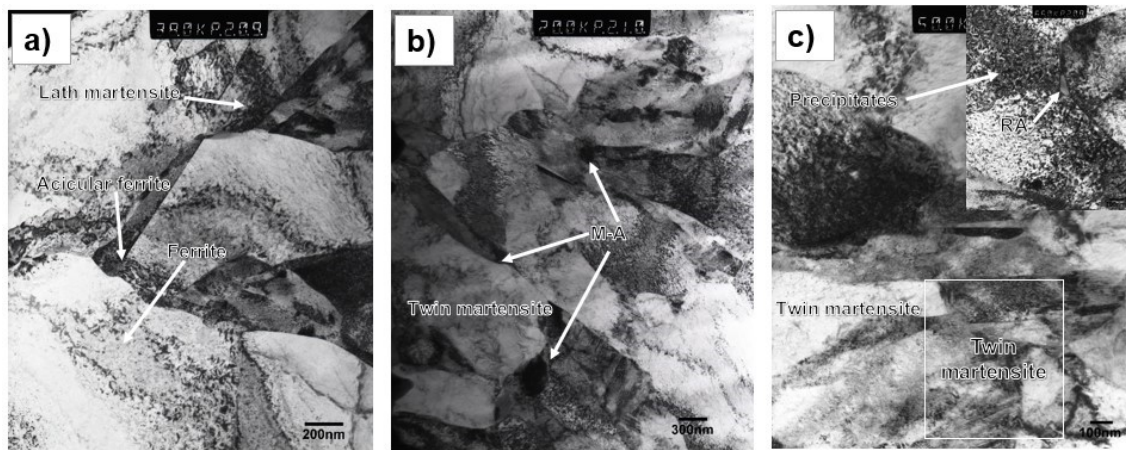


Figure 4-30 (a-c) Bright TEM micrographs of WAAM single pass ER120S-G + ER70S-6 (interface)

4.4.2 DIC with stress-strain field maps for mixed grades interactions

The effects of stress and strain on the mixed grades were analysed using the DIC set-up, as shown in Figure 3-9. Figure 4-31 shows the strain field map in ER70S-6 (single grade). At the onset of load application to the single grades, the strain was distributed uniformly along the entire structure, as shown in Figure 4-31(a & b). As the uniaxial tension continues, voids begin to build up at the region of stress concentrations. These are regions of weakest dislocation due to microstructural incompatibility, defects, and hard spots. This marked the beginning of plastic deformation on the structure as the atoms of the sample begin to agitate with the

continuous straining. Further application of uniaxial tension loading leads to yielding and necking, as shown in Figure 4-31(c & d) where it plastically deforms as the strength of the material can no longer sustain the loading and hence coalescence of voids and rapid failure occurred at this load. This form of plastic behaviour is unique to the single grades fractured in any direction.

Figure 4-32 show the strain field map for mixed grades fractured in the Z-direction. At the onset of load application to the mixed grades, the strain was distributed uniformly along the structure but was more pronounced on the lower strength material, as seen in Figure 4-32(a & b). With time, as the uniaxial tension continues, voids begin to build up at the region of stress concentrations (regions of weakest dislocation due to microstructural incompatibility, defects and hard spots), this marked the beginning of elastic deformation on the structure as the atoms of the sample begin to agitate with the continuous straining. Further application of uniaxial tension loading leads to yielding and necking, as shown in Figure 4-32(c & d). The failure is on the lower grade material of the matrix. The interface is stronger, as revealed in Figure 4-30, with the formation of twin martensite.

Figure 4-34, present the stress-strain field maps of mixed grades fractured in the X-direction. This mixed matrix behaves differently from the Z-direction loading. On the application of load, the strain was distributed more uniformly across the structure and held responsive by the stronger grade in the matrix. As the load increases, beyond the YS of the lower grade material, the lower grades begin to form voids at the stress concentrated points; these voids continue to grow on the lower grade material while the higher grade materials continue to hold the matrix in place, as seen in Figure 4-34(b & c). Further application of uniaxial tension loading leads to yielding where it plastically deforms as the strength of the stronger material can no longer sustain the loading and hence coalescence of voids and rapid failure occurred, as shown in Figure 4-34(d).

The young modulus derived for the mixed grades loaded in the Z-direction is higher than the mixed matrix in the X-direction. In Figure 4-15, ER120S-G + ER90S-B3 had the highest G value of 239 GPa; this is attributed to the higher

strength of the ER90S-B3 + ER120S-G mixed grade due to the presence of a harder martensite phase which strengthens the matrix and controls the deformation attributes of the ferrite phase by introducing internal stress caused by the strain gradient between the two phases [116]. The modulus derived for this multi-alloy grade loaded in the X-direction are ER70S-6 + ER90S-B3; 190GPa, ER90S-B3 + ER120S-G; 214GPa and ER70S-6 + ER120S-G; 210GPa. This is lower than the Z-direction, which maybe because of the orientation effects as a result of the matrix. However, a conclusion cannot be drawn from here because different sizes of the tensile piece were used for the different orientations and since E does not depend on geometry, further investigation needs to be carried out.

Figure 4-33 present the strain field distribution and the YS at fracture using a MATLAB code to extract the DIC analysis for the mixed grades interaction in Z loading directions. It is evident that no yield data was recorded in the ER120S-G segment of the structure (ER70S-6 + ER120S-G) (Figure 4-33(b)) and the YS at failure was about 387MPa comparing these to the ER120S-G + ER90S-B3 mixed grade structure which showed some yield data on both grades as seen in Figure 4-33(d), but fracture at higher YS of 645MPa (Figure 4-33(d)). Loading mixed grades in Z- the direction has shown here that the lower strength material governs the fracture and the interface is stronger than the lower grade in the metrics. Figure 4-35 present the strain field distribution and the YS at fracture using DIC analysis for the mixed grades interaction in X loading directions. Different modulus of elasticity was applied to the structure to evaluate their responses and how they affect the deformation pattern of the graded structure. Failure patterns were noted. It is evident from the strain map that yield data was observed in the entire mixed grades and the YS at fracture is over 460MPa, this is higher than mixed grades loaded in Z direction of the same material.

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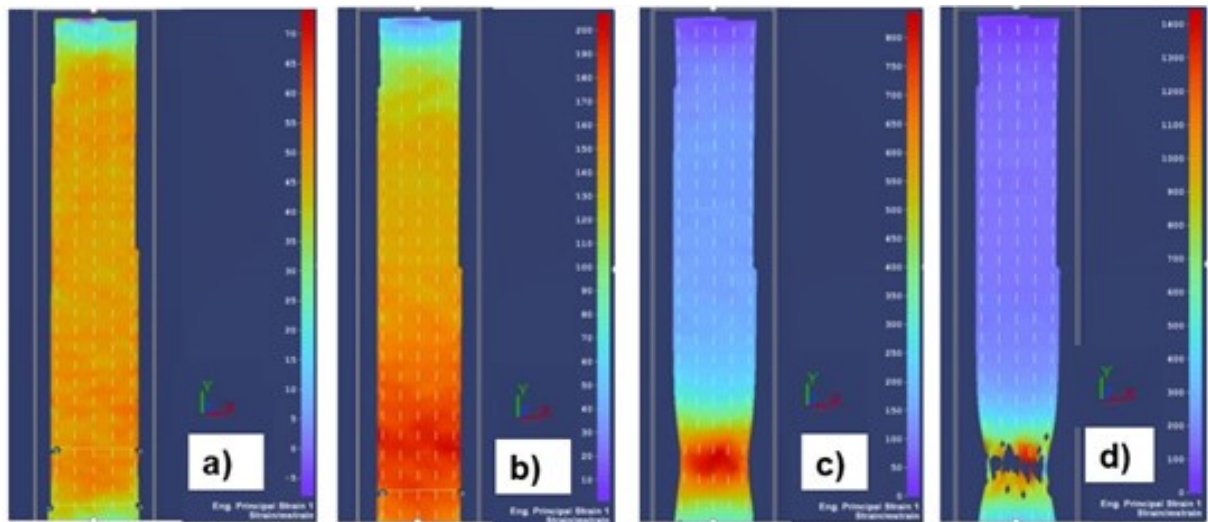


Figure 4-31 Strain field map in ER70S-6 (single grade)

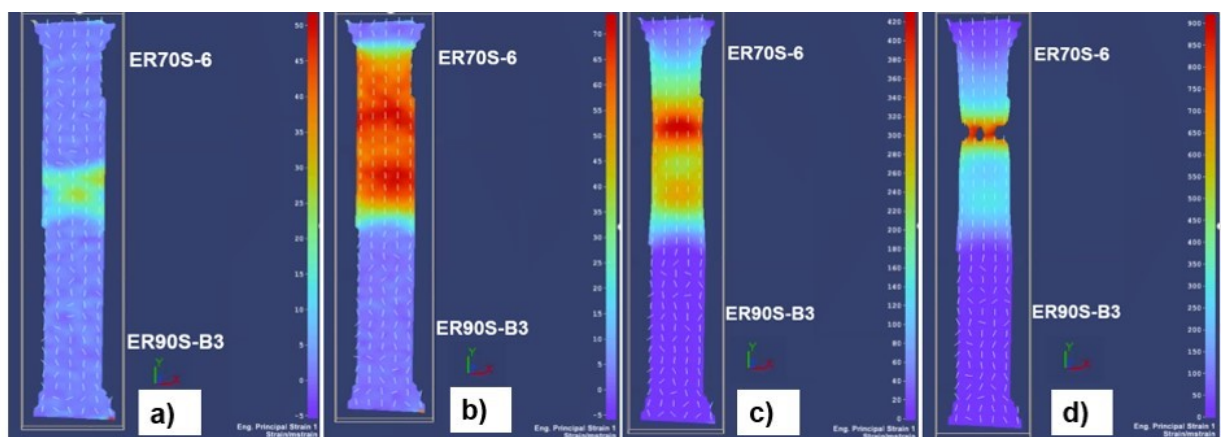


Figure 4-32 Strain field map in ER70S-6 + ER90S-B3 – Z

Tailored Deposition of Wire Grades

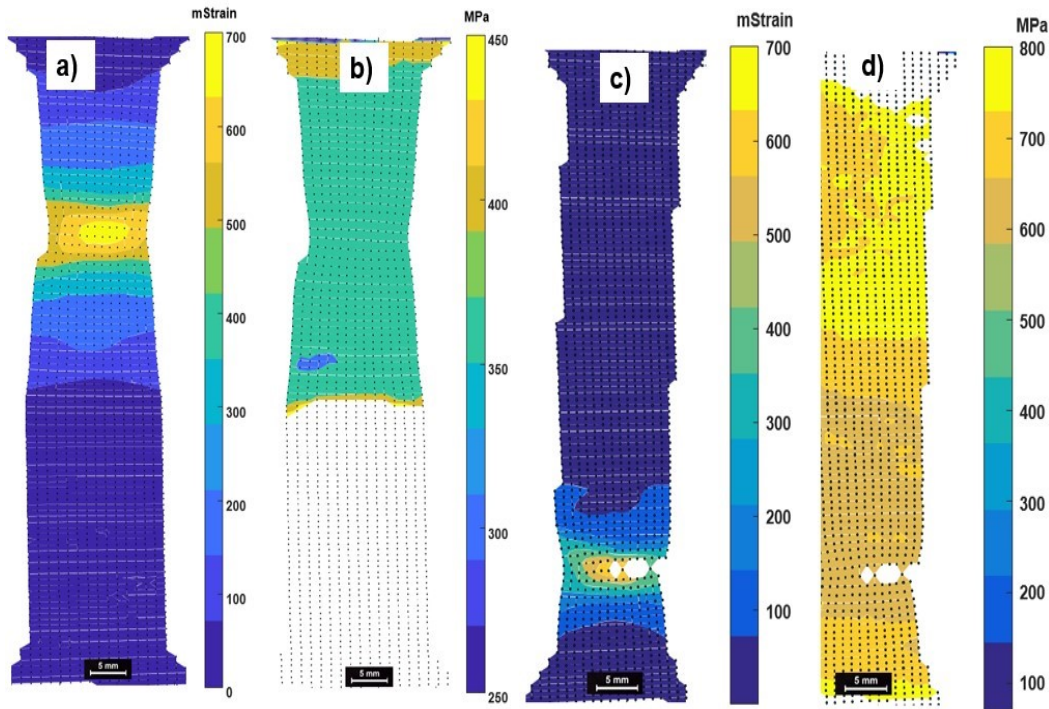


Figure 4-33 Stress-strain interaction in Z-direction

- (a) Max strain map (ER 70S-6 + ER120S-G) (b) YS map showing interface of ER70S-G & ER120S-G (c) max strain map (ER120S-g+ER90S-B3) (e) YS map showing interface of ER120S-G & ER90S-B3

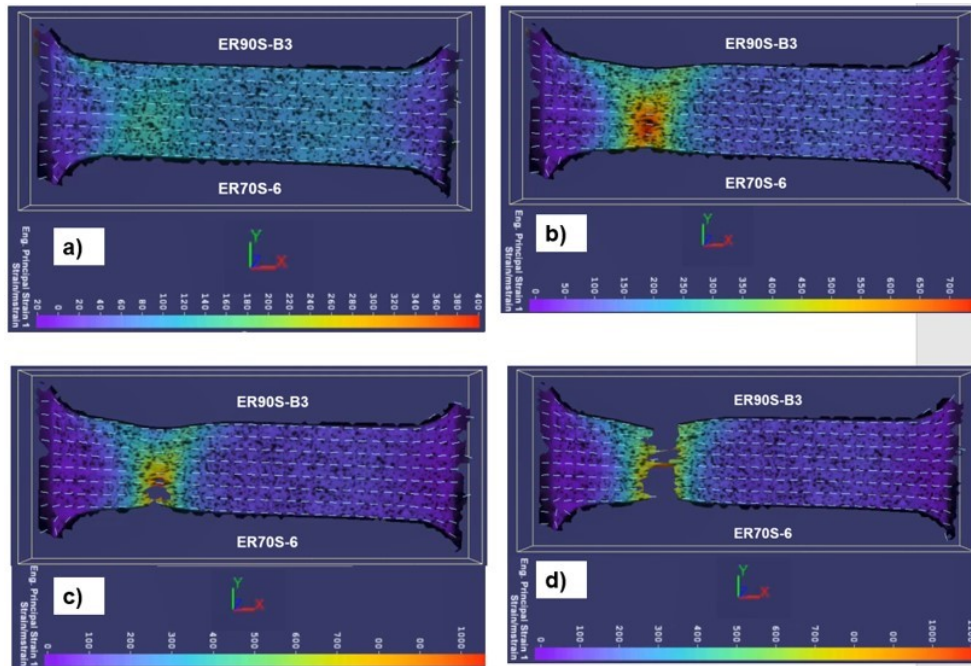


Figure 4-34 Strain field map in ER70S-6 + ER90S-B3 – X

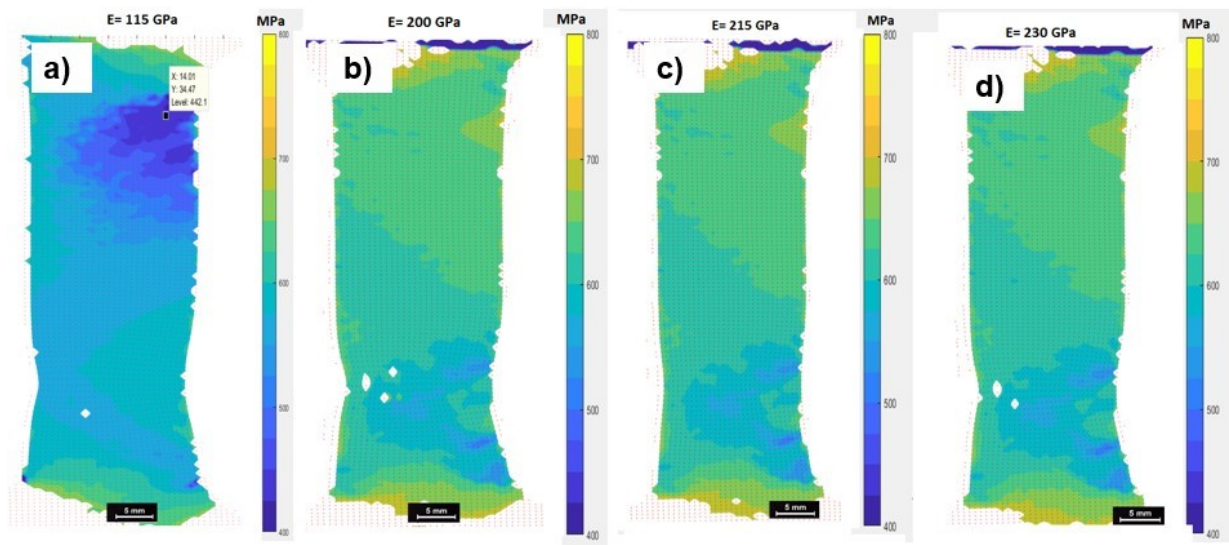


Figure 4-35 Yield stress interaction in ER120S-G + ER70S-6 X-direction

Young's modulus of (a) 115 GPa (b) E 200 GPa (c) 215 GPa (d) 230 GPa

4.4.3 Conclusion

In this chapter, the mechanical and microstructural properties of WAAM single pass single grade and WAAM single pass graded structures deposited using the standard Cold metal transfer process and machined in the X & Z direction respectively were evaluated. Also, the suitability of use in the as-deposited condition was considered. From this investigation, it was shown that:

1. The mechanical properties of WAAM single-pass single grades and mixed grades match well with the wrought counterparts.
2. The mechanical response of WAAM as-deposited single grade to tensile loading is similar to the machined condition, but with lower mechanical strength, this goes forth to show that WAAM components can be manufactured and deployed to the industry without further or with less machining, hence cost reduction and lead-time-saving strategy.
3. The microstructural evolution obtained showed, that similar weak texture was obtained with the single pass strategy using HSLA steels, hence the isotropic lath martensitic microstructure in random crystallographic

orientation. This shows that single pass deposition strategy is excellent with WAAM.

4. The microstructural attributes of the graded structure are like those obtained with the single grades. The results show a very similar texture type and generally weak texture intensity without any preferential crystallographic direction, hence isotropic in nature.
5. Graded deposition using WAAM is a key tool for topological optimisation, as the properties were higher than the results obtained in the single grade WAAM.
6. The disappearance of the Lüders band in graded structure with mild steel, which is an undesirable problem for the industries, is an indispensable strategy to optimise the functionality of applications. With a multi-alloy deposition strategy, the required strength of a segment of a component in a structure can be engineered using novel materials, thereby saving cost, time and wastage

The investigation in the present chapter, demonstrate the concept that using HSLA steel wires will be suitable for making bespoke components for the offshore energy sector. Chapter 5 will investigate further the influence of surface waviness on fatigue initiation strength of the as-deposited condition since it has a mild effect on tensile strength properties. WAAM process parameters causing surface waviness will be investigated to know how to reduce or eliminate the waviness, hence saving machining cost and production time. Subsequently, Chapter 6 will be investigating further the effects of surface rolling on the fatigue initiation properties of WAAM structures.

5 INFLUENCE OF SURFACE TOPOGRAPHY ON FATIGUE INITIATION

5.1 Introduction

Surface waviness (SW) is one of the major issues in wire plus arc additive manufactured (WAAM) components, as it would restrict the application of WAAM components in the as-built state due to geometric stress concentration factors from such waviness. Despite all the progress made in WAAM, detailed studies on the relationship between the stress concentration effect from SW and the fatigue initiation strength of an as-built component have not been done. To effectively and economically commercialise WAAM, the integrity and reliability of built components under specific loading conditions must be assured.

The first objective of this study is to evaluate how the WAAM deposition parameters can be modified to improve the fatigue initiation response of the as-deposited structure. And, the responses of different as-deposited structures to fatigue initiation strength for different SW levels. The second objective of this study is to evaluate the effects of deposition rate on the fatigue initiation strength. The third and last objective is to study the effects of stress concentration caused by SW on the fatigue initiation strength of as-built WAAM steel and also to evaluate the effect of elastic stress concentration factor (S_{el}) and effective stress concentration factor (S_{eff}) on the fatigue initiation strength. All analysis was carried out with a single-pass deposition strategy.

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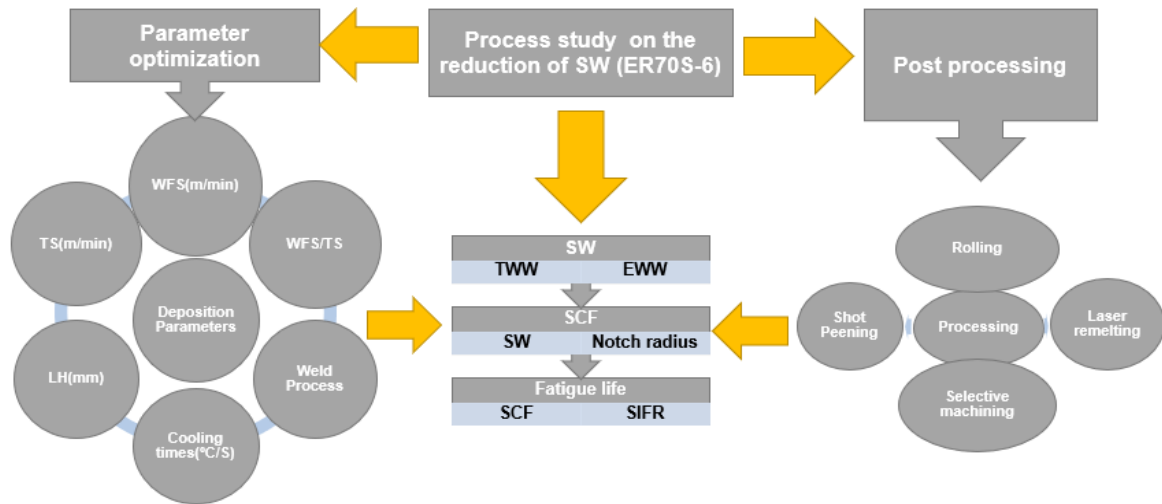


Figure 5-1 Study flow

5.2 Experimental procedure

The WAAM of the structures was carried out with standard cold metal transfer power sources, which is a modified GMA process based on a controlled dip transfer mode of metal transfer, as reported by [117]. The set-up shown in Figure 3-1 was used in this experimentation. Local shielding was used with Specs shield (Ar + 20%CO₂) at a flow rate of 15L/min. The base plates used as substrate in the experiment were rolled structural steel plates of S275 JR (EN10025) with dimensions of 350mm x 200mm x 15mm. The plates were secured by steel jig tightened to an aluminium base plate. The wires used in this experiment were ER70S-6 wires with diameters of 1.0mm and 1.2mm, with chemical composition and mechanical properties as shown in Table 3-1 and

Table 3-2. Single-pass deposition strategy, as described by [118], was used to deposit the layers, with every next layer deposited in the opposite direction. All deposition was done in the PA (1G/Down hand) position from the base of the plate with the torch and wire feeding in the vertical direction. In stage one of the studies, relationships between WAAM process parameters (TS, WFS, WFS/TS) and SW were established. Four trials (A, B, C & D) were carried out in stage two. The first five structures termed trial A (used to study the effect of layer height on SW) were built to a layer height (LH) of 0.8, 1.2, 1.6, 2.0 & 2.3mm to establish

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different SW levels. The subsequent built structures termed trial B (used to study the effect of SW on fatigue life in the as-deposited state) and trial C (used to study the impact of machining on fatigue life) were built with a TS of 0.4m/min, and wire feed speed to travel speed ratio (WFS/TS) of 10.0. Trial D (used to study the effect of wire diameter on SW & fatigue life in the as-deposited state) was built with a TS of 0.4m/min, and wire feed speed to travel speed ratio (WFS/TS) 16.25 respectively.

Figure 5-2 shows the deposited structure with build orientation. Trial A had six samples machined only in the X-direction, as shown in Figure 5-2 (a) (X1-X6). Trials B, C, and D had 12 samples each and were machined in the X- & Z-directions, as shown in Figure 5-2(a). Wire with a diameter of 1.0mm was used in all cases except for trial D, which was deposited with a 1.2mm wire diameter. 42 fatigue specimens in the shape of a standard dog-bone specimen were machined out. Trials A, B, and D (30 in number) were only machined at the grip end (42.73mm), as shown in Figure 5-2(b) hence called the as-built condition. Trial C samples (12 in number) were all machined and termed as machined specimens. 18 tensile test specimens were extracted from trials B, C and D (six from each) in the X-direction. Trial C specimens were all machined while trials B and D specimens were only machined at the grip end, as shown in Figure 5-2 (c).

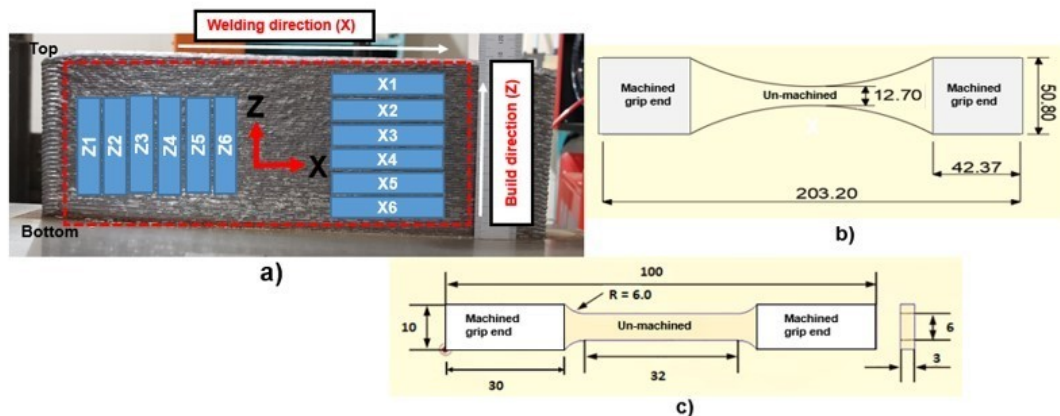


Figure 5-2 Built and test matrix

- (a) Deposited wall showing build orientations (b) Fatigue testing geometry (c) Tensile testing geometry, fatigue samples in as-deposited conditions are shown in appendices (Fig.2)

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The axial tension-tension fatigue testing was carried out as enumerated in Section 3.5.3. Fatigue response of as-built trial A specimens were established at a maximum stress (σ_{max}) of 343.7MPa (which is 70% of the machined ultimate tensile strength), minimum stress (σ_{min}) of 34.37MPa, stress amplitude ($\Delta\sigma$) of 154.35MPa and stress range of 308.7MPa. In order to obtain a maximum stress of 343.7MPa across the trial A specimen, P_{max} varies across the specimen due to its thickness variations. Fatigue responses of trial C specimen (machined condition), and trials B and D specimens (as-built condition) were established at 60-90% of the ultimate tensile strength (UTS) value (different stress levels) to build an S-N curve. All fatigue and tensile samples were prepared in accordance with Sections 3.5.3 and 3.5.1. The uni-axial tensile testing was performed in accordance with Section 3.5.1.

The cross-sections of the as-deposited specimen were mounted, ground and polished before etching with 5% Nital. A macro examination was carried out on each cross-section of the as-deposited specimen to analyse the SW. An optical microscope of 40X magnification was used to understand the profile of the waviness of the specimen. A scanning electron microscope of LYRA3 TESCAN was used to characterise the fractured surface.

The images of SW were imported to solid works to measure the depth, a and radius of curvature, ρ . An average of six measurements was taken from each specimen. A Bruker ContourGT was used for 3D optical profilometry; it utilises vertical scanning/white light interferometry. A typical topography scan was taken of the X-Z plane with a single image area of $1.2 \times 0.95\text{mm}$. A total area of $5 \times 5\text{mm}$ was scanned for every sample to characterise the surface roughness, as shown in Table 5-1 and Figure 5-3. The results were analysed using VISION64 software to visualise the 3D contour and to calculate the different surface roughness parameters of the machined trial C and as-deposited trials B & D.

Table 5-1 Surface Characterisation

Nomenclature	Description
R_a :	Mean surface roughness ($R_a = \frac{1}{n} \sum_{i=1}^n y_i $)
R_v :	Maximum valley depth
R_p :	Maximum peak height
R_t :	Maximum height of the profile ($R_p - R_v$)
R_q :	Root mean square surface roughness $R_q = \frac{1}{n} \sqrt{\sum_{i=1}^n y_i^2}$

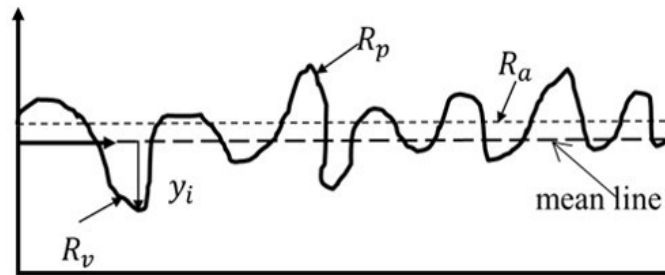


Figure 5-3 Surface profile

5.3 Results

5.3.1 Surface roughness Characterisation

The average surface roughness measurements of trials B & D samples are presented in Table 5-2 while the roughness map in 3-D is shown in Figure 5-5. The average surface roughness of the trial B specimen (as-deposited built with WFS of 4.0m/min and TS of 0.4m/min), and trial D specimen (the as-deposited built with WFS of 6.5m/min and TS of 0.4m/min) is characterised in terms of mean surface roughness (R_a), maximum valley depth (R_v), maximum peak height (R_p), maximum height of the profile (R_t), and root mean square surface roughness (R_q) as shown in Table 5-2. Scan images of B1 & B2 roughness in the build direction are shown in Figure 5-4.

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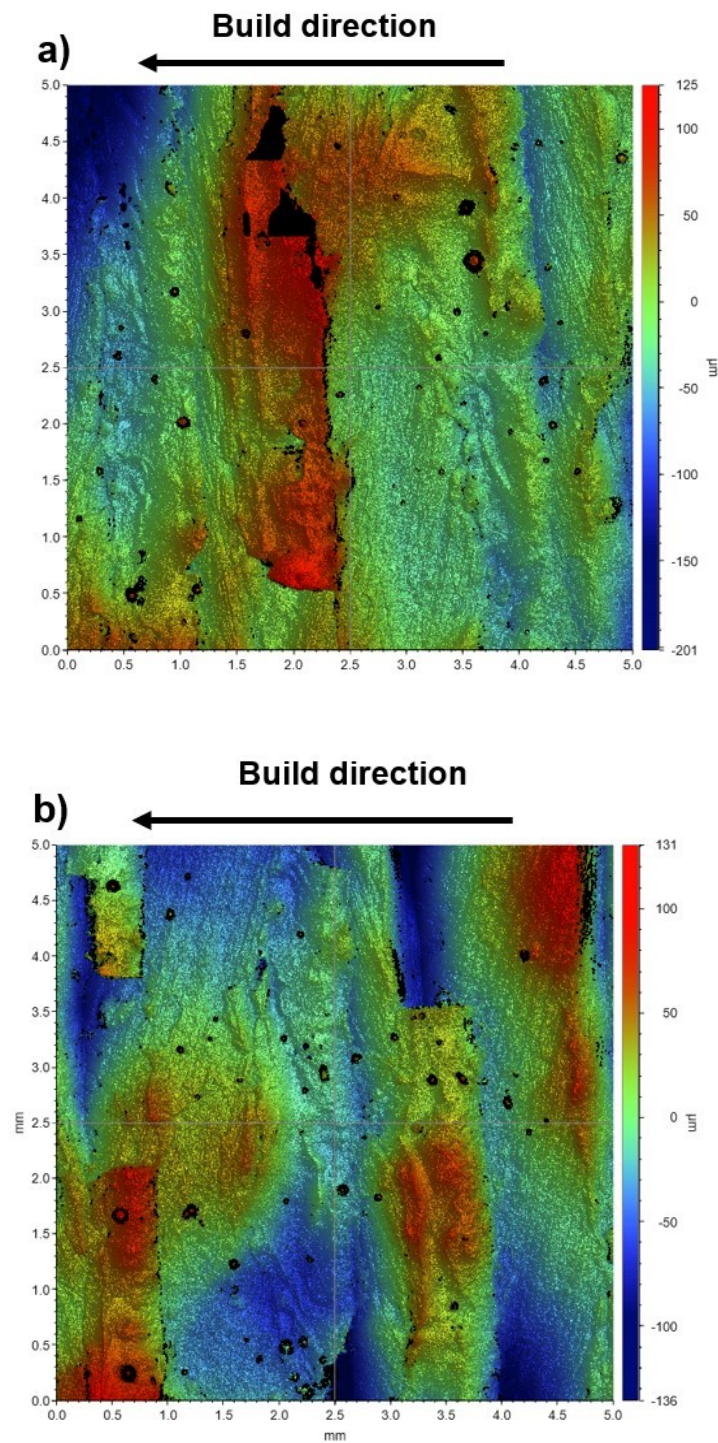


Figure 5-4 Surface profile

(a) $R_a: 30.62\mu\text{m}$ (b) $R_a: 33.49\mu\text{m}$

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Table 5-2 Average surface roughness values of as-deposited and machined

Wall	Surface roughness	Ra(μm)	Rp(μm)	Rq(μm)	Rt(μm)	Rv(μm)
B-1	ASD(SW: 0.18mm)	33.49	125.35	43.76	326.74	-201.38
B-2	ASD(SW:0.18mm)	30.62	130.5	37.37	266.13	-135.60
D-1	ASD(SW: 0.47mm)	100.77	313.73	122.76	683.87	-270.15
D-2	ASD(SW:0.47mm)	100.45	534.9	122.88	826.29	-291.36
C-1	Machined	1.32	41.09	1.89	51.21	-10.12

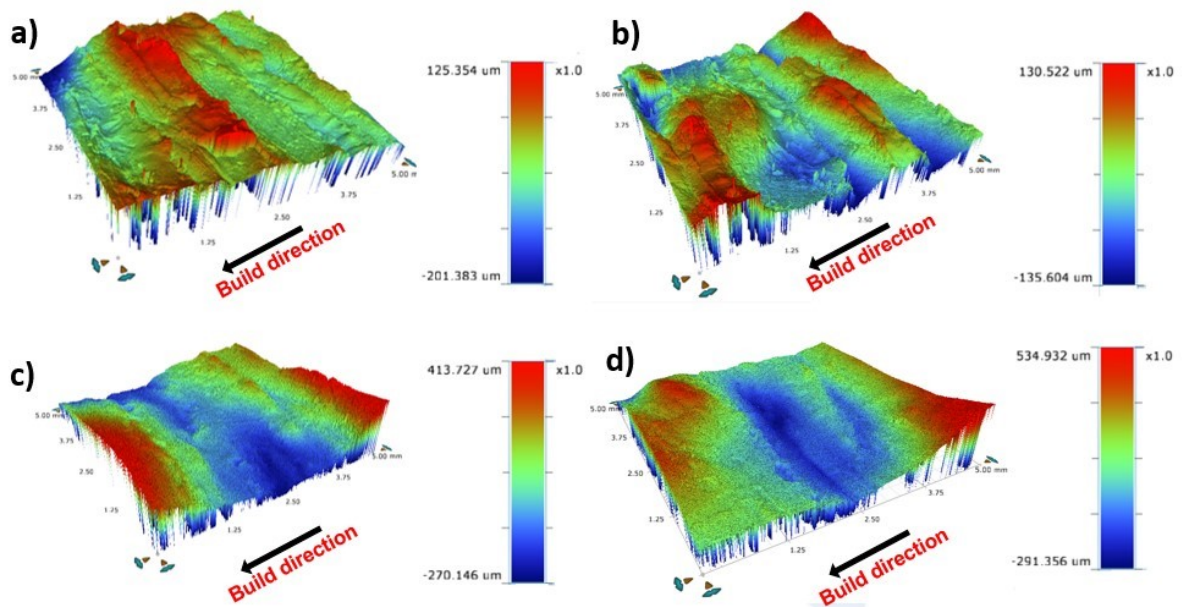


Figure 5-5 Roughness measurement

(a & b) Trial B (as-deposited with SW of 0.18mm) (Ra; 32.05 μm) (c & d) Trial D (as-deposited with SW of 0.47mm) (Ra; 100.61 μm)

5.3.2 Surface waviness, deposition rate and notch measurement

The as-deposited WAAM profile is shown in Figure 5-6(a & b). The etched cross-section of the as-deposited and the as-deposited schematic illustrating the notch depth, effective wall width (E_{WW}) and total wall width(T_{WW}), is shown in Figure

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5-6(c & d). The SW here was characterised as the average maximum peak to valley distance, measured on a cross-section of the deposit; this same approach was described by [119] as given in Eq (5-1). The relationship between LH, re-melt-ratio (RR), notch radius (ρ), E_{WW} , T_{WW} and SW on WAAM deposits is shown in Figure 5-7(a & b). According to [120], the relationship between (E_{WW}) and LH in the WAAM melt pool, as a result of the process parameter interaction, determines the final appearance of the WAAM as-deposited component.

$$S_w = \frac{T_{WW} - E_{WW}}{2} \quad (5-1)$$

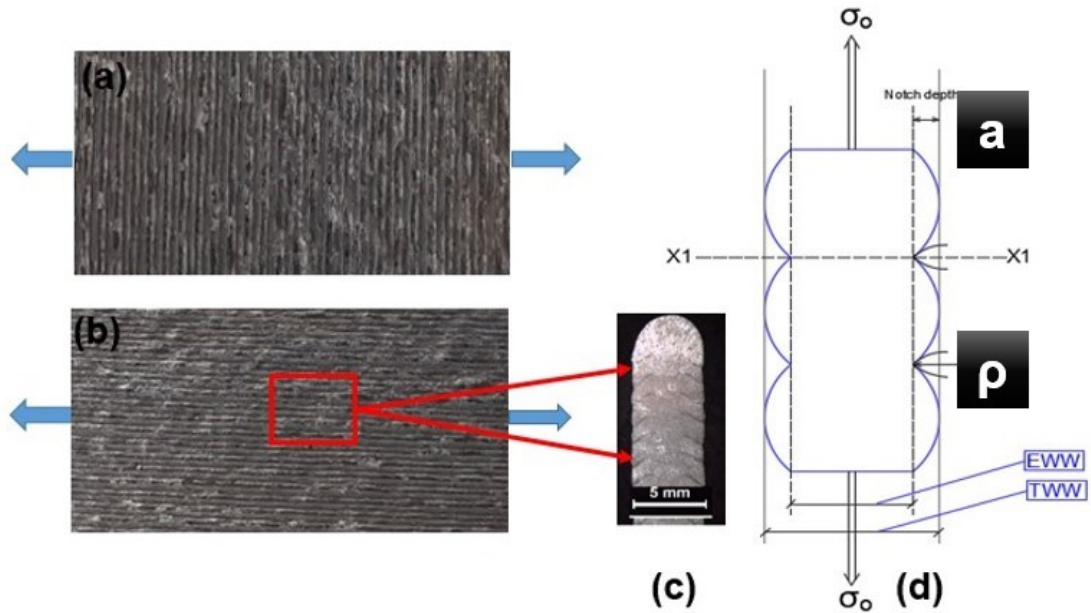


Figure 5-6 As-deposited WAAM profile

(a) As-deposited WAAM loading in building direction (Z-direction) (b) As-deposited WAAM loading in welding direction (X-direction) (c) Etched cross-section of as-deposited WAAM (d) As-deposited schematic illustrating SW parameters and notch depth

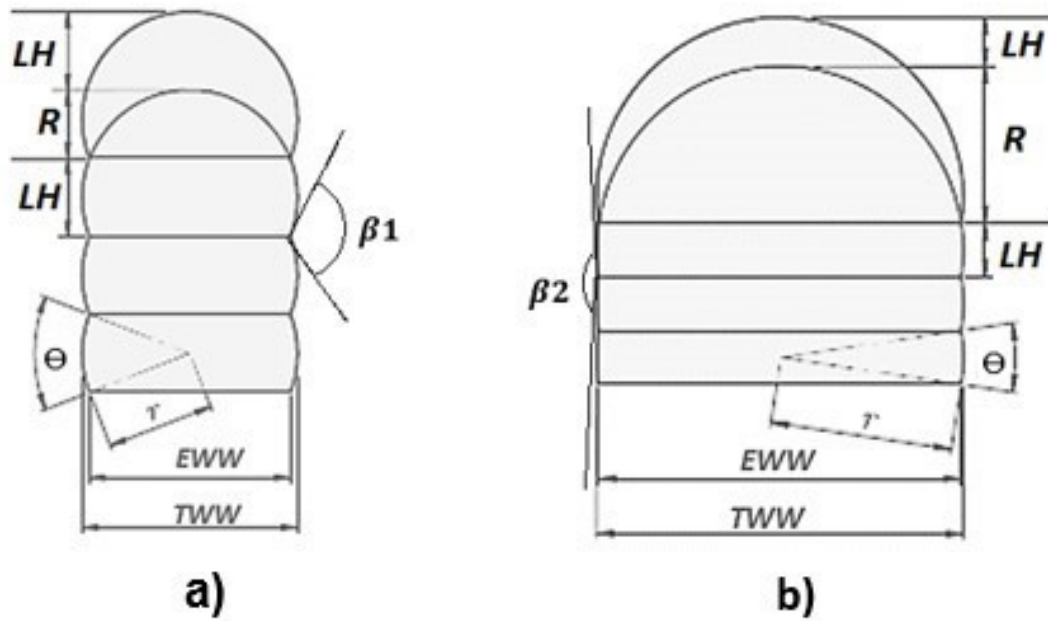


Figure 5-7 WAAM profile schematic illustration

(a) Built layer geometry with high LH (b) Built layer geometry with low LH
[120]

Figure 5-8 presents SW profiles of trials A, B & D samples with their LH. Calculating the SW using Eq.(5-1) above gives 0.18, 0.20, 0.22, 0.28 and 0.33mm respectively for trial A samples. Trial D specimens deposited with 1.2mm wire offered an average SW of 0.47mm. The SW is influenced by the deposition parameters, such as the ratio of wire feed speed to travel speed (WFS/TS), LH, wire diameter, cooling time and chemical composition of the wire used in the deposition [55]. This indicates that better SW could be obtained in WAAM parts by optimising wire size and deposition parameters as demonstrated here.

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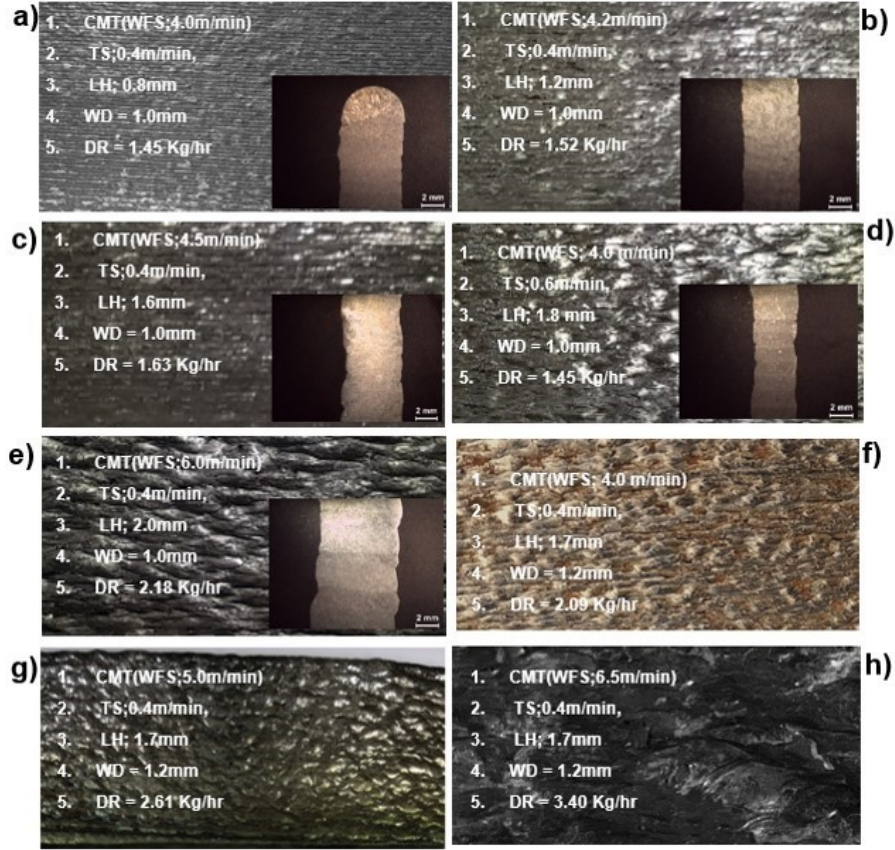


Figure 5-8 Surface profile of as-deposited

(a) SW of 0.18mm (LH: 0.8mm) (b) SW of 0.20mm (LH: 1.2mm) (c) SW of 0.22mm (LH: 1.6mm) (d) SW of 0.28mm (LH: 1.8mm) (e) SW of 0.33mm (LH: 2.0mm) (f) SW of 0.24mm (LH: 1.7mm), (g) SW of 0.30mm (LH: 1.7mm), (h) SW of 0.47mm (LH: 1.7mm)

$$D_{rate_{Kg/hr}} = \rho \cdot \left[\pi \cdot \left(\frac{CMT_wire\ diam}{2} \right)^2 \cdot CMT_wire_{wfs} \right] \quad (5-2)$$

The DR is computed using the relationship in Eq.(5-2), where the density of steel (ρ) is taken as $7700Kg/m^3$. CMT wire diameter and WFS were imputed as used. Table 5-3 shows the relationship between SW and DR.

Table 5-3 Relationship between SW and DR (Kg/hr)

Wire diam (mm)	Average SW (mm)	WFS (m/min)	DR (Kg/hr)
1.0	0.18	4.0	1.45
1.0	0.22	4.5	1.63
1.0	0.33	6.0	2.18
1.2	0.30	6.0	2.61
1.2	0.47	6.5	3.40

5.3.3 Determination of Elastic and Effective Stress Concentration Factors

Two parameters were used in this study to characterise the stress amplification states: the elastic stress concentration factor (S_{el}) which describes that of the as-deposited sample with SW, and the effective stress concentration factor (S_{eff}) which evaluates the stress concentration factor of the as-deposited in relation to the machined condition from the fatigue failure curves. Notch parameter measurements (notch depth a and radius ρ) for trials A & D specimens were measured and inputted into Eq. (5-3) to calculate the elastic stress concentration factor, as given by [121].

$$S_{el} = 1 + 2 \sqrt{\left(\frac{a}{\rho}\right)} \quad (5-3)$$

The average S_{el} values obtained for the as-deposited condition of trials A & D specimens are presented in Table 5-4.

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Table 5-4 the average value of SW and the resultant S_{el} for trials A & D samples

Trial/Wire diam(mm)	Average SW(mm)	S_{el}	LH
A/1.0	0.18	1.99	0.8
A/1.0	0.20	2.15	1.2
A/1.0	0.22	2.35	1.6
A/1.0	0.28	2.39	1.8
A/1.0	0.33	2.47	2.0
D/1.2	0.47	2.63	1.7

Trial D specimen deposited with a WFS of 6.5m/min with 1.2mm diameter wire had an average S_{el} value of 2.63.

5.3.4 Mechanical Properties

Table 5-5 summarises the average mechanical properties of WAAM in both machined and as-deposited condition in the X-direction. Higher UTS and YS values were observed in the machined condition compared to the as-deposited condition. It is evident that the surface condition reduces elongation, UTS, and YS.

Table 5-5 Average mechanical properties in both condition (X-direction)

Surface condition	YS(MPa)	UTS(MPa)	PCE (%)
ASD (SW: 0.47mm) (trial D)	280	382	28
ASD(SW:0.18mm) (trial B	307	405	30
Machined (WAAM, this study) trial C	376	491	38
Machined(WAAM) [58]	410	480	28

5.3.5 Fatigue Response

Fatigue responses of trials A, C & D specimens built with different SW levels were tested at a constant stress level of 70% for the machined condition UTS (491MPa) value, which is equivalent to 343.7MPa, and are presented in Figure 5-9. This stress level was chosen because the components are expected to withstand up to 70% of the maximum UTS level in service, and both machined and as-deposited will operate under the same loading condition. The ratio of the maximum applied stress in fatigue to the YS of the as-deposited condition ($\sigma_{max}|\sigma_{ys}$) is 1.227.

Fatigue cracks initiated at the notches perpendicular to the fatigue loading; this has a good agreement with the analysis results given by [78] who stated that fatigue crack initiation often starts from points of metallurgical incompatibility, such as notches. In the case of the X-direction specimen, there are peaks and valleys running parallel to the direction of loading, causing plastic incompatibility and stress concentrations due to higher strain. Taking a near-surface crack as a micro notch with a crack depth of a (mm) and a total crack length of $d + a$, when d is taken as the SW, the stress intensity factor (K_I) for cracks in notches can be likened to non-propagating cracks, as discussed by [122]. This relationship was also used by Chapetti to estimate the driving force for any given geometrical loading and condition [79]. The relationship is given by Eq. (5-4)

$$\Delta K_I = Y\Delta\sigma\sqrt{\pi(sw + a)} \quad \text{or} \quad \Delta K_I = Y\Delta\sigma\sqrt{\pi(sw)} \quad (5-4)$$

where $\Delta\sigma$ is the stress amplitude, SW is the maximum notch depth, a is the physical crack depth and is equal to zero in this case, and Y is the geometric correction factor. In this experimental work the Y value is taken to be 1.1 for the geometry [124].

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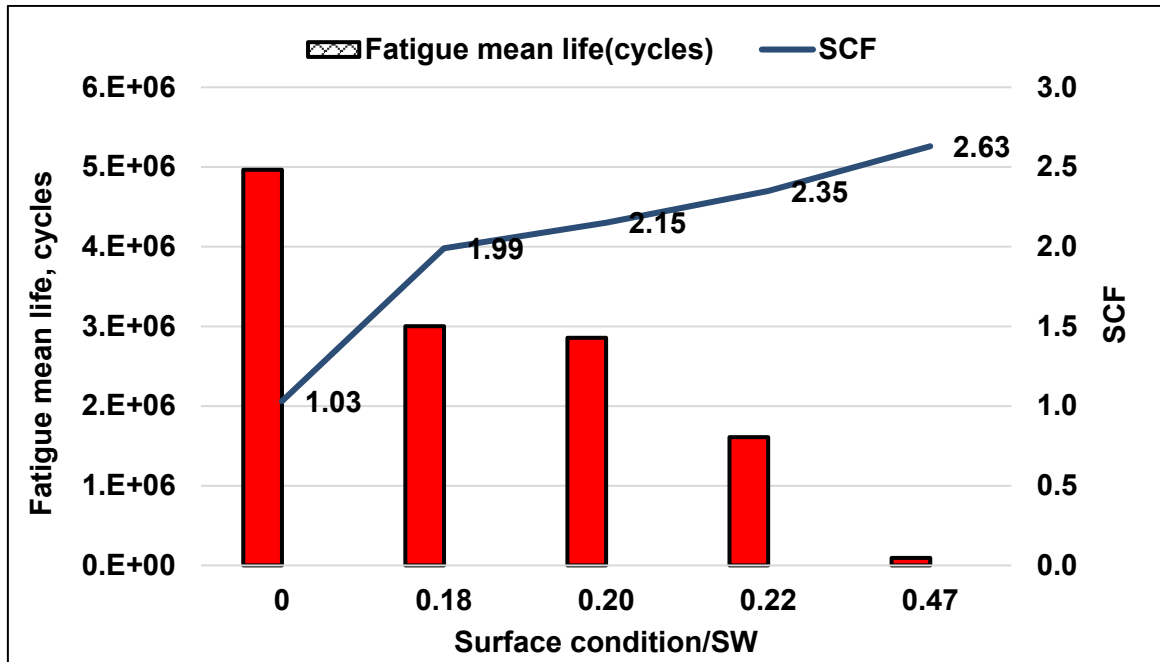


Figure 5-9 Correlation of fatigue life with surface condition/SCF at constant stress level (343.7MPa)

5.4 Discussion of results

5.4.1 The effect of WAAM process parameters on SW

5.4.1.1 Effects of TS on SW

Figure 5-10 presents the effects of TS on SW. It can be seen that TS has a strong influence on SW, particularly at lower TS regimes. Figure 5-11 presents the deposited layer profile for both WFS and TS variations. The higher SW observed is due to the deposition of the larger cross-sectional area obtained when using lower TS at constant WFS, which is as a result of the high HI incurred within the beads and ultimately leads to a higher WFS/TS ratio. Increasing the TS leads to the deposition of less volume of material per unit length.

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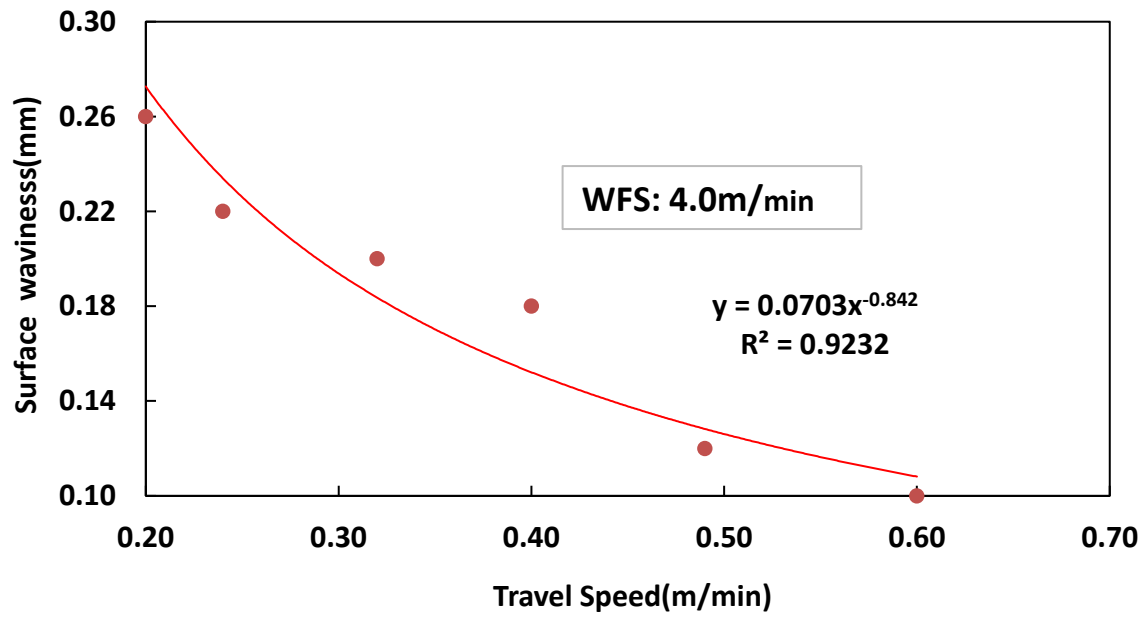


Figure 5-10 TS against SW

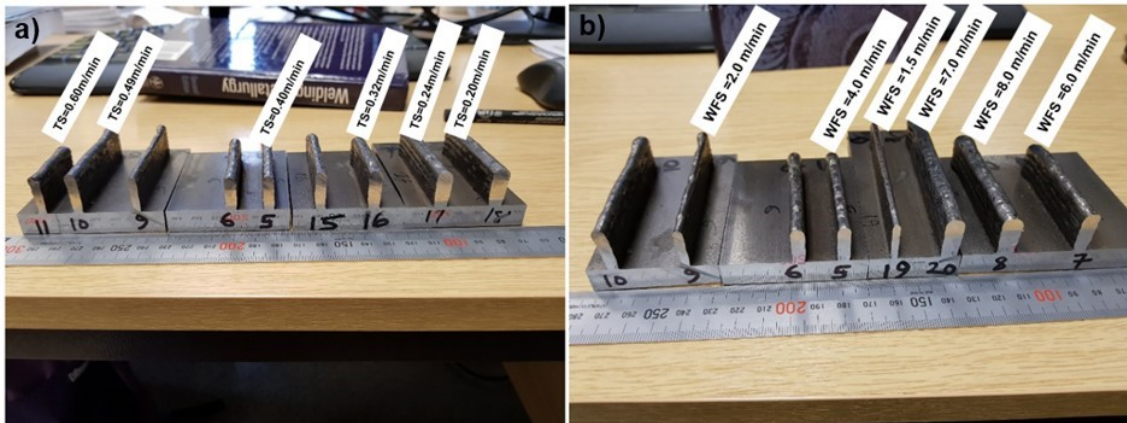


Figure 5-11 Deposited layer profile

(a) Effects of TS (b) Effects of WFS

5.4.1.2 Effect of WFS on SW

Figure 5-12 presents the variation of WFS against SW. WFS has a strong influence on SW, particularly the higher WFS regimes. Increasing WFS leads to the deposition of larger cross-sectional areas as a result of the increase in the arc pressure and droplet impact.

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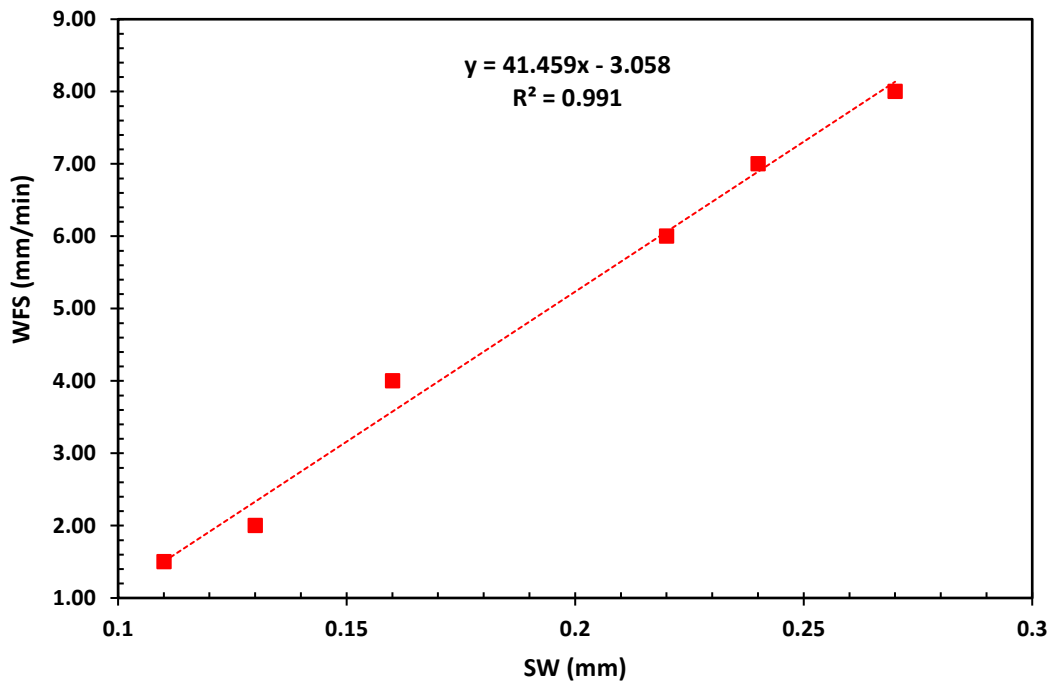


Figure 5-12 WFS against SW

5.4.1.3 Effect of layer height on SW (formation mechanism)

Figure 5-13 presents the effect of LH on SW for as-deposited. It can be seen from the plot that the increase in LH leads to a corresponding increase in SW. LH plays a vital role in the mechanism of SW formation in WAAM. Two key formation mechanisms were identified by Geng et al. as being the re-melting and wetting (spreading mechanism) [125]. A higher re-melting ratio, as a result of more heat input into the melt pool and deeper penetration of the melt pool, which is usually accompanied with low LH, is responsible for the low SW producing a wider EWW wall, as seen in Figure 5-7(b). Wetting and spreading modes are accompanied by higher LH. The low re-melting ratio results in a shallow melt pool that cannot penetrate the previous deposit; hence, a thin wall with lower EWW is produced, as shown in Figure 5-7(a). As-deposited built with LH of 0.8mm had EWW and TWW of 4.73mm and 5.09mm respectively, slightly higher than as-deposited built with LH of 1.2mm, which also had EWW and TWW of 4.70mm and 5.10mm respectively. Another contributing factor to the increased SW is the TS. As-

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deposited built with LH of 1.8mm and TS of 0.6m/min had EWW and TWW of 3.42mm and 3.98mm (thinner wall) as a result of faster TS, which gave a SW of 0.28mm

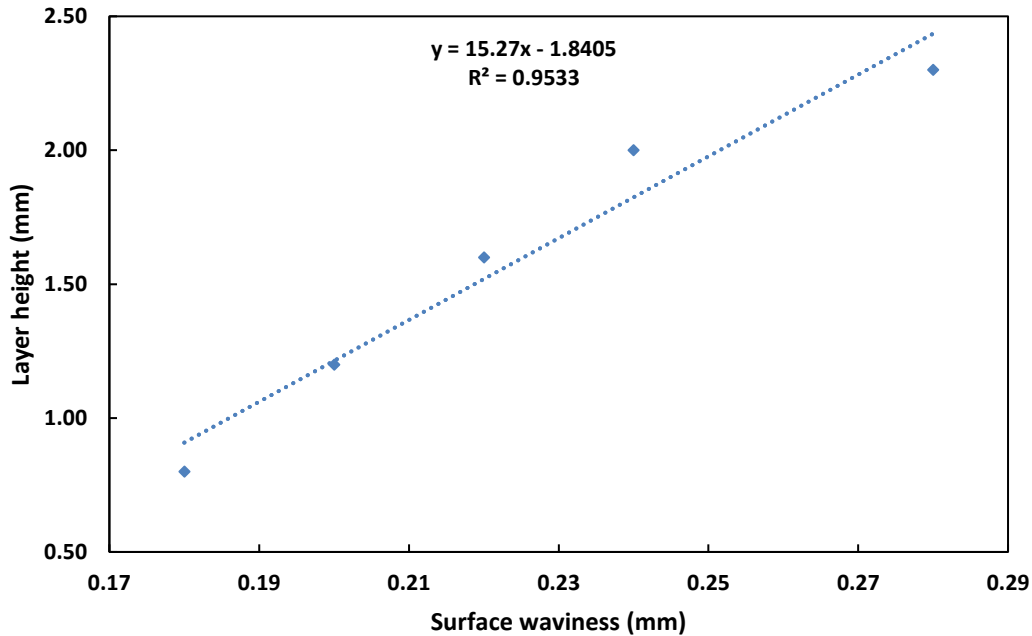


Figure 5-13 Effects of layer height on SW of ASD ER70S-6

Increasing the TS to 0.6m/min leads to increased SW as a result of the increase in arc instability and deposition of less volume of material per unit length. It was also noticed in work done by [55] that the increasing of TS beyond 0.42m/min leads to spatter formation and disturbed melt pool. As-deposited with LH of 2.0mm built with WFS of 6.0m/min had EWW and TWW of 6.68mm and 7.34mm with an SW of 0.33mm. Increasing WFS at constant TS of 0.4m/min leads to increasing SW as a result of the deposition of more volume of material per unit length due to the increase in arc pressure and droplet impact. The thermal cushioning effects between the wire and the previous deposition hinder the heat transfer to previous layers. Hence there is a low re-melting ratio [54] compared to the as-deposited condition with the same TS and LH of 0.8mm Figure 5-13 substantiates the variation in measured SW with the LH increment.

5.4.2 The influence of SW on Mechanical properties

Figure 5-14 presents the stress-strain curve for the machined and as-deposited conditions of ER70S-6 machined in the X-direction. The as-deposited condition showed lower UTS, YS, and % elongation (PCE) compared to the machined condition. The UTS, YS and PCE reduction factors, as a result of the as-deposited condition (with SW of 0.47mm and 0.18mm), are 22%, 23.6%, and 26.32% with 17.5%, 18.35%, and 21.05% respectively. Hence, higher mechanical properties could be achieved with the as-deposited condition built with a lower SW using lower LH, as justified by the tensile response of sample B (SW: 0.18mm). The ratio of the YS/SCF of the as-deposited samples were 159MPa and 120MPa as a result of the SCF effects when compared to the machined with YS of 367MPa. If the as-deposited SCF is applied to the machined condition in both case, there will be a YS reduction of 189MPa and 154MPa which is 18 & 28% more.

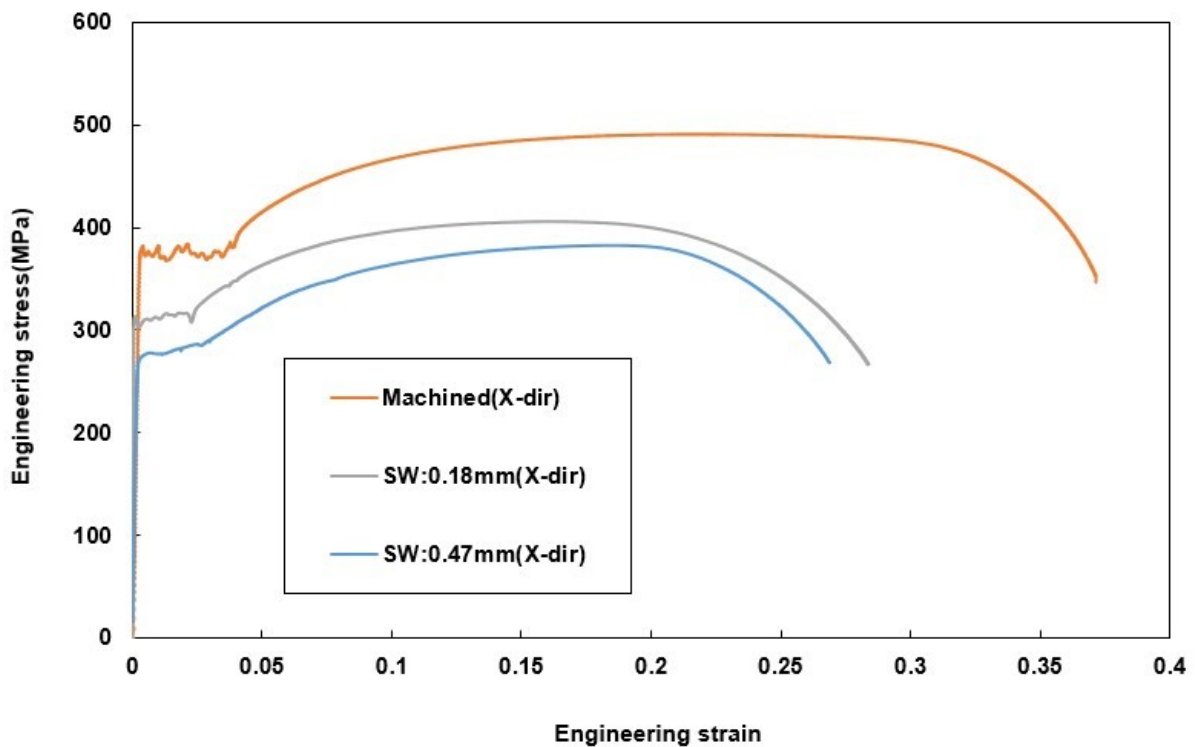


Figure 5-14 Stress-strain curve of the machine and ASD ER70S-6

5.4.3 Effect of Stress concentration and notch radius on SW

Figure 5-15 shows an explicit dependency of SCF on the notch radius. The SCF increases with smaller notch radius due to a relatively sharper change in geometric features between two consecutive layers. Figure 5-16 presents the variation of notch radius with SW of as-deposited WAAM, and Figure 5-17 shows the relationship between SCF and SW of as-deposited ER70S-6. The plot in Figure 5-16 indicates that the notch radius (ρ) increases as the SW reduces and this appears to follow a power law function ($\rho = 0.0603 SW^{-0.94}$) and is inversely proportional. The higher re-melting ratio produced by lower LH leads to a higher external angle between layers hence ($\beta_2 > \beta_1$), as shown in Figure 5-7. This is responsible for the larger notch radius. At low values of SW, the notch radius values are large, as shown in Figure 5-16. In Figure 5-17, SCF is shown as a linear function of SW and can be obtained using the equation $SCF = 4.7465 SW + 1.0913$.

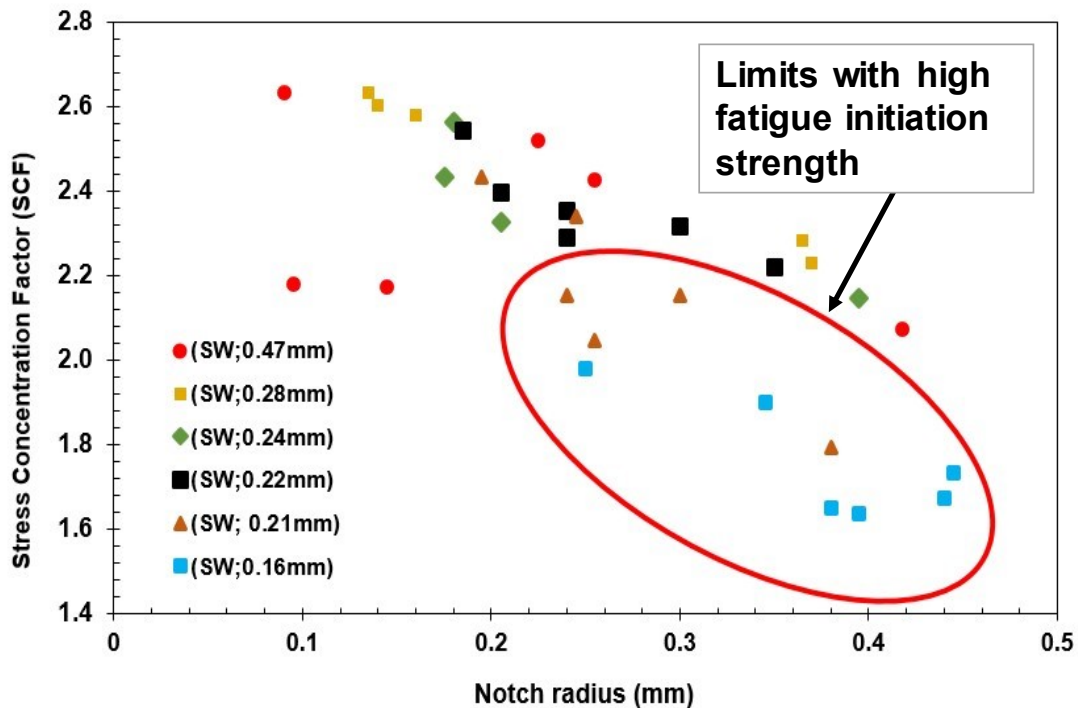


Figure 5-15 Dependence of SCF on notch radius

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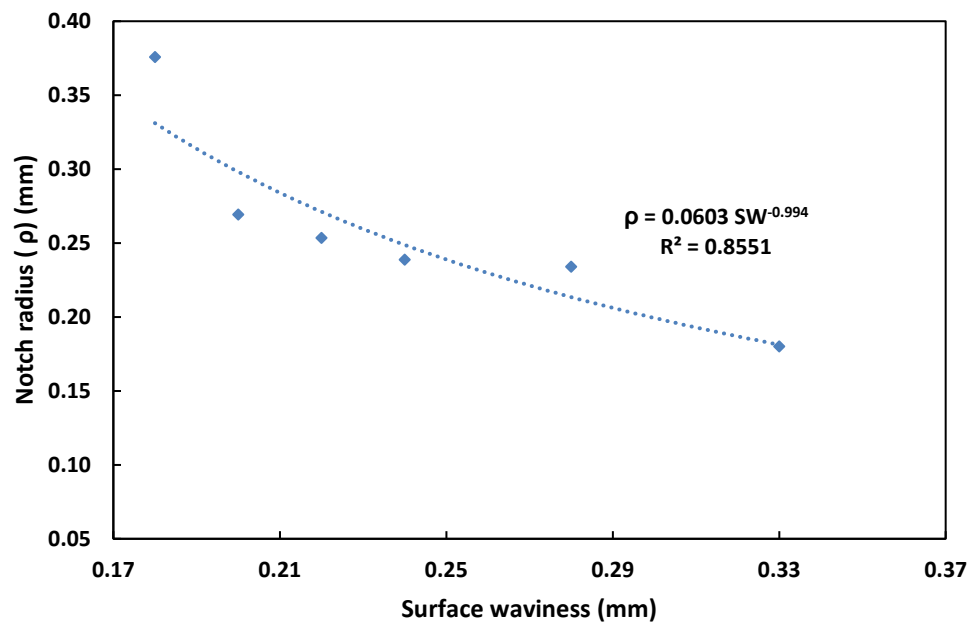


Figure 5-16 Changes in notch radius with variation in SW of ASD E70S-6

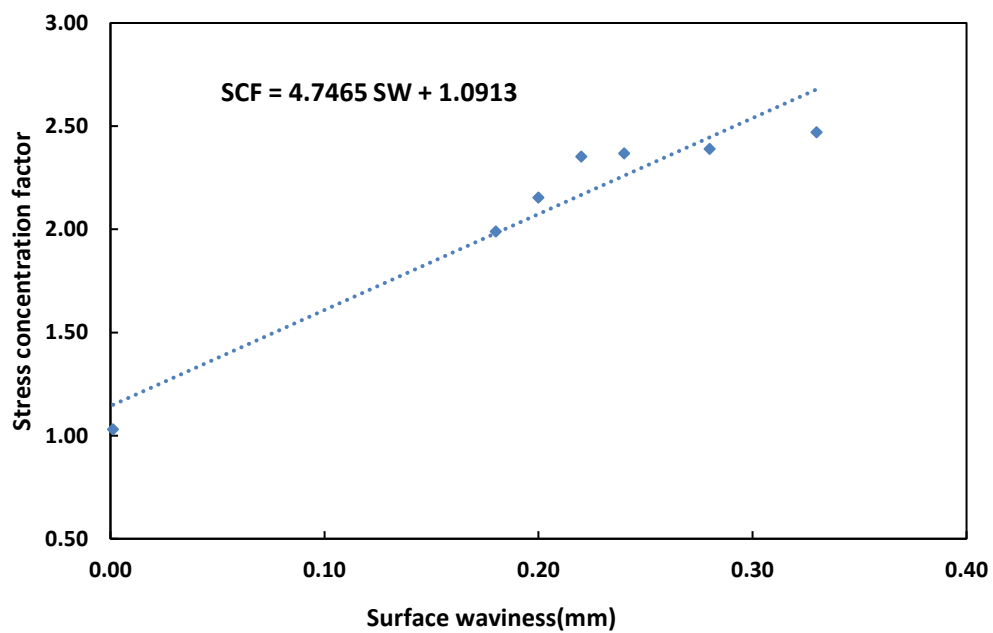


Figure 5-17 Relationship between SW of ER70S-6 and SCF

5.4.4 The effect of SW on the fatigue initiation life of as-deposited ER70S-6 at a constant stress level of 343.7MPa

Figure 5-18 presents the effects of SW on the fatigue life of as-deposited ER70S-6 at a constant stress (σ) level of 343.7MPa. It shows that the mean fatigue life of a WAAM as-deposited structure may be described using a power-law with fatigue data plotted on a log scale and SW plotted as a linear scale. This implies that a reduction in SW will greatly improve the fatigue life, which appears to agree with the claim of [126] who explained that the fatigue life of an AM is affected by surface features such as micro notches. The effect of reducing SW from 0.20mm to 0.18mm improved the fatigue life of the as-deposited sample by 37%. This type of chart will be of immense benefits to the industry to predict the functionality of a component with a defined SW before manufacturing.

Increasing the fatigue life of as-deposited mild steel by reducing SW entails optimising the process parameters to obtain low LH with optimum WFS and TS, which will produce heat input within the tolerance of the material being deposited. It is clearly shown that reducing SW by optimising the process parameters for a particular material produces a noticeable beneficial effect on the fatigue behaviour of WAAM as-deposited. However, the disadvantage of optimising deposition parameters to have lower SW, which gives a lower DR, cannot be ignored. There is a need to have a compromise between SW and DR to fully harness the benefits of WAAM. No doubt, determination of the SW and SCF helps to predict the life of as-deposited WAAM mild steel and to decide whether or not to machine to reduce cost.

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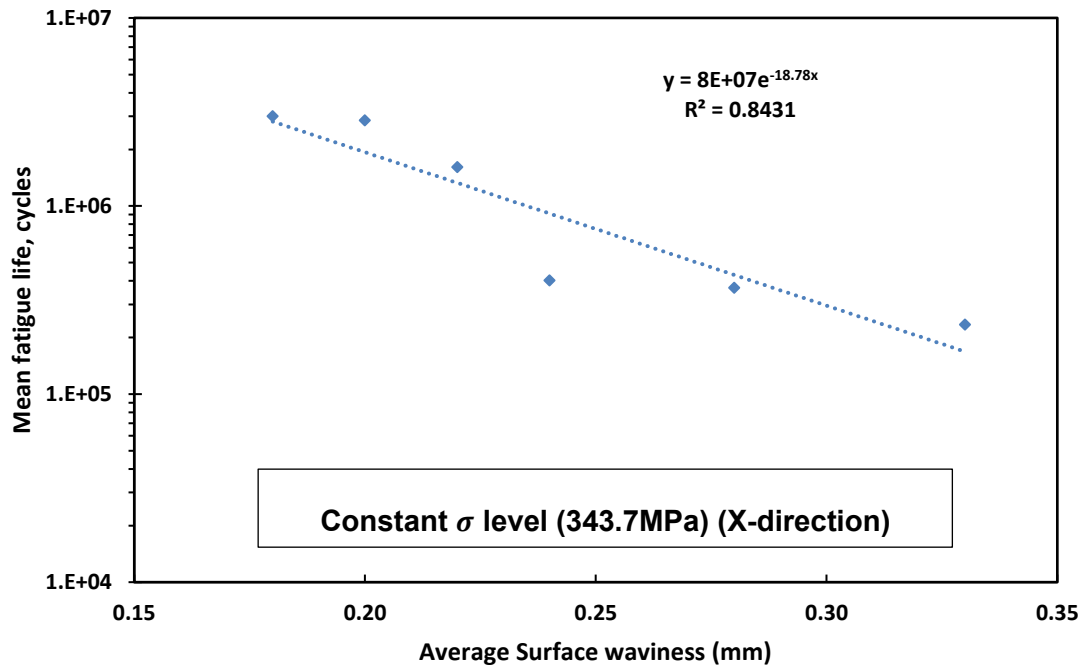


Figure 5-18 Effects of SW on fatigue life of ASD ER70S-6

5.4.5 Effect of SW and stress concentration factor on SIFR

Figure 5-19 presents the effects of SW on the SIFR of as-deposited ER70S-6, and Figure 5-20 shows the effects of SCF on the SIFR of as-deposited ER70S-6. With the maximum stress of 343.7MPa applied across the various samples with different SW, as shown in Figure 5-9, the driving force was higher for the as-deposited condition with SW of 0.47mm ($5.05\text{MPa}\sqrt{m}$), hence the lower fatigue life of 94,163 cycles obtained proves that higher local stress was established at the notch root and the value was larger than that of the lower machined condition. Hence, a higher driving force at the notch is responsible for the lower fatigue crack initiation life, and higher fatigue crack growth rate. The machined condition without SW showed a very low driving force of ($0.96\text{MPa}\sqrt{m}$) which leads to a higher fatigue life of 4,965,081 cycles. The plot of SW vs. SIFR curve is a power law, which shows an increase in SIFR with SW. Consequently, a lower SW of 0.18mm obtained with the as-deposited case leads to a relatively low driving force of $3.91\text{MPa}\sqrt{m}$ and a higher fatigue life of 3,002,003 cycles. Generally, SIFR

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increases as the SW increases. Figure 5-20 shows a linear relationship between SIFR and SCF.

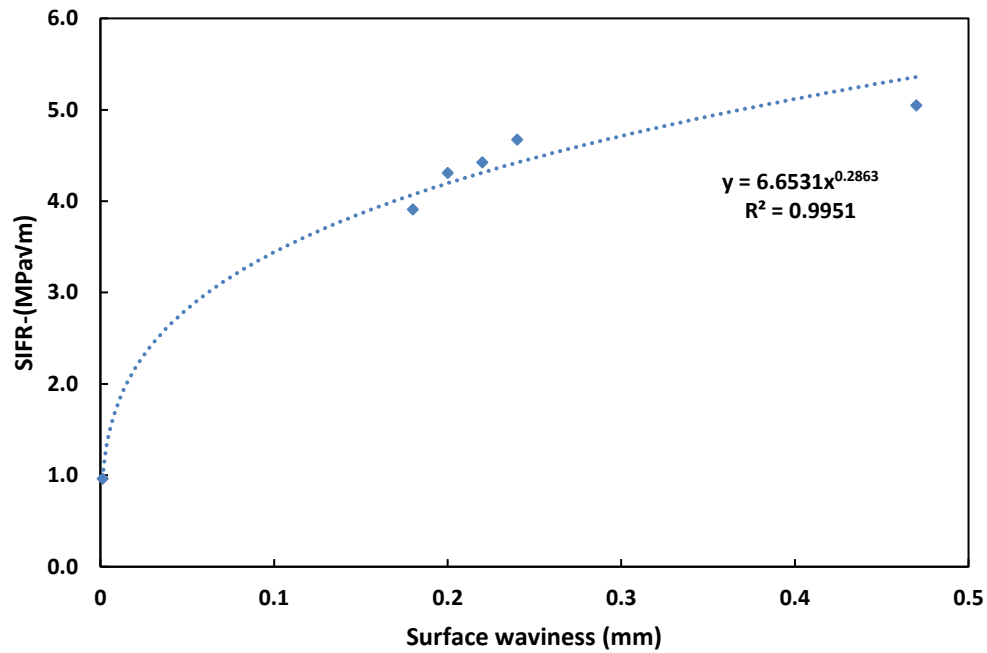


Figure 5-19 Effects of SW on SIFR of ASD ER70S-6

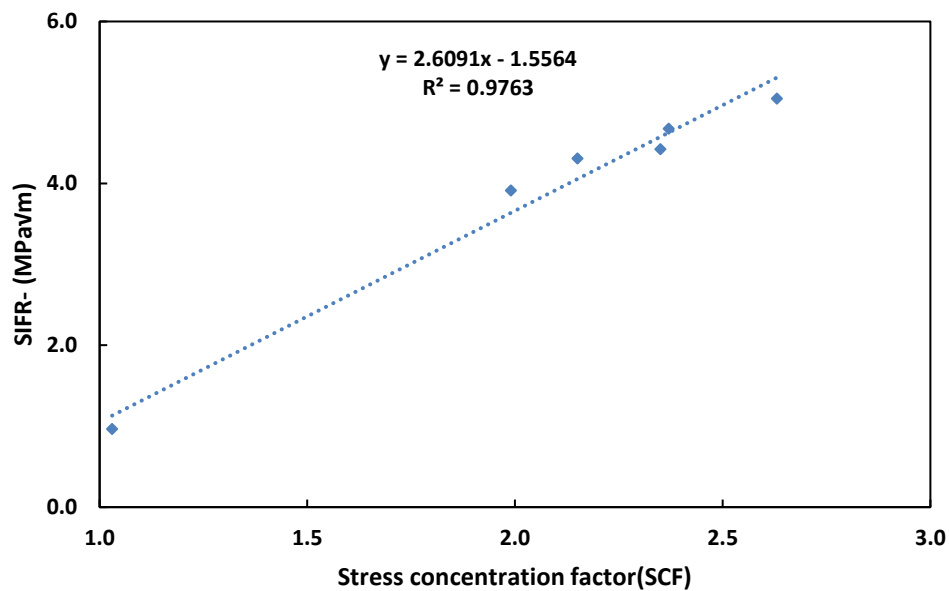


Figure 5-20 Effects of SCF on SIFR of ASD ER70S-6

5.4.6 The Effect of SW on WAAM deposition rate and Fatigue life at different stress levels

Figure 5-21 presents the effects of SW on the DR. Figure 5-22 shows the fatigue responses of trial C specimen (R_a : $1.32\mu\text{m}$) machined condition, trial B specimen (R_a : $32.1\mu\text{m}$ & SW: 0.18mm) and trial D specimen (R_a : $100.6\mu\text{m}$ & SW: 0.47mm) as-deposited condition, established at different stress levels using 60 - 90% of the machined UTS value to build the S-N curve. Fatigue loading was done, as shown in Figure 5-6 (a) (Z- & X-direction). It is clear from Figure 5-21 that WAAM process parameters such as WFS and WD, which are responsible for increasing DR directly, affect the SW and R_a of the final built structure.

Increasing the WFS produces more volume of materials per unit length as a result of an increase in arc pressure and droplet impact (molten pool instability) which leads to increasing SW & R_a ; this is also heightened with higher WD. In Figure 5-22, the effects of increasing SW on the fatigue response of as-deposited as a result of increased WD are obvious, as SW of 0.47mm deposited with WD of 1.2mm (DR = 3.40Kg/hr) performed less well compared to SW of 0.18mm deposited with 1.0mm WD (DR= 1.45Kg/hr). Hence WAAM built parameters which control DR, and SW are responsible for the fatigue behaviour of as-deposited structures.

The machined condition (R_a : $1.32\mu\text{m}$) performed better than the as-deposited conditions due to the absence of SW. The as-deposited condition with R_a : $32.1\mu\text{m}$ & SW: 0.18mm performed better than the as-deposited condition with R_a : $100.6\mu\text{m}$ & SW: 0.47mm due to the severity of SCF at higher SW & R_a . It is worth noting from Figure 5-20 that fatigue loading in the Z-direction, Figure 5-6(a), for the as-deposited presented the worst fatigue life for both rough surfaces because the crack propagation rate is more severe when the SW is orientated perpendicularly to the dynamic loading direction.

Loading in the X-direction, as shown in Figure 5-6(b), is more superior to fatigue life as the crack is less reactive when the SW is orientated parallel to the dynamic load. At a maximum stress level of 393MPa , the fatigue life of the as-deposited (R_a : $32.1\mu\text{m}$ & SW: 0.18mm) machined in X- & Z-directions was 65 & 288 times

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better than the fatigue life of the as-deposited with (Ra: 100.6 μ m & SW: 0.47mm). The machined condition survived the same stress level at higher cycles before failure. The fatigue response of WAAM as-deposited parts is controlled by the WAAM deposition parameters and can be improved by reducing the SW and Ra by using low LH and DR (WFS & WD).

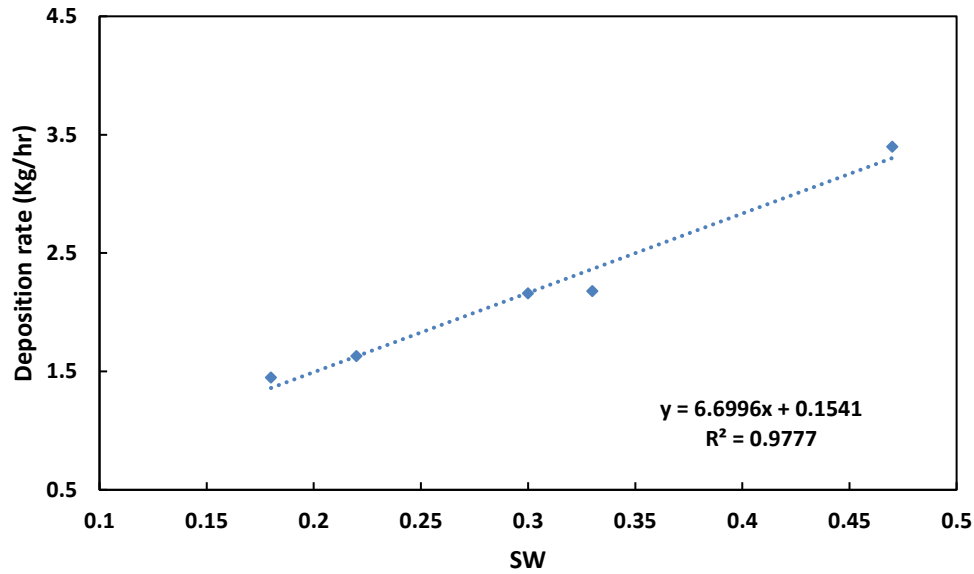


Figure 5-21 Effects of SW on the deposition rate of single pass bead

The effective stress concentration factor (S_{eff}) was evaluated experimentally from the fatigue failure curves. According to [127], (S_{eff}) is expressed as

$$S_{eff} = \frac{\text{Failure stress of the machined specimen}}{\text{Failure stress of the as – deposited specimen}} \quad (5-5)$$

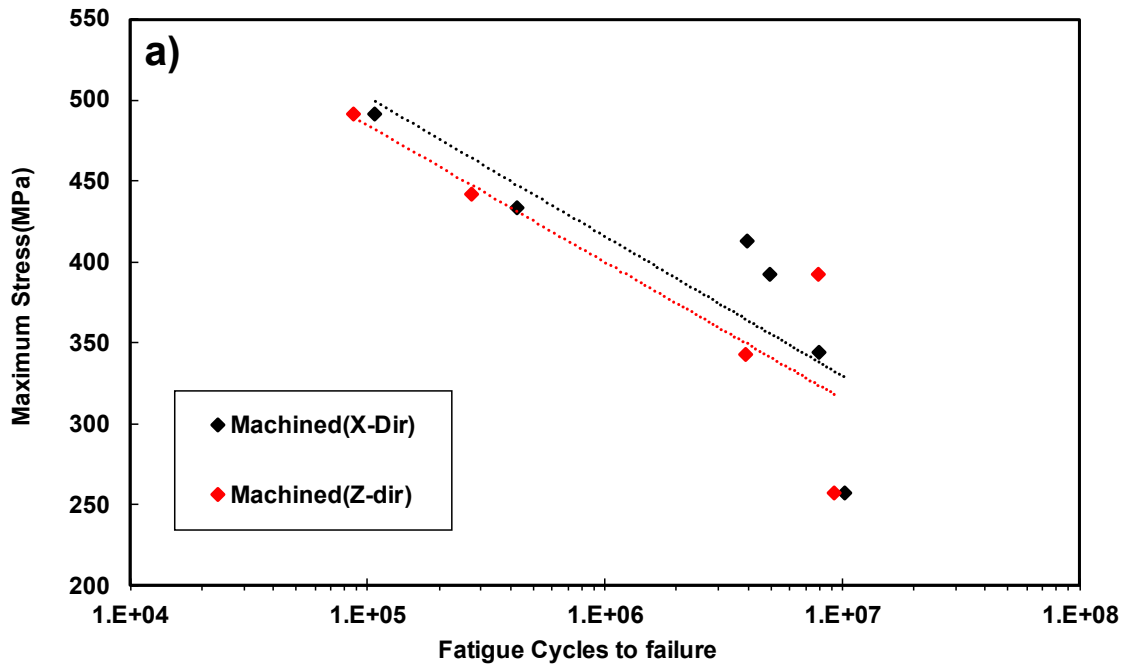
The average S_{eff} evaluated from Figure 5-22 for the as-deposited condition for trial B (Ra: 32.1 μ m & SW: 0.18mm) and D (Ra: 100.6 μ m & SW: 0.47mm) specimens is 1.29 and 2.33 respectively. It can be inferred from this study that the SW affects both the elastic and effective stress concentration factors. The

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average S_{eff} value (1.29 & 2.33) obtained in Figure 5-22 for as-deposited with (Ra: 32.1 μ m & SW: 0.18mm) & (Ra: 100.6 μ m & SW: 0.47mm) is less than the average S_{el} value (1.99 & 2.63) calculated in Table 5-4. The notch sensitivity index [79] is expressed as,

$$q = (S_{eff} - 1) / (S_{el} - 1) \quad (5-6)$$

Estimated q indices for the two SW values (0.18mm & 0.47mm) for trials B and D are 0.29 and 0.82, respectively. The value of q for the as-deposited condition is a function of the SW. The as-deposited having lower SW (0.18mm) showed a q improvement of 64.6%, compared with the as-deposited having an SW of 0.47mm. This implies that notch parameters due to SW are less severe as SW or Ra tends to zero. It is likely that a sensitivity index close to zero may be obtained by optimising the SW parameters.



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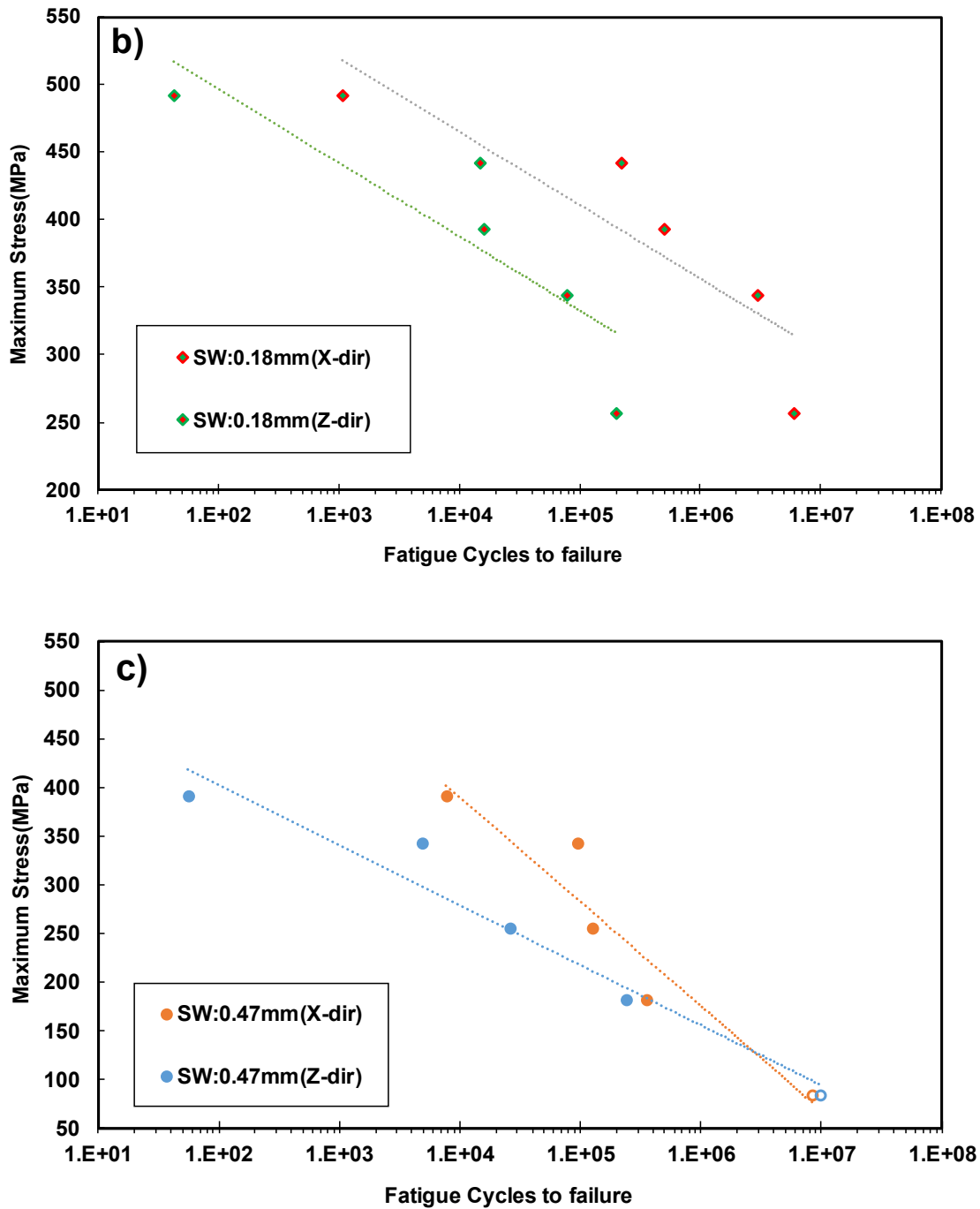


Figure 5-22 Effects of SW on fatigue life X Vs Z direction (ER70S-6)

(a) Machined condition (b) As-deposited (SW: 0.18mm) (c) As-deposited (SW: 0.47mm)

5.4.7 Fracture morphology and details

Fractographs of the fatigue fracture surface of the machined ($R_a: 1.32\mu\text{m}$) and as-deposited condition ($R_a: 32.1\mu\text{m}$) fractured in the Z-direction and X-direction

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are presented in Figure 5-23. An examination of the fractured surfaces using LYRA3 TESCAN revealed that the failure in the machined condition, which exhibited higher fatigue life (4,965,081 cycles) at 393MPa, was as a result of the gas pore. The crack initiation mechanism/path in the machined condition is from the weakened path formed by the nucleation of spherical pores; perhaps introduced during WAAM processes, as shown in Figure 5-23(a), the effects of SW have been eliminated by machining.

The fatigue failure mechanism is by ductile striation; the striations are believed to be the distance moved by the crack per cycle [128]. The various as-deposited conditions have essentially the same microstructural features. The significant reduction in fatigue life for the as-deposited sample is primarily due to the presence of notches (SW) which act as stress raisers. The cracks initiated at the valley (pit) of the SW and propagated towards the centre of the specimen. At a critical length, where the remaining sample ligament could no longer support the increased stress, then plastic collapse occurred. The crack initiation mechanism and path in the as-deposited condition is from the weakened valley with the maximum stress concentration between two layers that form the SW profile (point with the sharpest notch radius and large notch depth).

The radius and depth of the notch (SW) is a major contributing factor to the severity of the stress concentration factor. The closer the factor to unity, the better the functionality of the surface. The Z-direction fracture exhibited lower fatigue life compared to the X-direction fracture because SW across the layers serves as a plastic incompatibility region where cracks grow parallel to the SW and perpendicular to the applied load at a faster rate. At a σ_{max} of 442MPa, the X-direction failed at 495,119 cycles while the Z-direction failed at 15,894 cycles, as shown in Figure 5-23(c & d). In the absence of process inherent defects, fatigue performance of the as-deposited structure is dominated by crack initiation from the weakest point on the structure.

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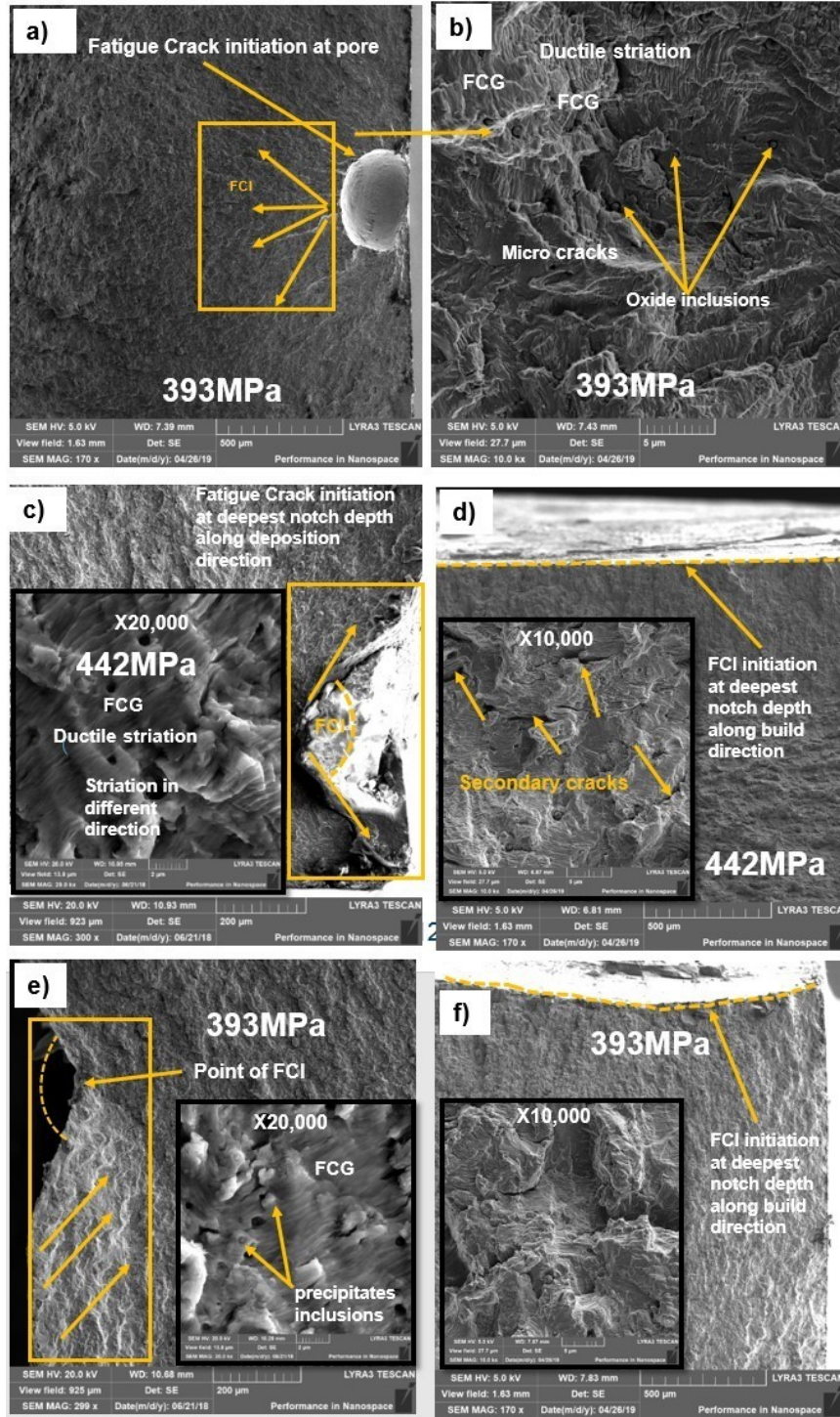


Figure 5-23 SEM fractographs of WAAM ER70S-6

(a-b) Machined condition, $\sigma_{max} = 492MPa$, $N_f = 107,698$ cycles, as-deposited condition ($Ra:32.1\mu m$) (c) (X-direction): $\sigma_{max} = 442MPa$, $N_f = 318,145$ cycles (d) (Z-direction): $\sigma_{max} = 442MPa$, $N_f = 14,576$ cycles, (e) (X-direction): $\sigma_{max} = 393MPa$, $N_f = 495,119$ cycles, (f) (Z-direction): $\sigma_{max} = 393MPa$, $N_f = 15,894$ cycles

5.4.8 Conclusions

In this chapter, the effects of WAAM process parameters on SW of as-deposited structures, and the effect of SW on the dynamic mechanical properties of the same structure have been investigated. The influence of machining on the fatigue life of the same components was also evaluated. From this investigation, it was shown that:

1. The fatigue response of WAAM as-deposited parts is controlled by the WAAM deposition parameters. Low layer height and wire feed speed, with small wire diameter and moderate deposition rate, improve the surface waviness and hence the stress concentration factor. The loading direction of the WAAM as-deposited structures is paramount to the fatigue initiation life.
2. The larger notch radius achieved with lower layer height is effective in reducing both the elastic and effective stress concentration factor which, in turn, enhances the fatigue initiation strength of the as-deposited condition.
3. The fatigue initiation life of the WAAM as-deposited structure varies with SW by a power law, and layer height varies linearly with surface waviness, which is detrimental both to the mechanical and fatigue initiation strength of the as-deposited WAAM HSLA structure.
4. The usefulness of surface waviness and stress concentration factor determination helps to predict the life of the as-deposited WAAM built structure and whether to machine, and the extent to machine to save cost.
5. The fatigue damage mechanism for the WAAM as-deposited component was through crack initiation at points of stress concentration (stress raisers) due to surface waviness, growth, and the coalescence of micro voids around inherent processing defects or second phase particles and then plastic collapse of the remaining sample ligament that could no longer support the increased stress.
6. Surface waviness is not as bad as such, but the geometry of the deposited structure must be controlled.

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7. For fatigue sensitive components, there is an economic balance between deposition rate and fatigue initiation life. Or post-weld machining is recommended.

The investigation in the present chapter displayed that WAAM process parameters are responsible for the fatigue initiation behaviour of WAAM as-deposited structures and that reduction of surface waviness by process parameter optimisation will greatly improve the fatigue life of as-deposited WAAM components. Since surface waviness is directly influenced by the deposition rate, low surface waviness can only be achieved with a low deposition rate in the single pass strategy. This will defeat the whole essence of cost-saving with the deposition of bespoke components in the as-deposited condition because the company will be faced with low productivity.

Chapter 6 will investigate an alternative cold work process that can be combined with WAAM high deposition, knowing fully well that it leads to high SW, to reduce the effect of SW on the fatigue initiation strength of the single-pass WAAM as-deposited structures, hence still meet its requirements for use in as-deposited condition and high productivity gains.

6 EFFECT OF SURFACE ROLLING METHODS ON WAAM PROPERTIES

6.1 Introduction

All the proposed techniques and studies for improving the SW of WAAM components were only able to minimise but not eliminate the entire effects. SW, which acts as a stress raiser, reducing the tensile and fatigue life initiation of as-deposited WAAM component, is still a major concern to the industry. The process parameter optimisation method of reducing SW (as studied in Chapter 5) is limited by low deposition rate (low productivity), and this refutes the intents of using WAAM as a low-cost AM process. To fully harmonise and harness the benefits of the WAAM process, there should be a compromise between SW, productivity, and machining. Gains from high productivity should cover the machining cost and make a break-even point for the industry. The high price of jigs and fixtures for machining, coupled with the cost of cutting tools, is also limiting the use of machining as a support to the high productivity advantages of WAAM. Hence there is a need for a competitive cold working process that will enhance the SW reduction while maintaining a high productivity level.

To the best knowledge of the author, there is no current work showing the effects of the reduction in SW by the introduction of compressive stresses by rolling on a WAAM built component, and how this affects its mechanical and fatigue initiation strength when used in the as-deposited condition.

The first objective of this thesis was to evaluate the effectiveness of cold side rolling and top interpass rolling in reducing the SW.

The second and last objectives are to evaluate the effect of compressive residual stress on the mechanical and fatigue initiation strength of the as-deposited plus rolled component. A correlation with their microstructural variation will also be made.

6.2 Experimental procedure

A standard cold metal transfer (CMT) welding process using a Fronius CMT (VR 7000 CMT) power source, as presented in Figure 3-1, was used to deposit six WAAM structures of ER70S-6 steel with a dimension of 245mm x 120mm x 5.0mm, and the SW of the deposited structure evaluated as per Eq. (5-1) is 0.18mm. Both sides of the deposited walls were rolled with a constant travelling TS of 500mm/min parallel to the welding direction (X-direction) with different loads varying between 50 and 160kN applied on different wall specimens, using a small flat roller with an effective width of 20mm and a diameter of 120mm. The set-up and testing rig are shown in Figure 3-2. The wire composition and deposition parameters are shown in Table 3-1 and Table 6-1.

The WAAM structures in as-deposited and rolled-condition were transverse cross-sectioned for metallographic analysis, and micro-hardness measurement was done as per Sections 3.4.1 and 3.4.2. Micro and macro images of the etched surface were obtained using an optical microscope and deployed to the Axio Vision software to reveal the SW parameters (effective wall width, total wall width, and notch radius). The fracture surfaces were characterised, as discussed in Section 3.4.3. The X-ray diffraction (XRD) patterns of WAAM as-deposited (70 - A) and WAAM as-deposited + rolled (160KN) were characterised as discussed in Section 3.4.6. All tests were done at ambient temperature. The tensile test was performed in accordance with Section 3.5.1 on all tensile samples, as given in Table 6-2. A gauge length of 32mm was used to calculate the elongation. A strain rate of 1mm/min was used.

The fatigue test was performed in accordance with Section 3.5.3 on all fatigue samples, as given in Table 6-3. A tension - tension fatigue test was carried out with a frequency of 10HZ and a stress ratio = 0.1 using the machine equipped with a 250kN load cell. The fatigue response of the different conditions was established at different stress levels to build an S-N curve. The test programme was set up to compare the mechanical properties between the three fatigue conditions as detailed in Table 6-3. These samples were subjected to the same

Effect of Surface Rolling Methods on WAAM Properties

loading conditions as in-service. The build orientation schematic with tensile and fatigue testing geometry is showed in Figure 6-1.

Table 6-1 WAAM deposition parameters

Parameters	Values
CTWD	11.5 mm
Wire diameter	1.0 mm
Shielding gas	80% Ar + 20% C_{O_2}
Flow rate	15 L/min
WFS	4.0 m/min
TS	0.4 m/min
WFS/TS	10
Layer height (LH)	1.6 mm

Table 6-2 Tensile samples

Tensile sample condition	Surface waviness (mm)	Number of samples
As-deposited	0.18	3
Machined	N/A	3
As-deposited + rolled	0.08, 0.13, 0.17	9
As-deposited + rolled + machined	N/A	3

Table 6-3 Fatigue samples

Fatigue samples	Surface waviness (mm)	Number of samples
As-deposited	0.18	5
Machined	N/A	6
As-deposited + rolled	0.08	3

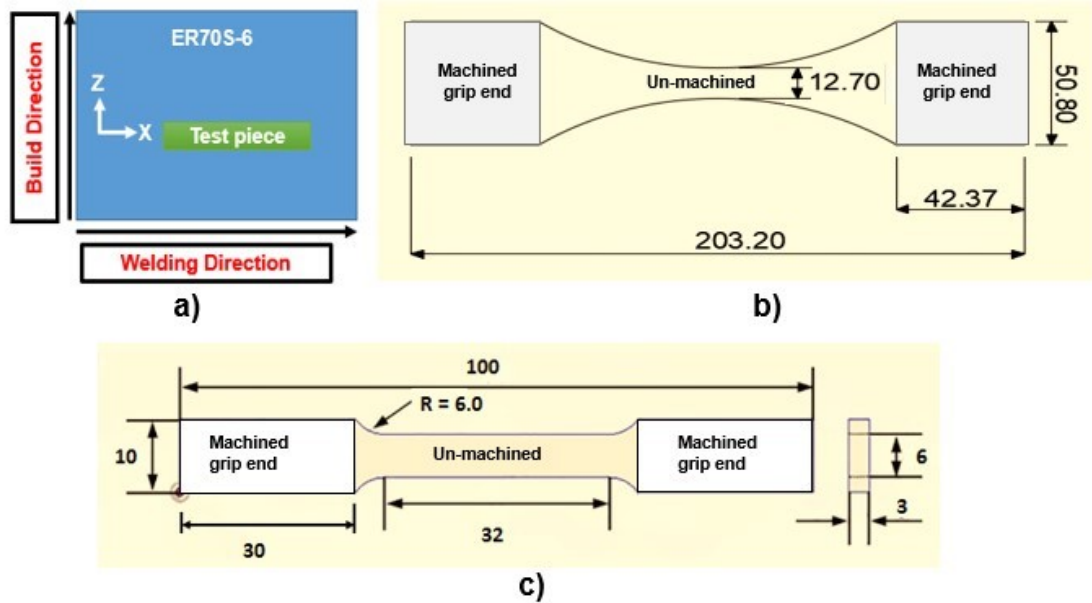


Figure 6-1 Deposition Matrix

(a) Build orientation (b) Fatigue testing geometry (c) Tensile testing geometry

6.3 Results

6.3.1 Surface waviness, stress concentration factors (as deposited and rolled)

Figure 6-2 presents the macros of the as-deposited + top rolled WAAM deposit, while Figure 6-3 presents the macros of the as-deposited + side rolled WAAM structures. In top rolling, TWW increases as the applied load increases due to the effect of bulk material displacement under compressive stress. This bulk material displacement in the case of the top rolling is not suitable for SW reduction, as seen in Figure 6-2, where 50kN & 75kN load application increased the SW from 0.22 – 0.26mm. Top rolling increases the deposited width and reduces the deposited height. This is quite unlike the side rolling methodology, where the rolling load reduces the deposited width and increases the deposited height. Side rolling is more effective in reducing the SW and compelling the deposited layers to conform to the shape of the roller. Side rolling also shows distinctive layer bands, as shown in Figure 6-3(f). The subsequent analysis was done with side rolling.

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The TWW, as shown in Figure 5-6(d), decreases due to lateral elongation of the structure as seen in Figure 6-3, whereas the SW decreases from 0.18 – 0.008mm as the load increases. Figure 6-4 presents the side rolled surface profile, and Figure 6-5 presents the variation in SW and SCF of both conditions and the relationship between rolling load P (kN) and SW (mm). The measured SW of the rolled WAAM samples, as shown in Figure 6-5, revealed that the SW decreases with the increase in the rolling load. The measured SW of the as-deposited WAAM sample was 0.18mm without rolling; this further reduces to 0.08mm with a load of 160kN. The peaks of the as-deposited samples are reduced as the rolling load increases because of plastic deformation, as shown in Figure 6-4(b-d). The SCF is calculated using the relationship in Eq. (5-3)

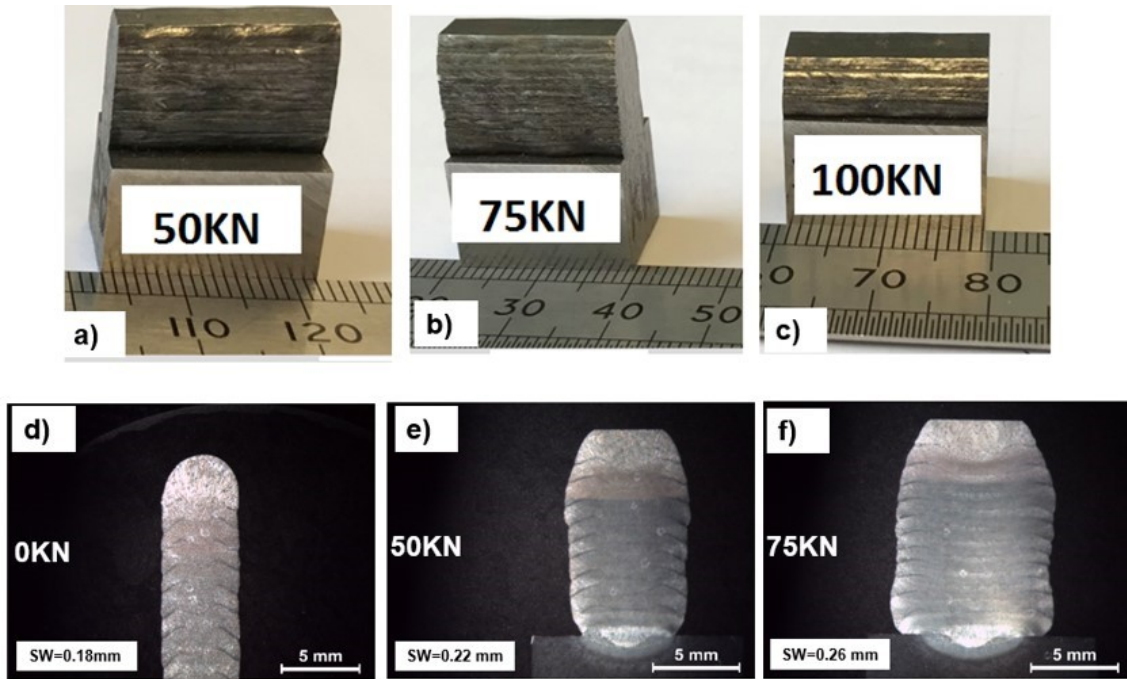


Figure 6-2 Macros showing the effects of top rolling on as-deposited SW

(a) 50kN (b) 75kN (c) 100kN (d) 0kN (e) Macro-50kN (f) Macro-75kN

Effect of Surface Rolling Methods on WAAM Properties

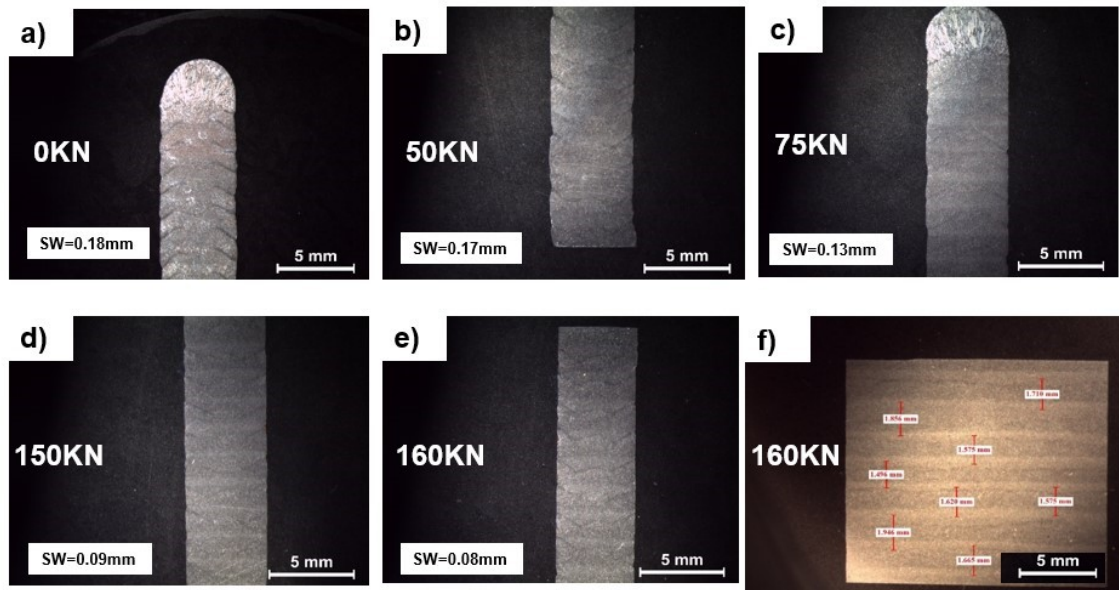


Figure 6-3 Macros showing the effects of side rolling on as-deposited SW

(a) 0kN (b) 50kN (c) 75kN (d) 150kN (e) 160kN (f) 160kN – layer bands

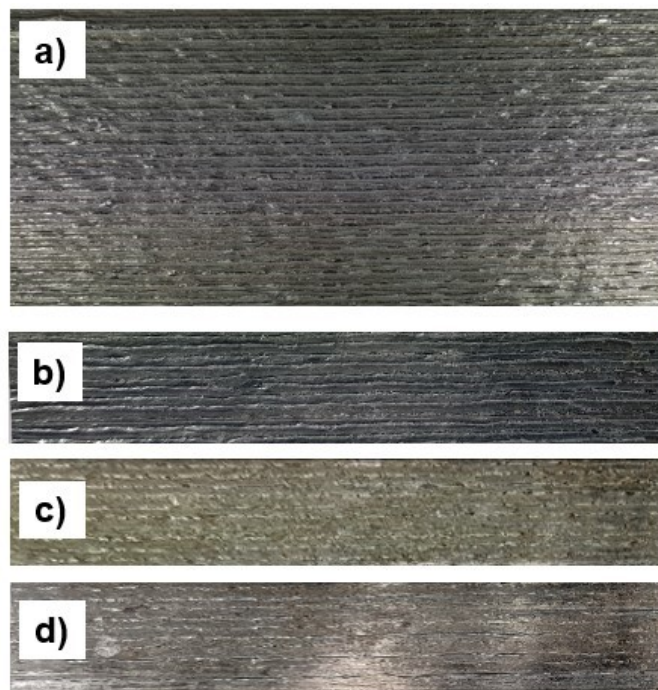


Figure 6-4 Surface profile

(a) ASD (b) ASD + Rolled (75kN) (c) ASD + rolled (150kN) (d) ASD + rolled (160kN)

Effect of Surface Rolling Methods on WAAM Properties

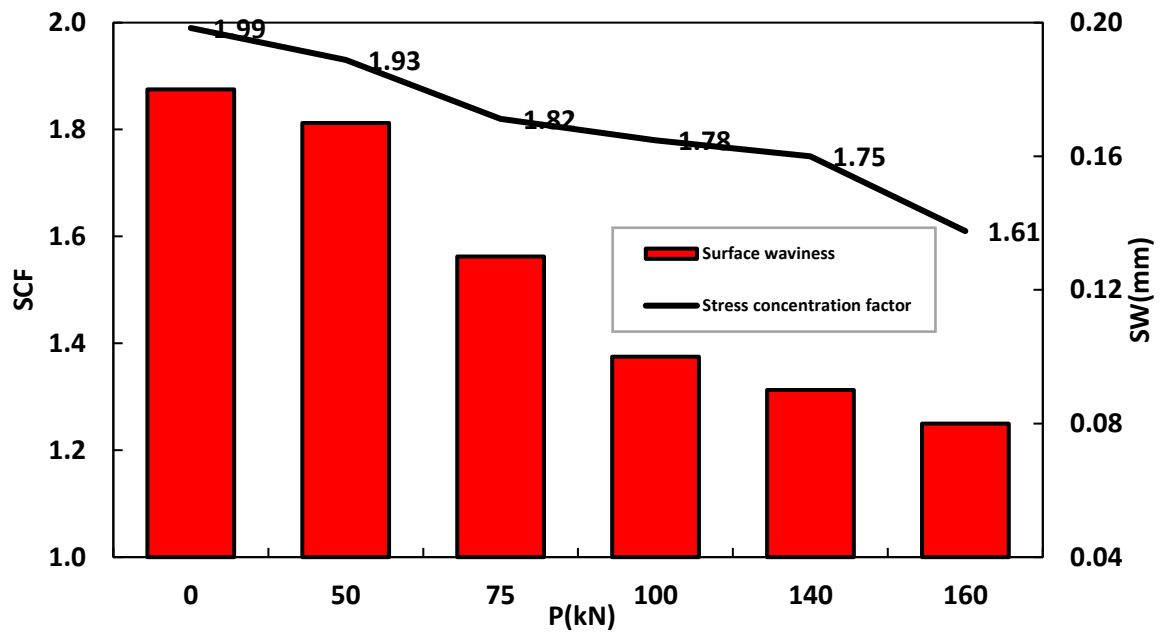


Figure 6-5 Variation in SCF and SW of WAAM ASD and side rolled condition

6.3.2 Monotonic properties of side rolled - tensile strength

Figure 6-6 presents the stress-strain curves obtained from the tensile sample in Table 6-2. The UTS and YS values are on the increase as the rolling load increases. It is important to note that the as-deposited condition with SW of 0.18mm yielded the least strength values. The as-deposited + rolled + machined condition without any SW showed the highest UTS and YS with the PCE, as shown in Figure 6-6, so it is evident that SW reduces the static properties. The interrupted yielding phenomenon called the Lüders band elongation tends to disappear due to work hardening rate effects incurred from the plastic deformation process [113,114]. This interrupted yielding decreases with the rolling load, as seen in Figure 6-6. The as-deposited + machined condition showed higher PCE compared to the as-deposited + rolled. The interaction plot for rolling load versus UTS and SW is shown in Figure 6-7.

Effect of Surface Rolling Methods on WAAM Properties

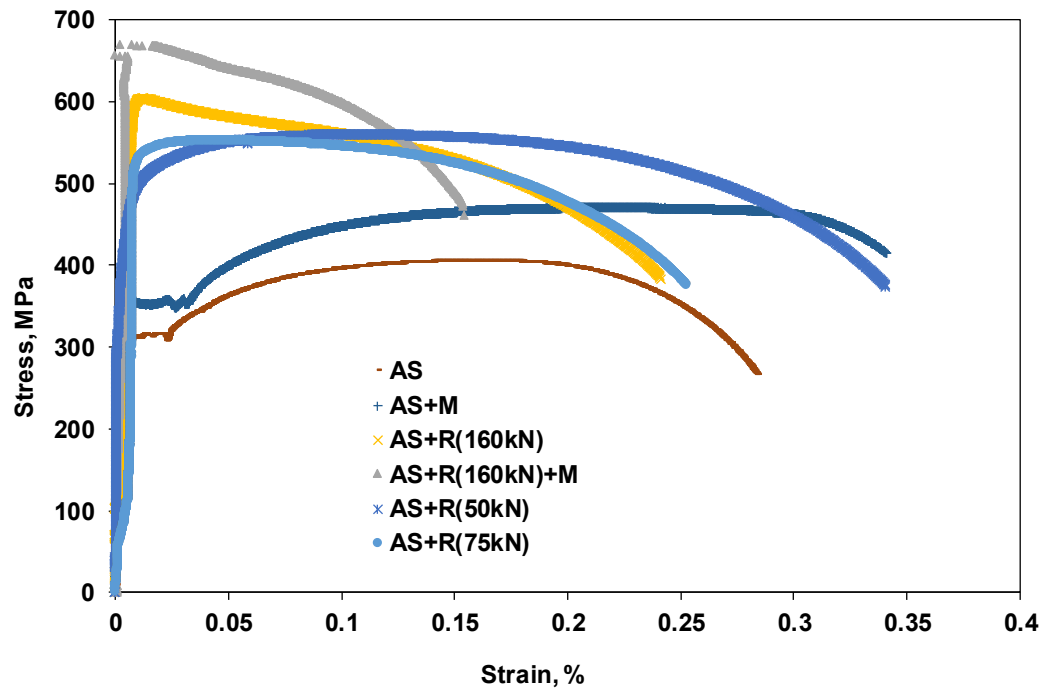


Figure 6-6 Stress-strain curve of various conditions of WAAM ER70S-6

Additional stress-strain curves are found in appendices (Fig.8)

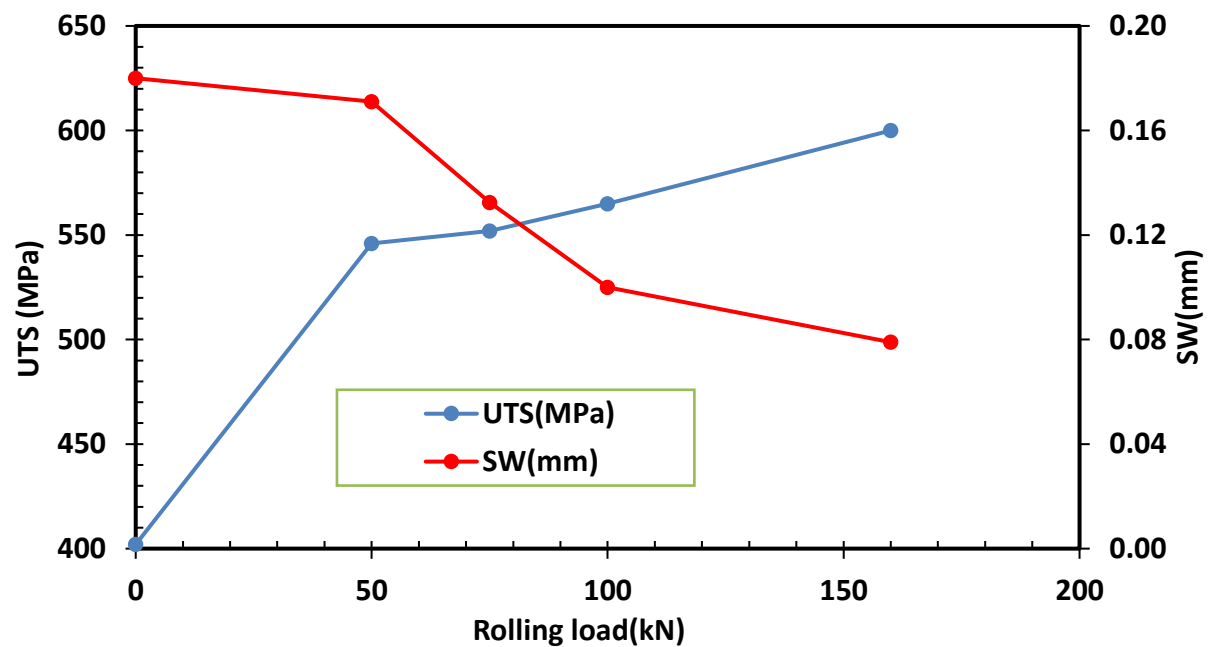


Figure 6-7 Interaction plot for rolling load vs. UTS and SW

6.3.3 Microstructural and hardness variation

6.3.3.1 Top rolling

The hardness maps for the various top rolled load conditions are shown in Figure 6-8. The SEM images showing the effect of rolling on the microstructure are shown in Figure 6-12. The top rolling has more impact on hardness, as seen in Figure 6-8(a & b) as a result of the microstructural changes enforced by the plastic deformation. Top rolling results in increased dislocation density within the microstructure by recrystallising the LAGs to large numbers of HAGs, as shown in Figure 6-12. The degree of microstructural changes taking place in the rolled samples is commensurate with the rolling load.

Top rolling is more effective in microstructural refinement than side rolling. The side rolled experience grain elongation in the rolling direction as shown in Figure 6-11(a-d) and Figure 6-12(a-b), and there is no grain refinement as such. There is strength and hardness strengthening from the grain size reduction in top rolling; this is demonstrated in Figure 6-8(a & b) where the average hardness for 75 & 50kN top rolled is 330 & 310HV respectively. This is a 46 & 41% increase from the side rolled with the same load. Rolling has a dual effect of reducing the grain sizes in the rolled direction and increasing the hardness. Top rolling increases the EWW, but is not effective in reducing the SW as seen in Figure 6-2; hence, further study with side rolling was undertaken.

6.3.3.2 Side rolling

The hardness versus depth plots across the through thickness direction for both as-deposited and the various rolled conditions are shown in Figure 6-9. The SEM images showing the effect of rolling on the microstructure are shown in Figure 6-10, Figure 6-11 and Figure 6-12. The average hardness of the as-deposited + rolled specimens increases with the rolling load. The 160kN samples show the maximum hardness of $233 \pm 1.5\text{HV}$ at both rolled surfaces, while the 50 and 75kN samples show maximum hardness of 225 ± 1.0 and $219 \pm 1.5\text{HV}$ at both rolled surfaces, respectively. Both as-deposited + rolled conditions showed minimum hardness at the midpoint of the sample.

Effect of Surface Rolling Methods on WAAM Properties

The as-deposited condition showed a maximum hardness of $186 \pm 2.0\text{HV}$, which is 25.26% lower than the maximum obtained hardness obtained with 160kN rolled condition. The increase in hardness with the rolling load is a result of more dislocation density being generated by the higher rolling load, which work-hardens the material [129], as exemplified in Figure 6-6. In Figure 6-10, the as-deposited has an average grain size of $14.5\mu\text{m}$, which is bigger than the as-deposited + rolled in Figure 6-11 and the top rolled in Figure 6-12, which had grain sizes decreasing as the rolling load increased.

The 160kN side rolled sample had elongated grains with an average size of $8.6\mu\text{m}$ while the 75kN top rolled has equiaxed grains with a size of $5.7\mu\text{m}$. Rolling plastically deforms the surface, increases the dislocation density and elongates the grains. In the 75 and 50kN side rolled condition, rolling load, the roller contact was not sufficient, and the middle portion was not work hardened. It has also been reported that rolling caused austenite nucleation and provided further grain size reduction [106]. In this study, rolling plastically deforms the peaks of the SW, shown in Figure 5-6(c & d), into surfaces, shown in Figure 6-4, hence reducing the EWW and the SW as well. This is also in support of the claim by Colegrove et al. [106] and Hönnige et al.[130]. The plastic deformation induced on the surface, which is a function of the rolling load reduces not only SW as observed here but also the residual stress. Plastic deformation by rolling has been proven to improve the mechanical properties of WAAM deposits [57].

Effect of Surface Rolling Methods on WAAM Properties

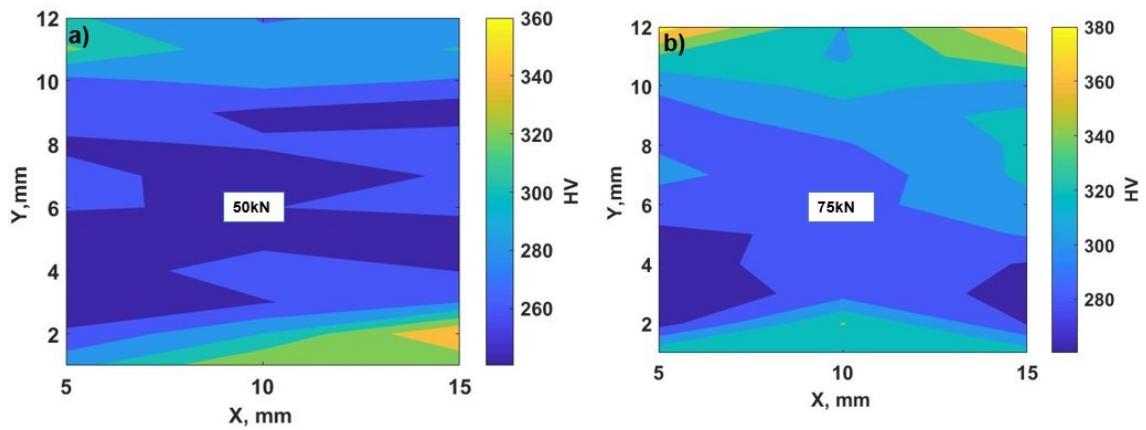


Figure 6-8 Effect of rolling load (kN) on hardness variation of top rolling

Rolling load (a) 50kN (b) 75kN

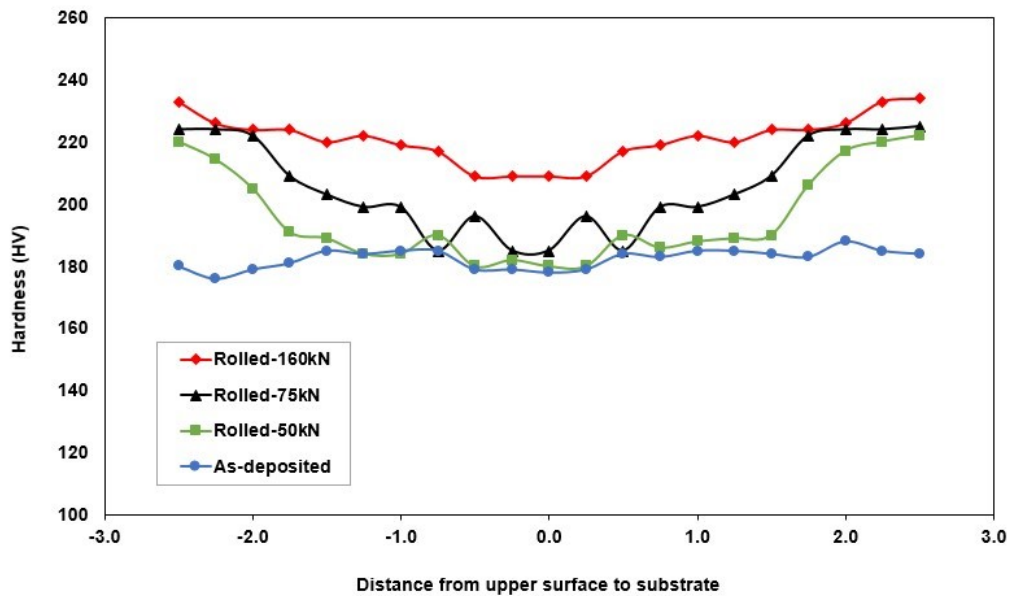


Figure 6-9 Effect of rolling load and hardness variation across the through thickness direction (side rolling on both faces)

Effect of Surface Rolling Methods on WAAM Properties

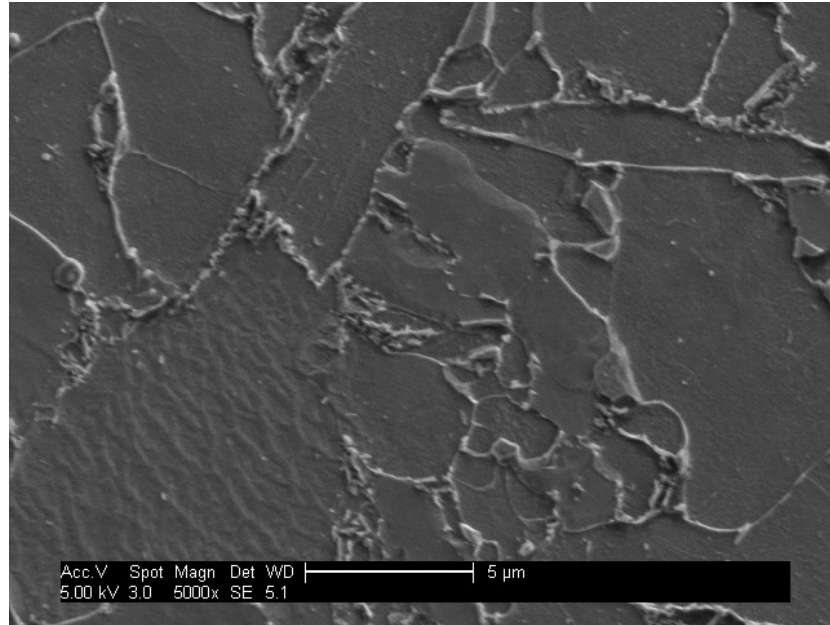


Figure 6-10 SEM image of as-deposited

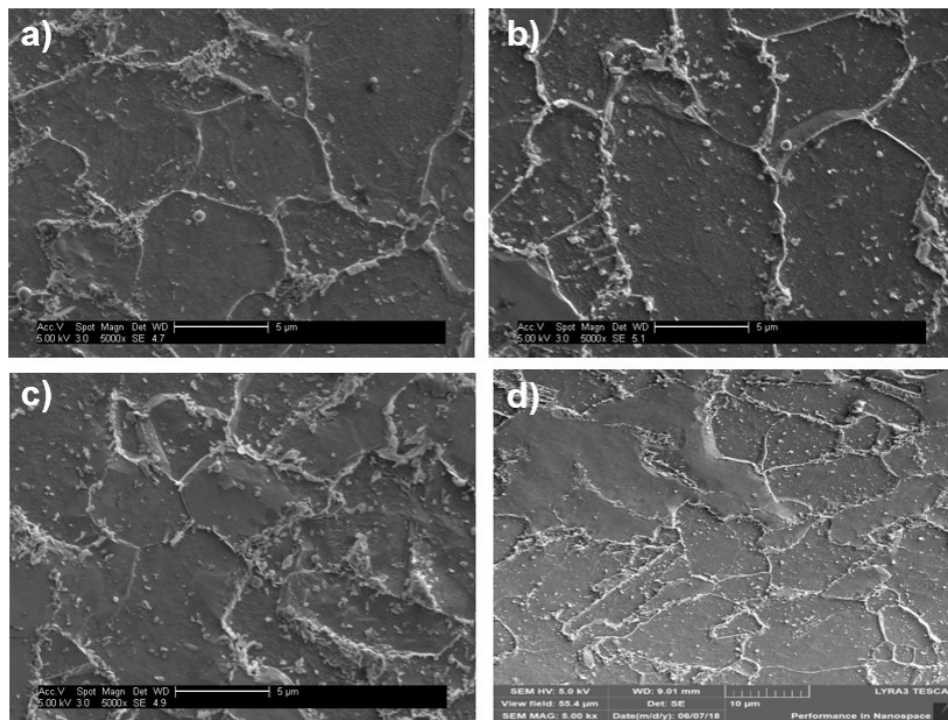


Figure 6-11 SEM images of side rolling

- (a) As-deposited + rolled (75kN) (b) As-deposited + rolled (100kN) (c) As-deposited + rolled (140kN) (d) As-deposited + rolled (160kN)

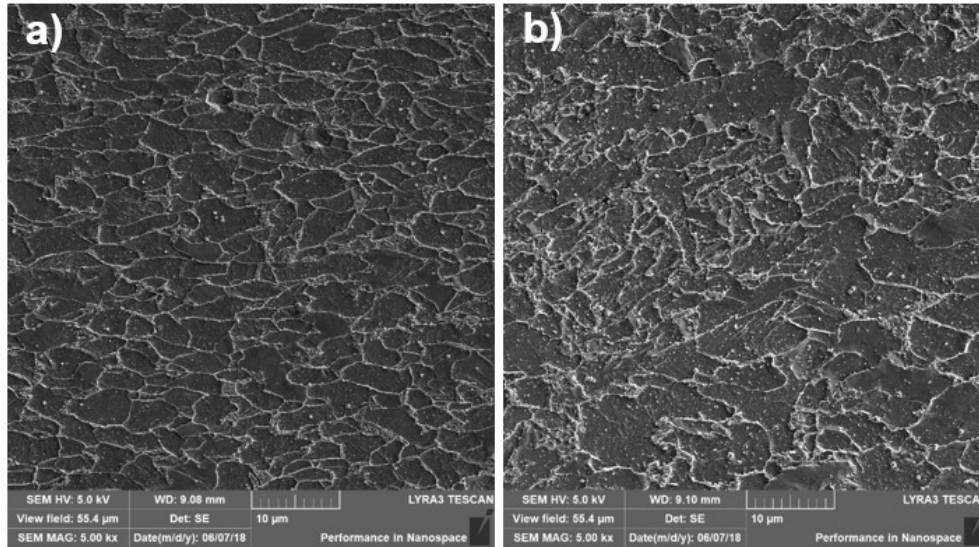


Figure 6-12 SEM images of Interpass top rolling

(a) 75kN (b) 50kN

6.4 Discussion of results

6.4.1 Macroscopic failure mechanism of as deposited and rolled WAAM structures

Figure 6-13 presents the effect of rolling on notch radius and SCF. This is an extension of Figure 5-15. Rolling is a useful tool in increasing the notch radius and thereby, reduction in SCF and SW. The effect of notch radius on SW for both rolled, and the as-deposited condition is presented in Figure 6-14 while Figure 6-15 shows the results of SCF on SW of both rolled and as-deposited conditions. It is evident from this study that the notch radius, (ρ) increases as the SW reduces, as shown in Figure 6-13. Low values of SW give rise to a corresponding increase in notch radius as shown in Figure 6-13. Rolling reduces the notch depth and increases the notch radius. A smaller notch radius is responsible for a high-stress concentration between the layers as a result of the poor transition between the layers, hence tensile and fatigue life reduction. In Figure 6-5, the stress concentration factor is shown as a linear function of SW; both are responsible for the fatigue behaviour of the as-deposited condition. The rolled condition shows a smooth transition between the layers, as shown in Figure 6-3(d & e). Failure in WAAM as deposited is always from this point of weakness.

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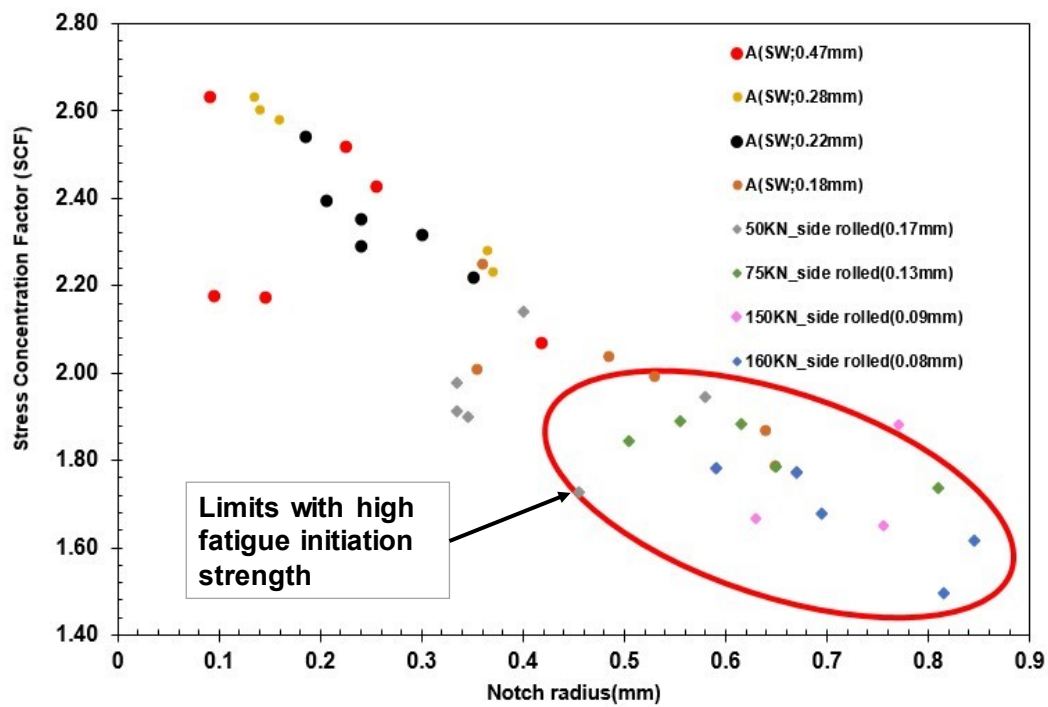


Figure 6-13 effects of rolling on notch radius & SCF

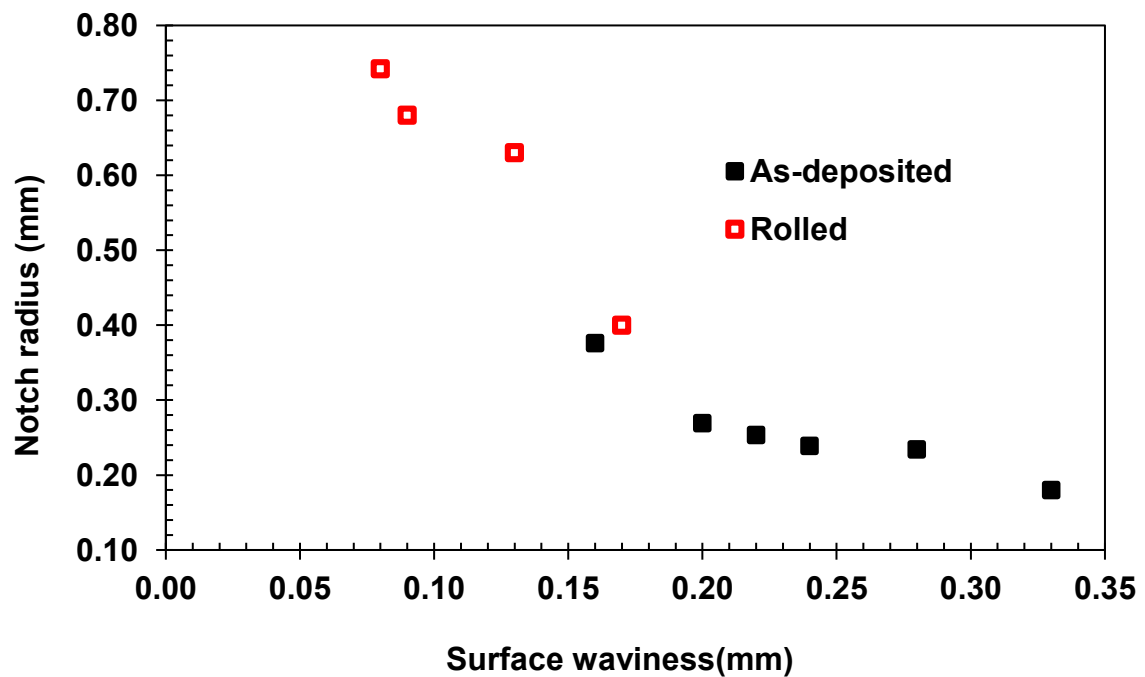


Figure 6-14 Parameters effects; notch radius vs. SW for both ASD and rolled condition

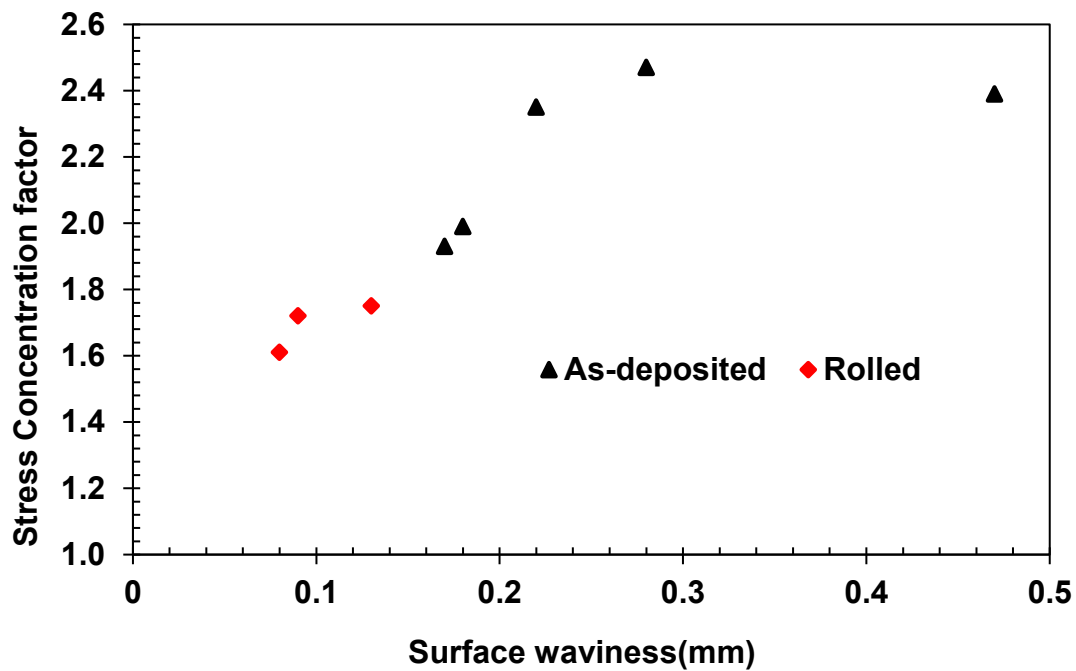


Figure 6-15 Parameters effects; SCF vs. SW on both rolled and ASD

6.4.2 Tensile properties and fracture morphology

Figure 6-16 shows the static tensile test values of WAAM as-deposited, machined, and rolled in different conditions. The tensile fracture morphology in the various conditions is presented in Figure 6-17. It is evident from the results that the UTS and YS are in direct correlation with the level of SW. High SW leads to reduced tensile properties. The SW serves as points of fracture initiation – the higher the SW, the quicker to failure the sample will be. It is evident from Figure 6-6 that the application of rolling, which reduced the SW, also enhanced the UTS and YS of the WAAM deposit. Rolling with a 160KN force produced a 55% and 49.25% enhancement on the SW and UTS respectively, while a rolling load of 75KN produced 27% and 3.9% in SW and UTS enhancement respectively. But PCE, on the other hand, is decreased as the rolling load increased. There was a decrease of 37% in the PCE when a rolled load of 160KN was applied to the as-deposited condition. Rolling induced strain results in the reduction of thickness and increased in length in the rolling direction as a result of plastic flow.

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This is the reason why the grains observed in the side rolled samples with different loads were more elongated with only little changes in grain size compared to the top rolled samples (Figure 6-11, Figure 6-12 and Figure 6-12). The machined condition without SW showed higher UTS and YS compared to the as-deposited condition. The as-deposited + rolled + machined condition produce a UTS and YS of 665MPa and 608MPa respectively, but with the lowest PCE of 15%, due to work hardening and resulting reduction of the plastic region, as shown in Figure 6-6.

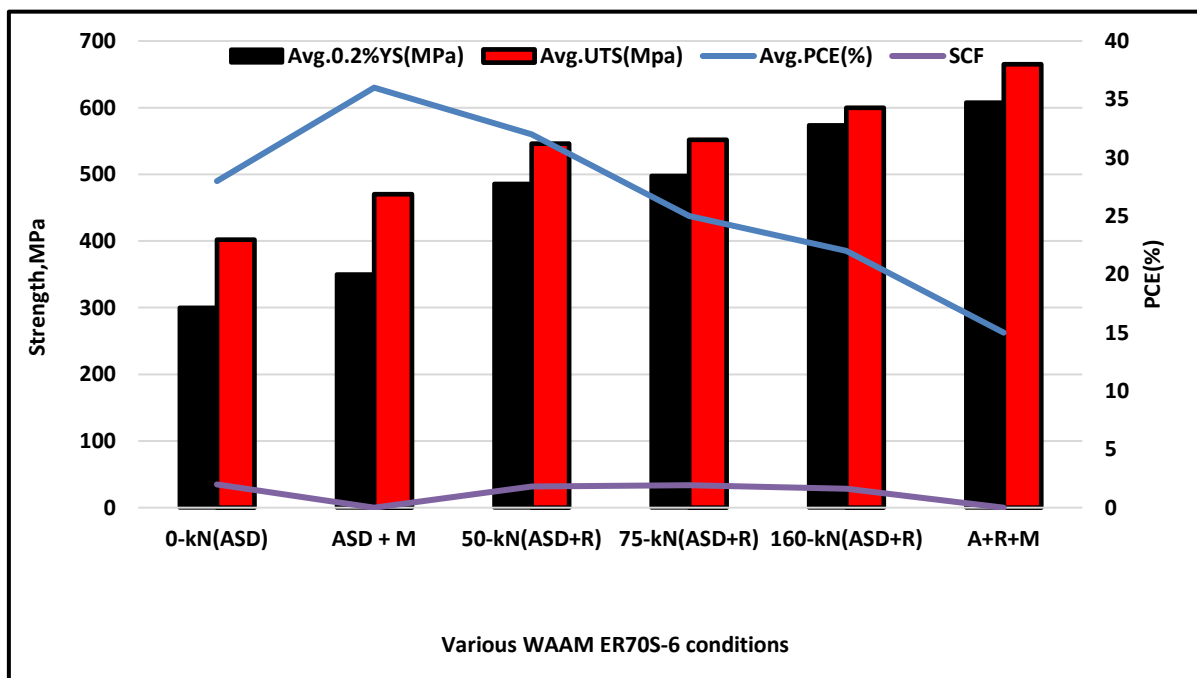


Figure 6-16 Average strength test values of WAAM ASD, ASD + R and ASD + M

The fracture morphology of the as-deposited structures shown in Figure 6-17(a) is characterised by conical equiaxed dimples formed from microvoid coalescence rupture, indicating a high level of ductile tearing with low strength. Precipitates in the form of inclusions are also seen. The dimples in the as-deposited are more oval and ductile compared to the as-deposited + rolled condition (Figure 6-17(c)). The as-deposited + rolled + machined condition as seen in Figure 6-17(b) has smaller and deeper dimples, hence the higher strength with reducing ductility. The as-deposited + rolled condition as shown in Figure 6-17(c) is dominated by a mixture of large and small elongated dimples as a result of plastic deformation, enhancing the elongation of the grain structure as a result of that plastic

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deformation. Hence the elongated dimples lead to high UTS and YS, with reduced ductility compared to the as-deposited condition. The machined condition in Figure 6-17(d) has smaller interwoven dimples as a result of the high fracture resistance it possesses, hence its high ductility compared to as-deposited and rolled condition.

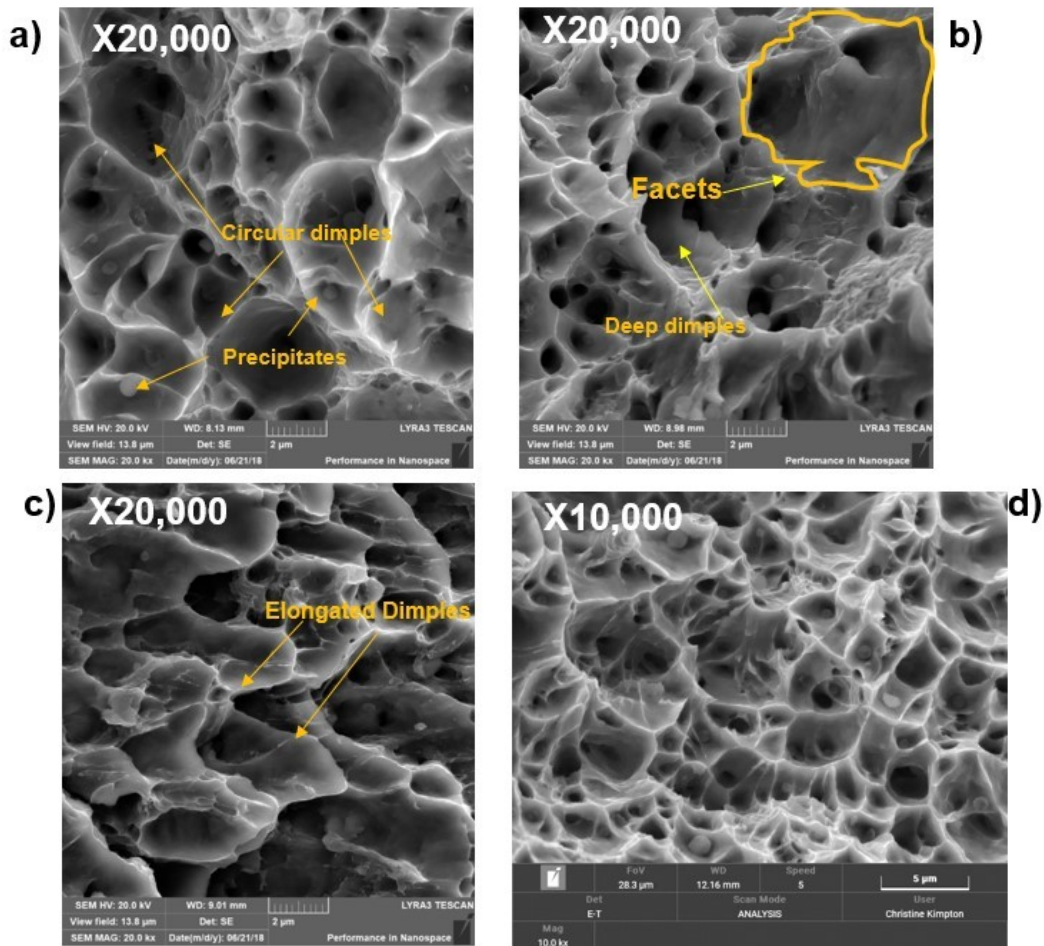


Figure 6-17 Tensile fracture morphology of WAAM ER70S-6

(a) As-deposited (b) As-deposited + rolled (160KN) + Machined (c) As-deposited + rolled (160KN) (d) Machined

6.4.3 Influence of cold work on microstructure (XRD & TEM analysis)

The WAAM layer by layer deposition with top rolling leads to grain refinement of the subsequent layer; the top rolling microstructure showed distinctive, refined grains, with the 75kN load giving more refined and equiaxed grains, as shown in

Figure 6-12(a). The XRD patterns of the as-deposited (70-A) and side rolled (160kN) are shown in Figure 6-18. Ferrite colonies were observed in both samples with no appearance of the austenite phase, i.e. no phase transformation after the sample was subjected to side rolling. At diffraction angles (2θ) of 45.18° , 65.31° and 82.71° , the samples exhibit peaks pertaining to Fe (110), Fe (200), Fe (211) planes, whereas a little broadening of the peaks pertaining to all three planes is observed for the rolled sample (Figure 6-18(b-d)).

The broadening of the peaks experienced by the rolling sample could be ascribed to the structural refinement in the orientation of the sample. The rolled sample (160kN) experienced a higher intensity in the XRD pattern as compared to the as-deposited. Using Scherrer's equation, the volume of the phases was calculated to be 0.02263 nm^3 , 0.02322 nm^3 , and 0.02326 nm^3 , for (110), (200), and (211) phases, respectively. Due to the peak broadening as shown in the XRD analysis (Figure 6-18) and structural refinement in orientation observed in the rolled-160kN sample, compressive residual stress was induced which increases the dislocation cluster density and volume fraction of the phases.

This could be responsible for the high hardness experienced by the rolled-160kN sample in Figure 6-9. Furthermore, the degree of dislocation in the rolled-160kN sample is strongly dependent on the level of deformation induced by the rolling process. The improved hardness of the rolled condition is due to work hardening; this enhances both the strength and fatigue life of the as-deposited. Figure 6-19 and Figure 6-20 show the transmission electron microscope (TEM) images of the as-deposited and rolled (160kN) WAAM steel. In the as-deposited condition (Figure 6-19(a-c)), thin-like-shaped cementite particles are observed inside the lath with an increased dislocation density of the cementite. Beside, PGF was produced. Dislocation at the lath boundaries was also produced with sub-grain boundaries (nanoscale) or AF (Figure 6-19(b-c)).

Furthermore, a well-developed dislocation substructure is observed in the volume of ferrite crystals in the as-deposited samples. The thin lath-like particles of cementite appeared in the ferrite lathes by the beginning of deformation (Figure 6-19(a-c)). The TEM images in Figure 6-20(a-c) reveal the production of grains

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which were separated by high-angle grain boundaries. Most of the grains in the as-rolled samples are heavily strained and contain a high density of dislocations, as indicated in Figure 6-20(a-b). Figure 6-20(f) shows the diffraction pattern of surrounding ferrite in the matrix.

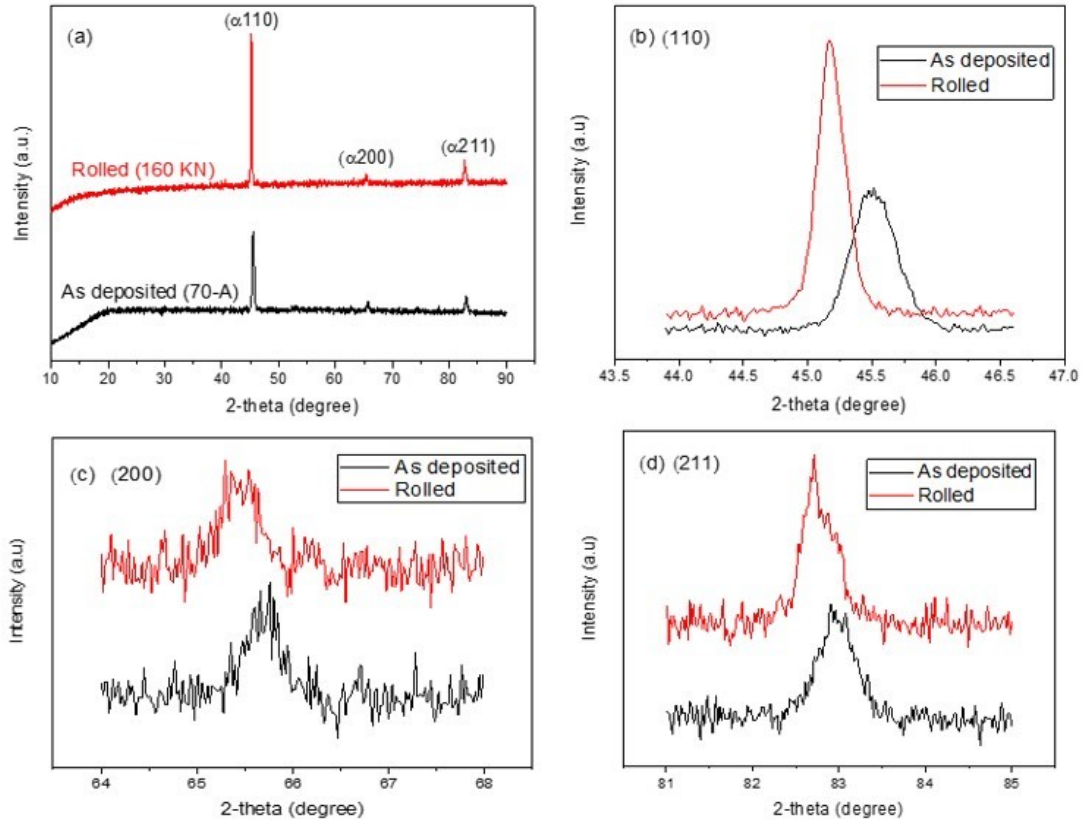


Figure 6-18 XRD Pattern

Untreated (70-A) and treated (160KN) mild steel (a); detailed XRD patterns for the peaks: Fe (110) (b), Fe (200) (c), Fe (211) (d)

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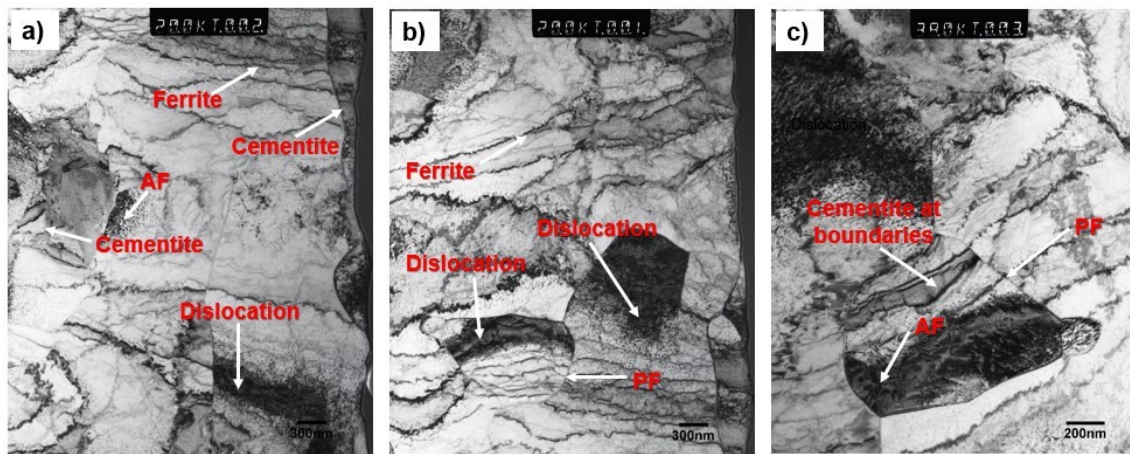


Figure 6-19 Bright TEM micrographs of as-deposited

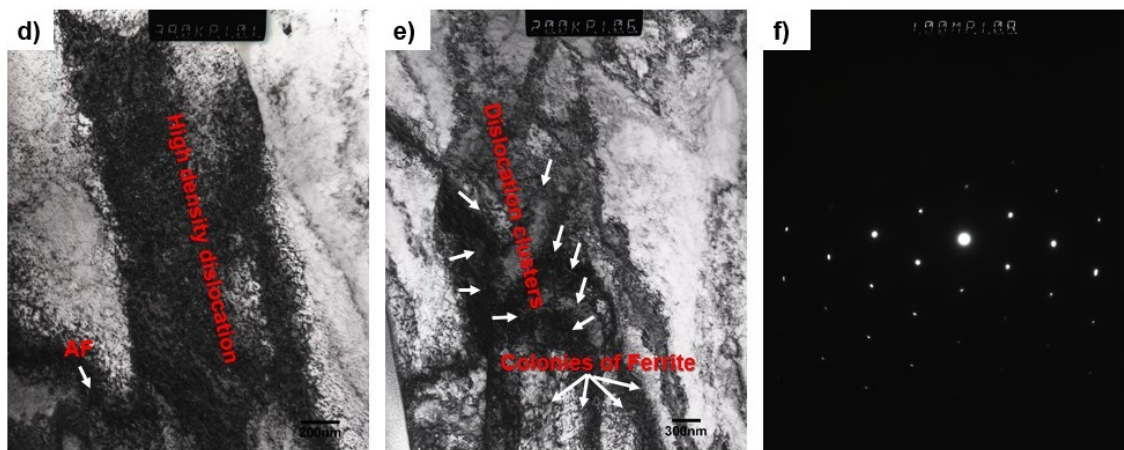


Figure 6-20 Bright TEM micrographs of as-deposited + rolled (160kN)

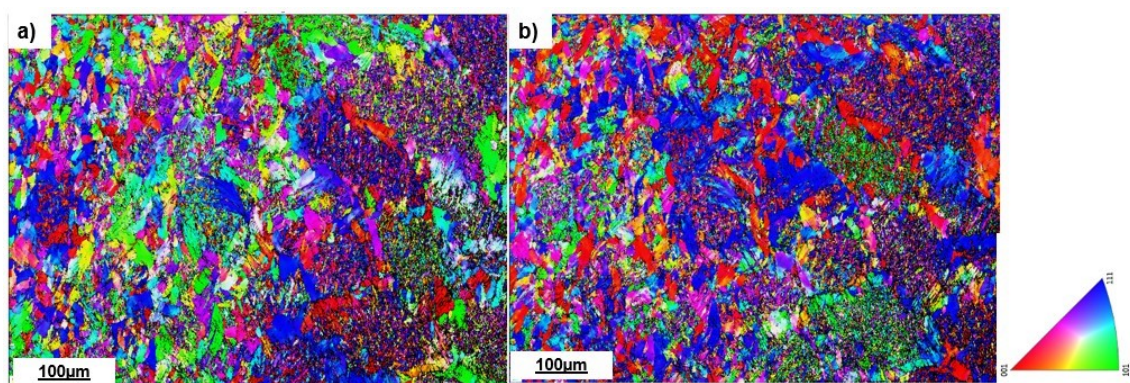


Figure 6-21 EBSD maps top rolled (75kN)

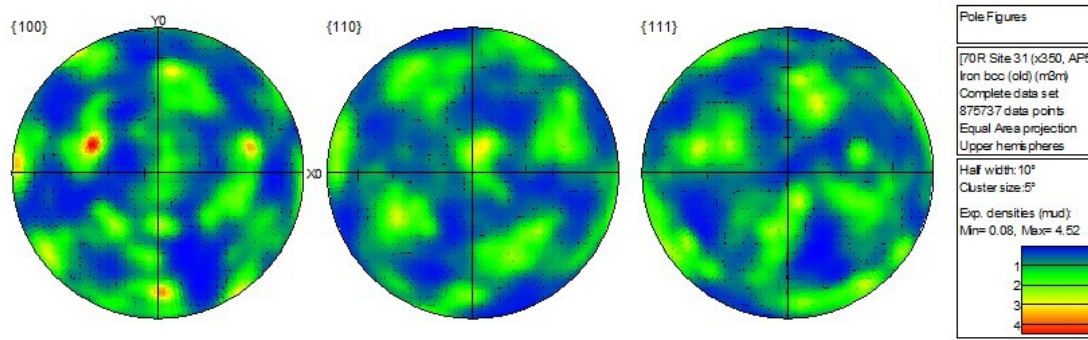


Figure 6-22 Pole figures (75kN)

6.4.4 Microscopic failure mechanism and effect of rolling on fatigue

Figure 6-21 & Figure 6-22 present the EBSD map and pole figure for the top rolled (75kN) condition. Rolling induced recrystallisation was responsible for the change in the grain morphology; the grain size is smaller with a weakened texture morphology than as-deposited ER70S-6, as shown in Figure 4-25(a & b) with a grain size of 14.9 μm . The high cycle fatigue behaviour of the WAAM as-deposited, machined, and as-deposited + rolled is displayed in Figure 6-23. The as-deposited condition has an SW of 0.18mm (180 μm), while the as-deposited + rolled condition has an SW of 0.08mm (80 μm) when rolled with a force of 160kN.

The machined condition showed a better fatigue limit than other conditions, but generally, rolling improved the fatigue limit of the as-deposited samples. Figure 6-24 presents the fractography of WAAM ER70S-6 in the machined condition, Figure 6-25 also shows the fractographs of the as-deposited condition, while the fractographic analysis of the as-deposited + rolled condition is displayed in Figure 6-26. Crack initiation mechanism and failure path in the machined condition is essentially by a ductile fracture through microvoid coalescence from the weakened path formed by nucleation of hard spots in the microstructure as a result of WAAM deposition process, as shown in Figure 6-24(a). Silicon and manganese inclusions are formed as a result of the oxidising potential of the shielding gas used for the deposition, and the relatively high heating and rapid cooling of the WAAM process [131]. These inclusions serve as hard spots with high-stress concentration and areas of quick failure initiation for the machined

Effect of Surface Rolling Methods on WAAM Properties

condition as the effect of SW has been eliminated by machining [58]. The fatigue crack growth in the machined condition is based on applied stress conditions and the nature of the microstructure [132]. The ductility of this material also supports the fatigue crack growth rate behaviour, as shown in Figure 6-24(a & b). The striation spacing is closer than that observed in the as-deposited condition shown in Figure 6-25; hence, the higher fatigue life of 107,698 cycles at maximum stress of 492MPa was spent.

The main function of the added micro-alloying elements (Ti, Cu, and Zr) as shown in Table 3-1 is for precipitation strengthening and the promotion of acicular ferrite formation, which induces microstructural refinement [133], hence the behaviour of the material. The significant reduction in fatigue life for the as-deposited sample is primarily due to the presence of notches (SW) which act as stress raisers. Cracks initiated at the trough (pit) of the SW and propagated towards the centre of the specimen, at critical length where the remaining sample ligament could no longer support the increasing stress, resulting in a plastic collapse occurring.

Maximum stress concentration occurs between any two layers that form the SW profile, as shown in Figure 6-25(a). Increasing the notch radius and decreasing the depths of the SW with plastic deformation (rolling) minimises the stress concentration effects. The lower the SCF of a WAAM as-deposited surface, the more resistance to crack initiation it has, and this leads to better overall integrity that will be observed under both monotonic and cyclic loading conditions. Initiation of cracks during fatigue loading is usually from the outside and propagates inward till failure, as shown in Figure 6-25 (b). The fractography also reveals multiple crack initiation sites at the surface, independently of the stress amplitude condition.

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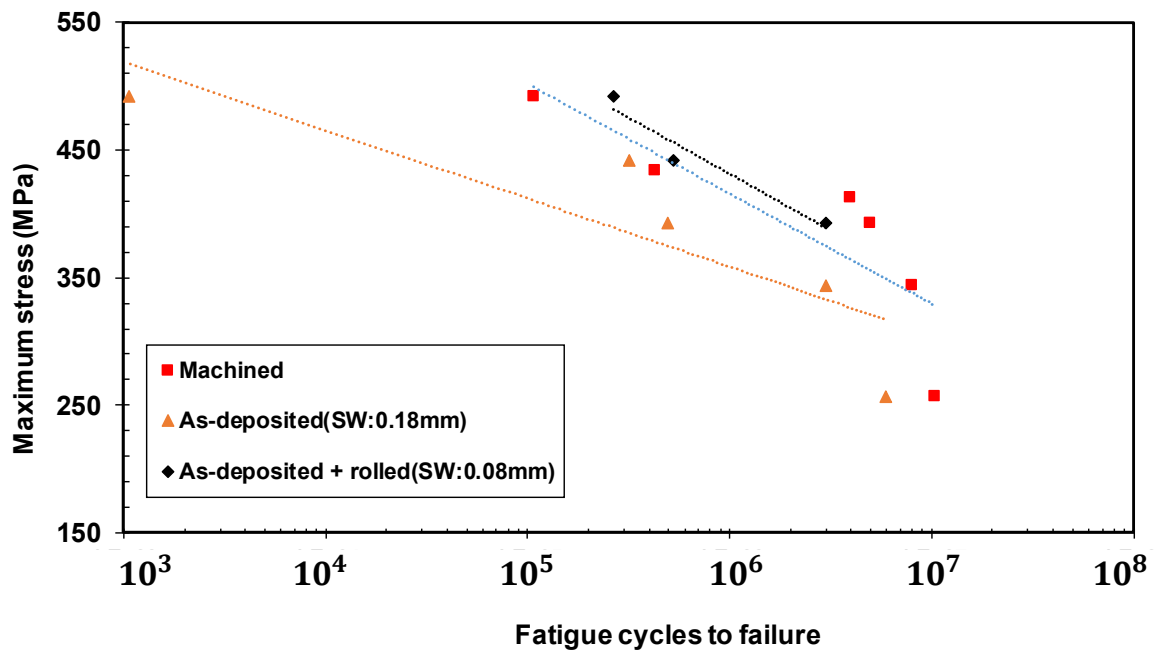


Figure 6-23 High cycle fatigue properties (R=0.1) of WAAM ER70S-6(X-direction)

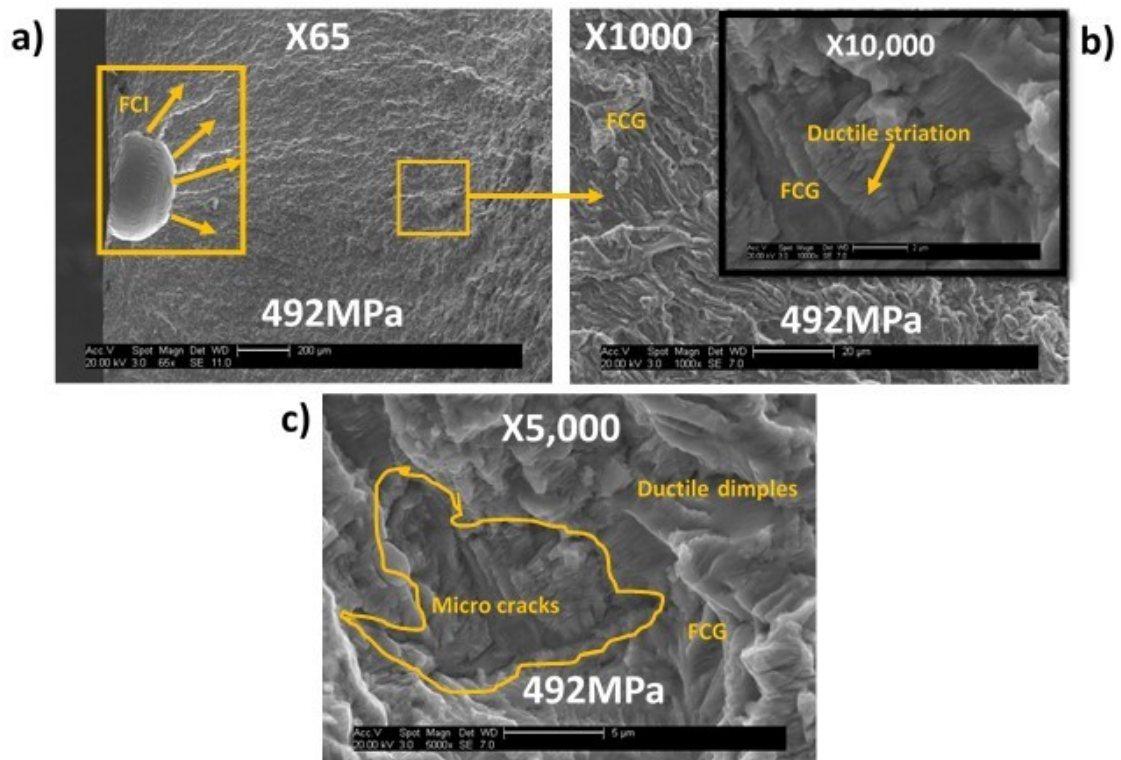


Figure 6-24 SEM Fractography of WAAM ER70S-6 in the machined condition

(a-c) $\sigma_{max} = 492\text{MPa}$, $N_f = 107,698$ cycles

Effect of Surface Rolling Methods on WAAM Properties

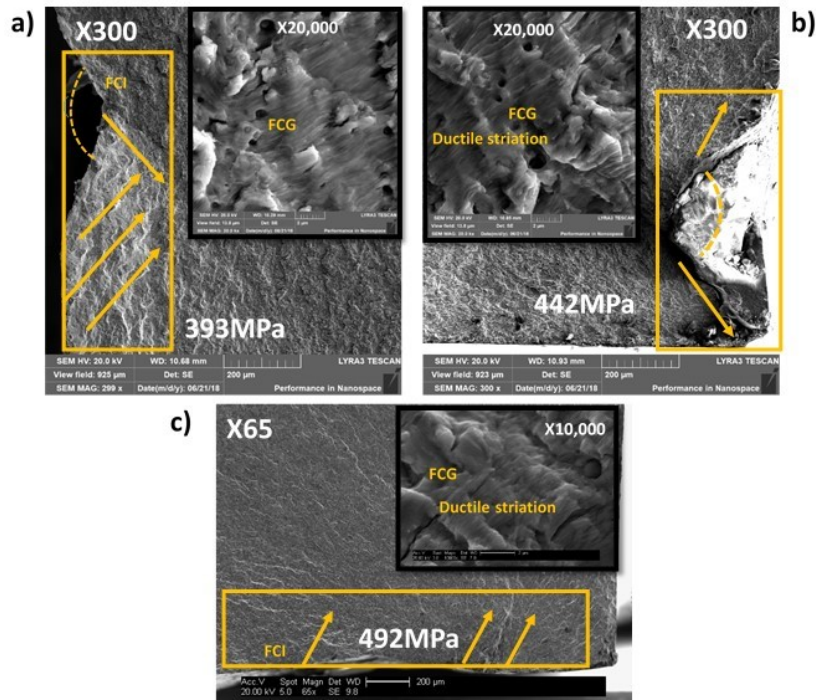


Figure 6-25 SEM Fractography of WAAM ER70S-6 (as-deposited)

- (a) $\sigma_{max} = 393MPa$, $N_f = 495,119$ cycles (b) $\sigma_{max} = 442MPa$, $N_f = 318,145$ cycles (c) $\sigma_{max} = 492MPa$, $N_f = 1,069$ cycles

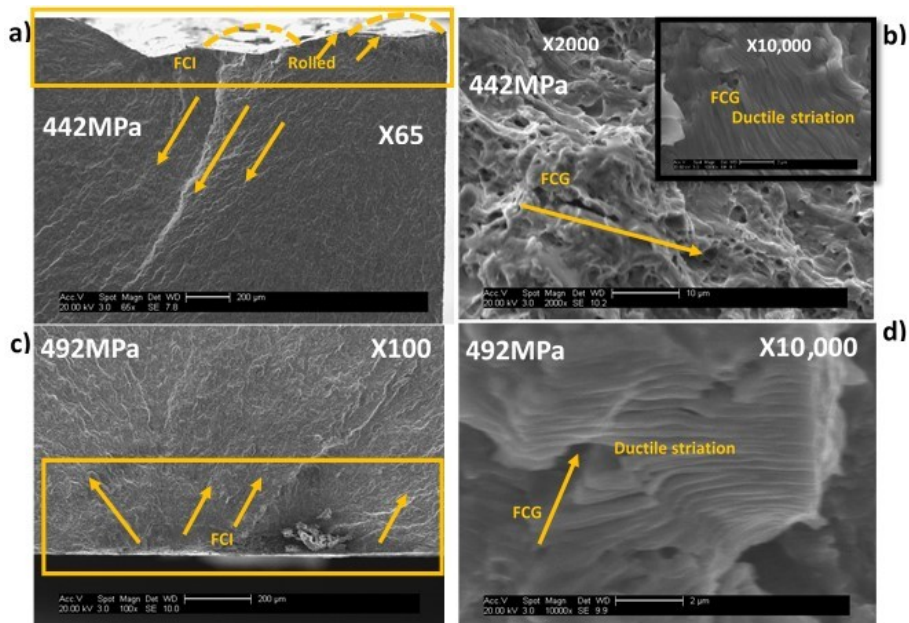


Figure 6-26 SEM Fractography of WAAM ER70S-6 rolled

- (As-deposited + rolled_160KN) (a-b) $\sigma_{max} = 442MPa$, $N_f = 530,319$ cycles (c-d) $\sigma_{max} = 492MPa$, $N_f = 266,856$ cycles (c)

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The closer the included angle between two layers to 180° , as shown in Figure 6-25(c), the greater the resistance to fatigue crack failure. The fatigue crack growth rate of the WAAM as-deposited part is dependent on the stress condition and geometry of the notch (SW). The fractography in the as-deposited condition indicates that the striation spacing is dependent on the stress level applied to the as-deposited condition. The lowest fatigue life of 1,069 cycles was observed at the maximum stress level of 492MPa in the as-deposited condition as a result of SW, hence lower than the machine condition at the same stress level.

The crack initiation site and path in the as-deposited + rolled WAAM surfaces is between any two weak layers with maximum stress concentration where their curvature was not effectively reduced, as shown Figure 6-26(a & c). The notch radius is increased by the application of rolling as shown in Figure 6-26(a) where the depth between two layers is reduced, hence the reduction of the stress concentration factor on the rolled surface. The higher rolling load is required for lower SW, as shown in Figure 6-5. The as-deposited + rolled had a fatigue life improvement of 66% at a stress level of 442MPa as compared to the as-deposited condition.

The as-deposited + rolled showed a higher fatigue life of 266,856 cycles at a stress level of 492MPa, which is higher than the as-deposited condition (1,069 cycles) and the machined condition (107,698 cycles) even at the same stress level. Most of the grains in the rolled samples are heavily strained and contain a high density of dislocations as indicated in Figure 6-20(a & b), hence improvement in fatigue life. Rolling improves the usefulness of the WAAM as-deposited part by plastically deforming the peaks of the SW, thus increasing notch radius and decreasing the depths, which are the points of stress raisers. The deformed notches and valleys and the surface hardening effects are responsible for delaying the initiation of cracks when subjected to static and dynamic loading. It has also been reported that compressive stress from rolling retards cracks propagation in a structure.

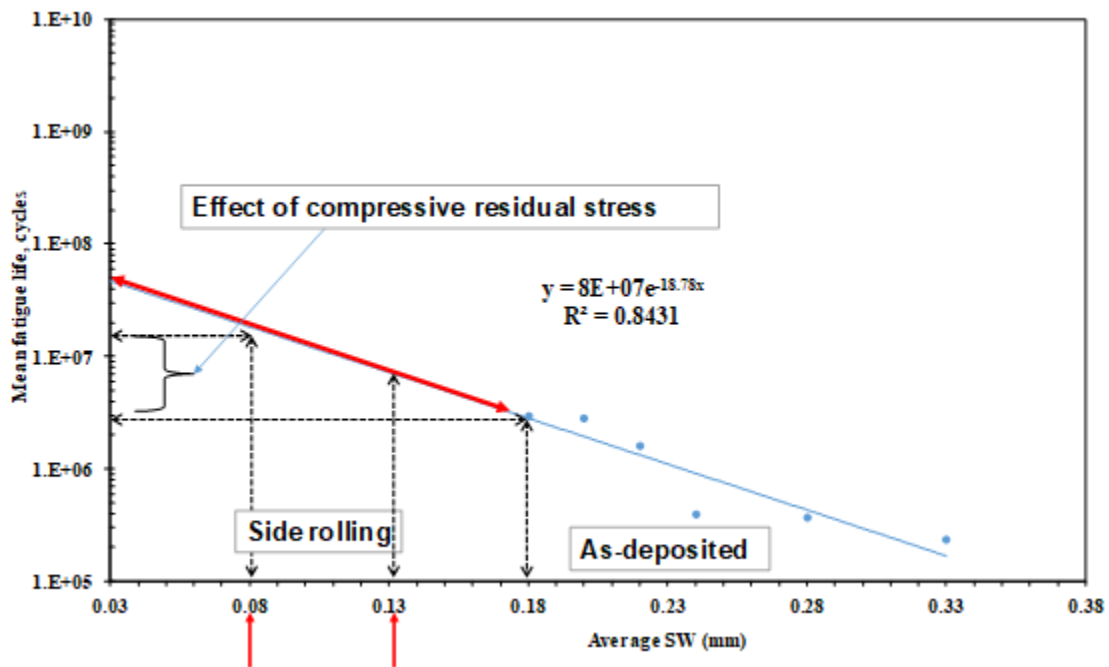


Figure 6-27 Prediction of fatigue life due to SW reduction at a constant stress level of 343.7MPa(X-direction)

If the mean fatigue life versus average surface waviness plot in Figure 5-18 is extrapolated backwards to see the effect of reducing SW by rolling (0.18 - 0.08 mm), the improved fatigue life observed in the new plot in Figure 6-27 shows a life enhancement beyond 10^6 Cycles as a result of the compressive residual stress and the notch radius increment. The fatigue life of as-deposited + rolled surface of 0.08 mm which was not experimented in Figure 6-23 was predicted from Figure 6-27 to be over 10^7 cycles, hence the usefulness of the plot.

6.4.5 Conclusions

In this chapter, the functionality of wire + arc additive manufactured mild steel in the as-deposited condition was enhanced by rolling to improve its surface finish and fatigue life. From the investigation, it was shown that:

- High SW is detrimental to the mechanical strength and fatigue life of an as-deposited WAAM steel structure.

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- There is a possibility of the introduction of a surface compressive residual stress state which would lengthen the life of a component.
- Reduction of the SW by rolling also resulted in an increase of notch radius between two layers in WAAM. Increase in notch radius is also responsible for fatigue life improvement due to lowering in stress concentration.
- WAAM mild steel components could be used in the as-deposited condition for engineering application when rolling is applied, which would improve WAAM's strength and reduce its SW.
- Side rolling at room temperature results in the lowering of SW and also increases its strength due to work hardening.
- Rolling is environmentally friendly compared to machining because of its residue that could pollute the environment.

The investigation in the present chapter proved that functional surfaces could be made out of WAAM deposit by side rolling. WAAM mild steel components could be used in the as-deposited condition for engineering application when rolling is applied to reduce its SW. Chapter 7 will be investigating further the fracture and fatigue behaviour of the machined condition of these WAAM structures build with oscillatory deposition strategy.

7 FRACTURE AND FATIGUE BEHAVIOUR OF SINGLE AND OSCILLATORY PASS DEPOSITION METHODS ON WAAM PROPERTIES

7.1 Introduction

Regardless of the efforts made to characterise the wire + arc additive manufactured structural steel components, all current research focuses on single pass WAAM deposited steel components with a thickness not exceeding 10mm using single-pass strategy. While some of the research only focuses on the mechanical and fracture toughness strength of mild steels using oscillatory and parallel deposition strategy to build structures thicker than 10mm.

This chapter builds further on the work done in chapter 4 and 6. The first objective of this chapter is to understand the basic microstructural changes between single-pass and oscillatory pass strategies.

The second objective in this chapter is to understand how the microstructural variation, as a result of the layer band, affects the fracture resistance in the build and welding direction. This investigation also points out the fracture mechanics in terms of J, CTOD and failure mode of the WAAM oscillatory deposited structure.

The third objective is to understand the relationship between the fatigue crack growth rate of WAAM HSLA steel and wrought S355 steels.

The fourth and last objective is to understand the fatigue life of this WAAM HSLA steel.

7.2 Experimental procedure

7.2.1 Oscillatory build strategy

The CMT-WAAM of the structures was carried out with a standard cold metal transfer welding process, which is a modified gas metal arc welding process based on a controlled dip transfer mode mechanism [134]. A Fronius CMT (VR 7000 CMT) power source with a wire feeder and a CMT torch connected to an IRB 2400 ABB robot is used, as shown in Figure 3-1. Local shielding was done with spec shield, light (80%Ar + 20%CO₂) for ER70S-6 and ER120S-G and spec shield heavy (Ar +2%O₂) for ER90S-B3 at a flow rate of 15L/min. The base plate used in the experiment is rolled structural steel plates of S275 JR (EN10025) with dimensions of 350mm x 200mm x 15mm. The plates are clamped by steel jig tightened to an aluminium base plate. The wire diameter used was 1.2mm. The deposition was done in the PA (1G/Down hand) position from the base of the plate with the torch and wire feeding in the vertical direction. BÖHLER ER120S-G and ER90S-B3 with Lincoln ER70S-6, as shown in Table 3-1, were used.

Table 7-1 Oscillatory WAAM parameters

Strategy	Current (amps)	Voltage (Volts)	HI (kJ/mm)	WFS (m/min)
Oscillatory pass	157 - 165	13.3- 15.4	0.313 - 0.42	6.5-7.5
Strategy	TS (mm/sec)	GPF (secs)	WFS/TS	CTWD
Oscillatory pass	6.67 - 10.0	1.4	16.25 - 20	11.5 - 13

The oscillatory pass strategies, as shown in Figure 7-1, were used to deposit layer by layer onto the substrate with the bi-directional oscillatory movement of the torch. A portable arc monitor was used to monitor the weld parameters. The parameters in Table 7-1 were used with a dwell time of 60secs. For each successive layer deposition, a layer height increment of 3.5mm of the torch distance from the previously deposited layer was targeted for the oscillatory pass. For the oscillatory strategy, two walls each were built with the wires, totalling six

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walls for the three different grades studied, with a dimension of 290mm x 120mm x 24.5mm.

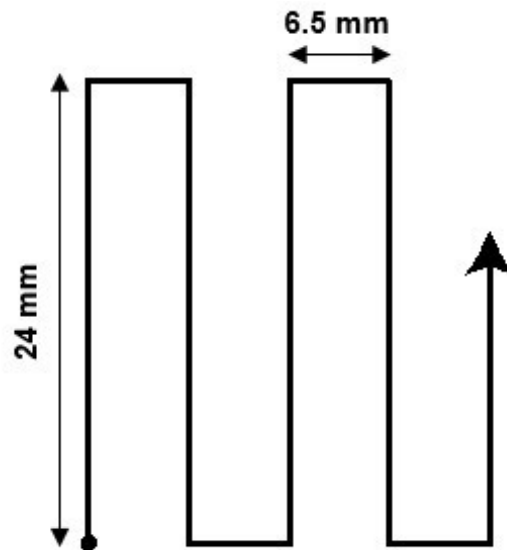


Figure 7-1 Oscillatory Strategy

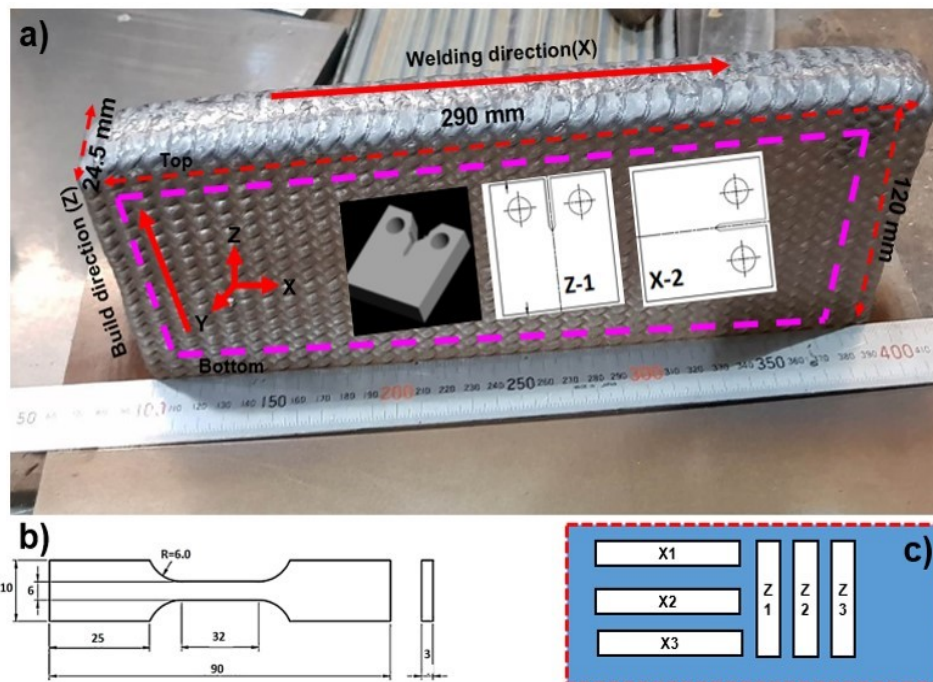


Figure 7-2 Oscillatory deposited wall geometry

(a) C (T) machining matrix (b) Tensile sample geometry (c) Tensile specimen machining matrix

7.2.2 Mechanical and microstructural characterisation methodology

Eighteen tensile specimens (six from each grade), as shown in Figure 7-2(c), were cut from the oscillatory wall in the build and welding direction; all specimens were prepared following ASTM E8/E8M-16a standards. The uni-axial tensile testing was performed with an electro-mechanical controlled Instron 5500R machine with a load cell of 30kN using a strain rate of 1mm/min. The metallography samples were extracted from the as-deposited (AD) walls and prepared with a Buehler Metaserve 250 grinder and polisher machine.

The ground and polished samples were etched in 2% Nital (Nitric acid and Ethanol) for 10 seconds to reveal the microstructure. The electron backscattered diffraction (EBSD) was carried out to characterise the grains and texture. The grain sizes were determined using the line intercept method. Scanning Electron Microscope (SEM, FEI XL 30-SFEG) equipped with energy dispersive spectrometry detector (EDS) was used to characterise the fracture toughness surfaces, fatigue crack growth path and elemental composition of inclusions and carbides in the WAAM deposit. The Vickers hardness testing was done at room temperature using the Zwick/Roell hardness testing machine. An average of 42 indents were made on the polished surface along the build direction using a 300g load at a spacing of 0.5mm.

7.2.3 Fracture toughness characterisation following EPFM

The oscillatory build walls with a thickness of 24.5mm were machined, as shown in Figure 7-2(a). The C (T) specimens were cut and machined from two different orientations of the deposited structure. The first orientation was from the Z-direction (grooved notch perpendicular to the layer band and welding direction), as shown in Figure 7-5(a). The groove notch was placed in two different locations on the second orientation (grooved notch parallel to the layer bands and welding direction). In the first location, the grooved notch was in between a layer (intra layer), as shown in Figure 7-5(b) and in the second location, the grooved notch was placed on the layer band boundary between two layers (interlayer) as shown

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in Figure 7-5(c). Three C (T) specimens each with dimensions of 60mm x 62.5mm x 16mm, were prepared from each grade according to the ASTM E1820 standard using electronic discharge machining (EDM) as shown in Figure 7-3(a). Fatigue pre-cracking and fracture toughness test as per J-integral was conducted with an Instron 8801 servo-hydraulic machine with a load cell of 100kN. An initial crack starter of ($a_o = 19\text{mm}$) was machined by EDM. The fatigue pre-cracking was done with a frequency of 5Hz to attain a fatigue crack size range (total average length of the crack starter configuration + fatigue crack) of 0.45W - 0.70W using the K-decreasing approach with the maximum stress intensity in the first step limited to $K_{max} = (\sigma^f Y_s | \sigma^T Y_s) (0.063 \sigma^f Y_s M P a \sqrt{m})$, where $\sigma^f Y_s$ & $\sigma^T Y_s$ are the material yield stresses at fatigue pre-crack and test temperature. This was done to introduce a sharp crack tip without allowing a significant plastic zone to develop ahead of the starter crack tip.

The pre-cracked C (T) samples were latter side grooved to 0.2B, where B is the final thickness of the C (T) specimen (20% of total thickness) on both faces of the sample along the pre-cracked path to further introduce dominant plane constraint to minimise plastic zone formation, as shown in Figure 7-4(b). All specimens had width, total thickness and net thickness of W: 50mm, B: 16mm, and B_n : 12.8mm, respectively. The fracture toughness test was performed by applying sequences of loading and partial unloading at specified intervals. The load and load line displacement (LLD) data were measured using a clip gauge attached to the crack mouth of the C (T) specimen.

The test was performed under LLD control mode with 5-minutes holding time, followed by 20% unloading at each peak load. The unloading slopes, which are linear and independent of prior plastic deformation, were used to estimate the instantaneous crack length at each unloading increment using the elastic compliance relationship. All tests were performed at room temperature with a loading and unloading rate of 0.50mm/min and an LLD increment of 0.125mm.

Fracture and Fatigue Behaviour of WAAM Components

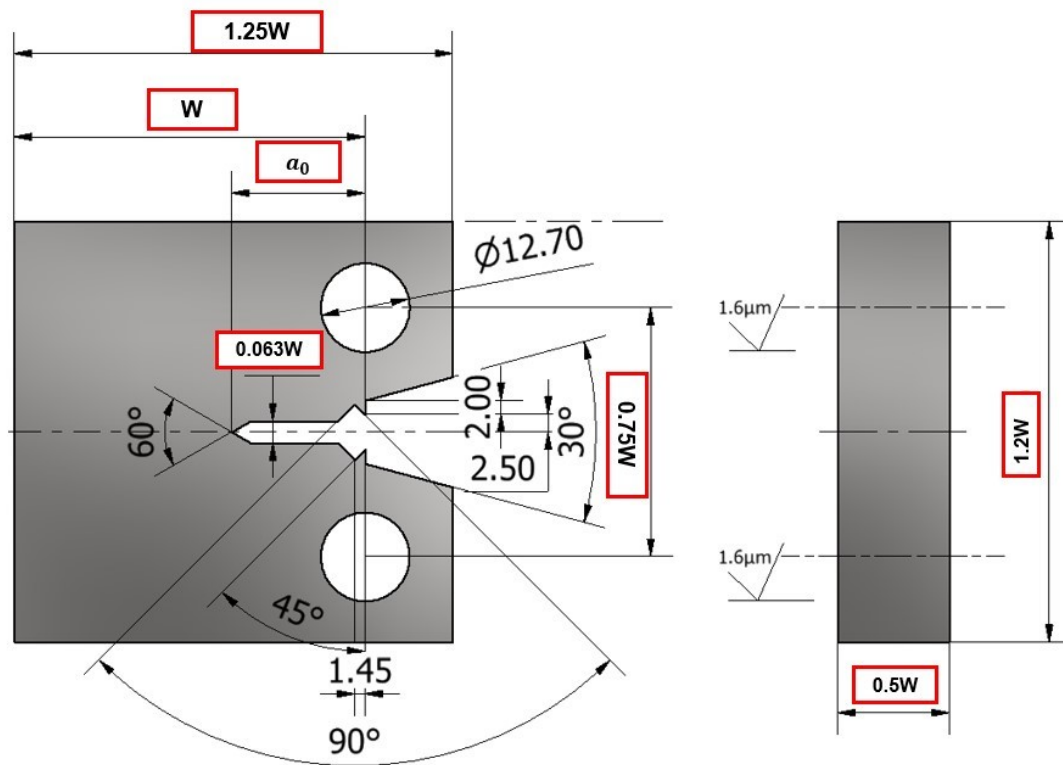


Figure 7-3 C (T) specimen in accordance with ASTM E1820-17

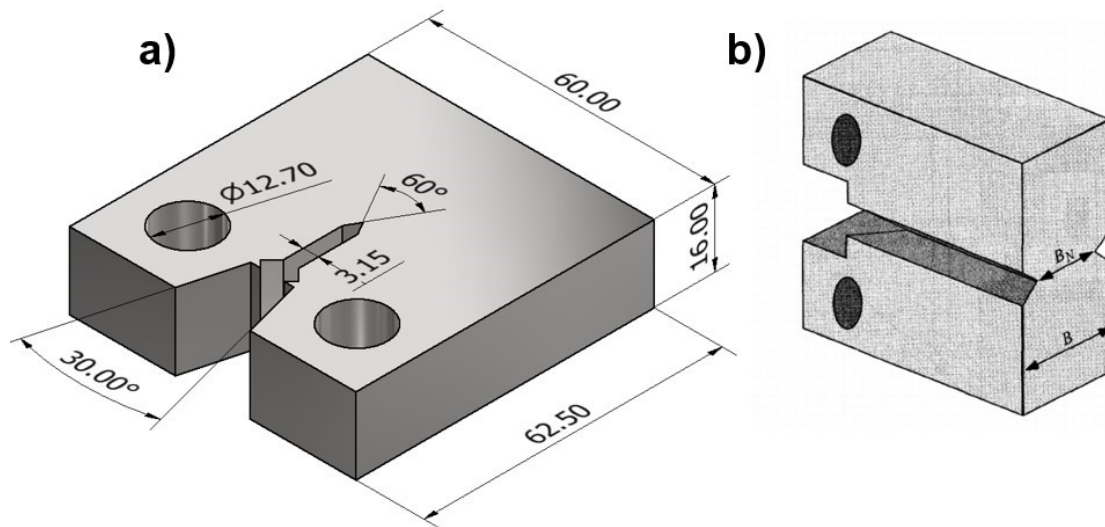


Figure 7-4 (a) C (T) specimen in accordance to ASTM E1820-17 (b) Side groove geometry

Fracture and Fatigue Behaviour of WAAM Components

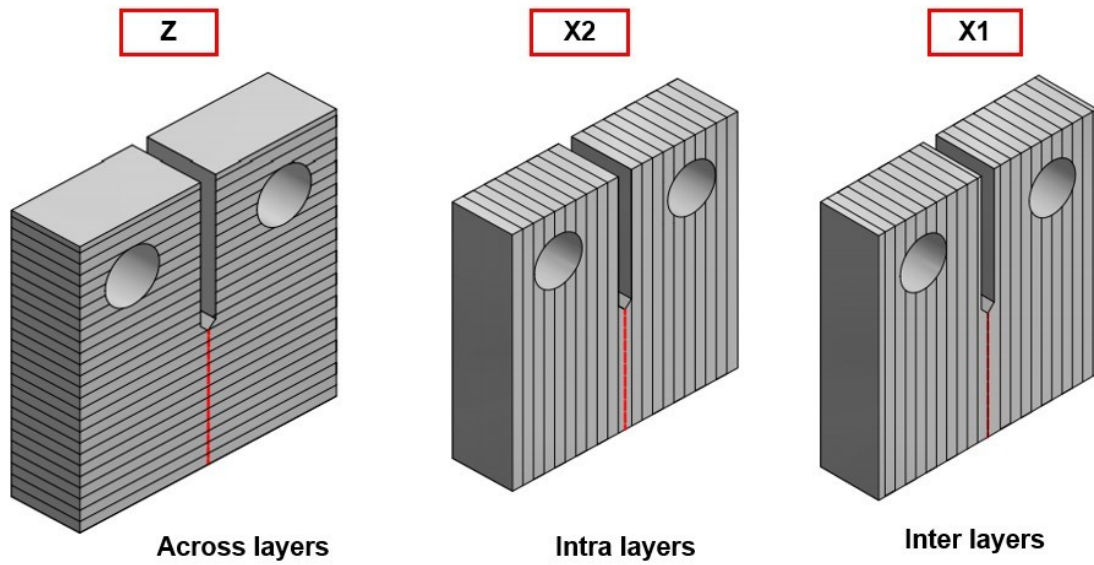


Figure 7-5 J_{1c} fracture toughness

(a) Across the layer (b) Along the layer – in between a layer (intralayer) (c) along the layer – on the layer band boundary between two layers (interlayer)

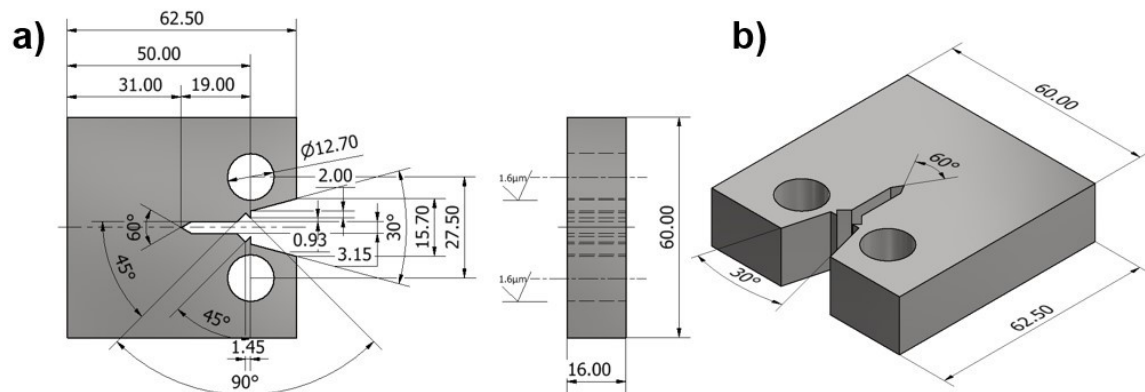


Figure 7-6 (a) C (T) specimen in accordance with ASTM E647-15 (b) Isometry details

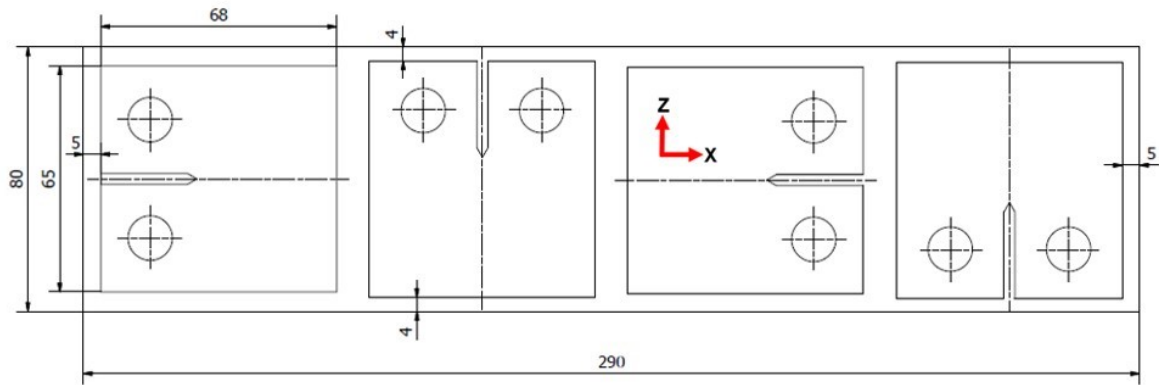


Figure 7-7 C (T) specimen machined outline

7.2.4 Fatigue crack growth rate characterisation method

Figure 7-6 presents the FCGR C (T) specimen samples used in the test. An Instron 8801 servo-hydraulic test rig (100kN frame capacity) was used. The FCGR test was conducted in air using a constant amplitude sinusoidal waveform, as shown in Figure 7-8. Specimens were pre-cracked using the K-decreasing method and a load ratio of $R = 0.1$ and loading frequency of 5Hz; the test was conducted under load control mode. The machine load cell was calibrated dynamically as recommended in BS standard [135] to a level of 1.5 times the maximum test load. The crack length was monitored using Baumer digital cameras. Fatigue data were acquired after the initial crack length exceeded 2.54mm (minimum). To eliminate crack initiation effects, P_{max} reduction was stepped in at 18, 16, 14, and 10kN. The pre-cracked length for a given C (T) sample was at least 1.6mm and no more than 3.5mm.

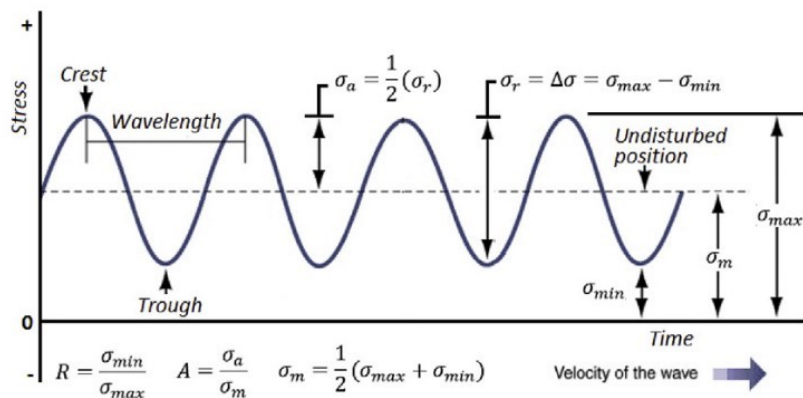


Figure 7-8 Sinusoidal waveform (constant amplitude cyclic loading)

7.2.5 Fatigue life characterisation method

The fatigue life determination was done with an Instron 8801 servo-hydraulic test rig with a capacity of 250kN. All the samples were tested with a sinusoidal waveform, as shown in Figure 7-8, under an applied stress ratio of $R = 0.1$. The fatigue strength was determined by the staircase method in which at least six pairs of the specimen were tested according to ISO 12107:2003 to raise the confidence level. A frequency of 15Hz was utilised. The fractured surface of the failed fatigue specimen was examined with the LYRA TESCAN as described in Section 3.4.3. The commencement stresses used for the fatigue testing of these samples are the average UTS values obtained from the static tensile test.

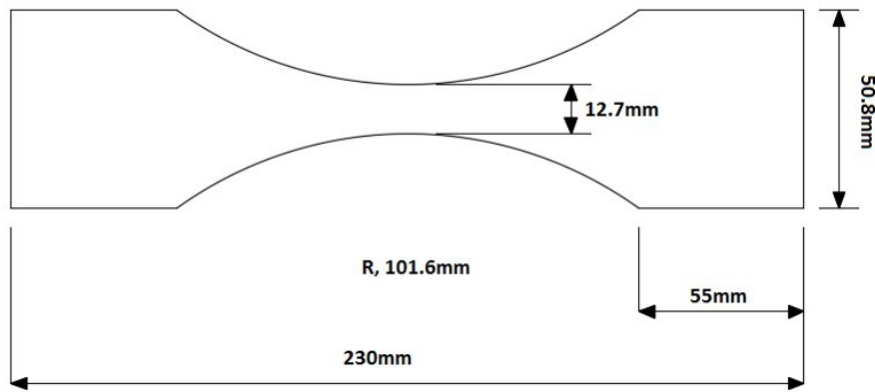


Figure 7-9 Fatigue sample following ASTM E466-18

7.2.6 Equations employed in data analysis for fracture toughness according to EPFM

The fatigue pre-cracking was done as per ASTM E1820 using a force value based on the force P_m .

$$P_m = \frac{0.4 B b o^2 \sigma_y}{2w + a_0} \quad (7-1)$$

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The fracture toughness of ductile material can be characterised using the elastic-plastic fracture mechanics parameter J as per ASTM E1820. J can be divided into an elastic component $J_{elastic}$ and a plastic component $J_{plastic}$.

$$J = J_{el} + J_{pl} \quad (7-2)$$

$$J_{(i)} = \frac{K_{(i)}^2 (1 - \nu^2)}{E} + \left(\frac{\kappa_{pl} A_{pl}}{B_N b_o} \right) \quad (7-3)$$

where $J_{(i)}$ is the instantaneous fracture resistance energy, $K_{(i)}$ is the instantaneous stress intensity factor calculated from the Eqs. (7-4) (7-5) below, E is the Young's modulus, ν is the Poisson's ratio, B_N is the net specimen thickness between the side grooves, b_o is the uncracked ligament, $(W - a_0)$ and $\kappa_{pl} = 2 + 0.522 b_o / W$ (*CT specimen*). a_0 is the initial crack length and W is the specimen width. A_{pl} is the plastic area under the load versus displacement curve, as described in ASTM E 1820.

$$K_{(i)} = \frac{P_{(i)}}{\sqrt{B_N b_o} \sqrt{W}} \cdot f\left(\frac{a_i}{W}\right) \text{ (MPa}\sqrt{\text{m}}) \quad (7-4)$$

$$f\left(\frac{a_i}{W}\right) = \left(2 + \frac{a_i}{W}\right) \left[0.886 + 4.64 \frac{a_i}{W} - 13.32 \left(\frac{a_i}{W}\right)^2 + 14.72 \left(\frac{a_i}{W}\right)^3 - 5.6 \left(\frac{a_i}{W}\right)^4 \right] / \left(1 - \frac{a_i}{W}\right)^{3/2} \quad (7-5)$$

The J versus Δa (incremental crack growth) also known as the J-R curve, was generated to find the fracture toughness properties of the WAAM HSLA steel. A blunting line was constructed in the J-R curve using Eq. (7-6), the slope of which

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was used for the exclusion lines at Δa offset of 0.15mm, 0.20mm and Δa_{max} which is the maximum allowable crack size in fracture analysis.

$$J = 2\sigma_y \Delta a, \text{ where } \sigma_y = \frac{\sigma_{UTS} + \sigma_y}{2} \quad (7-6)$$

The fracture toughness testing was done here using a single specimen approach where the specimen is subjected to a sequence of loading and unloading. The instantaneous crack length at each peak load, a_i was estimated using Eq. (7-7)

$$\frac{a_i}{W} = 1.000196 - 4.06319u + 11.242U^2 - 106.043U^3 + 464.335U^4 - 650.677U^5 \quad (7-7)$$

where;

$$U = \frac{1}{(B_e E C_{c(i)})^{1/2} + 1} \quad (7-8)$$

In Eq. (7-8) $C_{c(i)}$ is the load line crack opening elastic compliance and B_e is the specimen effective thickness calculated by

$$B_e = B - (B - B_N)^2 / B \quad (7-9)$$

where B is the specimen thickness

The CTOD (δ), which also denotes the fracture toughness of the WAAM deposit near the onset of the stable crack extension from the pre-existing fatigue crack is characterised using Eq. (7-10). δ can be added as the plastic and elastic components ($\delta = \delta_{elastic} + \delta_{plastic}$). Similarly, a $\delta - \Delta a$ resistance curve was generated. A blunting line of $\delta = 1.4\Delta a$ was constructed to generate exclusion lines at Δa offset of 0.15mm, 0.20mm and 1.5mm.

$$\delta(CTOD) = \frac{J_i}{m\sigma_y} \quad (7-10)$$

where J_i value is from Eq. (7-3) with $a = a_i$, the current crack size

$$m = A_0 - A_1 \left(\frac{\sigma_{YS}}{\sigma_{TS}} \right) + A_2 \left(\frac{\sigma_{YS}}{\sigma_{TS}} \right)^2 - A_3 \left(\frac{\sigma_{YS}}{\sigma_{TS}} \right)^3 \quad (7-11)$$

where $A_0 = 3.62$, $A_1 = 4.21$, $A_2 = 4.33$, $A_3 = 2.00$

7.2.7 Equation employed in fatigue crack growth rate analysis

The LEFM stress intensity (Δk) was calculated using Eq. (7-13), the same approach was used in the calculation of the crack propagation rate, where the values of N , a were measured by the machine and camera. The crack length, CFCGR, crack increment per cycle (da/dN), vs. the stress intensity factor range (SIFR), were plotted on a log-log graph. The (da/dN) was evaluated using the seven-point incremental polynomial method. The shape factor was calculated using Eq. (7-12)

$$f\left(\frac{a}{w}\right) = \left(2 + \frac{a}{w}\right) \left[0.886 + 4.64 \frac{a}{w} - 13.32 \left(\frac{a}{w}\right)^2 + 14.72 \left(\frac{a}{w}\right)^3 - 5.6 \left(\frac{a}{w}\right)^4 \right] / \left(1 - \frac{a}{w}\right)^{3/2} \quad (7-12)$$

$$\Delta K = \frac{\Delta P}{B\sqrt{w}} \cdot f\left(\frac{a}{w}\right) \text{ (MPa}\sqrt{\text{m}}) \quad (7-13)$$

where ΔP is the applied cyclic load range, B is the specimen thickness (16mm), W is the width of the specimen, a is the crack length, $f\left(\frac{a}{w}\right)$ is the specimen shape function. The existing crack in a material grows above a certain critical combination of nominal stress range and crack length known as the

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threshold, (ΔK_{th}) . Since region II (usually the linear part) in the log-log axes is generally used for design purpose, the equation describing this region is called the Paris law as given herein Eq. (7-14). This is obtained from a straight line drawn through the linear part of the curve.

$$\frac{da}{dN} = C \Delta K^m \quad (7-14)$$

where C and m are empirical power-law constants found by fitting a regression line to the test data.

$$\frac{da}{dN} = C \left(\frac{\Delta P}{B\sqrt{w}} X f\left(\frac{a}{w}\right) (MPa\sqrt{m}) \right)^m \quad (7-15)$$

$$\begin{aligned} \frac{da}{dN} = C \left(\frac{\Delta P}{B\sqrt{w}} \cdot \left(2 + \frac{a}{w}\right) \left[0.886 + 4.64 \frac{a}{w} - 13.32 \left(\frac{a}{w}\right)^2 + 14.72 \left(\frac{a}{w}\right)^3 \right. \right. \\ \left. \left. - 5.6 \left(\frac{a}{w}\right)^4 \right] / \left(1 - \frac{a}{w}\right)^{3/2} \right) (MPa\sqrt{m}) \right)^m \end{aligned} \quad (7-16)$$

$$\frac{\Delta a_i}{\Delta N_i} = \left(\frac{a_{i+1} - a_i}{N_{i+1} - N_i} \right) \quad (7-17)$$

7.2.8 Equation employed in fatigue life analysis

The S-N fatigue data were collected under laboratory conditions using the actual loading condition where the oscillatory stress is superimposed, as shown in Figure 7-8. The stress range, amplitude, mean and R are given by Eq. (7-18), (7-19), (7-20) and (7-21). A tension-tension cycle with R = 0.1 was used for the studies.

$$\sigma_r = \sigma_{max} - \sigma_{min} \quad (7-18)$$

$$\sigma_a = \frac{\sigma_{max} - \sigma_{min}}{2} \quad (7-19)$$

$$\sigma_m = \frac{\sigma_{max} + \sigma_{min}}{2} \quad (7-20)$$

$$R = \frac{\sigma_{min}}{\sigma_{max}} \quad (7-21)$$

7.3 Results

7.3.1 Mechanical and microstructural properties in WAAM Oscillatory deposition strategy

The variation in UTS and YS across the three WAAM steel structures in the X and Z direction using the oscillatory deposition strategies are shown in Figure 7-10. The minimum UTS and YS of these WAAM deposited structures are comparable to their wrought categories. A typical comparison of the anisotropic behaviour of the oscillated WAAM steel deposit is shown in the stress-strain curve presented in Figure 7-10. It is clear from the plot that the strength properties of the oscillatory strategy are higher in the X-direction, but ductility is better in the Z-direction.

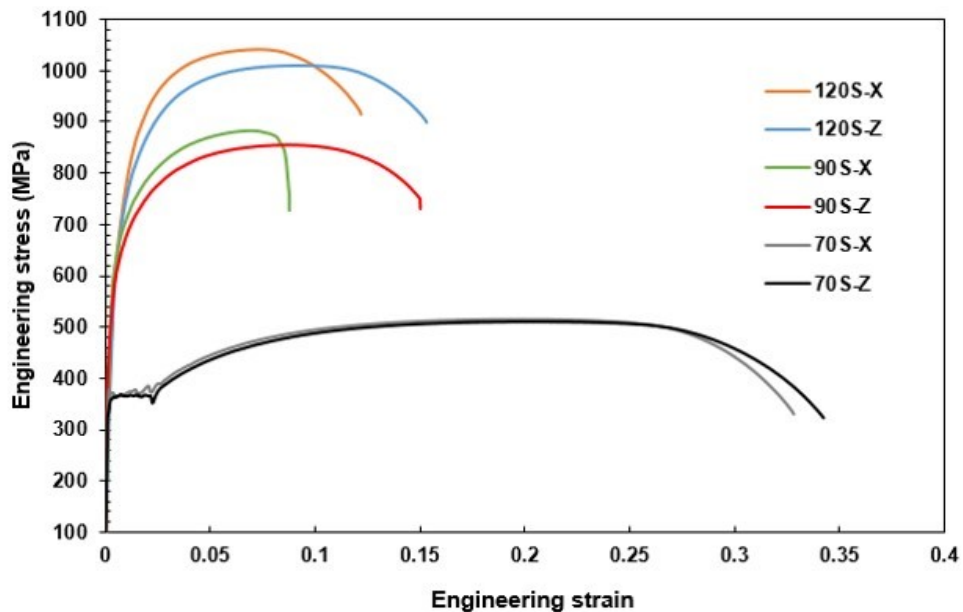


Figure 7-10 The stress-strain curve obtained in the oscillatory strategy

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Figure 7-11 present the hardness variation in the WAAM oscillated structure, while the microstructural variation obtained is shown in Figure 7-12, Figure 7-13 and Figure 7-14. The average hardness obtained for the ER70S-6, ER90S-B3, and ER120S-G is 268HV, 262.3HV, and 326.56HV, respectively. The prolonged thermal cycle and the increased austenite grain size are responsible for the increases in hardenability of these materials. A similar situation has been reported in [137,138]. The layer stacking and thermal cyclic heating which causes the hardening and softening of adjacent layers is responsible for the hardness variation, as seen in Figure 7-11; similar hardness trends were reported in a WAAM structure built with HSLA steel and maraging steel [51][55]. The ER120S-G microstructure, as revealed in the SP Figure 4-20(e & f), presents a significantly martensitic lath (ML), and quasi polygonal ferrite (QPF) formed as a result of fast cooling, while the OS (Figure 7-14) contains some martensitic austenite (M-A), acicular ferrite (AF) and intercritical ferrite (IF) formed as a result of the slow cooling rate due to the process.

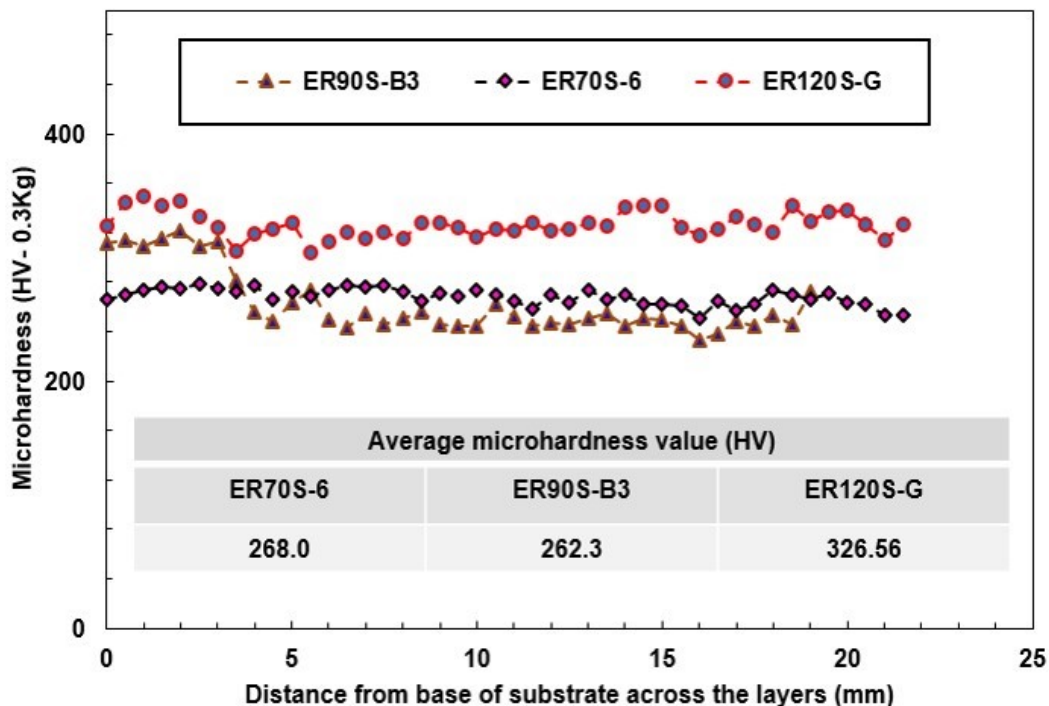


Figure 7-11 Hardness variation in WAAM oscillated strategy

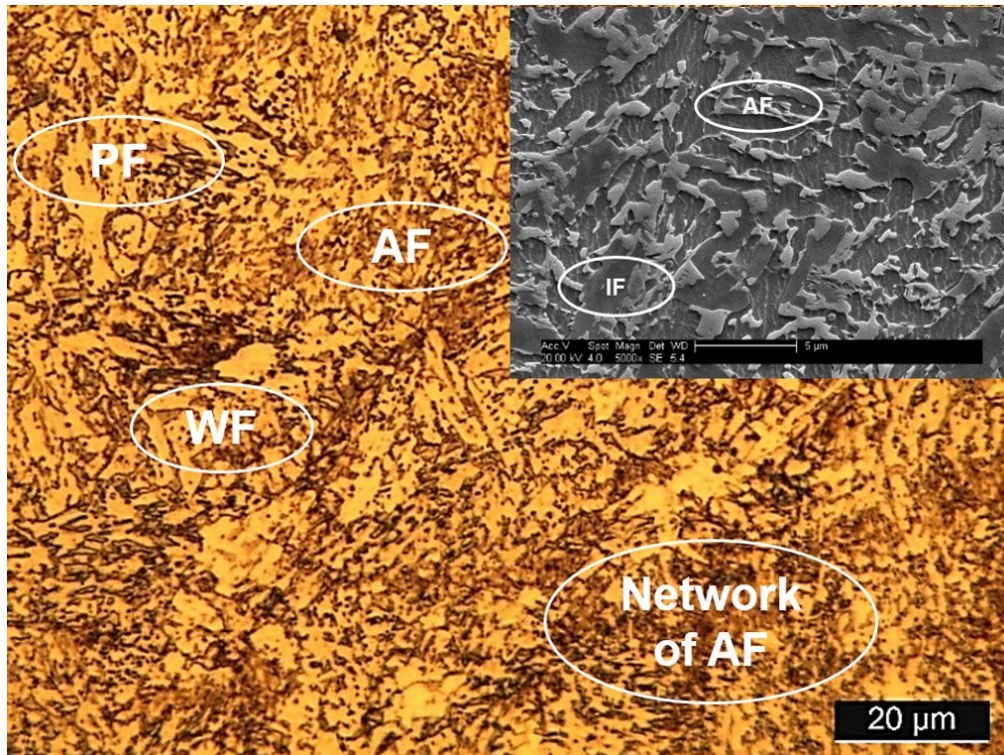


Figure 7-12 ER70S-6 WAAM oscillated

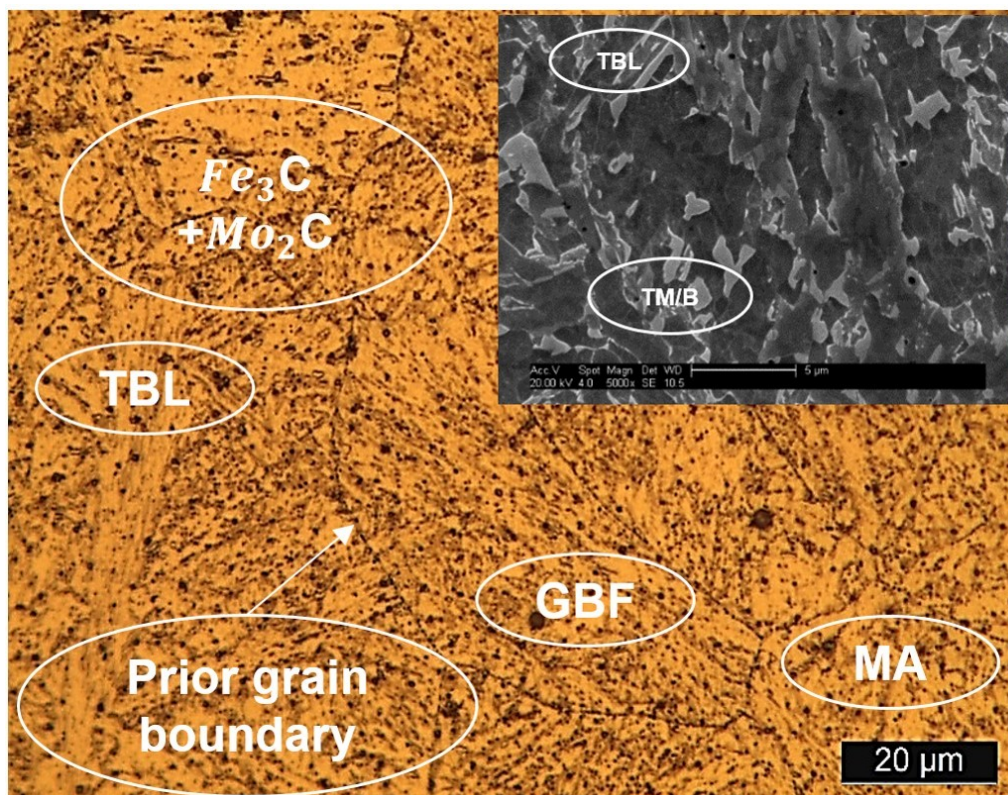


Figure 7-13 ER90S-B3 WAAM oscillated

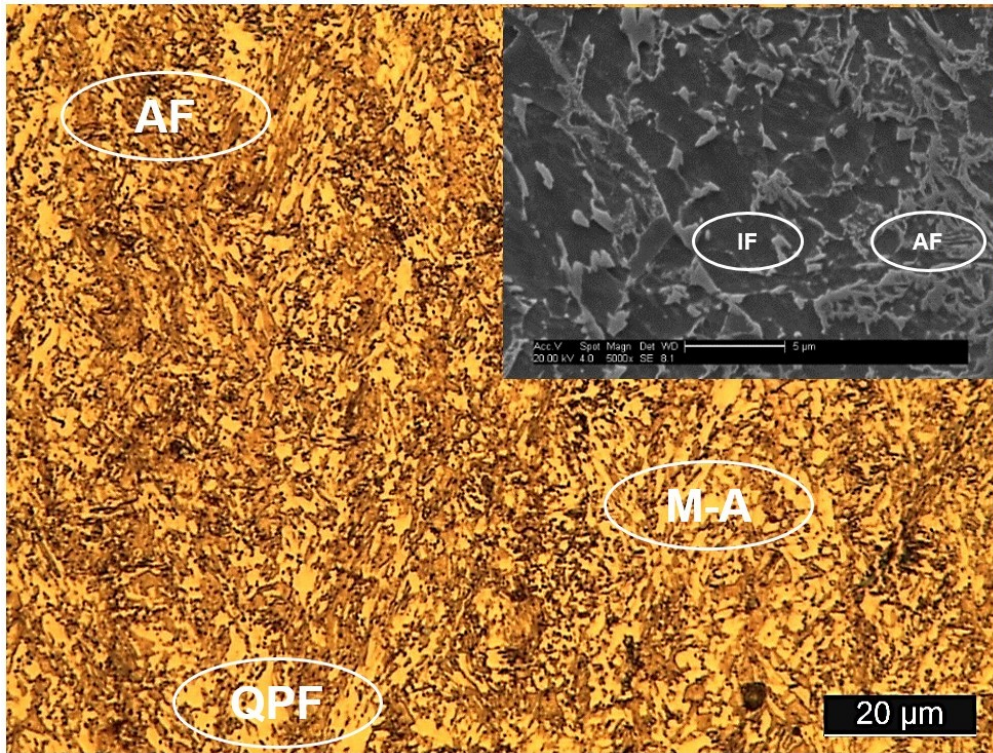


Figure 7-14 ER120S-G WAAM Oscillated

7.3.2 J_{Ic} / CTOD data for WAAM HSLA steel

C (T) specimen were used for the testing, although when compared with SEN(B) specimen, they have slight higher constraint and would give lower fracture toughness for the same material thickness and temperature compared to SEN(B). Hence results obtained here could not be compared to other works done with SEN(B) samples. Figure 7-15 (a & b) presents the loading and unloading load-line displacement (LLD) and J-R resistance curve based on J-integral. Figure 7-16 presents the typical J_{Ic} plot with exclusion lines and the J_q values in the different crack orientations for the various WAAM deposited steel components. Figure 7-17 and Figure 7-18 present the displayed anisotropy in J_q and CTOD. The maximum J_q values which represent the crack initiation toughness corresponding to the various WAAM deposited steels components are presented in Table 7-2.

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Figure 7-19 presents the macro image showing layer bands and fractured cracked path in WAAM deposited steel components, as shown in Figure 7-5. Fracture in the Z-direction (across the layers) could only occur in crack path Z, while fracture in the X-direction (along the layers) could occur in crack path X1 or X2. The variation in toughness, as given by the J_q values, with respect to crack orientation is evident in the LLD and J-R curves.

The observed fracture toughness across the build direction (layer bands) present the highest resistance to fracture failure, followed by a fracture in the welding direction (X-direction). The X1 path through the hard zone showed the least toughness; the X2 path possesses more resistance to fracture than the X1 path, being the annealed, tempered and refined path. The J-R plots indicated higher resistance to crack propagation in the ER70S-6 WAAM component compared to other WAAM steels, as given in Figure 7-17.

The J_q value for ER90S-B3 WAAM component was higher in the Z-direction compared to the X-direction, although the highest J_q value for this component is significantly less compared to the ER70S-6 WAAM component. The average J_q values for ER120S-G and ER70S-6 were very close but significantly higher than the ER90S-B3 WAAM steel. The J_q value for ER120S-G in the Z-direction is about 1.1 & 2.3 times more than the value in X2 and X1 fracture orientation respectively, as reflected in Figure 7-17. It is worth noting that the J_q values are generally higher in the Z-direction as the annealed and tempered region of the layer band serves as a crack arrest region retarding the fracture; this was also reported with Ti-6Al-4V [141]. The CTOD values shown in Figure 7-18 follow the same trend as the J_q values.

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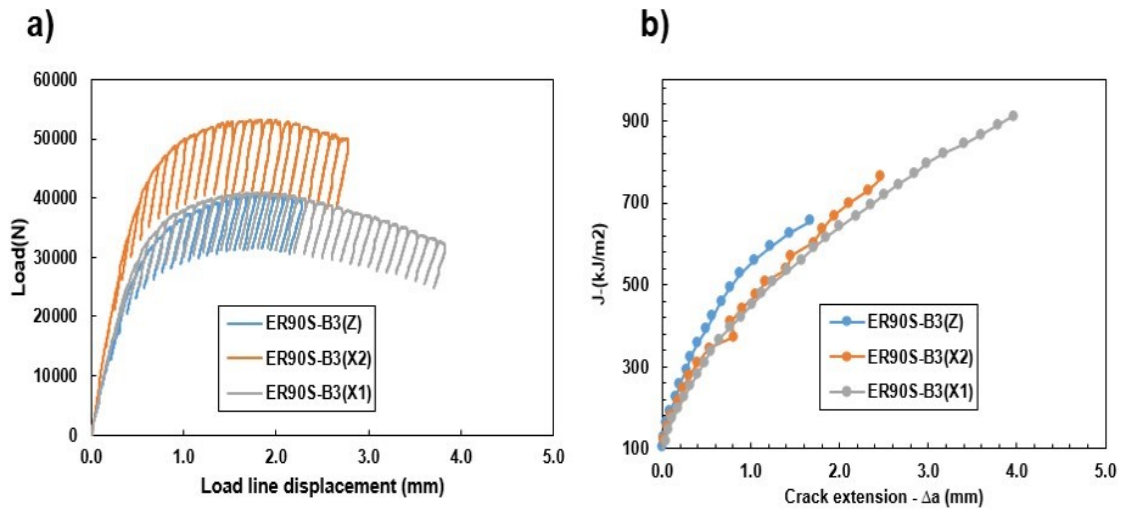


Figure 7-15 WAAM deposited structures of ER90S-B3

(a) Loading and unloading curve (b) J-R curves

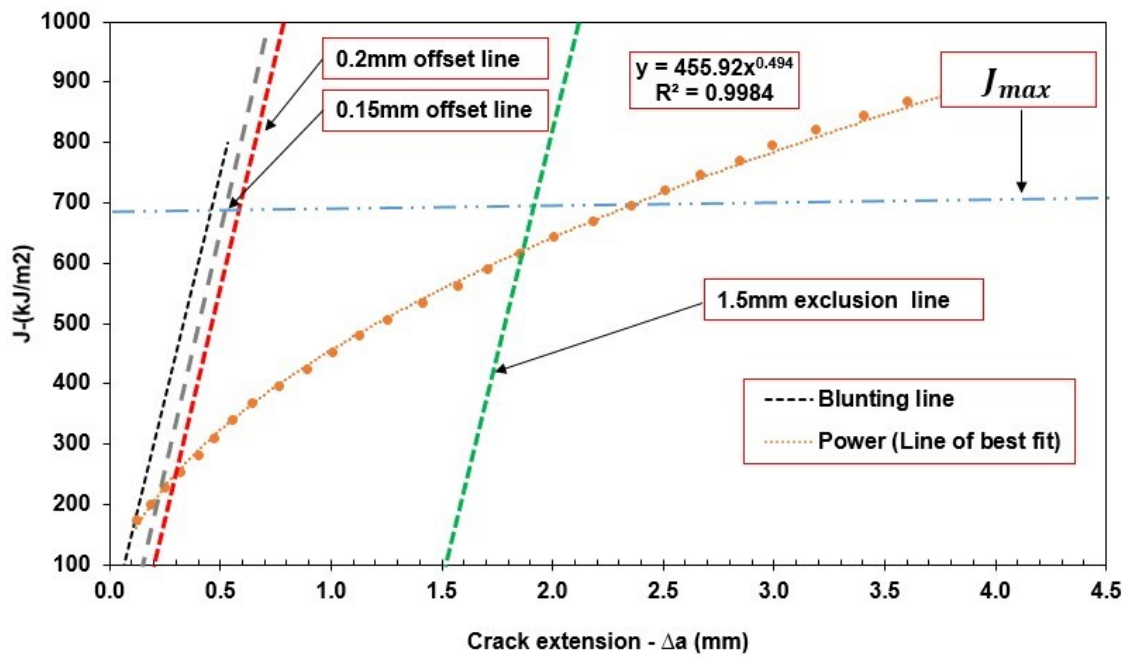


Figure 7-16 J_{IC} plot illustrating the exclusion lines, J_{max} & J_{IC} and J_q

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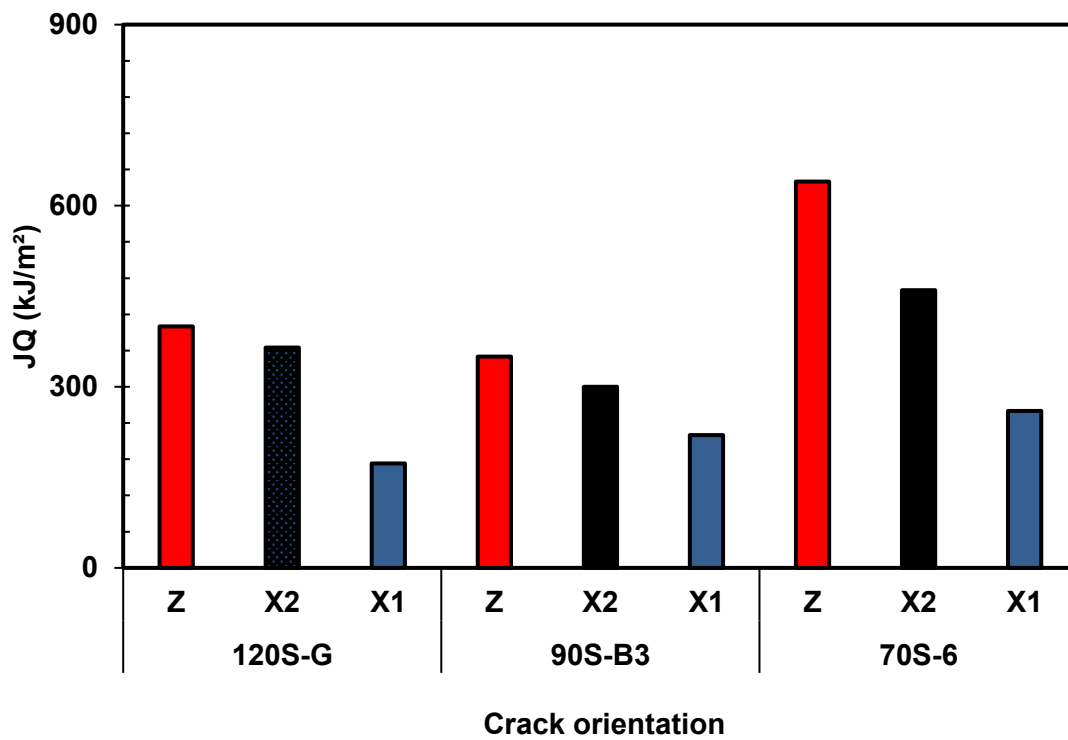


Figure 7-17 J_{IC} plot values for the different fracture orientations in WAAM

J1c testing in different orientations are shown in appendices (fig.4)

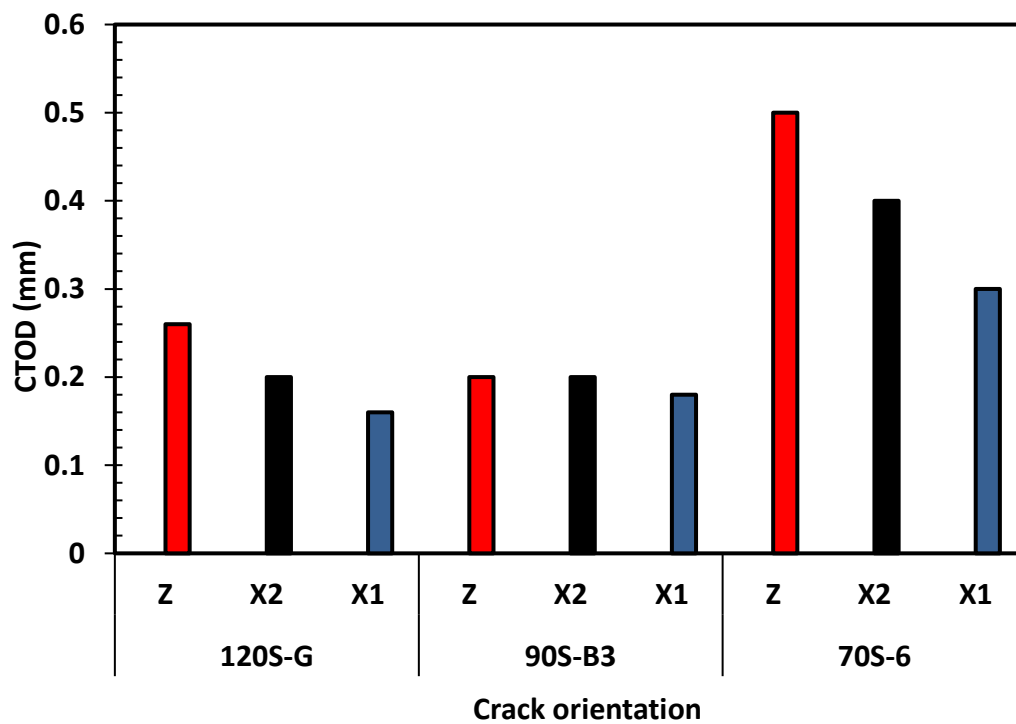


Figure 7-18 CTOD values for the different fracture orientations in WAAM

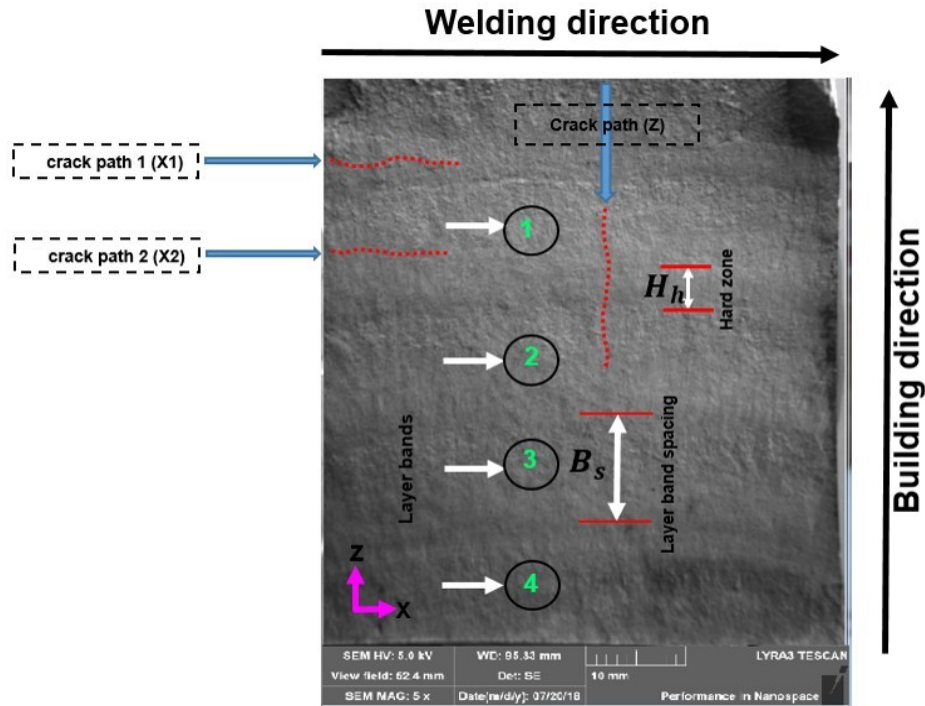


Figure 7-19 Macro images showing layer bands and fracture crack path

J1c fractured samples are shown in appendices (fig.5 & 6)

Table 7-2 Summary of fracture toughness test values

WAAM steel grade	Max, J_Q (kJ/m ²)	Crack orientation	Avg J_Q (kJ/m ²)	Max J1c (MPa.m)	C	X	Δa max	R^2
120s	400	Z-Dir	313	0.40	940.05	0.6127	2.8	0.99
90s	350	Z-Dir	290	0.35	541.73	0.4568	3.5	0.99
70s	640	Z-Dir	453	0.64	1148.5	0.451	2.56	0.98
S355G10	780	Wrought	780	0.78	1045.2	0.6621	2.4	0.99

7.3.3 Fatigue crack growth rate data for WAAM HSLA steel

Figure 7-20 presents the fatigue crack growth rate (FCGR) plots of the WAAM steels and S355 TMCP. Fatigue crack growth rate data for these WAAM steels fall within a relatively narrow scatter band, although they behave similarly to S355 TMCP.

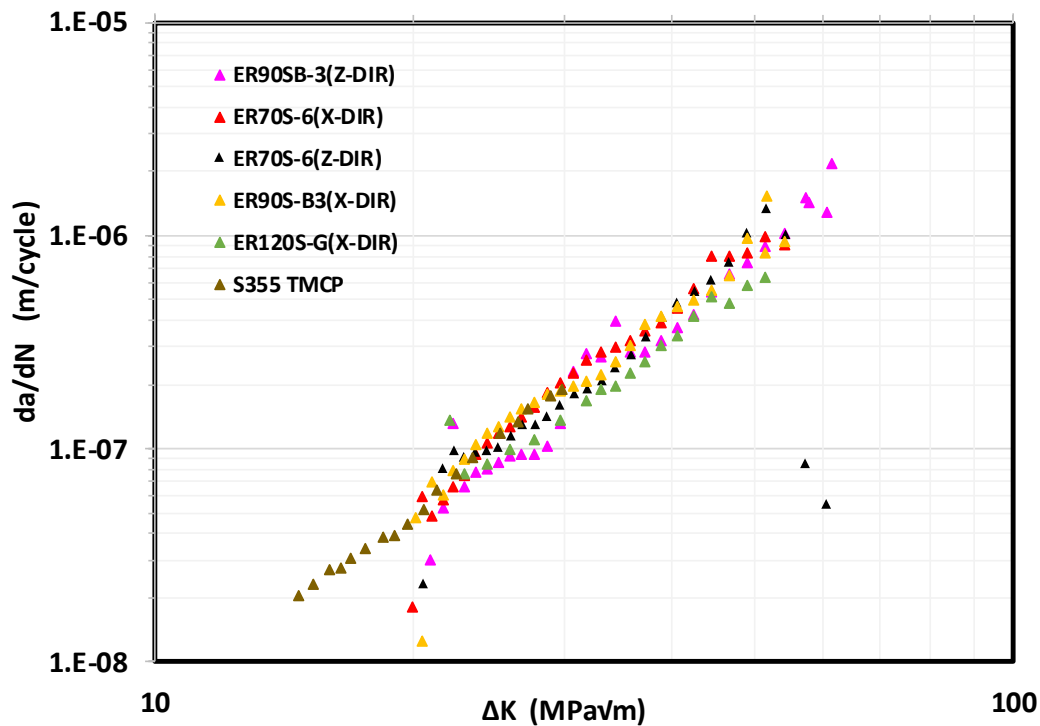


Figure 7-20 FCGR of WAAM HSLA steel

CT specimen for fatigue crack growth rate is shown in appendices (fig.3)

7.4 Discussions of results

7.4.1 Effects of WAAM deposition strategy- Oscillatory

The SP strategy, used in Chapter 4 (Figure 4-3 and Figure 4-4) which was deposited with layer by layer single bead molten steel, has one unique TS and cools very fast as a result of more heat conduction routes along the build; hence its limited deposited thickness, compared to the OS strategy, whose actual TS along the length of deposit is relatively slow. Despite having the same heat input, its advancement speed in the oscillation direction (wider width) makes it retain localised heat accumulation as a result of the torch oscillating at a location for a longer time. This is responsible for the variation in microstructure and mechanical properties. The key influencing factor in the mechanical behaviour of the WAAM

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deposit is the chemistry of the wires, which gives way to different microstructural formation as a result of the cooling rate imposed by the WAAM OS deposition strategy. The comparison of the relationship in mechanical properties between OS and the SP strategy are depicted in Figure 7-21 (a-e). The SP showed higher UTS and YS with ER120S-G & ER90S-B3 deposits, with average values of 987MPa and 870MPa, respectively. ER90S-B3 UTS, YS & PCE obtained here were typical of F22 wrought steel.

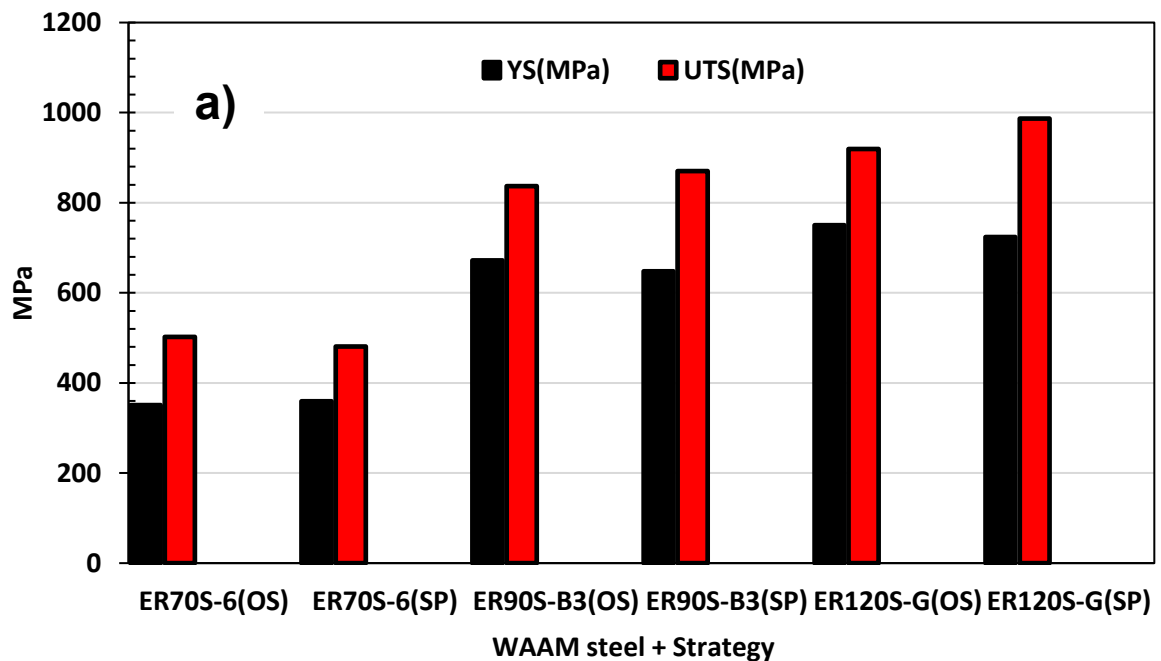
The ER70S-6 in the OS had higher UTS but lower YS and PCE; this is typical of A36 wrought [136]. It is also worth noting that the mechanical behaviour of WAAM deposit is due to the direction of heat flow, grain growth, layer stacking, and thermal gradient obtained during the deposition process. The Cr, Mo, and Ni in this alloy composition increased the hardenability, giving high strength non-equilibrium phases martensite and bainite (M&B). That is the reason why the trend of hardness as shown in Figure 7-21 (c) is on the increase from ER90S-B3 to ER120S-G. The wires used for this experiment have low carbon content for improved weldability, as shown in Table 3-1.

The microstructure of the ER90S-B3 OS, as presented by Figure 7-13, is coarser than the SP Figure 4-20(c & d)) with some clearly defined prior grain boundaries showing some grain boundary ferrite (GBF) decorated with finely dispersed carbides ($\text{Fe}_3\text{C} + \text{M}_{\text{O}_2}\text{C}$) as a result of the cooling rate and prior composition. This is detrimental to fracture toughness, as the structure provides a pathway for crack propagation [139]. The tempered bainite lath (TBL) was formed as a result of the successive softening and hardening of adjacent layers. The microstructure presented by the ER70S-6 SP (Figure 4-20 (a & b)) contains mostly polygonal ferrite (PF) and pearlite (P), while the OS (Figure 7-12) is coarser with some PF and a network of basket weave AF nucleated intragranularly from the non-metallic inclusion sites. This basket weave feature has been reported to be a hindrance to crack propagation but supportive of increasing fracture toughness [140].

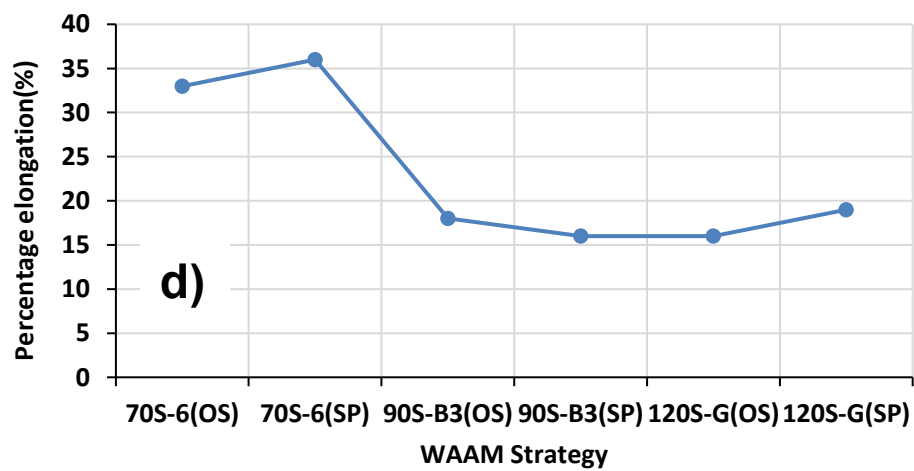
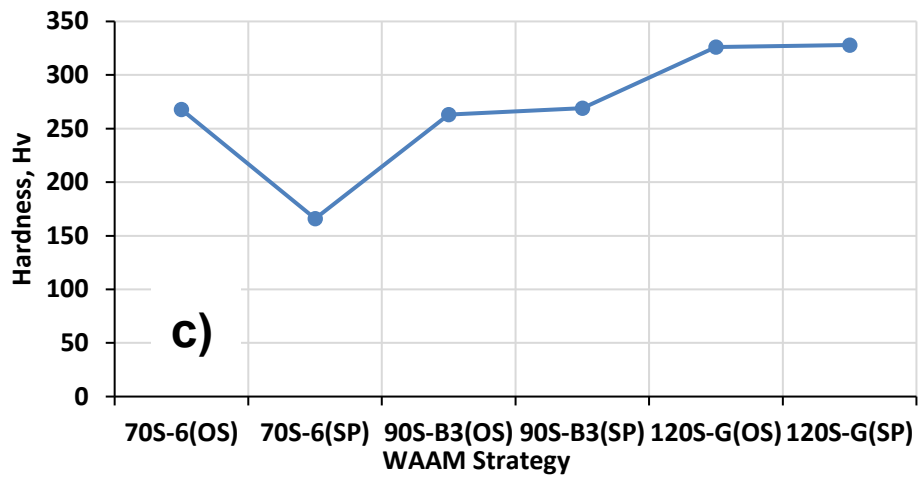
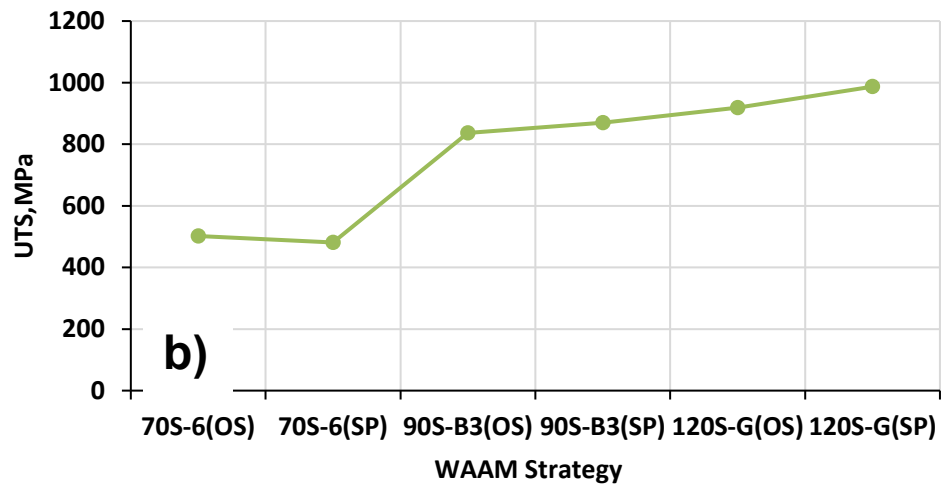
The grain sizes are larger in the OS than the SP as shown in Figure 7-21 (e) owing to the heat accumulation effects in the OS, which favours grain growth. In both deposition strategies, they are in the increasing order (ER120S-G, ER90S-

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B3 & ER70S-6). According to the Hall-Petch equation, relating yield strength and hardness to grain sizes, $\sigma_y = \sigma_o + Kd^{-1/2}$ and $H = H_o + K^1d^{-1/2}$ where σ_y , H & d are the yield strength, hardness and grain size and σ_o , H_o , K & K^1 are constants independent of grain size. A larger grain size will decrease the mechanical barrier to having a complete martensitic transformation; hence, the reduction of grain size leads to an increase in hardness and yield strength (fine grain strengthening mechanism). The carbide forming elements such as Cr & Mo in the original chemistry result in secondary hardening, which, in alliance with the refined dispersion of carbides, benefits the toughness. In the case of ER90S-B3, the heat accumulation in the oscillatory strategy leads to the formation of coarse carbides arranged in large lumps, which lower the fracture toughness in the re-melt zone. The precipitates further dissolved back into the solid solution with smaller uniform distributed carbides of $M_{O_2}C$ in the tempered region, thus create the recrystallised, refined, soft and homogeneous microstructure, which aids higher strength and fracture toughness as possessed by the refine zone.



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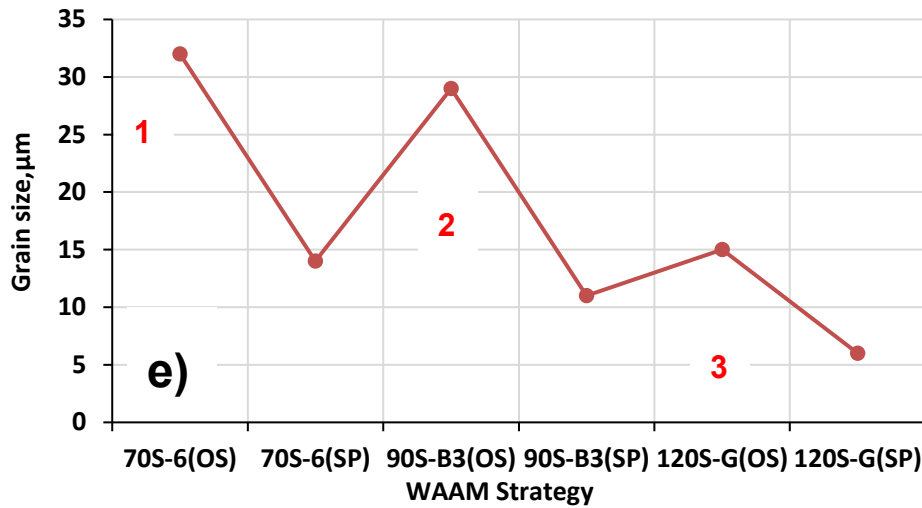


Figure 7-21 Comparison of Oscillatory with Single pass strategy

(a) average of strength for both strategies (b) UTS (c) Hardness (d) PCE (e) Grain size

7.4.2 The influence of oscillatory deposition strategy on microstructural changes and hardness variation

Figure 7-22 presents the microstructural changes taking place in the melt zone and refine zone of WAAM HSLA steel. The cyclic heating of these WAAM steels, due to the sequential layering strategy (oscillatory strategy), results in the variation in peak temperature and cooling rates of the adjacent layer in the building direction. The variation in peak temperature, which was as a result of the cooling rate, is dependent on the WFS, TS (heat input) and the oscillatory dwell time and leads to the formation of layer bands.

These bands have distinctive regions which, on subsequent cyclic heating, cause annealing and softening of the adjacent layers. This phenomenon was also reported for Ti-6Al-4V by [142] and further investigated by [143]. The layer band spacing, which is equivalent to the layer height, causes variability in microstructure and hardness of deposit, as shown in Figure 7-11, Figure 7-12, Figure 7-13 and Figure 7-14. This trend is also similar to phase changes in the

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HAZ of steel welding observed by [93]. The prolonged thermal cycle in the oscillatory strategy resulted in grain coarsening, which is absent in a single pass strategy, as observed in [53], and higher hardness as a result of an increase in austenite grain size, as reported by [147][148], which also increases hardenability.

The oscillatory strategy has a feature of increased thermal mass and a slower cooling rate, giving enough time for carbon to diffuse into the austenite. The re-melt zone in ER90S-B3 is composed of a lathy microstructure of martensite, martensitic austenitic constituent and grain boundary ferrite, while their refine region is composed of tempered martensite and bainite with bainitic pearlite, as shown in Figure 7-22(a & b). ER120S-G in the re-melt zone is comprised of Widmanstätten ferrite, lath martensite and lath austenite, while their refine region is composed of tempered martensite and acicular ferrite and grain boundary ferrite, as shown in Figure 7-22(c & d). The re-melt region in ER70S-6 is composed of quasi polygonal ferrite, martensite, and some acicular ferrite, while the refine region is composed of equiaxed ferrite, chains of acicular ferrite and Widmanstätten ferrite, as shown in Figure 7-22(e & f)

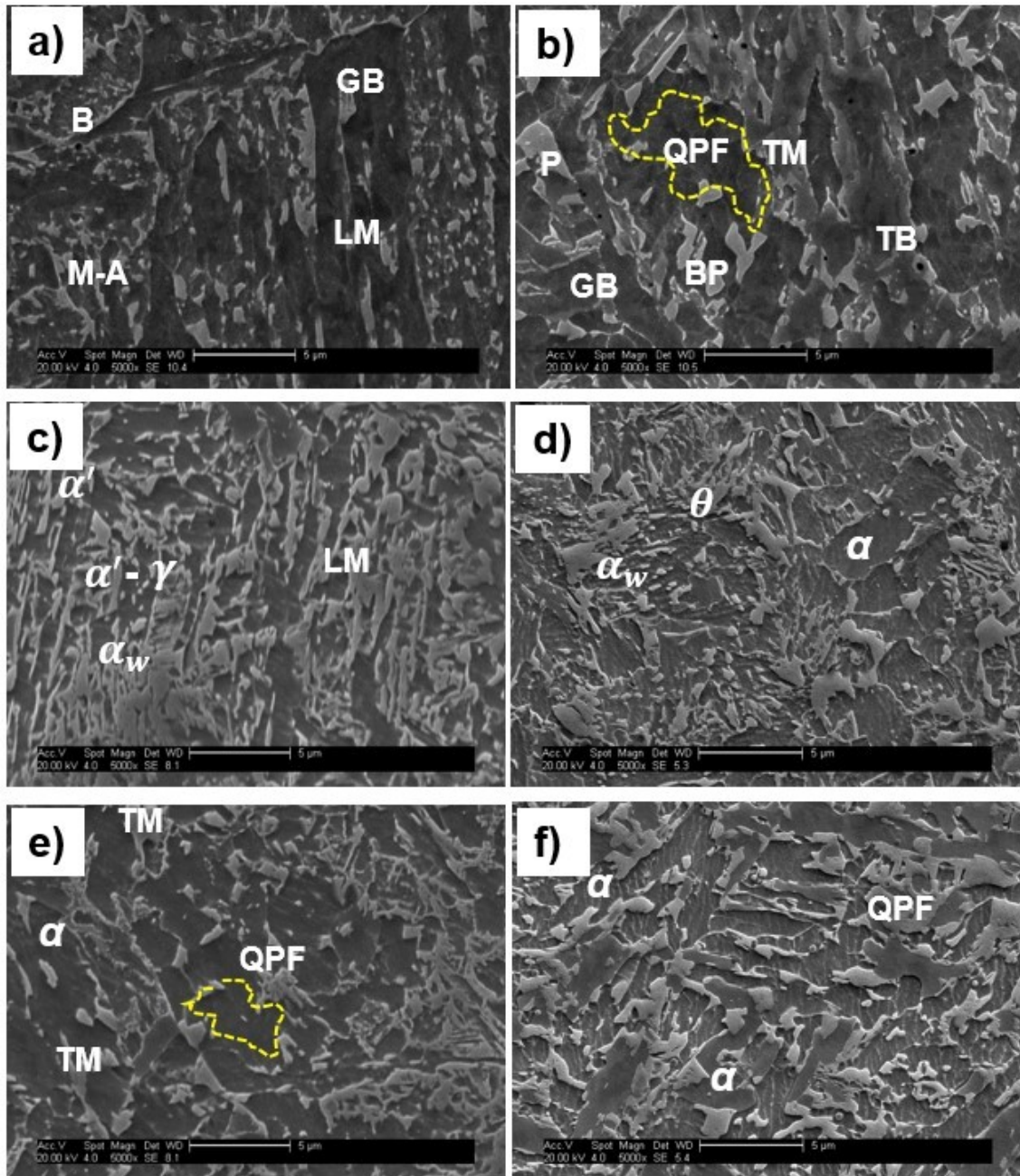


Figure 7-22 Characterised SEM optical microstructure of WAAM HSLA steel

(a-b) ER90S-B3 (c-d) ER120S-G (e-f) ER70S-6 (a, c & e) Re-melt zone (b, d & f) Refine region

7.4.3 Texture, crystallographic orientations and phase formation

EBSD characterisation was used to investigate the crystallographic orientation in the oscillatory deposit at specific points of interest to confirm the intrinsic relationship between the crystal growth direction and the thermal history experience by the deposited structure during the building. Figure 7-23 presents the EBSD maps showing ferritic, martensitic grain orientations from the refine region on the layer band and from the re-melt region close to the re-melt line in the layer bands. Figure 7-24 presents the corresponding $\{100\}$, $\{110\}$, $\{111\}$ pole figures.

The grain boundaries distribution and volume statistics of HAGBs and LAGBs is present in Figure 7-26 and Table 7-3. The pole figure misorientation angle of the BCC phases and the EBSD phase mapping are shown in Figure 7-25 & Figure 7-27. The legend showing the relationship between the crystal orientation and colours is inserted in the top upper left corner of Figure 7-23(a). The grain size characterisation in the oscillatory strategy of these WAAM HSLA steels is shown in Figure 7-28.

The re-melt zone in ER70S-6, ER90S-B3 & ER120S-G consists of epitaxial elongated grown lathy martensitic grains with average length greater than $60\mu\text{m}$ emanating from the solid-liquid interface. The average grain sizes in this zone ($38\mu\text{m}$, $34\mu\text{m}$ and $18\mu\text{m}$) as shown in Figure 7-28, are relatively higher than the average grain size in the refine zone ($26\mu\text{m}$, $24\mu\text{m}$ and $11\mu\text{m}$). The oscillatory grain sizes are much bigger and coarser than the single-pass grain sizes, as shown in Figure 4-24, owing to the enormous heat accumulation effects observed in the oscillatory pass.

The heat input in the single pass strategy is determined by the deposition parameters of the geometry, and since it operates in a linear motion, its cooling is more rapid. A strong texture intensity in the melt-zone is observed in the ER70S-6 & ER90S-B3, which indicated that an anisotropic martensitic lath microstructure in the random crystallographic texture was obtained as evident in the pole figure shown in Figure 7-24(a-b) & (c-d). The preferential grain growth direction during solidification in ER70S-6, ER90S-B3 and ER120S-G is the $\{100\}$ & $\{101\}$ orientation, and this is contrary to the single pass deposition strategy of

the ER90S-B3 and ER120S-G, as shown in Figure 4-25, where no preferential orientation was observed in ER90S-B3 & ER120S-G.

EBS for the phase identification reveals higher percentages of RA in the re-melt zones shown in Figure 7-27(a & c). This zone also possesses larger grain growth, as evident in Figure 7-31. According to the Hall-Petch equation relating yield strength to grain sizes, $\sigma_y = \sigma_o + Kd^{-1/2}$, where σ_y is the yield strength, d is the grain size, σ_o & K are the material constants. A larger grain size will decrease the mechanical barrier to have a complete martensitic transformation, and this is supportive of the reason why poor fracture toughness properties are noticed in this zone (X1), as evident in Figure 7-17 and Figure 7-18.

Retained austenite has been known to be detrimental to fracture mechanics properties. Hence the hardness of ER70S-6 in oscillatory strategy is considerably higher than the single pass deposition strategy, as shown in Figure 4-17, due to the consequences of this repetitive thermal cycle. The heat accumulation in the oscillatory strategy was responsible for the dissolution of the carbides and nitrides particles which are supposed to be effective for grain boundary peening and strengthening. Fe_3C and Mo_2C precipitates were observed in ER90S-B3.

In the refine zone, the grain size is smaller, although with a low volume of retained austenite. The precipitates dissolved back into solid solution, thus creating the recrystallised, refine, soft and homogeneous microstructure as noticed in Figure 7-22(b, d & f), hence the higher fracture toughness properties possessed by this zone (X2), as evident in Figure 7-17 and Figure 7-18.

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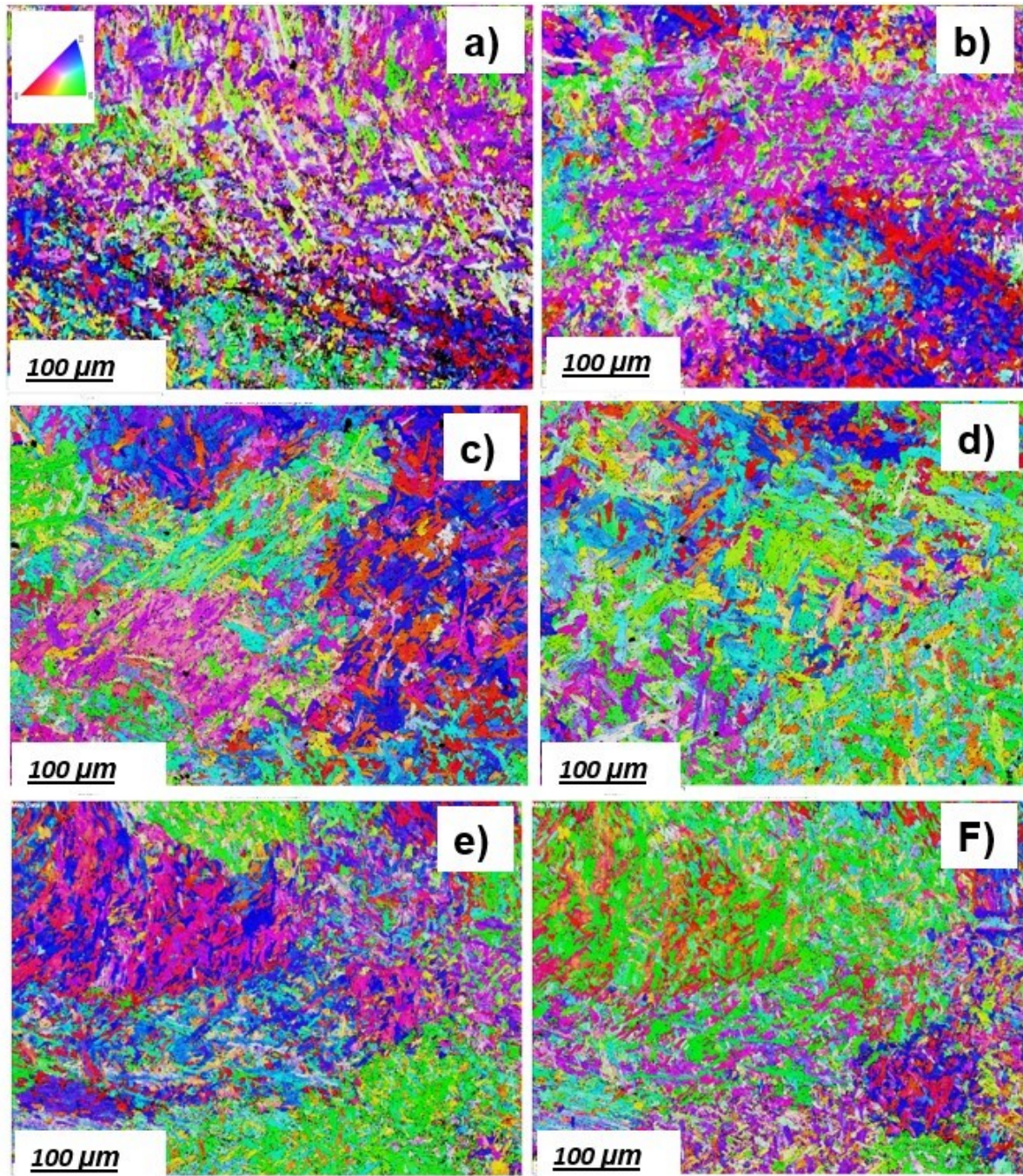


Figure 7-23 EBSD Maps showing martensitic orientation in

70s (a-b), 90s (c-d) and 120s (e-f): (a, c & e) re-melt zone; (b, d & f) refine region

Fracture and Fatigue Behaviour of WAAM Components

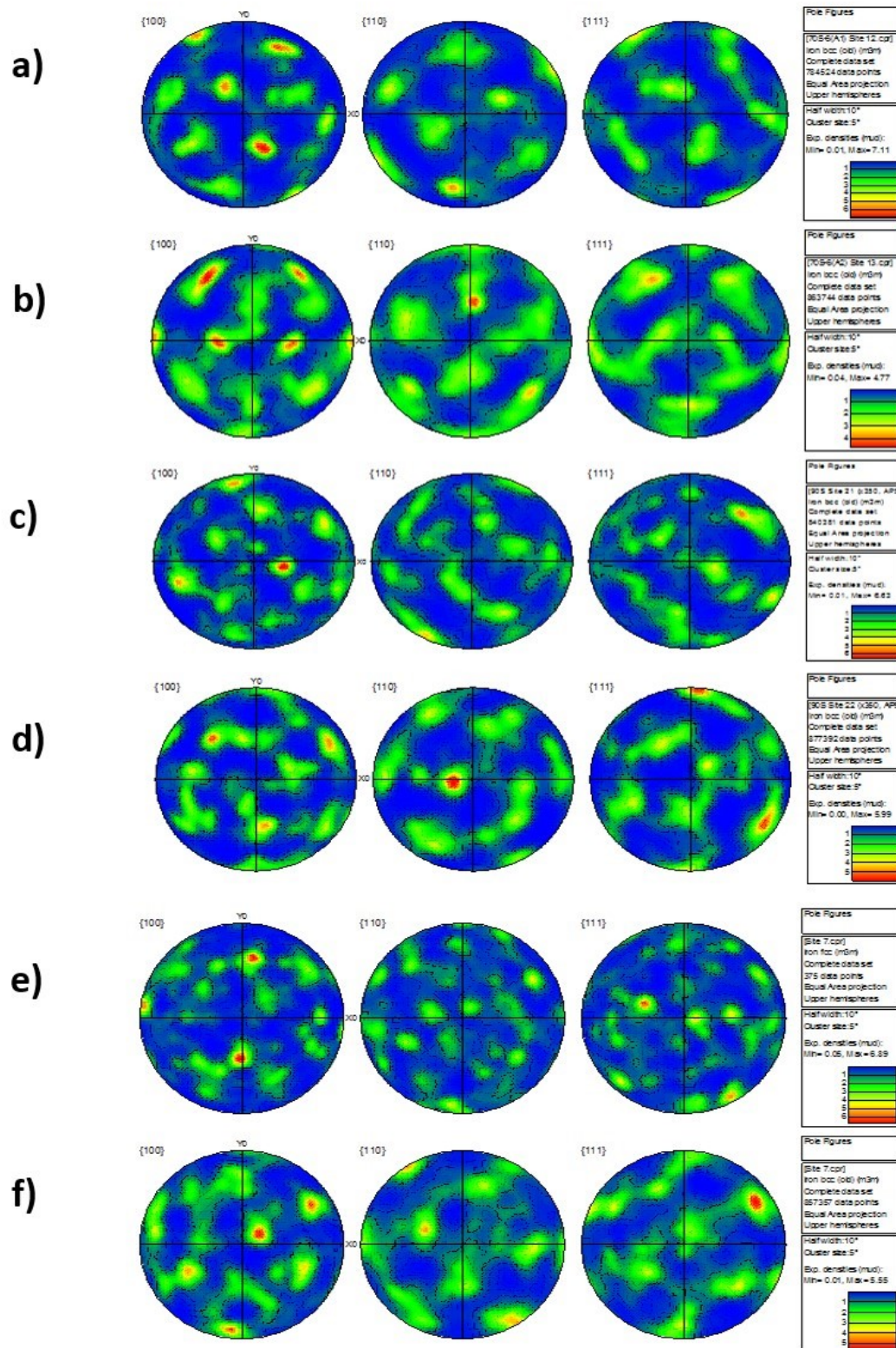


Figure 7-24 Pole figures of BCC phase in

70s (a-b), 90s (c-d) and 120s (e-f): (a, c & e) re-melt zone; (b, d & f) refine region

Fracture and Fatigue Behaviour of WAAM Components

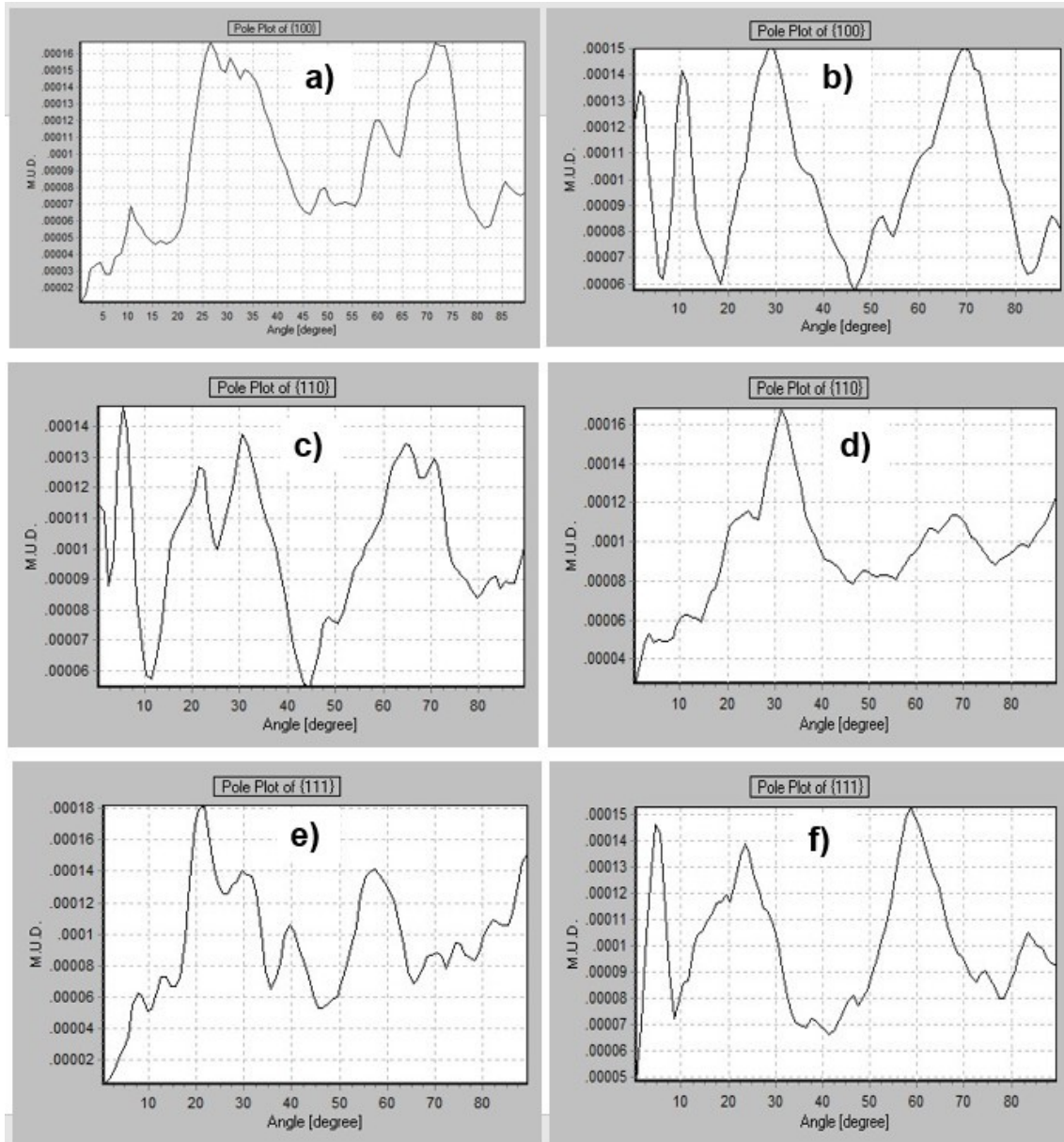


Figure 7-25 Pole figures misorientation angle of BCC phase in

70s-6 (a, c & e), 70s-6 (b, d & f): (a, c & e) re-melt zone; (b, d & f) refine region

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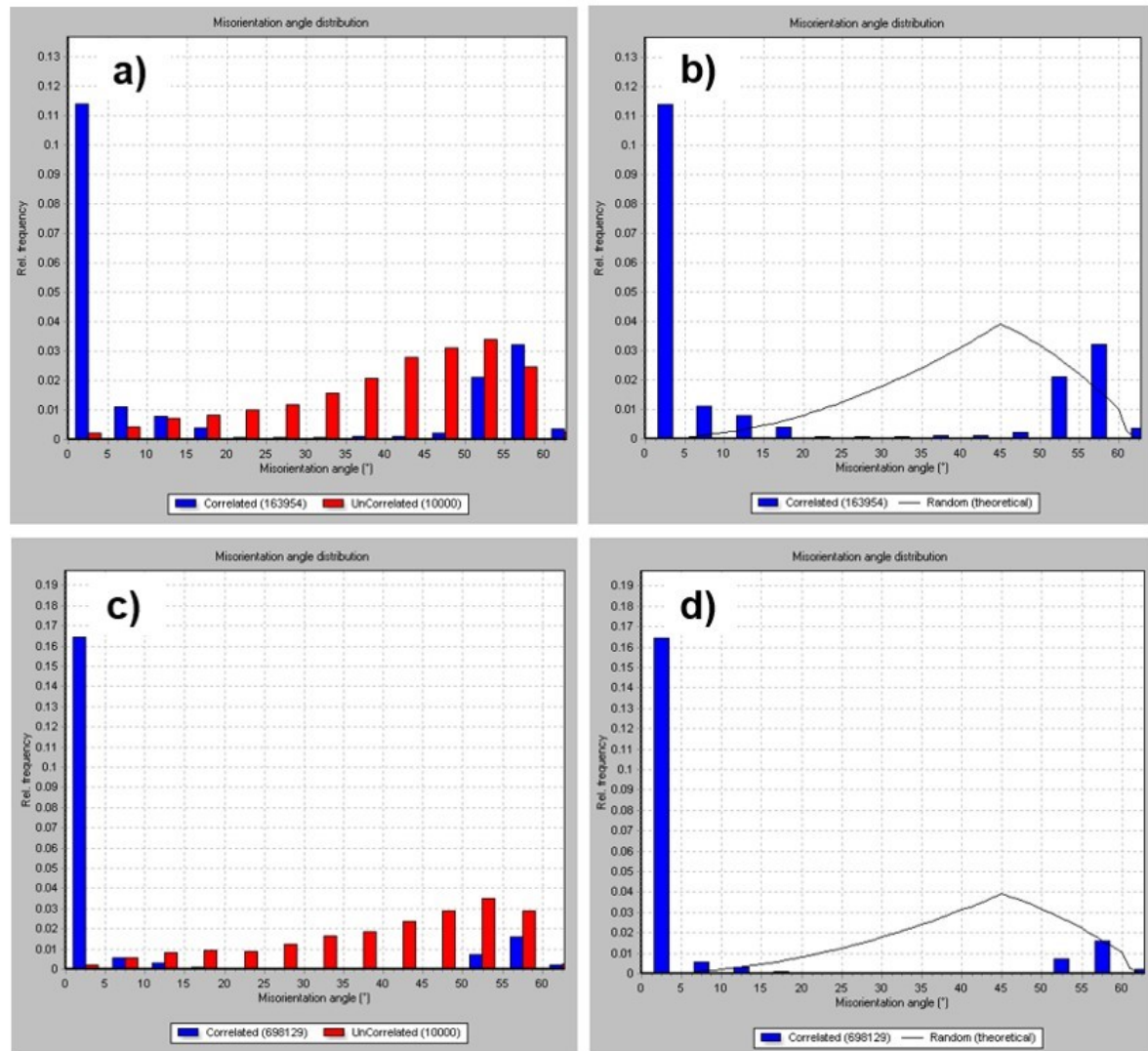


Figure 7-26 Grain boundaries distribution in hard zone

70s-6 (a & b), 90s-B3 (c & d)

Table 7-3 Relative frequency of HAGs and LAGBs in the hard zones

WAAM steels	Grain boundaries	
	LAGs (<15°C)	HAGs (>15°C)
70S-6	52.36	47.64
90S-B3	81.12	18.88
120S-G	87.69	12.31

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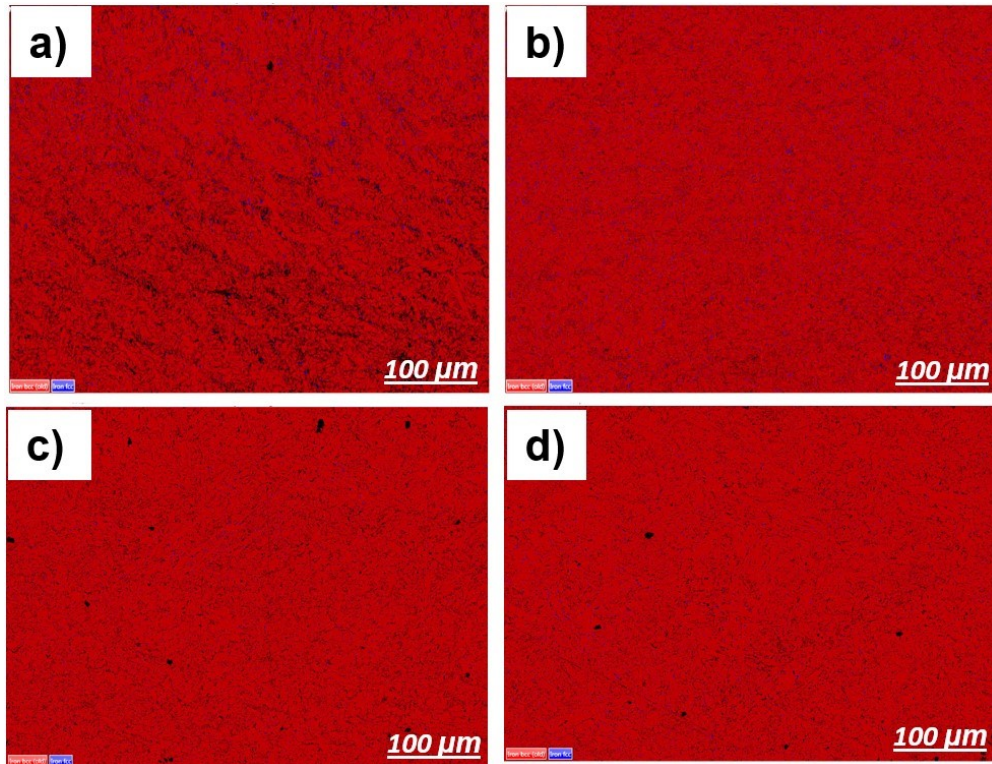


Figure 7-27 EBSD phase mapping for WAAM HSLA steels showing BCC (Martensite) and FCC (Austenite) in red and blue respectively.

(a-b) (70s), (c-d) (90s); (a & c) re-melt zone; (b & c) refine region

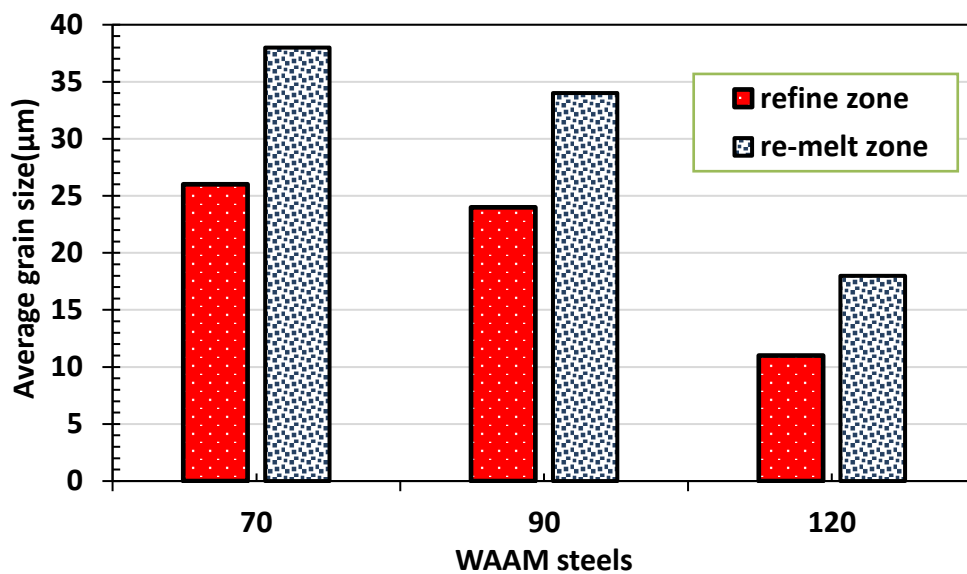


Figure 7-28 Grain size measurements for WAAM

(a) Refine zone (strips) (b) Re-melt zone (light)

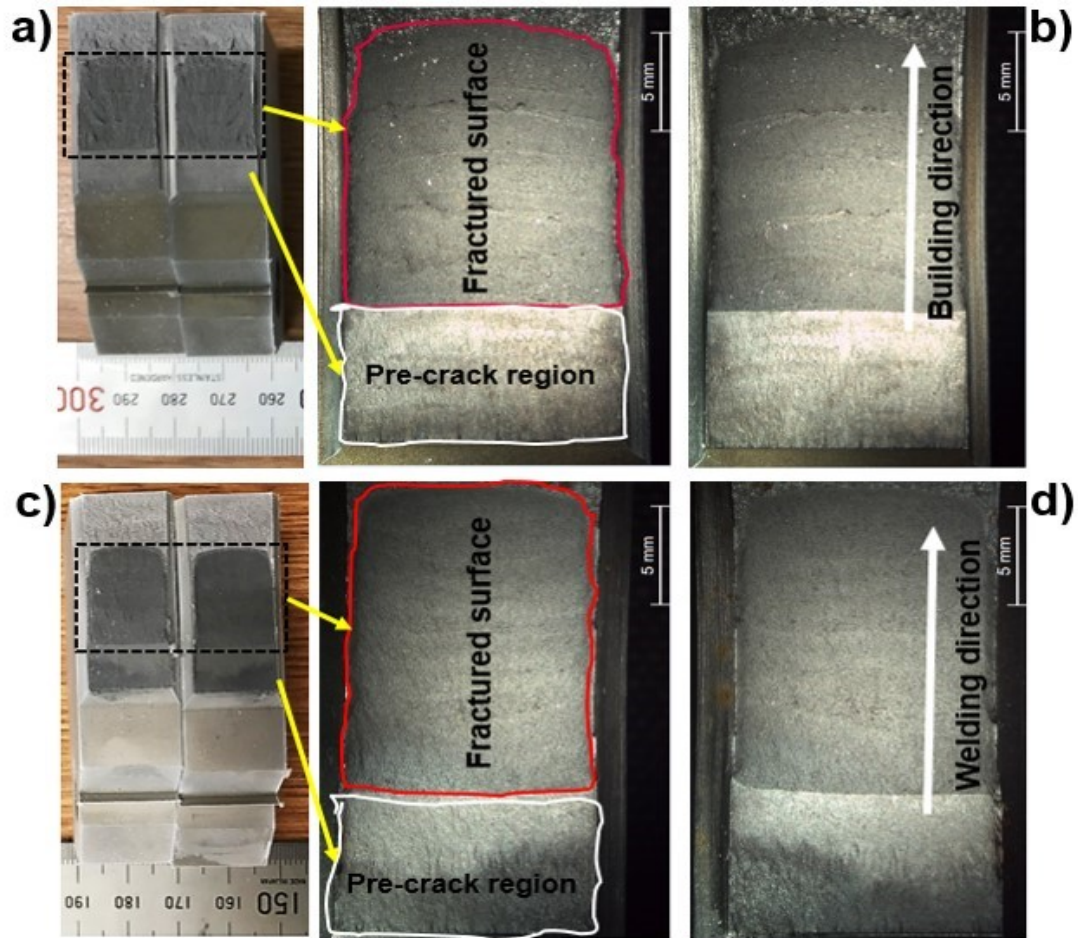


Figure 7-29 Fracture toughness crack path

7.4.4 The influence of crack orientation (Anisotropy) on fracture toughness (J_{1c} and CTOD) of WAAM steels – Z direction

The macro of the C (T) specimen fractured in the Z-direction (across the layers) is presented in Figure 7-29 (a - b), while that of the X-direction (along the layers) is presented in Figure 7-29 (c - d). The surface revealed ductile tearing with some resistance across the layer bands in the Z-direction. The layer bands contain some hard and soft zones, as shown in Figure 7-19. These two zones are the peak and valley in the hardness plot shown in Figure 7-11. The hard zones are

comprised of large grain size of a low proportion of low angle grain boundaries with misorientation angle between 2° and 15° , and the soft zones are comprised of small grain size of a high proportion of high angle grain boundaries with misorientation angle greater than 15° . The distance between two layers (band spacing: B_s) and the height of the coarse grain region (hard zones: H_h) which looks like the HAZ region in intercritically reheated weld could be the key to fracture toughness resistance of the WAAM OS steel structures. The toughness of these steels is explained with the formation/distribution of grain boundaries. HAGBs support low-temperature impact toughness. Cracks suffer a twist effect in HAGBs colonies; thus, the least crack propagation rate and the more energy absorbed by these colonies in fracture.

The soft zone contains ferrite, which promotes maximum resistance to crack growth propagation and crack arrest. At the hard zone, just at the re-melting interphase, there is a directional grain growth parallel to the build direction (Z-Dir), being the region, most affected by the temperature gradient of the solidifying liquid. The dendritic lath emanating from this liquid, solid interface is also influenced by the composition of the solute [154][155]. Although earlier result on WAAM showed this same trend [107], it is supposed that the soft region will lead to slower crack propagation through the build direction. Crack path roughness is also associated with the fracture in the Z-direction (across the layers) being the tortuous fracture path, as shown by Figure 7-29(a - b). The J_q recorded for the Z-direction were higher than the values recorded for the X-direction. The phenomenon discussed above was responsible for this difference in fracture resistance.

7.4.5 The influence of crack orientation (Anisotropy) on fracture toughness (J_{1c} and CTOD) of WAAM steels – X-direction

SEM details of WAAM microstructural features of 90S-B3 steel obtained in the X1 and X2 paths are shown in Figure 7-22. The crack path X2 is the softened region and crack path X1 is the harden microstructural region, as shown in Figure 7-19. The fracture C (T) specimen displayed in Figure 7-29(c - d) revealed ductile tearing with less resistance across the path due to the absence of layer bands.

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The crack path was constrained in these regions by side grooving, as shown in Figure 7-4(b). The path with the least resistance, as shown in the J-R plot, is the X1 path. This path contains an average grain size of $34\mu\text{m}$ with large bainite, and martensitic lath packet sizes greater than $60\mu\text{m}$. Some retained austenite and Widmanstätten ferrite, which is less resistant to dislocation movement is present, as shown in Figure 7-22(a). The X2 path possesses better fracture resistance as a result of the crack following the annealed and soft zone between two-layer bands. This region is refined with average grain sizes of $24\mu\text{m}$ with tempered bainitic and quasi polygonal ferrite, as shown in Figure 7-22(b), hence the improved toughness. In both cases, the subgrain sizes $<1\mu\text{m}$ are present. The preferential fracture path in the X-direction would have been the un-annealed region (X1 path) as claimed in the work of [141] if there were no side grooving.

7.4.6 The influence of microstructure, inclusions and precipitates on the fracture behaviour of WAAM oscillated HSLA steels

The fractured morphologies of the J_{IC} specimens analysed with EDS showing the inclusions and particles found in the fractured zone are shown in Figure 7-31 (a - f). As a result of heating and relatively slow cooling of the WAAM structure, there is a fine distribution of carbides, oxides and intermetallic compounds in the ferrite matrix. The presence of mainly fine non-equiaxed ferrite of an interwoven (basket weave) nature, called acicular ferrite, was found in the ER70S-6 WAAM deposit, as shown in Figure 7-12. The interlocking nature of acicular ferrite plus its fine grains promotes maximum fracture resistance, hence the maximum J_q value of $640\text{kJ}/\text{m}^2$. The Ti and Mn oxide inclusion in this steel deposit, as revealed by EDS scanning in Figure 7-31(a - b), gives rise to ferrite nucleation, and a significant amount of them form AF structures; this is consistent with the work of [146]. The work of [133] also revealed that Ti non-metallic inclusions in welds support acicular ferrite formation due to the replacement of MnSiO_3 by Mn_2TiO_4 and MnTiO_3 and finally Ti_2O_3 with some effectively depleted manganese (Mn) zones. In Figure 7-31(a - b), the high triaxial state stress was reflected in the high plasticity possessed by ER70S-6 WAAM deposit, and this was shown in the

copious equiaxed shaped, small deep and closer dimples in the fractured surface, hence, the decrease in dimple size as energy per unit area required to fracture increases. The Z path specimen had more and smaller dimples. The fracture mechanism is caused by ductile failure, and the smaller sized dimples depict its high ductility, hence the tougher it was compared to other WAAM deposited steels.

The microstructure presented by the ER90S-B3 is coarse, as shown in Figure 7-13. It contains martensite and austenite (M-A) constituents consisting of martensitic laths, upper bainite, and grain boundary ferrites. These hard phases are promoted by the presence of a high volume of Cr and Mo, as shown in Table 3-1. The formation of grain boundary ferrite and upper bainite provides an easy pathway for crack propagation, hence, the lowest J_q value of $350\text{kJ}/\text{m}^2$ compared to WAAM ER70S-6 & ER120S-G. Fine spherical carbide precipitates and inclusions of AlO, MnS, and Al-Si were observed in the fractured surface, as shown in Figure 7-31(c - d). It was also reported by [147] that the non-uniform micro straining of the matrix and dislocation locking caused by these precipitates are responsible for the fracture behaviour of these steels.

According to [148] $M_{23}C_6$ and M_7C_6 are the secondary precipitates found in the weld subzone that weakens the fracture path and leads to micro void generation. The large sized non-equiaxed intragranular dimples were as a result of the weak influence of the microstructure on the fracture behaviour. The dimple sizes in the Z and X paths are very much alike. The fracture mechanism is through micro-void coalescence with some quasi cleavage and dimple fracture. The fracture surface of WAAM ER120S-G revealed the presence of Ti, Al and Mn inclusions, as shown in Figure 7-31 (e-f). The reduced inclusion number as a result of aluminium retarding ferrite transformation leads to the presence of a low amount of acicular ferrite [149–151]. Precipitation and homogeneously dispersed carbides and oxides in the ferrite matrix were responsible for the improvement in strength and toughness of the WAAM ER120S-G steel. The fine inclusion particles act as nucleation sites for acicular ferrite formation [152]. This also possesses huge effects on grain boundary pinning, hence the reduced average grain size $<12\mu\text{m}$ and subgrains of $<1\mu\text{m}$. The micro constituents form interlocking structures in the microstructure that improve strength and toughness

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[95,97,152]. The combined presence of Ni (2.27%) and Mo (0.63%) in a certain proportion in the original chemistry of the wire aided the reduction of both second phase and grain boundary ferrite, hence the high J_q value of 400kJ/m^2 .

This is consistent with [136]. The presence of numerous large equiaxed dimples reflects the ductility attributes of this steel. WAAM of steel is a constructive process that could enhance the formation of different transformation products depending on the process deposition parameters and thermal mass control. The kinetics of ferrite nucleation and growth reaction at the austenite grain boundary and intragranular nucleation sites determine the extent of acicular ferrite formation. The fracture toughness of low alloy steel is dependent on the volume fraction of AF. Austenite grain size, alloy content and inclusion characteristics with WAAM deposition parameters are the factors controlling the large volume fraction of AF in WAAM steel deposits.

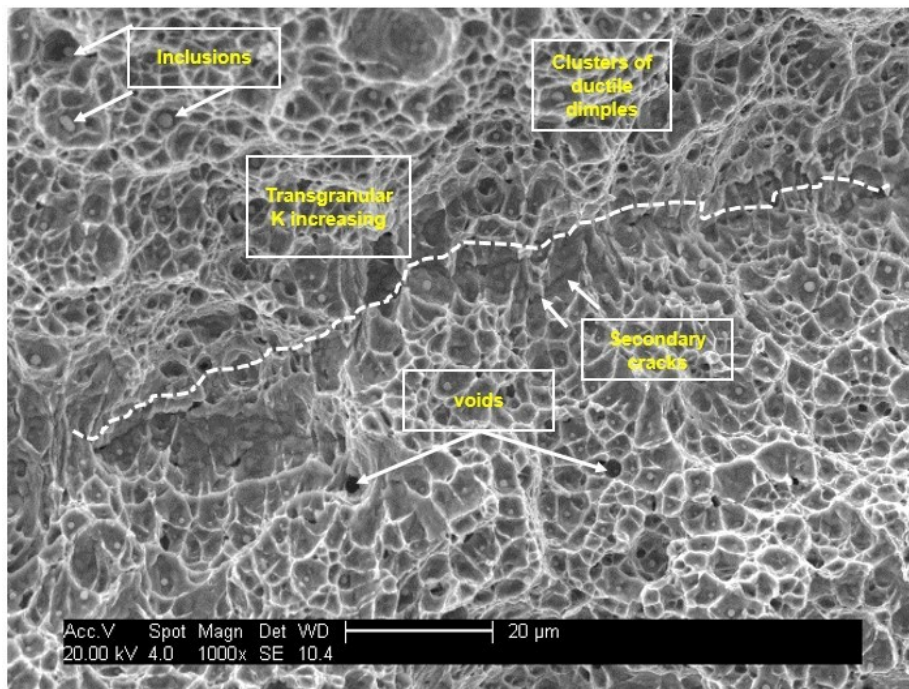
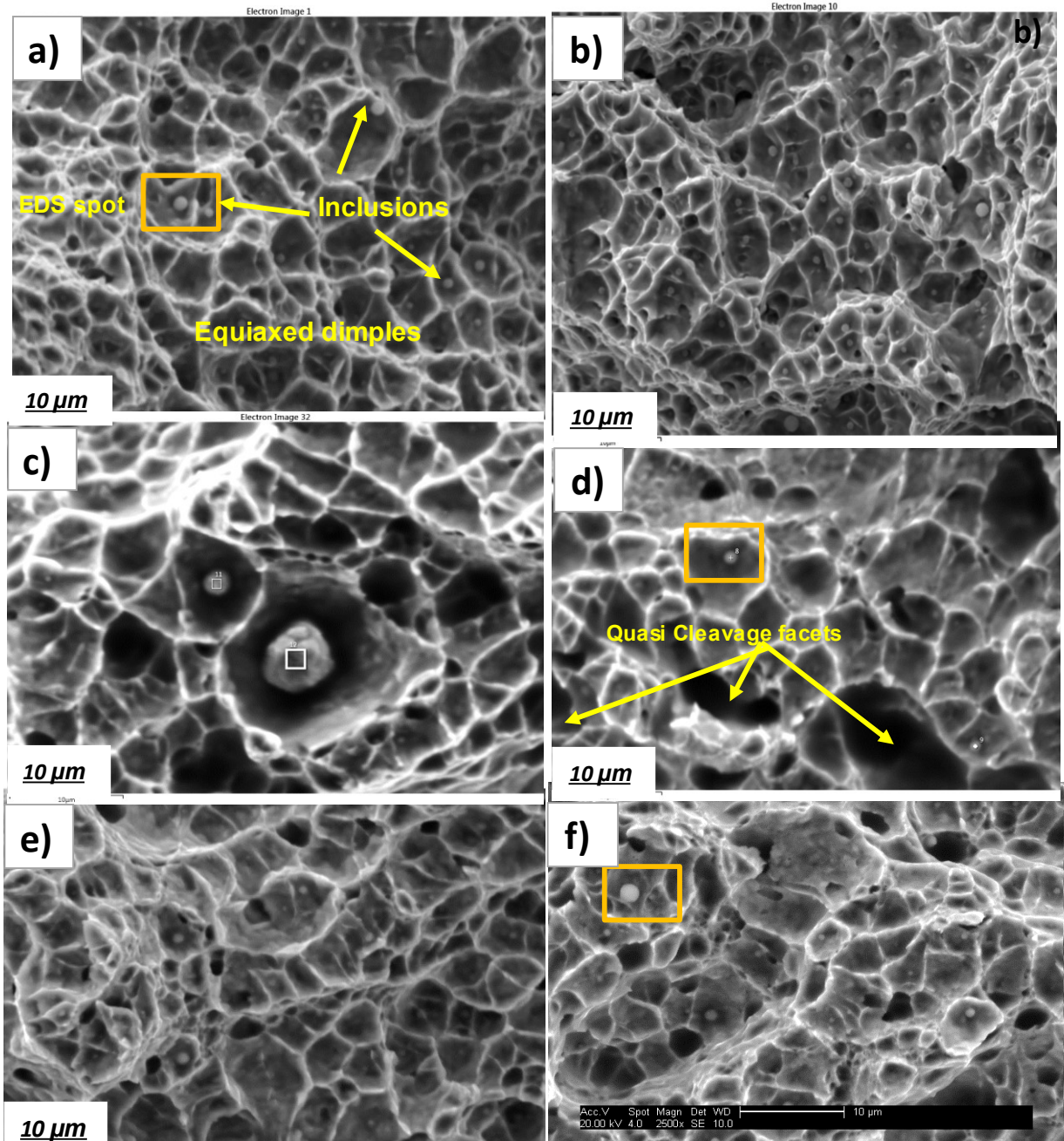


Figure 7-30 Crack propagation path in fracture toughness testing of ER70S-6

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Fracture and Fatigue Behaviour of WAAM Components

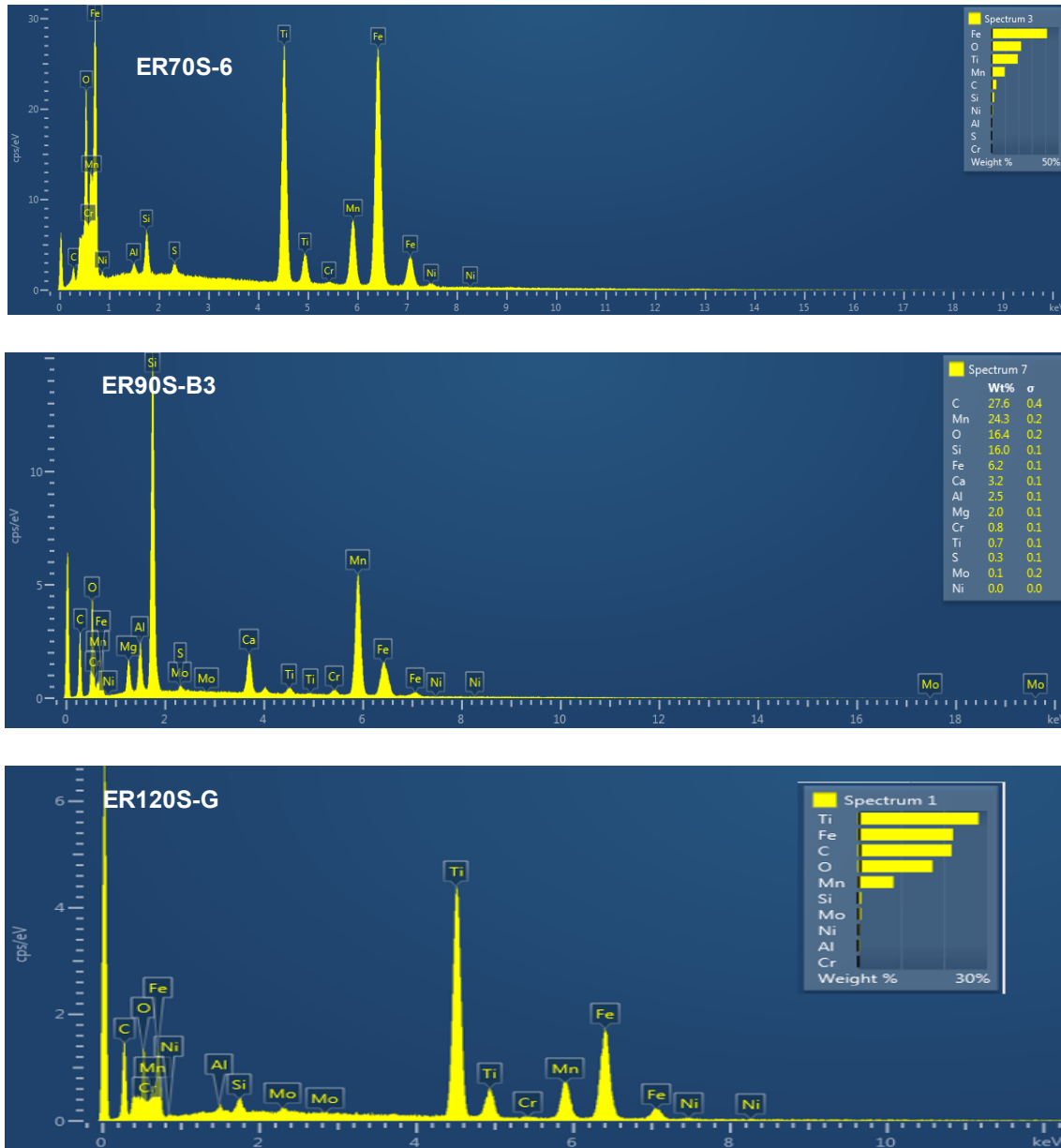


Figure 7-31 Fracture morphologies of WAAM steel

(a) ER 70S-6 (Z) (b) ER 70S-6 (X) (c) ER90S-B3 (Z) (d) ER90S-B3 (X) (e) ER120S-G (Z) (f) ER 120S-G (X)

7.4.7 Fracture toughness behaviour of wrought compared to WAAM steel (microstructural sensitivity)

The S355G10 steels with 0.06%C, 1.257% Mn, 0.33% Ni, 0.27% Si, 0.24% Cu, 0.03% Cr and 0.01% Mo with UTS: 545MPa, YS; 435MPa, E; 36% contain mainly

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ferrite-pearlite microstructure ($\alpha - P$) with a low volume fraction of pearlite in a ferrite matrix, as presented in Figure 7-32. The average grain size of S355G10, as measured in section 4.3.4, is $15.29\mu\text{m}$. This is similar to the coarse grain-sized steel (SM490) used in Chapetti et al. [79]. Despite having close composition to ER70S-6 HSLA steel, its %Ni composition is higher, and this aids the reduction of both the second phase and grain boundary formation which is responsible for the weak fracture. However the ER70S-6 oscillatory strategy, as shown in Figure 7-22(e & f), is composed of martensitic structures and retains austenite, hence the lower fracture toughness value compared to S355G10 as depicted in Table 7-2.

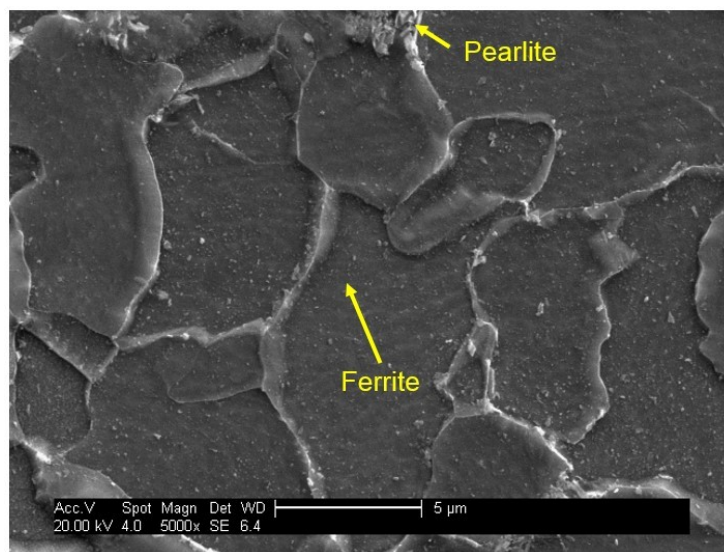


Figure 7-32 S355

7.4.8 Fatigue crack growth rate of wrought compared to WAAM parts (crack path sensitivity)

Table 7-4, Figure 7-6 and Figure 7-34 present the Paris law constants and FCGR plots of WAAM versus wrought. The result of this present study, enumerated in Table 7-4, is comparable to TMCP S355G10, EH36 and S355 normalised. The major difference in the WAAM HSLA steel is the sensitivity factor m , which indicates the severity of the growth rate to applied stress. This factor m is 3.26 & 3.31 in both the X- and Z-directions of ER70S-6. This is indicative of the ductility of the material. A comparison of the FCGR at SIFR of $15\text{--}45\text{MPa}\sqrt{\text{m}}$ showed that the WAAM steels performed better than the BS 7910:2013 recommended design mean value limit for FCGR for steel in air tests ($1.65\text{E-}11\Delta K^{3.0}$). It was also

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evident that at SIFR between 15-25MPa $\sqrt{m}$, the S355G10 performed better than the WAAM steels, but between SIFR of 25-45MPa $\sqrt{m}$, the WAAM steel performed better, as shown in Figure 7-34. This could be due to their phase's distribution, texture, volume fraction, composition and general morphology (microstructure).

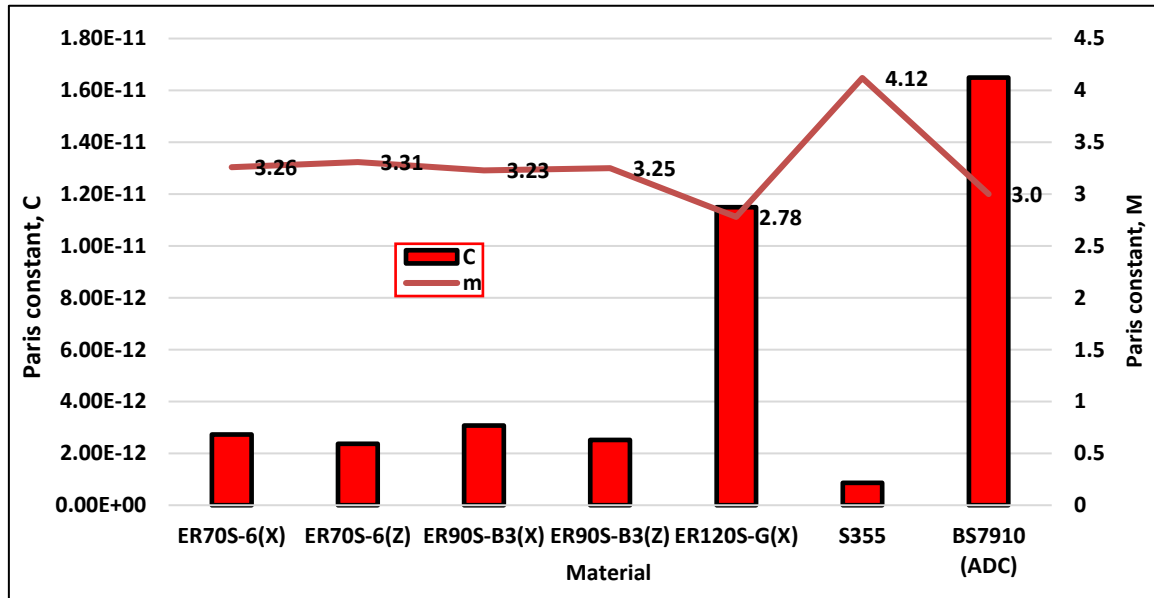


Figure 7-33 Relationship of WAAM steel Paris constant (C, M)

Table 7-4 Paris law constant C^* & M for WAAM steels and wrought

Material	Da/dN (m/cycle)	C^*	m	Ref
ER70S-6(WAAM)-X	$2.73E-12\Delta K^{3.26}$	$2.73E-12$	3.26	This study Present
ER70S-6(WAAM)-Z	$2.37E-12\Delta K^{3.31}$	$2.37E-12$	3.31	
ER90S-B3(WAAM)-X	$3.07E-12\Delta K^{3.23}$	$3.07E-12$	3.23	
ER90S-B3(WAAM)-Z	$2.52E-12\Delta K^{3.25}$	$2.52E-12$	3.25	
ER120S-G(WAAM)-X	$1.15E-11\Delta K^{2.78}$	$1.15E-11$	2.78	
S355G10+M(wrought)	$1.66E-13\Delta K^{4.12}$	$1.66E-13$	4.12	V. Igwemezie et al.
BS7910 air design curve	$1.65E-11\Delta K^{3.0}$	$1.65E-11$	3.00	V. Igwemezie et al.
SLIC (TMCP)	$2.83E-12\Delta K^{3.23}$	$2.83E-12$	3.23	A. Mehmanparast et al.
Tsay (EH36)	$3.07E-13\Delta K^{3.89}$	$3.07E-13$	3.89	Tsay et al.
Thorpe (Normalized)	$3.26E-12\Delta K^{3.27}$	$3.26E-12$	3.27	Thorpe
304L stainless steel(WAAM)-Z	$7.29E-11\Delta K^{4.12}$	$7.29E-11$	4.12	J.V Gordon et al.
304L stainless steel(WAAM)-X	$3.01E-10\Delta K^{3.79}$	$3.01E-10$	3.79	J.V Gordon et al.

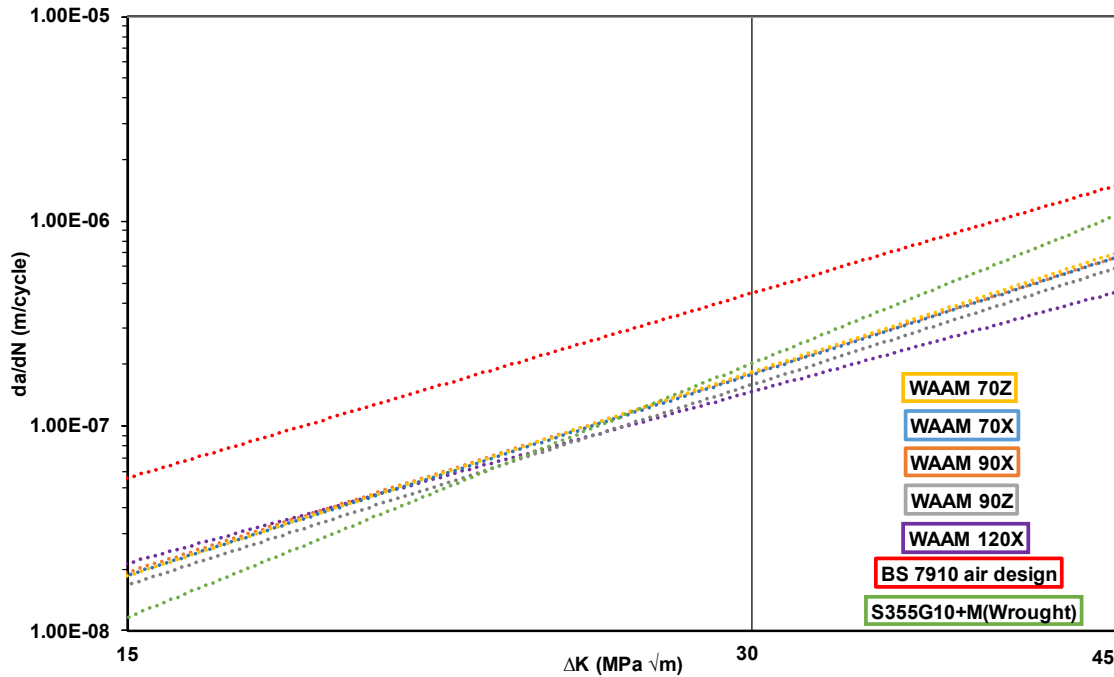


Figure 7-34 FCGR of WAAM steel and wrought

Figure 7-37 presents the FCGR vs. UFG TMCP and WAAM ER70S-6. To investigate the grain effect further, the FCGRs of two ultrafine-grained ferrite-pearlite steels (UFGS) produced by the TMCP method and WAAM process have been included in the analyses. The composition and loading condition of the UFGS is indicated in [82].

The difference in the FCGR for the WAAM steel and the UFGS can be attributed to two factors; phase morphology and residual stress believed to be inherent in WAAM parts as a result of the manufacturing process. Also, Chapetti found that stress ratio affected the UFGS in the range of 0.1 and 0.7; hence, the two-stress ratio was plotted. FCGRs at both ratios are still lower than those of WAAM steel (ER70S-6) and NR steels, although FCGRs in NR and WAAM steels appear to fall within a common band.

Even though WAAM products are believed to have a high level of residual stress, surprisingly they performed better than some of the wrought steels that may be judged as having less residual stress. This calls into question if residual stress is detrimental to FCGR in WAAM steels to the extent it is assumed presently, although the UFGS has a smaller grain size due to the thermo-mechanical

treatment carried out on the ferrite grains ($1\mu\text{m}$) as opposed to the WAAM steel grain size of $22\mu\text{m}$.

Then if we discount the grain size effect, the difference in FCGRs as observed in Figure 7-37 could be attributed to the phase morphology. At this point, one will be tempted to conclude that microstructure has a profound effect on FCGR in air, which is in support of the claim of [132]. This is conflicting with the conclusion that microstructure has little or no effect at ambient temperature.

7.4.9 Fatigue crack growth path in WAAM steel compared to ferrite-pearlite steels

Figure 7-35 and Figure 7-36 present the non-planar crack path in ferrite-pearlite steels and WAAM HSLA steels. What is obvious is that the crack path is non-planar for both mild steels and WAAM deposited steels, and the mode of fatigue crack propagation is not the same for all the mild steels. According to V. Igwemezie et al.[132], three phenomena were observed viz, crack deviation shown by the white arrow, crack branching as indicated by red triangle and formation of metal crumbs in-between branched and re-joined crack fronts as shown by white triangle, all of which are shown in Figure 7-35. The normalised grade of S355 (Figure 7-35(a)) and WAAM path (Figure 7-36) appear to have somewhat similar modes of crack propagation in the sense that the crack deviation angles are generally low. This may explain why the FCGRs for Normalised and rolled (NR), and WAAM are similar, as shown in Figure 7-37.

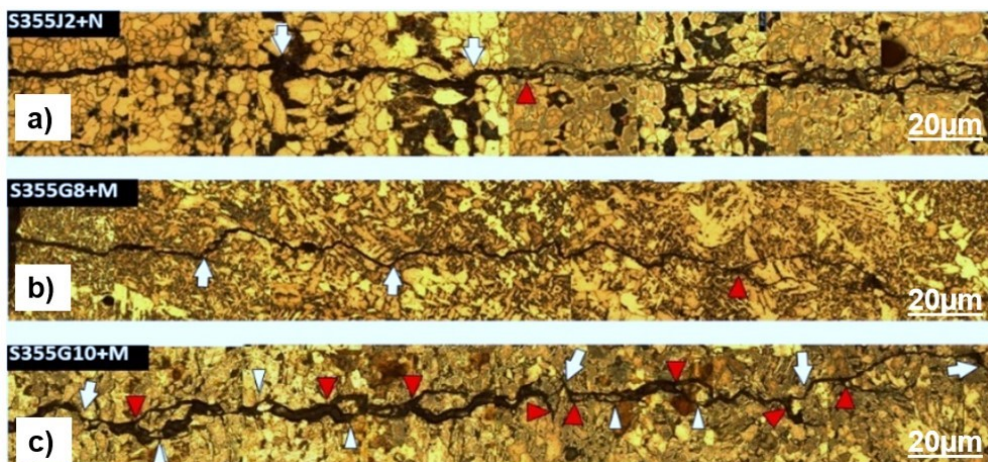


Figure 7-35 Non-planar crack path in ferrite-pearlite steels

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(a) S355J2 + N normalized rolled steel (b) S355G8 + M TMCP (c) S355G10 + TMCP [132]

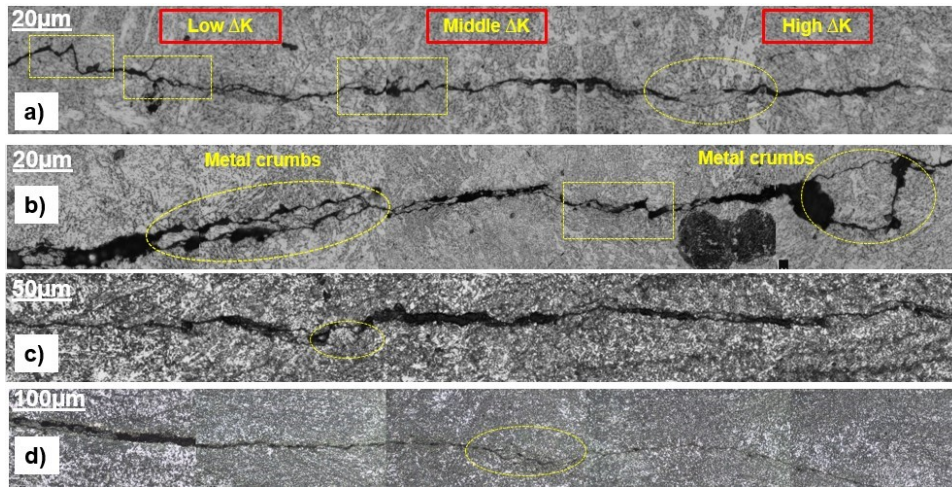


Figure 7-36 Non-planar crack path in WAAM steels

(a) ER90S-B3 (Z) (b) ER90S-B3 (X) (c) ER70S-6 (Z) (d) ER70S-6 (X)

The similarity of NR and WAAM appears to show that residual stress may not have such a strong effect on crack growth rate as currently believed. The branched crack fronts generally appear 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, and this happens when the other crack fronts encountered a more resisting phase. An important observation is that there were several branched crack fronts that re-joined to form one crack front, creating metal crumbs in-between the re-joined crack fronts. The tendency for crack branching and re-joining tends to increase with an increase in the driving force for crack propagation. This phenomenon can be seen clearly on the micrographs.

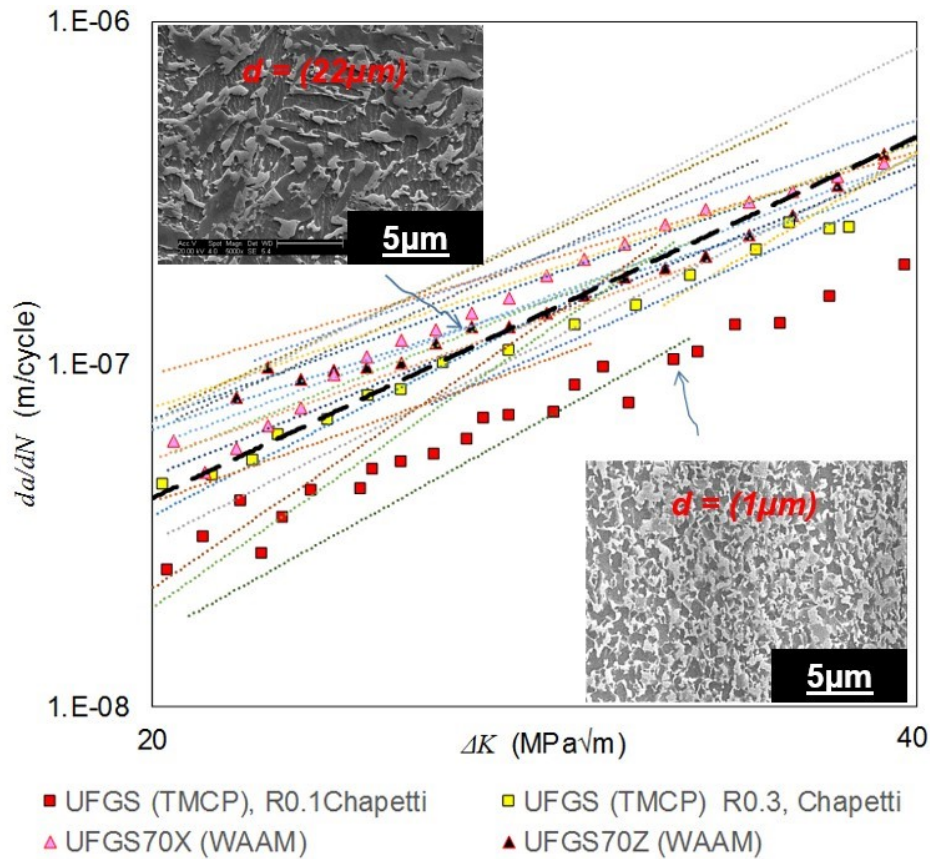


Figure 7-37 FCGR performance of ultrafine grain TMCP and WAAM ER70S-6

The low angle of deviation and parallel fibrous crack fronts in J2N and WAAM appear to contribute little to the FCGR retardation as their FCGRs still lie above the arbitrary line in Figure 7-37. The three phenomena mentioned occurred many more times in the G10, followed by G8. The angle of crack deviation is higher for the TMCP steels. In general, TMCP microstructures showed a high angle of crack deviation, branching and with the complex multiplanar crack front. The crack deviation, branching and creation of metal crumbs can reduce the effective stress at the local crack tips resulting in retardation of FCGR. The degree of crack tip stress shielding will vary with the angle of crack deviation and the nature of metal crumbs created.

A crumb will be effective in crack growth retardation if it re-orientes and causes a mismatch or wedging action. The tendency to re-orient will increase if the crumbs are smaller in a rough-round shape (as seen in the G10) rather than in elongated-plate form as seen clearly in WAAM micrograph (Figure 7-36(b)), although it is in

mixed mode. These two natures of metal crumbs will be a function of the steel microstructure.

The high angle deviation and nature of metal crumbs may explain why FCGR is lowest in the TMCP steels, as shown in Figure 7-37. As aforementioned, two more phenomena which are associated with crack retardation are crack closure and interlocking; however, the concept of crack closure cannot be rigorously applied to steel in all conditions, let alone when a fatigue test is done under plain strain condition – i.e. with little or no plastic zone at the crack tip. When the plastic zone is small, the effect of crack closure can then be ignored, or its consequences will be insignificant. The tendency for interlocking should be a function of the angle of crack deviation, which is likely to increase with an increase in the angle of deviation.

Figure 7-38 presents the SEM images of the WAAM steel crack path. Note how fine-grained the WAAM steel is, compared to the TMCP steel. It is worth noting that the SEM micrographs and crack propagation modes of G8 and G10 are generally the same. This implies that the crack propagated in both transcrystalline and intercrystalline modes in air. The mode of crack propagation appears to depend on the local orientation of the phases immediately ahead of the crack tip. The microstructure appears to contain very little or no pearlite. We can observe mainly low angle crack deviation and branching in Figure 7-38. The metal crumbs formed in between the branched and re-joined crack front are plate-like and may not be able to re-orientate to reduce the crack-tip driving force.

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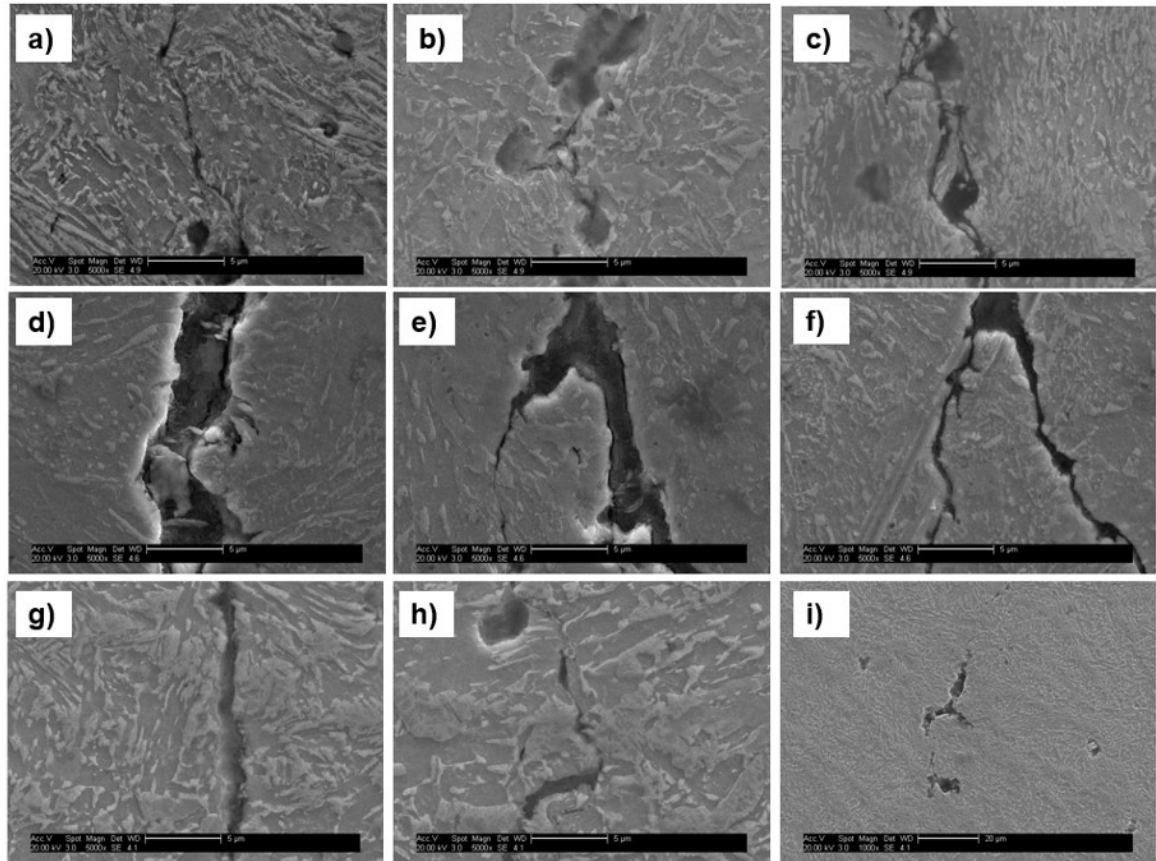


Figure 7-38 The SEM images of WAAM steels crack path

(a-c) ER70S-6 (d-f) ER90S-B3 (g-i) ER120S-G

This mode of crack propagation is similar to that of J2N and may explain why their FCGRs are similar. Another important observation is that there appears to be flaking off of metal pieces in the WAAM steel, thereby creating what appears as large spaces. This may be as a result of tensile residual stress inherent in the WAAM deposited structure. A preliminary assumption we can make at this level is that the angle of crack deviation, branching and nature of the metal crumb formed strongly and fundamentally affects the FCGR in WAAM steel in air. Also, I am tempted to state that the crack path in this low carbon low alloy and high strength steels microstructure deposited with WAAM is complex (not linear), and the microstructure of these deposits strongly influenced the FCGR throughout the da/dN vs. SIFR (including the Paris region) which is also in support of V. Igwemezie et al.[132] claims.

7.4.10 The fatigue life of WAAM HSLA steel

Figure 7-39 presents the results of the fatigue strength as a function of the number of cycles to fracture. The smooth fatigue limit (σ_{eR}) was defined as the maximum nominal stress amplitude under which the specimen endured more than 10^7 cycles. In these studies, the WAAM HSLA steel gave σ_{eR} of 160MPa for ER70S-6, 257MPa for ER90S-B3, and 307MPa for ER120S-G. The σ_{eR} limit is in consonant with their grain sizes, as presented in Figure 4-24.

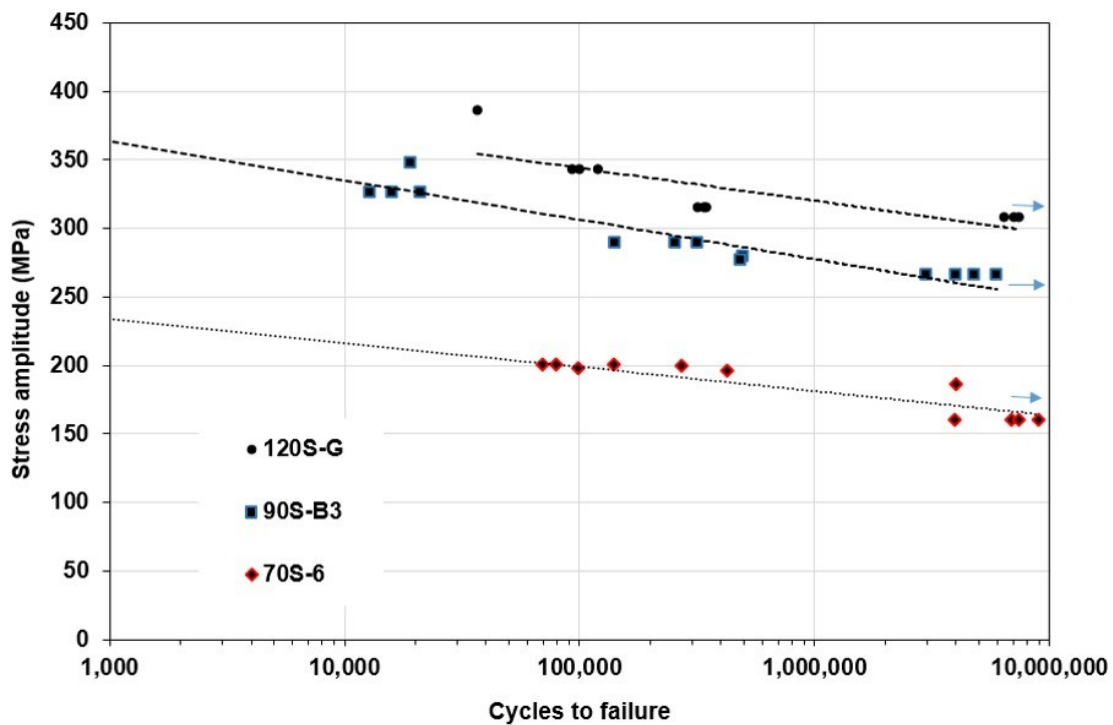


Figure 7-39 Fatigue life of WAAM steels in X-direction

7.4.11 Conclusion

In this chapter, the effect of WAAM oscillatory deposition strategy on the crack microstructural interaction in a structure has been investigated; it was shown that:

1. The preferential grain growth direction during solidification in oscillated WAAM HSLA steel is the $\{100\}$ & $\{101\}$ orientation. This is contrary to the single pass deposition strategy where no preferential orientation was observed in ER90S-B3 & ER120S-G. A strong texture intensity in the melt-

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zone is observed in the ER70S-6 & ER90S-B3, which indicated that an anisotropic martensitic lath microstructure both in random crystallographic texture was obtained.

2. The possible cause of variability in fracture toughness in the WAAM steel deposits is the hardening and softening regions in the layer bands caused by the variation in peak temperature and cooling rates of the adjacent layer in the building direction. The fracture toughness across the Z-direction possesses the most resistance to fracture failure as a result of the reduced grain sizes, due to grain refinement and increases in the density of grain boundaries in the Z-direction. The fracture toughness along the welding direction (intralayer and interlayer) possesses the least resistance to fracture failure as a result of increasing grain size in the welding direction and tendency of the crack propagating in the most vulnerable part of the microstructure, i.e. the hardened zone in the layer band. This trend is different with the fatigue crack growth rate where the mild steel WAAM shows the most consistent similarity in growth rate in both X and Z direction, while the HSLA steel WAAM shows higher fatigue crack growth rate in the X direction compared to the Z-direction.
3. The layer band spacing and the thickness of the re-melt zone are key parameters that control the fracture toughness of WAAM steel in both directions. The dominating mechanism of failure in the WAAM steel is ductile fracture bound with large and small dimples with some fine dispersed carbides and oxides in the ferrite matrix.
4. The crack path in ferrite-pearlite steel and WAAM HSLA steel is complex (not linear). The angle of crack deviation, branching and nature of the metal crumb formed strongly, and fundamentally affect the FCGR in ferrite-pearlite and WAAM HSLA steel in air. The microstructure of ferrite-pearlite and WAAM steels strongly influenced the FCGR throughout the da/dN vs. SIFR (including the Paris region).

The investigation in the present chapter was demonstrated that bespoke components of low carbon low alloy and high strength steel with higher thickness

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could be made from WAAM using oscillatory deposition strategy and that better fracture toughness, fatigue crack growth rate and mechanical properties could be obtained if the heat accumulation phenomenon is taken care of by developing a novel approach to independently controlling the heat built up.

Chapter 8 will be developing a novel wire assisted WAAM process to control accumulated and build up heat in oscillatory and other strategies to deposited structures with original grain refinement and wrought specifications. This novel approach will reduce heat build-up effects, microstructural variability, consequently residual stress and elimination of base substrate dilution.

8 ADDING COLD WIRE TO IMPROVE WAAM DEPOSITION

8.1 Introduction

There are three significant challenges to the increase in productivity of WAAM using the tandem gas metal arc power source (Tandem GMAW), namely: the uncertainty in the mechanical properties; the structural integrity of the final product due the slow cooling of the deposited structure as a result of the heat accumulation, which caused variability in the deposited microstructure. The high cost of special jigs and fixtures required with this process to cater for the enormous residual stress in-built as a result of heat accumulation. The cost of the additional power source and power usage is another limiting factor to the use of tandem GMAW.

To the author's best knowledge, there are currently no suitable studies that has proposed a solution to this microstructural variation emanating from the thermal heat effects when depositing at a higher rate. This study will harness the low-cost benefits of the PGMAW process for WAAM application. The challenges posed by the deposition of structures with intersections using the WAAM process, as shown in Figure 8-1 would be solved with CW addition. The addition of CW to a PGMAW power source would ease off some of the disadvantages of using a PGMAW power source for WAAM deposition. Preliminary investigations shown in Table 8-1 with the depositions of beads on plate, as shown in Figure 8-2, were used to illustrate these merits of CW addition into the PGMAW WAAM process. The objective of this study is enumerated below.

1. The first objective of this research is to determine the suitability of feeding cold wire into WAAM melt pool and the optimised orientation to use.
2. The second objective of this research is to determine the process envelope suitable for cold wire feeding into the WAAM melt pool and how the bead characteristics affect WAAM deposited structure.
3. The third and fourth objectives are to determine the effects of feeding the cold wire into the WAAM melt pool at a varying deposition and constant deposition rate.

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4. The fifth objective is to determine the mechanical and microstructural properties of optimised deposition rate and compared with WAAM properties deposit in section 4 and 7.

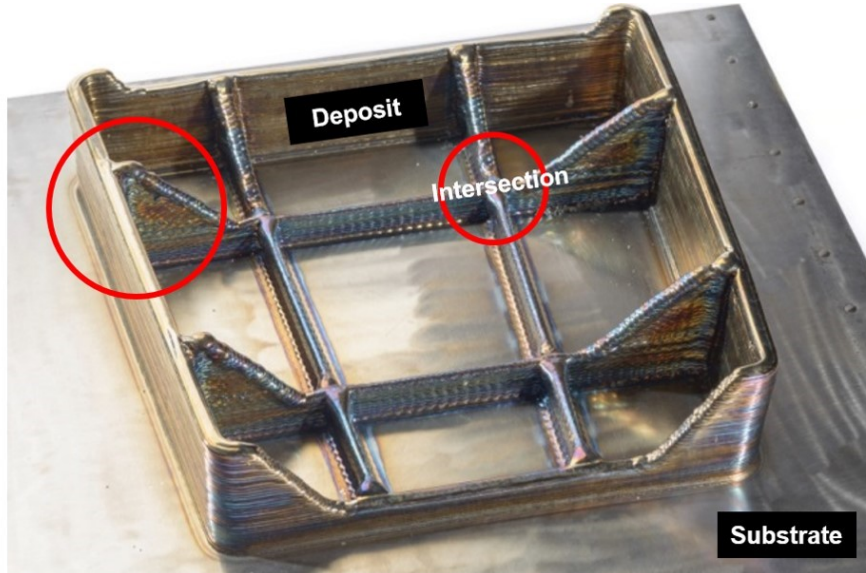


Figure 8-1 WAAM deposit illustrating the challenges

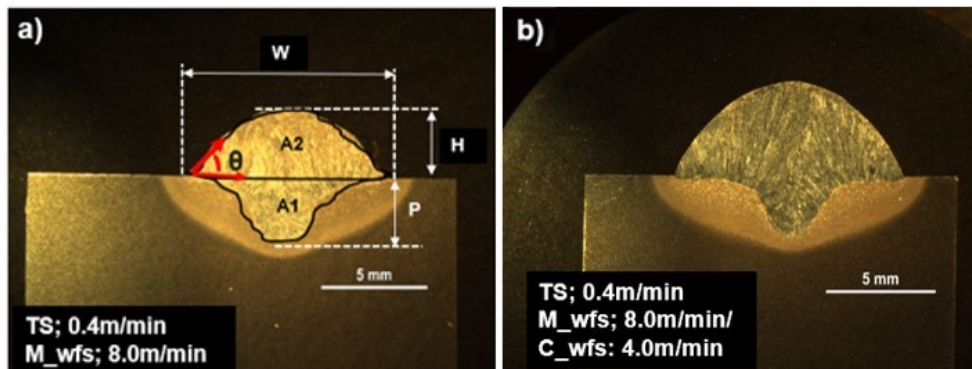


Figure 8-2 Comparison of WAAM and WAAAM deposition with PGMA

- (a) (WFS, 8.0m/min, TS: 0.4m/min) (PGMAW only) (b) PGMAW + CW (TWFS: 15.0m/min, TS: 0.4m/min) front feeding

Table 8-1 Comparison of CW addition potentials

Bead parameters	PGMAW	FF	SF	Diff	Remarks
Penetration-P(mm)	3.16	1.37	1.53	57%	+
Bead width-BW(mm)	9.76	11.02	11.88	22%	+
Bead height-BH(mm)	3.42	4.18	4.09	19.5%	+
Aspect ratio-AR	0.35	0.38	0.34	5.7	+
Re-melt ratio-RR (%)	34.5	14.19	10.69	69%	+
Contact angle(°)-CA	46.62°	67.3°	56.4	21%	-
Deposition rate-DR(Kg/hr)	4.18	6.27	6.27	50%	+
Heat affected zone area HAZ (mm ²)	19.86	16.21	12.6	37%	+

8.2 Experimental methodology

8.2.1 Set-up procedure

In this experiment, S275RJ plates of dimensions 450mm x 150mm and thickness of 12mm were used as the base material, ER90S-B3-Böhler wire with a diameter of 1.2mm was used for both the PGMA and CW feed. The typical chemical composition of the wire is shown in Table 3-1. A K2426-5 LF-74 Lincoln wire feeder was used to feed into the CW. The PGMA torch was connected to an ABB six-axis robot. Figure 3-3 shows the set-up and connection to the robot. Bead on plate deposition was conducted in the flat position. The contact tip to work distance (CTWD) was established with some trials to determine the suitable region to work with. A 98:2 mixture of Ar and O₂ (Argo shield heavy) was used as the shielding gas, which was supplied at a flow rate of 15-25L/min.

Suitable feeding direction was determined by placing the CW at the rear, front and side of the PGMA wire into the melt pool, as shown in Figure 8-3. The torch was set perpendicular to the substrate. Feeding of the PGMA was done at 90° while a clamp holder, which allows the CW to be fed at different angles (45°, 15° & 30°) to the base of the substrate, was utilised. An arc monitoring camera was

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connected to the clamp assembly. Fronius CMT VR 7000 power sources, with a synergic database for PGMA combining the required process mode, wire diameter with type, and shielding gas, were used. A portable arc monitoring device (AMV4000) was also connected across the terminal for arc parameter monitoring. Before the commencement of the experiment, the relationship between the WFS (m/min) of the CW feeder and the reading from the AMV4000 was established. The CW calibration chart is shown in Figure 3-4.

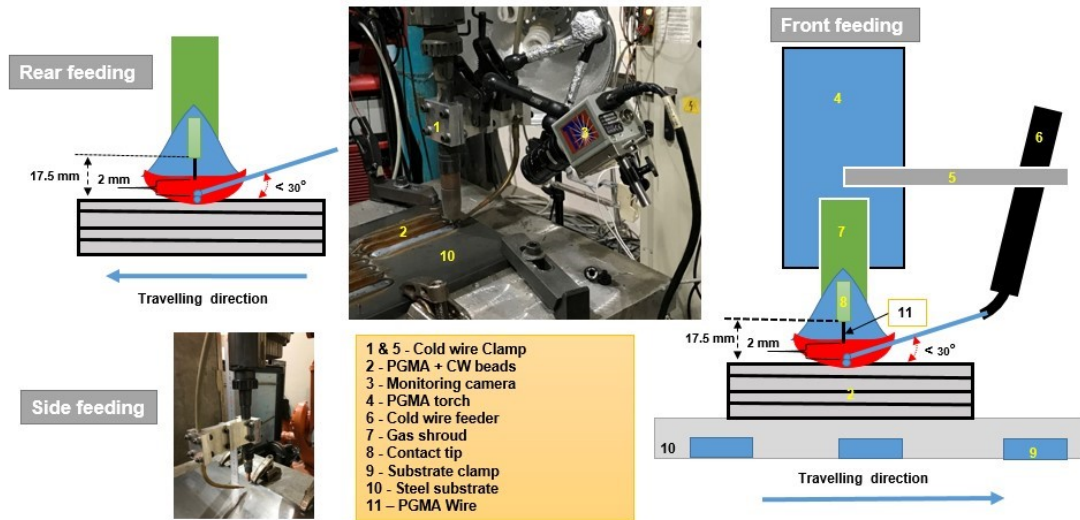


Figure 8-3 Schematic representation of WAAAM

In the first stage of the experiment, a suitable build envelop was determined for PGMA beads on plate alone without the CW. In the second stage of the research, suitable CW feeding direction and angle of feed was established. In the third stage, a determined PGMA WFS of 8.0m/min from the build envelop in stage 1 was used as a constant WFS while CW was being fed at different WFS while simultaneously increasing the travel speed (TS) of the deposition from 0.4, 0.6 and 0.8m/min. In the fourth stage, the effects of feeding CW into the PGMA bead at a constant deposition rate of 6.27Kg/hr and different WFS ratio ($PGMA_{WFS}|CW_{WFS}$) were studied at different TS of 0.4 and 0.6m/min.

In the fifth stage of the experiment, the best combination of PGMA + CW (WAAAM) deposited bead parameters, as shown in Table 8-2 were used to deposit a 6-layers structure for with the front feeding (FF) and 39-layers for the

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side feeding (SF). The control sample was deposited with PGMA alone, with a WFS of 8.0m/min and 12.0m/min on a TS of 0.4m/min respectively. Figure 8-4 and Figure 8-5 present the front and rear feeding schematic and built structure with side feeding. Figure 8-6 presents the WFS trend to show the stability of the process.

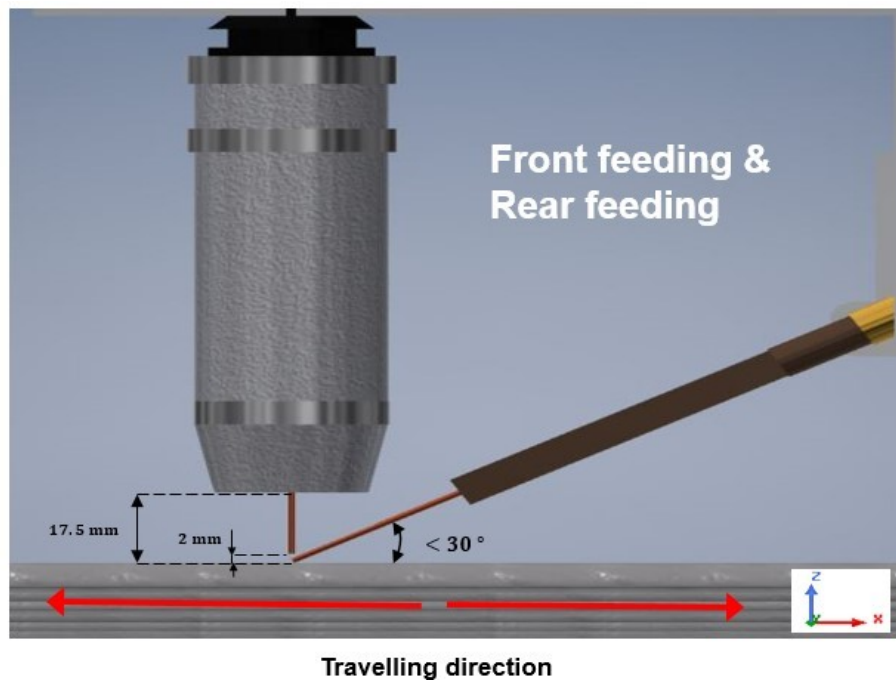


Figure 8-4 Front and rear feeding configuration

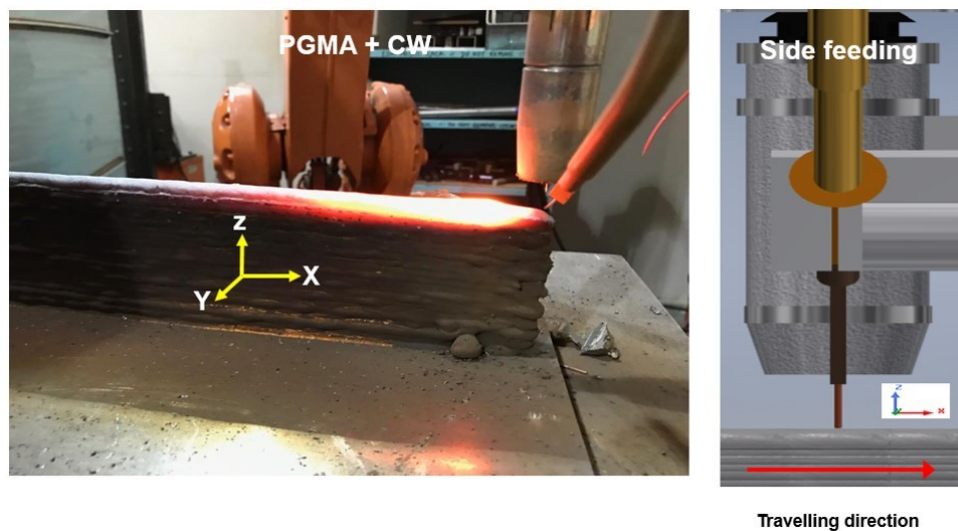


Figure 8-5 build process and side feeding configuration

Table 8-2 Build parameter for WAAAM structure

WAAAM parameters	Front feeding	Side feeding
PGMAW WFS (m/min)	7.65	6.0
CW WFS (m/min)	4.0	4.2
TWFS (m/min)	12.0	10.2
CW-WFS/PGMA-WFS	0.52	0.7
Robot travelling speed (TS) (m/min)	0.4	0.4
Interpass temperature (°C)	<100	<100
Deposition current (I) (Amps)	226	185
Deposition voltage (U) (volts)	25.4	22.5
Heat input (kJ/mm)	0.861	0.624
Gas flow rate (L/min)	25	20
Layer height (mm)	3.5	3.5
CTWD	17.5	17.5
Effective heat (kJ/mm)	0.448	0.437

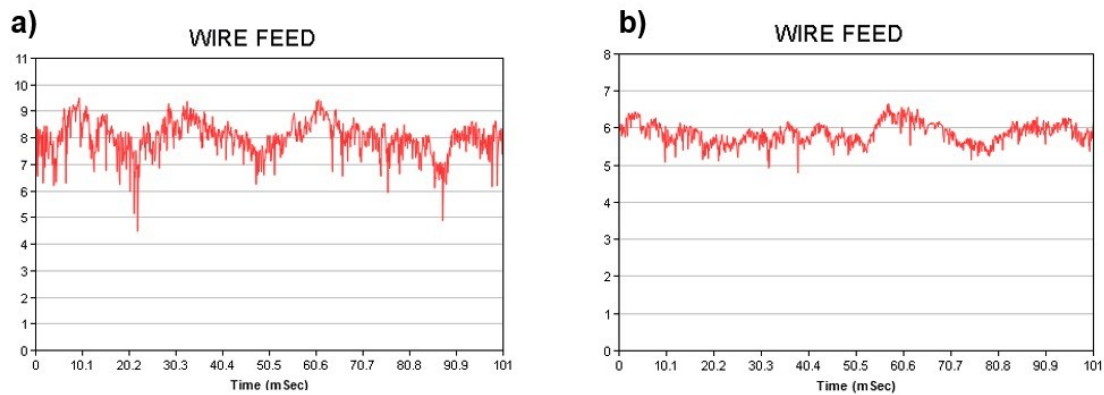


Figure 8-6 Energy trend to show process stability

(a) front feeding (b) side feeding

8.2.2 Beads profile, microstructural and hardness characterisation

The WAAAM beads from stages 1, 3, & 4 and the control sample of the experiment were transversely cross sectioned from the wall for metallographic analysis with the subsequent procedure consisting of grinding, polishing and

etching in 2% Nital. Macro images of the etched WAAAM beads were obtained using the optical microscope (Optiphot Nikon) and deployed to the Axio vision software to reveal the bead profile parameters, bead width (BW), bead height (BH), Penetration (P), Contact angle (CA), reinforcement area (A_2), penetration area (A_1), Re-melting ratio (RR), and aspect ratio (AR). The structure build in the fifth stage was cross-sectioned to obtain the microstructural variation along the deposit layers using a Scanning Electron Microscope (SEM FEI XL30-SFEG). The same system was used to characterise the tensile fractured surfaces and the Charpy impact toughness fractured surfaces. The micro-hardness is measured by the Zwick/Roell hardness tester along the deposit from the base to the top direction under a load of 0.5kg with 15secs holding time. The EBSD samples were gradually ground and polished for 30mins before quantitative analysis.

The deposition rate of the WAAAM bead was characterised with the relationship shown in Eq (8-1), where $D_{rate_{Kg/hr}}$ is the deposition rate, ρ is the density of wire, E is the deposition efficiency, PW is the PGMA, CW is the cold wire, WFS is the wire feed speed (m/min). The ratio of PGMA WFS to CW WFS is computed using Eq (8-2) while the ratio of total WFS (PGMA WFS & CW WFS) to robot travel speed (TS) is presented in Eq (8-3).

$$D_{rate_{Kg/hr}} = \rho \cdot E \left[\pi \cdot \left(\frac{PGMA_wire\ diam}{2} \right)^2 \cdot PGMAW_wire_{wfs} \right] + \rho \cdot E \left[\pi \cdot \left(\frac{CW_wire\ diam}{2} \right)^2 \cdot CW_wire_{wfs} \right] \quad (8-1)$$

$$C_R = \frac{PGMA_{wfs}}{CW_{wfs}} \quad (8-2)$$

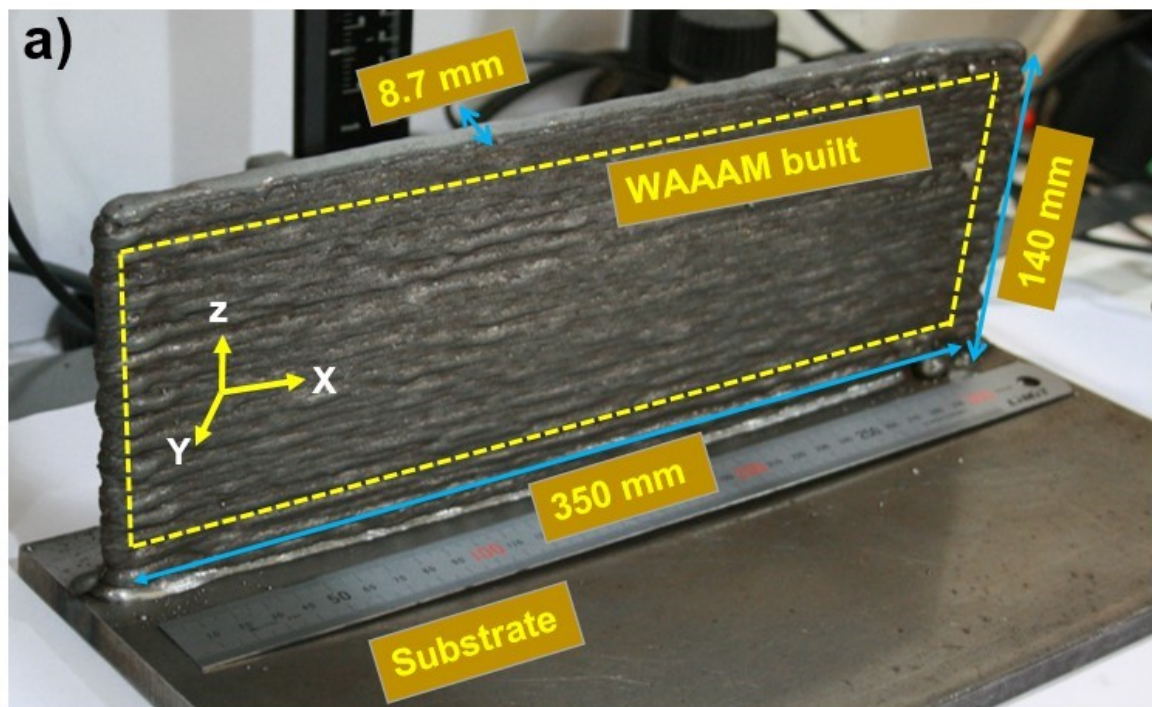
$$S_f = \frac{T_{wfs}}{T_s} \quad (8-3)$$

8.2.3 Mechanical and impact toughness characterisation

Tensile specimens were prepared according to ASTM E8/E8M standard in a practical design to assess the mechanical properties of the machined condition

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of the demonstrated part with an Instron 5500R electro-mechanical testing machine equipped with a load cell of 30kN, and crosshead speed (1mm/min) at room temperature (20°C). During the tensile test, the samples were equipped with reflective tape to measure the elongation (gauge length = 25mm) using a laser extensometer. Because of the limited thickness of the as-built wall, the adjusted dimensions of Charpy test samples were adjusted to 55 x 10 x 2.5mm and 55 x 10 x 5mm with a 2mm V-notch at the centre, according to ASTM A370. The tests were carried out at a temperature of -30°C using a Charpy Leeds Ls102DE impactor with a striker of 600J energy and 8mm radius. The build structure, testing matrix and tensile with Charpy geometry are shown in Figure 8-7(a, b & c).



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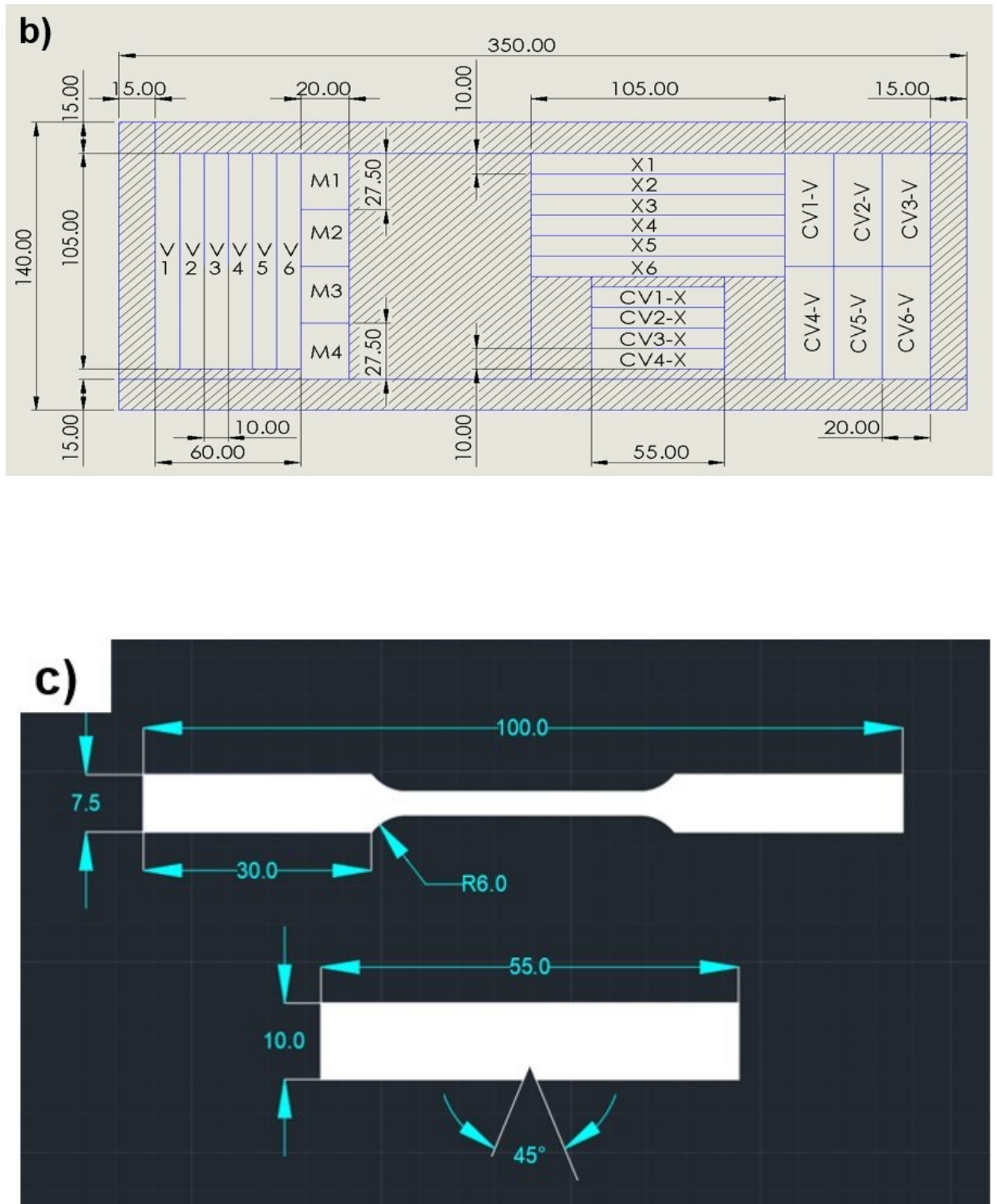


Figure 8-7 WAAAM deposited structure

(a) Build deposited wall (b) Testing matrix (c) Tensile sample; Charpy impact sample

8.3 Results

8.3.1 PGMA bead on plate process study

Figure 8-8 and Figure 8-9 present the PGMA bead on plate studied and the process stability video. The points with the red markings in Figure 8-8, are the combined regions of WFS and TS that would not give a successful bead shape and good appearance, as a result of melt pool interaction with the substrate as shown in Figure 8-9. At WFS of 6.7m/min and TS of 0.4m/min, the molten pool was stable and calm as a result of the appropriate arc pressure. But at WFS of 12.56m/min, TS of 0.8m/min and WFS of 9.98m/min, TS of 1.2m/min, the molten pool vibrates and hence is pushed away by the arc pressure giving rise to keyhole defects and lack of fusion beads with humping and sagging at the higher TS regimes, as shown in Figure 8-9(b & c). The subsequent CW addition studies were developed around this initial build envelope.

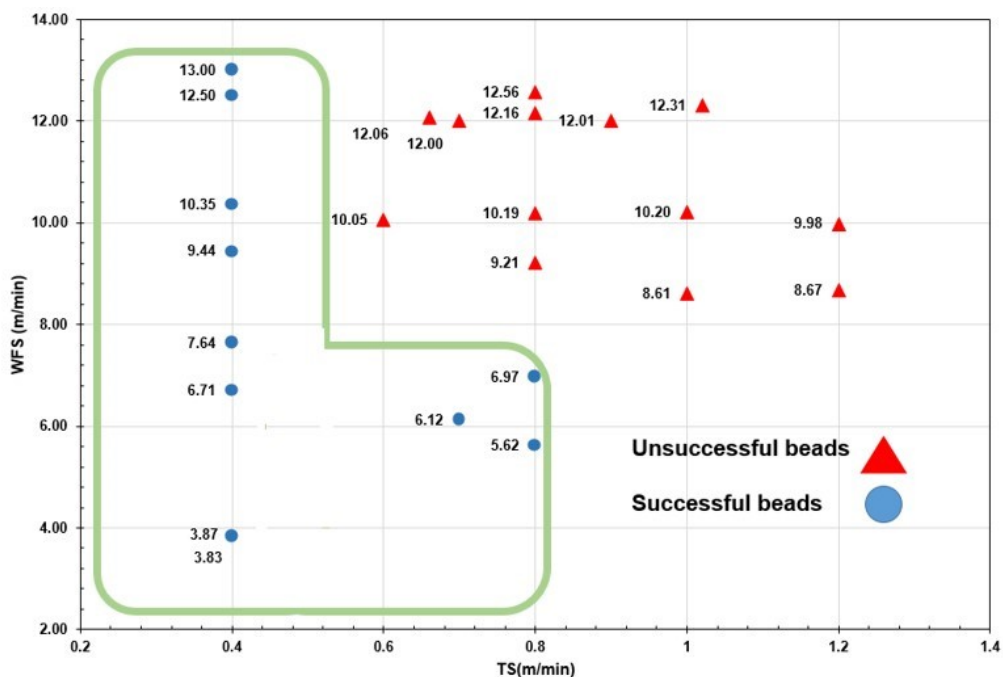


Figure 8-8 PGMA bead on plate studies (build envelope)

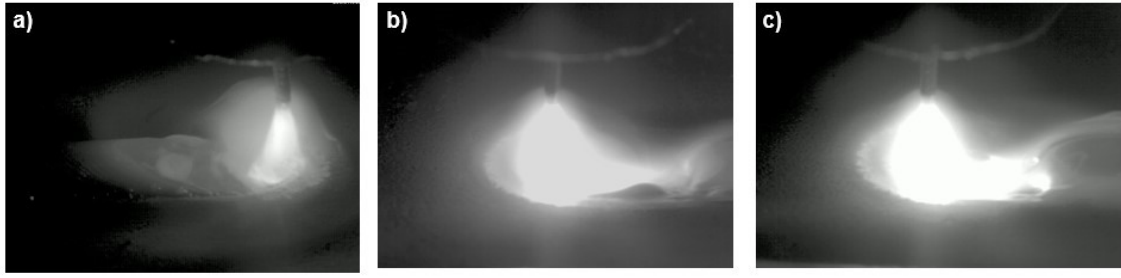


Figure 8-9 PGMA process stability video

(a) WFS; 6.7m/min, TS; 0.4m/min (b) WFS; 12.56m/min, TS; 0.8m/min (c) WFS; 9.98m/min, TS; 1.2m/min

8.3.2 WAAAM bead on plate studies and process envelope

Figure 8-10 and Figure 8-11 present the RF and FF details. The rear feeding of the CW is posed with the challenges of entering the hotter region of the melt pool without an appropriate droplet as shown in Figure 8-10(a), despite using a low feed angle ($<20^\circ$). Increasing the angle ($>20^\circ$) on the other hand, leads to the CW stroking the base plate at a higher CW WFS rate, as shown in Figure 8-10(b). The front feeding is more suitable at higher CTWD (17.5mm) as shown in Figure 8-11(b), as lower CTWD leads to the destabilisation of the pulse arc and spatter beads this might not be suitable for WAAAM.

The building envelope of parameter selection for the WAAAM deposition beads is presented in Table 8-3. Figure 8-12, Figure 8-13 & Figure 8-14 display the physical features of the deposited WAAAM beads. In Figure 8-12 & Figure 8-13, the PGMA was kept at a constant feeding rate of 8.0m/min while the CW was being fed into the melt pool at different feed rates. Figure 8-12 (a-n), carried out at a TS of 0.4m/min, showed different bead attributes from Figure 8-13 (a - g) & Figure 8-13 (h - n), which were carried out at a TS of 0.6m/min & 0.8m/min respectively. Spatter found in Figure 8-12 (a - b) deposits were as a result of feeding the CW into the arc (wire positioning is sensitive to the process). Porosity found in Figure 8-12 (d - j) deposits were as a result of inadequate shielding of the melt pool (15L/min) (high volume of shielding gas is essential for a good

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WAAAM bead appearance void of porosity). Hence a special shielding arrangement is required.

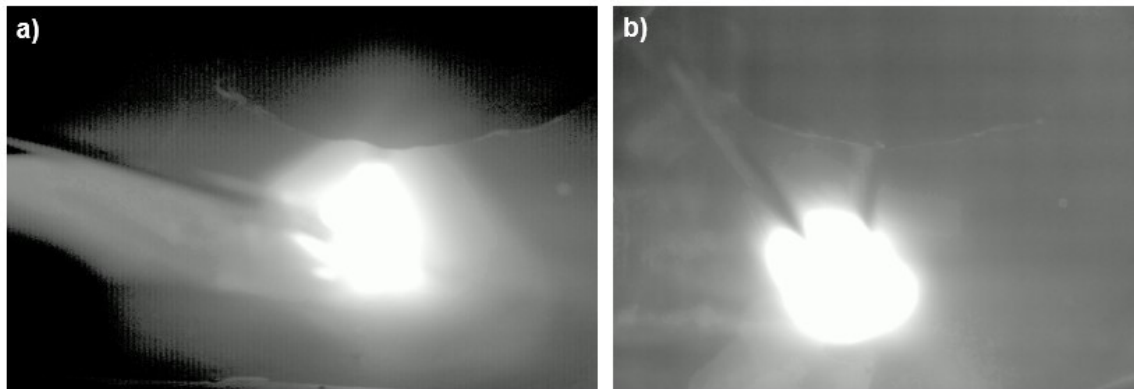


Figure 8-10 Rear feeding trials (angle of feed)

(a) Feeding angle $<20^\circ$ (b) Feeding angle $>45^\circ$

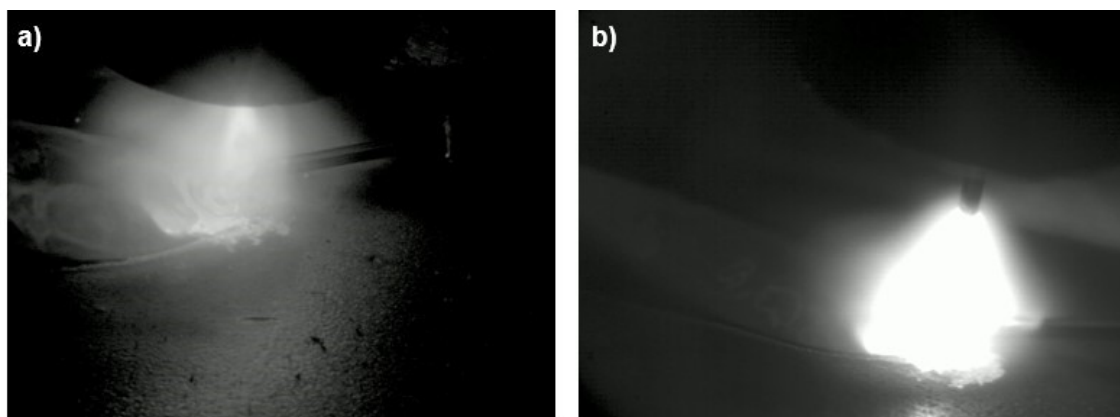


Figure 8-11 Front feeding trials

(a) CTWD: 13mm (b) CTWD: 17.5mm

Table 8-3 Feeding CW into PGMA at constant WFS of 8.0m/min

Trial parameters	Cold wire WFS (m/min)				
	4.0	5.0	6.0	7.0	8.0
TS (m/min)	4.0	5.0	6.0	7.0	8.0
0.4	Excellent	Excellent	Excellent	Excellent	Excellent
0.6	Excellent	Good	Good	fair	fair
0.8	Excellent	fair	fair	poor	poor

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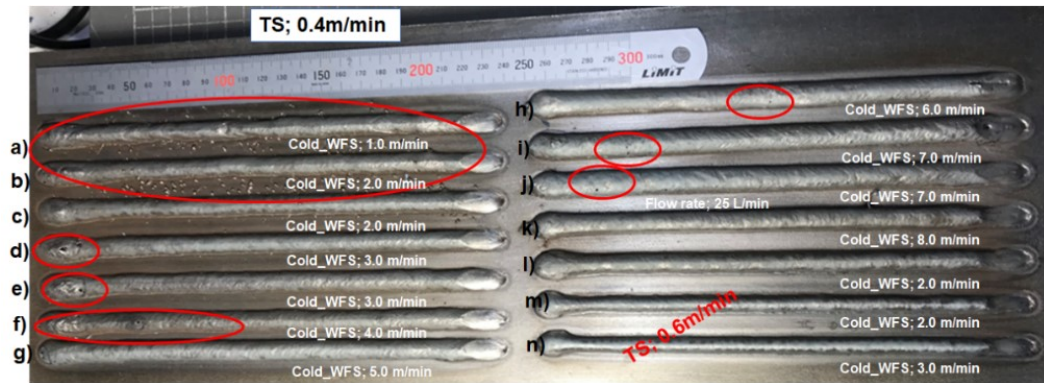


Figure 8-12 Bead profile of WAAAM (PGMA WFS at a constant rate of 8.0m/min, TS: 0.4m/min)



Figure 8-13 Bead profile of WAAAM (PGMA WFS at a constant rate of 8.0m/min)
(a-g), TS: 0.6m/min, (h-n), TS: 0.8m/min



Figure 8-14 Bead profile of WAAAM (constant deposition rate of 6.27Kg/hr)
(a-f), TS: 0.6m/min, (g-l), TS: 0.4m/min

In Figure 8-13 (a-g), it was also evident that the CW could not be fed beyond 8.0m/min because it was characterised by the periodic undulation of the weld bead (hump and valley). The occurrence of these features in a WAAAM deposit is a pointer to the fact that the limit of either the WFS or the TS has been crossed. Other observed features which proved that the limits were passed were the undercuts observed. In Figure 8-13 (h-n), CW feeding was only possible up to a WFS of 5m/min. Feeding CW at WFS of 5m/min & 6m/min at TS of 0.8m/min gave irregular bead profile with undercuts as presented. Feeding CW at WFS of 7m/min & 8m/min was characterised by stocking of the wire in the melt pool, humping, valleys and undercuts. An optimised build envelope design for this WAAAM process (feeding CW into PGMA on constant WFS of 8.0m/min) is shown in Table 8-3.

It is evident from these studies that as the CW WFS increases above certain limits, the formation of discontinuous WAAAM beads with defects starts to manifest, and this becomes worse at higher TS. In Figure 8-14(f & l), the physical appearances of the WAAAM bead at the same WFS (6.0m/min) on both TS of 0.4m/min & 0.6m/min showed that there was insufficient heat to melt the CW, hence the erratic wire protrusion from the melt pool, incomplete fusion and poor melt pool stability. Feeding of the PGMA & CW at WFS of 7.0m/min & 5.0m/min with TS of 0.4m/min resulted in bead which showed a lack of fusion at the toe of the bead and undercut while at TS of 0.6m/min it showed a lack of fusion at the toe blend also. In general, the bead profiles were representative of the WFS ratios.

8.3.3 WAAAM bead profile characterisation and trends

Figure 8-15 presents the macros of bead characteristics of WAAAM deposited at different deposition rates. Macros showing the bead characteristics of WAAAM deposited at a constant deposition rate are presented in Figure 8-16. It is evident from Figure 8-15 that the deposited bead aspect ratio (AR) changes as the deposition rate increases. As the cooling ratio (C_R) ratio reduces by the incremental increase in CW WFS (at constant PGMA WFS of 8.0m/min), the deposition rate increases but the fusion at the surface and penetration reduces

and is more severe at faster TS, as seen in Figure 8-15(l). Bead fusion and penetration is a major factor to be considered in WAAAM of parts since it is linked to the base material HAZ area, residual stress, distortion, and shrinkage. Cooling ratios C_R ratios of 1.6, 1.14 & 1.0 at TS of 0.4m/min, as shown in Figure 8-15 (b, c, & d), will be suitable for WAAAM high productivity.

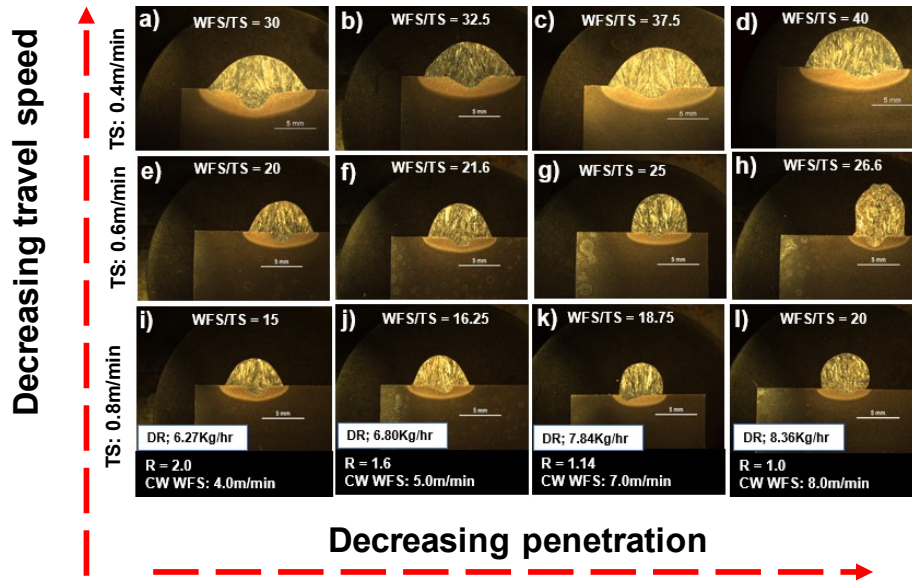


Figure 8-15 WAAAM bead profile at different deposition rate
(a-d), TS: 0.4m/min, (e-h), TS: 0.6m/min, (i-l), TS: 0.8m/min

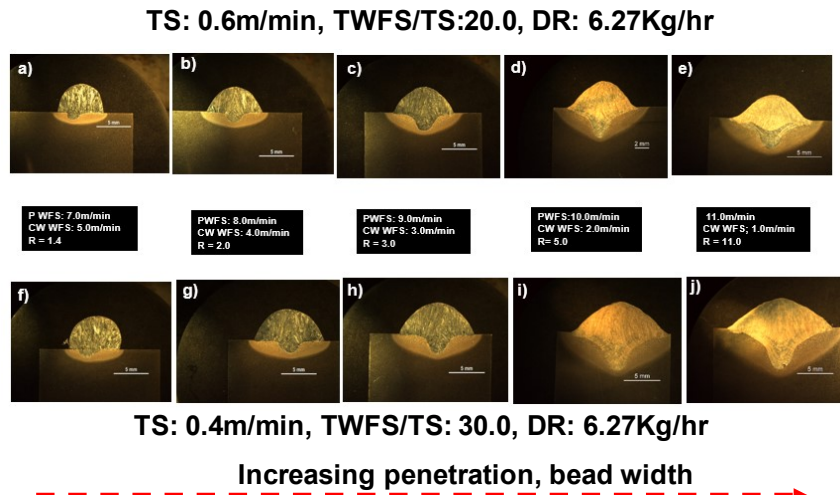


Figure 8-16 WAAM bead at constant deposition rate (6.27Kg/hr)
(a-e), TS: 0.6m/min, TWFS/TS: 20.0, (f-g), TS: 0.4m/min, TWFS/TS: 30.0

In Figure 8-16, the WAAAM bead AR changes as the cooling ratio reduce by the incremental increase in CW WFS and a corresponding decrease in PGMA WFS (achieving a constant deposition rate of 6.27Kg/hr), cooling ratio increases as the bead characteristics improve in fusion at the base and penetration. The beads' profiles are also distinguished by their shape factor(S_f). A higher shape factor obtained at lower TS of 0.4m/min produces WAAAM beads that are more coherent and fused.

8.3.4 Mechanical properties, microstructure and hardness trends

Figure 8-17 presents the stress-strain curve for the various WAAM & WAAAM feed strategies. Figure 8-18 and Figure 8-19, present the microstructural and SEM images at various locations on the build orientation, wrought $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel. The EDS elemental analysis of spotted points is shown in Figure 8-20 and Table 8-4. Hardness variation in front feeding and side feeding sectioned specimens is shown in Figure 8-21 and Figure 8-22. Obviously, the WAAAM front feeding structure performed better than the WAAAM side feeding structure but with low PCE: UTS, YS and PCE of $972 \pm 3\text{MPa}$, $730 \pm 3\text{MPa}$ & 21% as against $935 \pm 5\text{MPa}$, $648 \pm 2\text{MPa}$ & 27%. However, some porosity was seen in side feeding as a result of the less shielding gas used, which might be responsible for the lower strength. Both front and side feeding mechanical results were better than the CMT and PGMAW deposit, as shown in Figure 8-17. There was 14.8%, 12.6% and 58.9% improvement in UTS, YS & PCE respectively. This was a result of their microstructure variation.

The WAAM microstructure in Figure 4-20(c & d) and Figure 7-13 possesses different microstructural features from the WAAAM microstructures. The front feeding microstructure consists of mainly bainitic ferrite matrix and martensitic austenite (M-A) constituents. T1 in Figure 8-18(a) displayed the presence of elongated twinned M-A constituent distributed in a bainitic ferrite matrix, T2 in Figure 8-18(b) displayed the presence of blocky lath bainite with some prior austenite grain boundaries and T3 in Figure 8-18(c) displayed lower bainite and lath bainite in a ferrite matrix. The general microstructural evolution is due to the fast heat mop-up from the deposits as a result of cold wire addition. These results

were better than wrought as a result of the presence of refine and low temperature bainite within the ferrite matrix, unlike F-22 material that contains tempered bainite, ferrite and carbides [153]. The SEM images in Figure 8-19 also show that the dispersion of carbides was more uniform in side and front feeding and that these carbides did not form lumps, as in the case of WAAM, hence the mechanical properties.

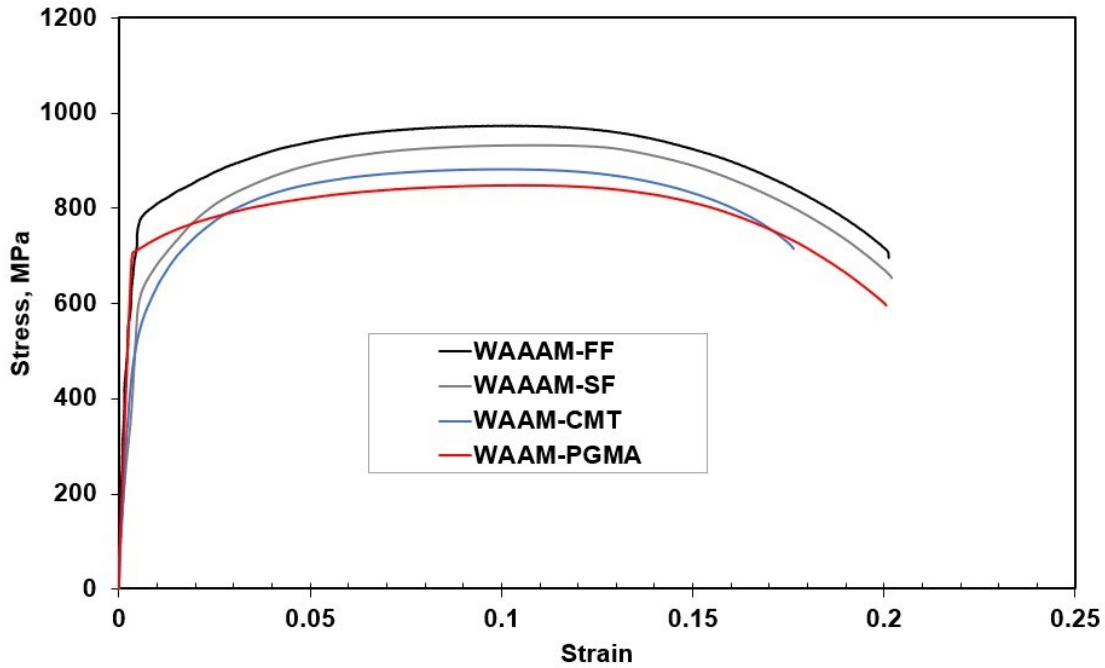


Figure 8-17 Stress-Strain Curve for WAAAM and WAAM strategies

The cyclic heating of this Cr-Mo WAAAM steel due to the sequential layering (single bead strategy) results in the variation in peak temperature and cooling rates of the adjacent layer in the building direction. The difference in peak temperature, which was as a result of the cooling rate, is dependent on the WFS, TS, and dwell time. This leads to the formation of layer bands as seen in Figure 8-21 and Figure 8-22. These bands have distinctive regions on subsequent cyclic heating. This cyclic heating caused annealing and softening of the adjacent layers and hence the variation in hardness along the build direction as seen in Figure 8-21 and Figure 8-22. This phenomenon was also reported for Ti-6Al-4V by [142] and further investigated by [143].

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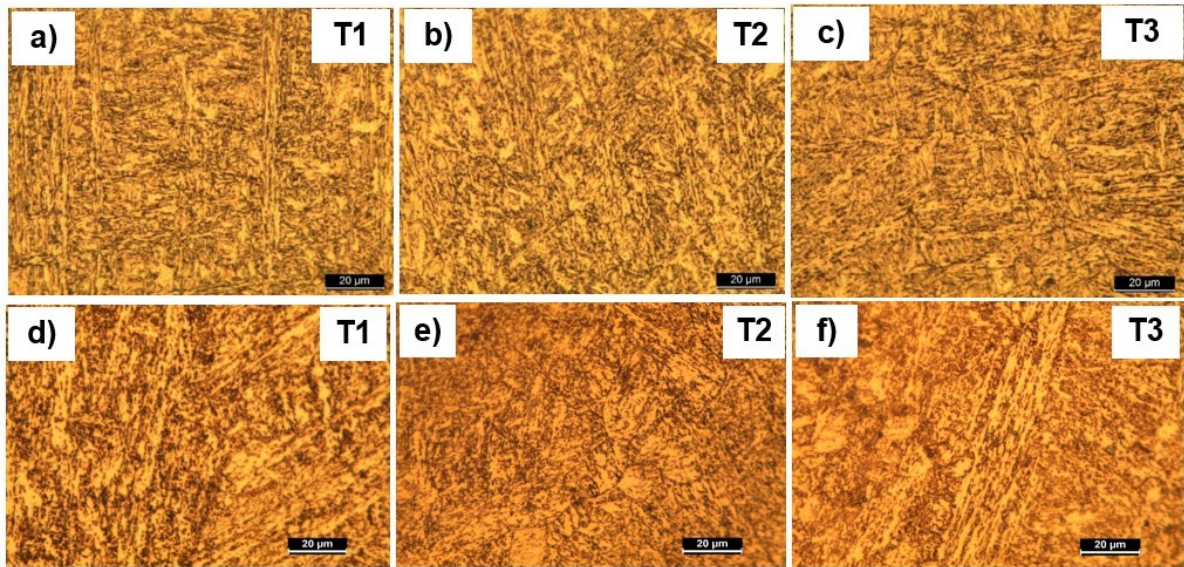


Figure 8-18 WAAAM build microstructural variation

Front feeding (a) Top region (b) Middle region (c) Base region. Side feeding (d) Top region (e) Middle region (f) Base region

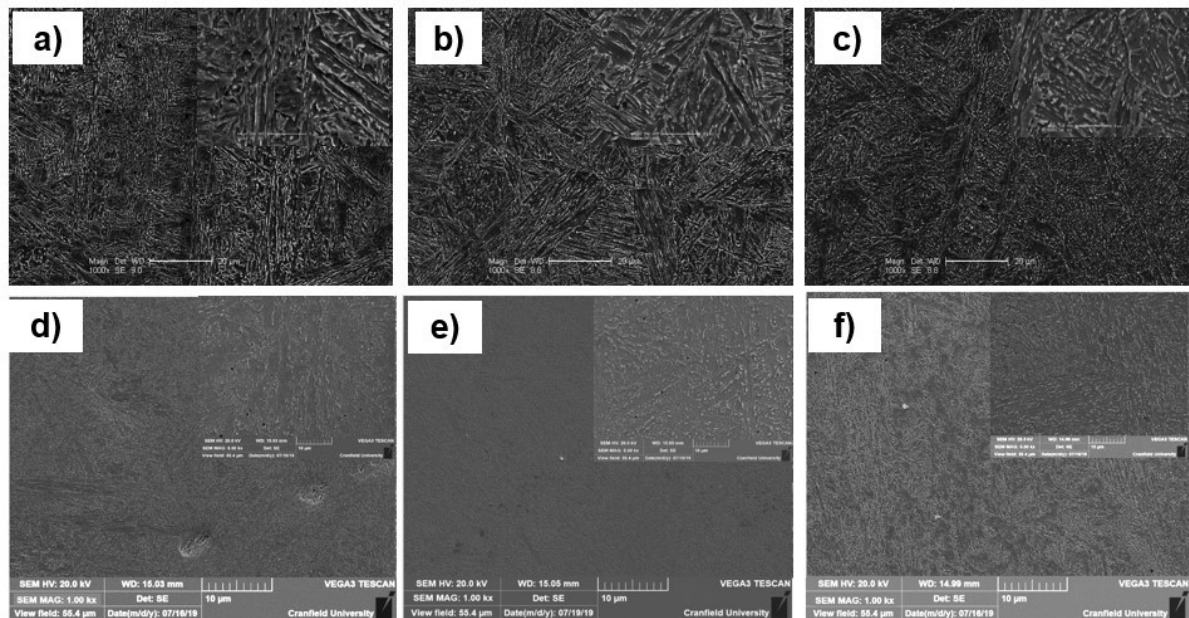


Figure 8-19 WAAAM build microstructural variation - SEM

Front feeding (a) Top region (b) Middle region (c) Base region. Side feeding (d) Top region (e) Middle region (f) Base region

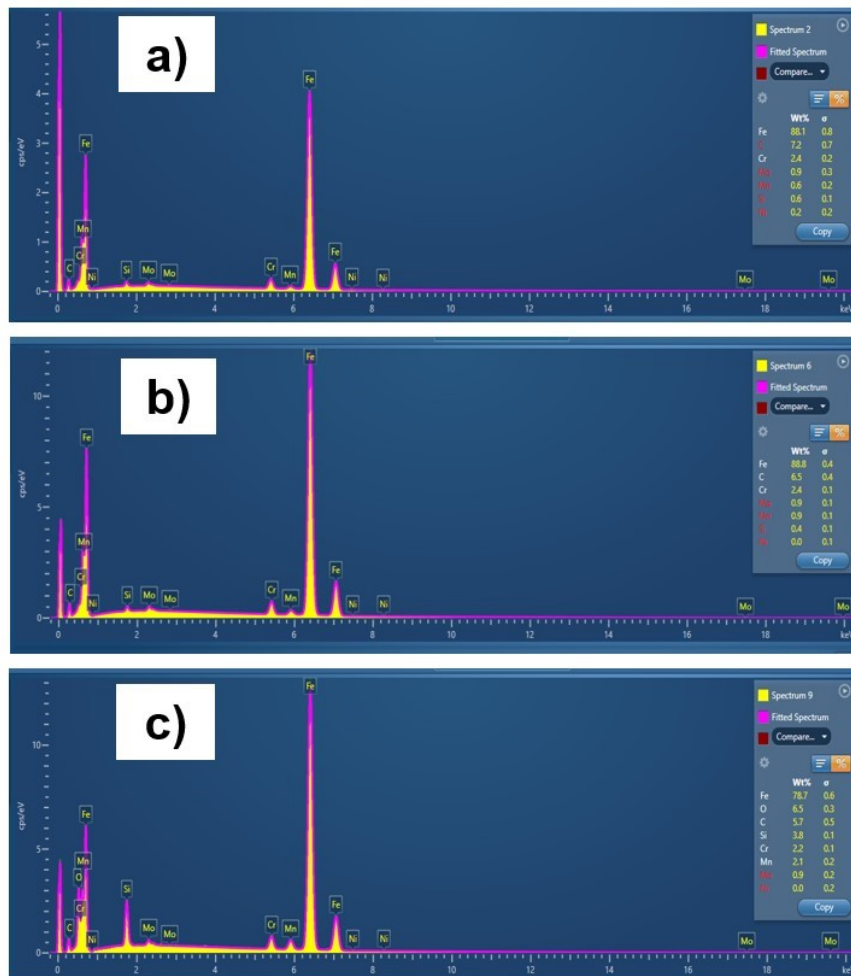


Figure 8-20 EDS line scan of microstructure – Side feeding

Table 8-4 Elemental composition of WAAAM deposit on EDS scan

Spectrum	Fe	C	Cr	Mo	Mn	Si	Ni	O
2	88.1	7.2	2.4	0.9	0.6	0.6	0.2	
6	88.8	6.5	2.4	0.9	0.9	0.4		
9	78.7	5.7	2.2	0.9	2.1	3.8		6.5

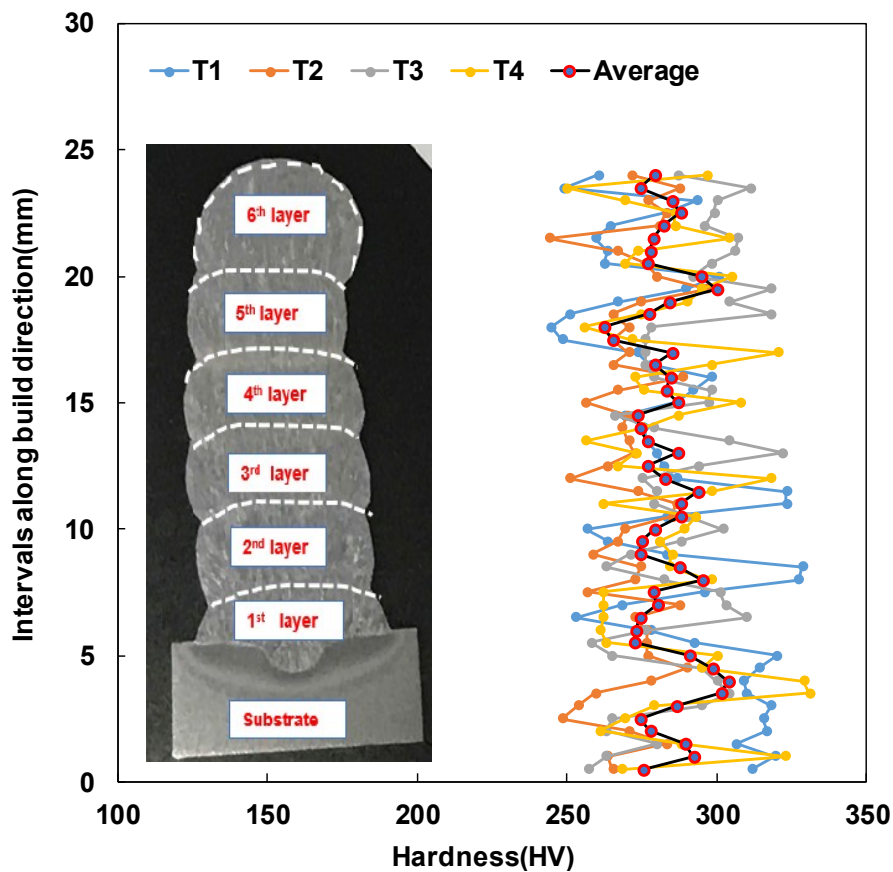


Figure 8-21 WAAAM hardness profile - front feeding

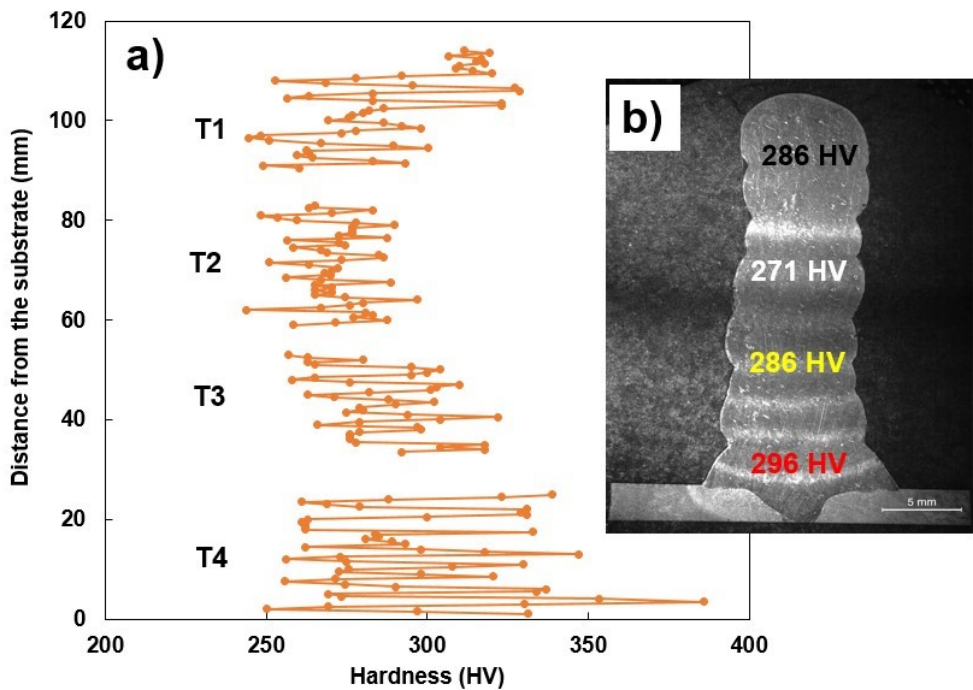


Figure 8-22 WAAAM hardness profile - side feeding

8.4 Discussion of results

8.4.1 Effect of CW on WAAAM bead profile at varying deposition rates

8.4.1.1 Front feeding + PGMA WFS of 8.0m/min

Figure 8-23 presents the effect of FF of CW on the bead profile at constant PGMAW WFS of 8.0m/min. Figure 8-24 shows the arc interactions of front feeding in extreme cases. According to the Young Dupré equation:

$$\cos \theta_c = \frac{\gamma_{SG} - \gamma_{SL}}{\gamma_{LG}} \quad (8-4)$$

where θ_c is the contact angle between the melting pool and substrate, γ_{SL} is the interfacial tension of the melting pool to substrate interface, γ_{SG} is the interfacial tension of the air to substrate interface, γ_{LG} is the interfacial tension of the air to melting pool interface. Since the deposition was done in a controlled environment, γ_{SG} and γ_{LG} are relatively constant. Hence the surface tension between the melt pool and surface is inversely proportional to the activity of the liquid melt pool [154]. Contact angle (CA) is dependent on the CW feed rate (WFS), and temperature of the melt pool. The higher contact angle is obtained at higher CW WFS, as seen in Figure 8-23(a). The thermal cushioning effect induced by the cold wire decreases as the CW WFS increases due to the formation of the colder melt pool and consequently colder base metal. The resultant effects of this lead to an increase in interfacial tension and viscosity of the molten pool and a decrease in the wetting behaviour [54].

Hence higher contact angle is obtained at a lower cooling ratio. The bead wetting behaviour is also reduced, as depicted in Figure 8-15(c-d)(g-h)&(k-l). Higher contact angle is not suitable for WAAM, as the transition between layers will be too sharp leading to a high-stress concentration factor. Re-melting ratio (RR) also decreases as the CW WFS increases, as shown in Figure 8-23(b). Less effective heat is available at the higher regimes of CW WFS. Penetration is a significant requirement for increasing re-melting ratio.

Penetration (P) decreases as the CW WFS increases, as seen in Figure 8-23(c). The CW takes up the available excess heat to melt, more heat is available at low CW WFS, hence the high penetration in this regime. It can be seen from Figure 8-23(c) that TS has a substantial effect on the penetration of the WAAAM deposition; at certain CW WFS there will be no or less penetration, as seen in Figure 8-15(c-d)(g-h)&(k-l). WAAAM does not require too much penetration and too shallow a deposit as a result of the collapsing of the previous layer, fusion defects such as lack of sidewall fusion, root keyhole defects [54] and unfused beads.

In Figure 8-23(d), the aspect ratio (AR) increases as the CW WFS increases. Higher bead height and lower bead width at lower regimes of cooling ratio are responsible for the higher aspect ratio. Lower bead height and higher bead width at higher regimes of cooling ratio lead to a lower aspect ratio as a result of an increase of more arc pressure, more wettability of the molten pool and less volume of the cold wire. It is also evident from Figure 8-23(d) that the effect of TS is significant on aspect ratio since more time is available for melt pool, cold wire and base plate interaction sustained by the higher heat input.

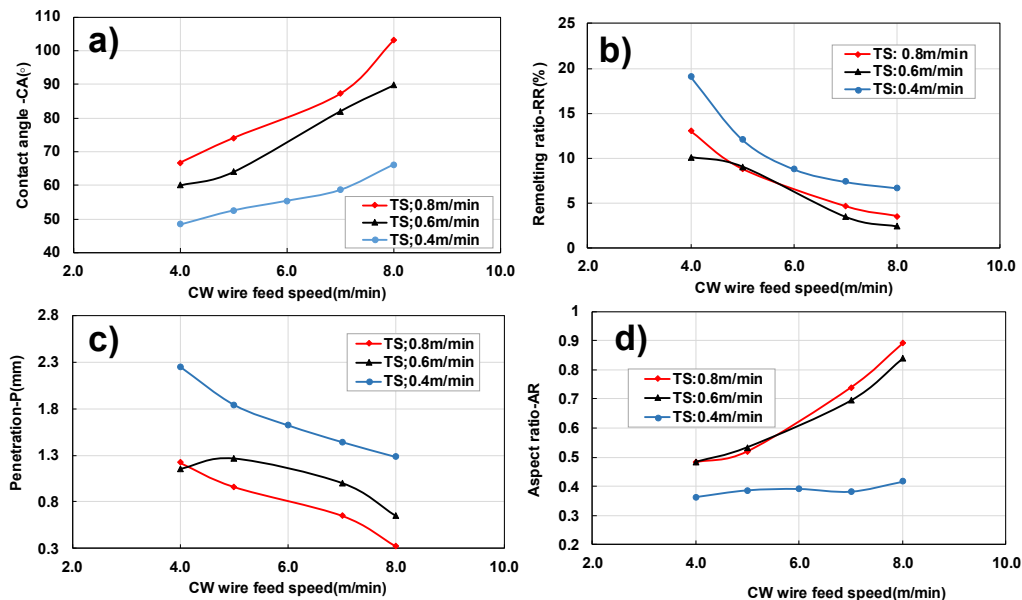


Figure 8-23 Bead characteristics with front feeding using PGMA WFS of 8.0m/min

(a) Contact angle (b) Re-melt ratio (c) Penetration (d) Aspect ratio

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WAAAM bead deposited with CW WFS of 8.0m/min and TS of 0.8m/min, as shown in Figure 8-24(a-c), has an arc characterised with keyhole, lack of fusion (sidewall), and undercut defects, as shown in Figure 8-14(l). The CW WFS was far beyond the limit the melt pool could accept due to the fast TS and less heat available to the melt pool, hence sagging of beads and protruding wire, as shown in Figure 8-13(m). The improper bead shape observed led to a contact angle value of 84.6° and a high aspect ratio of 0.737. This bead shape is not suitable for WAAAM deposition. WAAAM bead deposited with CW WFS of 8.0m/min and TS of 0.6m/min as shown in Figure 8-24(d-f) is characterised by humping and sagging beads with an undercut, lack of fusion as shown in Figure 8-14(f). The cross-section from the staggered bead shown in Figure 8-14(h) leads to a contact angle value of $>90^\circ$, hence not suitable for WAAAM deposition. Conclusively, moderate penetration & re-melting ratio with low contact angle & aspect ratio are required with WAAAM to build actual components.

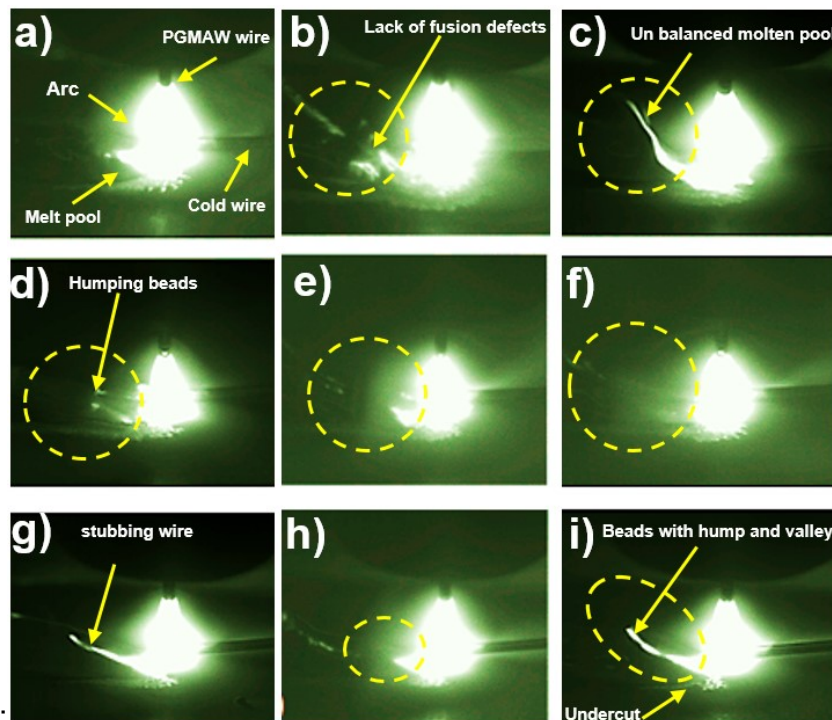


Figure 8-24 Arc interaction

(a-c) PGMA WFS 8.0m/min, CW WFS 8.0m/min, TS: 0.8m/min, (d-f) PGMA WFS 8.0m/min, CW WFS 8.0m/min, TS: 0.6m/min (g-i) PGMA WFS 6.0m/min, CW WFS: 6.0m/min, TS: 0.6m/min

8.4.1.2 Side feeding + PGMA WFS 8.0m/min and 12.0m/min

Figure 8-25 presents the WAAAM bead characteristics of side feeding with PGMA WFS of 8.0m/min. Figure 8-26 also shows the WAAAM bead characteristics at a higher PGMA WFS of 12.0m/min. The contact angle in side feeding follows the same trend with front feeding, as shown in Figure 8-23(a), increasing from the low regimes of CW WFS to high regimes of CW WFS. The contact angle at higher TS in the side feeding shows lower values compared to front feeding. However, the contact angle is on the increase at higher regimes of CW for both feedings. The contact angle at higher PGMA WFS (12.0m/min) is relatively lower than the contact angle for lower PGMA WFS (8.0m/min), as seen in Figure 8-26(a).

The increase in wettability of the molten pool, as a result of arc pressure increase governed by the high WFS, is responsible for this low contact angle despite increasing the CW feeding. Bead height in both high and low PGMA WFS shows an increasing trend as more CW is being added, as seen in Figure 8-25(d) and Figure 8-26(d). Bead width in the low PGMA WFS (8.0m/min) is decreasing as the CW WFS increases (Figure 8-25(b)), unlike the PGMA WFS of 12.0m/min, where the bead width is increasing (Figure 8-26(b)). This is as a result of the availability of more heat with the 12.0m/min PGMA, hence a less viscous molten pool that spreads faster regardless of the CW addition, except at the level where CW intake saturation is reached. The aspect ratio obtained in the side feeding is lower than the front feeding, due to the offset configuration arrangement of both feeding directions. The penetration achieved is a function of the PGMA WFS and the CW WFS. Penetration in the 12.0m/min at various CW feedings were all higher than Penetration obtained with 8.0m/min at the same CW WFS, as seen in Figure 8-25(c) and Figure 8-26(c).

WAAAM deposition is suitable with less penetration. Penetration in the front feeding strategy is higher than the side feeding strategy, even at the same CW feeding rate. Side feeding gives 60% reduction in penetration at a CW WFS of 8.0m/min when compared to front feeding at the same PGMA WFS of 8.0m/min. This means that side feeding might have more potential in building WAAAM AM structures than front feeding. Secondly, the regimes of PGMA WFS could be

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used to change the aspect ratio (bead height and width), penetration and contact angle of the WAAAM deposition process, thus controlling the deposition profile, heat built-up (effective heat) and SW (although not studied here).

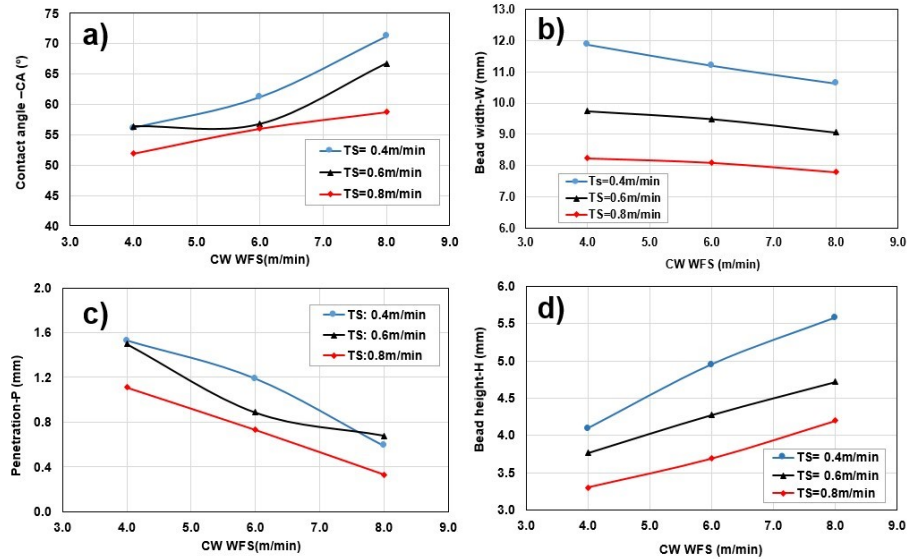


Figure 8-25 Bead characteristics with side feeding using PGMA WFS of 8.0m/min

(a) Contact angle (b) Bead width (c) Penetration (d) Bead height

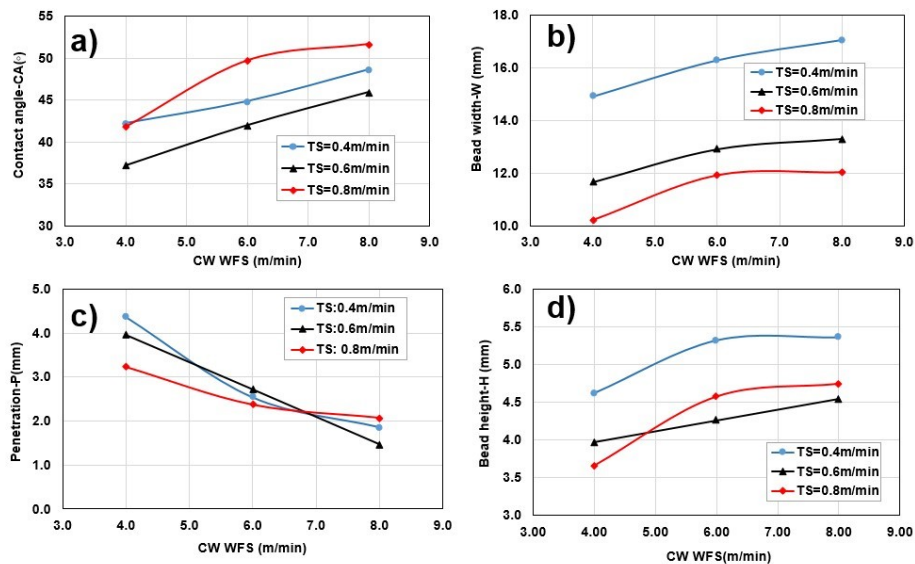


Figure 8-26 Bead characteristics with side feeding using PGMA WFS of 12.0m/min

(a) Contact angle (b) Bead width (c) Penetration (d) Bead width

8.4.2 Effect of WAAAM bead characteristics at different cooling ratio and a constant deposition rate of 6.27Kg/hr.

Figure 8-27 and Figure 8-28 present the effects of WAAAM bead on cooling ratio (C_R) as define in Eq (8-2) at a constant deposition rate of 6.27Kg/hr. it was found that the bead width is lower at a low and intermediate value of C_R . The effects of interaction between bead width and shape factor as defined in Eq (8-2) (S_f) are evident in Figure 8-16 (a-j). BW gradually increase with an increase in C_R as seen in Figure 8-27 (d). Bead width is low at generally lower regimes of C_R ratio as a result of faster cooling of the melt pool and the non-wettability behaviour of the melt pool in this regime as a result of a higher proportion of CW in the mixture (in-feed) which cushions the thermal effects.

This is unlikely to be the case at higher regimes of C_R where more heat is available even under the lesser CW feeding rate, hence enough wettability is obtained in this region. The C_R also has a strong influence on bead height (BH) especially in the lower regime of C_R . The faster cooling of the melt pool, as a result of the high volume of CW and lower arc temperature, leads to a reduction in the wettability behaviour of the melt as a result of the heightened interfacial tension and viscosity of the melt pool in these regimes. This is responsible for the reduction in bead width and increases in the bead height since there is no room for a molten pool to flow and wet the base plate.

It is also evident from Figure 8-27 (b) and Figure 8-28(a) that the C_R has a higher influence on bead width and lower influence on aspect ratio, especially at lower C_R where wettability is an issue. The aspect ratio drops rapidly with increasing C_R as evident in Figure 8-28(a). Aspect ratio follows the same trends as bed width and bead height phenomena. Higher bead height and lower bead width at lower regimes of C_R gave way to the higher aspect ratio while lower bead height and higher bead width at higher regimes of C_R produced a lower aspect ratio as a result of an increase of more arc pressure, more wettability of the molten pool and low volume of the CW addition.

Penetration increases as the C_R increases; it is less sensitive to the (S_f) as shown in Figure 8-28(c). At lower regimes of C_R , where we have more CW in-feed, the arc pressure is being forced to the minimal, continuous increase of CW feeding beyond the accepting capability which will result in weld beads with staggered and protruding defects with lack of penetration.

The re-melting ratio (RR) and dilution (D) are dependent on the C_R ; RR & D show little sensitivity to the (S_f) at lower regimes and higher susceptibility to (S_f) at higher regimes of (S_f) . More re-melting ratio and dilution are seen at higher regimes of (S_f) due to more available heat input (more arc pressure) at this regime and lesser cold wire feeding. Hence the melt pool can quickly take a lesser volume of feed wire. High penetration is also available in these regions. The extremely high CW feed rate of 6.0m/min and PGMA feed rate of 6.0m/min (6.27Kg/hr deposition) with TS of 0.6m/min revealed stubbing of wires, staggering beads, undercut and lack of fusion defects, as shown in Figure 8-14(f) as a result of CW WFS having crossed its acceptance limit since there is insufficient heat to melt them both.

Contact angle as seen in Figure 8-27(a) decreases as C_R increases; it is less sensitive to (S_f) at lower regimes of C_R . The reduction in the wettability behaviour of the melt pool at the lower region of C_R , which heightened the interfacial tension and viscosity of the melt pool, leads to bead height increase and consequently increase in contact angle. The higher contact angle is not suitable for additive manufacture, as the wall will be too rough. Optimising the deposition process using C_R of 2 - 6 will give low contact angle and consequently low SW and low SCF, which is one of the key challenges with high deposition using one wire with the PGMA process or in tandem with two wires. Here the increase in productivity by increasing WFS incurs high SW while in WAAAM (PGMA + CW) increases in productivity at specific C_R deter high SW.

8.4.3 WAAAM uniqueness – effective heat and dilution

One of the attainments with the CW addition is the reduction in the amount of base material dilution. High dilution of the base material, which is as a result of

the large HAZ area and penetration, has a direct link with increased residual stress, distortion, and shrinkage of the base substrate, which could be expensive in the case of Ti and other special alloy materials, but with WAAAM, the excess heat that causes this is the mop-up by the CW addition. The effective Heat input is decreasing as the volume or % of CW increases in the deposition, as shown in Figure 8-28(c & d), this is the novel key discovery in this study. The WAAAM process serves to control heat input effects on deposited structures, reduction in base substrate dilution due to reduction in the HAZ area, reduction in residual stress, distortion and shrinkages of parts as a result of the flexibility of independently controlling the heat input effects of deposits. All the above adepts are shortcomings from the current power sources used of WAAM-AM.

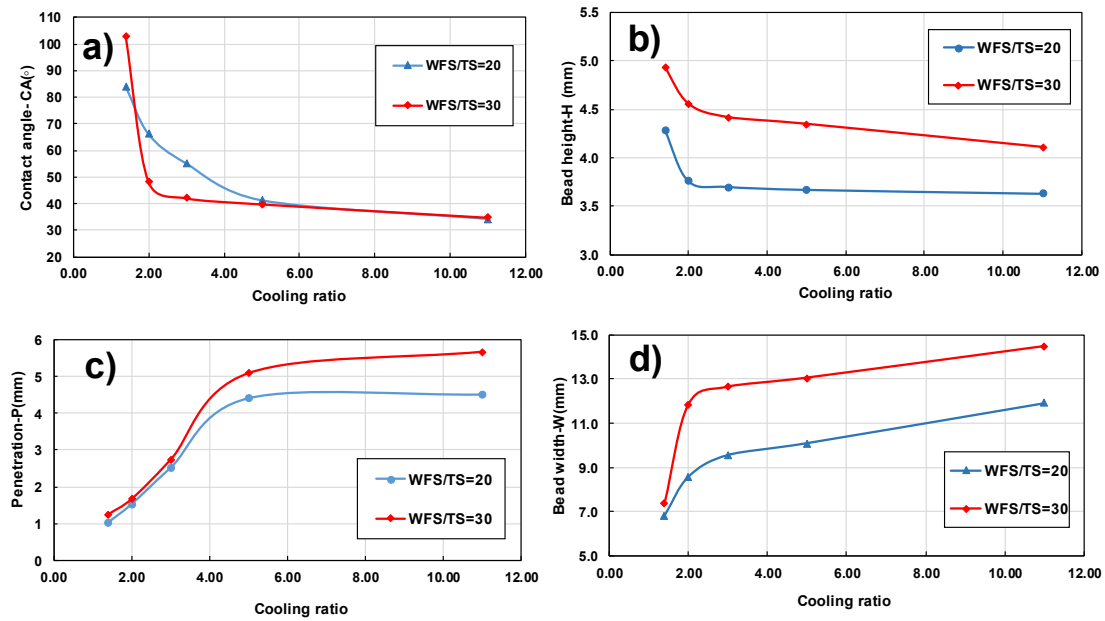


Figure 8-27 Effects of WAAAM bead characteristics at different R-ratio

(a) Contact angle (b) Bead height (c) Penetration (d) Bead width

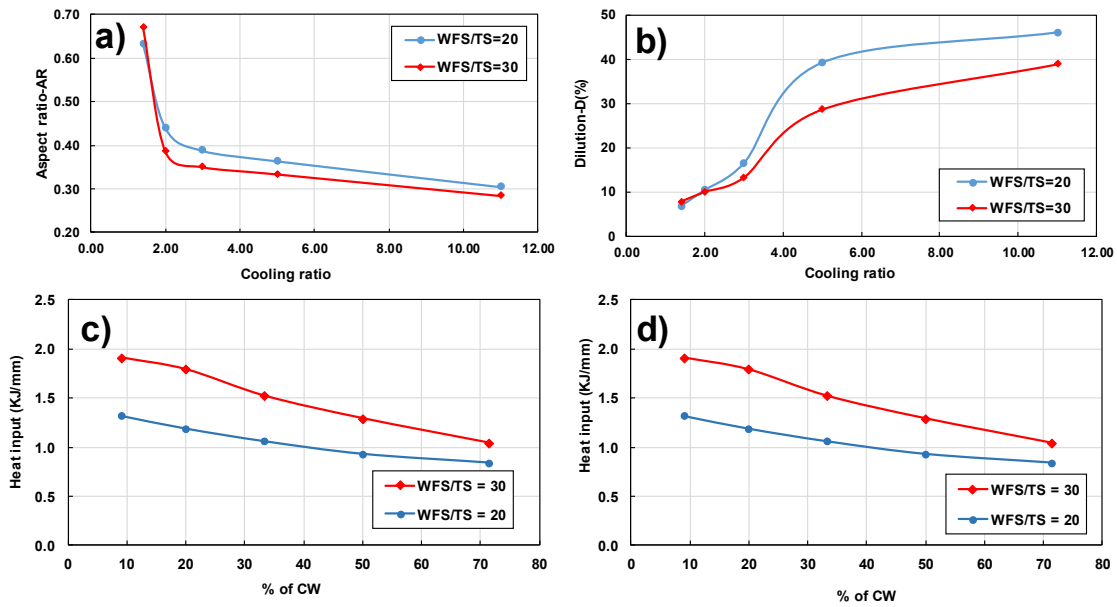


Figure 8-28 Effects of WAAAM bead characteristics at different cooling-ratio

(a)Aspect ratio (b) Dilution (c) Heat input vs. %CW (d) Heat input vs. CW WFS

8.4.4 Effect of grain morphology and texture

The EBSD classification was used to investigate the crystallographic orientation to confirm the relationship between the crystal growth direction and the thermal history of the WAAAM structure during deposition. The EBSD inverse pole figure (IPF) in Figure 8-29 presents the bainitic growth orientation in the top, refine zone and re-melt zone. The legend showing the relationship between the crystal orientation and the colours is inserted in the top right corner. Figure 8-30 shows the corresponding BCC and FCC pole figures. Figure 8-31 presents the higher magnification (X4) of the re-melt zone (Figure 8-29(d)) in a BCC structure for clarity.

The microstructure in the top zone consists of elongated ferrite and martensite oriented parallel to the build direction with length reaching 90 μ m. The microstructure in the refine zone consists of refine and equiaxed ferrite grains with low temperature bainite with average BCC and FCC grain sizes of 12.6 μ m and 0.368 μ m. Both zones have a similar texture with no preferential crystallographic direction during solidification. The re-melt zone consists of

bainite and martensitic laths with a maximum length of 109 μ m growing epitaxially in a random direction.

They show strong texture in both the FCC and the BCC, as depicted in Figure 8-30(a & b) with the {101} plane being the preferential crystallographic direction during solidification. The UTS, YS & PCE of the WAAAM part were 130MPa and 84MPa and 5% higher than the WAAM part. This is due to the promotion of refining the martensite and bainite brought about by the CW addition. The formation of twin martensite in the granular bainite, as reflected in the BCC structure in Figure 8-29(d) & Figure 8-31, is another contributing factor to the unique strength induced by carbon partitioning during austenite to bainite transformation [155].

Table 8-5 Volume fractions of HAGBs and LAGBs

Location	Grain boundaries	
	LAGBs(<15°)	HAGBs($\geq$ 15°)
WAAAM deposit		
Top region	73.0	27.0
In between layers	69.7	30.3
Re-melt region	72.4	27.6

The ductility in the WAAAM structure is better than the WAAM structure as a result of the presence of soft tempered ferrite phases enhanced by the CW addition, hence called transformation induced plasticity process [107]. Figure 8-32 & Figure 8-33 present the grain boundary distribution in the WAAAM and WAAM deposit. Figure 8-34 and Table 8-5 present the grain boundaries distribution and the volume fractions of HAGBs and LAGBs in the WAAAM built. The WAAAM deposit has more high angle grain boundaries (HAGBs) with misorientation $>15^\circ$ compared to WAAM deposit with more low angle grain boundaries (LAGBs) with misorientation $<15^\circ$ as shown in Table 7-3(ER90S-B3). The presence of more HAGBs in the WAAAM was responsible for the improved impact toughness values, as the crack along the HAGBs is twisted to slow the crack propagation path, hence better toughness. [156][157]

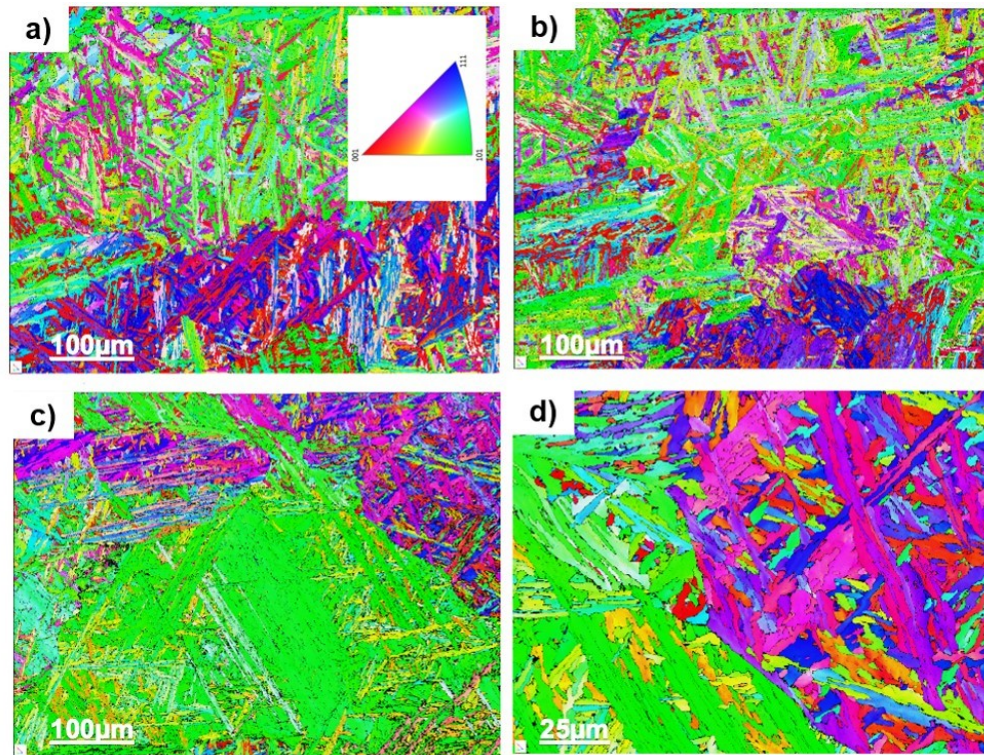


Figure 8-29 Microstructural distinguishing features in WAAAM ER90S-B3

(a) Top region (b) Refine zone (c) Re-melt zone (d) Intercritical ferrite and martensitic lath structures

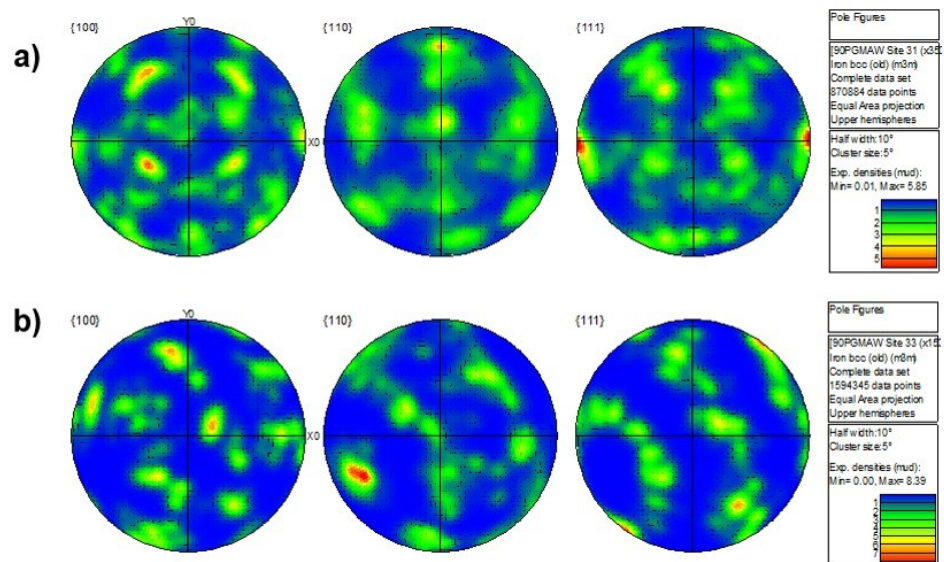


Figure 8-30 Pole figures depicting the strong {100} texture across the re-melt zone

(a) BCC phase (b) FCC phase

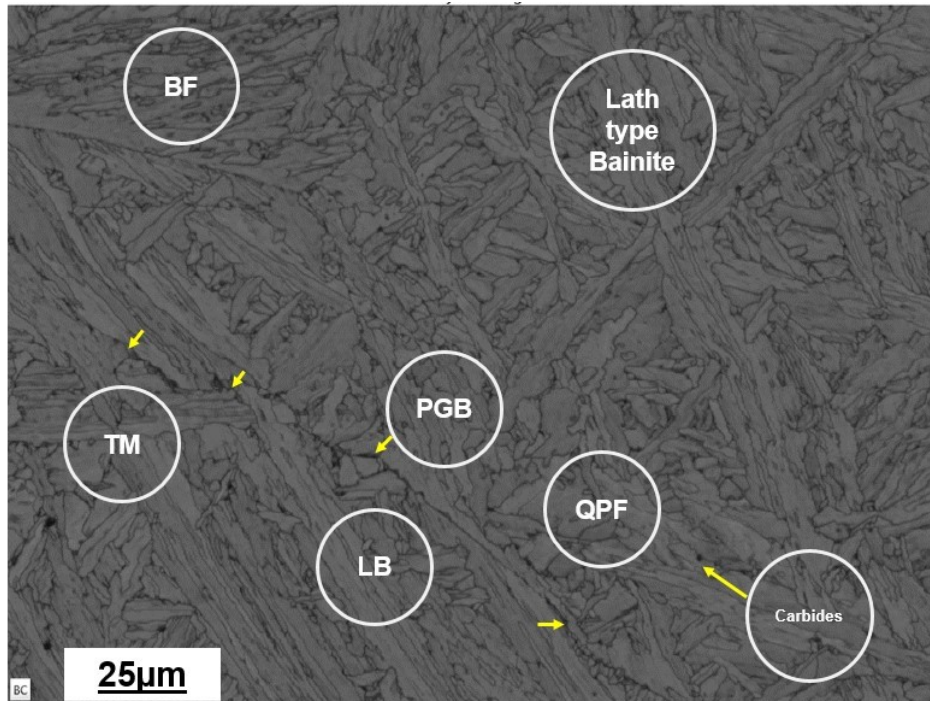


Figure 8-31 Grain boundaries distribution (LAGBs & HAGBs)

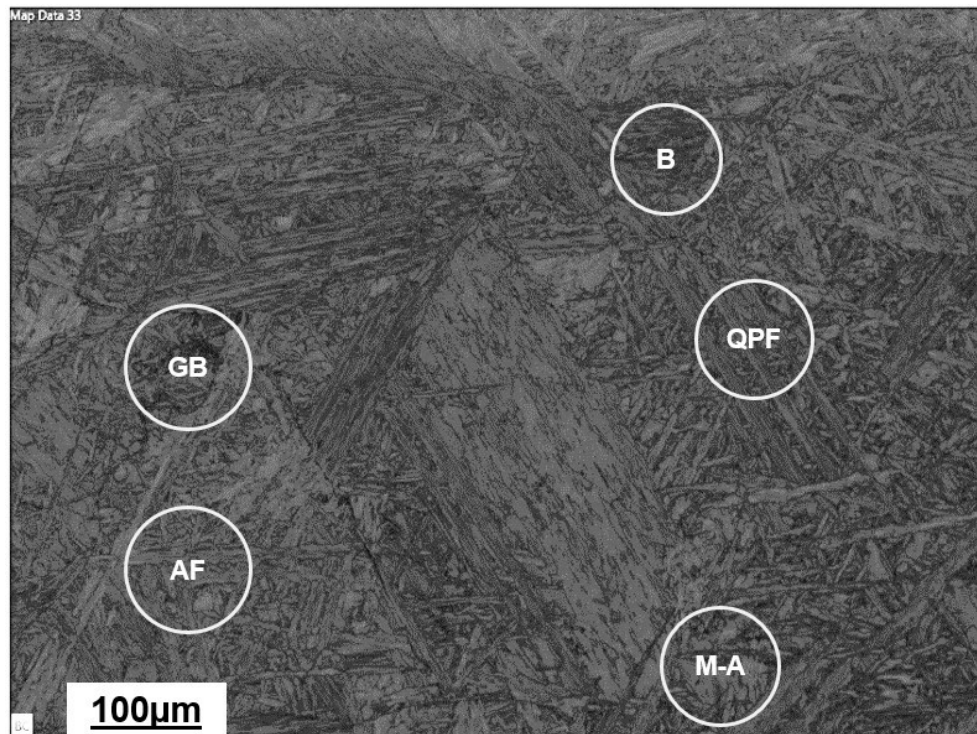


Figure 8-32 Grain boundaries distribution (LAGBs & HAGBs) (WAAAM)

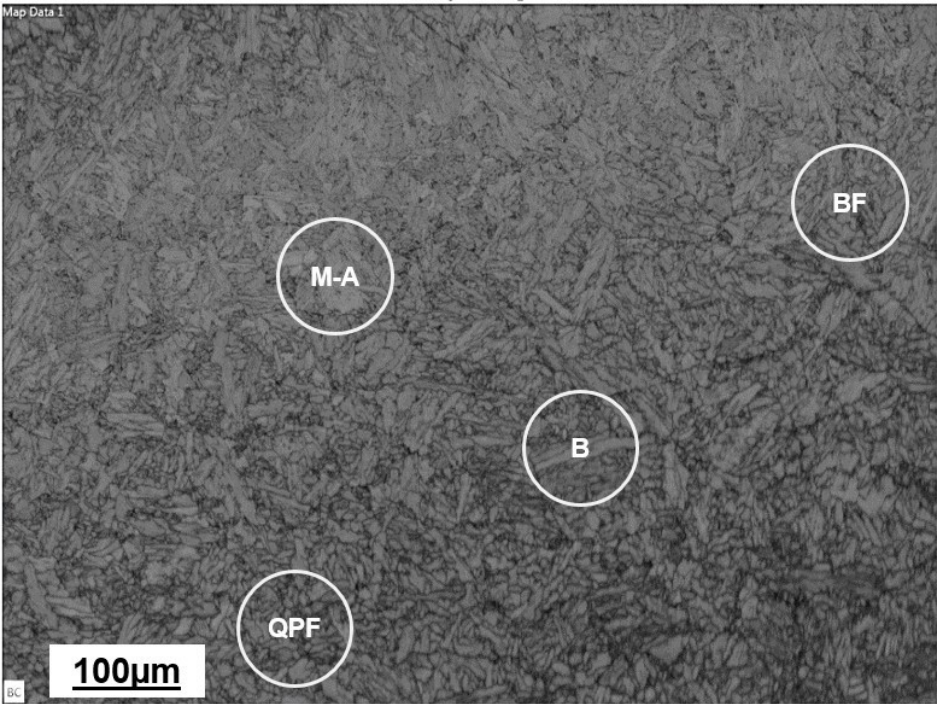


Figure 8-33 Grain boundaries distribution (LAGBS & HAGBS) (WAAM)

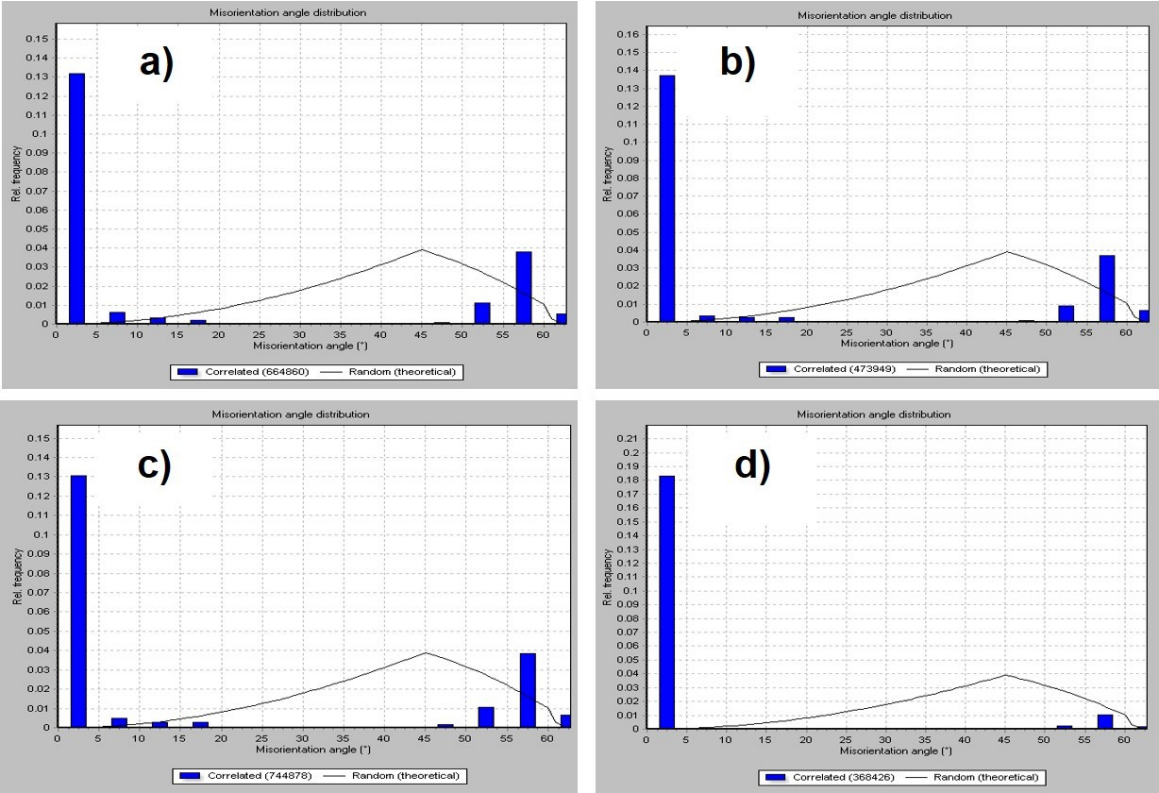


Figure 8-34 Grain boundaries distribution

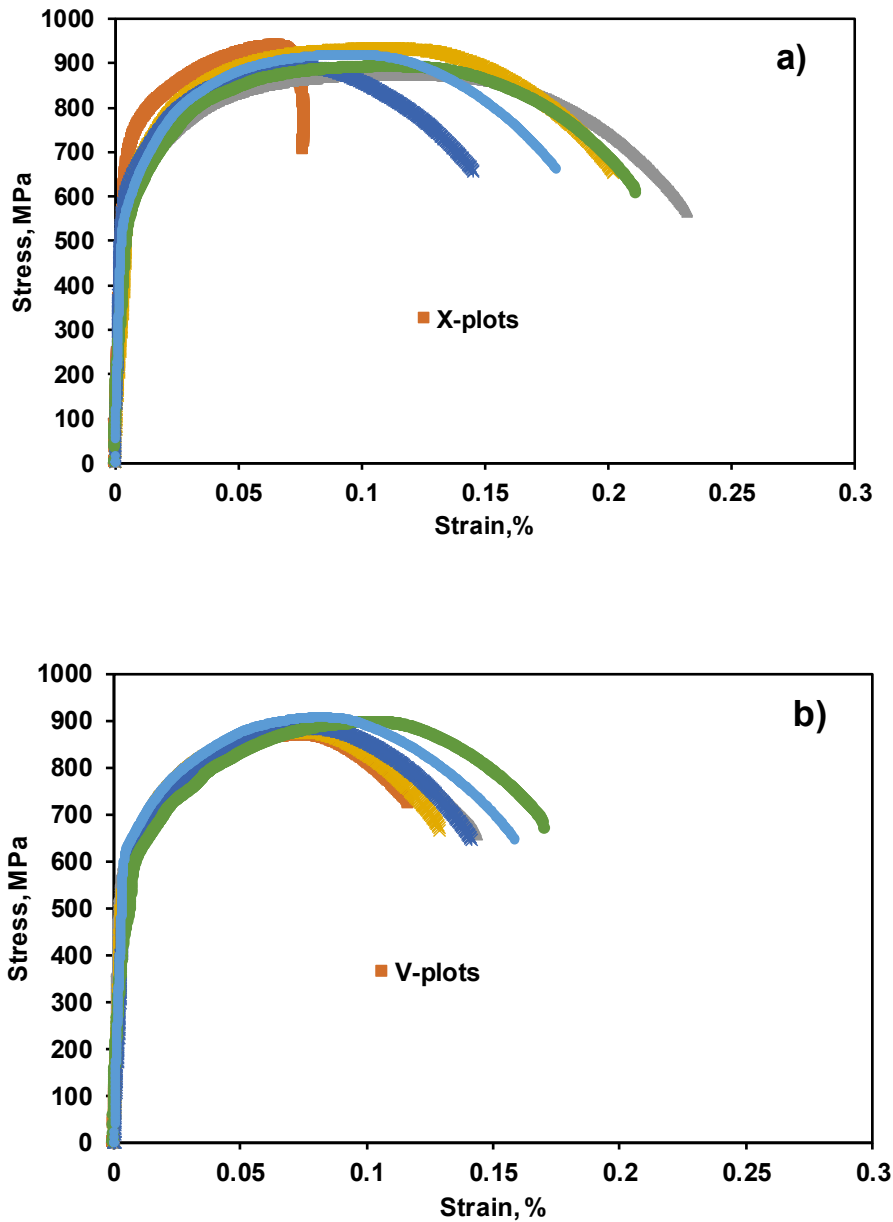
8.4.5 Effect of microstructural evolution on tensile strength and impact toughness

Figure 8-35 presents the anisotropic mechanical properties of WAAAM – 2.4Cr-1Mo steel. The average mechanical properties of both WAAAM strategies and WAAM are compared in Figure 8-36. The tensile fracture morphologies of the WAAAM structure are shown in Figure 8-37. The CW addition, which mops up the available heat from the melt pool, refines and eliminates the adverse effect of high heat input on the microstructure and hence slow cooling. It is believed that the deposits also received second phase strengthening due to the incomplete decomposition of M-A constituents [158].

The tensile fracture surface of the WAAAM samples showed typical ductile fracture mode with a high density of small dimples, which is one of the reasons for the improved mechanical properties. It is believed that the impact toughness of 2.4Cr-1Mo steel is dependent on crack initiation and propagation rate which is always affected by the distribution of dispersed carbides, precipitates in the form of inclusions, hard M-A constituents and bainite grain size [154][159][160].

Figure 8-38 presents the impact toughness relationship between the WAAAM and WAAM process and wrought. The WAAAM impact toughness fractured morphologies and EDS count are depicted in Figure 8-39. The Charpy impact value is higher in the WAAAM sample than the CMT WAAM deposits, as seen in Figure 8-38. This is due to the slow cooling and improvement in ductility of the WAAAM over other deposition strategy and methodology. The fractured surface of the FF (Figure 8-39(d)) of the impact toughness samples showed some inclusion sites, and the electron spectroscopy analysis of the first spot (EDS spot 1) revealed some inclusion of MnS, AlO, and Al-Si. It was reported that [147] the non-uniform straining of the matrix and dislocation locking caused by these secondary precipitates and hard phases are responsible for the fracture behaviour of this Cr-Mo steel. This could be responsible for the scatter obtained in the hardness plot as shown in Figure 8-21 and Figure 8-22. This is always prone to the base of the substrate in AM. This scatter is as a result of high thermal

cooling at the base of the substrate occasioned by the surface of the substrate and the higher WFS used for the first 3 layers to initiate the deposition. The scatter begins to disappear as the built height increases as seen in the Figure 8-21, hence uniform cooling begins as expected, which is the reason for WAAAM methodology). The same trend is observed in Figure 8-22. Average hardness of plot along the build direction in Figure 8-21 show a deviation of ± 8.5 HV deviation from the mean. And this trend is expected to reduce to ± 4.0 HV as the build height continues to increase. This is achieved by tweaking the volume of cold wire feed into the melt pool.



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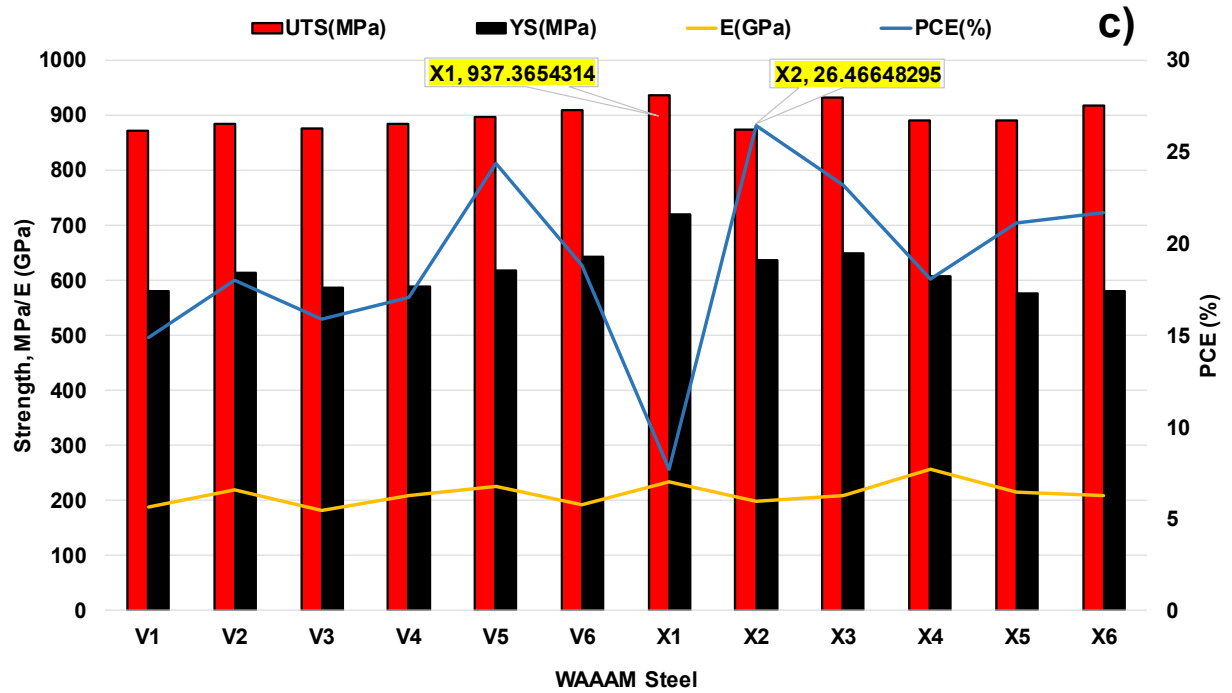


Figure 8-35 Mechanical Anisotropy of WAAAM (side feeding) – 2.4Cr-1Mo steel

(a) Stress-strain plot X-direction (b) Stress-strain plot Z-direction (c) all orientation plot

The fractured surface of the SF (Figure 8-39(b & c)) showed some hard phases which were likely sites of high-stress concentrations in the microstructure during fracture initiation and hence aid quasi-cleavage failure with resultant poor fracture toughness compared to FF. However, the shielding volume (20L/min) used with the SF was not sufficient, as some porosities were found on the microstructures in Figure 8-19(d-f). The fractured surface showed both ductile fracture with numerous small dimples and quasi-cleavage fracture at the sites of the pore areas.

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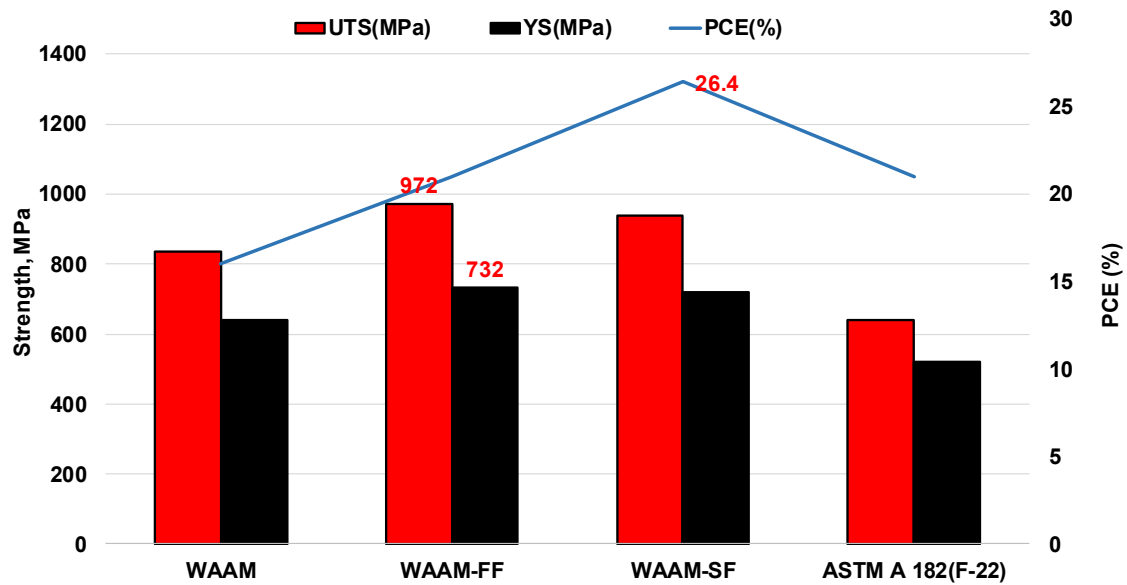


Figure 8-36 Average mechanical properties of WAAAM – 2.4Cr-1Mo steel

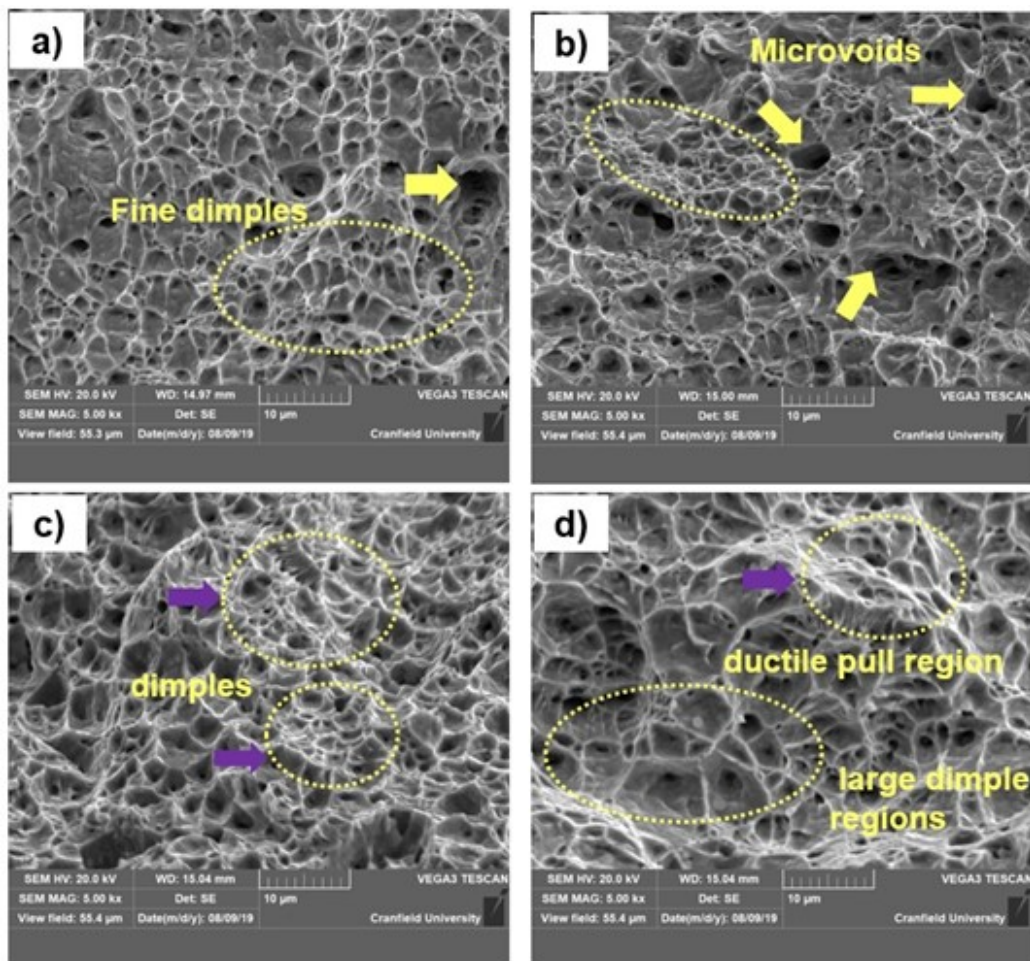


Figure 8-37 WAAAM tensile fracture morphologies

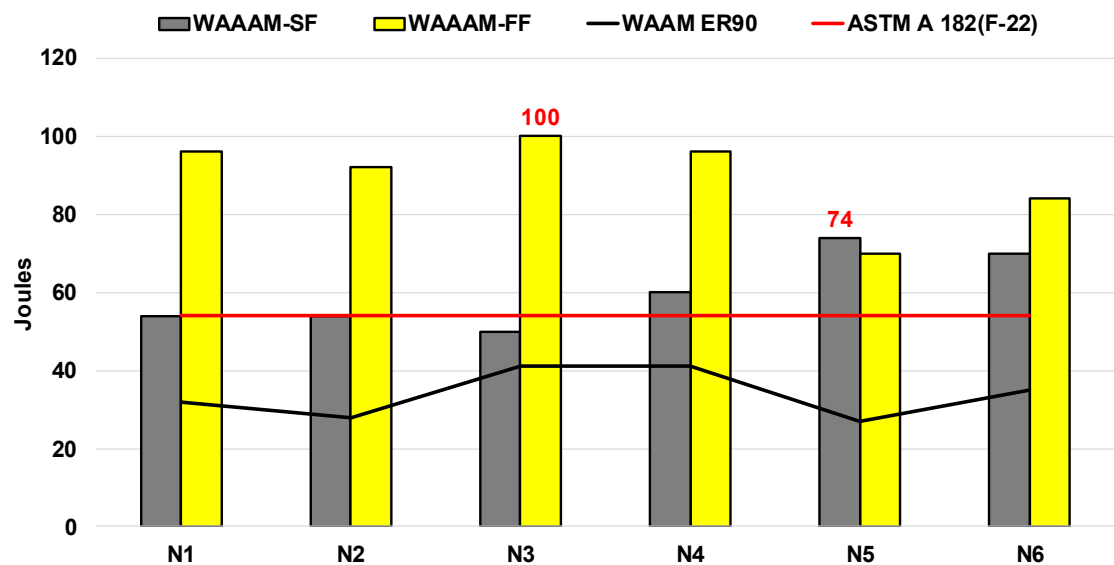
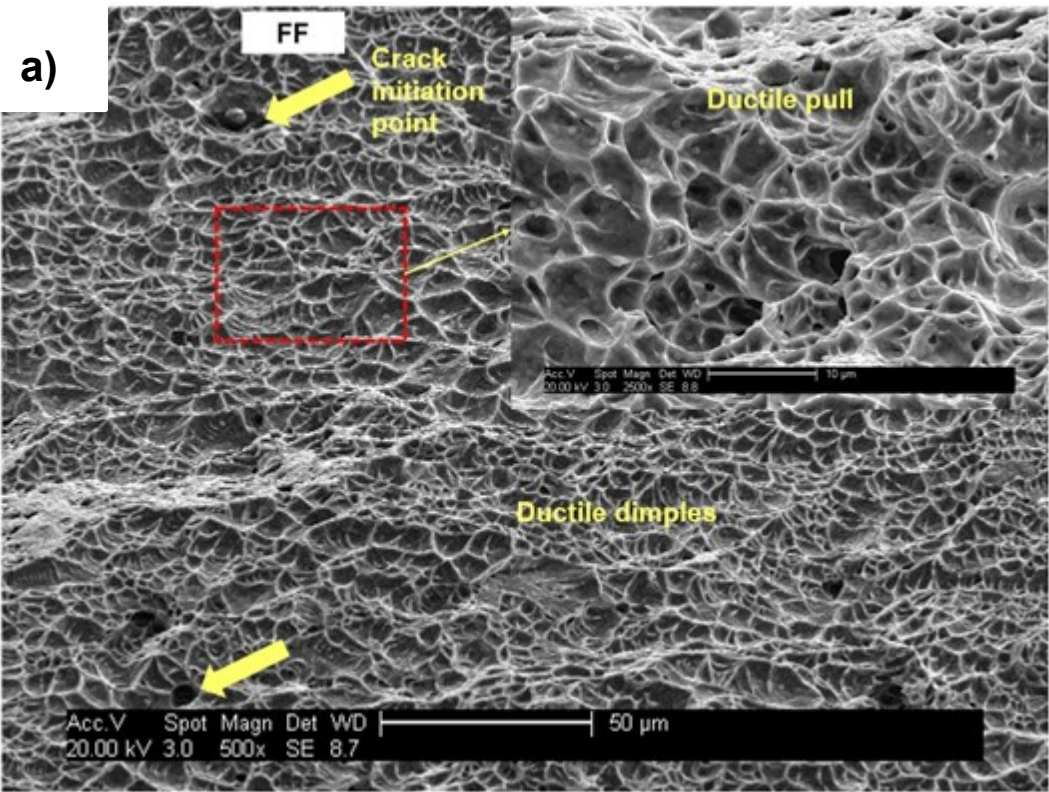
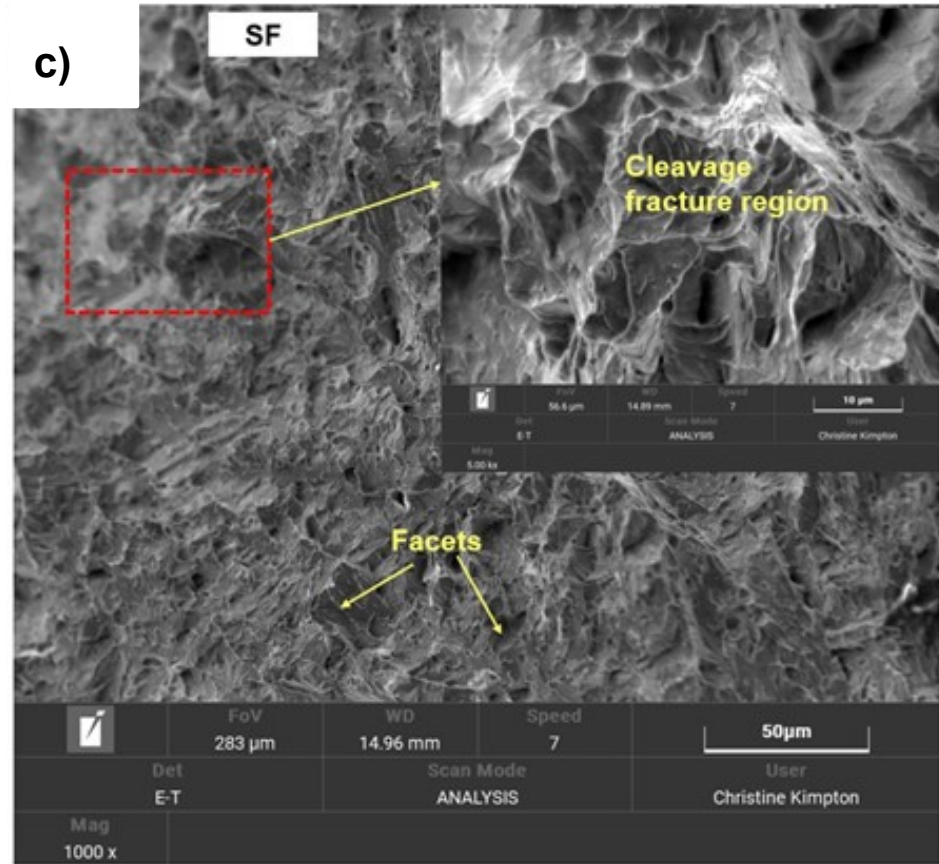
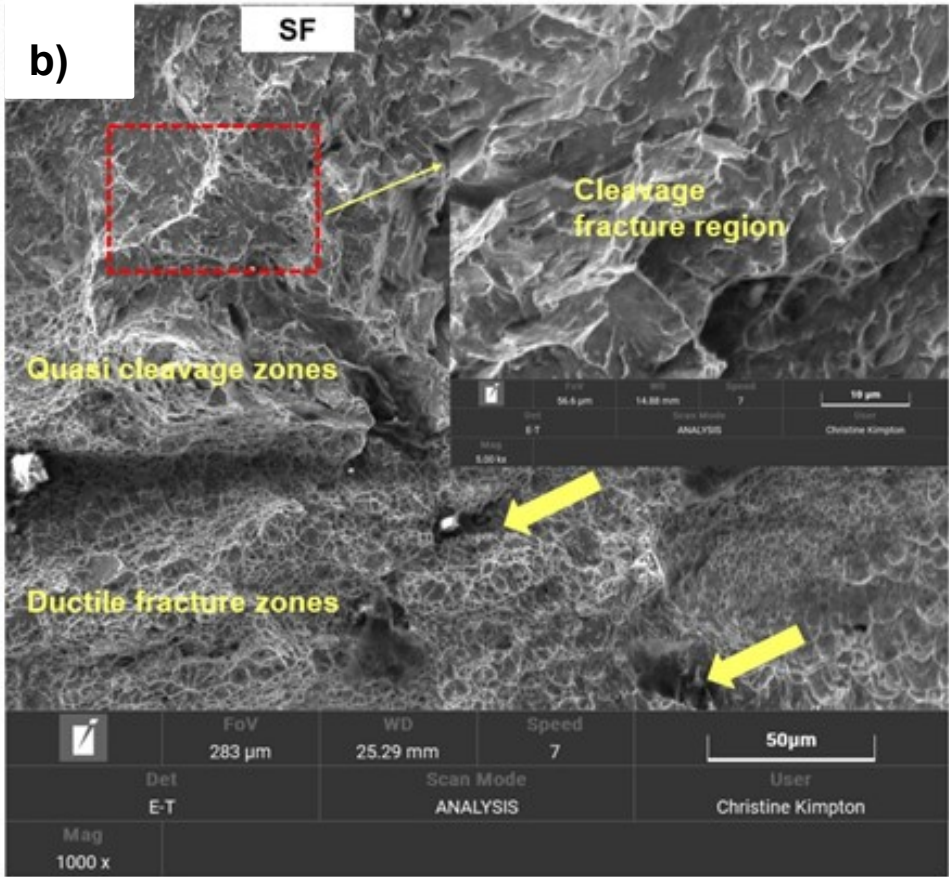


Figure 8-38 WAAAM vs WAAM impact toughness at -30°C





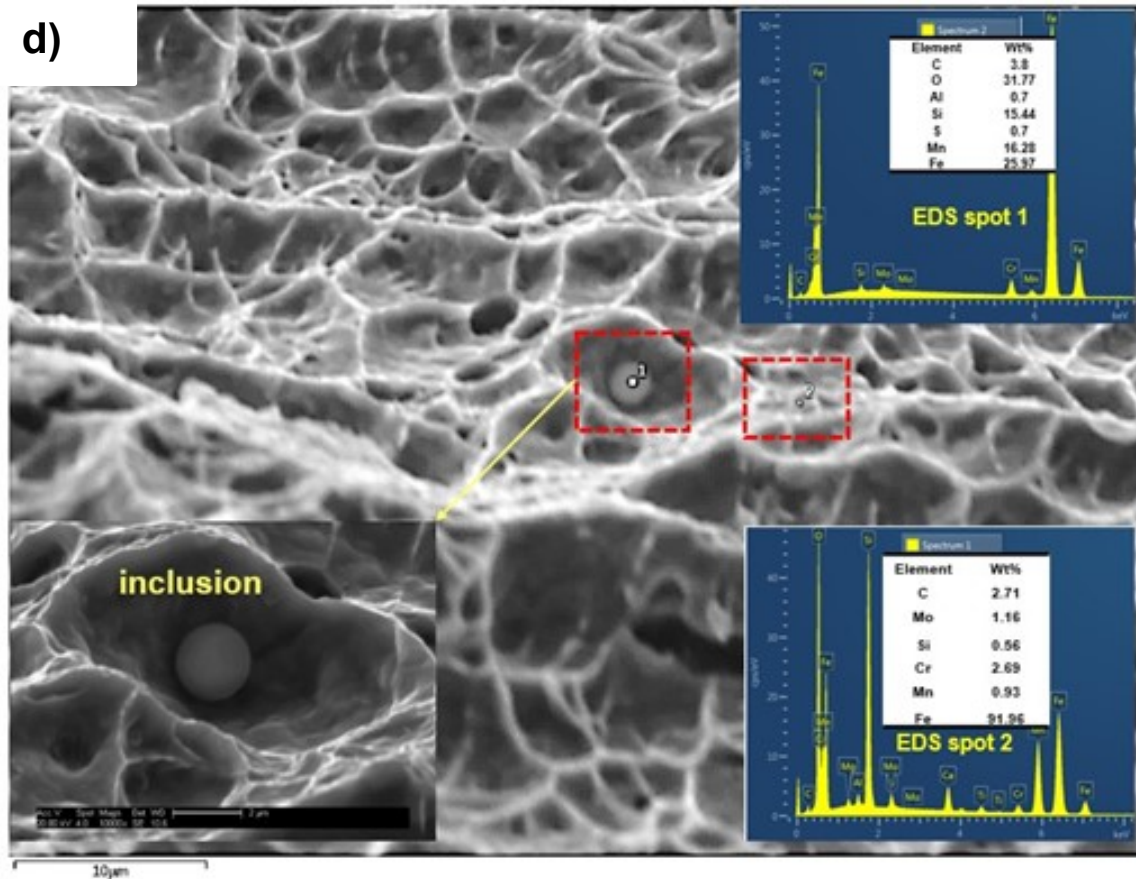


Figure 8-39 PGMAW + CW impact toughness fractured morphologies and EDS count

(a) front feeding (b) Side feeding-V (c) side feeding-V (d) EDS analysis of precipitates

8.4.6 Conclusion

In this chapter, a novel approach to contending with the disadvantages of build-up heat effects with high deposition using PGMAW power source for the building of WAAM structures by independently controlling the heat input effects by addition of cold wire using any of the deposition strategies was developed. Appropriate deposition parameters (PGMAW WFS; 8.0/6.0m/min, CW WFS; 4.0/4.2m/min & TS; 0.4m/min) were selected from the studies and used to build the demonstration part. The microstructure, mechanical properties, impact toughness properties, texture and grain morphology of the WAAAM structure were systematically investigated. It was shown that:

1. High productivity deposition of 6.27/5.23Kg/hr was achieved with the novel process without any defects in the structure.
2. The mechanical properties of the WAAAM structure using 2.4Cr-1Mo alloy (ER90S-B3) were increasingly heightened giving a maximum value of UTS of 972MPa, YS; 730MPa & elongation of 27% which is higher than properties obtained from PGMAW and CMT WAAM.
3. The fracture toughness surface showed good fracture behaviour at -30°C giving an average impact toughness of 96 joules as compared to WAAM of the same alloy with average impact toughness of 35 joules
4. The WAAAM microstructure is composed of elongated finer twinned low temperature bainitic and martensitic laths with less M-A constituent distributed in a ferrite matrix as a result of the fast cooling due to the lower effective heat input induced by the cold wire. Hence the improvement in the impact fracture toughness compared to the wrought value.
5. With the WAAAM process, the effect of accumulated and build up heat in oscillatory and other deposition strategies has been reduced to a minimum to deposited structures with original grain refinement and wrought specifications.
6. One of the advances with the CW addition is the reduction in the amount of base material dilution which also reduces the heat-affected zone (HAZ) area. High dilution of the base material and larger HAZ area has a direct link with increased residual stress, distortion, and shrinkage. The excess heat that causes this is mop-up by the CW addition.
7. Cost-saving and mechanical properties enhancement from using WAAAM by utilising one power source will outweigh the corresponding advantages of high productivity with tandem PGMA.
8. Appropriate WAAAM bead parameters for building structures were determined for different optimised rates.
9. Great process stability was achieved with the wire position, TRIM control, and CTWD.

9 SUMMARY, CONCLUSION, AND RECOMMENDATION

9.1 Introduction

This chapter presents a summary of the research findings from the five technical chapters. The summary is followed by concluding remarks that state the outline of this work in the context of knowledge in the field that has been expanded by this work.

9.2 Summary

A research has been made of the suitability of tailoring mild and HSLA steel wires for WAAM application. Although, the mechanical and microstructural variation of WAAM single grade and graded structures matches well with the wire manufacturers specification and they are deemed suitability for making bespoke components but not suitable for use in the as-deposited condition because of surface waviness. It was discovered that WAAM process parameters are responsible for the fatigue initiation behaviour of WAAM as-deposited structures and that reduction of surface waviness by process parameter optimisation and by cold side rolling improved the fatigue life of as-deposited WAAM components.

The fatigue crack growth rate and fracture toughness of the WAAM structural steels deposited using the oscillatory strategy showed some anisotropy variation in the properties, which was as a result of the microstructural variation encountered as a result of the heat accumulation effects of the strategy. This was overcome by the addition of cold wire into the WAAM melt pool to absorb the excess heat built up and control the cooling rate as desired. This novel strategy reduced the microstructural variability, consequently residual stress, elimination of base substrate dilution and was responsible for grain refinement and mechanical properties enhancement to wrought specifications.

9.3 Conclusion

1. Multi-alloy deposition using WAAM is a key tool for topological optimisation, as the properties were higher than the result obtained with the single grade WAAM. The disappearance of the Lüders band in mixed grade with mild steel, which is an undesirable problem for the industries is an indispensable strategy to optimise the functionality of applications. With a multi-alloy deposition strategy, the required strength of a segment of a component in a structure can be engineered using novel materials, thereby saving cost, time and wastage.
2. The mechanical properties in the single pass, single grade and mixed grades match well with the wrought counterparts. The Charpy impact toughness strength of the WAAM steel structure meets the specification except for 2.4Cr-1Mo steel which was below specification. Although with the advent of cold wire addition, the 2.4Cr-1Mo WAAAM deposit overmatched the specification. Their fracture toughness strength was within acceptable range but could be better if CW application was initiated in the built.
3. The fatigue response of WAAM as-deposited parts is controlled by the WAAM deposition parameters. Low layer height and WFS, with small wire diameter and moderate deposition rate leads to a low SW with reduce stress concentration factor. The loading direction of the WAAM as-deposited structures is paramount to the fatigue performance. Surface waviness is detrimental both to the static and fatigue performance of the as-deposited WAAM steel structure.
4. Side rolling is more effective in reducing SW but reduces the deposited width and increases the deposited height, whereas top rolling does the reverse. Reduction of the SW is achieved by the increase of notch radius of the valley between two SW in WAAM as a result of rolling. Compressive

Summary, Conclusion and Recommendation

stress induced by rolling on the as-deposited condition reduces the SW from 0.18mm to 0.08mm and produces an increase in hardness, tensile strength and fatigue strength due to an increased volume of heavily strained high-density dislocations ferrite lath.

5. The preferential grain growth direction during solidification in oscillated WAAM HSLA steel is the {100} & {101} orientation. A strong texture intensity in the re-melt zone is observed in the ER70S-6 & ER90S-B3, which indicated that an anisotropic martensitic lath microstructure was obtained. But in the single pass deposition strategy, no preferential orientation was observed in ER90S-B3 & ER120S-G.
6. The possible cause of variability in fracture toughness in the WAAM steel deposits is the hardening and softening regions in the layer bands caused by the variation in peak temperature and cooling rates of the adjacent layers in the building direction. The fracture toughness across the Z-direction possesses the most resistance to fracture failure as a result of the reduced grain sizes due to grain refinement and increases in the density of grain boundaries in the Z-direction. The fracture toughness along the welding direction possesses the least resistance to fracture failure as a result of increasing grain size in the welding direction and tendency for the crack to propagate in the most vulnerable part of the microstructure, the hardened zone in the layer band.
7. The layer band spacing and the thickness of the re-melt zone are key parameters that control the fracture toughness of WAAM steel in both directions. The dominant mechanism of failure in the WAAM steel is ductile fracture bound with large and small dimples with some fine dispersed carbides and oxides in the ferrite matrix.
8. The crack path in ferrite-pearlite steel and WAAM HSLA steel is complex (not linear). The angle of crack deviation, branching and nature of the

Summary, Conclusion and Recommendation

metal crumb formed strongly and fundamentally affected the FCGR in ferrite-pearlite and WAAM HSLA steel in the air. The microstructure of ferrite-pearlite and WAAM steels strongly influenced the FCGR throughout the da/dN vs. SIFR (including Paris region).

9. This novel approach of CW addition to wire WAAM, called WAAAM controls the effect of heat accumulation in oscillatory and other strategies to deposited structures with original grain refinement and wrought specification. This novel approach reduces residual stress and the elimination of base substrate dilution. The WAAAM microstructure is composed of elongated finer twinned low temperature bainitic and martensitic laths with less M-A constituent distributed in a ferrite matrix as. Hence the improvement in the impact fracture toughness compared to the wrought value.

9.4 Recommendations

To make high integrity steel structures using WAAM technology, the following actions are recommended;

1. Tailored the wire grades to where those properties are needed in the build, in terms of location and orientation.
2. Utilise side rolling to reduce surface waviness and improve the fatigue life for application under cyclic loading
3. Use an oscillating arc deposition pattern to improve strength and toughness
4. The addition of cold wire to the deposition melt pool to improve the thermal gradient by mopping away the excessive heat, hence better surface waviness and to improve the mechanical properties and deposition rate.

9.5 Future works

The following further work is recommended;

1. WAAAM strategy with side feeding for building components should be evaluated with different feed angles, TRIM, and CTWD value different from this study to ascertain process flexibility for with this approach.
2. Extended residual stress measurement on WAAM and WAAAM structures should be evaluated to see the great advantage of the WAAAM process.
3. Different routes of post-weld heat treatment should be carried out on the HSLA steel WAAM depositions made by oscillatory and single-pass deposition to evaluate any difference in fracture mechanics and fatigue crack growth rate properties.
4. Fatigue life determination for the Z-direction samples of the WAAM mild and HSLA steel should be evaluated.
5. Extended study on the effect of robotic hammer peening in reducing WAAM as-deposited surface waviness should be evaluated as an option to complement rolling at WAAM intersections.
6. Since strain hardening exponent are incorporated in both the elastic and plastic component of CTOD for ASTM E1820, and ASTM E1820 underestimates the ISO 12135. A re-evaluation of the CTOD value should be done with ISO 12135

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APPENDICES

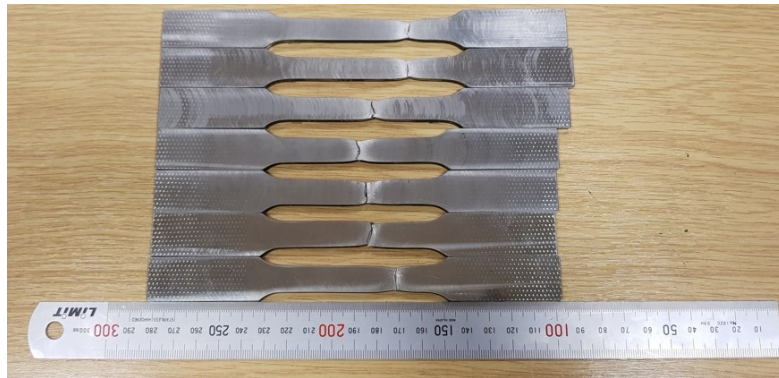


Figure 1 – Tensile fractured specimens

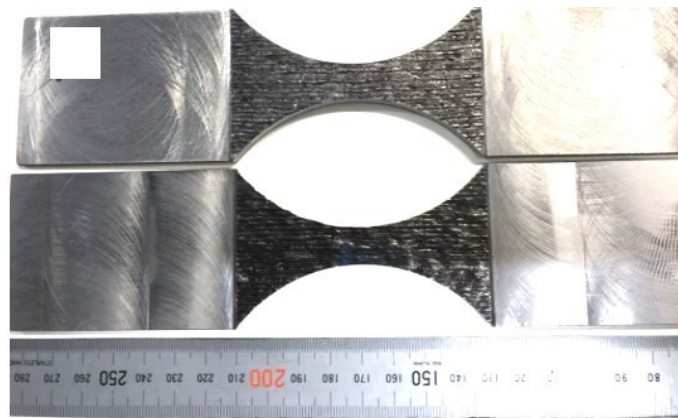


Figure 2 – Fatigue samples- as-deposited condition

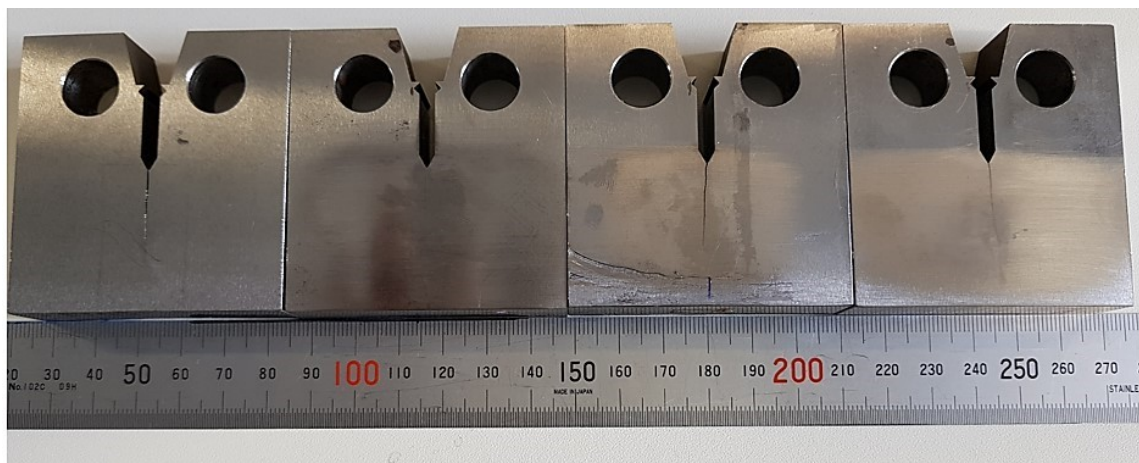


Figure 3 – CT specimens for fatigue crack growth -FCGR

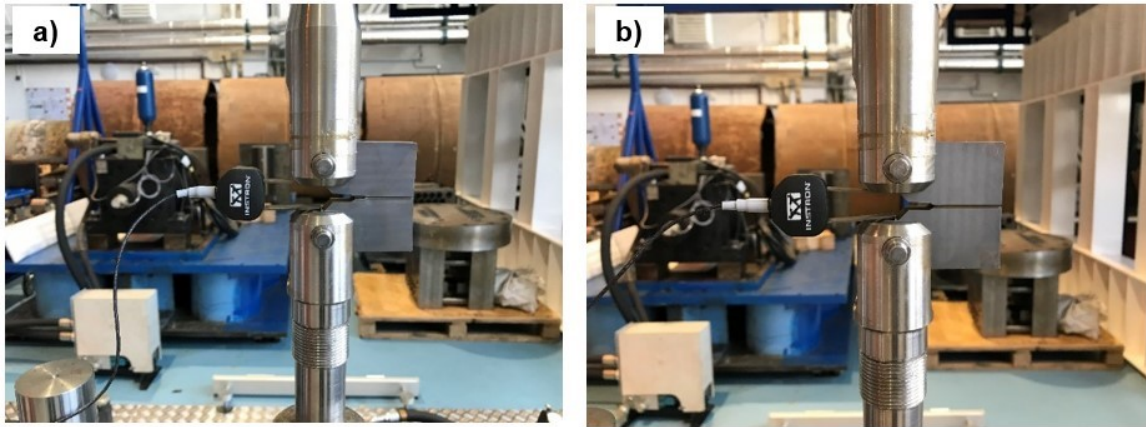


Figure 4 – (a) J1c testing along the layers (b) J1c testing across the layers

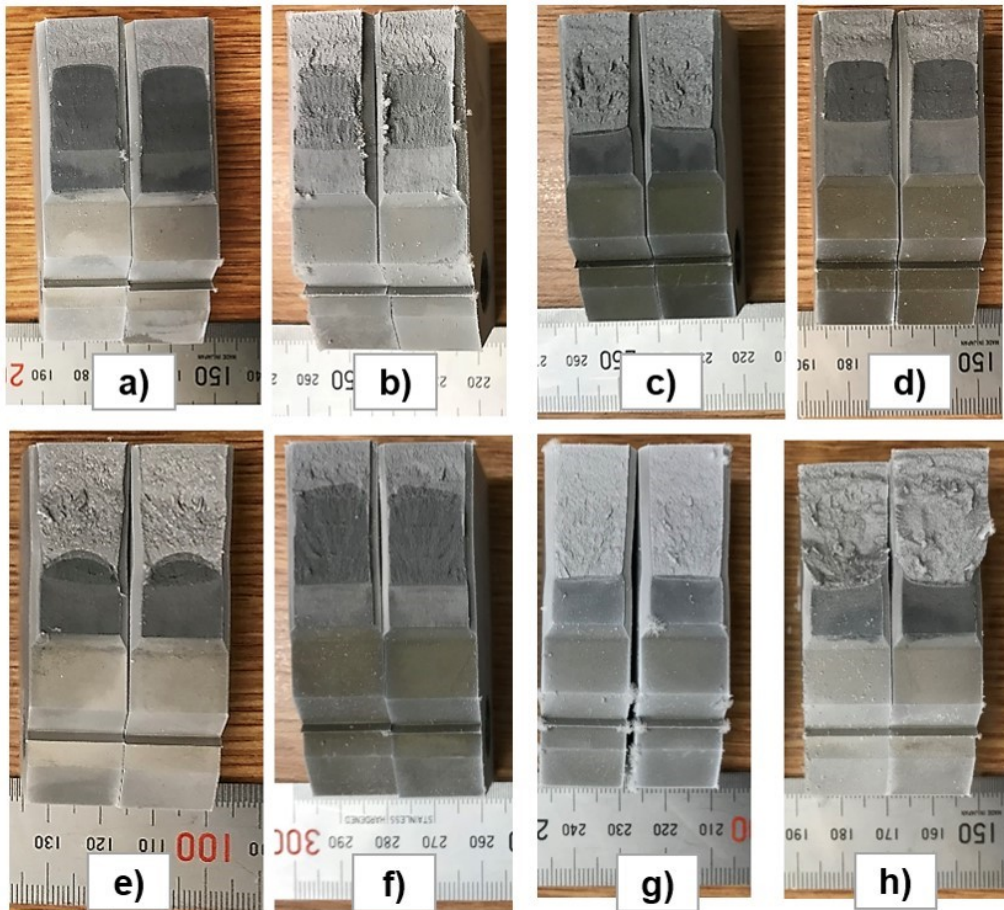


Figure 5 – (a-h) Fractured CT specimen from J1c testing showing notch faces

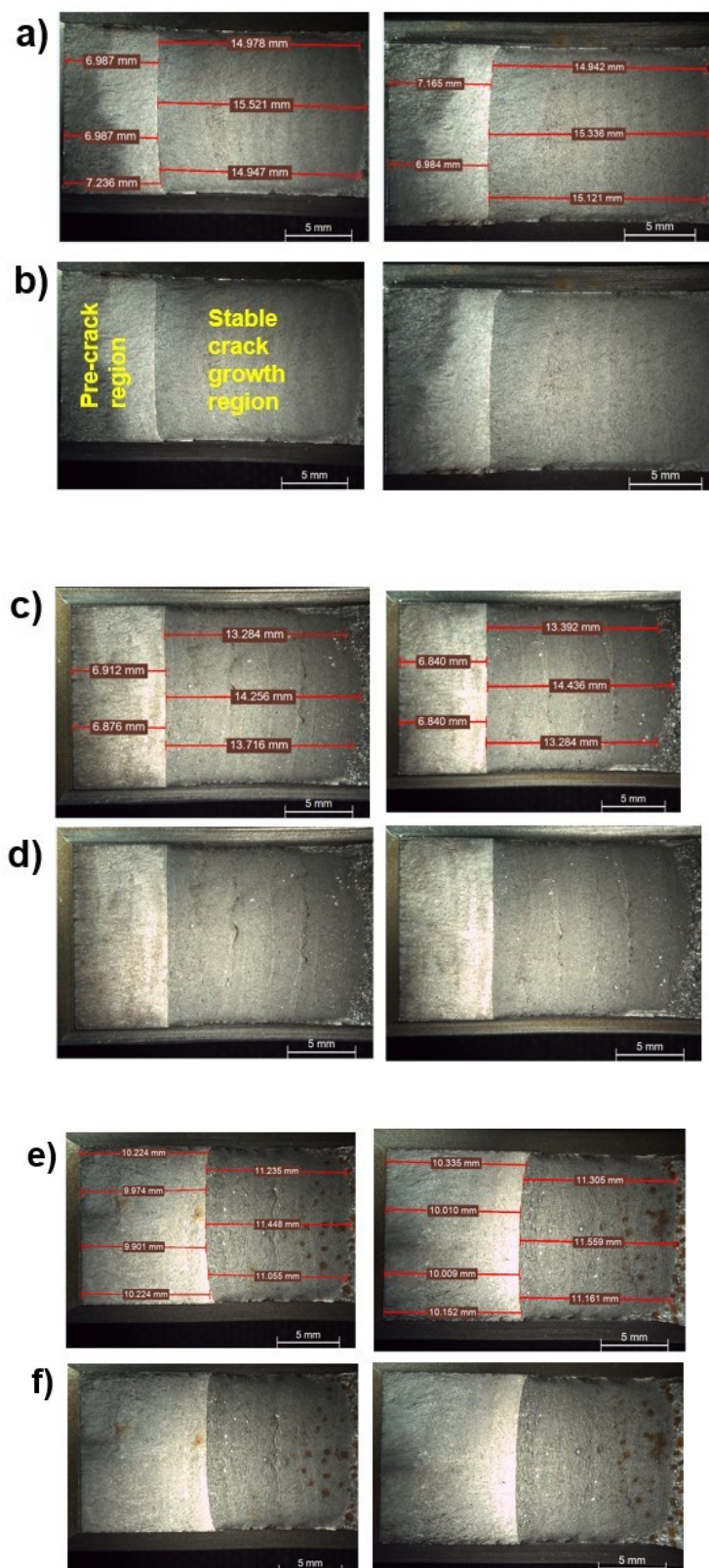


Figure 6 – (a-f) Fractured surfaces from toughness testing showing pre-crack and stable crack growth region



Figure 7 – WAAM built wall

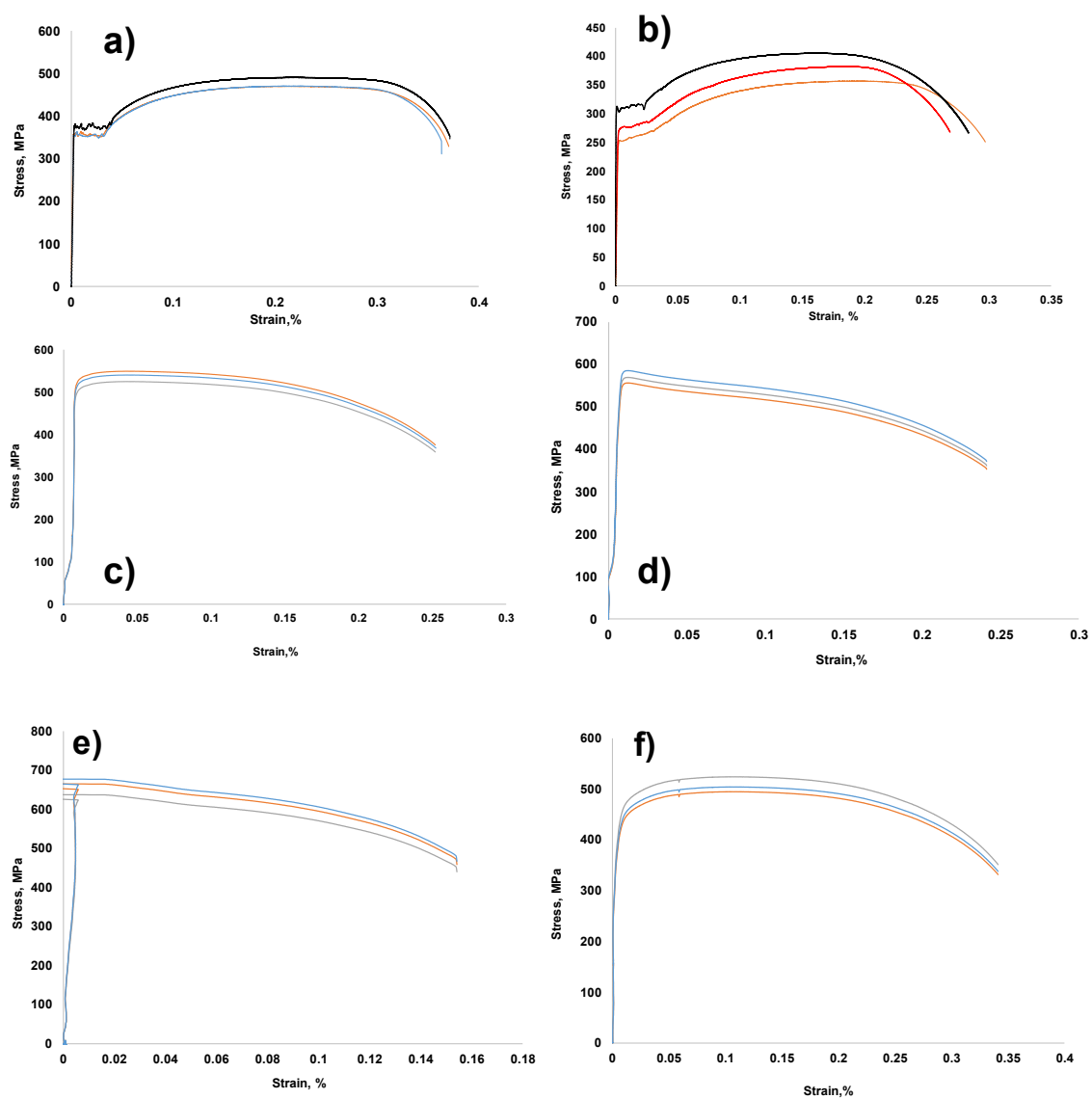


Figure 8 – Stress- strain curve

voestalpine Böhler Welding UK Ltd.

704108

voestalpine Bohler Welding UK Ltd.

Registered Office:
European Business Park
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OLDBURY
West Midlands B69 2BN
Telephone:
0121 569 7700 (Switchboard)
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Web Site: www.voestalpine.com/welding/uk
E-mail: sales@voestalpine.com

P W P Industrial

Inspection certificate 3.1

Jaguar House
MK41 0EP Bedford
United Kingdom

as per : EN 10204
No. : 2017-2012181350-900002-014
Rev. 0 Page 1 of 1

PO no.	PO28850	of	07.04.2017
Order no.	1012158097		
Delivery note/pos./splitt	2012181350/000020/900002	of	07.04.2017
Product	GMAW wire electrode		204751
Trade name	BÖHLER CM 2-IG		23974
Standard designation	EN ISO 21952-A - G CrMo2Si		1073A00G
	AWS A5.28: ER90S-G [ER90S-B3(mod.)]		0120
Dimension	1,2 mm		
Heat no.	102758		
Quantity	36,0 KG		

Chemical composition in % of the product

C	Si	Mn	P	S	Cr	Mo	Ni	V	Cu	Nb				
0,08	0,58	0,89	0,005	0,002	2,4	0,99	< 0,1	< 0,01	< 0,1	< 0,01				

Mechanical properties

EN 10204 - 2.2

Tensile test							
T	ReL / Rp 0,2 MPa	Rp 1,0 MPa	Rm MPa	A (Lo = 5d) %	Z %	WBH PWHT	Remarks
20°C	≥ 400		≥ 500	≥ 22		720°C/2h	
Impact test							
T	Impact energy KV / J	Average KV / J	Lateral expansion mm	Shear fracture %	WBH PWHT	Remarks	
20°C	≥ 47				720°C/2h		

Town
Oldbury

Date
07.04.2017

This certificate was issued by DP-equipment and does not require signature.

Authorized representative
Stuart Thomas

voestalpine
ONE STEP AHEAD.

Figure 9 – Wire certificates- ER90S-B3

voestalpine Böhler Welding UK Ltd.

70410B

voestalpine Bohler Welding UK Ltd.

Registered Office:
European Business Park
Taylors Lane
OLDBURY
West Midlands B69 2BN
Telephone:
0121 569 7700 (Switchboard)
0121 569 7710 (Sales)
0121 569 7707 (Accounts)
Web Site: www.voestalpine.com/welding/uk
E-mail: sales@voestalpine.com

P W P Industrial

Inspection certificate 3.1

Jaguar House
MK41 0EP Bedford
United Kingdom

as per : EN 10204
No. : 2017-2012181350-900001-014
Rev. 0 Page 1 of 1

PO no.	PO28850	of	07.04.2017
Order no.	1012158097		
Delivery note/pos./splitt	2012181350/000010/900001	of	07.04.2017
Product	GMAW wire electrode		204751
Trade name	UNION X 90		17684
Standard designation	EN ISO 16834-A - G 89 6 M21 Mn4Ni2CrMo		1H49C003
	AWS A5.28: ER120S-G		0120
Dimension	1,2 mm		
Heat no.	102232		
Quantity	54,0 KG		

Chemical composition in % of the product

C	Si	Mn	P	S	Cr	Mo	Ni	V	Cu	Ti	Al	Zr		
0,10	0,79	1,77	0,008	0,010	0,35	0,63	2,27	< 0,01	0,07	0,07	< 0,01	< 0,01		

Mechanical properties

EN 10204 - 2.2

Tensile test							
T	ReL / Rp 0,2 MPa	Rp 1,0 MPa	Rm MPa	A (Lo = 5d) %	Z %	WBH PWHT	Remarks
20°C	≥ 890		950 - 1180	≥ 15			M21
Impact test							
T	Impact energy KV / J	Average KV / J	Lateral expansion mm	Shear fracture %	WBH PWHT	Remarks	
-60°C	≥ 47					M21	
20°C	≥ 90						

Town
Oldbury

Date
07.04.2017

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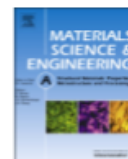
Figure 10 – Wire certificates-ER120S-G



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Analysis of fracture toughness properties of wire + arc additive manufactured high strength low alloy structural steel components

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ARTICLE INFO

Keywords:

High strength low alloy steel (HSLA)
CMT wire + arc additive manufacturing (CMT-WAAM)
 J_{IC} fracture toughness
Layer band spacing
Microstructural variation

ABSTRACT

The uncertainty surrounding the fracture behaviour of CMT-WAAM deposited steel, in terms of crack tip condition (J and CTOD) needed to cause crack tip extension, has made this manufacturing technique unpopular to date. Fracture toughness parameters are crucial in the structural integrity assessment of components and structures in various industries for assessing the suitability of a manufacturing process and material. In the offshore wind industry, the EN-GJS-400-18-LT ductile cast grade for the mainframe and hub has lower fracture toughness resistance for its high strength grade. Its high weight level affects the Eigen frequency of the tower and imposes high installation cost incurred from heavy lifting equipment usage. Poor fracture toughness is currently a challenge for wind turbine manufacturers in the quest for a cleaner and cheaper energy in the form of offshore wind. In this study, CMT-WAAM is used in depositing steel components with an oscillatory and single pass deposition strategy. The effects of microstructural variation, as a result of layer by layer deposition and the layer band spacing, on the fracture resistance in the build and welding direction was shown here. The fracture mechanics and failure mode of the WAAM deposited parts were investigated. The microstructural variation, again as a result of the layer by layer deposition and the layer band spacing, are the key parameters that control the fracture toughness of WAAM steel. Anisotropic behaviour in the J_{IC} values was observed between both fracture orientations. The constructive transformation mechanism of the WAAM oscillatory process made way for intragranular nucleation of acicular ferrite on the Ti containing inclusion, thereby improving the toughness of the ER70S-6 deposit with a unique microstructure and J_{IC} value of 640 kJ/m^2 .

1. Introduction

The pursuit for cleaner energy generation and the vast availability of offshore wind energy has called for the need to improve the structural materials currently being used and the manufacturing process [1]. Wind power is currently the world's fastest growing source of energy. Owing to the current trend for the quest for renewable wind energy by the UK government, a target was set in Vision 2020 [2]. A typical wind turbine is comprised of 10–25 tons of ductile grade EN-GJS-400-18-LT. This ductile cast grade is used for casting the rotor, hub and mainframe before machining [3,4]. The key issue with the casting process is the endangering of the environment by its high carbon footprint coupled with its high cost of manufacturing involvement. The second key issues are associated with the material's response; high strength ductile cast grades respond poorly to fracture resistance, in addition to their high weight level, which affects the Eigen frequency of a turbine. This is likely to limit its application in larger wind turbines as more

exploration of offshore wind is imminent. In order to build larger and lighter wind turbines to be used in deep waters, lighter materials with enhanced fracture toughness and strength properties will be needed.

The use of modern high strength low alloy (HSLA) steel for the wind industry will allow for the reduction in weight and cost of some specially made components [5]. The cost saving from installation and reduced inspection times as a result of damage tolerance material (HSLA steels) usage is enormous. HSLA steels are micro-alloyed with Ti, V, Nb and Al, in addition to the low carbon content, increase strength and toughness are obtained through unique steel strengthening mechanisms, e.g. grain size refinement [6]. Other elements such as Mo and Cr are useful in promoting bainite formation while Mn and Ni increase the hardenability of the weld metal by helping in the formation of more acicular ferrite, and Ni on its own helps in solid solution strengthening. The volume fraction of any of this microstructure, i.e. acicular ferrite, granular bainite, polygonal ferrite, pearlite, and martensite, form in the microstructure of HSLA steel depends on the proportion of alloying

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Wire plus arc additive manufactured functional steel surfaces enhanced by rolling

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ARTICLE INFO

Keywords:
Functional surfaces
Fatigue life
Tensile strength
WAAM
Rolling

ABSTRACT

Surface waviness (SW) is one of the major problems confronting the economical use of as-deposited components made with the wire plus arc additive manufactured (WAAM) process. The SW acts as a stress raiser, thereby reducing the tensile properties and fatigue life of the component. In this study, the effect of compressive residual stress on the mechanical and fatigue behaviour of the as-deposited WAAM mild steel component was carried out using a process which combines deposition and rolling on the WAAM component surface. The fractured faces and microstructure were characterised by a scanning electron and optical microscope. The microstructural changes were characterised by X-ray diffraction techniques. The results revealed that an increase in the notch radius and compressive stress induced by rolling on the as-deposited condition reduced the SW from 0.18 to 0.08 mm with a reduction of stress concentration, consequently reducing crack initiation and propagation and improving fatigue life. This work shows that rolling has a dual effect on as-deposited WAAM mild steel components with a new functional surface.

1. Introduction

Currently, offshore and onshore wind energy generation constitute about 8.01% and 9.1% of the entire energy generation in the UK, respectively. This has led to the reduction of CO₂ emissions in the UK by 12 million tonnes [1]. The Vision 2020 projection of the UK Department of Energy and Climate Change is to reduce further CO₂ emissions by 10% by the year 2020 [2]. With this mandate, the wind turbine original equipment manufacturers (OEMs) must look for solutions and cutting-edge technologies to remain at a competitive advantage over other forms of energy producers. The current subtractive manufacturing (SM) process used in the mainframe and hub has some drawbacks such as material wastage, increased cost and higher production lead time of materials manufactured through a sand casting process before machining.

Additive manufacturing (AM) is a method of building parts and components, layer by layer using different energy sources. The development of AM processes has been on the increase because of their numerous advantages compared to SM. The use of AM can drastically reduce material wastage and production lead times while allowing the building of complex shapes with less expensive tooling requirements and hence cost savings [3–6]. AM methods have been developed in recent times to optimise geometrical features, deposition rates,

dimensional accuracy, layer thickness and surface finish [7]. Despite all these improvements, only a rough near net shaped surface (final outlook finish) can be produced [8].

One of the newest low-cost AM processes is wire plus arc additive manufacture (WAAM), which involves building a structure with a consumable wire in a layer by layer manner in an open or atmospherically controlled chamber. The WAAM process has a higher deposition rate than other AM processes, close to 10 kg/h for steels [9,10], with better material utilisation rate than others but has a limited ability to produce finer surface finish features [11]. Rougher surface finish in WAAM is a result of overheating and heat accumulation effects which reduce the surface tension and make the molten pool unstable. The WAAM process parameters such as travel speed (TS), wire feed speed (WFS), interpass temperatures and the ratio of WFS to TS are responsible for the heat accumulation effects [12–14]. Besides the SW challenges, WAAM has been known to be affected by unevenly distributed stresses generated during deposition, which might require further post-heat treatment or cold work to relieve these stresses and improve their mechanical properties [15]. Path planning sequencing and deposition strategies compensation with built parameter optimisation could be effective in reducing the residual stresses [16,17]. A 500 µm surface waviness (SW) encountered in WAAM is a serious problem, limiting the use of WAAM as-deposited structures [5]. The as-

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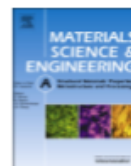
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Microstructure and mechanical properties of TOP-TIG-wire and arc additive manufactured super duplex stainless steel (ER2594)

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ARTICLE INFO

Keywords:

TOP-TIG
Wire and arc additive manufacturing
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CrN
Cr2N
Anisotropy
Nitrogen work hardening

ABSTRACT

As the excellent combination of mechanical properties and corrosion resistance for super duplex stainless steel, a prospective method – Wire and Arc Additive Manufacturing – for fabricating this material was proposed, and a wall component was deposited in this study. The microstructure of the as-deposited wall was carefully analyzed along with the variation of mechanical properties. The results revealed that, in the wall-body, the austenite/ferrite phase balance was broken by the overgrowing the austenite phase. During this process, the intergranular secondary austenite leading the increase of austenite phase together with some contributions made by the precipitation of intragranular secondary austenite. Propagation of the intermetallic phases, chi and sigma phase, was not the major reason for the low impact toughness in the last layer area and the root region. Instead, the presence of CrN and “inclusions” (Cr2N and impurities) took the main responsibility not only in the impact toughness but also the ductility. The anisotropic analysis revealed that the UTS and elongation appeared distinct difference in vertical and horizontal direction samples. The varieties in YS were eliminated by the nitrogen work hardening effect to a large extent.

1. Introduction

Super duplex stainless steels (SDSSs) are widely used in the marine, petrochemical, oil and gas industries owing to their excellent properties in strength and corrosion resistance [1]. The proper balance ratio of austenite to ferrite phase (50/50) is the guarantee of these remarkable properties. However, the higher alloy content and accurate thermo-mechanical processing of SDSS-wrought products have led to higher costs than normal austenitic stainless steels. Thus, it is essential to explore proper manufacturing processes for fabrication of SDSS structural parts.

The wire + arc additive manufacturing (WAAM) technology, a rapid fabrication technology based on layer-by-layer construction [2], is outstanding with high printing efficiency, full density, premium mechanical properties, low cost and relatively high productivity [3]. So far, the WAAM processes predominantly use gas metal arc welding

(GMAW), tungsten inner gas welding (TIG/GTAW), cold metal transfer welding (CMT) and plasma arc welding (PAW) for melting the feeding material [4]. Among these, TIG-WAAM provides high-level process control and supplies very clean deposition [5]. A certain amount of high-quality metal components with different materials were produced [6–9]. This promising method reveals high expectations in the area of manufacturing super duplex stainless steel components. Nevertheless, there are few pieces of literature focusing on the integration of such alloy and process. Hence, in this study, some fundamental concerns were carried out.

There are two types of the additive manufacturing method for DSS in recent years, including selective laser melting (SLM) [10–12] and CMT-WAAM [13,14]. The microstructure built by SLM contained nearly 100% of the ferrite phase. However, 30% of ferrite phase was found by Posch [13], and between 15 and 27% was reported by Eriksson [14] with the heat input ranging from 0.4 to 0.87 kJ/mm after using the CMT-

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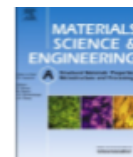
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Critical assessment of the fatigue crack growth rate sensitivity to material microstructure in ferrite-pearlite steels in air and marine environment



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ARTICLE INFO

Keywords:

Grain refinement
Microstructure
Corrosion-fatigue
Seawater
BS7910

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 fatigue crack growth (FCG) in air and seawater. The data from the tests were then compared with results of several studies on α -P steels of similar compositions and microconstituents in the literature. 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 belief that microstructure has little or a 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 a 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 a strong retarding effect on FCG in α -P steels at low stress intensity factor range (SIFR). This finding has a fundamental implication for analytical and numerical models that wholly depend on the mechanical properties of the steel and assumes plane horizontal crack propagation front in predicting FCGR in steels. Such models are likely to fail in predicting FCGR even in a subgrade of the same grade of steel.

1. Introduction

There are many studies that have claimed that microstructure has little or a negligible effect on the intermediate fatigue crack growth rate (FCGR) in metallic materials provided that the material is not in a corrosive medium [1–7]. It appears no study has proved beyond doubt 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. The regions of microstructural and other factors' influence are commonly presented as in Fig. 1.

Mechanical properties have also been found to strongly influence

the threshold and region of accelerated failure. Stonesifer [11] 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 Fig. 1, microstructure, environment and mean stress are reported to have a large influence on the near threshold behaviour of metallic materials [5,6,8,11–13]. For the final failure region, microstructure, mean stress and thickness have a large influence while the environmental influence is small.

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 [14]. 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 [15,16]. This statement is not strictly correct as steel materials, no matter how well they are produced

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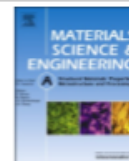
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Improving mechanical properties of wire plus arc additively manufactured maraging steel through plastic deformation enhanced aging response

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ARTICLE INFO

Keywords:

Wire plus arc additive manufacture
Plastic deformation
Maraging steel
Dislocation density
Age hardening

ABSTRACT

Maraging steel gains ultrahigh strength through aging; however, wire plus arc additively manufactured maraging steel features a columnar-dendritic structure with associated segregation and shows a much less pronounced aging response. In this paper, plastic deformation was introduced through interpass cold rolling during the layer-by-layer deposition process. After aging, the ultimate tensile strength (UTS) was improved substantially from 1410 MPa (unrolled) to 1750 MPa (50 kN rolled). Rolling induced partial recrystallization to break the dendritic structure and form high-angle grain boundaries, which promoted the atoms diffusion to enable a more uniform solutionizing process and improved the subsequent aging response by 105–110%. The main contribution of overall strengthening of the rolled alloy was attributed to the effective aging process, accounting for more than 95% of the entire strength increase.

1. Introduction

Maraging steel is a family of low-carbon (below 0.03%) steel that combines ultrahigh strength with excellent fracture toughness [1]. Maraging steel derives the high strength by the dispersion of large amount of nanosized precipitates from the age hardening of an extremely ductile body-centred cubic (bcc) lath martensitic structure. The unique hardening mechanism results in strengths up to 3450 MPa [2], as compared to other steels with similar crystallographic structures. The high strength-to-weight ratio, ease of machining in the annealed condition, superior dimension stability during age hardening, and good weldability make maraging steel widely used in aerospace and tooling industry.

Additive Manufacture (AM) is a novel approach for building near-net-shape components in a layer-by-layer manner [3]. In Directed Energy Deposition (DED) AM, the feedstock (powder or wire) is melted by a heat source (laser, electron beam or arc), and the deposition path is controlled by a robot or a CNC system to form the desired geometry. Compared to conventional subtractive manufacture (SM), AM is attractive for the fabrication of metallic components due to substantially reduced material wastage of expensive metals (such as Ti-6Al-4V, Inconel superalloys and maraging steels), shortened lead times and greater design freedom [4].

Despite these advantages, the implementation of AM in engineering

applications requires the AM parts to possess equivalent or even superior mechanical properties to the SM counterparts. Currently, the benefits above are not realised due to the limitation in properties, and AM is mostly exploited in powder bed processes where the design capability is utilised to provide weight savings. The conventional metal billet or forging production procedure involves melting, casting and subsequent thermomechanical processing, through which the entire billet or forging undergoes a uniform processing and ends up with a relatively homogeneous microstructure and largely isotropic characteristics [5]. Due to the inherent characteristics of DED AM, the deposited material undergoes a unique processing featuring localised melting, rapid solidification and non-uniform heating-cooling cycles. These conditions often result in a spatially differential microstructure of dendritic nature associated with segregation and largely anisotropic properties in comparison to that attained in a conventionally processed material [6]. For example, in AM of Ti-6Al-4V alloy, undesirable, coarse-columnar, primary β grain structures have been widely reported to result in a strong texture and mechanical anisotropy in an AM component in contrast to the fine equiaxed grains, nearly random grain orientation and isotropic mechanical properties in a wrought alloy [7,8].

Most of the AM research on maraging steels focussed on the laser/powder based process [1,9–14]; however, the solidification could result in non-steady amorphous-nanocrystalline composite

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1 Article

2 Effects of wire plus arc additive manufacture 3 deposition parameters on surface waviness and 4 fatigue behaviour of as-deposited steel

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13 **Abstract:** Surface waviness (SW) is one of the major issues in wire plus arc additive manufactured
14 (WAAM) components, as it would restrict the application of WAAM components in the as-built
15 state due to geometric stress concentration factors from such waviness. In this study, the effects of
16 WAAM deposition parameters on SW and the corresponding effects of stress concentration, caused
17 by SW, on the fatigue life of as-deposited WAAM steel structure was evaluated. And the effect of
18 elastic stress concentration factor (S_{el}) and effective stress concentration factor (S_{eff}) on the fatigue
19 properties was determined. The geometric factors from SW were used to evaluate the S_{el} while the
20 fatigue failure curves were used to determine S_{eff} . The results show that SW is detrimental to the
21 dynamic mechanical properties of as-deposited components. The larger notch radius achieved with
22 lower layer height is effective in reducing both the S_{el} and S_{eff} which, in turn, enhances the
23 fatigue life of the as-deposited condition. This study shows that WAAM steel components could be
24 used in the as-deposited condition in engineering applications if deposition process parameters are
25 such that the SW is reduced.

26 **Keywords:** Wire + arc additive manufacture; stress concentration factor; fatigue life; surface
27 waviness; as-deposited condition; process parameters

29 1. Introduction |

30 Additive manufacturing (AM) of solid geometrical objects is achieved through layer by layer
31 deposition using a heat source, with manipulation done by a robot or numerical control system using
32 either powder or wire as feed material [1]. Significant material and cost savings can be achieved by
33 AM when compared to computer numerical control (CNC) milling or machining from forgings and
34 castings, where up to 95% of materials is removed from the billets [2]. AM is categorized into powder
35 bed, powder feed, and wire feed systems. Powder bed systems are limited in build volume and
36 deposition rate (DR), while powder feed systems have build volume greater than $1.2m^3$ [3]. Powder-
37 feed/-bed systems are capable of producing parts with high geometrical accuracy with a layer
38 thickness of 20 -100 μm and dimensional accuracy of $\pm 0.05 mm$ [4]. Inconel 625 components built
39 with the selective laser melting (powder bed system) achieve a top and side surface roughness of
40 0.009 - 0.010 mm [5] and the direct laser sintering of Cu-based metal powder (powder feed system)
41 used to produce the insert of a mobile phone cover had a surface roughness of 0.014 -0.016 mm
42 [6]. The limited DR (0.6g/min) of the direct metal deposition process of the powder feed system has