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**An Investigation into the Fatigue and Corrosion Fatigue Properties of Two High
Strength Low Alloy Steels and their HAZ's.**

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Abstract

Testing was performed on three point bend specimens machined from parent plate and welded samples machined such that the crack propagated down the heat affected zone. Specimens were subjected to tension-tension loading using a stress ratio of 0.6 and a testing frequency of 0.5 Hz. Tests were performed in air and in a synthetic sea water solution. Three levels of corrosion protection were investigated, freely corroding, protected at -800 mV and over protected at -1100 mV. The resulting crack growth rate data were plotted on a log scale against the log of the stress intensity range. This made comparisons between different conditions and materials easier. Both high strength steels and the heat affected zones associated with their weldments compared favourably with BS4360:50D structural steel when tested in air and with a cathodic potential of -800 mV. Using over protection conditions the materials showed a susceptibility to hydrogen concentration.

A particularly sensitive technique was used to investigate the effects of heat affected zone microstructures on the crack propagation rate. This was achieved using a continuous monitoring gauge mounted onto the side of the specimen. The gauge was used to obtain the crack length and a computer was coupled to the fatigue machine to control the applied loads. The resulting data showed that there was a variation in growth rates between different weldment regions which was in turn influenced by the weld heat input used. The other finding was that when low heat inputs were used for welding there was rapid increase in crack growth rate associated with the coarse grained region, which was attributed to the aligned carbides of this region which act to stop crack blunting. This effect was only seen in the 1.5 kJ/mm weld, higher heat inputs cause a break up of these carbides and hence the crack is able to blunt.

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Nomenclature

S	Stress Range
N	Number of Cycles to Failure
K	Stress Intensity Factor
$K_{\max}$	Maximum Stress Intensity
$K_{\min}$	Minimum Stress Intensity
ΔK	Stress Intensity Range
ΔK_{th}	Stress Intensity Threshold
K_{op}	Opening Stress Intensity
ΔK_{eff}	Effective Stress Intensity Range
K_{c}	Fracture Toughness
R	Stress Ratio
$\frac{da}{dN}$	Crack Growth Per Cycle
a	Crack Length
r	Polar Coordinate
Θ	Polar Coordinate
r_y	Plastic Zone Size
E_{corr}	Corrosion Potential
i_{corr}	Corrosion Current

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1. Introduction

The discovery of oil and natural gas off the coast of the United Kingdom, coupled with the increase in oil prices in the early 1970's, saw an increase in the development of offshore engineering. Up until this time oil had been extracted in areas where there was no problem of environment, i.e. the desert or the environment was not overly aggressive as in the Gulf of Mexico.

The North Sea thus posed problems both in the severity of the loading and the aggressive nature of the environment. The subsequent jacket designs used were only suitable for relatively shallow waters. The majority of these structures were fabricated from carbon-manganese steels conforming to BS4360 grade 50D. These structures and steels have performed adequately in these shallow waters. There is now a trend, as the shallow water deposits are depleted, to go towards the deeper waters and consequently the design problems become complex. One major consideration is weight. Consequently topside facilities have been fabricated from higher strength steels which give either a greater weight bearing capacity or for the same weight bearing capacity there is a lower weight penalty. It would appear inevitable that the deeper waters of the North Sea will require exploration in the very near future. Any deposits found will need extremely large jackets in order to withdraw the oil and gas. With the loads such a jacket would be required to withstand it is probable that the higher strength steels used on the topside facilities, or similar steel types, will be employed in jacket construction. It is therefore necessary that some of the properties of these high strength steels be investigated prior to their use to assess their suitability.

One of the major problems encountered by an offshore structure is the repeated cyclic loading of the aggressive environment. It is therefore seen that there is a problem both of fatigue loading and corrosion. It is possible to design against fatigue to an extent and to prevent corrosion by cathodic protection but it is important to know what the consequences of the combined action of a fatigue loading situation within a corrosive environment will be.

It is likely, even when cathodic protection is applied, that there will be regions where protection is not supplied together with regions where too much protection is applied. Thus it is necessary to investigate the range of protection levels from freely corroding through to over protection.

2. Literature Survey

It has long been known that if a cyclic load pattern is applied to a material then the material will eventually fail, even though the applied stresses are below the yield stress of the material (1). In general there are two approaches to describing the fatigue resistance of a material, the stress range to give failure in a number of cycles, S-N curves; and the fracture mechanics approach.

2.1 S-N Curves

The production of S-N curves usually involves the use of smooth specimens to which a stress range, S , is applied cyclically until the specimen fails after a number of cycles, N . It is usual to plot the stress range against the log of the number of cycles to failure. With smooth specimens up to 90% of the life can be spent in the initiation of fatigue cracks and the test is regarded as a measure of the resistance to fatigue crack initiation. With steels, it is common for a plateau region to be seen, called the fatigue limit, which corresponds to the stress range below which the material does not show failure by a fatigue mechanism. If a welded specimen is used it is generally observed that the number of cycles to failure is reduced. The recent development of higher strength steels together with improvements in welding techniques and surface contouring has meant that subsurface defects have become the determining factor for the fatigue limit as it is these defects which will act as crack initiation sites (2).

The effect of a corrosive environment on the S-N curve for steels is to depress the curve with the fatigue limit often being completely suppressed as seen in Figure 1. However the fracture mechanics method is concerned with the resistance to crack growth and was more appropriate in this context.

This project was mainly concerned with the effects of environment and welding and it was decided that the S-N approach was not entirely suitable for use in this work although it is recognised that S-N curves are still widely used in design codes for welded structures in a corrosive environment (3).

2.2 Fracture Mechanics Approach

The fracture mechanics approach to describing a material's fatigue characteristics is a relatively recent development. Many workers in the 1950's and early 1960's had tried to

relate the crack growth rate to the crack length, a . Various relationships have been proposed but it was Paris and Erdogan (4) who described the crack growth in terms of a single parameter, the stress intensity factor, K . The simple Paris relationship is:

$$\frac{da}{dN} = C \Delta K^m$$

Where $\frac{da}{dN}$ is the crack growth per cycle, which is usually called the crack growth rate; ΔK is the stress intensity range, i.e. $K_{\max} - K_{\min}$, and C and m are constants dependent upon the testing conditions and material.

2.2.1 Fracture Mechanics

If the region in the neighbourhood of the crack tip is considered, then it is possible to produce relationships to calculate the stresses present. These are given below (5) for the mode I crack opening case.

$$\sigma_x = \frac{K}{\sqrt{2\pi r}} \cos \frac{\Theta}{2} \left(1 - \sin \frac{\Theta}{2} \sin 3 \frac{\Theta}{2} \right)$$

$$\sigma_y = \frac{K}{\sqrt{2\pi r}} \cos \frac{\Theta}{2} \left(1 + \sin \frac{\Theta}{2} \cos 3 \frac{\Theta}{2} \right)$$

$$\tau_{xy} = \frac{K}{\sqrt{2\pi r}} \cos \frac{\Theta}{2} \sin \frac{\Theta}{2} \cos 3 \frac{\Theta}{2}$$

which can be summarised as follows:-

$$\sigma_{ij} = \frac{K}{\sqrt{2\pi r}} f_{ij}(\Theta)$$

Where r and Θ are the polar co-ordinates with respect to the crack tip and the general form of the stress intensity, K , is given by:-

$$K = \sigma \sqrt{\pi a}$$

Here σ is the applied stress and a is the crack length. The above relationship is for an infinite plate and thus for finite dimensions it is necessary to incorporate a geometric factor into the equation, the above equation thus becomes:-

$$K = Q\sigma\sqrt{\pi a}$$

As stated earlier, fatigue involves the application of a cyclic loading pattern and therefore a corresponding change in the stress intensity factor is produced. The stress intensity at the crack tip will therefore be influenced by the maximum and minimum applied loads and as a consequence the maximum and minimum stress intensity factors, $K_{\max}$ and $K_{\min}$ respectively. Thus it is seen that a stress intensity range, ΔK , is produced.

$$\Delta K = K_{\max} - K_{\min}$$

Also of importance is the ratio $K_{\min}:K_{\max}$, this ratio is known as the stress ratio, R .

$$R = \frac{K_{\min}}{K_{\max}}$$

It is also found that most elastic materials do exhibit some plasticity and a small region of the tip of the crack may contain plastically deformed material. However, if this region is small in comparison with the crack length the plasticity can be considered to have a limited effect on the stress intensity and linear elastic fracture mechanics equations are still applicable.

Two general equations are frequently quoted for the size of the plastic zone, r_y . One is for the plane stress situation, the second for plane strain (6).

$$r_y = \frac{1}{2\pi} \left(\frac{K_{\max}}{\sigma_y} \right)^2 \quad \text{Plane stress}$$

$$r_y = \frac{1}{6\pi} \left(\frac{K_{\max}}{\sigma_y} \right)^2 \quad \text{Plane strain}$$

The reason for the difference is the difference in stress state at a free surface compared to the state of stress away from the free surface.

2.3 Fatigue Plots

Conventionally these are plots of the rate of crack growth, usually taken as the increment in crack length per cycle of applied load, plotted at the corresponding stress intensity range, these plots are on log-log scales.

It is usual to find that the curve produced is sigmoidal in shape as seen in Figure 2 (7).

In region A a stress intensity range ΔK_{th} is seen, this value corresponds to the value below which crack growth is assumed not to occur, generally taken to be 10^{-9} m/cycle. It is not uncommon for workers to quote values of 10^{-11} m/cycle as the threshold limit, but as this value is less than the material's inter-atomic spacing it will have no physical meaning. An increase in ΔK from this value shows a rapid increase in crack growth rate for a small increase in ΔK . The curve then shows a decrease in gradient prior to entering region B of the curve. Region B is usually assumed to be linear and represented by the Paris equation given earlier. At higher ΔK values the crack growth rate shows a sharp increase as final failure is approached although in many practical experiments the upper limit K_{max} does not reach K_C or K_{IC} . This results in the sample undergoing a plastic hinging rather than complete sample failure.

2.4 Fatigue Thresholds

The threshold region of Figure 2 is region A. In this particular region the main problem in the production of reliable, low frequency data is the time taken for testing to standard procedures (8).

In general most workers agree with Ritchie that there is an effect of microstructure (9-11), however, the method by which the microstructure influences the threshold is not entirely clear. Two ways in which the microstructure exerts an influence have been proposed. Most workers agree that grain size has an influence with larger grains giving higher threshold values (9, 12-15). The only other microstructure related parameter that has an influence on the threshold value is the material yield strength, with higher yield strengths giving lower threshold values. This is not an unexpected effect since yield strength is related to grain size through the Hall-Petch equation (16, 17).

The influence of mean stress is a direct effect of the applied stress ratio. There is general agreement that an increase in stress ratio will bring about a decrease in threshold value (18-21). The general, although not universally accepted view, is that the stress ratio effect is due to crack closure phenomena (22-24) originally proposed by Elber (25).

The third influencing factor is the environment and because this work is concerned with a sea water environment the review of the literature will be confined to sea water environments.

Generally a change of environment from air to sea water results in a lowering of the threshold value, however, there are a few reported exceptions. Vosikovsky (21) reported for an X70 pipeline steel there was no apparent change in threshold value when the environment was changed, although there was a reduction in threshold as the stress ratio increased in both environments.

2.5 Crack Propagation

Region B of Figure 2 has received considerable attention in recent years. In region B the mechanism for crack propagation has been explained as the formation of striations. Whilst for face centred cubic structures, like aluminium, large regions of striations are easily observed, in body centred cubic structures, such as iron, the regions are generally observed to be neither large nor frequent. Thus, other mechanisms must take place in addition to striation formation.

2.5.1 Striation Formation

The number of models now proposed for striation formation has become excessive, but the three models conventionally accepted or from which adaptations are made are: the Laird and Smith crack blunting model (26), the Tomkins model (27) and the Pelloux alternating shear model (28).

The Laird and Smith model is schematically shown in Figure 3. From this it can be seen that from zero load the application of a tensile load causes the crack to open and slip starts to occur. Continued tensile loading causes the slip to increase and the crack appears to blunt. Reverse loading causes the material to buckle at the crack tip and reverse slip to occur. The crack becomes fully closed at maximum compression prior to reopening on the application of subsequent cycles.

Crack blunting has been observed by other workers (29, 30) and was explained by Tomkins to occur by simultaneous shear (27). However, the blunting process is usually followed by the formation of a sharp crack appearing at some point on the blunted crack front thus maintaining an essentially sharp crack. This apparent resharping of the crack is supported by Wanhill (31) who points out that when an oxidising environment is present, any fresh

metal surface will be readily oxidised whilst simultaneously undergoing continued te
straining. Fracture of the oxide would then maintain a sharp crack by initiating new shear
bands.

The Tomkins model for striation formation is shown schematically in Figure 4. Tomkins reported seeing both fine and coarse striations and the model proposed tried to explain this phenomenon. The larger striation spacing was shown to be related to the maximum crack tip opening displacement, whilst the finer striations had a lower limit of approximately $0.05\text{ }\mu\text{m}$ and above this showed good agreement with the macroscopic growth rate.

The mechanism proposed was shear decohesion. The model requires simultaneous shear band operation to describe the crack opening because alternate shear would result in a lack of striation symmetry.

At low growth rates, Tomkins states that one striation is produced over a number of cycles. This is in direct contradiction to the work by Schijve (32) who used program loading to show that one stress cycle produced one striation. Tomkins explained the formation of a single striation by the following mechanism. During each cycle the maximum crack tip opening is achieved and crack growth can be continuous if the crack tip shear planes accommodate a new portion of crack surface during each cycle. If insufficient surface is produced to form a striation then an accumulation process operates and a striation is only formed when sufficient surface has been accumulated.

Tomkins did not take account of possible microstructural interference which may affect the formation of a striation and may well account for the differences in striations produced compared to the number of cycles applied.

For high crack growth rates the model proposes that several striations are produced for each load cycle because several new crack tip shear planes come into operation.

The third model for the formation of striations differs from these models. Pelloux (28) proposed an alternating shear model rather than the simultaneous shear proposed for the other models. Figure 5 shows how crack extension occurs by alternating shear. Pelloux assumes that shear will occur along a plane and will continue until strain hardening occurs on the plane, it then becomes easier for shear to proceed on a plane that has not been deformed.

In theory the deformation should be completely reversible although in practice oxidation will limit the amount of reversibility of the crack tip deformation. Thus it would be expected that striations would be absent when a material is fatigued in a vacuum. This has been

shown not to be the case by the use of an internally generated notch which was unaffected by environmental factors (33).

The Pelloux alternating shear model is supported by other investigators (34, 35).

The above brief review of striation formation illustrates how complex and difficult the fatigue process is to describe on a microscopic level. Of the different theories put forward the most unlikely for tension-tension loading is the Laird and Smith crack blunting model. This model necessitates the compressive part of the cycle to cause buckling of the crack tip and reverse slip. It is however possible that the plastic deformation at the crack tip which is not accommodated by the elastic surrounding may cause buckling, but it is unlikely that reverse slip will occur.

The Pelloux model of alternating shear would appear to be a logical picture of striation formation and if the formation of striations in an oxidising environment is considered then the limitation of reverse slip by the oxide will account for the striation formation, this model does not account for the formation of striations observed in non-oxidising conditions.

The Tomkins model probably explains the formation of striations in the most accurate way. However, the proposed mechanism of a number of cycles needed to form a single striation is less satisfactory.

It is worth noting again that striations are not observed in general to a great extent in steels. Thus it may be that other failure modes operate. In this case it is seen that the striation spacing exceeds the macroscopic growth rate. Thus it is probable that locally the crack will propagate at an enhanced local rate when growing by a striation mechanism until restrained by some microstructural feature or the crack branches, which will effectively reduce the macroscopic crack propagation rate.

2.5.2 Other Methods of Growth of Fatigue Cracks

Stage II is reported as being a striation mode of crack growth. However, it has been found that additional modes of failure, usually associated with static loading, may occur. These being cleavage, microvoid coalescence and intergranular cracking.

A combined mode of fatigue crack propagation involving striations and micro cleavage, was observed by Richards (36) in a coarse grained silicon iron. The cleavage plane was thought to be the [110] or at least very close to it because the cleavage plane was visible with an optical microscope.

Ritchie and Knott (37) found that when a material was tested below its ductile-brittle transition temperature, crack growth rates were increased and the mechanism was found to be cleavage nucleated at brittle grain boundaries.

Fatigue cracks involving cleavage resulted in enhanced crack growth rates compared to a purely striation mode. Heald et al (38) showed that cleavage was due to the pearlite and its coarseness. The higher growth rates were achieved by the cleavage of cementite during fatigue crack growth.

Microvoid coalescence is generally seen to occur where $\frac{da}{dN}$ is influenced by K_{max} (39).

Various explanations were put forward to explain this.

The first proposal was that a K_{max} effect was a result of the mechanics of plane stress fatigue. If this were true then the effect would have been observed in all materials, but no significant effect has been observed.

The second possibility was that the K_{max} effect depended upon the coincidence of the void coalescence mechanism with plane stress conditions. However, Crooker et al (40) showed a K_{max} effect when void coalescence occurred under plane strain conditions. Although inconclusive, these reports suggest that the fracture mechanism is the dominant influence in the K_{max} effect.

The third possibility is that K_{max} is associated with the fracture mode transition alone. It is probable that the fracture mode transition is related, in some way, with the departure from plane strain conditions and the accompanying change in the distribution of stress and strain around the crack tip.

Where the crack growth rate is small compared with the microvoid size, it is reasonable to suggest that crack growth will be accompanied by striation formation in plane stress and plain strain conditions. Conversely, where the crack growth rate is large compared with the microvoid size, both states will produce a dimpled fracture surface (41). Forsyth and Ryder (42) proposed a mechanism for fatigue crack propagation by void coalescence in which voids form ahead of the main crack and eventually link up by thinning of bridging material during subsequent cyclic loading.

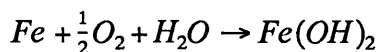
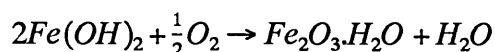
An intergranular mode of crack propagation is reported to occur most frequently in quenched and tempered high strength steels where fatigue cracks tend to follow the prior austenite grain boundaries at low stress intensity ranges and K_{max} values.

Intergranular crack propagation is encouraged by the presence of water and thus testing in air will enable the process to occur. Intergranular crack propagation is usually accompanied by an increase in growth rate and there is evidence that this mechanism is encouraged by impurity elements (43) which reduce the grain boundary strength. Several investigations have shown a decreasing probability of intergranular crack propagation with increasing stress intensities (43).

2.6 Marine Corrosion

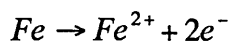
Aqueous corrosion occurs naturally by an electrochemical mechanism if a steel is immersed in water. On a corroding metal there will be both anodic and cathodic reaction sites. At an anodic site an oxidation reaction occurs, this reaction involves the loss of electrons. In order for this reaction to occur it is necessary for a cathodic reaction to occur simultaneously. The cathodic reaction is therefore a process in which the electrons produced by the anodic reaction are consumed.

In the case of steel, the general corrosion reaction is:-



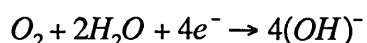
Iron in water forms iron hydroxide as a precursor to hydrated ferric oxide (rust). As the reaction is electrochemical it will occur as two half-cell reactions.

The anodic reaction, which forms one half of the above equation is:-

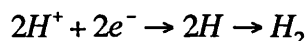


The metal ionises and passes into the electrolyte solution and the electrons are taken into the metal. These electrons are thus available for the cathodic reaction.

There are two cathodic reactions of importance. In neutral or alkaline solutions oxygen is reduced to hydroxyl ions:-



but in acidic solutions hydrogen ions are reduced to hydrogen gas:-



There is a case in aerated acidic solutions in which the two cathodic reactions may act simultaneously although they will be in direct competition.

In the case of sea water the oxygen half cell reaction is most important, although in localised acid regions, e.g. in a crevice, the hydrogen ion reaction may become important.

When corrosion takes place it is usual to find the corrosion product, i.e. rust, forming between the anode and cathode, Figure 6. If corrosion takes place uniformly over the surface the corrosion rate will be minimal. If however a localised region is anodic to a large cathodic region then it is expected that the anodic region will dissolve away rapidly in order to supply the cathodic demand for electrons. In this case severe local pits are produced which act as stress concentrations which can be initiation points for crack growth.

With parent plate steels the corrosion rate tends to be uniform over the surface. If, however, a microstructural discontinuity, such as that introduced by welding, is present it is not surprising to find that there may be localised changes in corrosion rates.

A weldment may corrode uniformly or may be susceptible to varying degrees of preferential corrosion. Figure 7 summarises the different corrosion types. The weld metal may corrode more (Figure 7C) or less (Figure 7B) than the base plate as a result of differences in composition or metallurgical condition. Also the base plate may show preferential corrosion (Figures 7D and 7E) as a result of metallurgical changes caused by the thermal cycles (44).

Although, as seen, the joining of metals using a welding process may result in localised corrosion, it is also known that different locations on an offshore structure are subject to different corrosion rates, Figure 8.

It is seen from Figure 8 that the bulk of an offshore structure has a corrosion rate of 0.15 mm/year. There are two regions where corrosion rates are enhanced from this bulk corrosion rate. The base of the structure is piled into the sea bed and this causes a differential aeration cell to be set up. In addition there may be bacterial affects and the scouring action of the sand will continually expose fresh metal thus accounting for the enhanced rate. The enhanced corrosion rate seen at the mud line is overshadowed by the problem encountered

in the splash zone which is subjected to alternating wet and dry cycles together with a mechanical affect from wave motion. The result being a ready supply of oxygen to maintain the corrosion rate at an enhanced rate, approaching 1 mm/year.

2.7 Other Factors Affecting the Corrosion Rate

2.7.1 Depth

In general the corrosion rate will decrease with increasing water depth because the oxygen content is reduced at increasing depths, although sulphate reducing bacteria may have a significant effect when the oxygen content becomes very low. The water depth may also have other effects such as on temperature, which decreases with increasing depth. Other effects such as water pressure, caused by increasing depths, have not been documented. Any effects of water depth would probably not apply to a great extent in the relatively shallow waters of the North Sea.

2.7.2 Temperature

It would be expected, from an electrochemical kinetics point of view, that a decrease in water temperature would cause a decrease in the corrosion rate, but this does not appear to be the case since structures in the warmer parts of the world do not suffer from greater corrosion rates. It has been shown that temperature does have an effect on the free corrosion potential. The lower the temperature, the less negative the free corrosion potential (45). Although, to put these changes into perspective, the change in free corrosion potential was only 40 mV for a 50°C temperature change. The effect of temperature on the electrochemical potential is seen in the Nernst equation;-

$$E = E_o + \frac{RT}{nF} \ln \left(\frac{a_{react}}{a_{prod}} \right)$$

2.7.3 Water Velocity

The flow rate has a great influence on the corrosion of steels in oxygen containing electrolytes, as it determines the diffusion of oxygen to the surface and the removal of corrosion products. An increase in the flow rate causes an increase in the corrosion of the steel. It does however appear that the flow rate effect shows a levelling off as the velocity approaches 6 m/s (46) as shown in Figure 9.

2.8 Corrosion Control

Aqueous corrosion is an electrochemical process and there are various methods available to reduce the corrosion rates to acceptable levels. It is theoretically possible to reduce the corrosion rate down to almost zero but this can cause other, more serious problems.

Painting, as used topside on offshore structures, provides a protective coating but this coating will contain defects if preparation is not adequate. Painting is also expensive, especially if done offshore. It is not the cost of the paint that causes the high costs but other factors such as scaffolding, etc. Duncan (47) estimated that the cost of painting the tidal zone of a North Sea platform was £300/m² in 1984. However, painting is generally not used for the submerged parts of the platform and was not used in this work and so will not be considered any further.

The alternative to painting the submerged regions of a platform is cathodic protection. The dissolution of a metal in a solution is an anodic process and it is generally agreed that the application of a cathodic potential will cause the anode reaction to decrease and eventually cease which is shown in Figure 10.

If no protection is applied to a specimen it is seen that the specimen would have a mixed corrosion potential E_{corr} and will be corroding at a rate equivalent to i_{corr} . If a cathodic potential is applied such that the specimen potential is depressed to E_1 it is seen that the corresponding i_{corr} is reduced and hence the corrosion rate is reduced. Further depression of the potential will cause i_{corr} to decrease until the situation at E_2 is reached where i_{corr} becomes zero and hence corrosion ceases. Any further reductions in potential ensure corrosion does not occur but then hydrogen evolution becomes a problem, these problems will be discussed later.

Cathodic polarisation is achieved in two ways:-

2.8.1 Sacrificial Anodes

Sacrificial anodes make use of the principles of bimetallic corrosion. If two metals are connected such that there is both an electron path and an ion path (electrolyte) then a cell will be set up. In this cell one metal will have a greater tendency to form ions. This tendency to form ions is represented by a more negative free corrosion potential. Table 1 shows the galvanic series in sea water.

Thus if we want to cathodically protect a steel it is necessary to use a material which will preferentially corrode, i.e. a material with a lower free corrosion potential. In order to sacrificially protect steel, three metals or alloys of these metals have generally been used; magnesium, aluminium and zinc. Sacrificial anodes are evaluated partly by their capacity, this is the number of ampere - hours a kilogram of material can supply. However, the theoretical capacity is not normally achieved and the actual capacity will be controlled by the efficiency (48). Sacrificial anodes have both advantages and disadvantages. The main advantage of sacrificial anodes is that once immersed protection will begin immediately, provided enough anodes are present, and will continue to protect the steel until the anode material area becomes too small to supply the required current or the anode falls off. The disadvantages of the sacrificial anodes are that they add a lot of weight to the structure and increase drag forces thus increasing both the static and dynamic loads on the structure. There is also the possibility that many of the anodes will become detached during the initial piling operations used to secure the structure in position. Any retrofit of consumed or detached anodes in order to extend the operational life of the structure is very expensive.

2.8.2 Impressed Current Systems

Although the principles underlying impressed current cathodic protection are similar to those of the sacrificial method, there are some important differences.

It is necessary in sacrificial systems that a good electrical contact exists between the anode and cathode, although there needs to be a small stand-off between the two, but in the impressed current system the anode is often shielded by an insulating layer to avoid excessive currents around the anode. There are even cases where, for offshore structures, the anode is sited up to 100 metres from the structure on the sea bed. This is thought to give a better potential distribution (49).

Unlike the sacrificial anode system, impressed current systems normally use anodes which are virtually non-consumable, this alleviates the problems of having to renew large quantities of anodes at regular intervals. It is also possible, with the use of electronics, to make the system self-regulating which is a significant advantage over the sacrificial anode system where there is no control of the current supply once the anodes have been installed.

Impressed current systems usually use a rectifier to convert the normally available alternating current into direct current of the required voltage. Many different materials have been used as anodes ranging from scrap iron to platinised titanium, although modern systems do tend to use the materials which are consumed slowest. Table 2 lists some anode materials and their properties.

There are, however, disadvantages to the impressed current system. In harsh conditions, such as the North Sea, the mechanical vulnerability of impressed current systems makes them a second choice to sacrificial anodes. Also for any work requiring divers to operate close to the jacket the impressed current system must be turned off for safety. For both sacrificial anode and impressed current cathodic protection a major factor in the design of the system is the current density required to fully protect a structure. Not surprisingly the current density will vary according to conditions, e.g. the roughness of the sea, and higher currents will be required initially in order to polarize the structure but as calcareous deposits form the current requirement will decrease because the calcareous deposit will act as a barrier between the steel and the sea water (50). Table 3 gives the recommended minimum design current densities for protection of fixed offshore steel platforms (51)

2.9 Corrosion Fatigue

It is fair to say that the addition of a corrosive environment to the problem of fatigue crack propagation does, in the majority of cases, enhance the effects of the testing variables.

For a given steel and environment the important testing variables are: the stress intensity range, ΔK ; the stress ratio, R ; the testing frequency; and the environment temperature.

2.9.1 Stress Intensity Range, ΔK

The derivation of ΔK was briefly discussed earlier and it was shown that ΔK was the difference between the maximum and minimum stress intensity factors.

The literature is unambiguous about the effects of ΔK on the growth rates: the growth rate increases with increasing ΔK (52-55). It has also been conventional to define the central portion of the typical sigmoidal fracture mechanics plot (Figure 2) as being linear. Whilst under some conditions, such as in air, the linear relationship may be an adequate description of the variation of crack growth rate with ΔK , there are many situations where the linear relationship is not capable of accurately describing the crack growth. These situations are typical of corrosion fatigue in over protection conditions or in other environments which are exceptionally aggressive.

Having noted that the relationship between $\log$ (growth rate) and $\log (\Delta K)$ is not always linear, it is still felt that the plots of $\log$ (growth rate) against $\log (\Delta K)$ are still an excellent method by which to compare different materials and assess the effects of environment on the crack growth rate.

2.9.2 Stress Ratio, R

The stress ratio, R, is the ratio of minimum to maximum applied stress and by definition is also the ratio of minimum to maximum stress intensity. The effects of stress ratio have been investigated by numerous authors and the general consensus is that with higher stress ratios the growth rate for a given ΔK value is seen to increase (21,56,57). It is through the simple relationship:-

$$\Delta K = (1-R)K_{\max}$$

that ΔK is related to R. This shows that for the same $K_{\max}$ but different R values, variations in ΔK can be produced, which implies that $K_{\max}$ is the important parameter and the same $K_{\max}$ values should give the same crack growth rates regardless of the ΔK value.

Whilst in the threshold region the effects are well accepted in the Paris region there is conflict over whether the stress ratio does have a significant effect. Many of the stress ratio effects are linked with the phenomena of crack closure proposed by Elber (25). In brief, Elber proposed that the plasticity caused by the application of a load was sufficient to cause the crack to close prematurely when the load was removed and the actual ΔK applied would be less than that applied by calculation because a certain load was required to open the crack before crack extension could begin. This leads to the term ΔK_{eff} where:

$$\Delta K_{\text{eff}} = K_{\max} - K_{\text{op}}$$

where K_{op} is the stress intensity to cause the crack to start opening. Various relationships have been put forward to calculate ΔK_{eff} from the traditional ΔK and all these relationships involve the stress ratio, R. A typical empirical relationship put forward by Elber for an aluminium alloy was:

$$\Delta K_{\text{eff}} = \Delta K(0.5 + 0.4R)$$

It does appear from the literature that there are different relationships for different materials and to list them here would be both time consuming and serve no particular purpose as it is felt that at high stress ratios closure is not a significant variable and the testing reported here was done at a high stress ratio (0.6).

2.9.3 Test Frequency

In air the testing frequency has limited effect as a significant variable. It has been shown that at higher frequencies the S-N curves give greater fatigue lives than the lower frequencies (58). For crack propagation it would appear that it is only when the frequency becomes very low and the stress ratio high that the crack propagation rate is affected (59). The frequency becomes a very important variable when a corrosive environment is present. It has also been shown that the effect of frequency is also dependent on the waveform (60). A schematic representation of a waveform together with the various times associated with the waveform is shown in Figure 11 (61).

The effect of frequency on crack growth in corrosive environments, like in air, is the lower the frequency the faster the growth rate. However, if the waveform is included in the assessment then the frequency effects can be seen to differ.

Corrosion is proportional to time and hence a frequency effect on corrosion fatigue is expected. It does appear from the literature that the crack growth rate is mainly proportional to the stress rise time, T_r , of the waveform.

Atkinson and Lindley (61) determined the effect of frequency on crack growth for the case $\Delta K = K_{max}$, i.e. $R = 0$, and found that the crack growth rate is highest at a frequency of 0.01 Hz, with the triangular waveform as shown in Figure 12.

The effect of rise time is shown in Figure 13 where it is seen that the square waveform and negative saw tooth produce the same crack growth rates which implies that growth rate is not proportional to total cycle time but to rise time. The highest growth rates occurred at a rise time of 100 seconds. Increased rise times then started to give reduced growth rates.

Atkinson and Lindley (61) also looked at the effects of hold times at maximum stress and showed that for two different rise times $\frac{da}{dN}$ was not significantly changed with increased hold times. The effect of hold time can however vary and does appear to be dependent upon whether K_{max} is above or below K_{ISCC} , the stress corrosion cracking threshold.

There are definite connections between corrosion fatigue and stress corrosion cracking. Staehle (62) showed that if the mean stress is high and the amplitude is low the conditions of stress corrosion cracking start to become a dominant feature. This is shown schematically in Figure 14.

Figure 15 shows the effect of hold time on the corrosion fatigue crack growth rates above and below K_{ISCC} .

The frequency has also been shown to affect the fracture mode (63). It does appear that intergranular cracking occurs predominantly at low frequencies as shown in Figure 16.

It is thus important that the expected cyclic waveform and a realistic frequency are used in the testing of materials.

2.9.4 The effects of Temperature

The effects of temperature on marine corrosion were reviewed earlier and whilst it is accepted that temperature does have an influence on the corrosion rate this section will be confined to the literature concerned with the effects of temperature on the fatigue crack propagation rate only.

There is a definite trend in the effect of temperature on corrosion fatigue, in that an increase in temperature results in an increase in crack growth rate.

It has been reported that the fatigue life of mild steel can be reduced by increasing the temperature from 15°C to 45°C in artificial sea water where as in air tests show no appreciable temperature effect in the above range (64).

However, Jaske et al. (57) conclude that 'except for possibly important effects on cathodic protection, temperature variations in the range 4 to 20°C are expected to have only a small effect on corrosion fatigue behaviour'. It has also been stated that calcareous deposits do not form as readily at temperatures below 5°C (65). This would imply that any effects of calcareous deposits whether beneficial or detrimental will be greatly reduced below 5°C.

2.9.5 The effects of pH

In the practical situation of a sea water environment, the bulk pH will not significantly vary from the expected pH of about 8. There is however every likelihood that the pH within a fatigue crack will vary. In order to assess the possible effects of pH changes within a crack it is necessary to review the literature concerned with bulk environment changes.

It is reported that changes in the pH in the range 4-10 in a salt solution result in little or no change in fatigue life (66,67). At the extremes of the above range it was stated that below pH 4 fatigue life decreased rapidly whilst at pH above 10 the fatigue life returned to almost that of air (45). It is thus clear that bulk changes in pH in sea water will have little effect on the fatigue life in the range $4 < \text{pH} < 10$. However, the region of importance

is within the crack where there is evidence of acidic conditions (68). More recent studies on the effects of pH on the fatigue crack propagation rate, rather than the fatigue limit, show that in general the crack propagation rate increases as the pH decreases (69).

2.9.6 The effects of Microstructure

In addition to the effects of the testing parameters and the environment there is now a considerable amount of literature indicating that the material microstructure may also have an effect on the fatigue crack growth rate.

The microstructure is an accepted influencing parameter in regions A and C of Figure 2 but in region B there are still conflicting reports over whether or not microstructure plays a significant role in determining the magnitude of the observed fatigue crack growth rate. Lindley et al. (70) reviewed the literature and concluded 'a study of the fatigue fracture surfaces of metals having a wide variety of microstructures and tested over a wide range of ΔK and K_{max} has shown four general mechanisms of growth, namely, striation formation, cleavage, void coalescence and intergranular separation. A plot of fatigue crack growth rate against ΔK has been found to be sigmoidal in nature, Figure 2. In the mid ΔK region (region B) where striation growth occurs, there is little influence of microstructure, mean stress, dilute environment and specimen geometry on fatigue crack growth.'

The above conclusion appears to be based on the data of Clark and Wessel (71). This data shows a wide range of steels in which cracks have propagated by either striation formation or one of the other mechanisms quoted. The striation mechanism generally gave the lowest crack growth rates but to say there is no influence of microstructure when striation formation is the mechanism of crack advance would appear to be hasty considering that variations in growth rate of up to a factor of five are readily observed from the Clark and Wessel data.

Lindley et al. do concede that when microcleavage, void coalescence or intergranular separation occur, the growth rate is not adequately described by the simple Paris equation. It is worth noting that Laird (72) admits that striations are very difficult to find in steels. This must imply that striation formation is not necessarily the primary mechanism of crack propagation and it would be expected that other mechanisms dominate the crack advance process. However in this work extra care was taken in looking for fatigue striations and consequently a high proportion of the fracture surface was seen to contain striations.

Consideration of the Lindley et al. (66) equation for non-striation advance of the crack does imply that microstructure plays a role in the determination of crack growth rate.

$$\frac{da}{dN} = A \left(\frac{\Delta K^4}{\sigma_i (K_c - K_{max})} \right)^n$$

Where A is a material constant; σ_i is a strength parameter; K_{max} the maximum stress intensity in the fatigue cycle; K_c the fracture toughness of the material. Two terms in the above equation are influenced to a large extent by microstructure, i.e. σ_i and K_c , and the microstructure must surely play a part in the determination of the crack growth rate. There is now more literature which agrees that microstructure is important to the fatigue crack growth in region B of Figure 2.

Barsom (73) noted that in ferrite-pearlite steels there was a significant influence of the pearlite colonies on the fatigue crack propagation. Barsom proposed that fatigue crack growth might depend on; the strength of pearlite relative to the ferrite, the interface bond between the two, on the size and orientation of the pearlite colonies and on the distribution density. Barsom suggests that pearlite colonies, because they are stronger than the ferrite matrix, will retard crack growth. He also proposes that the pearlite colonies will be left as unbroken ligaments as the crack advances. This will have the effect of reducing the stress intensity factor at the crack tip. These ligaments will fail when they are loaded to their tensile instability. Barsom does not discuss the influence of the pearlite lamellae orientation.

Work by Aita and Weertman (74) follows on from the work of Barsom. Aita and Weertman show not only the effect of pearlite content on the growth rate but also the effect on the exponent m in the Paris equation.

Figure 17 shows the crack growth per cycle for different iron-carbon alloys at $10 \text{ MPa}\sqrt{m}$. The value of $10 \text{ MPa}\sqrt{m}$ was chosen by these workers because crack extension was by a ductile mechanism for all alloys at this stress intensity range.

Generally it is seen that there is a reduction in growth rate with an increase in brittle micro constituent, however, there is one alloy which is in direct contrast to the rest. The exception was a 0.64% C steel which had been oil quenched and then tempered. The lower fatigue crack growth rate was explained by the arrangement of the carbides developed during the tempering process in which the iron carbide particles formed a network, partially enclosing the ferrite cells.

Aita and Weertman (74) also showed that the brittle constituents had little effect on the Paris exponent m as shown in Figure 18.

The exception was the alloy containing 80 volume percent brittle micro constituent. In this alloy the slip path was considered small and fatigue failure was predominantly by void formation and growth around flaws in the interlamellar ferrite.

Whilst microstructure has been seen to have a direct affect on the fatigue crack propagation rate there is another indirect consequence of microstructural variations. Microstructural changes can result in increases in yield strength which in turn leads to a reduction in section thickness required to support the current design loads of conventional steels. Investigations into the affects of section thickness on the fatigue limit have led to the incorporation of a correction factor in the design codes (75). Since the incorporation of a correction factor in the codes various studies have focused on the thickness effect and have shown the design codes to be conservative in their allowances (76). The thickness effect has also been shown to be dependent upon the welded attachment size and the loading mode as well as the main plate thickness (77).

2.10 Stress Corrosion Cracking

Any discussion of corrosion fatigue of high strength steels must include the possibility of stress corrosion cracking (SCC).

There is an inter-relationship between corrosion fatigue, SCC and hydrogen embrittlement. Hockenhull (78) summarises the requirements for SCC as:

1. Susceptible alloy composition
2. Susceptible microstructure
3. Specific aggressive environment
4. Sustained tensile stress.

Consideration of the typical SCC growth rate as a function of K_I , the applied stress intensity in mode I, shows there is a similarity to that obtained for fatigue, although in region II there is a plateau region where the crack growth appears independent of K_I compared to fatigue where a linear dependence is conventionally accepted. This is illustrated in Figure 19.

Regions I and III show growth rate dependence upon K_I . Region I shows a threshold, $K_{I,SCC}$, below which the propagation of stress corrosion cracks is assumed not to occur, though for SCC this clearly depends on the patience of the observer!

Conventionally the susceptibility to SCC for steels has been assessed by the material yield strength with higher yield strengths resulting in greater susceptibility. Jones (79) shows this graphically and Figure 20 shows that the introduction of a pre-crack does enhance the susceptibility.

Figure 20 would imply that materials with yield strengths below 500 MPa would show little SCC, however, when hydrogen is present in an aqueous environment, such as cathodic over protection in a corrosion fatigue situation (80-82) SCC is known to occur.

Thus, the question should be does hydrogen embrittle steel? The answer to this question is yet to be conclusively answered and at present, according to Hirth (83), there are nine theories to account for the effect of hydrogen.

Basically these theories fall into two categories: the internal pressure theory; and the decreasing strength theory.

The internal pressure theory, originally put forward by Zapffe (84), proposed that embrittlement was due to the recombination of atomic hydrogen to form molecular hydrogen at defects such as inclusions, and would be accompanied by a large pressure change which could cause expansion of microvoids and microcracks.

Van Leeuwen (85) points out that high concentrations of hydrogen may occur in localised regions due to the tendency of hydrogen to diffuse to and accumulate at sites with high triaxial stresses. Thus from Sieverts law high pressures of hydrogen are predicted. A good example of the effects of pressure is the localised build up of hydrogen at MnS inclusions which causes debonding of the interface between the inclusion and the metal matrix.

Van Leeuwen (85) also concludes that an internal pressure and a gross section stress is necessary to cause cracking and the presence of an internal stress will reduce the magnitude of the external stress required for cracking.

The second category is the decreasing strength theory which proposes that the presence of hydrogen dissolved in the metal matrix lowers the cohesive strength of the iron lattice.

A popular theory of hydrogen cracking, proposed by Troiano (86) and subsequently developed by numerous authors (87-89), assumed that hydrogen cracking is controlled by the long range diffusion of hydrogen in the stress field associated with a notch. The stress field is the driving force for the increase in hydrogen concentration and if concentrations reach a critical level fracture occurs.

It has been found that cracking occurs some distance in front of a machined notch, this distance being determined by the notch radius (90). In the case of a crack tip the hydrogen

concentration will be a maximum at or near the elastic plastic boundary. The boundary distance will be dependent upon the applied stress and will be the deciding factor for the extension of the hydrogen induced fracture.

Raj and Varadan (91) propose a similar model for high strength steels. In their model the crack propagates by the initiation of a secondary crack ahead of the crack tip, and by the interaction of this secondary with the main crack. The secondary crack forms in the region of maximum tensile stress in the stress field of a sharp crack with small scale yielding. Raj and Varadan propose that the initiation of this secondary crack is both stress and time dependent because a threshold stress is required to break the Fe-H-Fe bonds in the grain boundary and an incubation time is required to form a two dimensional cluster of hydrogen atoms in the grain boundary which is large enough for local fracture. Hence crack propagation is discontinuous and proceeds by a series of jumps.

Tomkins (92) examines the mechanics implications of the diffusion model for corrosion fatigue involving the embrittlement by microfracture ahead of the crack, this is shown schematically in Figure 21.

Tomkins states that the depth (r_{d1} , r_{d2} , r_{d3}) ahead of the crack that critical concentrations of hydrogen will form is dependent upon both the effective hydrogen concentration and the time available. Depending on the K_{max} and ΔK level operating this may or may not be within the critical stressed zone, r , within the plastic zone.

If the depth of hydrogen concentration is less than the critical stressed zone size microcracks of length r_d can form independently of K_{max} and ΔK giving the plateau regions seen in Figure 21.

If the depth is larger than the critically stressed zone the microcrack size is equal to r and $\frac{da}{dN}$ is proportional to $(K_{max})^2$.

The permeability of hydrogen in steel is determined to a large extent by the form taken by the carbon in the steel. Tomkins states that ferrite/pearlite structures are more susceptible than bainitic structures. A more comprehensive study, as shown in Table 4, lists the results of several investigations, in which different steel compositions were used (93). Each steel was capable of being heat treated to several structures.

It is seen that martensite has the lowest permeability, bainite is lower in permeability than pearlite but there is a variation between the two in respect of coarse or fine structure permeability.

2.11 Cathodic Protection

Cathodic protection has found widespread use in the offshore industry. It is possible to calculate that the application of a potential of 0.2 to 0.3 V more negative than the free corrosion potential will reduce the corrosion rate of a carbon steel by 99%. On this basis a typical potential level of -0.78 V (SCE) is suggested for adequate protection (94).

It is however practically impossible to maintain a uniform protection level over an entire structure and there will be regions of over protection and similarly regions of under protection.

Scott and Silvester (56) found that a cathodic potential of -700 mV (Ag/AgCl) gave the lowest crack growth rates in the range -0.63 to -1.1 V. It should be remembered that crack growth rate testing was performed on notched and pre-fatigued specimens and no initiation time was required. Figure 22 summarises the effect of cathodic potential on the crack propagation rate.

Consideration of S-N curves shows that in general the more negative the potential the higher the fatigue limit. Hooper and Hartt (95) found that a potential between -1.1 and -1.3 V (SCE) gave the highest fatigue limit. Tubby and Booth (96) found that under simulated North Sea conditions the fatigue strength of two welded high strength steels was influenced by the level of protection applied to the material. Fatigue testing with no protection caused a reduction in the fatigue limit, but when a cathodic potential of -850 mV was applied the fatigue limit increased from that observed with no cathodic protection. The fatigue limit observed with a cathodic potential of -850 mV was still lower than the fatigue limits of the in air sample.

Since this work was concerned solely with notched and pre-fatigued specimens, only the cathodic potential in notched specimens will be considered from now on.

In the absence of an applied cathodic potential corrosion will occur. At the tip of the fatigue crack, because this region is highly stressed, anodic dissolution is a possibility. The application of a cathodic potential will suppress the rate of corrosion and the general trend is for the crack propagation rate to return to approximately the rates observed in air.

Scott and Silvester (56) attributed this effect to the wedging of the crack by the deposition of calcareous deposits which had the effect of increasing the minimum crack opening displacement at a greater rate than the maximum crack opening displacement increased.

The above explanation is open to debate because firstly the calcareous deposit that is produced is not the hard compound usually examined in the dried form but rather a "mushy" compound which will have little load-bearing capacity. Secondly, the production of

calcareous deposits at a potential of -800 mV will exceed that of a potential of -700 mV and should, if indeed crack wedging occurs, apply a greater wedging force and hence decrease the crack growth rate to levels below those of the -700 mV protection potential. Continued reduction in potential results in a rapid increase in crack growth rate. The cause of this increase in growth rate is undoubtedly hydrogen embrittlement.

Two models have been forwarded to account for the effects of hydrogen embrittlement of the crack growth rate in corrosion fatigue.

Landes and Wei (97) have proposed a superposition model which attempts to predict the crack propagation rates from the algebraic sum of the mechanical fatigue component and an environmentally controlled parameter that may be determined independently.

The equation for this model is:-

$$\left(\frac{da}{dN} \right)_c = \left(\frac{da}{dN} \right)_{env} + \left(\frac{da}{dN} \right)_r$$

where c is corrosion fatigue, env is the environmental component and r is cycle dependent. Landes and Wei further postulated that:-

$$\left(\frac{da}{dN} \right)_{env} = \int_t \left(\frac{da}{dt} \right) K(t) dt$$

where K(t) is a function of the fatigue cycle load waveform over a time period t. This leads to the final equation:-

$$\left(\frac{da}{dN} \right)_c = \left(\frac{da}{dN} \right)_r + \int_t \left(\frac{da}{dt} \right) K(t) dt$$

This equation was used satisfactorily to predict the corrosion fatigue crack growth rates of several high strength steels in both hydrogen and humid air at low stress ratios.

Tomkins (92) dismissed this stating that their success was probably due to the dominance of the environmental term and indications are that the reference term $\left(\frac{da}{dN} \right)_r$ forms only a base line and there is little evidence of real summation.

The second model was proposed by Austen and Walker (89) and is known as the 'process competition model'. Unlike the Landes and Wei model, Austen and Walker propose that the processes of crack advance are competitive rather than additive. This suggests that the fatigue crack will propagate by the fastest mechanism available.

It is known that the three major forms of sub-critical crack growth: fatigue, corrosion fatigue and SCC can all be represented in fracture mechanics terms since the parameter governing the rate of crack advance is the stress intensity parameter, K , or its cyclic range, ΔK .

The Austen and Walker model suggests that the stress corrosion threshold, K_{ISCC} , is the separation point between the different mechanisms of crack advance, i.e. 'true' corrosion fatigue and stress corrosion fatigue.

True corrosion fatigue identifies the behaviour where fatigue crack growth rates are enhanced by the presence of an aggressive environment through a synergistic action of corrosion and cyclic loading. In order for true corrosion fatigue to occur ΔK must be below K_{ISCC} and the material must not stress corrode.

Stress corrosion fatigue describes the phenomenon of static load stress corrosion under fatigue conditions and this occurs whenever the stress intensity in the fatigue cycle is above K_{ISCC} .

Figure 23 shows both the true corrosion fatigue and the stress corrosion fatigue together with the behaviour that relates to material/environment systems which exhibit stress corrosion fatigue above K_{ISCC} and also true corrosion fatigue throughout.

Therefore the Austen and Walker model is more likely to accurately predict the crack growth rates of materials which do suffer from SCC when tested under corrosion fatigue conditions.

2.12 Welded Joints

Earlier sections have made brief mentions of the effects of welding on the fatigue strengths of a material. The general conclusion being that the welding process greatly reduces the fatigue life of a component because it introduces defects and surface discontinuities which can act as crack initiation points. The influence of surface discontinuities can be reduced by suitable profiling of the weld (99, 100).

If high strength steels are to be of benefit then there must be no loss of fatigue strength when the material is welded. However, Maddox (101) states: "The fatigue strength of welded details, characterised by mean S-N curves, were common to structural steels and

independent of their tensile properties. It follows that design codes are also common. In fact this situation applies to welded steels in general, including high strength or special steels which can otherwise be welded satisfactorily. Such behaviour was explained by reference to the initiation and propagation of fatigue cracks. It is observed that the tensile strength of steel has little effect on the rate of propagation of a crack. The fact that unwelded specimens show benefit from increases of tensile strength must therefore derive from the existence of crack initiation, or incubation, periods in addition to that required for propagation, equally, the constant fatigue strength of welded details implies that their cyclic life is determined essentially by propagation alone. This has been observed to be the situation for joints in class D and below. Such behaviour is attributable to the presence of pre-existing crack like flaws, such as the inherent intrusions at weld toes, which, in effect represent a condition in which crack initiation is achieved. Welded high strength steels offer, therefore, no intrinsic advantage in terms of fatigue strength and it is necessary to assess design criteria carefully to determine when they may be used with benefit in the context of fatigue loading."

Bigonnet et al (102) disagree with the above statement by Maddox and state that high strength steels and improved welding procedures allow the fatigue strength to rise substantially.

However, because defects are inherent and the nucleation period of the fatigue crack will be effectively removed, it is probably more beneficial to investigate the fatigue crack propagation in a weldment.

There are three regions of a weldment: the parent plate, the heat affected zone (HAZ) and the weld metal. Ekman (103) compares the crack propagation data from all three regions of a structural steel AB2525, which is similar in both composition and strength to BS4360:50D. Figure 24 shows the comparison between the fatigue crack propagation data for the three regions.

As can be seen, the parent plate gave the highest growth rates whilst the HAZ gave the lowest with the weld metal intermediate between these two. There is also a considerable degree of overlap of the data.

In contrast, Ohta et al. (104) found that the base metal gave the lowest growth rates, whilst the HAZ and weld metal gave similar crack growth rates in a HT80 with a 2.1 kJ/mm multipass weld. In the work of Ohta et al. the specimen was centre cracked and the authors attributed the different growth rates between different regions to the tensile residual stresses in the plate.

There is obviously some confusion over the effect of welding on the fatigue crack propagation rate. However, it does appear that residual stresses are credited with causing the variations in growth rates. It is worth noting that there can be no net residual stresses and thus different plate or weld regions may give different crack propagation rates. However in a fabricated structure it is necessary for a crack to propagate through both tensile and compressive regions and so any benefits of residual stresses will be limited.

If a corrosive environment is present it might be expected that the variation in fatigue life as measured by S-N curves for the different regions might not be so great. This is because the initiation time for the parent plate material can be significantly reduced by the defects produced by corrosion which may act as stress concentrations.

In notched samples, Vosikovsky (105) found that at low stress ratios the crack growth rate of the base metal can be considered to be a conservative estimate of the crack growth rate in the weld metal but at high stress ratios the crack growth rate in the weld metal is greater than in the parent plate. Vosikovsky attributed this to the boundaries of the columnar dendrites in the weld metal being susceptible to environmental cracking.

Work by Knight (106) does show that the weld metal crack growth rates obtained from weldments of two high strength steels had similar crack growth rates to the parent plates when tested in a sodium chloride solution. These crack growth rates were compared to the parent plate in air crack growth rates and were seen to be considerably higher than the in air crack growth rates.

In an attempt to discover the effects of different regions of the weldment on the fatigue crack growth rates, Dowse and Richards (107) used a simple constant ΔK test. A multipass weld was used to produce the required HAZ for their work, although there is no mention of the heat input used. To maintain a constant ΔK they employed a method of manually reducing the applied load every 0.5 mm. of crack growth.

If the growth rates obtained from Ducol W30 (a low alloy steel) are considered it is seen, from Figure 25, that at a ΔK of $38.8 \text{ MPa}\sqrt{\text{m}}$ there is a slight decrease in crack growth rates as the crack propagates into the HAZ and reaches a minimum crack growth rate approximately 3 mm from the fusion boundary. Continued crack growth shows that the rate increases as the fusion boundary is approached. The growth rate at the fusion boundary is slightly greater than the growth rate in the parent plate and the growth rate remains constant into the weld metal.

Increasing the ΔK shows that a similar trend is obtained, although the growth rates are higher, as would be expected. However, there is a much larger variation in crack growth rate than was seen in the lower ΔK test.

In the case of the mild steel weldment the crack was grown from both the parent plate to the weld metal and from the weld metal to the parent plate.

For the growth from the parent plate into the weld metal Dowse and Richards found that the crack growth rate showed a general decrease as the crack approached the HAZ, this decrease continued as the crack propagated through the HAZ. However, it was found that the crack would not propagate through the fusion boundary and instead was deflected by the boundary.

On propagating in the reverse direction the crack growth rate was seen to be constant in the weld metal but showed a decrease in crack growth rate as the fusion boundary was crossed. Further propagation showed the HAZ crack growth rate to increase until a constant crack growth rate was observed.

Dowse and Richards explained all the observed features of the crack growth rate variations in terms of the plastic zone around the crack tip. The reason why the crack shows a variation in the rate of crack growth is because the crack tip plastic zone is progressively decreased as the crack propagates towards the weld metal but in the opposite direction the plastic zone is increasing.

There appear to be a few points that are not satisfactorily explained by Dowse and Richards:

1. The growth rate in the weld metal was only slightly lower than in the parent plate but the hardness variation and hence yield strength and plastic zone size was large.
2. The growth rate in the HAZ, when the crack propagates away from the weld metal, appeared to attain a constant rate whereas in the opposite direction the crack growth rate decreased. A constant crack growth rate was not consistent with the variations in hardness observed and either the growth rate of the crack propagating away from the weld metal should have increased or, more likely, the crack growth rate of the crack propagating towards the weld metal should have been more constant until a distance approximately 1 mm from the fusion boundary at which point the plastic zone would have been affected by the large increase in hardness observed. At this point the crack growth rate should have decreased.
3. The measured crack growth rate in the HAZ of the sample with the crack propagating away from the weld metal was significantly higher than that observed in the mild steel parent plate which had a lower hardness value.

For the Ducol steel the plastic zone argument appears to be more valid. The crack starts to decelerate approximately 4 mm from the edge of the HAZ which corresponds to the size of the plastic zone in the high ΔK test and increases as the hardness ahead of the crack decreases.

However, the crack growth rate in the weld metal is greater than in the parent plate which is in direct contradiction to the plastic zone argument.

It is difficult to give other possible reasons for the variations in crack growth rates observed due to the lack of information quoted by Dowse and Richards. Thus it can only be speculated that the variation in plastic zone size may only play a part in the observed variations in crack growth rate.

3. Experimental

3.1 Materials

Two high strength steels were used in this investigation, one produced by a quench and temper route, the other by a controlled rolling process.

3.1.1 Superelso 500

Superelso 500 is a quenched and tempered steel with a low carbon content and low levels of sulphur and phosphorus. The heat treatment process involves heating the plate to 925°C for 15 minutes then water quenching followed by a tempering process at 625°C for 30 minutes before the plate is air cooled. Chemical analysis of the material by NEI showed reasonable agreement with the manufacturer's analysis although some of the alloying additions show some variance (Table 5). Mechanical properties of the material show that in the parent plate the yield strength is of the order of 530 MPa whilst Charpy V-notch energies of approximately 90 J at -50°C show that the material has good impact resistance (Table 6). Welding of the plate causes a slight reduction in yield strength but the toughness and impact energies of the material remains high (Table 7).

The material was supplied as plate with a section thickness of 60 mm. The typical microstructure of the plate was seen to be equi-axed ferrite within a mixed bainite matrix.

3.1.2 Mannesman X70

Mannesman X70 is a controlled rolled micro-alloyed steel with a low carbon content. Heat treatment involves initial rolling at temperatures of about 1150°C. Section thickness is reduced with subsequent passes at lower rolling temperatures with the final passes being carried out at about 750°C, followed by air cooling. The resulting microstructure of the plate consists of ferrite together with a mixture of pearlite and bainite.

Chemical composition of the steel shows good agreement between the manufacturer's analysis and NEI analysis (Table 8). Mechanical properties of the steel are given in Tables 9 and 10. The material shows good impact properties together with a yield strength of approximately 500 MPa although tensile tests have shown the plate to have a yield of 550 MPa (108). No manufacturers data is available for the effect of welding.

3.2 Welding Procedure

Welding of the plates was done using a 3 kJ/mm heat input. The high heat input was chosen to simulate the type of heat inputs used in platform construction. Welding was performed on an ESAB submerged arc welding machine with a moving head. Plate preparation involved grinding one face to a 35° angle to produce a half K joint (Figure 26).

The heat input was obtained using the following parameters:-

Welding Process	Submerged arc (SAW)
Joint Type	Butt
Bevel	Half K V = 35°
Back Gouging	Yes
Filler Wire	SD3Mo 4 mm: C 0.05 - 0.08, Mn 1.2 - 1.5, SiO 0.25 - 0.4, Mo 0.5
Flux	OP121TT 15% (SiO ₂ + TiO ₂), 40% (CaO + MgO), 20% (Al ₂ O ₃ + MnO), 25% CaF ₂
Pre-Heat	None
Interpass	100°C
Current	500 A
Voltage	30 V
Weld Speed	300 mm/min.
Heat Input	3 kJ/mm
PWHT	None

The submerged arc machine was also used to place a single bead onto the Superelso 500 plate. Three heat inputs were used in order to give a range of HAZ. The heat inputs were obtained using the following parameters:-

	1.5 kJ/mm	3.0 kJ/mm	6.0 kJ/mm
Welding Process	SAW	SAW	SAW
Joint Type	Bead	Bead	Bead
Filler Wire	SD3Mo	SD3Mo	SD3Mo
Flux	OP121TT	OP121TT	OP121TT
Pre-Heat	None	None	None
Interpass	N/A	N/A	N/A
Current	537 A	507 A	519 A
Voltage	28.75 V	29.75 V	29.0 V
Weld Speed	600 mm/min	294 mm/min	147 mm/min
Heat Input	1.5 kJ/mm	3.1 kJ/mm	6.1 kJ/mm
PWHT	None	None	None

3.3 Fatigue and Corrosion Fatigue Testing Machines

Both in air and corrosion fatigue testing were carried out on purpose built three point loading machines. The in air testing machine was an electro-hydraulic machine whereas the corrosion fatigue machine was an electro-mechanical machine. Both machines operate in the same way except in the load source.

Control of the hydraulic machine is obtained by means of a load-controlled closed loop servo control system. A sinusoidal output signal is fed from a function generator, through an error measuring circuit to the servo amplifier which is connected to the servo hydraulic control valve which is used to control the flow rate and direction of the hydraulic fluid to the hydraulic ram.

When hydraulic pressure is applied to the ram the ram moves until resistance to movement is met. Once resistance is encountered a load is applied, the magnitude of which is registered by the load cell. Under load the load cell produces an output voltage which is amplified and fed to the error measuring circuit. The output from the error measuring circuit is used to reduce the input of the servo amplifier. The reduced input is amplified and loading continues until the output signal from the load cell cancels that of the function generator.

The electro-mechanical machine works on the same principle except that instead of a hydraulic control valve controlling the load an electric motor is used but the control systems and feed back are the same.

Both fatigue machines produce excellent load control but the electro-mechanical machine is limited to a maximum frequency of 1.0 Hz because of the time required to mechanically load and unload the system, especially with larger crack lengths where the compliance is large.

3.4 Constant ΔK Fatigue Machine

The constant ΔK fatigue machine was an electro mechanical machine similar to the machine used for corrosion fatigue testing. In addition to the electronics used to run the machine with a sinusoidal waveform, a computer was linked, to attenuate the output signal from the fatigue machine such that a constant ΔK could be applied. The constant ΔK was maintained by reducing both the amplitude and the mean load. A schematic diagram of the machine is shown in Figure 27.

3.5 Test Specimens

The specimens from the parent plates were cut such that the longitudinal direction of the specimen was parallel to the plate rolling direction. This enabled the notch to be positioned such that the fatigue crack would propagate in the short transverse direction (Figure 28).

Specimen geometry was in accordance with BS6835:1988 (109) as shown in Figure 29. All specimens were machined such that the specimen width was 24 mm. In the case of the Superelso plate and the Superelso welded joint, specimens were taken from both the top and the bottom of the plate.

Notch lengths of all constant load specimens were 8 mm whereas for the constant ΔK tests the notch length had to be limited to 2 mm in order that a pre-fatigue crack could be introduced.

3.6 Corrosion Fatigue Test Environment

All corrosion fatigue tests were carried out in a circulating synthetic sea water system (Figure 30). The sea water was made up by the addition of 360 g of synthetic sea water salt, produced by BDH, to 10 litres of distilled water. The composition of the sea water is given in Table 11 (110). The water was maintained at a constant 9°C by a water cooler and the pH was approximately 8. The pH of the solution was checked weekly and the sea water changed if the pH was outside the range 8 ± 0.5 .

The tank in which fatigue took place had a capacity of 6.75 litres and the water was circulated from a 60 litre supply at the rate of 1.7 litres/min. Full aeration was ensured by the use of an aerating block and air pump in the testing tank. Crack length measurement was possible with the use of Perspex windows in the fatigue tank which allowed direct observation of the fatigue crack and enabled crack measurement to be made with a travelling microscope.

3.7 Cathodic Protection

In order to vary the level of protection and because of the small specimen size, requiring a very small sacrificial anode, an impressed current system was used.

The protection apparatus consists of a stable power supply, a saturated calomel electrode (SCE), a platinum anode and a digital multimeter. Connection of the apparatus was as shown in Figure 31. In all cases, unless otherwise stated, the potential quoted will be referenced to the SCE.

It was necessary when the circuit was initially set up that the system was monitored regularly whilst polarization occurred. Once the system stabilised it was found that only small adjustments needed to be made to the voltage at infrequent intervals.

Pre-fatigue cracked specimens were subjected to the appropriate level of protection for a two week period prior to testing to allow any calcareous deposits to build up. Immediately prior to testing the specimen was placed on the anvils and a small tensile load of approximately 0.25 kN was applied to the specimen and the potential allowed to stabilise before fatigue testing commenced.

3.8 Pre-Fatiguing

All specimens contained an initial machined notch. However, it was necessary to introduce a sharp starting crack into the specimen which was done by a pre-fatiguing operation.

Pre-fatiguing was performed in air using a frequency of 18-20 Hz and a load ratio (R) of 0.1. It was necessary to initiate the crack using high loads but once the crack had initiated it was possible for the loads to be gradually reduced. Care was taken to ensure that after each load reduction the crack was allowed to grow a distance twice that of the previously induced plastic zone size. The final pre-fatiguing loads were chosen such that they were lower than the testing loads to be used, this ensured that the actual fatigue test would not be influenced in the initial stages of crack propagation by the plastic zone left by the pre-cracking.

All pre-fatigue crack lengths extended at least 3 mm from the start of the machined notch. After pre-fatiguing, all specimens not immediately used for testing, were kept in a desiccator prior to testing.

3.9 Constant Load Fatigue Tests

3.9.1 In air

In air fatigue testing was performed on an electro-hydraulic fatigue machine of the type previously described.

It was necessary for the specimen to be positioned carefully on the anvils such that when the ram was raised it was centrally positioned behind the notch. Once positioning was completed the static load was gradually increased to the mean level. The amplitude was then applied. Minor adjustments were made to the static and amplitude controls to obtain

the required loads of 8.4 kN ($P_{\max}$) and 5.0 kN ($P_{\min}$) giving a stress ratio of 0.6. The machine loads only required periodic adjustment to compensate for drift and the loads were maintained to ± 0.1 kN. The cyclic frequency used was 0.5 Hz.

The crack length was measured using a travelling microscope with a magnification of $\times 8$. The actual measurement was read from a Vernier scale on the side of the microscope. The length was found by moving the eye-piece until the cross wire was positioned at the crack tip when the crack was fully open. It was possible to do this because of the low frequency used. Crack length readings were taken every 2000 cycles at the start of the test but as the crack progressed it became necessary to increase the frequency of crack length readings. The accuracy of the crack length measurements using the Vernier scale was ± 0.005 mm. Crack length measurements were also taken using a strain gauge system in which the crack grew through a gauge mounted onto the specimen using an epoxy resin. The gauge output was connected to a y-t chart recorder to give a permanent crack length measurement. The monitoring circuit is given in Figure 32.

3.10 Corrosion Fatigue Testing

3.10.1 Free Corrosion

In the free corrosion fatigue tests the electro mechanical fatigue machine was used. Specimen positioning was done such that the centre anvil was positioned directly behind the notch. The static load and amplitude were applied as for the in air testing. Loads were again set at 8.4 kN and 5.0 kN with a testing frequency of 0.5 Hz. Crack length readings were measured optically and with strain gauges as for the in air tests.

3.10.2 Protected Samples

Specimens were positioned on the anvils and a small load of 0.25 kN applied. The specimen was connected into the protection circuit and the voltage allowed to stabilise. For the -800 mV protection level stabilisation occurred within an hour but for the -1100 mV protection level stabilisation took up to six hours. Once a stable voltage was obtained the static load and amplitude were applied. Two potentials were used in this investigation, -800 mV and -1100 mV. These potentials represent protection levels where corrosion is practically stopped but calcareous deposits are not present to a great extent and a level where over protection is applied and calcareous deposits rapidly build up.

The loads applied and cyclic frequency were as for the in air tests.

The crack lengths were measured only using a travelling microscope. In the over-protected case it was necessary to remove some of the calcareous deposits around the crack. This was achieved using a 1200 grade silicon carbide paper.

3.11 Constant ΔK Tests

Constant ΔK tests were performed on small scale test pieces similar to those used for the in air and corrosion fatigue tests. Accurate crack length measurement was achieved using a continuous gauge developed at Cranfield by Tidswell (111). The gauge consists of a backing of plastic onto which nickel chrome is deposited using a coating system. This is followed by the deposition of copper bus bars on to the NiCr. Figure 33 shows a schematic representation of a gauge mounted onto a specimen.

Calibration of the gauges was performed by growing a fatigue crack through a gauge and recording the resistance and the corresponding crack length, which was measured by a microscope with a magnification of x400.

Computer software was written to read the crack length. Using this crack length the loads required to give a stress intensity range of $20 \text{ MPa}\sqrt{m}$ were calculated assuming a stress ratio of 0.1 was to be used. The computer then gave an output voltage in the range 0 to 255 millivolts which was proportional to the required mean load and amplitude. The applied loads were then checked by the computer by reading the load cell voltage and any alterations were made to ensure the ΔK was maintained at $20 \text{ MPa}\sqrt{m} \pm 0.5 \text{ MPa}\sqrt{m}$. There may be a small error in both measuring the crack length and applying the loads but these should be limited such that an error of $\pm 0.5 \text{ MPa}\sqrt{m}$ is possibly an overestimate of the errors. The machine was set such that a frequency of 0.2 Hz was used. The computer stored both the number of cycles and corresponding crack length for later evaluation.

3.12 Crack Propagation Plots

3.12.1 Constant Load Tests

For both in air and corrosion fatigue experiments the following procedure was used to evaluate the fatigue crack growth.

Using the crack length data together with the corresponding number of cycles a curve was produced (Figure 34). Using this curve the crack growth rate was obtained by taking the tangent to the curve at points along the curve. The growth rate was calculated as the gradient of the tangent. The use of the tangent to the curve technique is not specified in

BS6835, which uses a statistical average value for calculating crack growth rate, but was considered both accurate and reliable as a technique for calculating crack growth rates where limited data were available such as during fatigue cycling through the night.

Using the crack length at the points where the tangents were evaluated, the stress intensity range was calculated using the equations given in BS6835:1988 for single edge notch specimens.

$$\Delta K = \frac{6\Delta P f\left(\frac{a}{w}\right)}{BW^{\frac{1}{2}}}$$

Where ΔP is the load range, B is the specimen thickness, W is the specimen width and $f\left(\frac{a}{w}\right)$ is a compliance function the equation of which is:-

$$f\left(\frac{a}{w}\right) = 1.93\left(\frac{a}{w}\right)^{\frac{1}{2}} - 3.07\left(\frac{a}{w}\right)^{\frac{3}{2}} + 14.53\left(\frac{a}{w}\right)^{\frac{5}{2}} - 25.11\left(\frac{a}{w}\right)^{\frac{7}{2}} + 25.80\left(\frac{a}{w}\right)^{\frac{9}{2}}$$

Using the calculated growth rate and stress intensity range, graphs of log growth rate against log stress intensity range are plotted.

3.12.2 Constant ΔK Tests

Due to the nature of a constant ΔK test it is not possible for the typical fracture mechanics type graph to be plotted. In the case of the constant ΔK tests the crack length, a, against number of cycles, N, curve was plotted. This enables any variations in growth rate to be readily seen as a change in gradient.

3.13 Metallography

It was necessary for metallographic examination that the fracture surface be removed from the bulk of the material. This was achieved by sawing 5 mm behind the fracture surface. The fracture surface was then sawn in half before being ultrasonically cleaned in gramisol. The surface was then dried and mounted in bakelite such that the profile of the central region of fracture surface was visible. Specimens were polished in the standard way before being etched in 2% Nital.

3.14 Fractography

All fracture surfaces were examined using a Cambridge Instruments Stereoscan 250 MkIII scanning electron microscope (SEM).

As in the section above the fracture surface was removed from the bulk of the specimen. In the case of the in air specimens the surface was ultrasonically cleaned before being dried and mounted on a SEM stub. For the specimens from corrosion fatigue experiments the surface was first washed under running water for 20 minutes followed by drying. The surfaces were then ultrasonically cleaned in distilled water followed by drying and finally ultrasonically cleaned in isopropyl alcohol and dried before being mounted on a SEM stub. This procedure was to ensure that all unattached debris and soluble salts were removed.

In the case of the over protected specimens it was found necessary to remove the calcareous deposits found on the fracture surface. This was achieved by the use of inhibited hydrochloric acid.

4. Results

4.1 Crack Growth Rate Data for Superelso 500 Parent Plate

The results for the Superelso 500 parent plate are shown in figures 35 to 38. These figures are logarithmic plots of crack growth rate ($\frac{da}{dN}$) versus stress intensity factor range (ΔK). Figures 36 to 38 also show the in air fatigue best fit line which serves as a reference line. This enables the effects of environment to be more clearly seen.

It was seen that there was a variety of responses to the electro chemical potential exhibited by the parent plate Superelso 500:-

1. The in air data, which serves as a reference, is adequately described by the Paris equation:

$$\frac{da}{dN} = 1.72 \times 10^{-12} \Delta K^{3.44}$$

and it is this Paris equation which is used for the subsequent comparisons.

2. Under free corrosion conditions the early fatigue crack growth was seen to be independent of ΔK . At a ΔK value of between 17 and 18 $\text{MPa}\sqrt{\text{m}}$ the growth rate was seen to accelerate and follow the more typical fatigue crack growth curve. This region of growth is described by the in air Paris relationship.
3. The application of a cathodic potential of -800 mV returns the low ΔK fatigue crack growth rates to values typical of those obtained in air. The Paris equation however was different to the in air equations and this showed that in the high ΔK region the cathodically protected specimen had lower crack growth rates than the in air specimen.
4. The application of a cathodic potential of -1100 mV increases the crack growth rate by up to a factor of ten at low ΔK values. The crack growth rate shows no dependence on the ΔK value on the fatigue crack growth rate and because the growth rate remained constant the crack growth rate enhancement, compared to the in air data, decreased as the ΔK value increased.

One specimen was tested with side covers on the sample to restrict the ingress of the environment. This was seen to have a limited effect although it would appear that the growth rate at low ΔK values is affected, the growth rate being slightly lower for the specimen with the side covers.

4.2 Crack Growth Rate Data for Superelso 500 HAZ

Figures 39 to 42 are the results for the Superelso 500 HAZ. As for the parent plate, the figures are logarithmic plots of $\frac{da}{dN}$ versus ΔK .

For comparison, the parent plate in air Paris relationship is included on all figures.

As was seen with the parent plate, a variety of responses to the electro-chemical potential were exhibited by the HAZ samples but these responses were not always the same as those shown by the parent plate.

1. The in air HAZ fatigue data did show a general trend for the crack growth rate to increase with increasing ΔK , although in one sample there was a sharp decrease in growth rate at a ΔK of $17.5 \text{ MPa}\sqrt{m}$. The growth rate reached a minimum at a ΔK of $18 \text{ MPa}\sqrt{m}$ before the growth rate rapidly returned to the values obtained from the other sample.

Comparison with the in air parent plate reference line shows that the HAZ exhibits similar crack growth rates at low ΔK values but at a high ΔK values the HAZ growth rate is lower than the parent plate in air. The Paris equation for the HAZ in air is:-

$$\frac{da}{dN} = 1.9 \times 10^{-12} \Delta K^{3.34}$$

but this is not an accurate description of the crack growth particularly in the ΔK region between 17 and $19 \text{ MPa}\sqrt{m}$.

2. Under free corrosion conditions the fatigue crack growth rate shows no dependence on the applied ΔK for the entire range of ΔK investigated. There was a variation in the value of the growth rate observed in different samples. It was therefore not possible to use a Paris relationship to describe the crack growth.
3. The application of a cathodic potential of -800 mV is seen to restore the fatigue crack growth to a linear relationship between $\log\left(\frac{da}{dN}\right)$ and $\log(\Delta K)$ with the in air parent plate Paris equation describing the relationship between crack growth rate and stress intensity factor range.
4. Application of a cathodic potential of -1100 mV results in enhanced crack growth rates which show little dependence on the applied ΔK .

4.3 Fatigue Crack Growth Data for Mannesman X70 HAZ

Figures 43 to 45 are the results for the Mannesman X70 HAZ at different electrochemical potentials. As for the Superelso 500 results the figures are logarithmic plots of $\frac{da}{dN}$ versus ΔK .

For comparison, the reference line is the Paris relationship obtained by Callister (108) from a Mannesman X80 parent plate.

1. Under free corrosion conditions the early fatigue crack growth is seen to be independent of ΔK and the growth rate was constant. At a ΔK of $17 \text{ MPa}\sqrt{\text{m}}$ the crack growth rate accelerates and follows a Paris relationship, the Paris equation being:-

$$\frac{da}{dN} = 3.06 \times 10^{-11} \Delta K^{2.33}$$

For all values of ΔK the crack growth rate for the free corrosion condition is lower than for the parent plate.

2. The application of a cathodic potential of -800 mV increases the free corrosion crack growth rates and returns them towards the values obtained from the parent plate. The Paris equation for the crack growth under a cathodic potential of -800 mV is:-

$$\frac{da}{dN} = 6.18 \times 10^{-11} \Delta K^{2.25}$$

This differs from that in air in the value of the slope m . Thus, at low values of ΔK the in air crack growth rates are lower than the cathodically protected crack growth rates but above $18 \text{ MPa}\sqrt{\text{m}}$ the crack growth rates of the cathodically protected specimen are lower than the in air rates.

3. The application of a cathodic potential of -1100 mV enhances the crack growth rates. There is a difference in the degree of enhancement between the two samples. One sample shows no dependence of ΔK on the crack growth rate and this sample has the greatest enhancement. The second sample shows some

dependence on the ΔK value and it is possible to use a Paris relationship to describe the crack growth rate at values of $\Delta K > 14 \text{ MPa}\sqrt{\text{m}}$. The Paris relationship for this sample is:-

$$\frac{da}{dN} = 7.77 \times 10^{-10} \Delta K^{1.80}$$

The growth rates for the two specimens converge at $\Delta K = 22 \text{ MPa}\sqrt{\text{m}}$.

4.4 Fracture Surface Analysis

4.4.1 Superelso 500 Parent Plate

The fracture surfaces of the specimens were examined using an SEM. The surfaces were examined initially with any surface deposits present to enable analysis of the surface deposits. These deposits were subsequently chemically removed to reveal the underlying fracture surface. Figures 46 to 54 show typical fracture surface regions taken from the Superelso 500 parent plate specimens under different cathodic potentials.

The in air fracture surface, figures 46 and 47, shows a mixed mechanism of crack propagation. There are two mechanisms observed, ductile decohesion and striation formation. Striation formation is seen to occur in small regions similar in size to the measured grain size. The striation spacing is very much larger than the macroscopic crack growth rate observed, this may be as much as two orders of magnitude in extreme cases. In the regions where striations are not present, the mechanism of crack propagation is ductile decohesion. It is in these regions that the larger secondary cracks are observed. Smaller secondary cracking is seen as a direct consequence of striation formation but the crack opening and number of these secondary cracks is limited.

Under free corrosion conditions the fracture surface again shows a mixed mechanism of crack propagation, figures 48 and 49. As observed in air, the two main mechanisms are striation formation and ductile decohesion. There are also a few very isolated regions where the crack propagated intergranularly rather than transgranularly, which is more typical. The fracture surface showed regions of corrosion debris. Analysis of the corrosion debris using an energy dispersive analysis system, which was attached to the SEM, shows that the principal element present was iron. Since the analysis system is not able to identify elements with atomic numbers less than 6 it can only be surmised that the compound will be iron oxide or iron hydroxide.

With a cathodic potential of -800 mV the fracture surface, figures 50 and 51, is seen to be coated with a layer of white deposit. Analysis of the deposit shows that magnesium is the principal constituent. It is possible to image the fracture surface through this deposit and this implies that the deposit is only a few microns thick because the SEM uses electrons generated just below the surface. Other deposits are observed on the fracture surface and analysis shows these to be the same as is seen on the free corrosion fracture surface, namely iron hydroxide.

The mechanisms of failure are the same as is seen on the in air fracture surface, i.e. ductile decohesion and striation formation. The fracture surface also shows regions of what appeared to be corrosion pits.

The application of a cathodic potential of -1100 mV results in a thick, tightly knit deposit, figure 52, which is found to contain principally magnesium and from analysis this is magnesium hydroxide. The magnesium hydroxide is formed in preference to calcium hydroxide due to the composition of the sea water solution used which contains five times more magnesium than calcium. The deposit is seen to decrease in thickness towards the crack tip. It should be noted that this deposit is not entirely typical of real offshore calcareous deposits and it is quite possible that the calcareous deposit chemistry obtained in real situations may differ.

Removal of the deposit using an inhibited HCl solution reveals that the mechanism of crack propagation is different from that observed in the other specimens. The crack propagates by a mixed mode and ductile decohesion is still evident but the primary mechanism would appear to be hydrogen induced quasi-cleavage, figure 53, such as may be seen in stress corrosion cracking. The cleavage facets are found to be similar in size to the grain size. Secondary cracking is not visible to a great extent on the fracture surface although some stress corrosion cracking is in evidence.

Increasing the stress intensity factor range shows that there was a decrease in the quantity of quasi-cleavage and an increase in the proportion of ductile decohesion, figure 54.

4.4.2 Superelso 500 HAZ

Figures 55 to 62 show the fracture surfaces taken from different regions on the Superelso 500 HAZ specimens under different cathodic potentials.

As is seen from the fracture surface of the parent plate tested in air, figures 55 and 56, the principal mechanism of failure is ductile decohesion. However, on the HAZ samples there are fewer regions of fatigue striations. Secondary cracking is a feature of the fracture surface and is seen to increase both in density and severity as the crack length increases.

Under freely corroding conditions, the fracture surface shows no unusual features although the number of secondary cracks does appear to be greater than the in air fracture surface, figures 57 and 58. The main mechanism of crack propagation is ductile decohesion and there is evidence of corrosion debris, i.e. iron hydroxide, on the fracture surface.

With a cathodic potential of -800 mV the HAZ sample is seen to have a thin deposit of magnesium hydroxide on the surface. The fracture surface, figures 59 and 60, shows few striations and, as is the case for the other HAZ samples, the mechanism for crack propagation is ductile decohesion. Also visible on the fracture surface are corrosion pits and corrosion debris, i.e. iron hydroxide.

The application of a potential of -1100 mV results in a thick deposit of magnesium hydroxide. Removal of this compound reveals that, unlike the parent plate, quasi-cleavage does not appear to be a mechanism by which the crack propagates, figures 61 and 62. Instead the mechanism appears to remain as ductile decohesion although from the texture of the fracture surface, the general ductility has reduced. Secondary cracking is observed but in comparison to the other cathodic potentials the number of secondary cracks is greatly reduced.

4.4.3 Mannesman X70 HAZ

The fracture surfaces for the Mannesman X70 HAZ taken under different cathodic potentials are shown in figures 63 to 68.

The fracture surface of the freely corroding specimen is similar in appearance to that obtained from the Superelso 500 HAZ, figures 63 and 64. The mechanism of crack propagation is ductile decohesion but no striations are visible. Secondary cracking is seen to occur at both high and low ΔK values although the density of this secondary cracking is not as high as is seen on the Superelso 500 HAZ specimens.

When a cathodic protection of -800 mV is applied, a thin deposit of magnesium hydroxide is observed on the fracture surface. Other surface deposits are observed and these are analysed as containing iron and for the reasons stated previously it can be surmised that

the deposit will be either iron oxide or iron hydroxide. No corrosion pitting is seen. The mechanism of crack propagation is ductile decohesion and, as is the case for the free corrosion specimen, there is an absence of striations, figures 65 and 66.

With the application of a cathodic potential of -1100 mV a thick layer of magnesium based compound is formed. As is the case for the previous fracture surfaces, this deposit is chemically removed. The underlying fracture surface shows no evidence of quasi-cleavage, instead, the fracture surface is similar to the Superelso 500 HAZ over protected sample, with a ductile decohesion mechanism of crack propagation, figures 67 and 68.

4.5 Constant ΔK Testing

Due to the nature of the constant ΔK test it is not possible to use the logarithmic plots used for the rising ΔK tests. Instead, the crack length is plotted against the number of elapsed cycles. It is then easy to observe any obvious changes in the crack growth rate.

Three different heat inputs were used to produce different microstructural gradients but for all tests the crack was grown from the weld metal through the HAZ and into the parent plate.

Typical microstructural regions of the weldment obtained when a single weld run is placed on a base plate of Superelso 500 are shown in figures 69 to 72.

4.5.1 1.5 kJ/mm Heat Input

Figure 73 shows the crack length versus number of cycles for the 1.5 kJ/mm heat input.

The early stage of crack growth in the weld metal occurs at a constant crack growth rate of 9×10^{-9} m/cycle. Continued crack growth sees the crack enter the coarse grained region and this change in microstructure is accompanied by a large change in crack growth rate to 5×10^{-7} m/cycle. The accelerated crack growth lasts for approximately 50 μm before the crack slows down to 2.2×10^{-8} m/cycle.

Figure 74 shows the edge profile of the region where the accelerated crack growth is observed. It is seen that the crack followed the aligned carbides of this region, even though this means a crack path which is not perpendicular to the applied stress.

The crack growth rate remains constant through the remainder of the HAZ. On leaving the HAZ and entering the parent plate the crack growth is seen to accelerate to a growth rate of 8.6×10^{-8} m/cycle at which it then remains constant.

Figure 75 shows the crack growth rate observed for different distances from the fusion line. Also shown on this figure is the hardness profile of the regions through which the crack is grown. This figure shows that the crack growth rate of the coarse grained region is exceptionally rapid but is a highly localised effect.

4.5.2 3.1 kJ/mm Heat Input

Figure 76 shows the crack length against the number of cycles for the 3.1 kJ/mm sample. Early crack growth occurs in the weld metal and a constant growth rate of 2.1×10^{-8} m/cycle is observed. As the crack entered the HAZ the crack length is seen to increase and maintain a new constant crack growth rate of 4.2×10^{-8} m/cycle. However, unlike the 1.5 kJ/mm sample there is no rapid increase in crack growth associated with the coarse grained HAZ. The crack growth rate remains constant through the HAZ but is seen to increase to 1×10^{-7} m/cycle as the crack enters the parent plate microstructure.

Figure 77 shows the crack growth rates observed for different distances from the fusion line together with the hardness profile. This shows clearly that different regions of the weldment have different crack growth rates associated with them.

4.5.3 6.1 kJ/mm Heat Input

The crack length against the number of cycles for the 6.1 kJ/mm sample is shown in Figure 78. As was seen in the 3.1 kJ/mm sample the crack growth rate in the weld metal is constant, the crack growth rate being 3×10^{-8} m/cycle.

The crack growth rate is seen to increase as the crack propagates into the HAZ. As with the 3.1 kJ/mm sample there is no rapid increase in crack growth rate associated with the coarse grained region. The crack growth rate remains constant at 1.3×10^{-7} m/cycle in the HAZ and only increases as the crack enters the parent plate material. The growth rate then remains constant at 2.3×10^{-7} m/cycle.

Figure 79 shows the crack growth rates observed for different distances from the fusion line together with the hardness profile. As was seen for the 3.1 kJ/mm sample, different regions have different crack growth rates associated with them.

5. Discussion

5.1 Fatigue Crack Propagation

Fatigue crack propagation was investigated under a number of different conditions, the main aim being to assess the selected material's suitability for future offshore use in the exploitation of natural fuels. The North Sea is a particularly aggressive environment and the combined loading associated with the wind and waves together with the harsh environment would indicate that corrosion fatigue would be a potential failure mechanism.

It is possible to reduce the extent of corrosion by cathodic protection but the protection on an offshore structure will never be uniform. Hence there are likely to be regions close to the anodes which are over protected and regions where the protection does not reach. For this reason different levels of potential were investigated. For completeness, and for a reference against which the effects of the aqueous environment could be gauged, the fatigue crack propagation under atmospheric conditions was investigated.

5.2 In Air Fatigue Crack Propagation

5.2.1. Superelso 500 Parent Plate

The Superelso 500 parent plate data showed some scatter, figure 35. Though not excessive, the scatter did result in variations in crack growth rate. These variations were limited to a factor of two and were probably caused by variations in microstructure and material properties induced in plate manufacture.

Four specimens were used to generate the data and from Figure 35 it was seen that three of the four specimens had similar crack growth rates for given ΔK values. The one test which had different crack growth rates compared to the others had a longer initial crack. All other conditions were the same and there was no obvious external reason for the slightly enhanced crack growth rates produced by the one specimen.

The crack profile of the Superelso 500 parent plate, taken on sections normal to the specimen thickness, showed the fatigue crack to have taken a transgranular path through the ferritic regions. The ferritic regions have the lowest flow stress of the microstructures present in these steels and will therefore offer the least resistance to the advancing crack. On a macroscopic level grain boundaries did not appear to have a significant influence on the crack propagation. Examination of the specimen which gave the enhanced crack growth rate showed that secondary cracking was present but the crack path was less tortuous than

that observed from the other specimens. Thus on a microscopic level the crack growth rates for all the Superelso 500 parent plate may have been the same, but the shorter crack path in one sample would make the macroscopic growth rate appear faster.

As a general trend secondary cracking was seen to increase both in the number of secondary cracks per unit area, and in the degree of crack opening of the secondary cracks as the value of ΔK increased, this has been reported previously. The ability of a material to form secondary cracks has also been associated with improvements in the fatigue crack growth resistance of high strength steels (112).

The fractographic appearance of the steel (Figures 46 and 47) showed that ductile decohesion features dominated the fracture surface. In addition, there were many regions of fatigue striations. The clearness and ease with which these areas of striations were found is unusual for steels and must be a consequence of the steel cleanliness, the homogeneous microstructure and the methodical approach employed in the search for striations. A comparison between the striation spacing and the measured crack growth rate showed the striation spacing to be at least an order of magnitude greater than the measured growth rate. High magnifications revealed that the striations had finer striations on them. These fine striations still gave higher growth rates than the measured macroscopic growth rates which agrees with the findings of Delhovren (113).

5.2.2 In Air Fatigue Crack Propagation of Superelso 500 HAZ

The fatigue data for the Superelso 500 HAZ (Figure 39) fatigue tested in air were described by a Paris relationship. The Paris relationship for the HAZ differs from that for the parent plate. At ΔK values below $15 \text{ MPa}\sqrt{\text{m}}$ the HAZ had faster crack growth rates whilst at ΔK values above $15 \text{ MPa}\sqrt{\text{m}}$ the HAZ had lower crack growth rates than the parent plate. Metallography of the crack profile showed the fatigue crack had been positioned such that propagation had progressed through a range of microstructures. This was not caused by bad notch positioning but was a consequence of trying to produce a straight HAZ whilst using multipass welding to join the two sections. Although there was a range of microstructures through which the fatigue crack propagated there was no conclusive evidence as to the effect of microstructure on the crack growth rate. In the coarser grained regions the crack path was straight and secondary cracking was limited. When secondary cracks did occur propagation usually extended through three or more grains. In the finer grained regions secondary cracking occurred more frequently than in the coarse grained regions, but the secondary cracks usually extended through only one or two grains before stopping.

Fractography revealed that for the range of microstructures encountered the dominant mechanism of failure was ductile decohesion (Figures 55 and 56). Unlike the parent plate material striations were very hard to find, but where they did occur they extended across a grain before the mechanism returned to ductile decohesion. Generally the HAZ fracture surface had a finer texture than that seen on the parent plate fracture surface.

The fatigue data from one sample did exhibit an unexpected drop in crack growth rate between ΔK values of 17 and 18 $\text{MPa}\sqrt{\text{m}}$. Continued fatigue loading saw the growth rate reach a minimum growth rate of 1×10^{-8} m/cycle before the crack accelerated. At a ΔK value of 18.5 $\text{MPa}\sqrt{\text{m}}$ the growth rate had returned to the anticipated levels of crack growth rate. The test was completed with no further growth rate fluctuations.

Metallography showed that the fatigue crack had been travelling through a region of heat affected weld metal when the growth rate changed. This region extended approximately 0.5 mm which was far greater than the distance over which the growth rate change had occurred. Fractography gave no indication as to the cause of the decrease in growth rate. This lack of evidence as to the cause of this sudden decrease in crack growth rate is possibly best explained by a small overload having been applied, which would temporarily reduce the crack growth rate as was seen. Overloads are thought to blunt the crack tip causing the crack to reinitiate in order to continue propagating (114).

5.3 Corrosion Fatigue

Fatigue is a complex problem which can be further complicated by the addition of a very aggressive environment such as sea water. Carbon steels are subject to aqueous corrosion in such an environment but the extent of this corrosion can be reduced to acceptable levels by the use of cathodic protection methods. The use of cathodic protection on offshore structures is widespread (115) but current densities are never uniform. Hence it is very likely that not all regions will be subjected to the measured level of applied potential.

5.3.1 Superelso 500 Parent Plate Under Free Corrosion Conditions

The fatigue data for the Superelso 500 parent plate fatigue tested under free corrosion conditions (Figure 56) showed a different trend than the parent plate tested in air. Initially the fatigue crack growth rate was independent of the applied stress intensity range, this independence continued until a ΔK value of 17 $\text{MPa}\sqrt{\text{m}}$ was applied. At ΔK values greater than 17 $\text{MPa}\sqrt{\text{m}}$ the growth rate followed a typical Paris relationship.

The crack growth data for ΔK values greater than $17 \text{ MPa}\sqrt{m}$ fell between the extremes of the in air data. The in air Paris equation would be a good representation of the crack growth rate for a given ΔK .

Examination of the fracture surface obtained from the Superelso 500 parent plate, fatigue tested in sea water with no cathodic protection, showed that the dominant mechanism for crack advance had been ductile decohesion (Figures 48 and 49). Striations were also found in many areas and the spacing was seen to be much greater than the measured macroscopic growth rate. The regions of striations were seen in areas where the crack growth rate appeared to be independent of the applied ΔK , and also in regions where the crack growth rate showed a dependence upon the applied ΔK . The fractographs also showed that secondary cracking was a major factor at both low and high ΔK values.

The frequency with which secondary cracking occurred was also visible from metallographic examination. The secondary cracks usually originated from the centre of a ferritic grain but were terminated by grain boundaries. Usually the depth of the secondary cracks was limited to one or two grains.

5.3.2 Superelso 500 HAZ Under Free Corrosion Conditions

Figure 40 showed that at low ΔK values, from $10 \text{ MPa}\sqrt{m}$, the fatigue crack growth rate showed a similar response to the parent plate in that there appeared to be no dependence of crack growth rate on the applied ΔK . Unlike the parent plate this independence did not change to a ΔK dependence as the crack length increased. This crack growth rate independence over the entire range of ΔK investigated was unusual and has not been noted before in the literature. Both samples investigated gave the same trend although the crack growth rates exhibited by the two samples were different.

It was unclear from fractographic and metallographic observations as to the cause of this unusual behaviour. Fractography showed that secondary cracking occurred extensively over the fracture surface and the primary mechanism for crack advance was ductile decohesion. Metallography showed a small difference in grain size between the two specimens, this having been caused by the thermal cycle of the welding process. These differences, compared to the parent plate, would not appear to adequately explain the observed crack growth rates for a rising ΔK test.

The lack of obvious reasons for this unusual result leads to speculation as to the cause:

- (i) if the fatigue machine was malfunctioning such that the load applied was reduced compared to the load thought to be applied the applied load must be reducing continuously as the crack length increased. This is not thought to be the case because it is unlikely that two tests would behave exactly the same, and the subsequent test performed at a different potential, -800 mV, showed a "normal" fatigue crack growth rate.
- (ii) if premature crack closure occurred, caused by residual stresses induced during the welding process, the ΔK value applied would be reduced, this does not adequately explain the constant crack growth rate observed because it would require the $\Delta K_{\text{eff}} = (K_{\text{max}} - K_{\text{op}})$ to reduce as the crack length increased.

It would appear that the observed crack growth rate independence from ΔK is particular to this one steel HAZ, possibly only under the specific heat input used, and to base further discussion around this result may be misleading.

5.3.3 Mannesman X70 HAZ

The Mannesman X70 HAZ tested under free corrosion conditions (Figure 43) behaved almost identically to the Superelso 500 parent plate. Initially there was an independence of the crack growth rate to the applied ΔK , but at a ΔK of $16 \text{ MPa}\sqrt{\text{m}}$ the crack growth rate became dependent upon the applied ΔK and the crack growth rate followed a Paris type relationship.

Comparison between the Mannesman X70 HAZ and the free corrosion conditions and the HAZ tested in air showed that the X70 HAZ has superior crack growth resistance when freely corroding. The in air test data was taken from other work (108). Comparison between the parent plate compositions and mechanical properties of the two parent plates used to fabricate the welds showed that the compositions were practically identical, but the yield strengths were supposedly different. The previous work used material designated X80 with a yield strength of 550 MPa whilst for this work the material was X70 with a manufacturer's yield strength of 514 MPa. Tensile tests performed on X70 have shown the yield strength to be higher than the manufacturer's data and almost identical to the X80 values. Whilst it is acknowledged that there was a slight variation in yield strength, the similarity of both composition and microstructure would make the comparisons between the X70 and X80 HAZ tests valid.

Fractography of the X70 HAZ free corrosion sample for a low ΔK region (Figure 63) showed that secondary cracking was extensive, even more so than on the Superelso 500 parent plate. Ductile decohesion was the main mechanism by which the crack advanced but there were very few regions of striations. At higher values of ΔK , where the crack growth rate was dependent upon the applied ΔK , striations were present although they did appear to have been affected by the environment. This was thought to have occurred after the striations had been formed and the environment had attacked the surface.

Metallography of the fatigue crack showed that whilst the notch position was supposed to be located such that the crack propagation would be in the coarse grained region, the actual notch position meant that the crack propagated in a microstructure of finer grain size than the coarse grained region of a HAZ. This finer grained microstructure was smaller than the grain size of the parent plate and this would possibly account for the superior crack growth resistance exhibited by the HAZ samples. The secondary cracks seen branching from the main crack were much deeper than seen on the Superelso 500 parent plate samples and from the degree of opening and the bluntness of the secondary cracks it appeared that aqueous corrosion had occurred down the secondary cracks and may have helped terminate the cracks by blunting the tip.

5.4 Cathodic Protection effects

5.4.1 Potential of -800 mV

Cathodic potentials of about -800 mV are generally used for impressed current protection systems because they reduced the corrosion rates to acceptable levels but at potentials of -800 mV the conditions for hydrogen discharge are met but there is rarely a problem of enhanced crack growth rates at this potential (115).

The application of a cathodic potential of -800 mV to the Superelso 500 parent plate (Figure 37), HAZ (Figure 41) and Mannesman X70 HAZ (Figure 44) resulted in the fatigue crack growth rates, for given ΔK values, being similar to those of the in air tests.

Metallography and fractography showed that in general the features observed in air were present for the protected samples. The crack path was always transgranular and short secondary cracks were visible although they were present in far fewer numbers than seen on the free corrosion samples. Striations were again clearly seen in the Superelso 500 parent plate (Figures 50 and 51) but were not generally found in either of the steel HAZ's (Figures 59, 60, 65 and 66). The most noticeable feature was the large number of corrosion

pits visible on the fracture surface of the Superelso 500 parent plate. The fracture surfaces of the HAZ did not show this type of pitting although environmental attack was seen to have occurred in isolated regions. Generally all fracture surfaces showed that failure had been caused by ductile decohesion although the fracture surface from the Superelso 500 HAZ did contain some isolated regions of brittle intergranular cracking.

In order to explain the possible mechanisms occurring it will be more beneficial if comparisons are made with the free corrosion condition rather than the in air condition. In the free corrosion condition evidence of stress corrosion cracking was found, however, apart from a few isolated regions on the Superelso 500 HAZ, no such evidence was observed on the samples protected at -800 mV. In addition, the Paris plots for the two conditions do vary quite considerably. In the free corrosion condition there were regions in which the crack was found to grow at rates which appeared independent of the applied ΔK value. In contrast, the Paris plots for the -800 mV condition showed the expected trend of an increase in crack growth rate with increasing ΔK value. The lack of evidence of stress corrosion and the relatively constant m value suggests that the following mechanism may explain the effects of a cathodic potential of -800 mV on the fatigue crack propagation of a high strength steel.

As mentioned above, the choice of -800 mV as a potential was to reduce the corrosion rates to acceptable levels whilst avoiding the production of hydrogen. It was, however, probable that a calcareous type deposit would be formed on the specimen surfaces, including the surfaces produced down into the fatigue crack, because the water temperature was above 5°C (65) and was found to be the case. Calcareous deposits in this instance may be extremely beneficial. The quantity of deposit will be sufficient for only a thin layer of deposit to be formed on the specimen surface, but even a thin layer was sufficient for a significant reduction in applied current requirements to maintain the required protection levels. Additionally, because the deposit was extremely alkaline, being composed mainly of magnesium hydroxide in the case of this work, it was likely that hydrogen ions present in the sea water would have reacted with the deposit and as a consequence the pH level would have risen slightly due to the decrease in hydrogen concentration. Any additional hydrogen ions which remain in solution would have the added difficulty of having to travel through the deposit in order to get to the steel. It is clear, therefore, that any hydrogen which does get into the steel will be in such small concentrations that its embrittling effect will be limited.

Thus the fatigue of a steel protected to a potential of -800 mV can be considered to occur by the same mechanisms as observed in air. The observed corrosion pits are likely to have been caused after the crack front had propagated through, rather than whilst the crack was propagating through these regions. These corrosion pits do highlight the fact that corrosion is not fully prevented by cathodic protection at these levels of potential and the applied potential may not always protect every region.

It is worth noting that these tests were performed on small scale laboratory test pieces rather than realistic specimen sizes or configurations. Consequently, the crack front length and crack depth is small and will generally allow good penetration of the cathodic potential and the environment from both down the crack and from the sides of the specimens.

Previous work by Millar (116) has looked at the effects of side covers to restrict the access of the environment. Millar found that there appeared to be no difference between the samples with side covers compared to those without. However, the use of small scale laboratory specimens does mean that this result will be limited in its significance.

In real situations fatigue cracks are more likely to occur in stress hot spots (117). These fatigue cracks tend to be very long compared to their depth and the applied loading would not necessarily be confined to the simple case of mode I opening. In real situations the depth of penetration of the potential will be dependent upon the crack opening displacement and it is highly likely that situations will exist where the crack tip is closer to a condition of free corrosion. It is therefore necessary, even when protection is applied, that inspection of structures is maintained and any cracks found are dealt with in their infancy.

5.4.2 Cathodic Protection Effects at a Potential of -1100 mV

It was found that the use of a cathodic potential of -1100 mV causes the crack growth rates of the Superelso 500 parent plate, HAZ and Mannesman X70 HAZ to be greatly enhanced compared to the crack growth rates measured under other conditions (Figures 38, 42 and 45). Crack growth rate enhancement was most significant at measured ΔK values of approximately $12 \text{ MPa}\sqrt{\text{m}}$, this value being the approximate starting ΔK value. The general trend was for the over protected samples to maintain their initial crack growth rate up to high ΔK values where the data tends to merge with that obtained from the in air tests. Metallography has shown that the crack path was very straight and that secondary cracking was rare. The crack still travelled transgranularly through the ferrite as a general rule. Very little else was evident from the crack profiles.

The fracture surfaces obtained from the over protected samples were found to be covered by a thick layer of calcareous type deposits, analysed to be a magnesium based compound most probably magnesium hydroxide.

The Superelso 500 parent plate fracture surface (Figure 52) showed that the hydrogen induced quasi-cleavage was the major fracture feature observed at low ΔK values. These regions were similar in size to the material grain size. Increases in ΔK resulted in a reduction in the number of regions of quasi-cleavage and an increase in the amount of ductile decohesion present. Also evident were stress corrosion cracks. These cracks differ from the normally seen fatigue secondary cracks in that the longitudinal axis of the stress corrosion crack is normal to the longitudinal axis of a fatigue secondary crack. There was no applied stress acting to cause this form of cracking and it must be a consequence of the crack propagating on two levels causing a slight tensile component which under stress corrosion conditions was able to pull the material apart and form a crack.

Fracture surfaces obtained from the HAZ samples (Figures 61, 62, 67 and 68) did not show the quasi-cleavage which was predominant on the Superelso 500 parent plate. In general the features on the HAZ samples were finer and the Mannesman X70 HAZ sample showed regions of dimples which appeared similar to those obtained when microvoid coalescence is the failure mode.

A well known consequence of applying a cathodic over potential is the production of hydrogen and it was this hydrogen which was undoubtedly the cause of the enhanced crack growth rate observed.

It is worthwhile considering briefly how hydrogen is taken into the lattice. Initially there is a physical adsorption of hydrogen gas where uncharged molecules are adsorbed. These uncharged molecules then dissociate into atoms and then ions because of the strong crystalline field; this is known as chemisorption (118).

Hydrogen ions, as screened protons, are exceptionally mobile in lattices, and in a body centred cubic lattice, such as iron, diffusion of hydrogen is rapid. In the case of the pre-soak the driving force for diffusion would be the hydrogen concentration gradient and it would be expected that with a two week pre-soak equilibration of the hydrogen concentration in the specimen should have occurred. If a stress was then applied to the sample the hydrogen would tend to diffuse down the stress gradient and be found at the edge of the plastic zone (85). Diffusion in this case would be aided by mobile dislocations which effectively increases the apparent diffusivity.

Once diffusion had occurred embrittlement could occur. Hydrogen can embrittle high strength steels and cause slow crack growth at stress intensities below the critical stress intensity for fast fracture. The susceptibility to hydrogen induced slow crack growth has been shown to decrease with decreasing strength levels (Figure 20) (79).

However, under cyclic loading even lower strength steels have shown the enhanced crack growth rates caused by hydrogen embrittlement and it was not altogether surprising that the steels used in this work, having strengths in excess of 500 MPa, showed hydrogen embrittlement effects.

Consideration of the mechanisms of embrittlement makes elimination of some of the mechanisms possible, although it is worth noting that whilst some of the mechanisms are discounted in this work they may be perfectly valid for other applications.

As mentioned earlier, at low values of ΔK , whilst brittle quasi-cleavage was seen on the fracture surface of some samples, there were also regions of ductile failure. The probable cause of this mixture of ductile and brittle failure modes is that for cleavage type failure to occur it is necessary for appropriately orientated cleavage planes to be available. This may also explain why cleavage was not seen in the HAZ samples.

It would be reasonable to assume that there could be more than one mechanism operating over the entire crack front, the specific mechanism will depend on the local grain orientation. If the case of the brittle cleavage type failure is considered, it is possible that one of two mechanisms apply. These two mechanisms are however in direct opposition to each other.

Lynch (119) proposes that the major effect of hydrogen is to promote the injection of dislocations at the crack tip whereas Gilman (120) suggests that chemisorption would tend to prevent dislocation emission at the crack tip. In the case of fatigue in which slip at the crack tip is a major factor, it is thought that the Lynch model may account for the brittle failure more adequately than that of Gilman. If dislocations were injected, pile up would occur more rapidly. It is also possible that hydrogen reduces the energy required to cause cleavage, i.e. decohesion is occurring. These two factors will therefore promote the failure of the material by brittle crack growth.

The fracture surface appearance of the over protected samples in which a cleavage mechanism of crack advance occurred, together with a decohesion mechanism of crack advance, in which material ductility appeared to have been reduced, would support the above mechanism.

Decohesion may also account for the ductile failure modes seen and the enhanced crack growth rates present. In the case of ductile failure it is possible that if the material ahead of the crack had been sufficiently weakened then the crack increment would be larger for a given ΔK value.

At higher ΔK values cleavage was reduced in quantity and the crack growth rates did not show the same degree of enhancement compared to the in air data. This may be explained by the increase in plastic zone size requiring the diffusing hydrogen to travel further to reach the edge of the plastic zone during each crack growth increment. It may be deduced therefore that at higher ΔK values hydrogen had a reduced effect.

5.5 Sources of Error

The crack length, a , was measured using a travelling microscope. The accuracy of this technique was ± 0.005 mm. The number of cycles, N , was counted electronically and the error was negligible. The tangent to the curve technique used to calculate the $\frac{da}{dN}$ values was the greatest source of error but a curve could be drawn through the data points with a reasonable level of accuracy and the resulting errors would be no greater than ten percent of the quoted values. The errors quoted above were far outweighed by the variations in growth rate caused by the material.

The prefatiguing operation ensured that the starter crack was very sharp. The plastic zone size introduced by the prefatiguing operation was smaller than that introduced by the fatigue test loads. Allowing the prefatigue plastic zone size to be greater than that associated with the fatigue test loads would have caused anomalous results in the early stages of the fatigue test as effectively an overload would have initially been present.

Once begun the fatigue test was run continuously until final failure occurred, final failure being when the specimen underwent large scale plastic deformation and a plastic hinge formed. Continuous running of the test was essential to obtain realistic results. Crack monitoring was achieved during the night by the use of crack monitoring gauges. These gauges are highly accurate but due to the spacings between gauge wires the resolution is limited to 0.25 mm. Although with limited resolution the gauges do enable the crack length to be accurately recorded on a y-t plotter, and together with the day time crack length readings, did make the $\frac{da}{dN}$ versus ΔK plots reliable.

5.6 Surface Deposits

In all tests, except those done in air, a surface deposit of some description was formed. The extent and nature of the deposit was determined by the testing conditions. There were two types of deposit formed; in the free corrosion condition a red/brown deposit was observed on the fracture surface and in the cathodically protected samples a white calcareous deposit was present on the fracture surface and the specimen surface.

If the deposit on the free corrosion specimen is considered first this deposit was initially formed as a greenish brown deposit which only lightly adhered to the specimen surface and for ease of crack measurement was brushed off using a nylon bristled brush. With time the deposit developed into a darker brown scale which was difficult to remove and for fracture work required chemical removal. The scales formed were most likely iron hydroxide which as the test progressed oxidised to ferric oxide.

Examination of the fracture surfaces has shown that the deposit decreases in quantity as crack length increases, the implication being that at the crack tip the metal surface was very similar to that produced in an in air test. The lack of scale at the crack tip will therefore provide no protection from environmental attack. It is therefore very likely that access of the hydrogen in the solution to the steel was not restricted.

The other deposit produced was a white calcareous deposit which was formed as a consequence of cathodic protection and was analysed using an energy dispersive analyser attached to the SEM. Analysis showed the deposit to be a magnesium based compound, probably magnesium hydroxide. The reason the deposit was magnesium based rather than calcium based was the sea water mixture chemistry as discussed earlier. The thickness of the scale was dependent both on the applied level of cathodic protection and the exposure time. More negative potentials gave thicker scales and the scale was seen to be thinner nearer the crack tip, although in the case of over protection, deposits were present to some extent even at the crack tip. The effects of calcareous deposits on the crack growth rate were seen to be both beneficial and detrimental depending upon the level of protection applied.

In the case of an applied potential of -800 mV only a thin surface deposit was present, even though the samples were pre-soaked for two weeks. Although thin, this surface coating was sufficient to prevent hydrogen ions from reaching the metal surface and was thus an excellent method of preventing hydrogen embrittlement. It may be, as discussed earlier, that because magnesium hydroxide is very alkaline, any hydrogen ions are neu-

tralised by reacting with the magnesium hydroxide. The level of applied potential was sufficient to reduce corrosion to acceptable levels but was not low enough to produce hydrogen and as such cathodically produced hydrogen was not a problem.

As a consequence of over protection screened protons would have evolved from the specimen surface. The thick scale covering the surface would prevent escape of these protons. It is probable that a large proportion would be adsorbed back into the steel because when they recombine to form hydrogen gas the calcareous deposit would prevent escape from the material surface. The hydrogen can then be readsorbed into the steel. Thus hydrogen embrittlement of the steel was a major problem.

Further decreases in potential to more negative values would have two effects: Firstly a thicker scale would be formed; but also the quantity of hydrogen evolved would increase. It would therefore be expected that at levels of over protection greater than -1100 mV crack growth rates would be enhanced even further. However, the greater production of hydrogen may well cause the scale to be fractured and thus allow the hydrogen gas to escape thereby reducing the hydrogen available for adsorption, this type of effect has been noted by Millar (116).

Calcareous deposits have been credited with the ability to increase the threshold for fatigue crack growth. Scott and Silvester (56) observed that for a starting stress intensity range less than $15 \text{ MPa}\sqrt{m}$ there was an apparent reduction in crack growth rate as the test was run. They attributed this to crack wedging by the calcareous deposits present. For ΔK values greater than $18 \text{ MPa}\sqrt{m}$ this effect was absent. Callister (108) observed a similar effect in Mannesman X80 parent plate and attributed the effect to crack wedging. However, once the apparent thresholding had been observed the tests were terminated and no data was produced for higher ΔK values.

If the build up of calcareous deposits is capable of effectively wedging a crack open to some extent, it must be assumed that wedging would occur some distance behind the crack front. The reason for this being that when calcareous deposits are formed initially they are not like the strong scale seen on the surface but are more like a thick liquid and their load bearing ability would be limited. Also it would require time for sufficient build up of deposit to effect the crack closure. If, assuming sufficient build up of calcareous deposits and premature closure of the crack does occur, what would the effects be?

Firstly the applied stress intensity range would be reduced and it would therefore be more appropriate to use a ΔK effective term, ΔK_{eff} . If the crack growth rates obtained were now

replotted against the ΔK_{eff} it would be seen that the crack growth rates would increase for increasing ΔK_{eff} values. Hence if the thresholding effect was real, it was only real in terms of ΔK_{eff} and this value would be considerably less than the ΔK value thought to be applied. As a consequence of the ΔK_{eff} term it should also be noted that the stress ratio, R , must also increase, assuming P_{max} is constant, resulting in the crack being held open for much longer periods at loads artificially produced by the wedging effects. The result of which would be enhanced, rather than reduced, crack growth rates when SCC conditions occur, particularly for materials with higher yield strengths which are more prone to SCC.

It would appear from Scott and Silvester's work that the specimens which appear to threshold were terminated once the thresholding process had started. If testing had continued there may have been three possible outcomes:-

1. If the calcareous deposit continues to build up it is possible that cyclic loading would eventually stop. Then the specimen would be in a state where fatigue crack growth cannot occur. Although it is worth noting that if there is a susceptibility to stress corrosion cracking this mechanism would now be able to produce crack extension, this in turn may be followed by crack advance using a fatigue mechanism.
2. If the build up continues at a rate such that the build up of calcareous deposits is able to reduce the ΔK_{eff} to a value less than the true threshold ΔK value faster than crack extension is occurring, then a true threshold would be achieved although even if fatigue thresholding has occurred, if the stress corrosion threshold is lower than the fatigue threshold then stress corrosion cracking may still occur.
3. If production of calcareous deposits cannot be maintained, or the build up site changes, then it is possible that as the crack extends, at a slow rate, eventually the ΔK_{eff} would tend to increase and the crack would start to accelerate. If this were the case then it would be expected that the enhanced crack growth rates normally observed would be seen again.

Callister (108) also obtained a similar result to Scott and Silvester (56) but used a much higher R ratio, 0.6 compared to 0.1. If crack wedging is to be accepted for this case it would require an exceptional rate of deposit build up of hard calcareous deposit to occur in order for the applied minimum load to be affected.

Other works on similar steels by Vosikovsky (20) has shown no such thresholding effect. The work done for this thesis, although done on HAZ of the Mannesman steel, also showed no such effects.

Fractography in this work also showed that whilst the calcareous deposit did show a thickening effect as the distance from the crack tip increased, there was no evidence that at any point the two surfaces of the fatigue crack had made contact. Whilst it is accepted that the HAZ may react differently, there is still considerable doubt in the author's opinion as to the reproducibility of Callister's results.

It is thought that in this work the calcareous deposits had only a limited effect, excluding their effect on hydrogen embrittlement mechanisms. Their effects would be limited to the current requirements for cathodic protection, and it is expected that the calcareous deposits would inhibit the flow of solution into the fatigue crack.

5.7 Mechanisms of Failure

5.7.1 Mechanisms of Striation Formation

Striation formation was discussed in an earlier section and the validity of each model to tension-tension loading was emphasised. However certain aspects of some models are important and these aspects will be highlighted.

The regions of striations observed were similar in size to the measured grain size of the material. The implication being that grain orientation is important to the formation of striations in Superelso 500. The importance of orientation is a consequence of the iron lattice which will allow slip to occur on $\{112\}$ and $\{110\}$ planes but only in the $\langle 111 \rangle$ direction, resulting in high stresses being required to cause slip.

If the situation occurs where high stresses exist and the grain orientation is suitable then slip would occur. Slip involves the movement of dislocations down the stress gradient. Dislocation movement would continue until the stress is removed or an obstacle is encountered, a typical obstacle would be a grain boundary. Upon reaching an obstacle, dislocations would start to pile up and the resistance to dislocation movement behind the obstacle would increase. As in the Tomkins model (27) slip would occur simultaneously on slip planes either side of the crack tip, but like the Pelloux model (28) strain hardening would occur. Continued load application would start to blunt the crack and if the loads became high enough failure would start to occur by mechanisms such as microvoid coalescence. Fatigue loading is not normally high enough to cause tensile type failures on anything more than a microscopic level. At the maximum applied fatigue load a situation

would exist where localised strain hardening had occurred on a specific slip plane and the crack had become blunt. At the crack tip the material would have been stressed through its yield point and some very localised plasticity would be present (5).

A reduction of the fatigue load would allow recovery of the elastic strain in the tensile ligament. Further load reductions would allow more elastic strain to be recovered but the plastic strain would not be recovered. In order to accommodate the extra ligament length caused by the plastic strain the ligament would buckle. Subsequent fatigue cycles would then cause the process to be repeated using the microscopic failure produced in the previous cycle as the initiation point.

Accurate description of fatigue striation formation can only be speculative at present because models can only be formulated from information obtained "after the event". Speculation concerning the formation of striations will continue until future developments allow the fatigue crack to be observed on a microscopic level as it is propagating. The SEM does allow the surface to be observed and with modifications the fatigue crack can be watched in situ. The limitations are that this is only a surface technique and, therefore, limited to a state of plane stress and the limit of magnification may well be too low.

The above discussions showed that the crack was propagating at different rates along the crack front, the local rate being dictated by the local microstructure. In regions where striations are formed the crack would advance ahead of the main crack until an obstacle to striation formation was encountered. The crack would then be restrained until sufficient energy was available to overcome the obstacle. It would appear from the SEM examination that the grain boundaries between adjacent grains were the primary obstacles to striation formation because of the discontinuity of appropriately orientated slip systems. In areas where striations were not formed the crack advanced by a ductile decohesion mechanism which gave slower crack growth rates than the striation mechanisms and in steels, even clean homogeneous steels such as Superelso 500, was the main mechanism by which the fatigue crack propagated.

5.7.2 Mechanisms of Failure Other Than Striations

Typically crack propagation in the Paris region of the $\log \left(\frac{da}{dN} \right)$ versus $\log (\Delta K)$ is said to occur by a striation mechanism. However, striations were not the primary features observed on the fracture surface. Instead, a mainly ductile decohesion mechanism was seen to be the dominant feature, except where hydrogen played an important role.

The formation of striations was discussed earlier but it is worth emphasising the major factors which enabled this rather fast mechanism of crack propagation to occur. Striations were observed, with some frequency, in Superelso 500 parent plate but were not present to anything near the same extent in the HAZ samples. The reason for this was the very homogeneous microstructure associated with this steel. In addition to a homogeneous microstructure the grains at the crack tip needed to be in a suitable orientation for slip to occur. Dislocation pile up would then occur and prevent further slip. As the crack blunted a pure tensile component was introduced causing microfracture of the locally strained ligament to occur, this microfracture then acted as the starting point for slip in the next fatigue cycle.

When striations are not formed the following mechanism is proposed to account for the process of ductile decohesion. If the orientation of the grain at the crack tip was such that there was no slip system available at, or near, the critical angle for maximum resolved shear to operate, then the degree of slip would be greatly reduced. As a consequence there will be a greater tensile component applied. It would be expected that some slip would occur and this would depend on the angles of the slip plane and orientation to the applied stress.

If an arbitrary slip angle is taken the schematic illustration will attempt to describe the processes which occurred as the fatigue load was applied. Initially the applied load was at a minimum and the crack was at its minimum opening (Figure 80a). Increasing the applied load activated a slip system and slip started to occur. At the same time the ligament held in the central crack region began to strain (Figure 80b). The degree of slip was limited by the angle of the slip system. Further loading caused more slip to occur and microfracture to occur in the strained ligament (Figure 80c). Increasing the load to the maximum applied increased the slip and increased the extent of the microfracture (Figure 80d). Reduction of the load allowed recovery of the elastic strain and general closure of the crack (Figure 80e). It would be expected if this simple explanation were fully able to explain the propagation of a fatigue crack, that there would be evidence of where slip had occurred. It should be noted that the degree of slip would be very limited but should be evident. However, no such evidence was seen.

The explanation for this is that where fracture occurs at the crack tip the crack may not cause fracture of all the material. Hence ligaments may be left behind the crack tip. This was initially proposed by Barsom (73) and subsequently supported by Delhovren (113) and evidence of ligament formation has been seen in this work.

These ligaments would have two effects:-

1. The presence of ligaments would prevent crack closure at minimum loads, although with an R ratio of 0.6 closure should have been minimal anyway.
2. Any tendency for closure would be resisted with the consequence of a force actually occurring in the opposite direction to the applied load and this force may be sufficient to cause reverse slip and hence eliminate the evidence of slip.

The above explanation is capable of explaining many of the different features associated with fatigue crack propagation. Most notable would be the variation in the rates of fatigue crack propagation, dependent on the grain orientation.

Another prominent feature was the secondary crack, which was a crack which had initiated and propagated only a small distance before being stopped. Many secondary cracks were seen in regions of striations although the amount of secondary cracking in regions where striations were absent indicated that secondary crack formation occurred at seemingly any time.

The formation of secondary cracks was thought to be a consequence of the hold up of crack propagation and the applied load looking for an alternative, easy region to cause failure. In striated regions, because this mechanism results in a fast crack growth rate, the crack front would be ahead of the remainder of the crack front. If the formation of striations was then prevented and the crack temporarily stops the applied load would look to act on weaker areas. If a weakness in a material was found close to the crack tip some of the applied energy to cause crack advance may produce crack extension of the newly forming crack. Hence, the energy available to overcome the obstacle would be reduced. However, as the secondary crack increases in length more energy would be required for crack propagation. When this situation is reached the original crack path through the obstacle may now provide the easiest route and the obstacle is broken down or avoided. The crack may then advance and will leave the secondary crack behind it. Thus whilst crack propagation is locally prevented the load will cause a secondary crack to form and grow. In regions where striations are absent it is possible that the ligaments formed would prevent crack advance by reducing the applied ΔK at the crack tip. Whilst the fatigue crack would try to fracture these ligaments any weaker areas may provide an alternative crack propagation path. If a weaker area was found a secondary crack may form in the manner described above. However, the crack would continue to try to advance in mode I, because this was the principal acting stress, and eventually the ligaments would fail and the crack would return all of its energy to advancing in the principal direction leaving the secondary crack. The secondary crack may form on a pre-slipped plane, if this process had caused a sharp

notch. Alternatively new slip systems may become viable due to the change in tensile direction caused by the ligaments and a new crack may grow. It is quite probable that whilst the secondary crack is propagating the ligament itself is subjected to a fatigue loading and would fail. Once failure occurs the crack will then propagate in the mode I direction again.

The use of a potential of -1100 mV to protect a sample caused a change in the mechanism fatigue crack propagation. The Superelso 500 parent plate fractograph (figure 52) showed clearly that large regions of quasi-cleavage dominated the fracture surface at low values of ΔK but at high ΔK values there was a decrease in the quantity of cleavage regions.

The following proposed mechanism was based on the Superelso 500 parent plate. The reasons for this are that the steel microstructure was homogeneous and the fracture features were clearly defined, whereas due to the welding process, modifications to the microstructure complicate the problem in the HAZ. The mechanisms of crack advance in the HAZ may be no different to those of the parent plate, just the resulting fracture features were not as clearly defined.

In the low ΔK regions hydrogen induced quasi-cleavage was the main fracture feature seen. It was also noted that these regions of cleavage were the same size as the material grain size. This was not unexpected because once initiated, a cleavage crack would rapidly propagate along the cleavage plane until halted by a grain boundary. Together with these cleaved regions were regions of ductile decohesion although the extent of ductility was very limited.

It was the presence of the ductile decohesion that indicates that the initial process must be similar to that previously discussed in that slip must initially start to occur. If now, as Lynch (119) proposed, dislocations are injected into the system we see that pile-up must occur rapidly and hence straightening of the ligament in the crack tip region. A tensile component would now exist across the crack tip. The embrittlement of the steel by the hydrogen would reduce the material's ability to withstand an applied strain and if a nucleation site for cleavage, such as a carbide, was present the crack would propagate rapidly across the grain. If no nucleation site was available the material would undergo microfracture in the manner described earlier but with the lack of ductility in the material the microcrack would not be blunted by plasticity effects and the extent of this fracture would be expected to be greater. The mechanism of failure described would result in a stop/start method of crack propagation. The extent of each cleavage burst was limited to one grain, because of the relaxation of the applied stress caused by the opening of the cleavage crack and the lack of suitably orientated cleavage planes in the adjacent grain.

At high ΔK values the crack growth rate showed a tendency to accelerate and together with the increase in ductile decohesion seen on the fracture surface this would imply that there was a change in mechanism away from stress corrosion cracking and towards fatigue crack propagation. If this were the case then the process competition model proposed by Austen and Walker (98) would seem to be an accurate general description of the processes occurring when an over protection was applied.

It was thought that because of the embrittlement process, the formation of the proposed ligaments will be severely limited in the regions of ductile decohesion and will be absent from regions of cleavage. If these ligaments did form it would be expected that there would be a reduction in the number of cycles that the ligament could withstand before failing. As these ligaments would only form in the ductile decohesion regions there was little likelihood that they could have played a part in the crack thresholding process put forward by various workers based on the crack wedging by calcareous deposits because at the lower values of ΔK the primary observed mode of failure was quasi-cleavage.

The HAZ of both Superelso 500 and Mannesman X70 did not exhibit the same quasi-cleavage fracture features shown by the Superelso 500 parent plate. The most likely cause would have been the changes in microstructure and possibly the changes in the carbide distribution.

Whilst it was intended that the fatigue crack be grown through the coarse grained HAZ, in practice the crack was frequently observed to grow in regions adjacent to the coarse grained region. This may well be the reason for the lack of cleavage on the fracture surface. The crack seen in the Superelso 500 HAZ fractographic sample was found to have grown through the refined weld metal. The cracks in Mannesman X70 HAZ were grown in the intercritical region. Both of these regions have high levels of toughness which may explain the lack of cleavage.

The influence of the change in carbide distribution was shown by the Mannesman X70 HAZ in which localised regions of microvoid coalescence were observed. Evidence of a tensile mode of failure must add credence to the mechanisms that proposed the material at the crack tip was subjected to a tensile component. Whilst cleavage is not a feature of the fracture surface obtained from the over protected HAZ samples, the lack of ductility is obvious. The lack of ductility caused by hydrogen embrittlement was the primary cause of the crack enhancement.

5.8 Microstructural Effects

In this section the effects will be confined to the constant load, rising ΔK , tests. The constant ΔK tests will be discussed later. In air testing of the Superelso 500 parent plate and Superelso 500 HAZ gave no definite indication of the effect of microstructure on the fatigue crack growth rates. The growth rates for both the parent plate microstructure and those from the HAZ are very similar at given ΔK values and it is difficult to speculate whether the microstructure plays a significant part in the fatigue crack growth rates observed. This is also true of the effect of microstructure on the fatigue crack propagation rate when the material is subjected to a cathodic protection of -800 mV.

In the case of a protection level of -1100 mV there was no difference in the plateau growth rates of the Superelso 500 parent plate and the Superelso 500 HAZ despite the differences in microstructure that the weld thermal cycle introduces. Different regions of the HAZ will reach different temperatures on the initial welding pass and on subsequent welding passes which will cause grain growth, grain refinement or have a tempering effect depending upon the local temperature reached. This lack of variation in crack growth rates between different regions indicated that the microstructure did not have an influence on the observed macroscopic growth rates. However, there was a trend for the HAZ to show a threshold effect at a ΔK of $14 \text{ MPa}\sqrt{\text{m}}$. The parent plate did not show this trend, although one specimen did show a small reduction in growth rate at low ΔK values, this threshold effect was thought to be a ΔK_{ISCC} threshold rather than a true fatigue threshold and it was quite possible that it was the difference in microstructure which accounted for this effect. The most noticeable effect of microstructure in the Superelso 500 parent plate is the presence of quasi-cleavage which is absent on the HAZ fracture surface. The lack of quasi-cleavage was a consequence of the changes in microstructure caused by the welding process.

A possible microstructural influence on the fatigue crack propagation rates when an over protection is applied to a sample was seen in the Mannesman X70 HAZ samples.

Two samples were tested but there was a difference in the growth rates observed. Whilst one sample showed the usual plateau type of crack growth rate, the other showed an apparent ΔK_{ISCC} at $12 \text{ MPa}\sqrt{\text{m}}$ before the growth rate followed a Paris relationship.

The only apparent difference between the two samples was the microstructure through which the fatigue crack propagated. Both materials were supposed to have been notched such that the fatigue crack propagated in the coarse grained region. However, the notch

in one sample was positioned such that the crack propagated closer to the intercritical region. The sample in which the crack propagated through the intercritical region gave the plateau type of fatigue crack growth.

5.9 Comparisons with Other Work

The potential of high strength steels for use in the construction of structures used to extract natural fuels from deep water locations has received considerable attention in recent years. Hence there are now numerous conferences and reports which have targeted this specific area. However, since the commencement of this work there has been a reversal in the presentation format of fatigue data. Consequently, the use of Paris type curves is on the decline and the use of S-N curves to portray the fatigue data has become the preferred format. The result of this is that there are problems in finding suitably presented works for comparison.

The comparisons will be split into three sections: a comparison of different quenched and tempered steels; a comparison of controlled rolled steels; and a general comparison between quenched and tempered, controlled rolled and structural steels.

5.9.1 Quenched and Tempered Steels

Comparison between this work with other reported work has proven difficult in that there is some confusion and contradiction in the literature. The most commonly investigated materials have been HT80, HY130 and HY100, all of which have higher yield strengths than the materials used in this work.

HY130 showed much higher growth rates than Superelso 500 parent plate in air (121). Fatigue in a corrosive environment showed an enhancement on the in air data when the steel was freely corroding. The application of a cathodic protection to HY130 reduced the fatigue crack propagation rates and the lowest growth rates were similar to the in air data. Further reductions in cathodic potential resulted in enhancement of the crack propagation rates up to the rates seen in this work. These effects are shown in Figure 81. HY100 showed increased corrosion fatigue growth rates compared with in air data. However, only two corrosion conditions appear to have been investigated; free corrosion and zinc coupled (122).

The trend was for the fatigue crack propagation rates to increase the more negative the potential. Coupling to a zinc anode gave a measured potential of -1030 mV and this gave the typical over protection form of curve, with the crack growth rates being enhanced and appearing to have propagated by a stress corrosion mode (Figure 82).

Corrosion fatigue testing of a submerged arc weldment of HY100 showed that generally the crack growth rates followed the same trends as the parent plate material although, compared to the parent plate, the growth rates were lower for given ΔK values (122).

Comparison of these data with those obtained from Superelso 500 was difficult because of the difference in locations of the fatigue crack. However, a very general comparison indicated that there was not an excessive variation between the two sets of data.

HT80 was the material with the most literature available and it was not surprising that there was some difference of opinion about the material's fatigue response. Masuda et al (123) and Ebara (124) found similar fatigue results for HT80. The growth rates, for given ΔK values, were similar to those obtained from Superelso 500 both in air and freely corroding, Figure 83. Other work showed HT80 to have lower fatigue crack growth rates for given ΔK values when tested in air. However, when tested under freely corroding and cathodic protection of -850 mV conditions the growth rates were seen to increase above those obtained in air, Figure 84 (125).

Crack growth rate enhancement was observed by both investigators when an over protection was used and both attributed this enhancement to hydrogen embrittlement. Millar (116) showed similar trends for welded samples of N-A-XTRA 70, except the growth rates were generally higher.

It would thus appear that Superelso 500 parent plate compares favourably with most other quenched and tempered steels when tested in air or in a corrosive environment.

The limited amount of data available for HAZ makes comparisons with Superelso 500 difficult but it did appear from the data that was available that the Superelso 500 HAZ did have good corrosion fatigue properties when a protection of -800 mV was applied. There did seem to be a susceptibility to hydrogen assisted cracking both at low and high hydrogen concentrations.

5.9.2 Controlled Rolled Steels

Comparison of the corrosion fatigue data obtained in this work with that in the literature was made easier because the majority of work had concentrated on X65 and X70 line pipe steels.

In general, both grades exhibited very similar trends with the in air data showing the lowest growth rates, free corrosion showing intermediate growth rates, and cathodically protected, using a coupled zinc anode, giving the fastest growth rates.

Under corrosion conditions X70 showed comparable growth rates to X65, but when tested in air the X70 growth rates were normally higher than those obtained from X65.

Data obtained from HAZ regions is totally lacking in the literature for both in air and corrosion fatigue conditions. The only other work found was that of Callister who looked at the HAZ in air. Callister (108) also did some work on the parent plate both in air and under corrosion conditions, this data is given in Figure 85.

The work on parent plate showed that at low values of ΔK the free corrosion condition gave lower growth rates than the in air condition although because the m value is higher for the corrosion condition there was a crossing of the data at a ΔK value of $17 \text{ MPa}\sqrt{m}$ and thereafter the free corrosion growth rates were higher than the in air values. A cathodic potential of -800 mV produced growth rates slightly higher than the in air growth rates but the Paris plots were parallel.

The only unexpected feature occurred when a protection of -1050 mV was applied, in this case an apparent thresholding occurred, and this anomaly has already been discussed.

The data produced by Callister for the HAZ in air showed that the growth rate followed a Paris type relationship. The growth rates were slightly enhanced compared to the in air growth rates. The HAZ corrosion fatigue data obtained in this work shows similar trends to those reported to the literature for parent plate materials. The use of a cathodic protection of -800 mV returns the corrosion fatigue data to approximately that obtained in air whilst an over protection causes enhancement of the crack growth rates. The free corrosion data is the only data which differs to the parent plate data. In the HAZ the crack growth rates obtained under free corrosion conditions were the lowest obtained from any condition at a given ΔK value.

5.9.3 Comparison Between High Strength Steels and Conventional Structural Steels

Corrosion fatigue of structural steels can be summarised as: increased growth rates are observed as the cathodic potential is decreased, i.e. P_v becomes more negative.

Comparisons between controlled rolled X70 and quenched and tempered Superelso 500 showed that in air there was little to choose between these two high strength steels or BS4360:50D, as is seen from Figure 86.

However, in the free corrosion or cathodically protected at -800 mV state both high strength steels exhibit lower growth rates than BS4360:50D. As was the case for the high strength steels, structural steel HAZ data was very limited. Some HAZ data is available (92,112) on a structural steel similar to BS4360:50D which showed that in air the growth rates were slightly lower than those obtained from the parent plate. Maddox (126) in an investigation of HAZ's of structural steels showed that the in air HAZ of a structural steel at similar crack growth rates to the parent plate. The effect of the material yield strength was seen to increase the HAZ crack growth rate. However, Maddox used weld simulations for the production of the test pieces, this can cause problems in interpretation of the data as simulations will only represent a specific thermal cycle. Actual HAZ's from welded samples may be different but it is thought that the general trends seen by Maddox should be representative.

In summary it can be shown that the use of high strength steels has particular advantages over BS4360:50D when used in a corrosive environment, Figure 87. These advantages are that in both the free corrosion and cathodically protected states crack growth rates are similar to the parent plate in air growth rates.

Welding of the high strength steels makes only a small difference to the fatigue crack growth rate and the HAZ material showed a similar response to the parent plate material. There is little to choose between the two high strength steels which is in direct contradiction to Callister (97) who stated that controlled rolled steels show better fatigue crack growth resistance than quenched and tempered steels.

5.10 Constant ΔK Testing

The earlier sections of this discussion have indicated just how difficult accurate notch location can be in multipass weldments. The reason for this is because the welding process produces a rapid variation in material microstructure over a very small increment of distance. If these different locations all had the same crack growth rates for given stress intensity ranges then notch and crack location would be of little importance. The earlier sections, and other works in the literature, have shown that different weldment regions can give variations in the observed fatigue crack growth rate. It is therefore relevant for different weldment regions to be investigated in order to assess the influence of detrimental material microstructures.

Traditionally the technique employed in producing specific microstructures has been a weld simulation, in which a material is heat treated through a specific thermal cycle to induce the required microstructure, this technique, whilst producing representative microstructures, does have draw backs. Only a specific microstructure can be introduced and there is always an element of doubt as to the accuracy of a simulation compared to the "real" weld microstructure.

The second technique is to use "real" welds to produce the microstructures, but the problem here is that the different regions only extend for very short distances, and for low heat input welds the entire HAZ may only be 1 to 2 mm in width. The use of "real" welds was considered to be of significantly more benefit than weld simulations and the use of continuous crack monitoring gauges, adhered to the sample surface, gave accurate crack measurement with excellent resolution over a distance of 6 to 7 mm.

In order for comparisons to be made between different weldment regions it was necessary to remove any external variables that may have influenced the crack growth rate. Thus it was considered that the use of a constant ΔK would make comparisons more valid. To apply a constant ΔK either the applied stress ratio must increase or the loads must be reduced to maintain that constant stress ratio. The use of continuous gauges coupled to a computer allowed the applied ΔK value to be kept constant with a high degree of accuracy by decreasing the applied loads as the crack length increased.

The main problem with this technique is that the crack length is only measured at the surface and consequently there may be some discrepancy between what was observed at the surface and what may occur in the centre of the sample. The use of this technique produced large quantities of accurate and reliable data for a small increment of crack growth.

The most interesting feature observed from the crack growth rates observed from the constant ΔK tests was for the low heat input weldment. In this weldment a sharp increase in crack growth rate was observed as the crack propagated from the weld metal and entered the coarse grained region of the HAZ. The growth rate in this region was seen to be 5×10^{-7} m/cycle and occurred over a distance of 50 μm . This distance is comparable to the grain size of the coarse HAZ. The observed growth rate of 5×10^{-7} m/cycle may be a conservative estimate of the crack growth rate in this region because the crack length was only recorded every 100 cycles. In the worst case the crack may have grown at a rate of 5×10^{-5} m/cycle if the crack had propagated through this 50 μm region in a single cycle. Metallographic examination of this region gave a clear indication of the reason for the accelerated crack growth rate. On entering the coarse grained region the fatigue crack had

been deflected by the aligned carbides running in a plane approximately 30° to the expected crack path. These carbides are continuous and exist on parallel planes separated by ferrite. The distance between the planes was small and the crack was given very little latitude and once the crack had entered this aligned microstructure it found it difficult to escape. The fatigue crack was also seen to cross a grain boundary where it was again deflected by the carbides of the grain and forced to travel down the carbide path. It is quite possible that one cycle gave 5×10^{-5} m/cycle of crack growth if the local region of microstructure was brittle and unable to cause crack blunting through plastic deformation. These regions of brittle material are known as local brittle zones (LBZ).

LBZ's are a problem known to occur in higher strength steel weldments. These local zones may be weak links which trigger brittle fracture when fracture toughness testing (127).

Consideration of the thermal cycle of a single pass weld shows that the material adjacent to the fusion line is rapidly heated to very near the material melting point and is then rapidly cooled giving rise to coarse grains in which LBZ's are usually located (128).

The lack of any rapid crack growth rate regions in the 3 and 6 kJ/mm heat inputs would indicate that LBZ's are a problem specific to low heat input welding. The reason for this will be the tempering process which occurs due to the amount of heat the higher heat input welds put into the material, this extra heat will take longer to disperse and the microstructures are affected by this heat.

The LBZ problem is very important when related to real structures. Although single pass welding is not used to join structural members there is a case similar to a single pass when the final pass is laid down. The situation is further complicated if multipass welding is considered. In this case a weld pass will be affected by any subsequent weld pass, the degree to which it is affected will vary depending upon the local peak temperature reached. Denys et al (129) have shown that in a low heat input weld susceptible coarse grained regions can occur regularly through a HAZ particularly down the fusion line of a multipass weld. Thus in low heat input welding using multiple passes a situation exists where numerous LBZ's can line up to give a rapid path for crack advance and possible catastrophic failure. The plate thickness also becomes a factor to be considered because it is known that with thicker plate sections there is a higher proportion of LBZ's and their size increases thus increasing the possibility of the fatigue crack impinging on a brittle zone. However it has been stated that in many types of tubular joint the fatigue crack will grow away from the LBZ's and so there is not a risk of unstable fracture (130).

In a real situation where a nodal joint is subjected to cyclic loading cracks are normally initiated at local hot spots (131). Growth then occurs radially at a far greater rate than the rate of growth through the tube wall. Currently the frequency of inspection of nodal joints depends on how critical the joint location is to structural integrity. Even so crack detection is not particularly successful, with only a 50% detection rate for cracks of range or length up to 20 mm (132). New techniques are currently being evaluated which claim confident detection of cracks 30 mm long and 3 mm deep (133).

It is necessary to put the affect of LBZ's into some kind of perspective with regard to fatigue crack propagation. The LBZ is located primarily in a coarse grained region of weldments and for example they occur every 0.5 mm through the HAZ and give accelerated crack growth over a distance of 50 μm . If a crack was 3 mm deep, and was not detected on a routine inspection and the next inspection was not due for 12 months then the crack would extend to a depth of 14.6 mm assuming a constant growth rate of 2.2×10^{-8} m/cycle at a frequency of $\frac{1}{6}$ Hz. Comparison with a higher heat input, for example 3 kJ/mm, under the same conditions but with no LBZ's the crack depth would be 25 mm based on a growth rate of 4.2×10^{-8} m/cycle. These figures for growth rate are consistent with the values obtained from the ΔK experiments performed in this work. In terms of fatigue, LBZ effects are far out weighed by heat input effects because of the increases in crack growth rates associated with higher heat input welding. The exception may be for the fatigue crack which propagates radially around the nodal joint and if the LBZ's line up radially it is quite possible that the fatigue crack would rapidly propagate around the joint.

Thus the important feature of LBZ's is in the single pass capping run where an LBZ may nucleate a fatigue crack prematurely or give a small crack a short period of enhanced crack growth. Due to the limitations of the present crack detection systems the early initiation of a fatigue crack will not be detected.

When considering the use of welding as a procedure for the joining of steels various factors need to be considered such as joint integrity and strength. Providing welding procedures are available which can satisfy all the mechanical requirements the main aspect of welding is the financial cost of doing the weld. Fabricators will naturally want to limit their costs to the minimum and hence welding will be automated where possible and the weld metal will be deposited at the fastest possible rate, i.e. using high heat inputs.

It was seen from the literature that the heat input as a variable which may affect the fatigue crack growth has received little or no attention, indeed three reviews do not cite a single paper in which heat input was investigated (134-136). Hence the use of the constant ΔK experiments to assess the effects of heat input would appear logical.

The weldment of a steel is conventionally split into three sections. The weld metal is the material which was molten during the welding process; the HAZ is the region adjacent to the weld metal which has been heated but has not reached its melting point; and the parent plate is the material which shows no effects of the weld heat.

Table 12 summarises the fatigue crack growth rates for the different weldment regions of the three heat inputs. There are three general features that are seen from the crack growth rates given in this table:-

- i. The fatigue crack growth rate increases as the crack propagates from the weld metal through the HAZ and into the parent plate.
- ii. The fatigue crack growth rate increases in all weldment regions as the heat input is increased.
- iii. The crack growth rate for the parent material of the weldment shows crack growth rates greater than the base material which has been unaffected by heat.

The different crack growth rates for the individual, and distinct, weldment regions would indicate that it is a residual stress effect which is causing the variations in crack growth rates. However, normally the weld metal is subjected to tensile residual stresses and the adjacent parent material has compressive residual stresses. Thus it would be expected that the weld metal would have the enhanced crack growth rate, not the parent material. Hence the usual reason for variations in fatigue properties of weldments, residual stresses, may not adequately explain the observed crack growth rates.

The hardness traverse across the weldment of each heat input all show similar trends, i.e. the weldment hardness is greater than that of the parent plate and the hardness reaches a peak on the parent plate side of the fusion line in the HAZ. The hardness is an indication of the material yield strength and hence should give an indication of the plastic zone size of the crack as it propagates through the weldment regions.

If, as Dowse and Richards (107) suggested, the plastic zone size is the controlling factor in the rate of crack propagation the crack growth rate should react in the following way:- As the crack propagated through the weld metal the plastic zone size would remain constant. However as the crack approached the fusion boundary the hardness value rises rapidly to a peak and the corresponding plastic zone size will be reduced. The effect should be a decrease in crack growth rate, this was not seen.

The general trend is for the hardness value to decrease as the distance from the fusion line increased, this should have resulted in a general increase in crack growth rate.

Consideration of the hardness values for the different heat inputs also indicated that the plastic zone size theory of Dowse and Richards is not readily applicable to the fatigue crack growth in a weldment.

At a heat input of 6.1 kJ/mm the hardness of the weld metal is only 20 Hv above that of the parent plate but the crack growth rate is 8 times as fast in the parent plate. Additionally the hardness of a HAZ for all heat inputs drops to the same level of approximately 205 Hv but the crack growth rates vary considerably.

Thus it would seem unlikely that the Dowse and Richards theory of decreasing crack growth rate with decreasing plastic zone size applies to the weldments tested here.

One major consequence of the weld thermal cycle was the effect on the local microstructure. Microstructural changes are not always considered to be significant for fatigue crack propagation and it would appear that there was no definite microstructural connection, apart from the localised region which contains the LBZ discussed earlier, seen in this work. Thus it would appear that there is no single overriding factor which dominates the fatigue crack growth rate through a specific heat input weldment.

The influence of the heat input appeared to show a definite trend of higher crack growth rates as a consequence of higher heat inputs, this was true for all regions. Generally the crack growth rate showed a two fold increase as the heat input was doubled.

The major benefit of higher heat inputs is the absence of the local region of rapid crack growth rate. The reason for this was the tempering of the coarse grained region such that the aligned carbides, seen in the 1.5 kJ/mm heat input, were broken up and the crack did not have a local brittle region to travel down.

The increase in crack growth rate, in all three regions of the weldment, with increasing heat input also casts doubt as to the influence of residual stresses on the fatigue crack growth rate. Residual stresses are a consequence of the welding process and are caused by the contraction of a molten metal as it solidifies. The resulting residual stresses would be tensile in the weld metal and compressive in the parent plate. The use of different heat inputs would affect the magnitude of the residual stresses. Higher heat inputs would give higher residual stresses but, as stated earlier, the residual tensile stresses would be in the weld metal and it would be this region which had the higher crack growth rates. The relaxation of any residual stresses, using a post weld heat treatment, would have given a more definite indication as to the influence of the residual stress on the fatigue crack growth rate. If residual stresses were found to be the cause the use of preheat may well reduce the fatigue crack growth rate.

The above discussion indicated that no single factor was dominant in determining the fatigue crack growth rate. However, the results did indicate quite conclusively that the fatigue crack growth rate increased as the heat input increased. It is worthwhile noting that the crack growth rate for the single bead and multipass weld were identical at a $\Delta K = 20 \text{ MPa } \sqrt{m}$. Hence the use of higher heat inputs for the deposition of weld metal may reduce welding time and reduce fabrication costs but in terms of fatigue crack propagation rate the result of using a high heat input for welding would be a reduction in fatigue life of a member which has a propagating crack.

6. Conclusions

- 1 A marked increase in fatigue crack growth rate was observed with respect to testing in air when cathodic potential levels of -1100 mV were applied, based on the work in this thesis on the parent plate and HAZ. This comparison at low ΔK values indicates that the crack growth rate was enhanced by a factor of ten. This research supports the fact that hydrogen, generated by cathodic potentials above those required to prevent rusting, embrittles high strength steels.
- 2 Whereas environmentally assisted fatigue crack growth rates under free corrosion conditions showed unusual fatigue crack propagation rates, the use of a cathodic potential of -800 mV restored the fatigue crack propagation rates to those obtained when the HSLA steel, or the HAZ of the HSLA steel, was fatigue tested in air. The use of a potential of -800 mV did not cause hydrogen to be produced in significant quantities but did prevent the steels from rusting. The rates of crack propagation are lower than those recorded in the reference literature for BS4360:50D structural steels. Provided over protection is avoided in high stress regions there is no evidence, based on corrosion fatigue properties, why these steels should not be incorporated into offshore structures.
- 3 The use of low heat input, single pass welds produced a very localised brittle region which was located in the coarse grained region of the HAZ. The presence of this embrittled region corresponded with a rapid increase in crack propagation rate over a 50 μm increment with crack growth rates of 5×10^{-7} m/cycle. The formation of a local brittle region was specific to the low heat input weld (1.5 kJ/mm) and was attributed to the rapid rate of cooling associated with the heat flow from the weld region into the bulk of the steel. This resulted in the microstructure with the aligned carbides remaining relatively untempered. At higher heat inputs it appeared that the carbides in the coarse grained region were broken up and the microstructure had been tempered by the higher heat input used.
- 4 The use of different heat inputs for the deposition of the weld metal in single pass welds had a significant affect on the fatigue crack growth rate, with higher heat inputs giving higher fatigue crack growth rates in each weldment region. This affect of heat input was investigated using single pass welds but the similarity between the fatigue crack growth rates obtained for the 3 kJ/mm single pass and 3 kJ/mm multipass weld indicated that the correlation between higher heat input welding and crack growth enhancement should also apply in multipass welding of offshore fabrications.

- 5 Although the crack propagation rate for each heat input was seen to vary for different weldment regions, in all cases the parent plate region gave the highest crack growth rate. Hence for these materials the parent plate fatigue crack propagation rates could be used as a basis for design and calculations to predict the rate of crack propagation through any weldment region. This will provide a moderate reserve factor of 2 on the crack propagation rate while the crack path crossed the HAZ/weldment. It is probable that a growing fatigue crack would spend the early post-initiation growth period in this region, but for there to be a marked discrepancy between the calculated and actual crack growth rates the fatigue crack would need to propagate through a significant amount of weldment compared to parent plate.
- 6 The microstructure had little influence on the macroscopic crack growth rate of the fatigue crack but there were indications that on a microscopic level the local grain orientation did influence the mechanism by which the crack propagated. The mechanism by which fatigue crack propagation occurred was primarily ductile decohesion for all materials and potentials. The use of cathodic potentials of -1100 mV caused a reduction in the ductility of the material and for the particular case of Superelso 500 parent plate, tested at low ΔK values at a frequency of 0.5 Hz, local hydrogen-induced quasi cleavage facets were clearly observed. A secondary mechanism of crack advance involving the formation of striations was observed in local areas on the Superelso 500 parent plate. The striation spacing was an order of magnitude greater than the macroscopic crack growth rate.
- 7 The production of magnesium-based deposits occurred when a cathodic potential was applied to a steel in synthetic seawater. Under corrosion fatigue conditions using a load ratio of 0.6 no crack wedging occurred. The quantity of deposit depended upon the applied level of protection, increasing as the potential decreased. It was postulated that the surface deposit had an influence on the movement of hydrogen ions. If potentials were at a level such that hydrogen ions were not produced the presence of a thin surface film prevented any hydrogen ions present in the seawater from entering the steel which was beneficial. However, at potentials where hydrogen is produced the deposit then restricted the escape of the hydrogen ions such that more entered the steel. Consequently the deposit now tended to have a deleterious affect. The practice used on some offshore structures of over protecting to build up the calcareous deposits to lower the required currents for protection may be beneficial. However, extreme care should be taken because of the possibility of enhanced corrosion fatigue crack growth rates from preexisting cracks during the period of excess current, as hydrogen may be retained in the steel microstructure.

7. Further Work

The evidence is that Superelso 500 and Mannesman X70 and their HAZ are susceptible to hydrogen cracking. This susceptibility needs to be assessed at ΔK values below $10 \text{ MPa} \sqrt{m}$ in order to determine the ΔK_{SCC} . Additionally the problem of weld cracking needs to be assessed using the control thermal severity technique.

The threshold ΔK value needs to be determined and to achieve a high level of accuracy the gauges used for the constant ΔK monitoring could be employed to measure any very small changes in crack length.

Due to the current trend of producing S-N curves as opposed to the production of crack growth rate data it would be of benefit to assess the properties of Superelso 500 and Mannesman X70 using the S-N technique.

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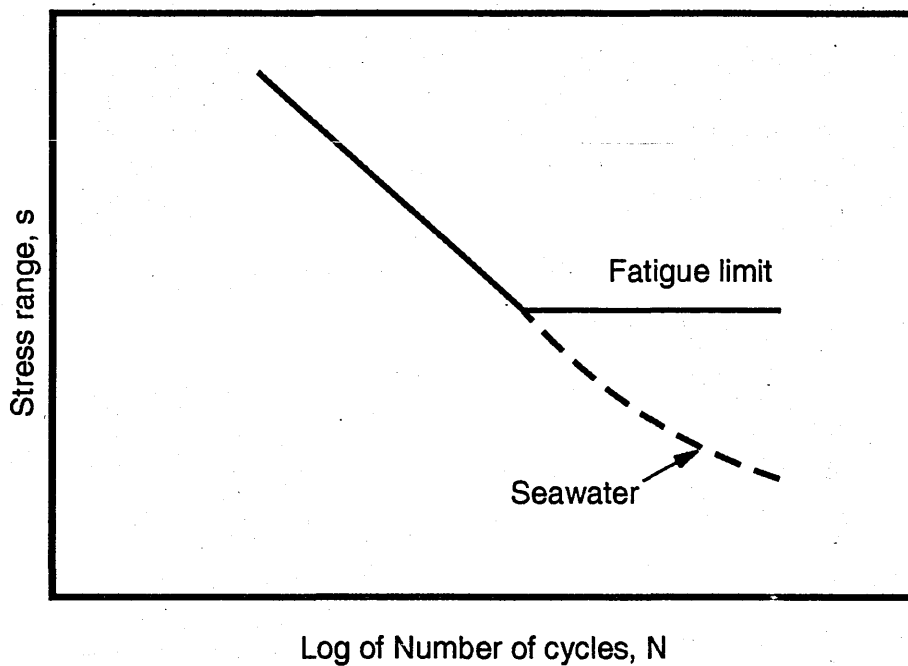


Figure 1 *The effect of sea water on the fatigue limit of a steel*

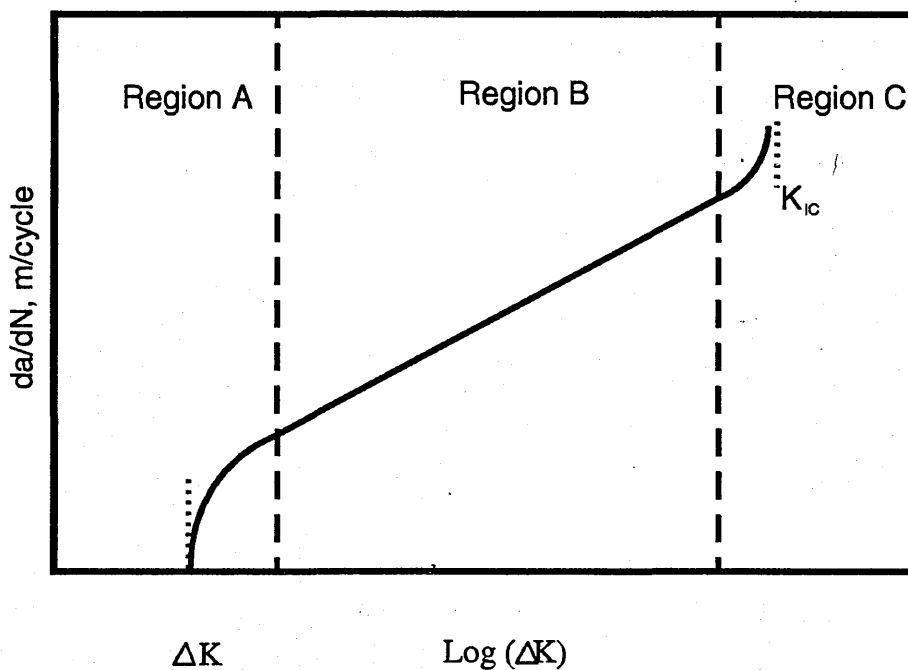
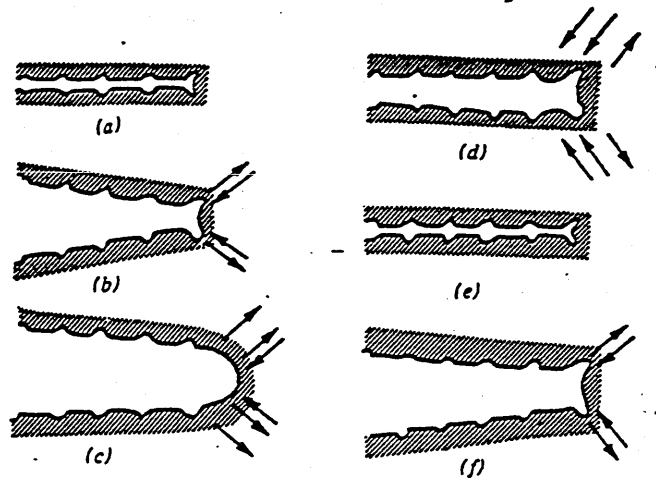


Figure 2 *The sigmoidal variation in fatigue crack propagation rate ($\frac{da}{dN}$) with alternating stress intensity (ΔK)*



—Schematic representation of the plastic blunting process of fatigue crack propagation in the Stage II mode: (a) zero load; (b) small tensile load; (c) maximum tensile load of the cycle; (d) small compressive load; (e) maximum compressive load of the cycle; and (f) small tensile load in the succeeding cycle. The double arrowheads in (c) and (d) signify the greater width of slip bands at the crack in these stages of the process. The stress axis is vertical.

Figure 3 Schematic representation of the plastic blunting process of fatigue crack propagation for the stage II mode.

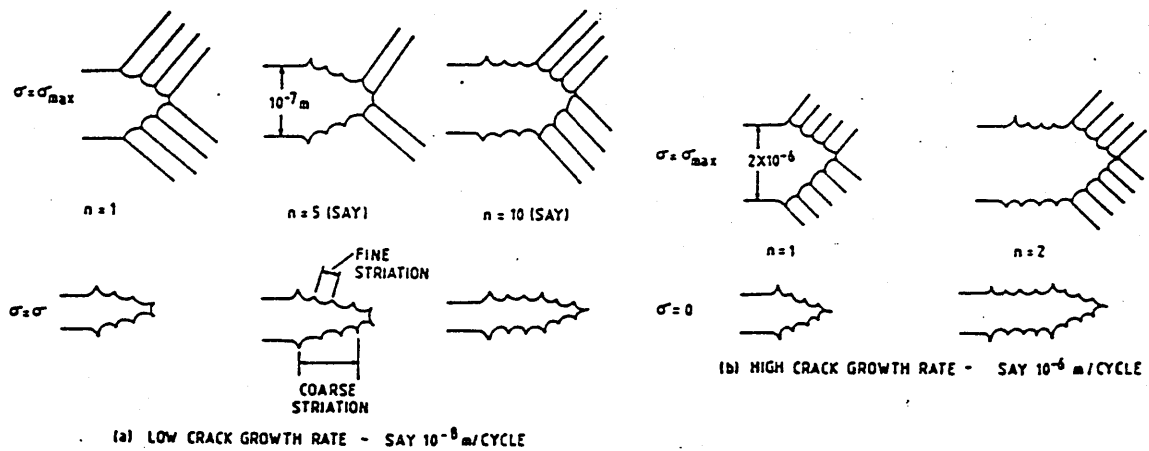


Figure 4 The Tomkins model of fatigue striation formation.

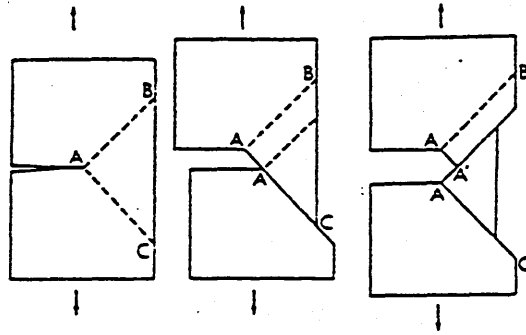


Figure 5 · *The Pelloux alternating shear mechanism of fatigue striation formation.*

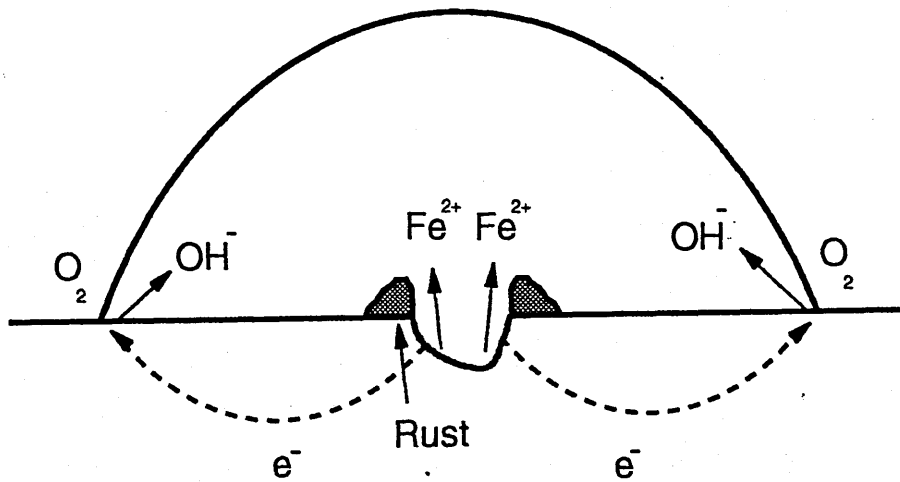


Figure 6 *Differential aeration causing rust to form under a water droplet.*

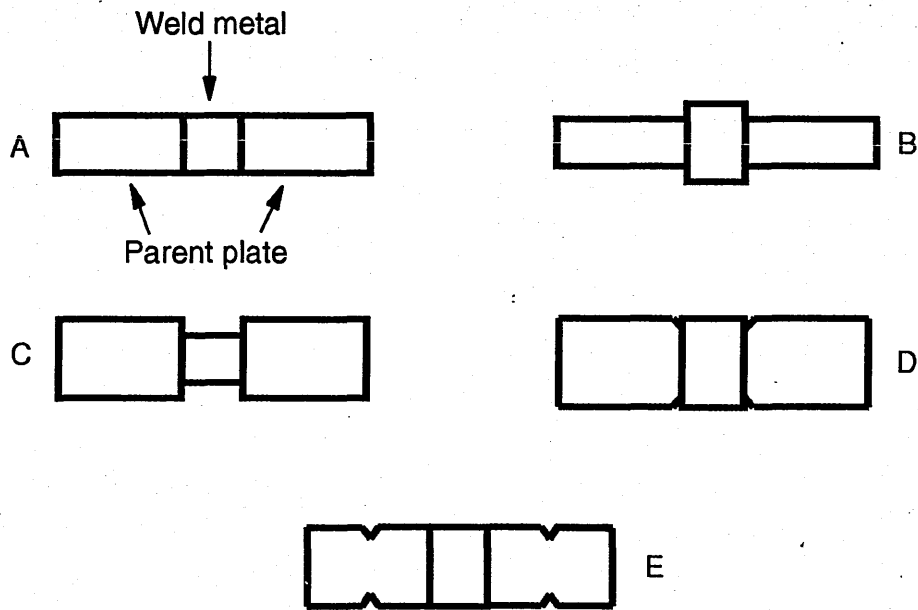


Figure 7 Different areas of weldment corrosion caused by variations in metallurgical conditions

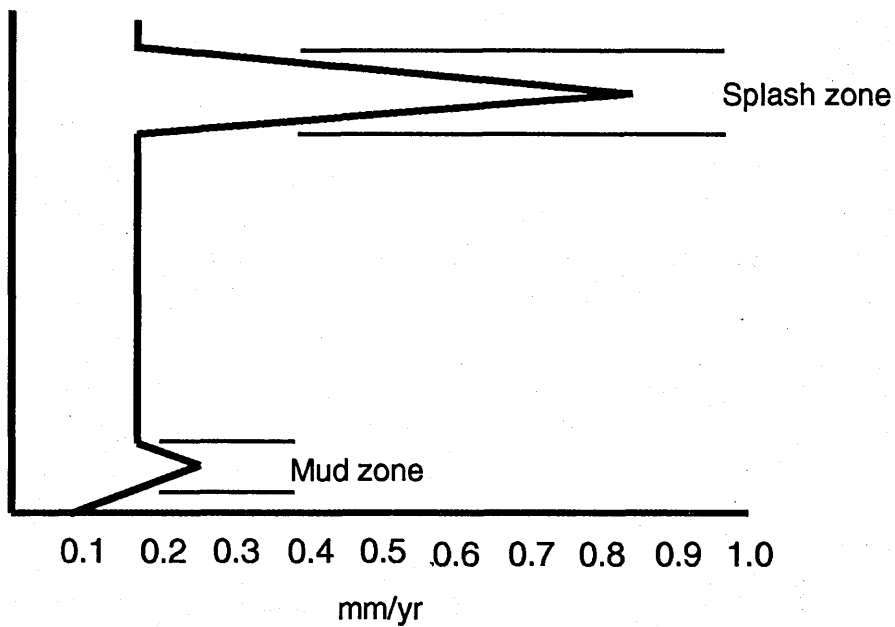


Figure 8 The effect of location on corrosion rates of an offshore structure.

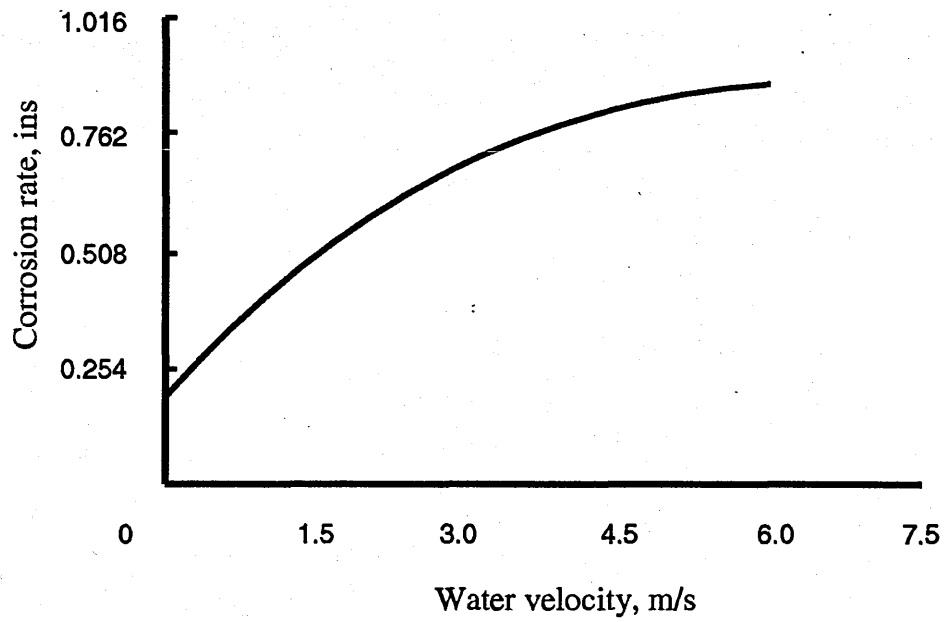


Figure 9 *The effect of velocity on the corrosion rate.*

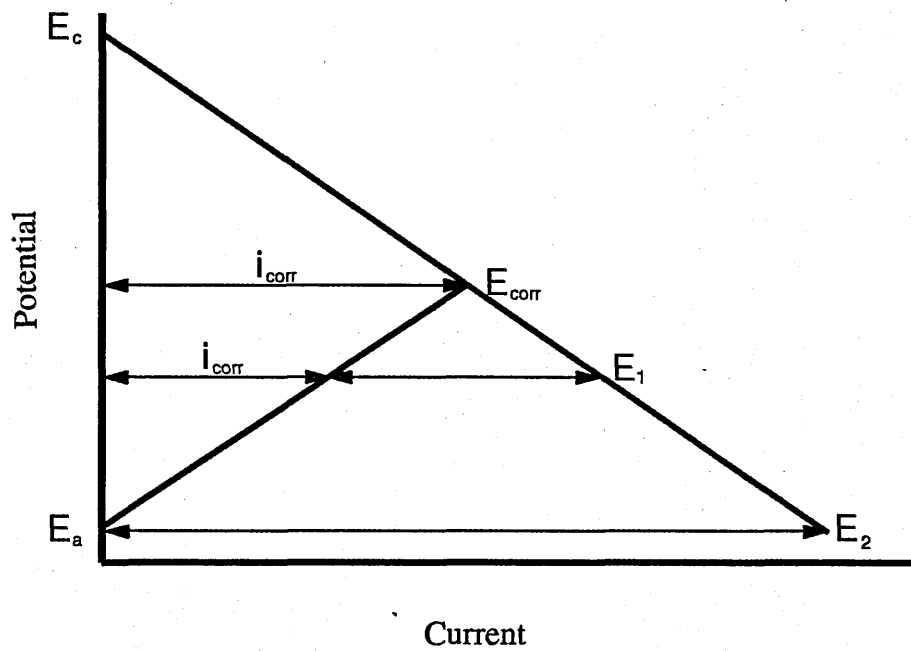


Figure 10 *Potential - Current diagram (Evans Diagram)*

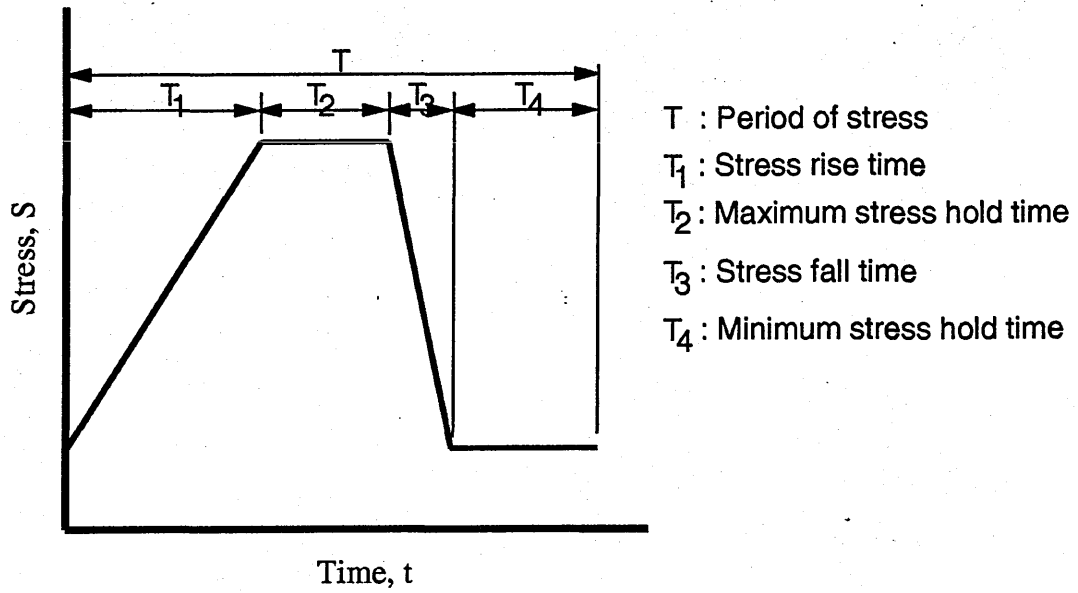


Figure 11 A schematic representation of a wave form.

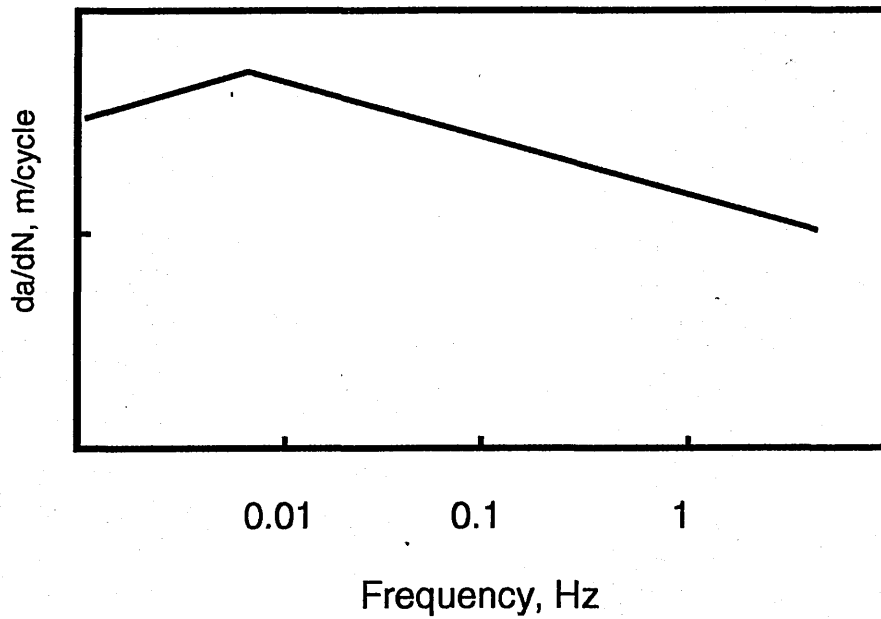


Figure 12 The effect of frequency on crack growth rate.

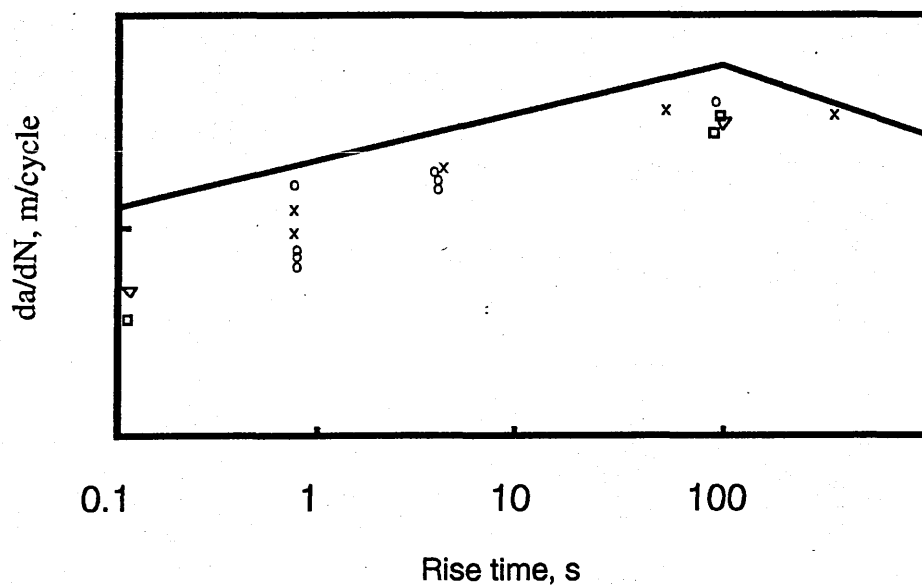


Figure 13 *The effect of rise time on crack growth rate.*

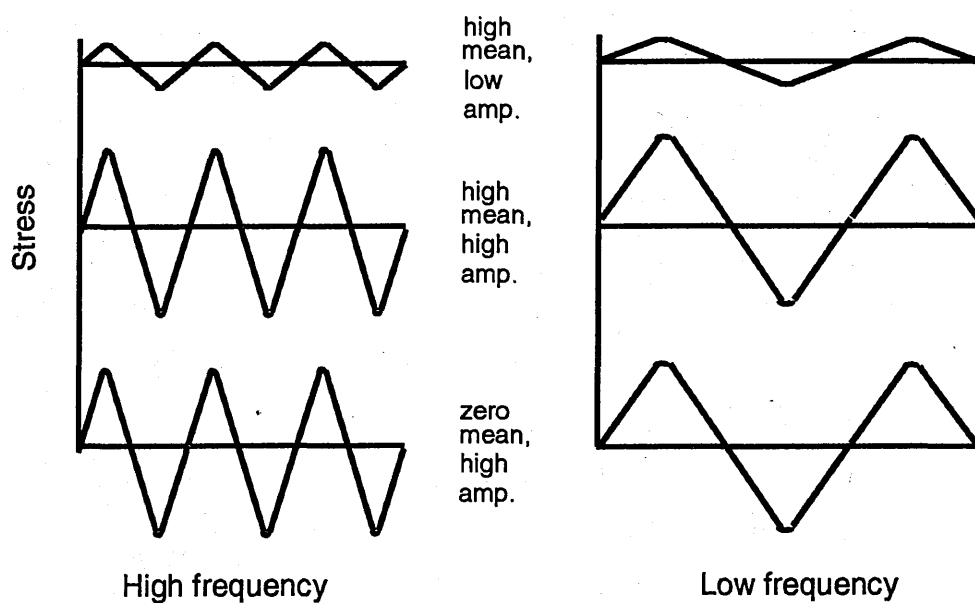


Figure 14 *The effect of means stress and amplitude producing an SCC failure rather than fatigue*

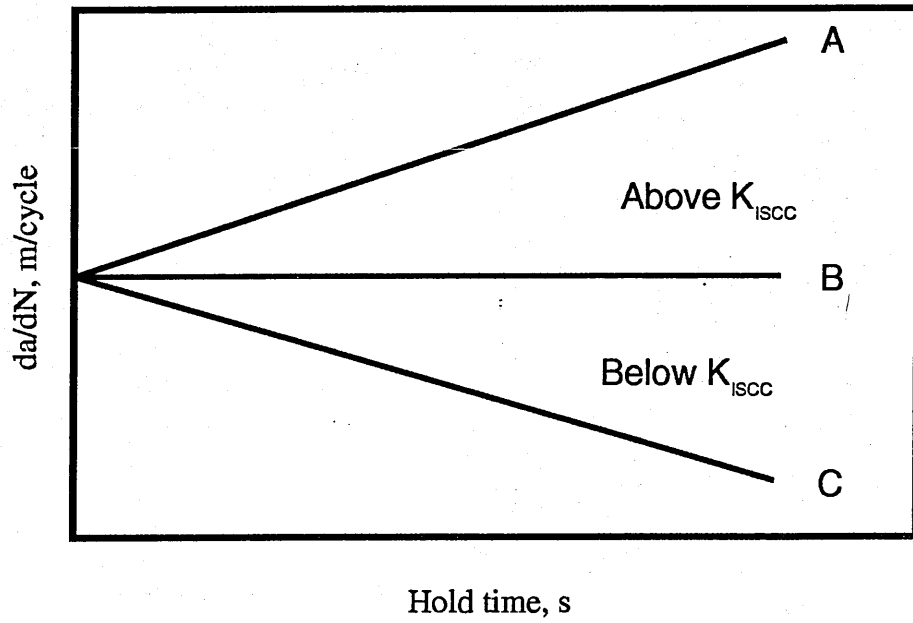


Figure 15 *The effect of hold time on the fatigue crack growth rate.*

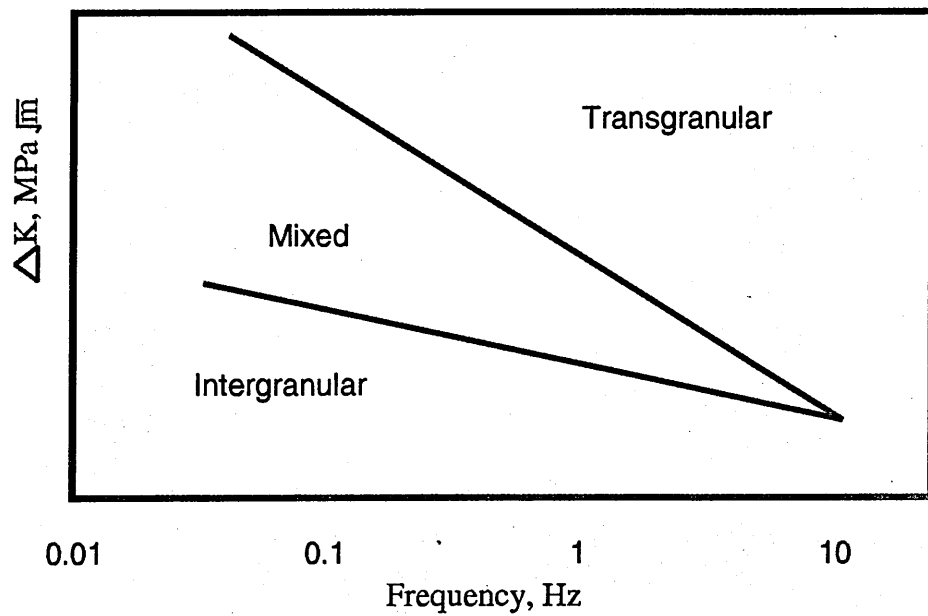


Figure 16 *The effect of frequency on fracture mode.*

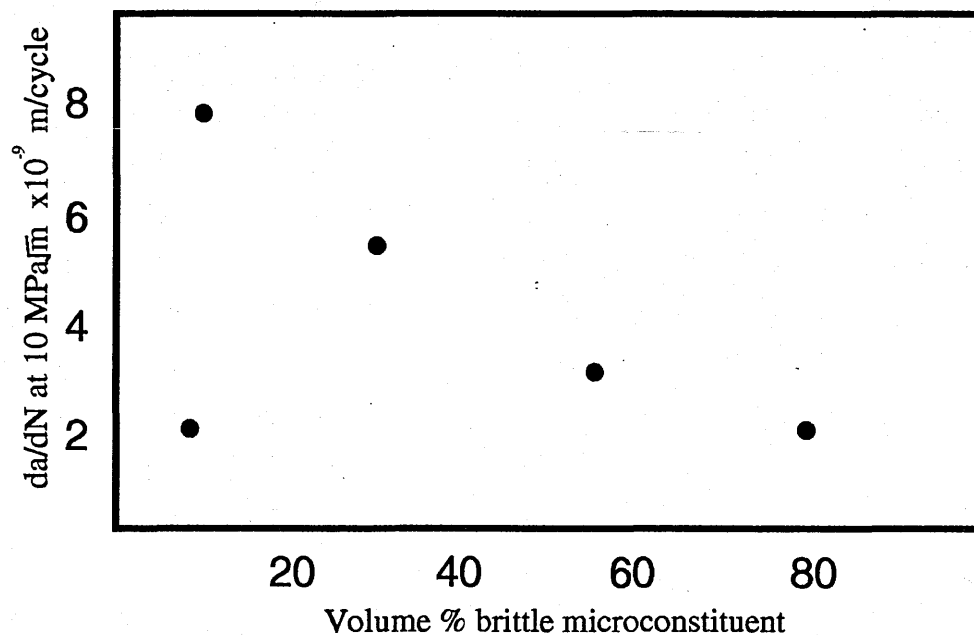


Figure 17 The crack growth per cycle for different iron-carbon alloys at $10 \text{ MPa}\sqrt{m}$

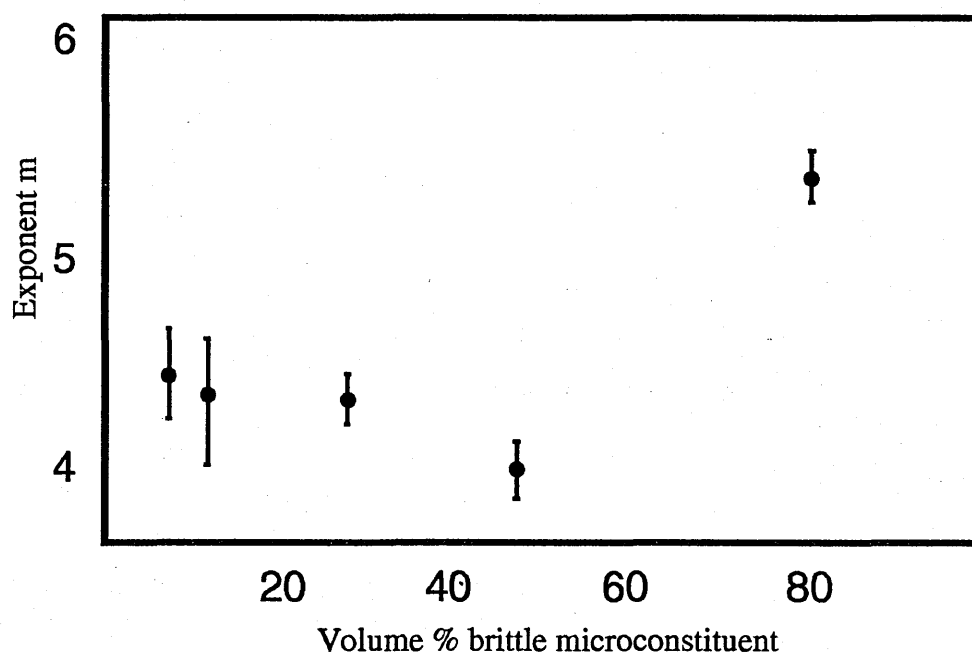


Figure 18 The effect of brittle constituent on the Paris exponent m .

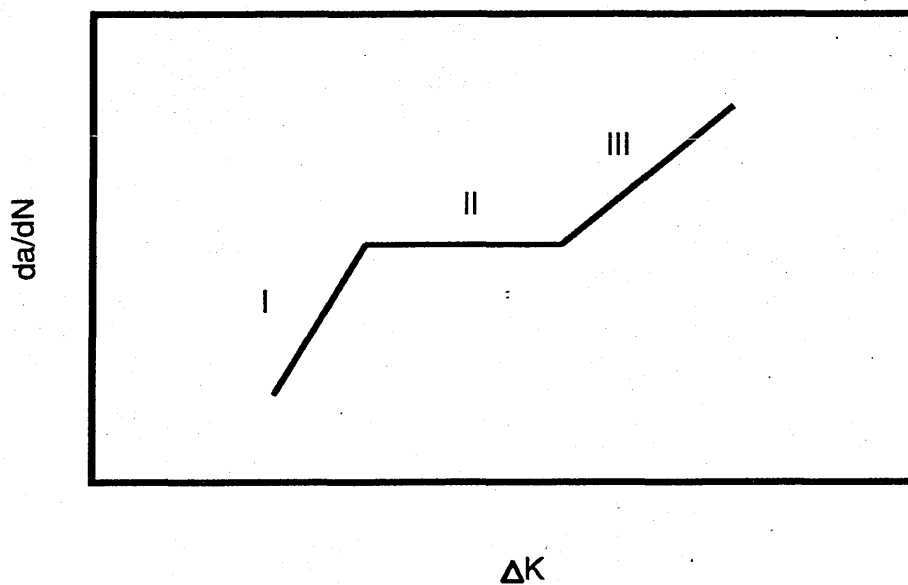


Figure 19 SCC crack growth as a function of applied ΔK .

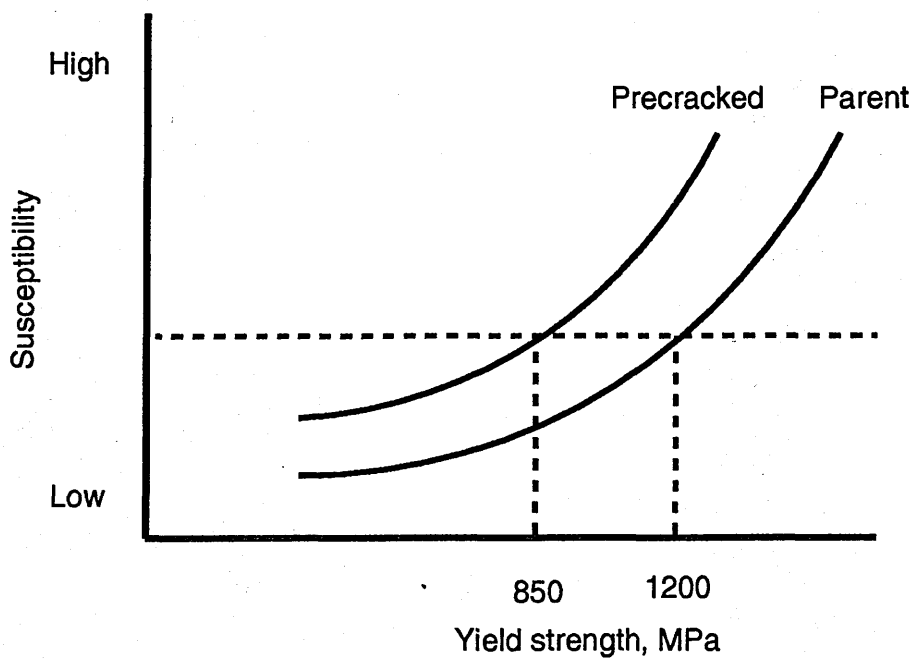


Figure 20 SCC susceptibility for pre-cracked and plain specimens

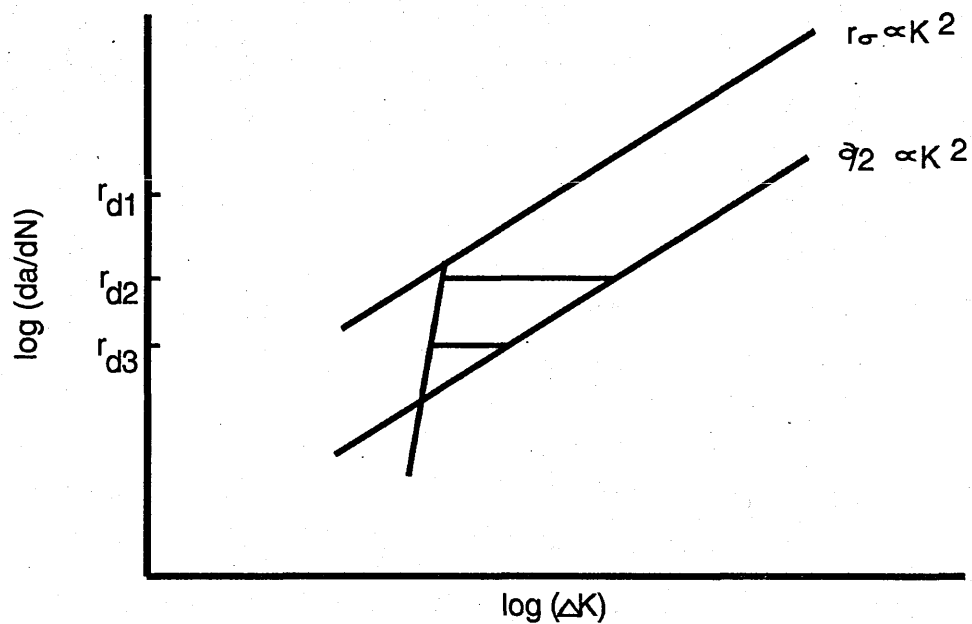


Figure 21 Schematic hydrogen embrittlement assisted fatigue crack growth.

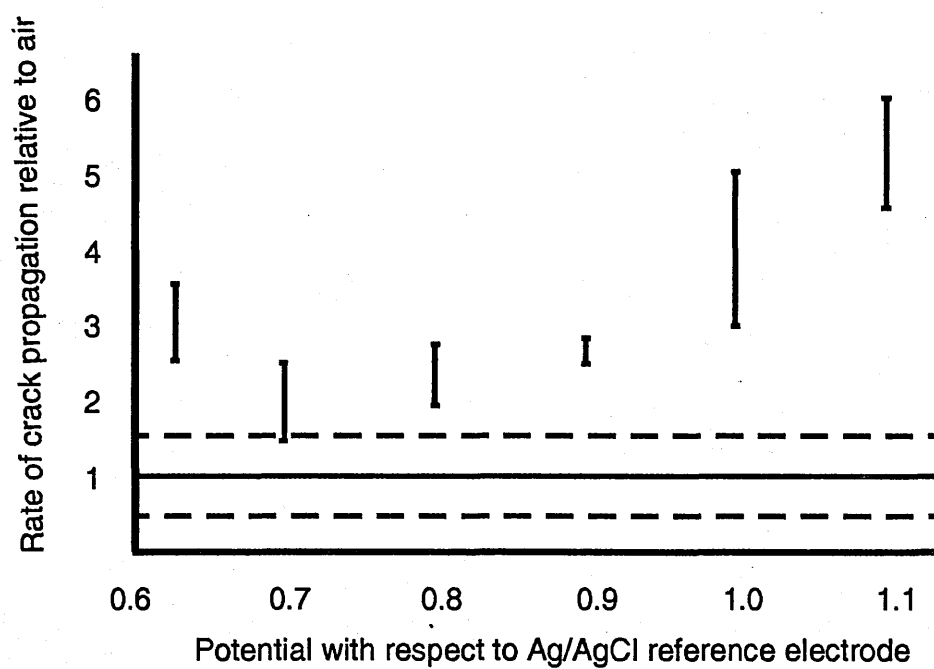


Figure 22 Summary of the effect of potential on the fatigue crack growth rate.

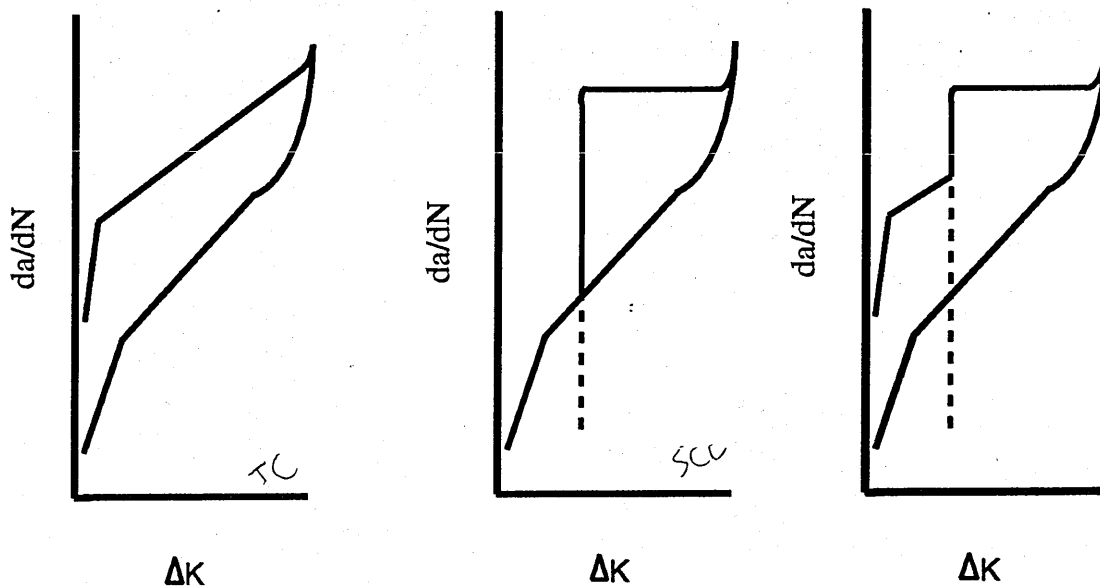


Figure 23 Schematic representation of the basic types of corrosion fatigue crack growth behaviour.

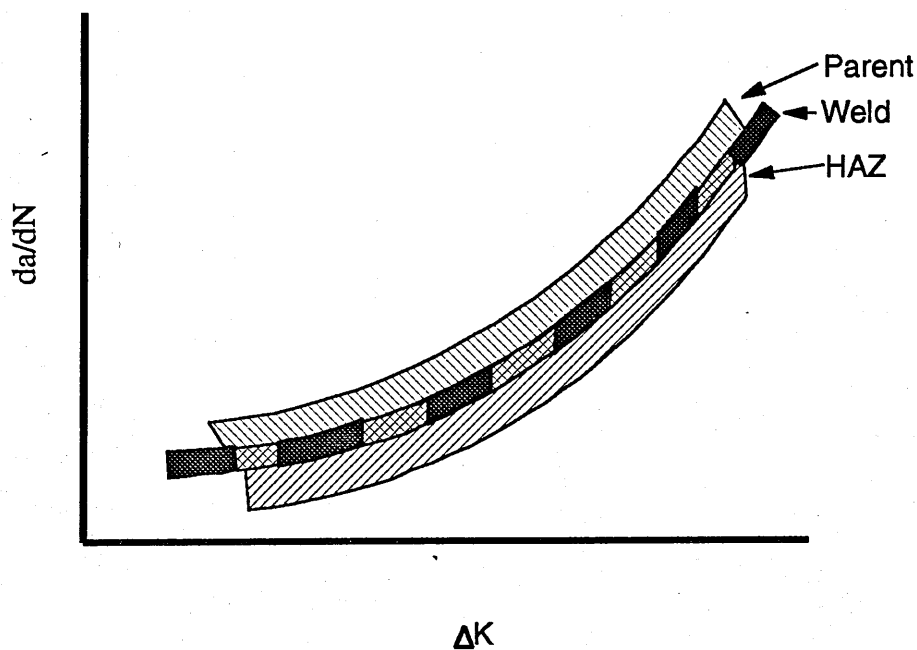


Figure 24 A comparison between the fatigue crack growth rates obtained from a structural steel weldment.

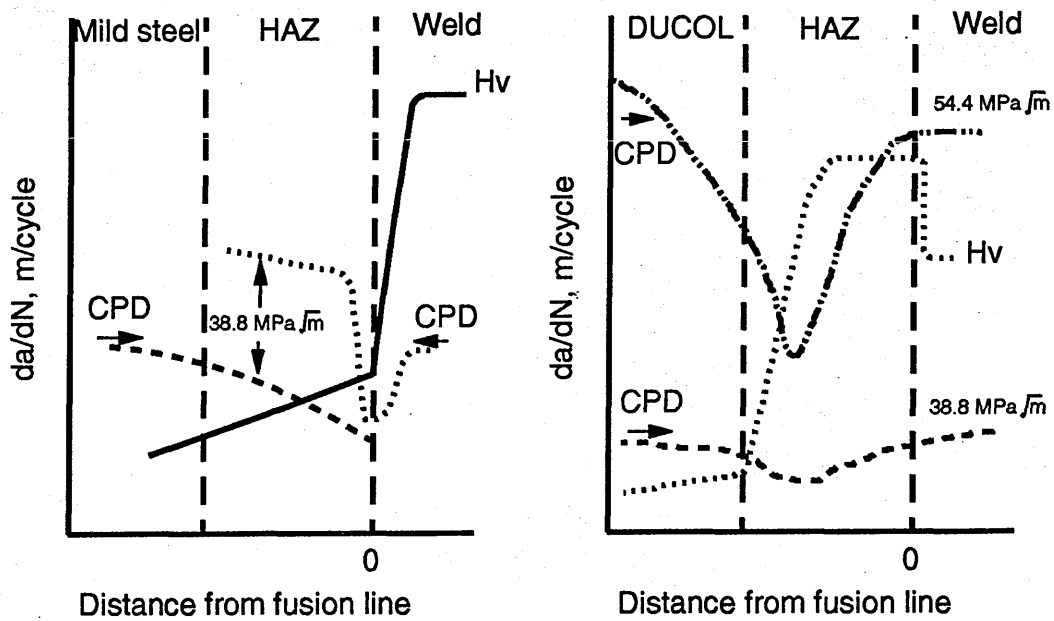


Figure 25 Crack growth rates as a function of distance from the fusion boundary and hardness.

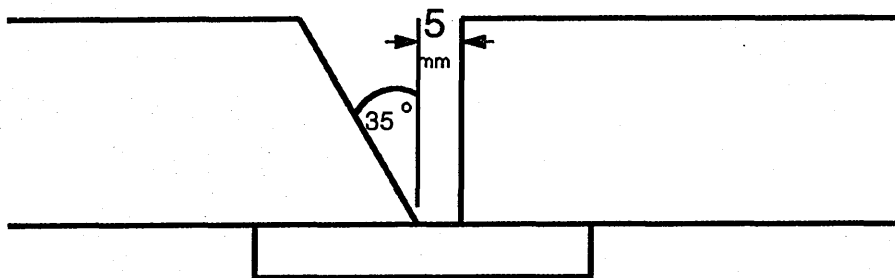


Figure 26 Weld Plate Preparation

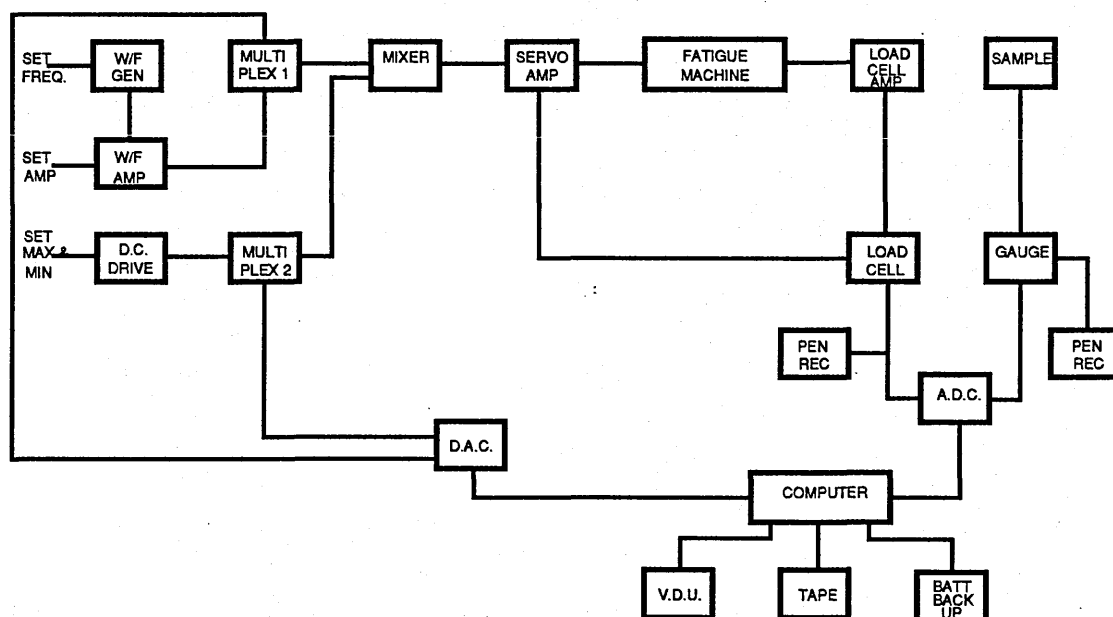


Figure 27 Schematic representation of the constant ΔK fatigue testing machine.

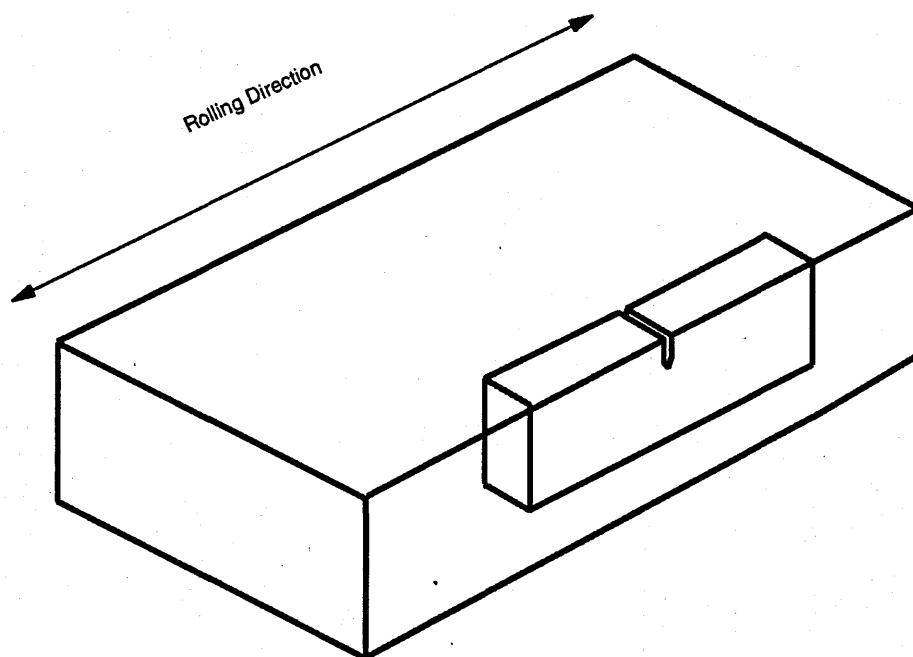


Figure 28 Notch position of a sample relative to the parent plate.

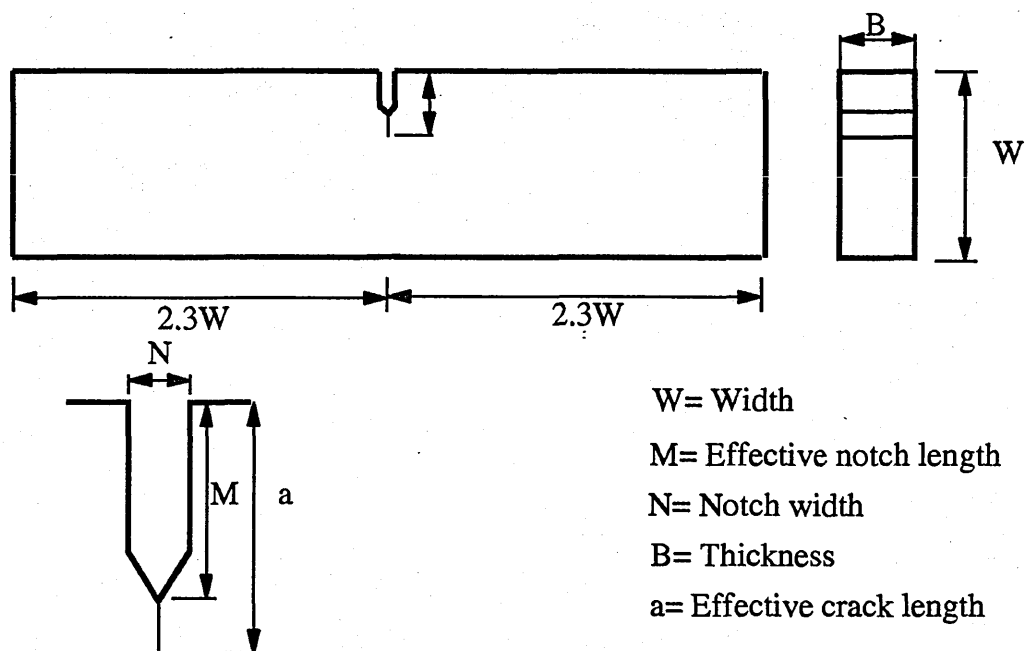


Figure 29 Specimen geometry according to BS6835:1988

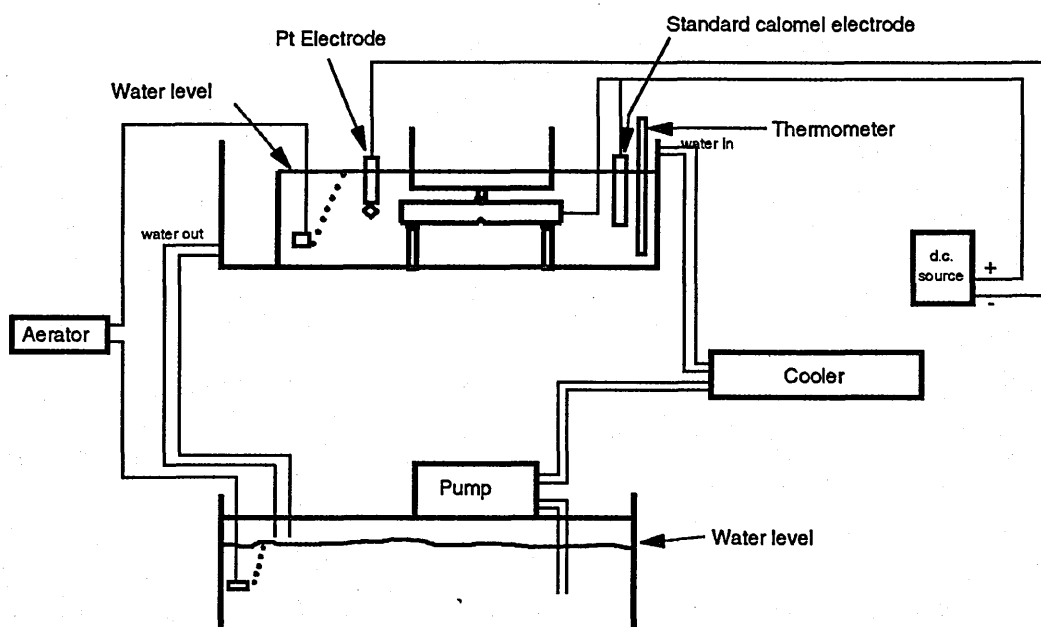


Figure 30 Circulating seawater system for corrosion fatigue.

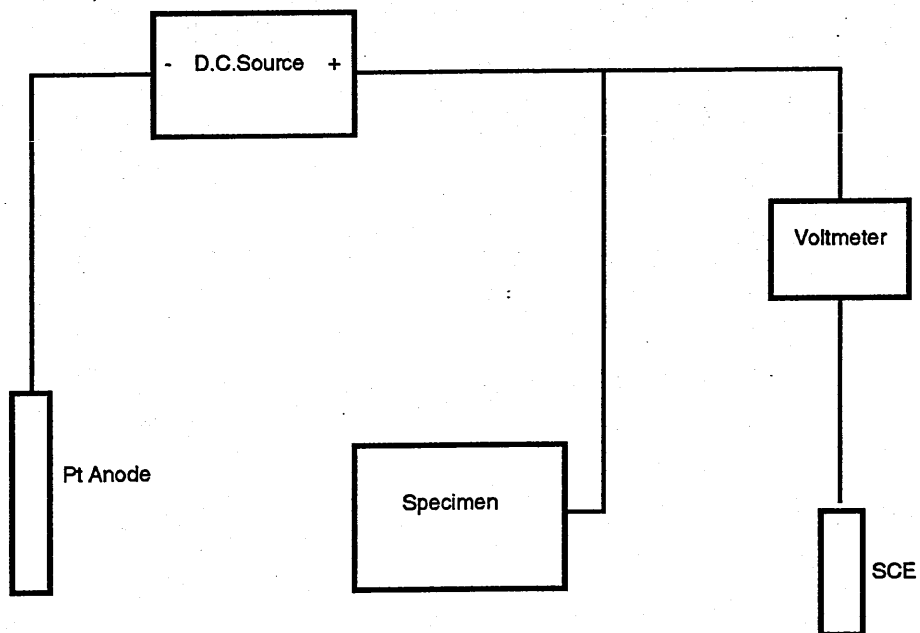


Figure 31 *Electrical connections for cathodic protection of specimens*

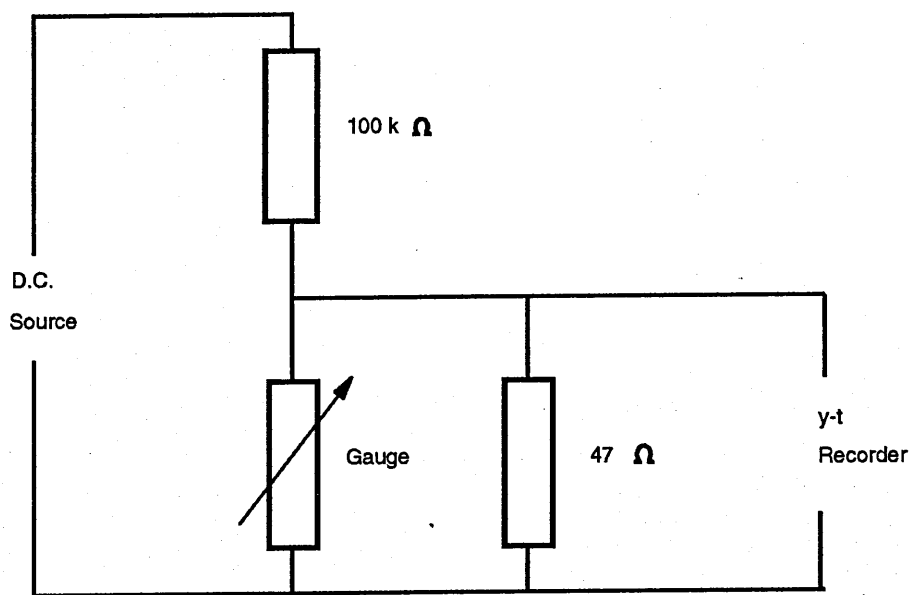


Figure 32 *Fatigue crack growth monitoring circuit.*

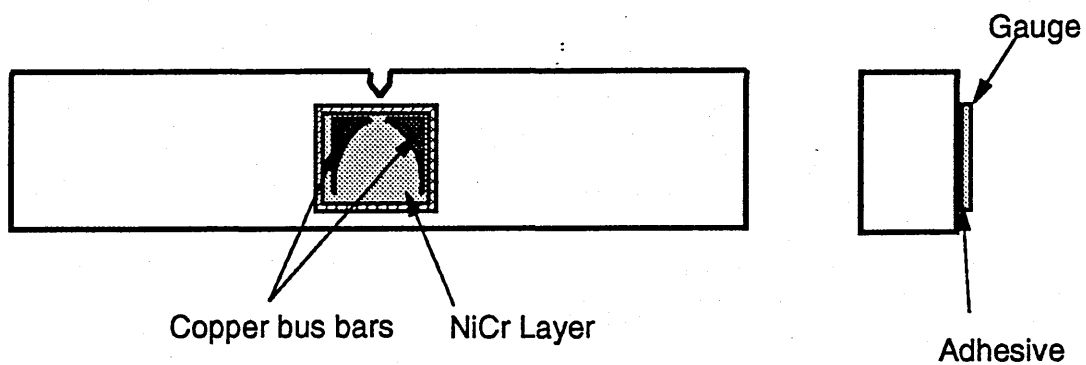


Figure 33 Location and geometry of constant ΔK crack monitoring gauges

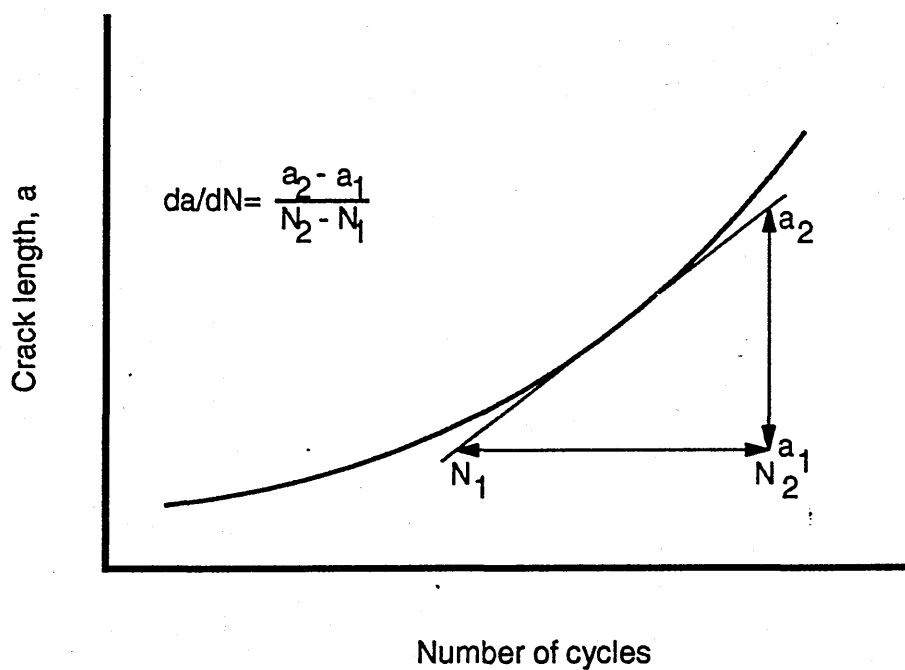


Figure 34 Crack length versus number of cycles curve.

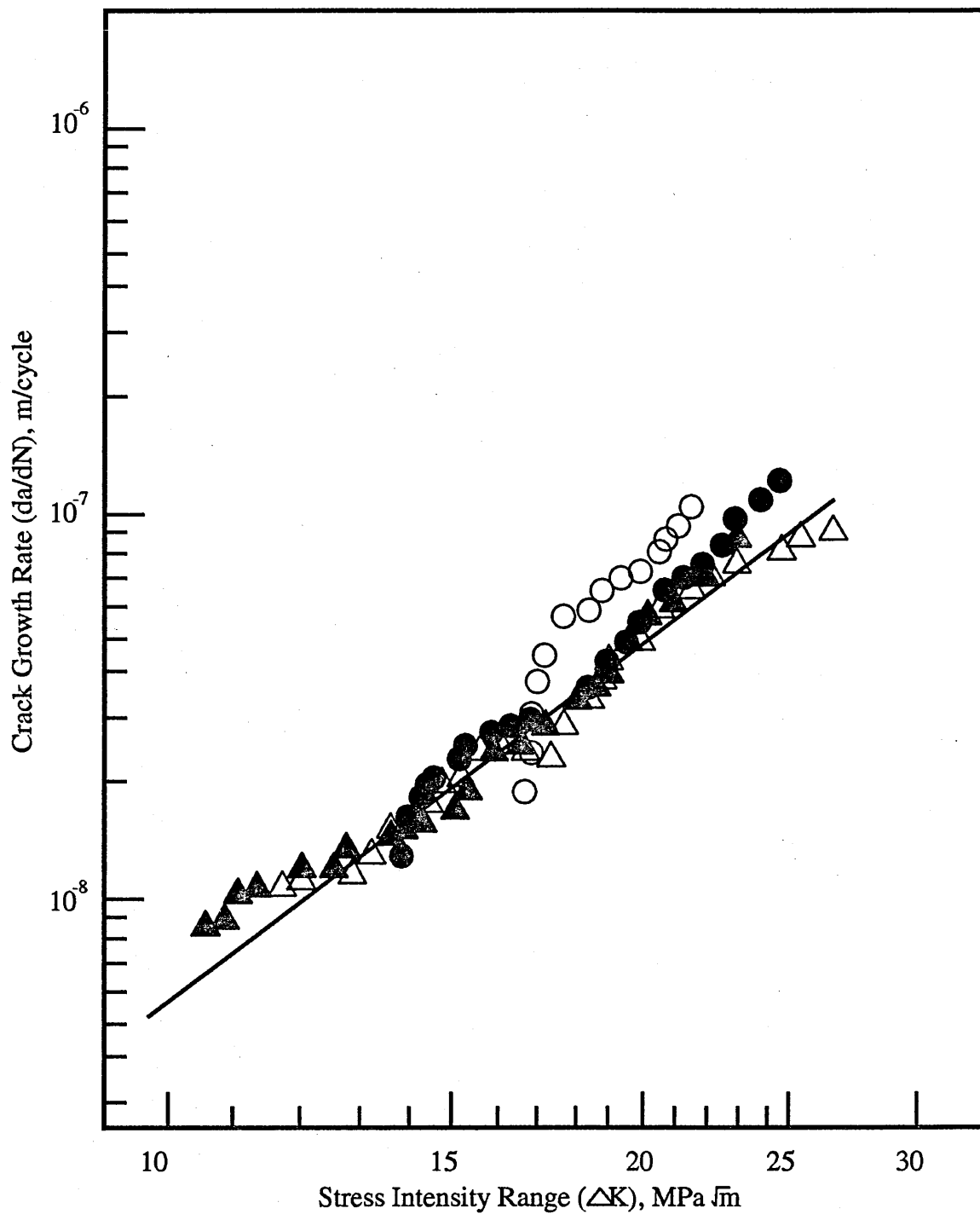


Figure 35 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 parent plate in-air.

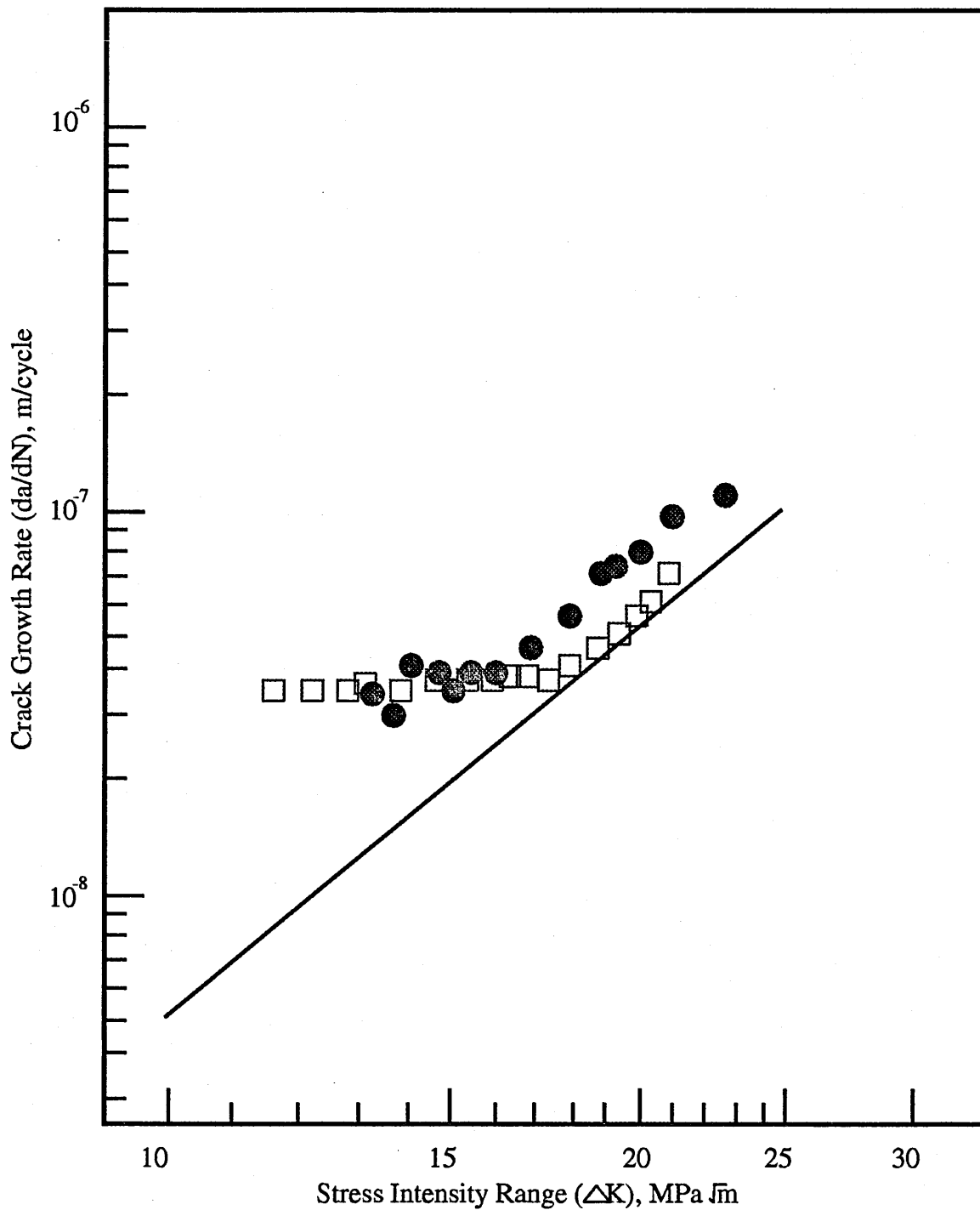


Figure 36 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 parent plate for free corrosion.

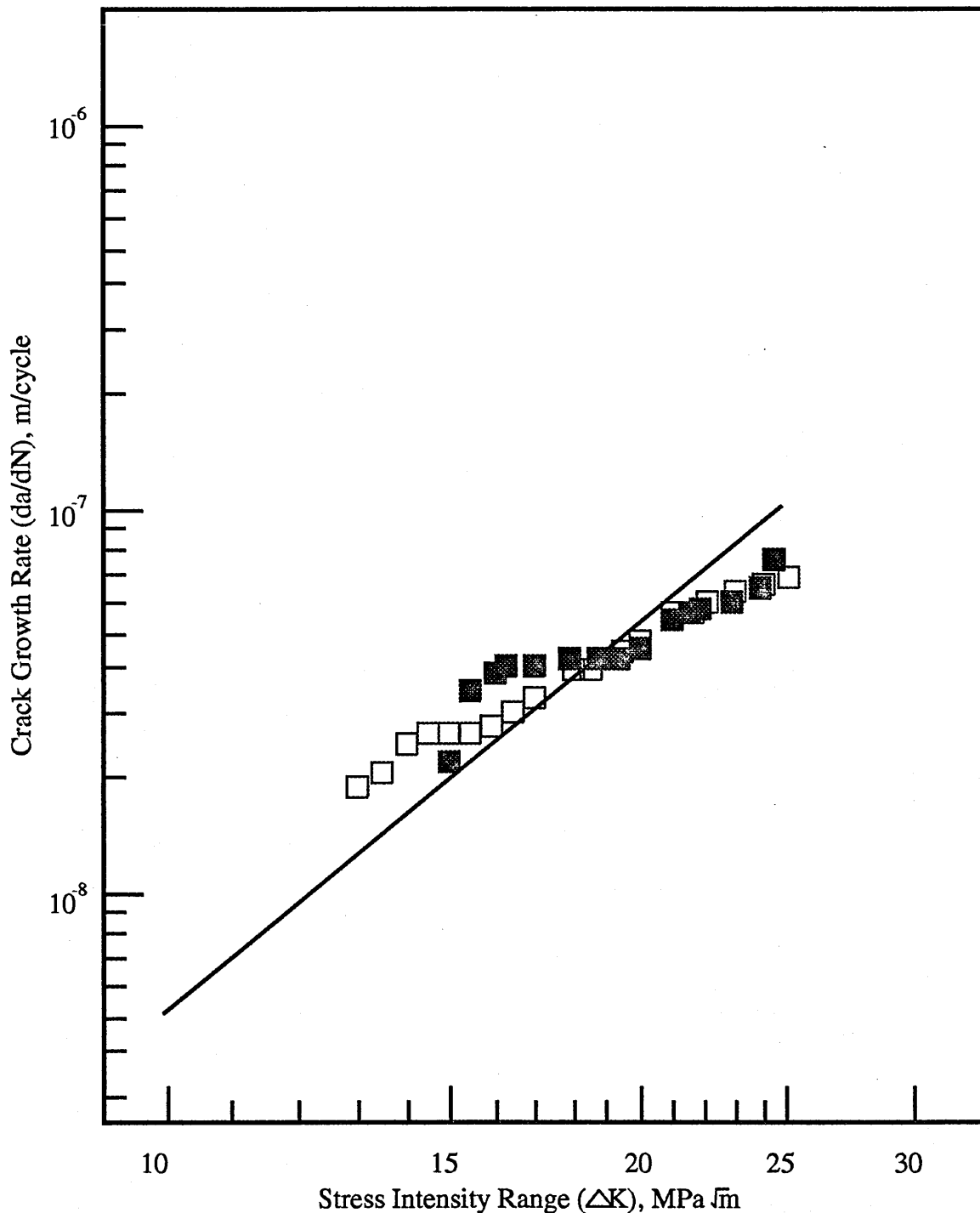


Figure 37 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 parent plate, $P_v = -800$ mV.

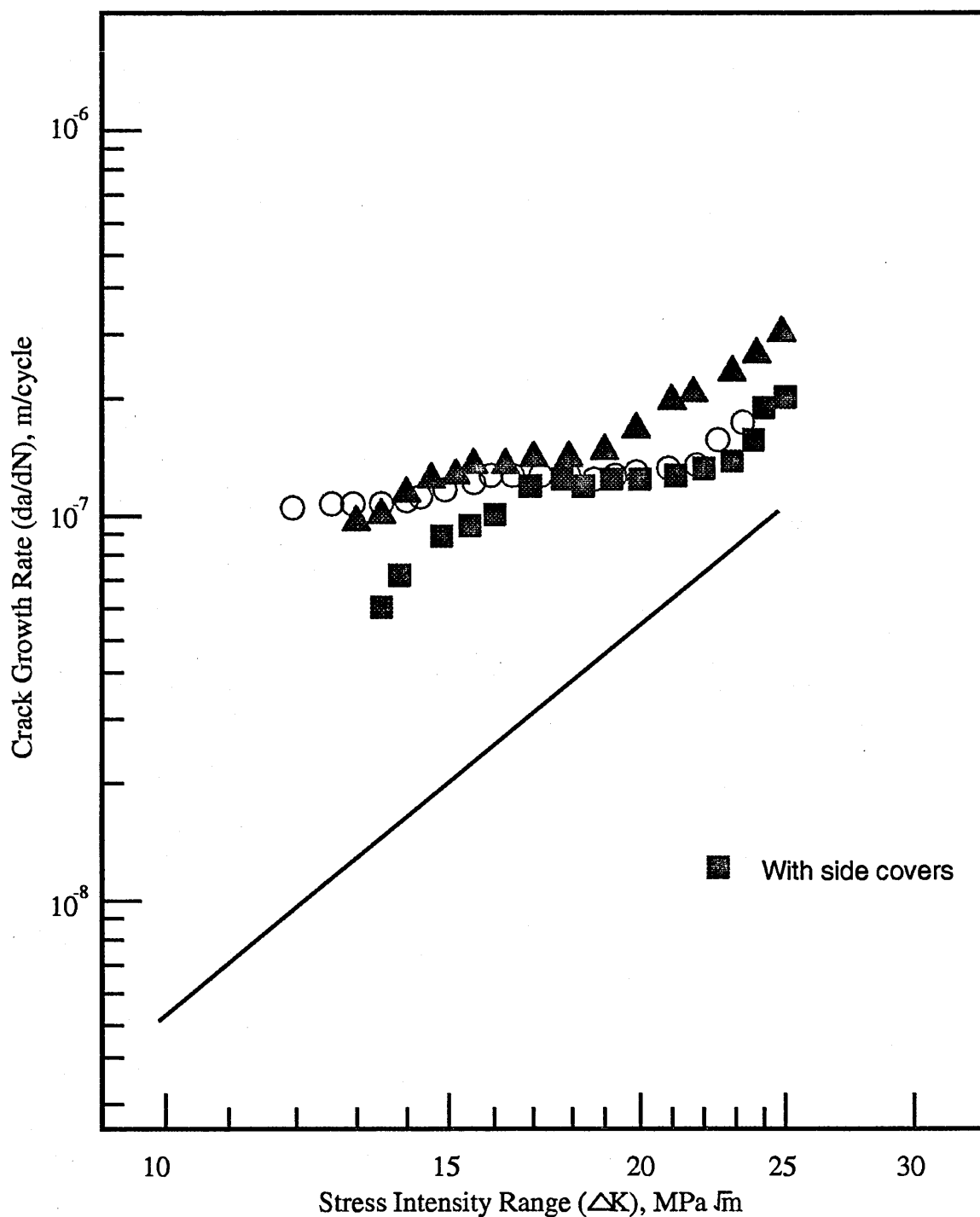


Figure 38 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 parent plate, $P_v = -1100$ mV.

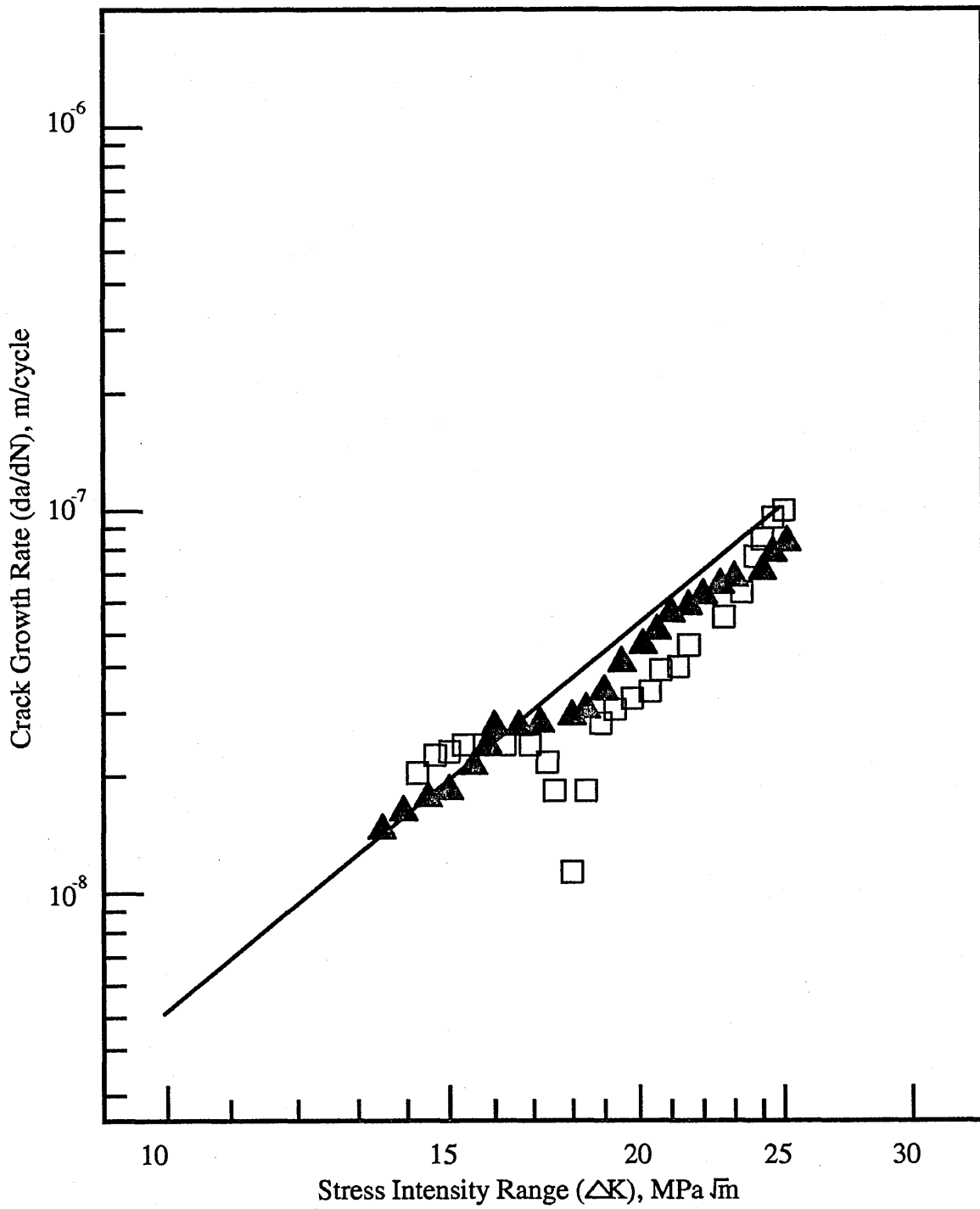


Figure 39 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 HAZ in-air.

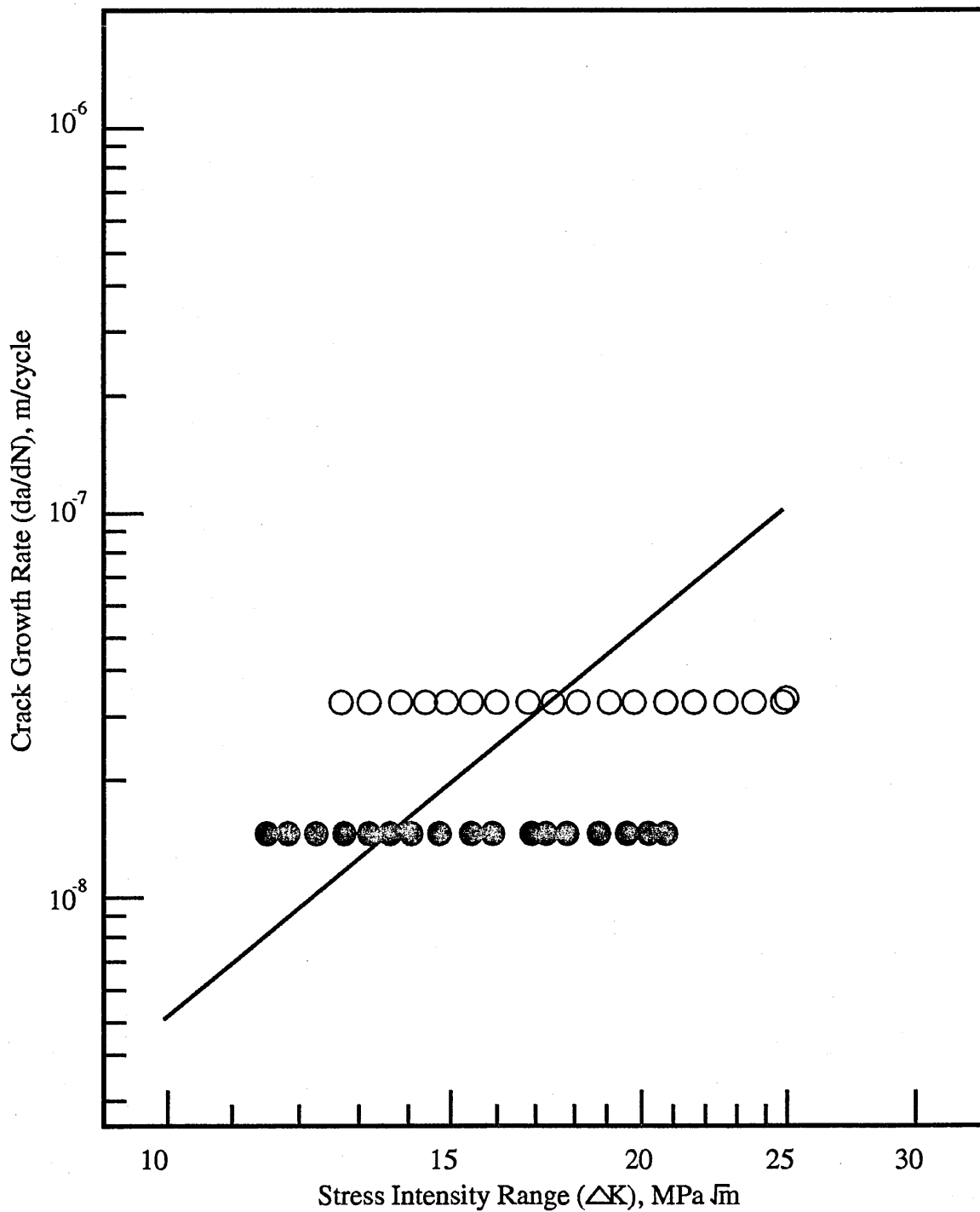


Figure 40 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 HAZ for free corrosion.

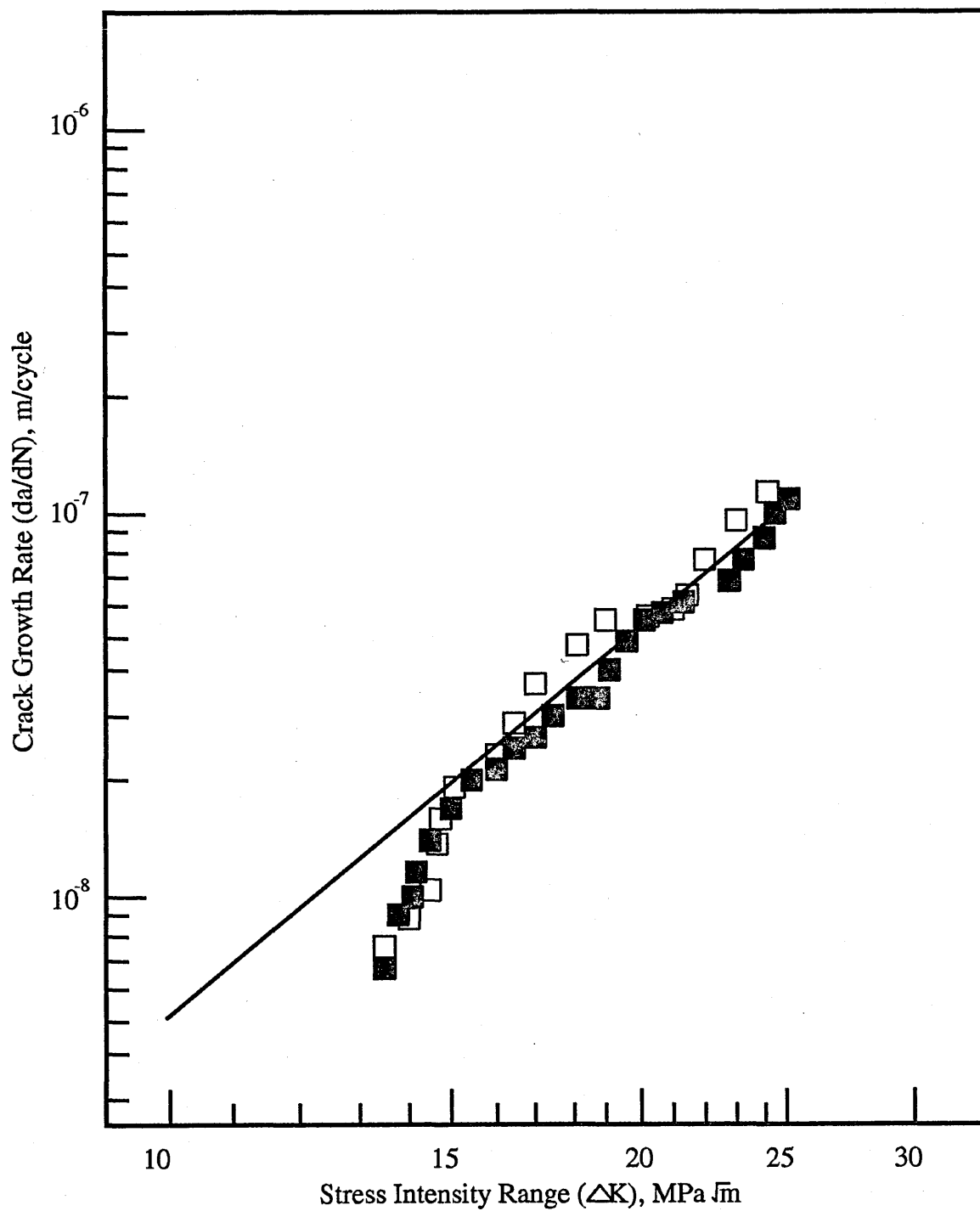


Figure 41 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 HAZ, $P_v = -800 \text{ mV}$.

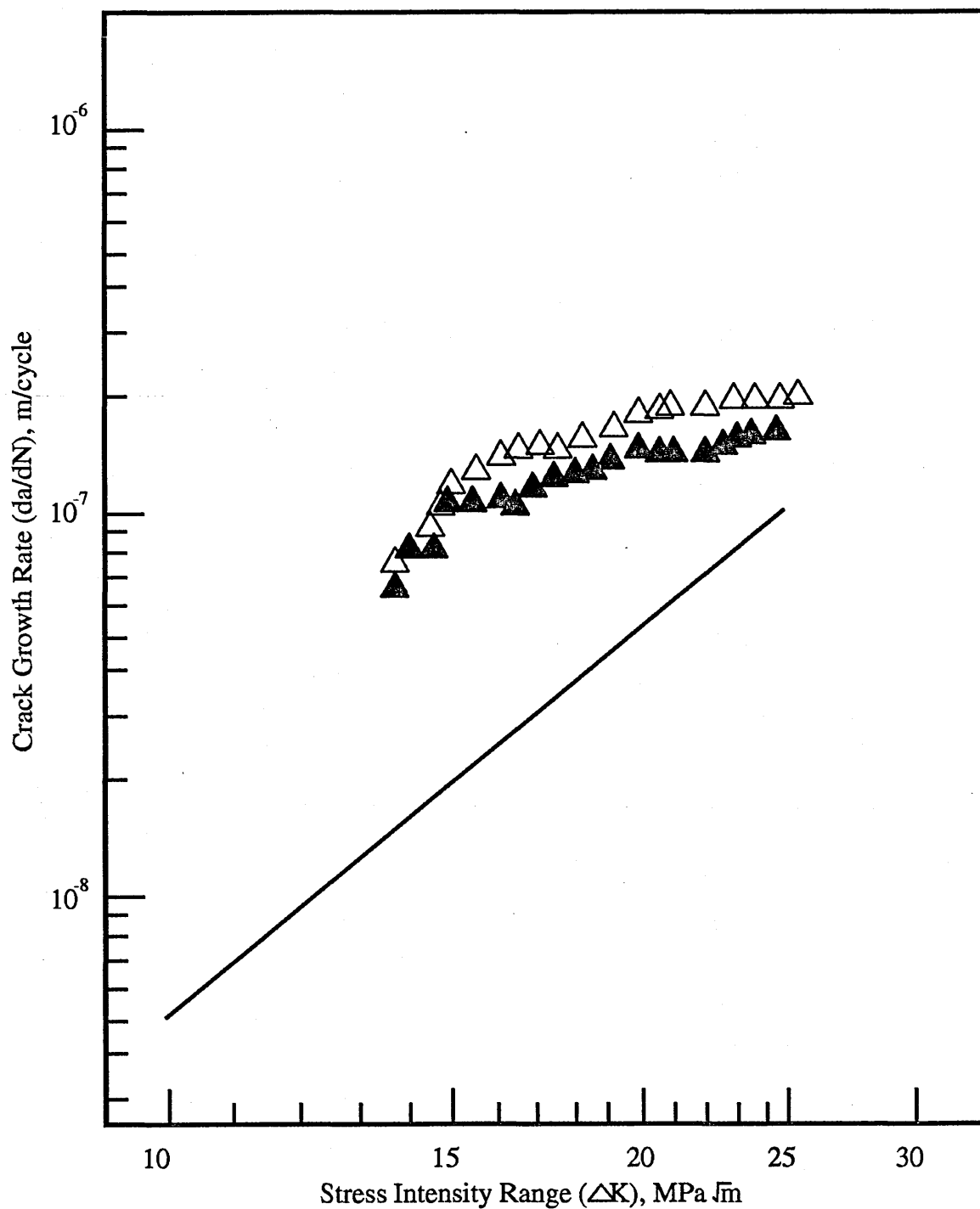


Figure 42 Graph of $\log da/dN$ versus $\log \Delta K$ for Superelso 500 HAZ, $P_v = -1100$ mV.

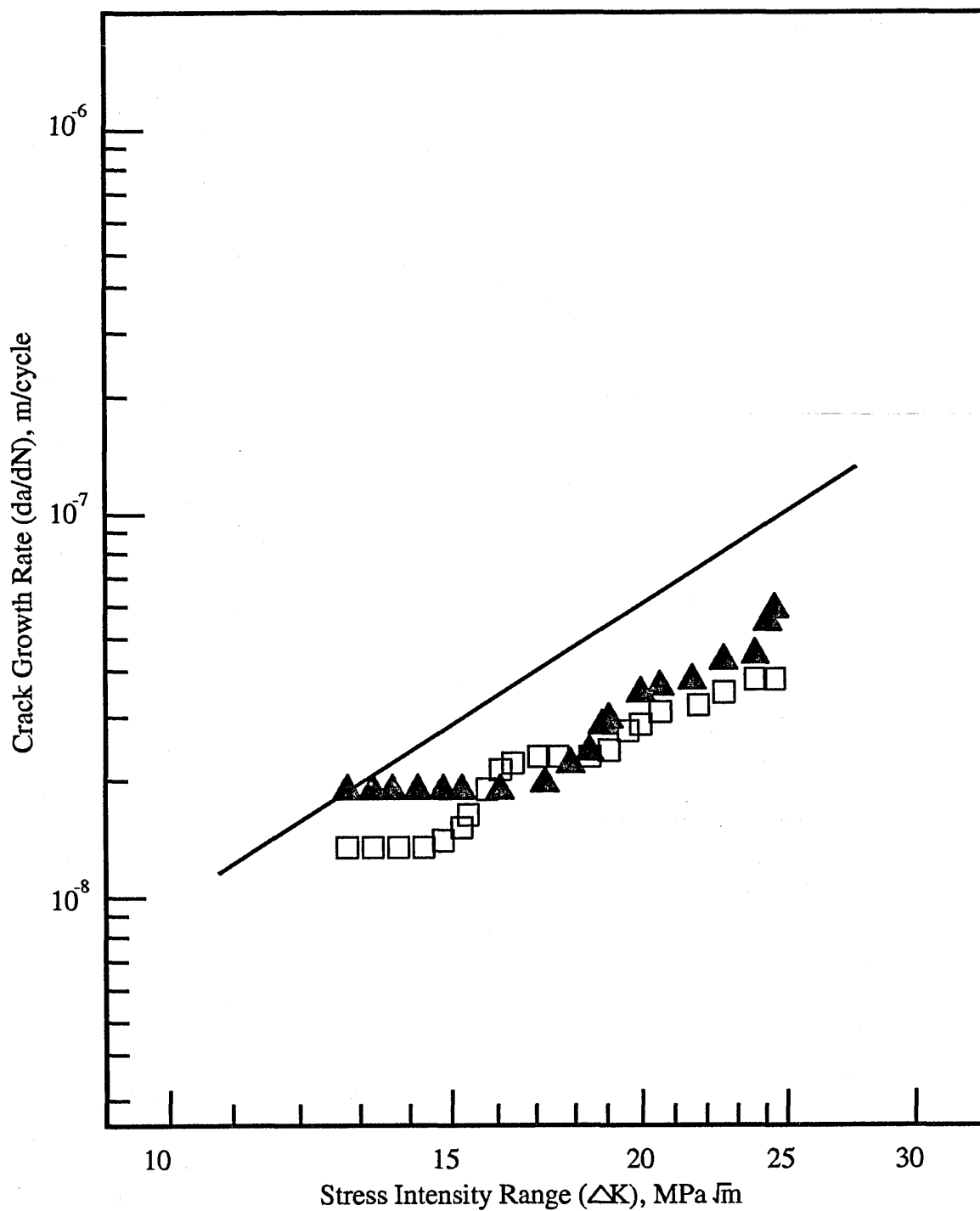


Figure 43 Graph of $\log da/dN$ versus $\log \Delta K$ for Mannesman X70 HAZ for free corrosion.

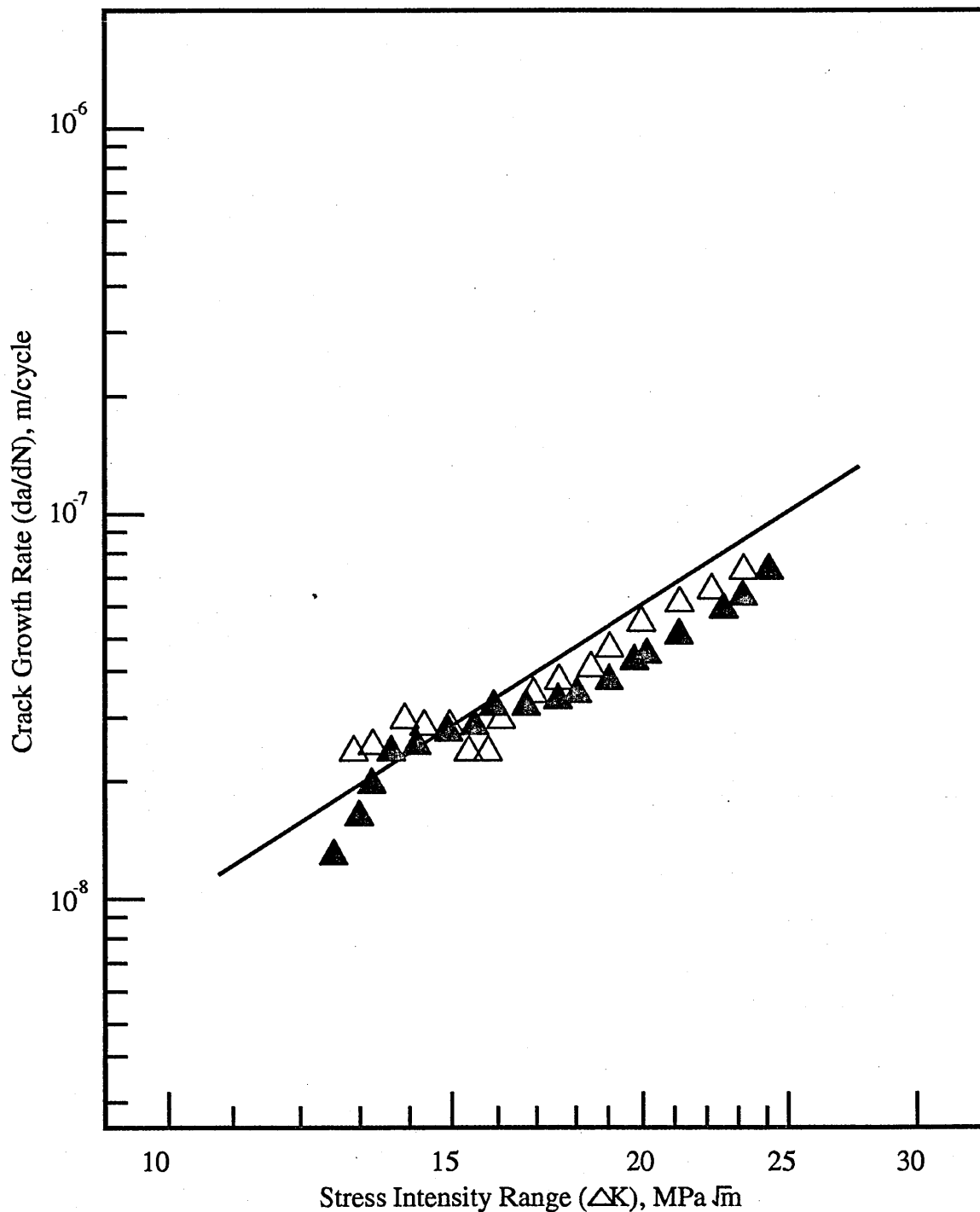


Figure 44 Graph of $\log da/dN$ versus $\log \Delta K$ for Mannesman X70 HAZ, $P_v = -800$ mV.

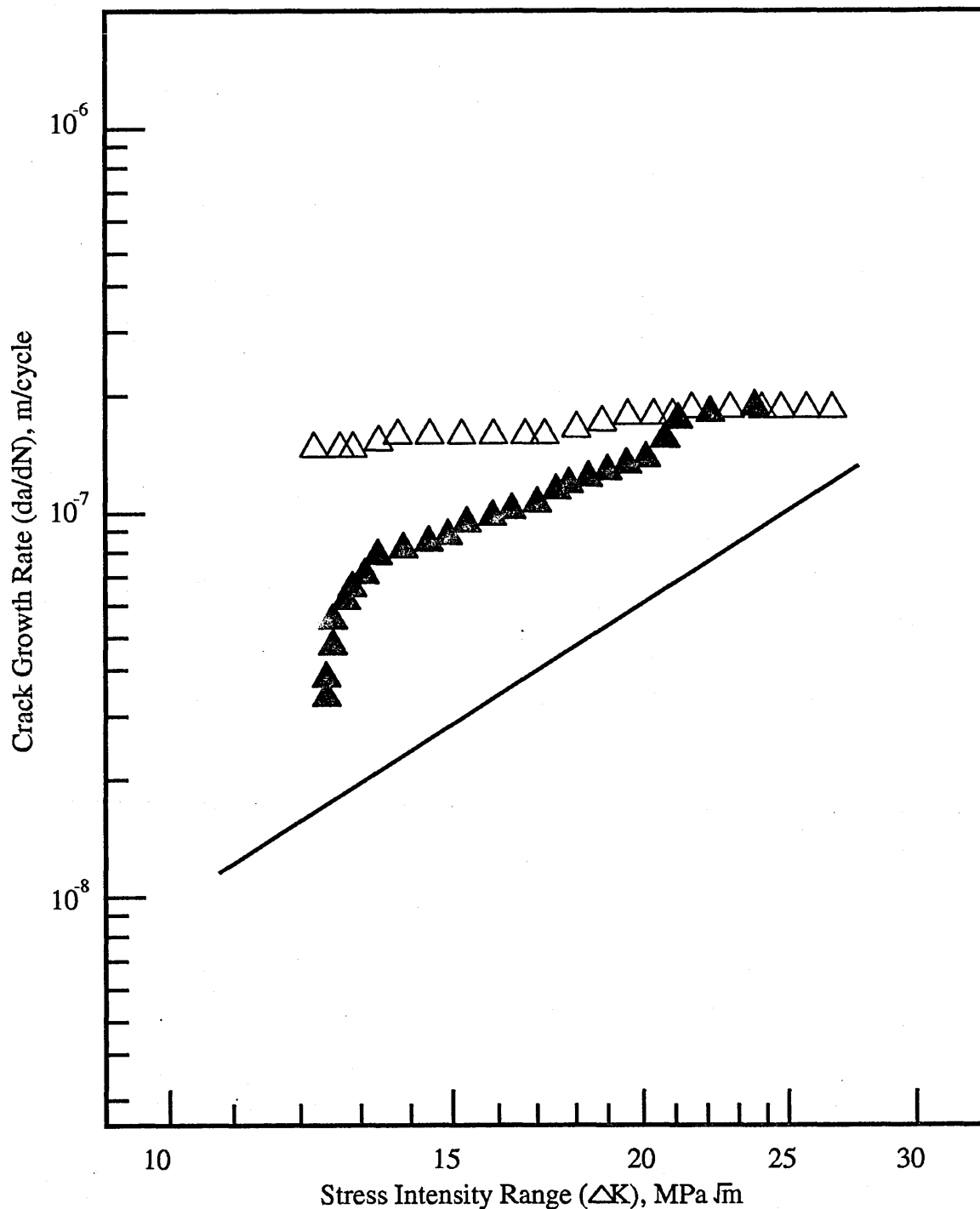
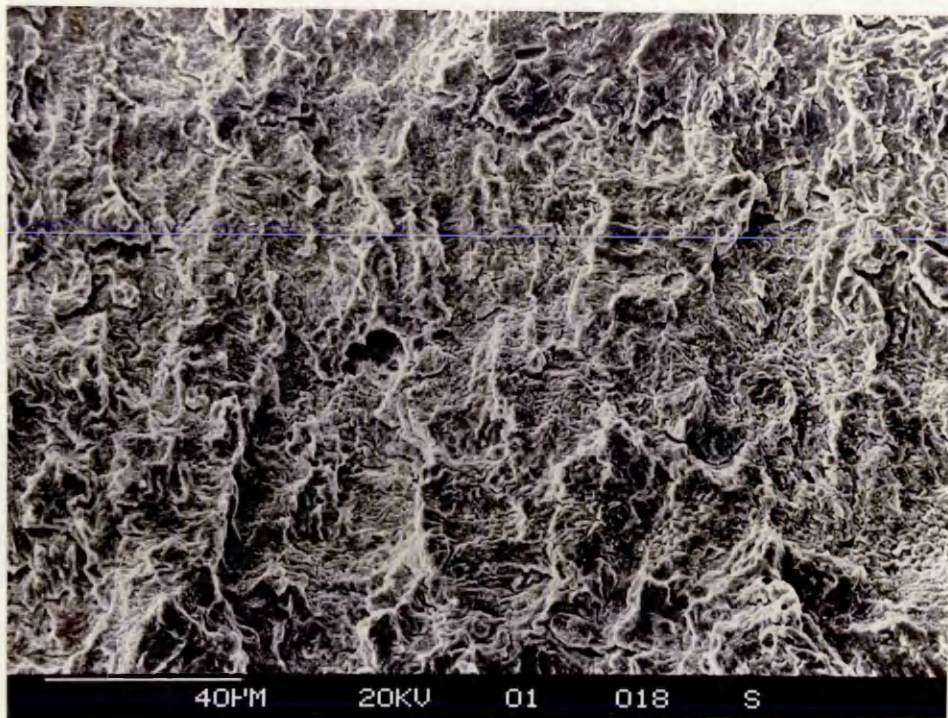
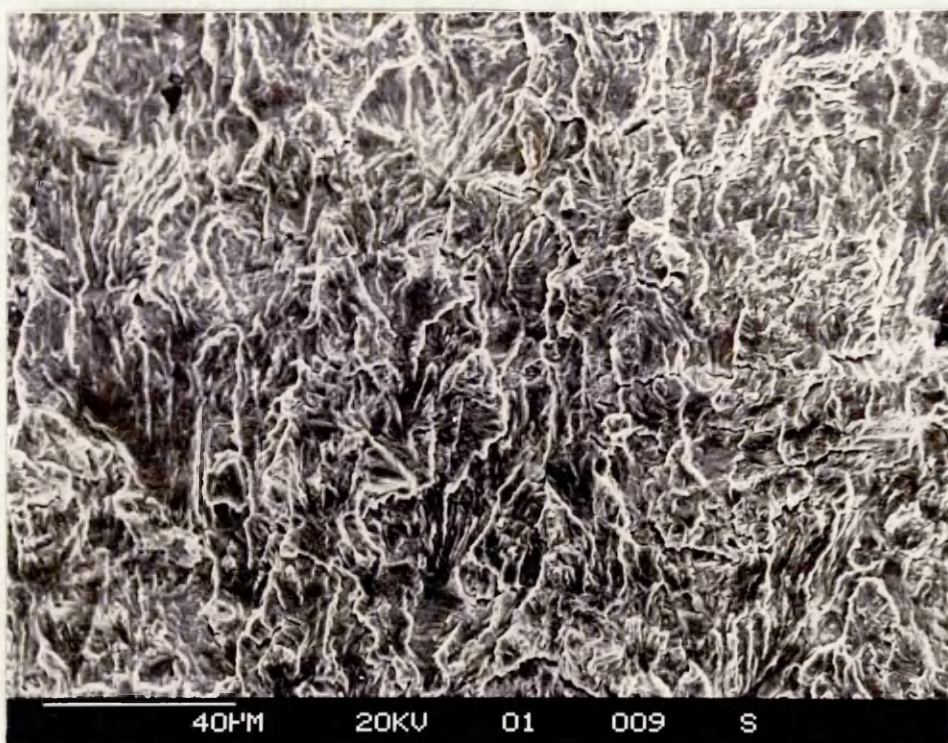


Figure 45 Graph of $\log da/dN$ versus $\log \Delta K$ for Mannesman X70 HAZ, $P_v = -1100$ mV.



X 625

Figure 46 *Fracture surface for Superelso 500 parent plate tested in-air, low ΔK .*



X 625

Figure 47 *Fracture surface for Superelso 500 parent plate tested in-air, high ΔK .*

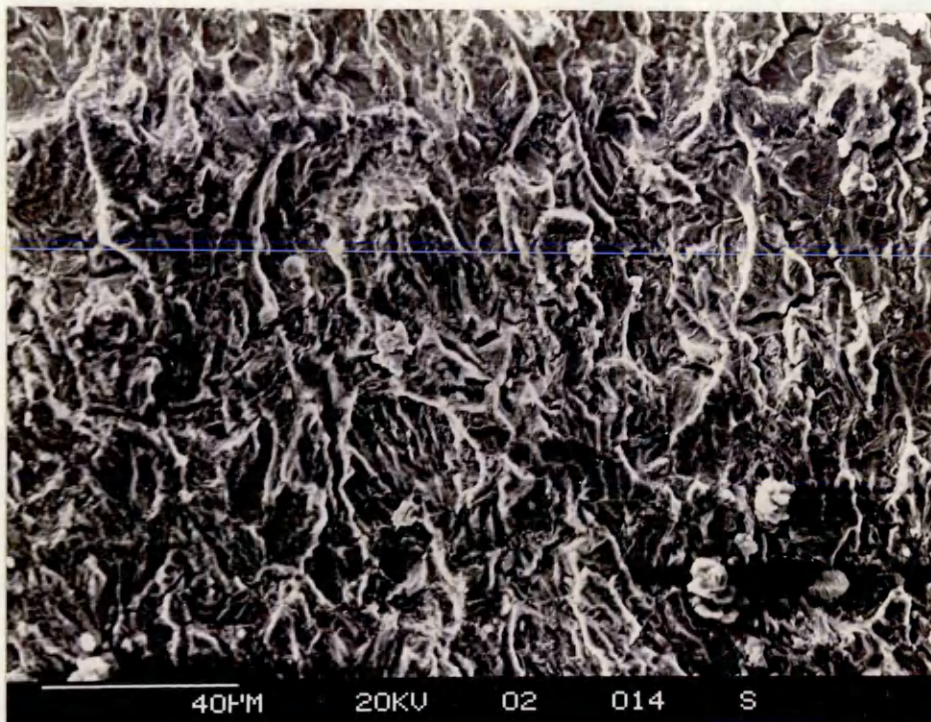


Figure 48 *Fracture surface for Superelso 500 parent plate tested under free corrosion conditions, low ΔK .*

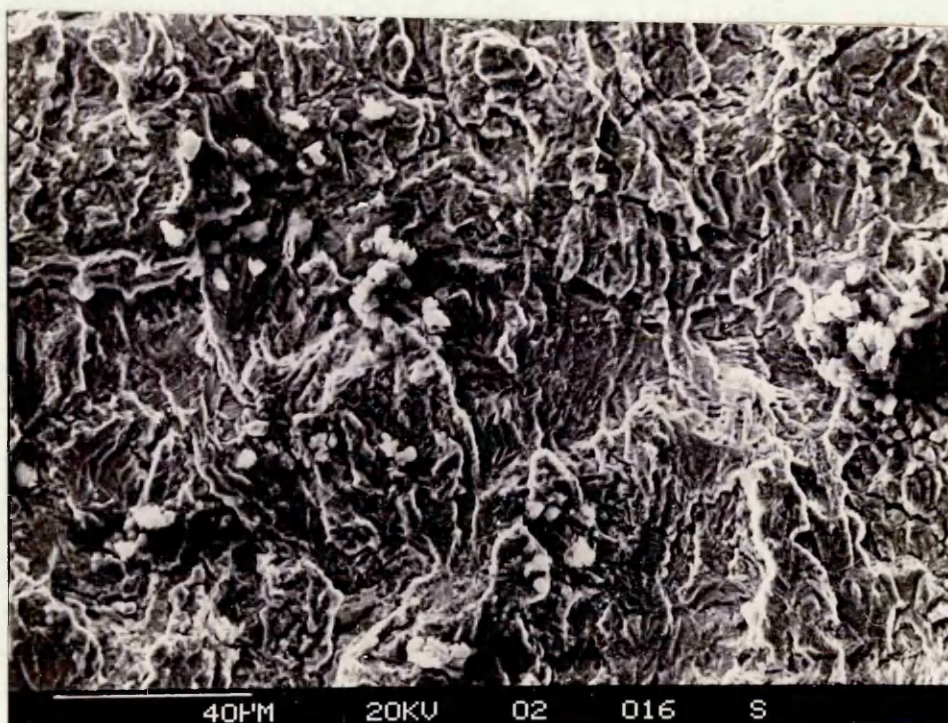
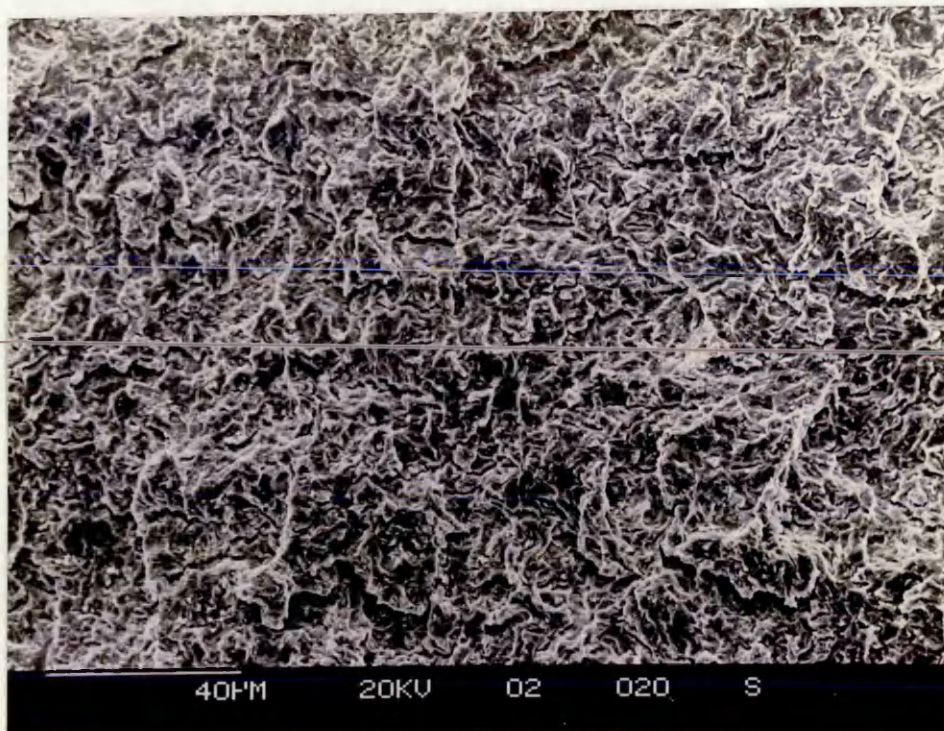
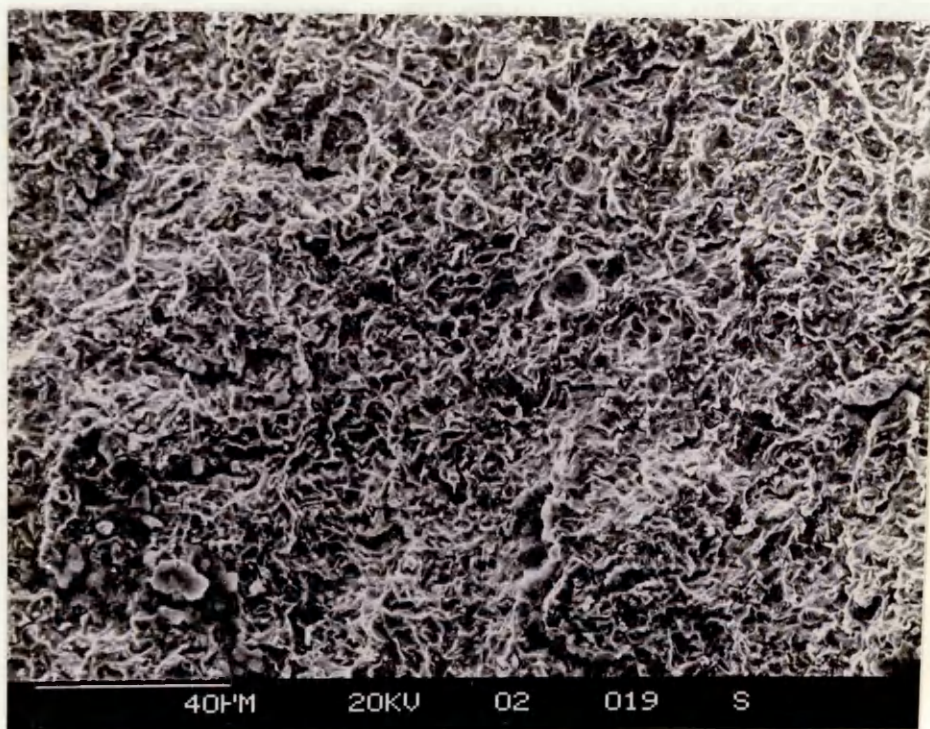


Figure 49 *Fracture surface for Superelso 500 parent plate tested under free corrosion conditions, high ΔK .*



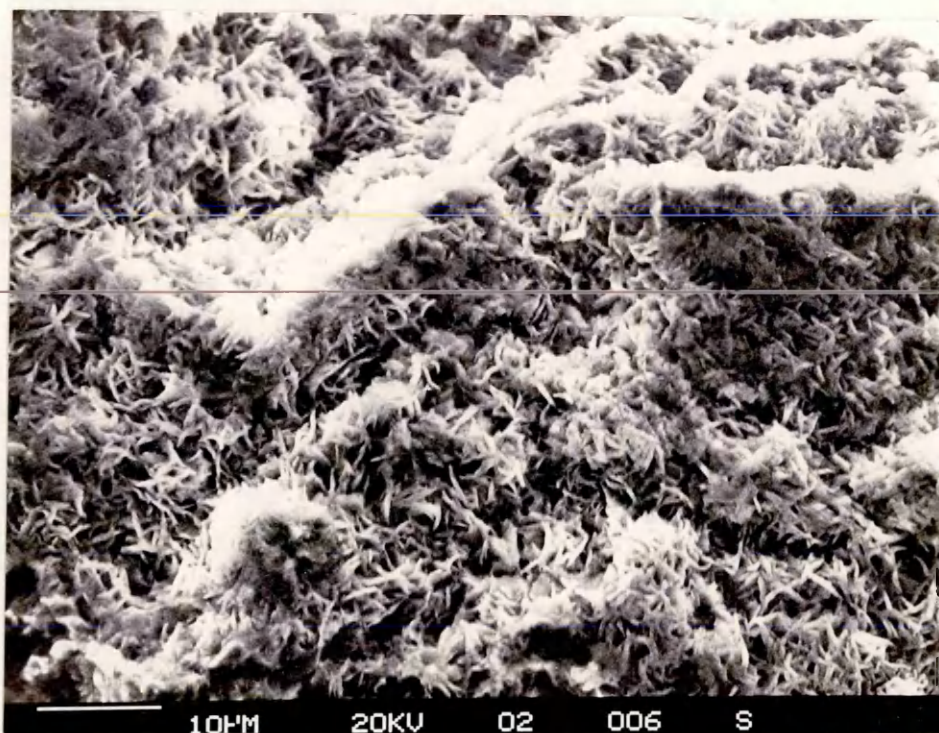
X 625

Figure 50 Fracture surface for Superelso 500 parent plate tested with an applied potential, $P_v = -800$ mV, low ΔK .



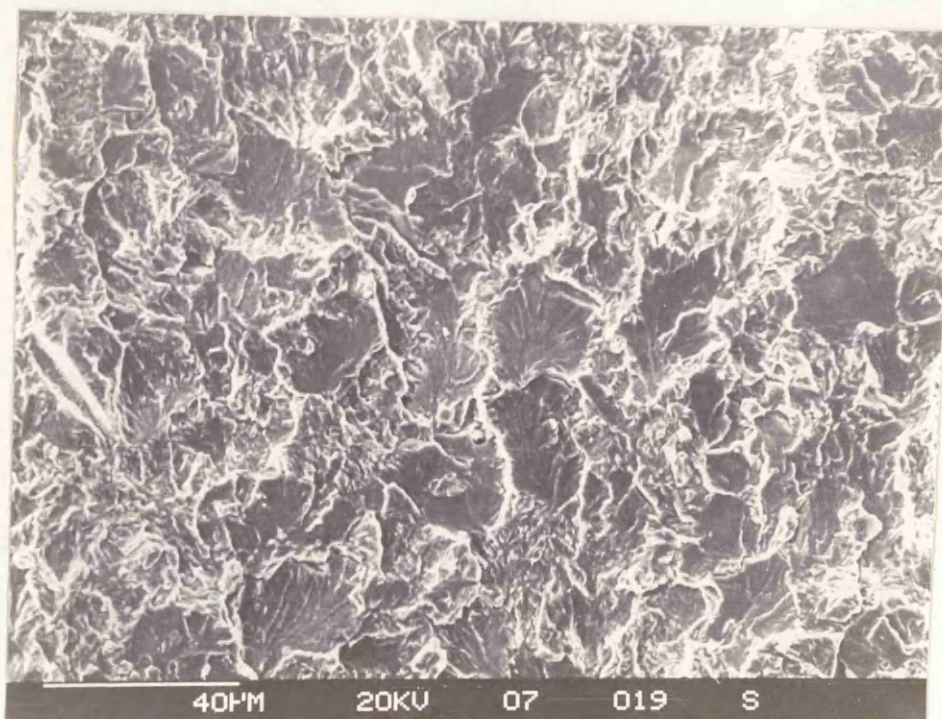
X 625

Figure 51 Fracture surface for Superelso 500 parent plate tested with an applied potential, $P_v = -800$ mV, high ΔK .



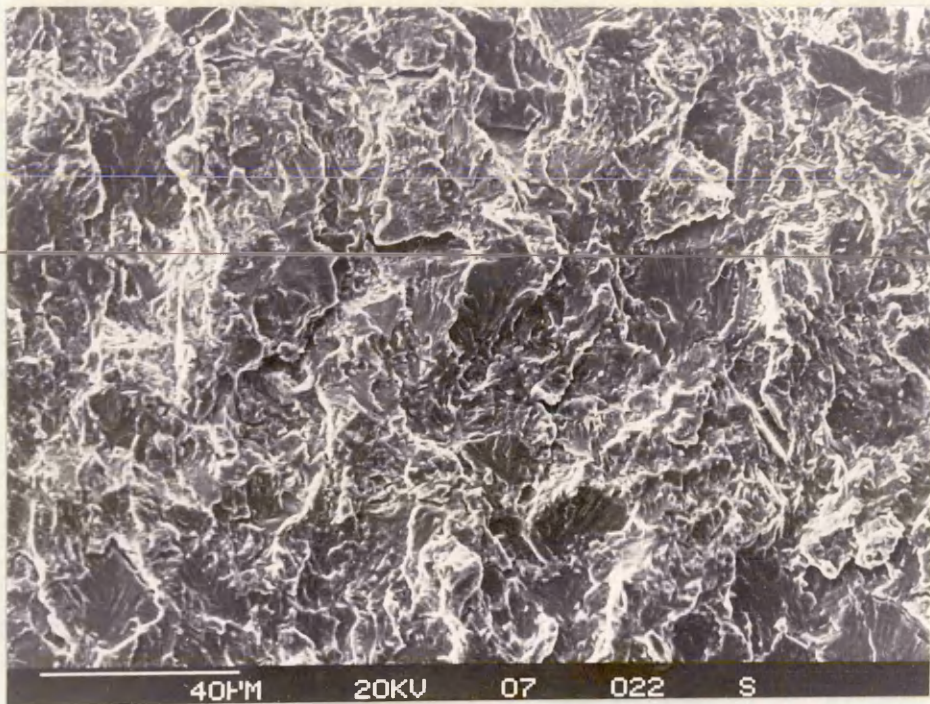
X1650

Figure 52 *Fracture surface for Superelso 500 parent plate tested with an applied potential, $P_v = -1100$ mV, low ΔK with calcareous deposit.*



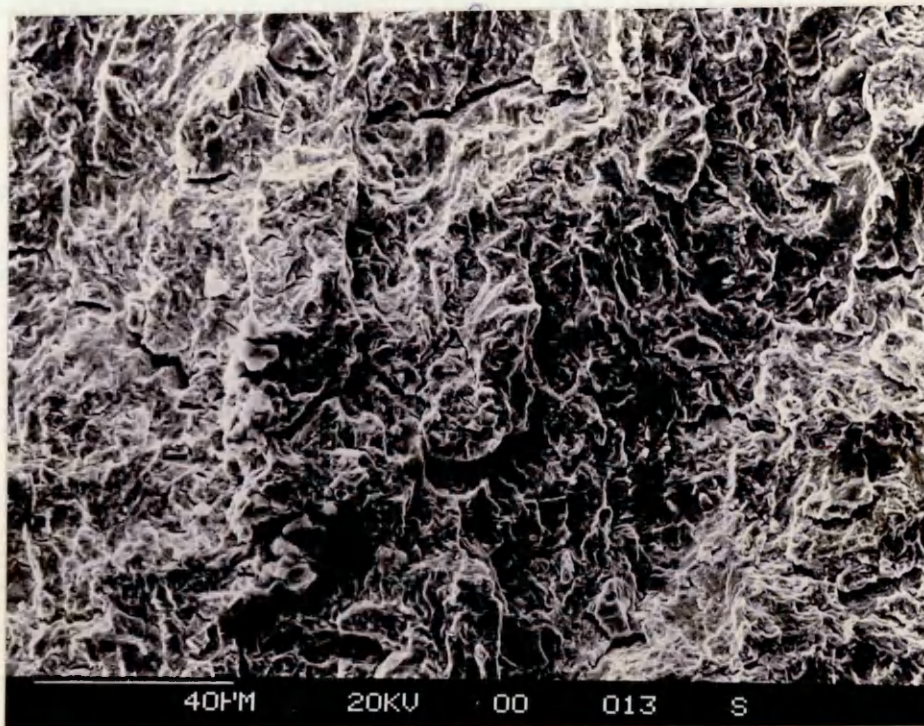
X625

Figure 53 *Fracture surface for Superelso 500 parent plate tested with an applied potential, $P_v = -1100$ mV, low ΔK with deposit removed.*



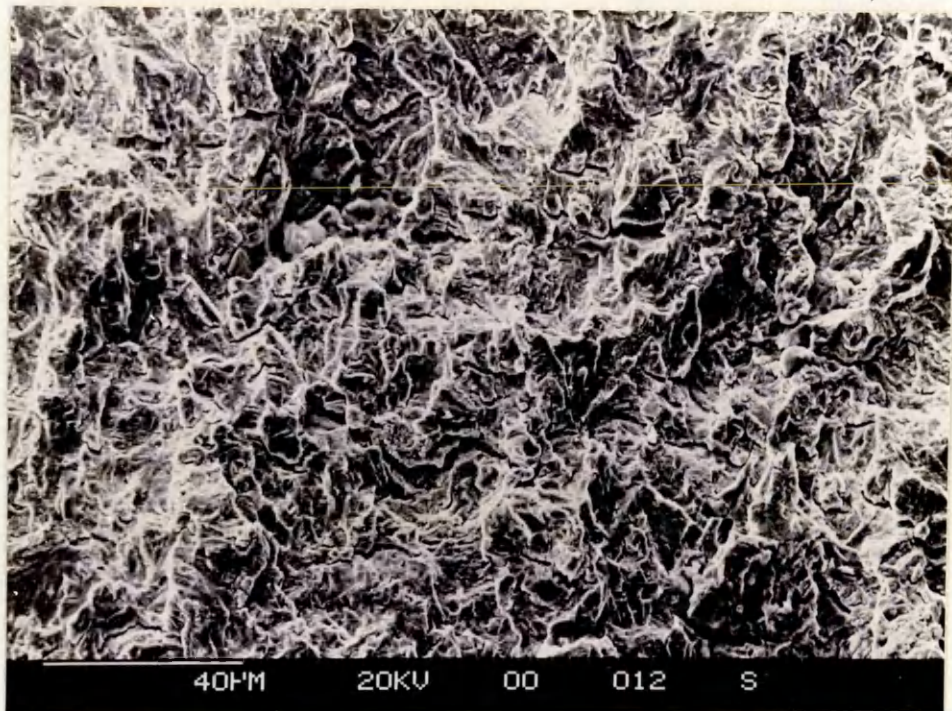
X 625

Figure 54 *Fracture surface for Superelso 500 parent plate tested with an applied potential, $P_v = -1100$ mV, high ΔK with calcareous deposit removed.*



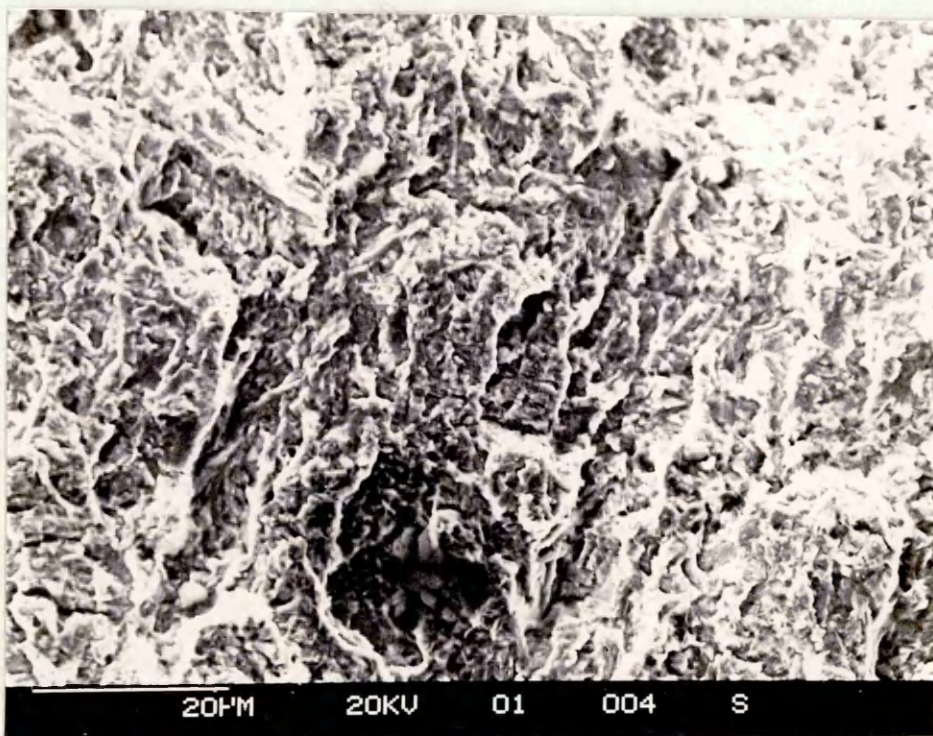
X 625

Figure 55 *Fracture surface for the Superelso 500 HAZ fatigue tested in-air, low ΔK .*



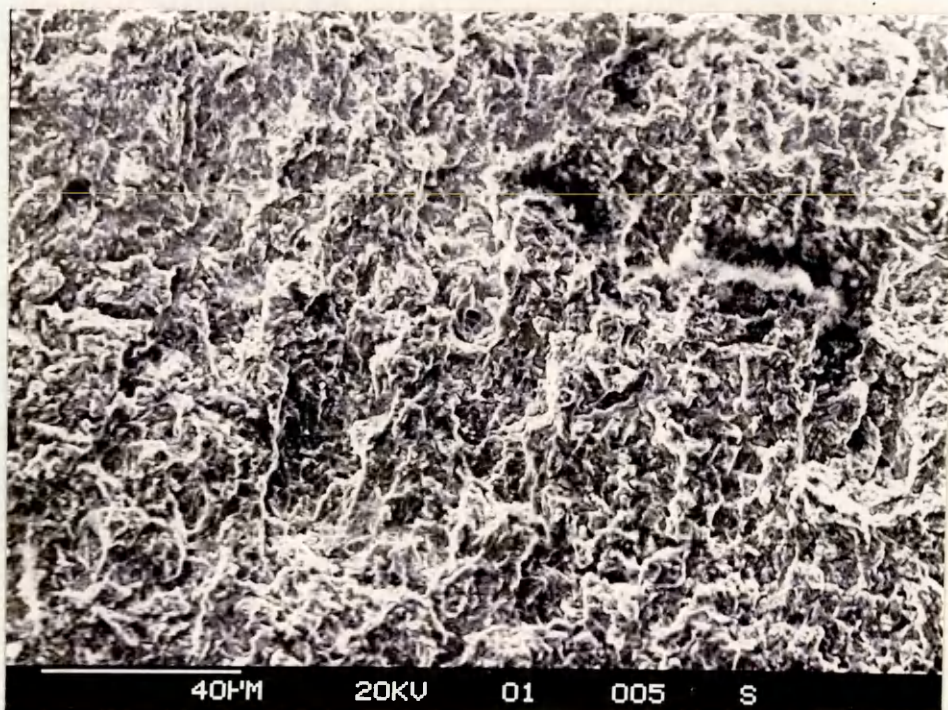
X 625

Figure 56 Fracture surface for the Superelso 500 HAZ fatigue tested in-air, high ΔK .



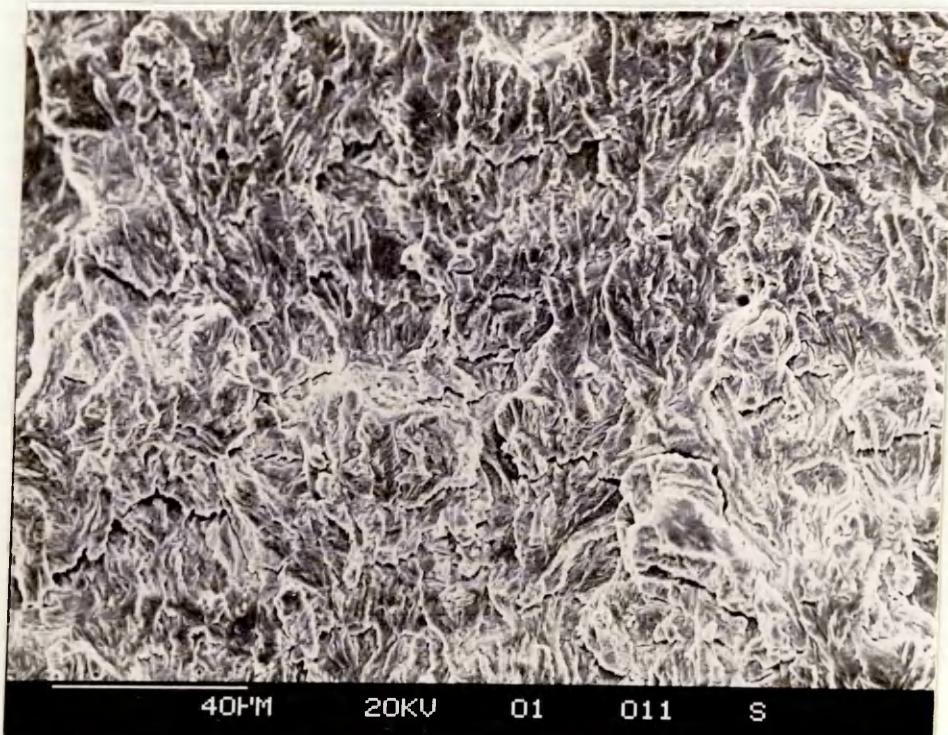
X 1300

Figure 57 Fracture surface for the Superelso 500 HAZ fatigue tested under free corrosion conditions, low ΔK .



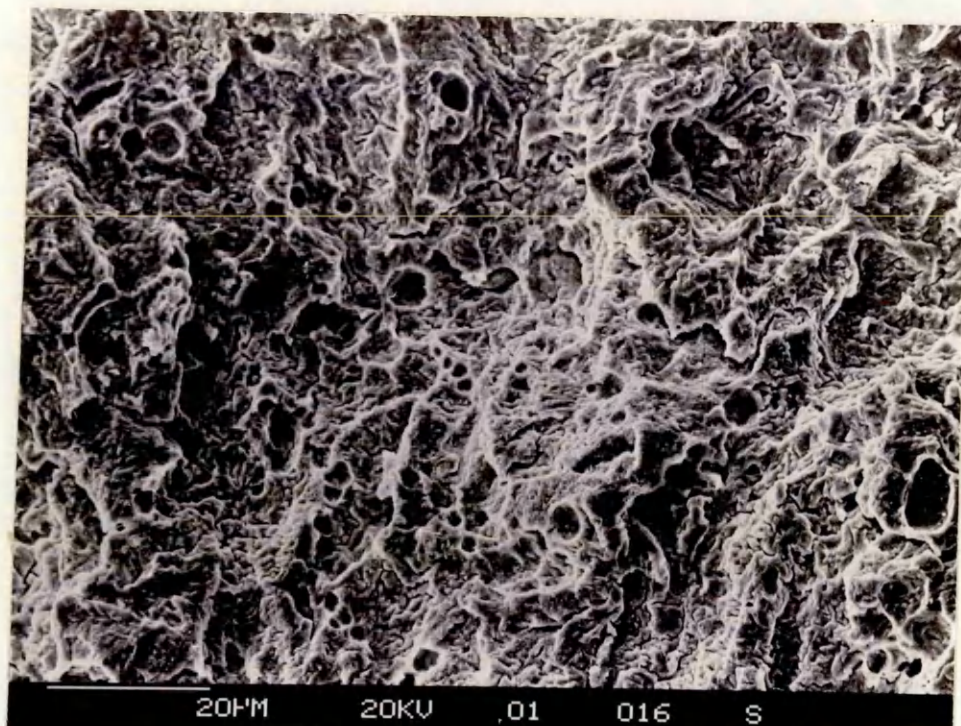
X 625

Figure 58 *Fracture surface for the Superelso 500 HAZ fatigue tested under free corrosion conditions, high ΔK .*



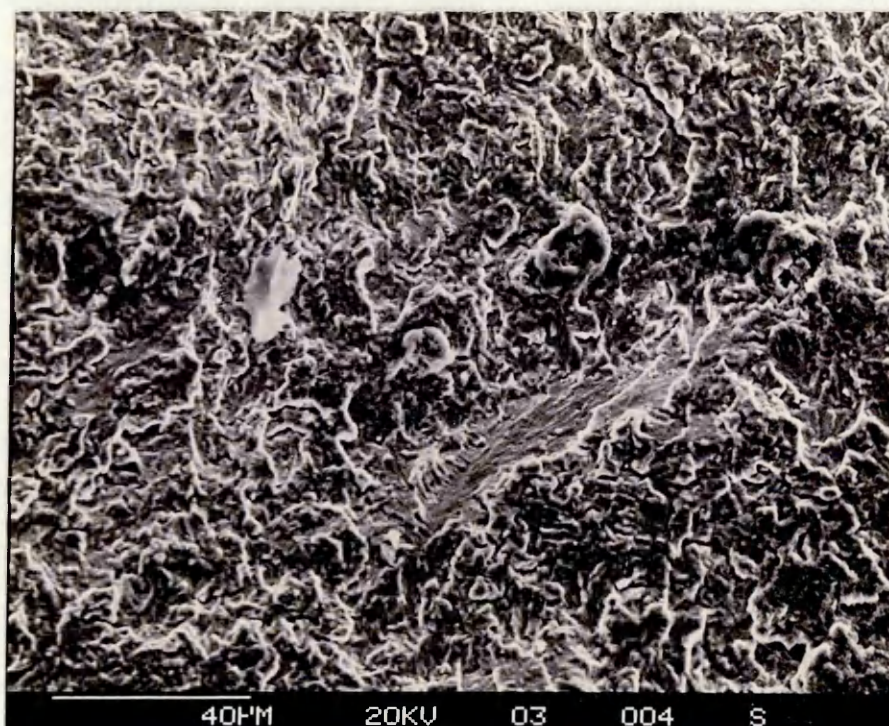
X 625

Figure 59 *Fracture surface for the Superelso 500 HAZ fatigue tested with an applied potential, $P_v = -800$ mV, low ΔK .*



X 1300

Figure 60 *Fracture surface for the Superelso 500 HAZ fatigue tested with an applied potential, $P_v = -800$ mV, high ΔK .*



X 625

Figure 61 *Fracture surface for the Superelso 500 HAZ fatigue tested with an applied potential, $P_v = -1100$ mV, low ΔK .*

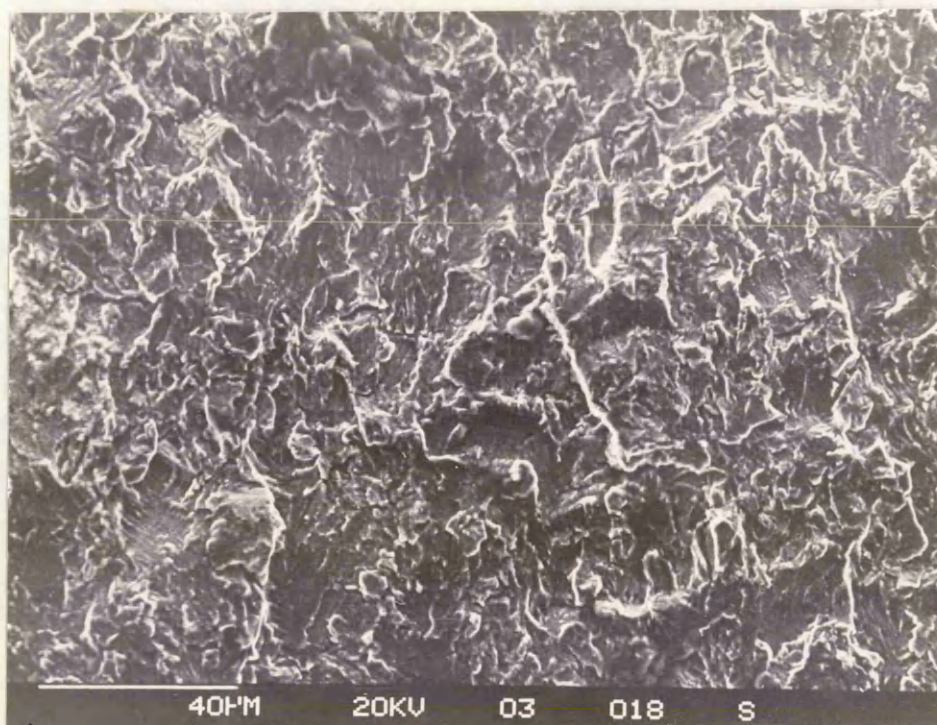


Figure 62 *Fracture surface for the Superelso 500 HAZ fatigue tested with an applied potential, $P_v = -1100$ mV, high ΔK .*

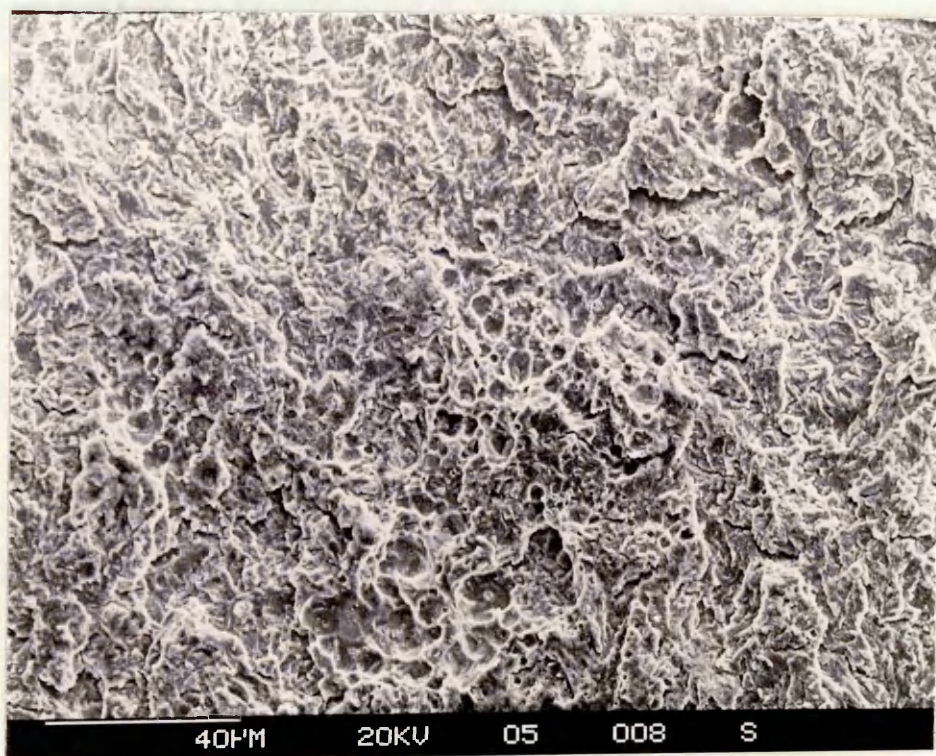
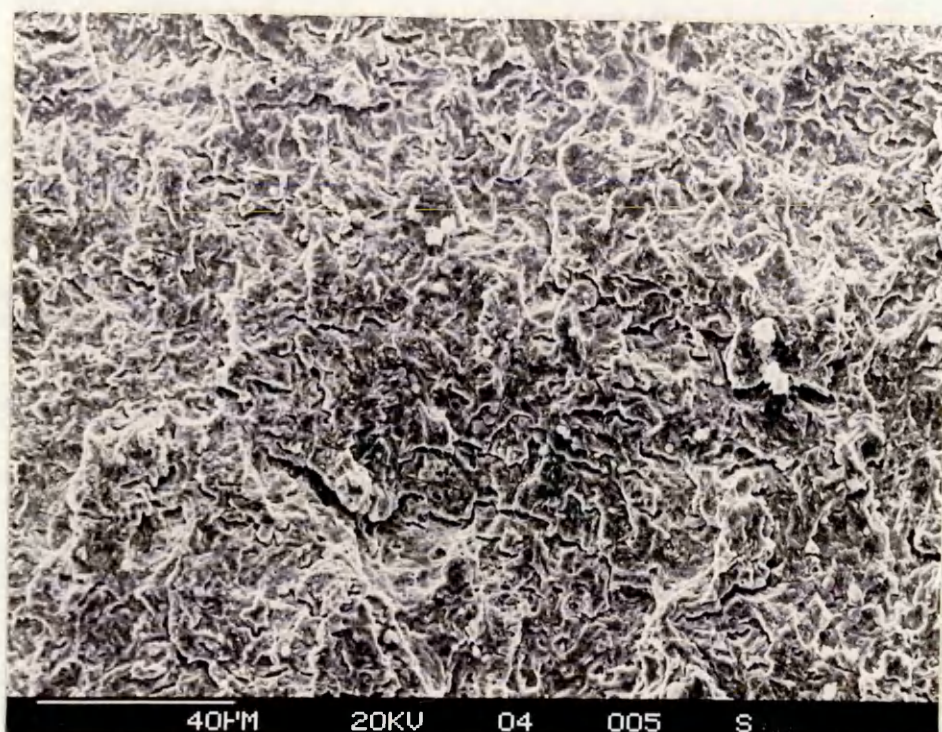
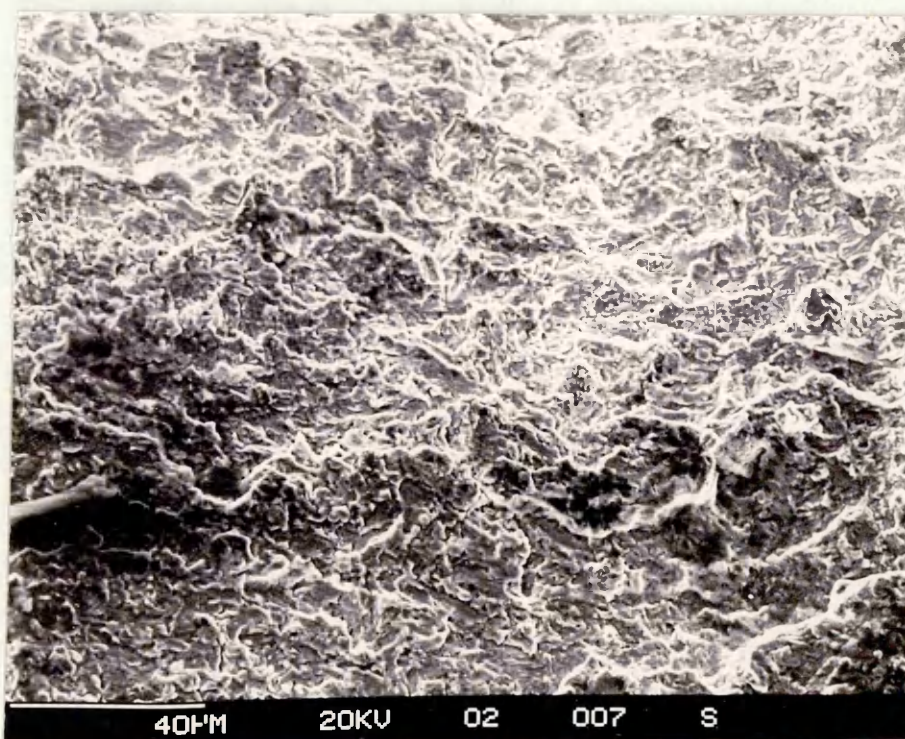


Figure 63 *Fracture surface for the Mannesman X70 HAZ under free corrosion conditions, low ΔK .*



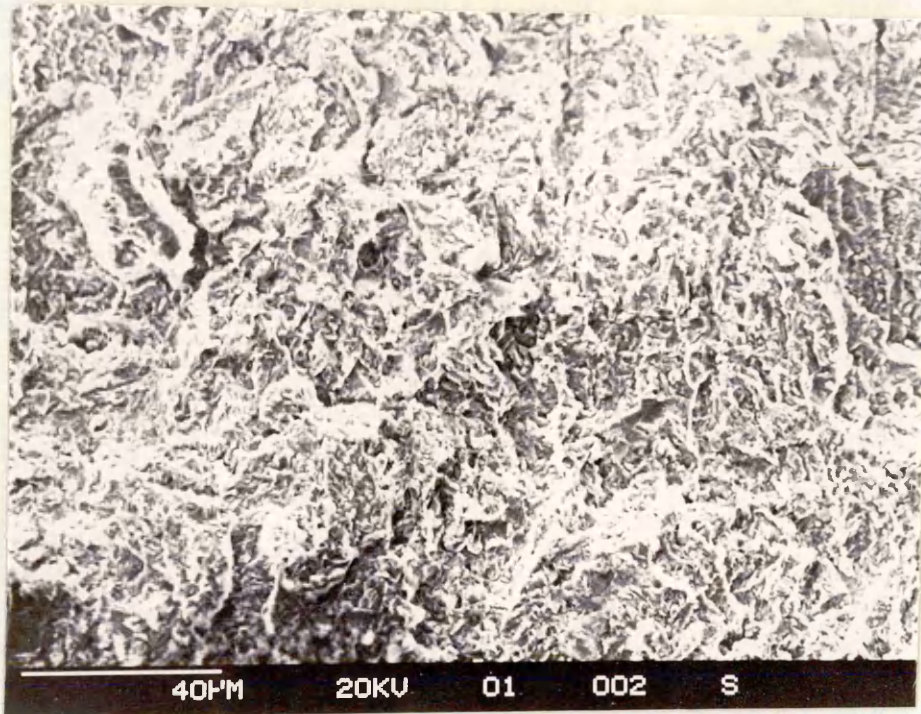
X 625

Figure 64 *Fracture surface for the Mannesman X70 HAZ under free corrosion conditions, high ΔK .*



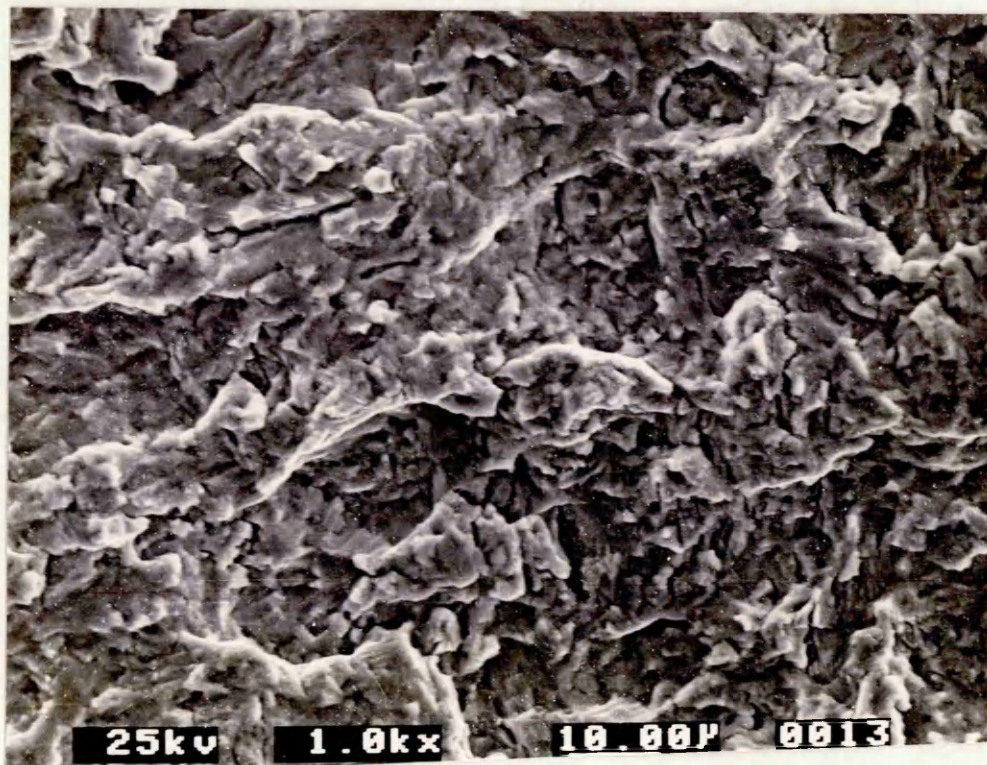
X 625

Figure 65 *Fracture surface for the Mannesman X70 HAZ with an applied potential, $P_v = -800$ mV, low ΔK .*



X 625

Figure 66 *Fracture surface for the Mannesman X70 HAZ with an applied potential, $P_v = -800$ mV, high ΔK .*



X 1200

Figure 67 *Fracture surface for the Mannesman X70 HAZ with an applied potential, $P_v = -1100$ mV, low ΔK .*

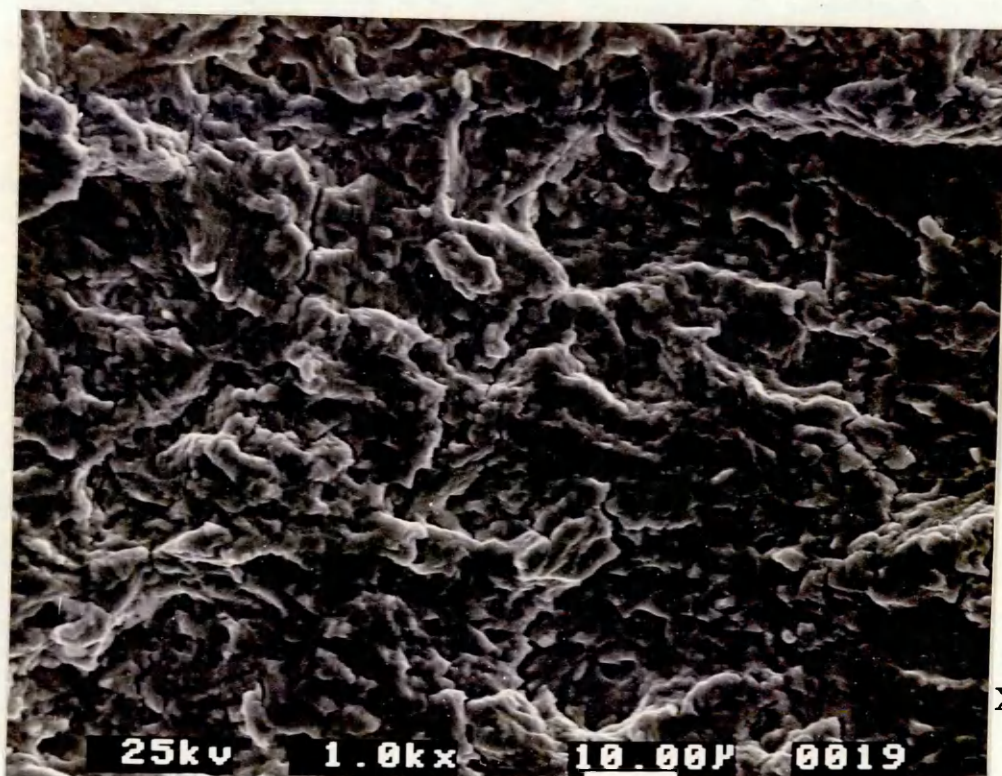


Figure 68 *Fracture surface for the Mannesman X70 HAZ with an applied potential, $P_v = -1100$ mV, high ΔK .*

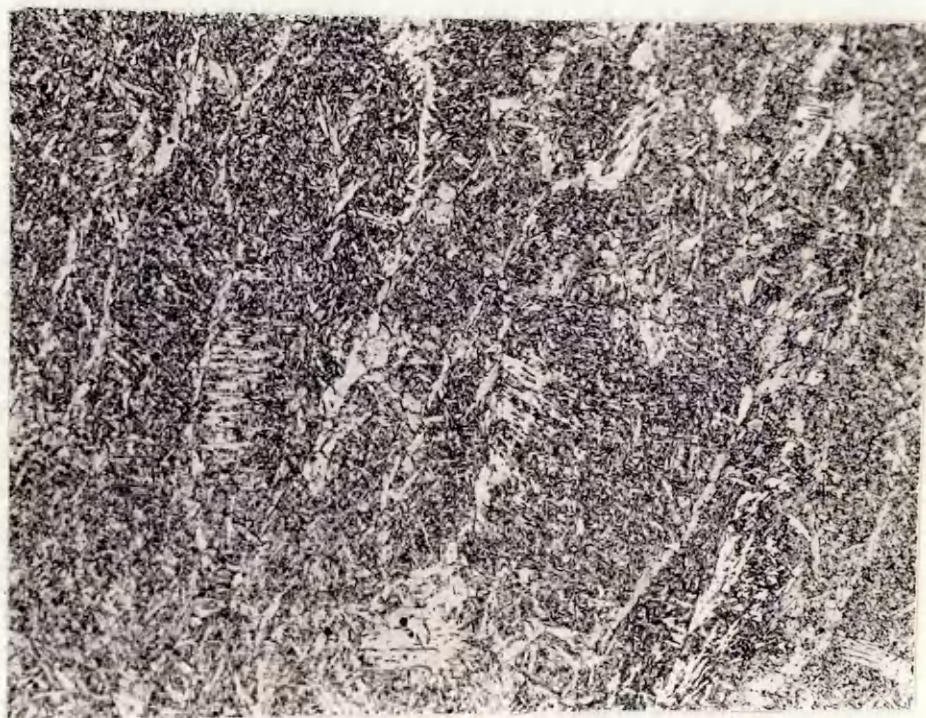
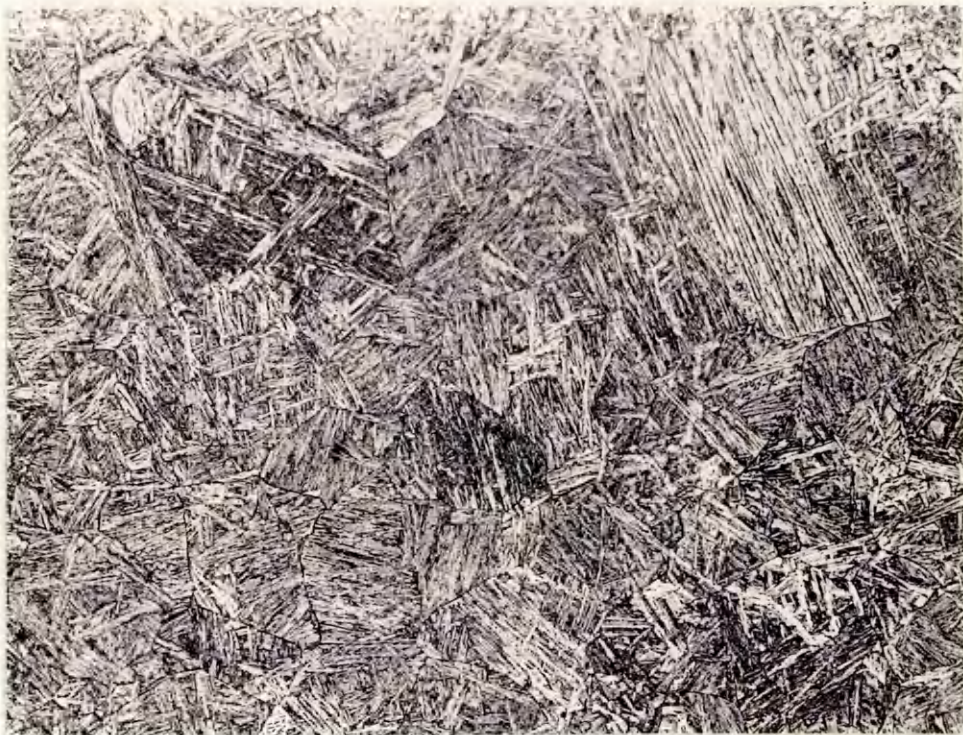
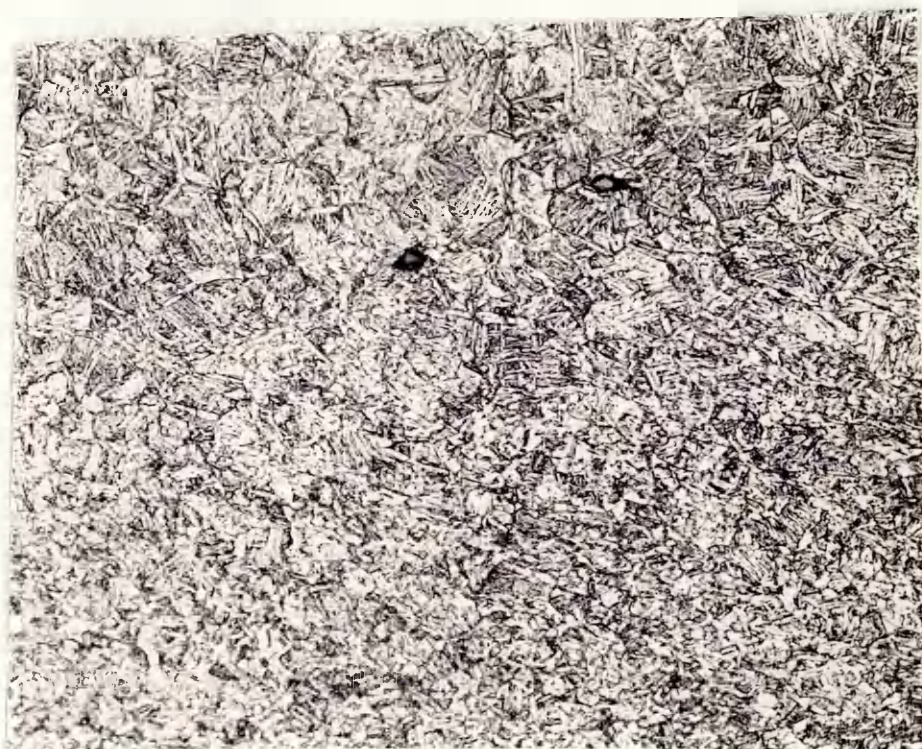


Figure 69 *As deposited weld metal.*



X 400

Figure 70 *Coarse grained HAZ.*



X 400

Figure 71 *Grain refined HAZ.*



X 400

Figure 72 HAZ/parent plate interface.

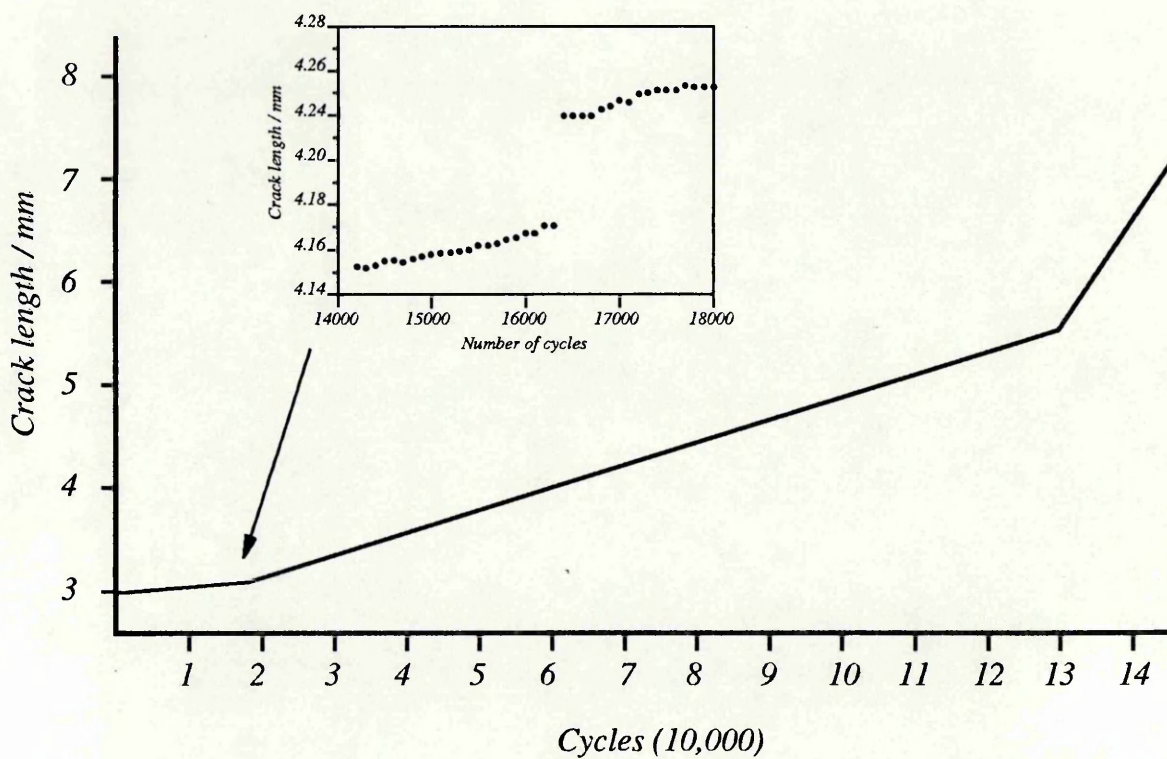


Figure 73 Crack length versus number of cycles for the 1.5 kJ/mm heat input.

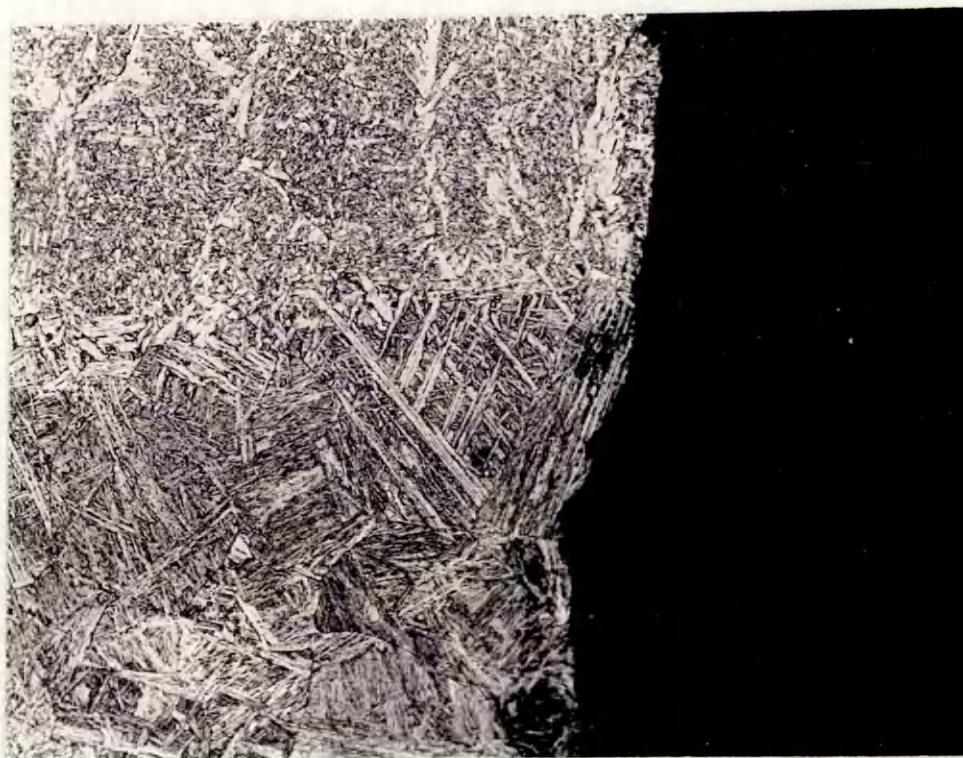


Figure 74 *The microstructure of the region in which the crack was observed to accelerate.*

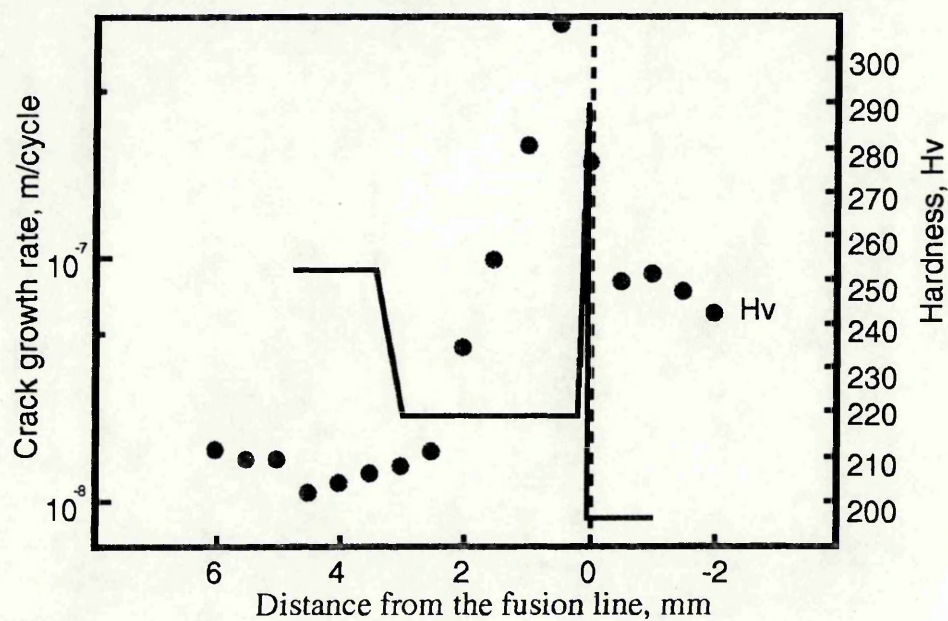


Figure 75 *The crack growth rate as a function of hardness and distance from the fusion boundary.*

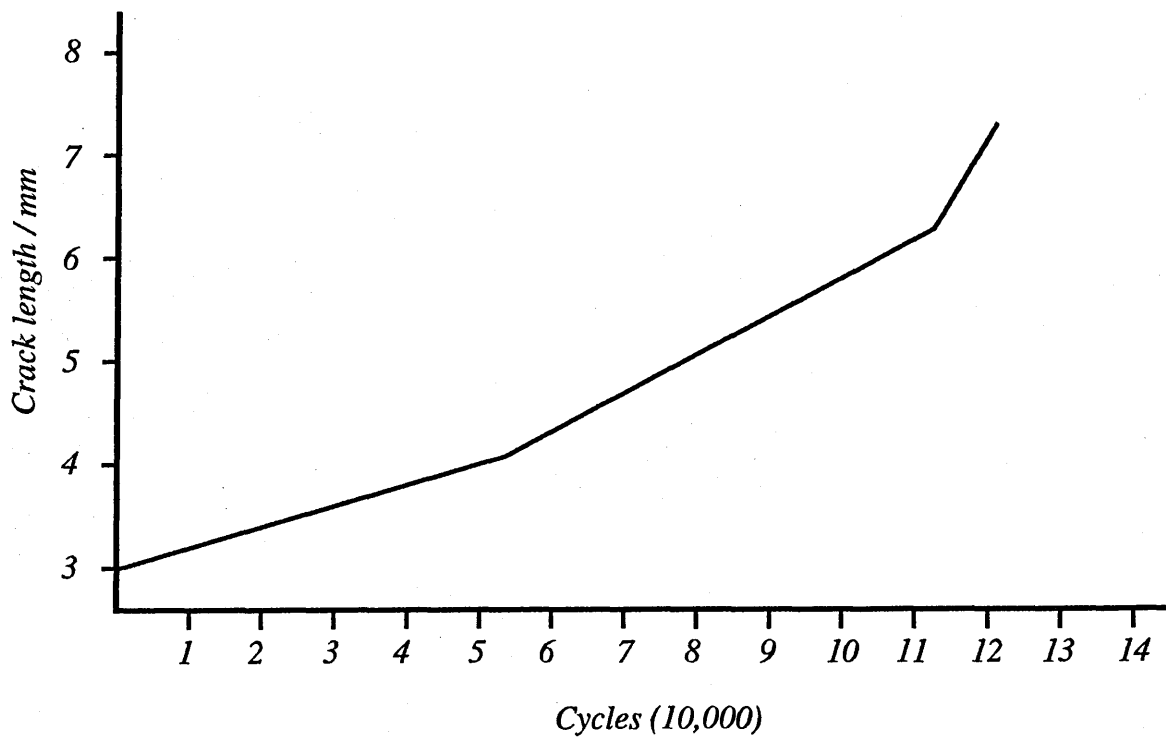


Figure 76 Crack length versus number of cycles for the 3.1 kJ/mm heat input.

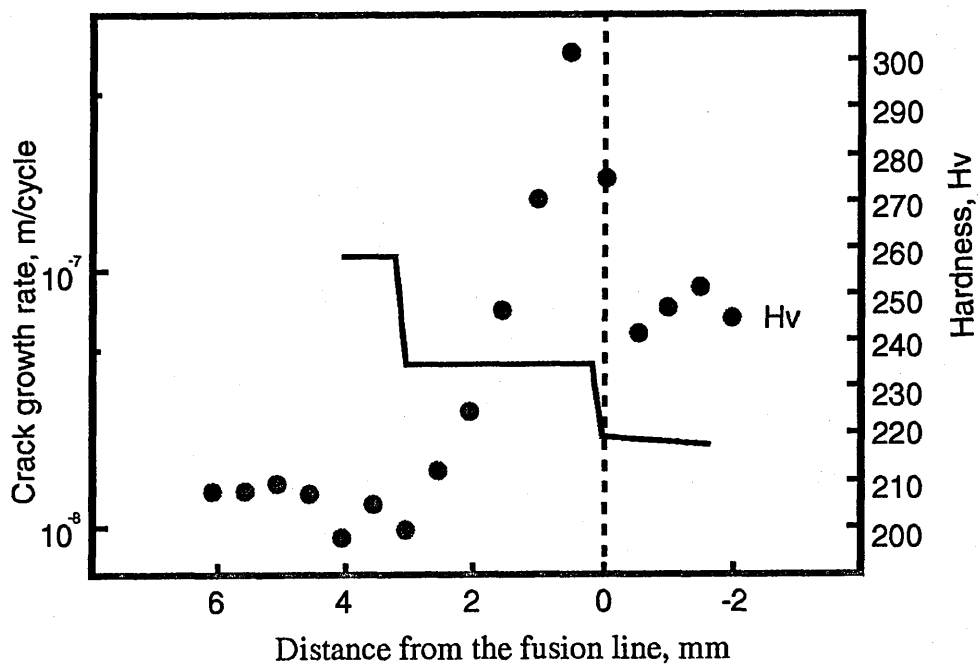


Figure 77 The crack growth rate as a function of hardness and distance from the fusion boundary.

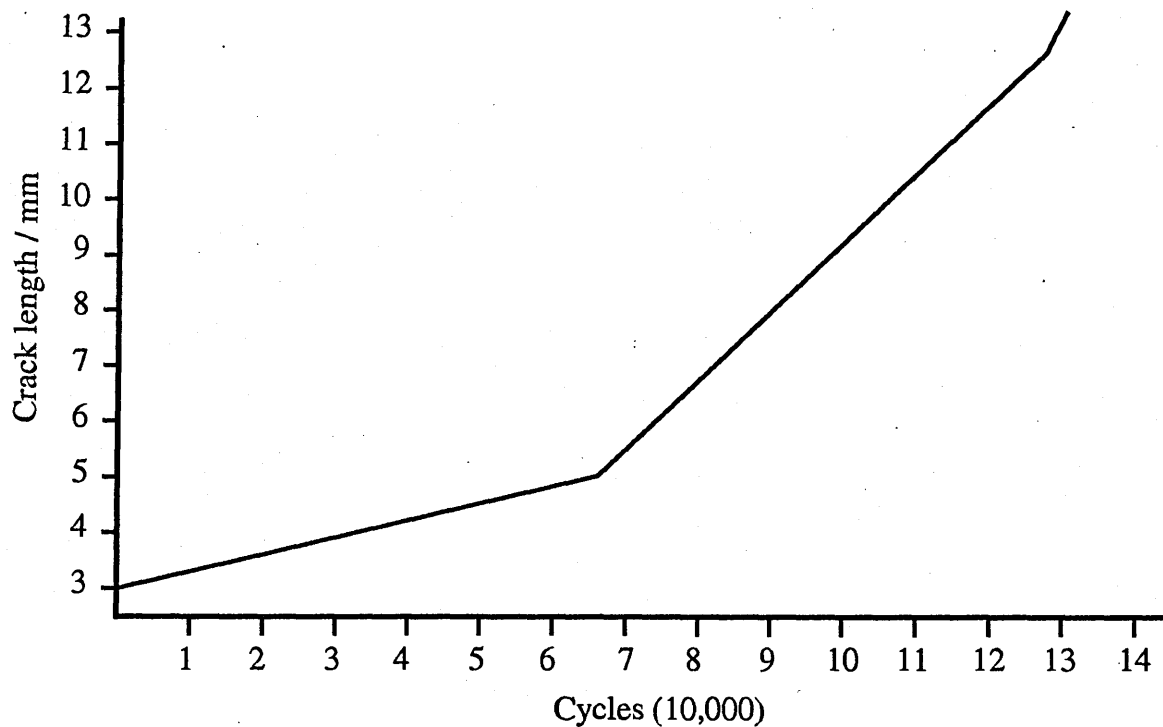


Figure 78 Crack length versus number of cycles for the 6.1 kJ/mm heat input.

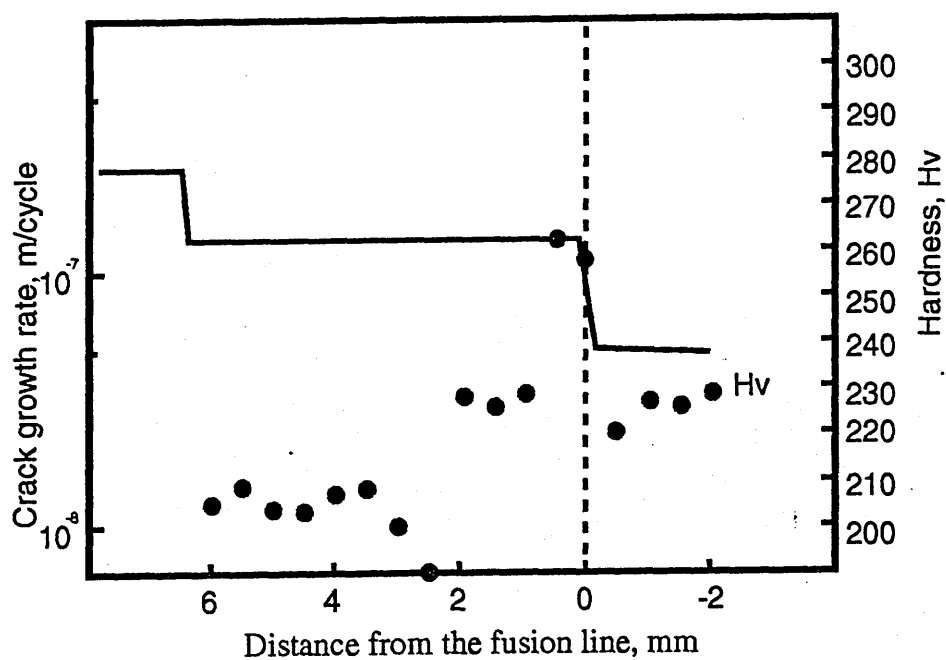


Figure 79 The crack growth rate as a function of hardness and distance from the fusion boundary.

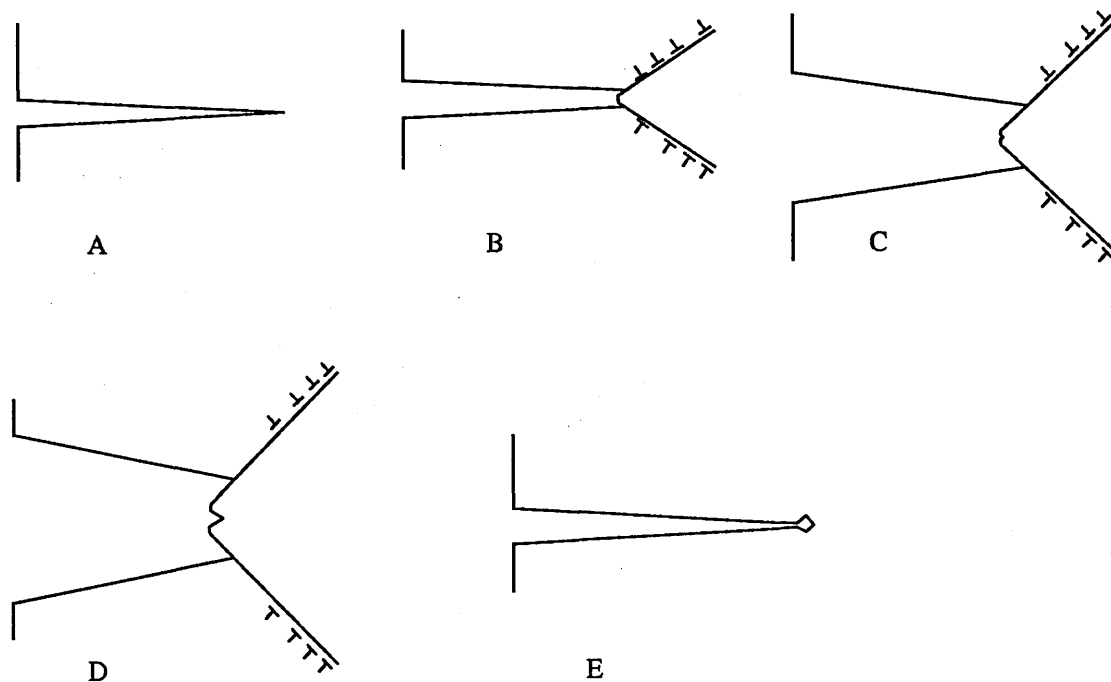


Figure 80 Schematic representation of the slip occurring to cause crack advance during a fatigue cycle.

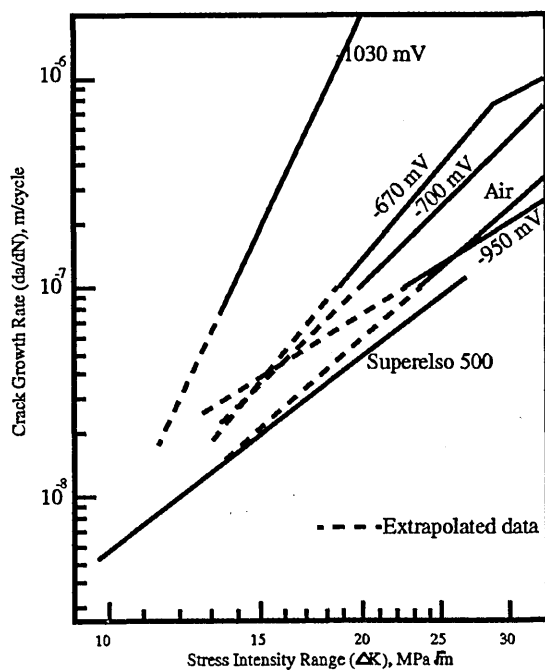


Figure 81 The effect of cathodic potential on the crack propagation rates of HY130

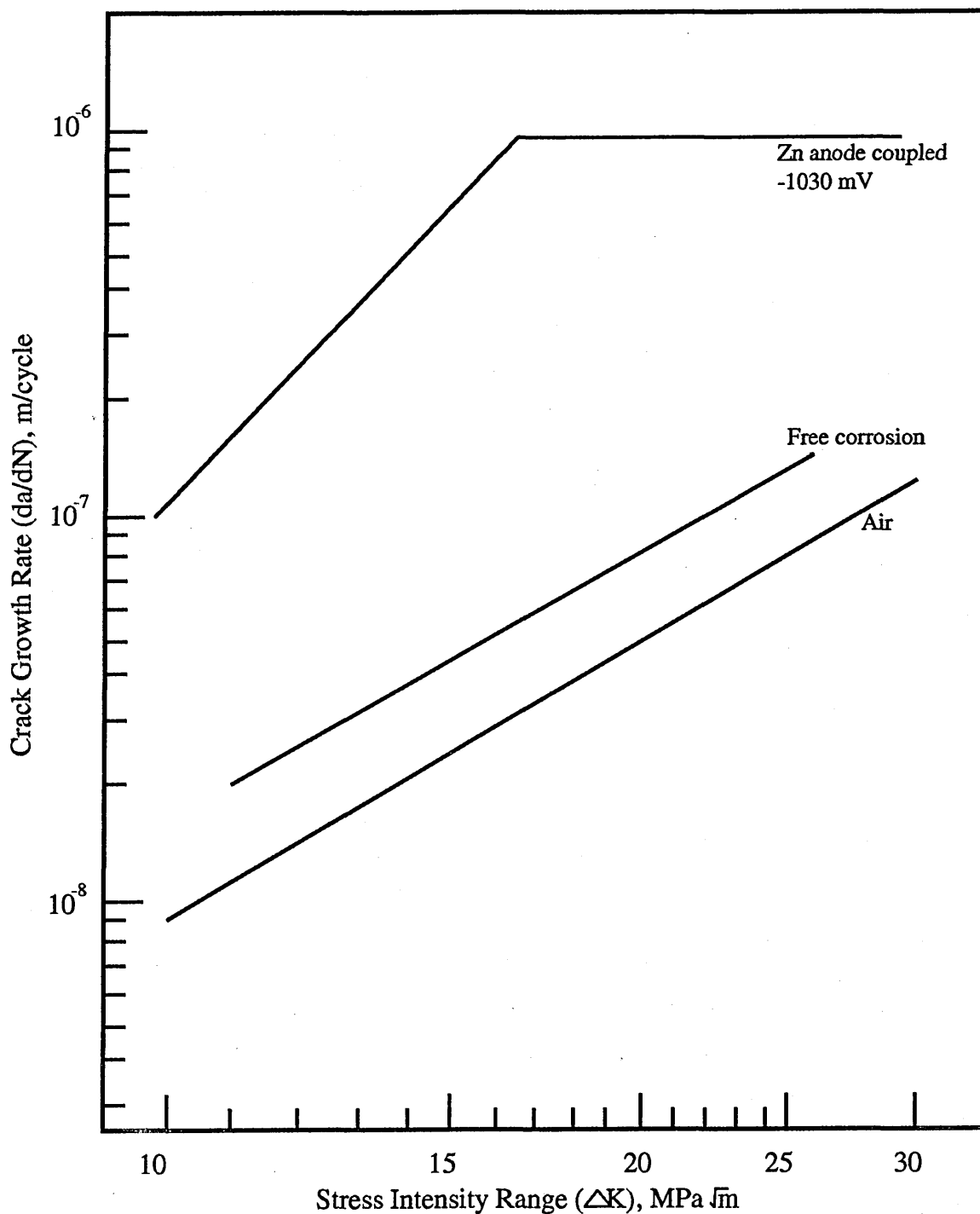


Figure 82 The effect of cathodic protection using a zinc sacrificial anode to provide the protection on a HY100 steel.

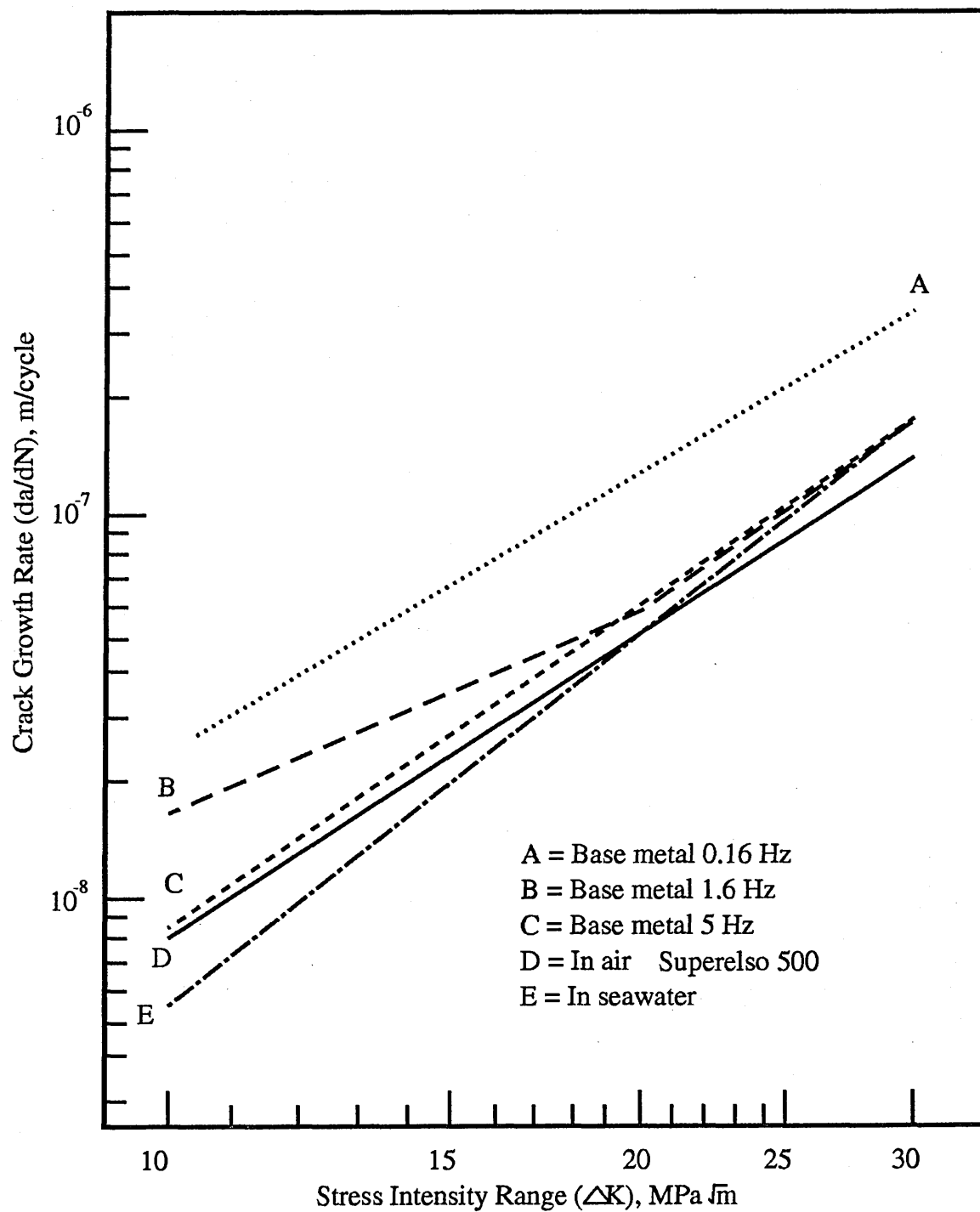


Figure 83 Crack growth rates for HT80 in-air and freely corroding.

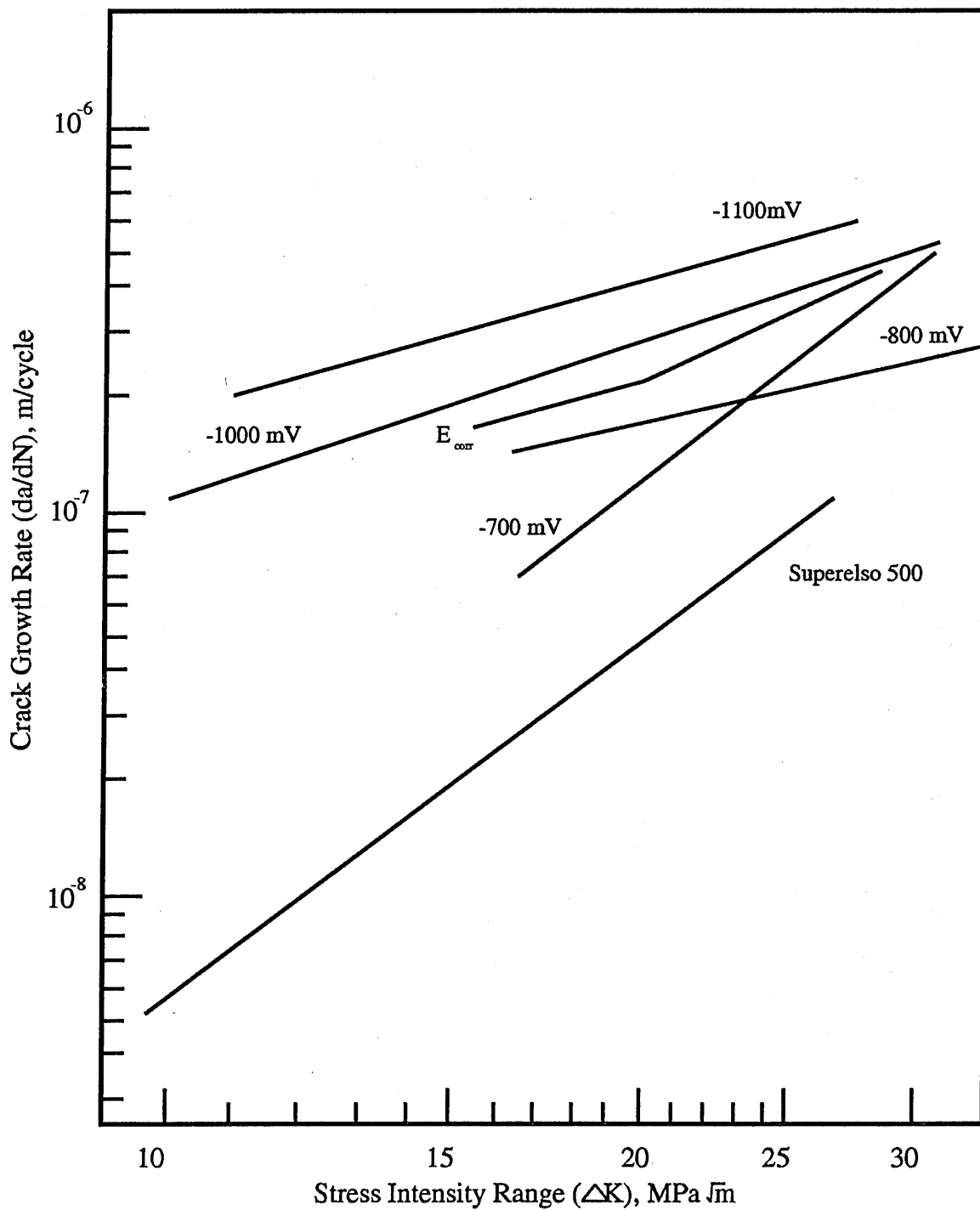


Figure 84 The crack growth rates for HT80 under cathodic protection.

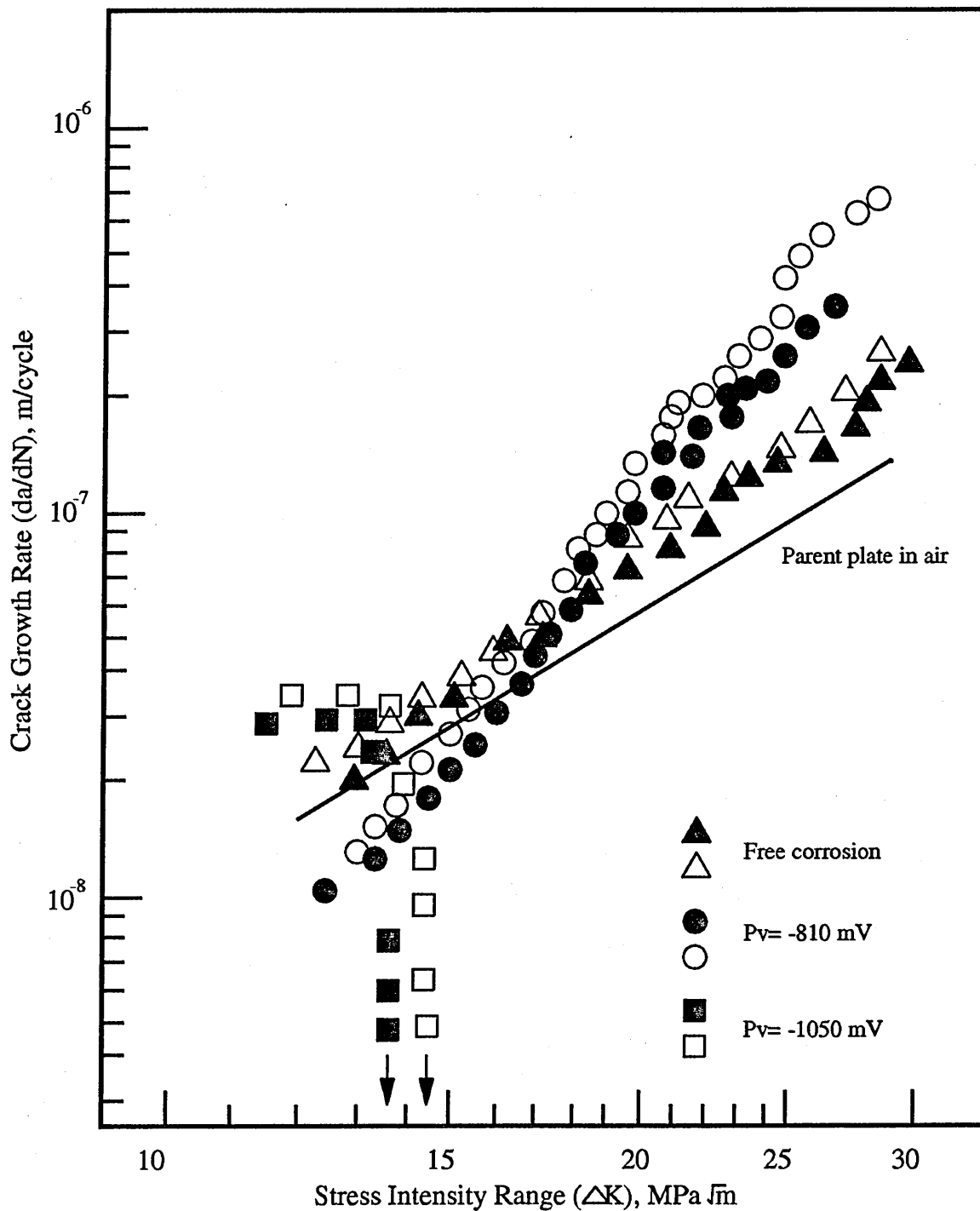


Figure 85 The crack growth rates for the Mannesman X80 parent plate under different corrosion conditions.

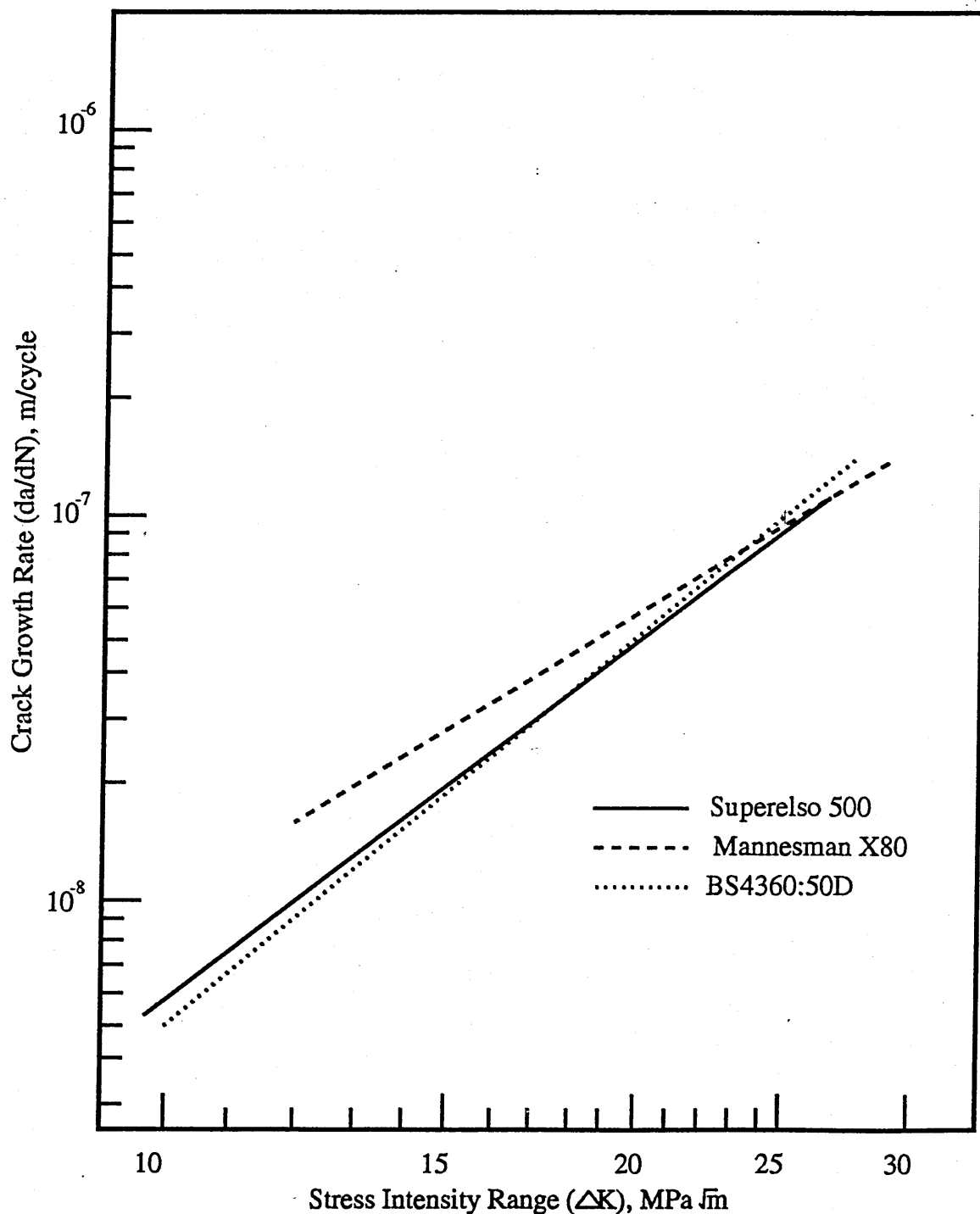


Figure 86 A comparison between X70, Superelso 500 and BS4360:50D fatigue tested in-air.

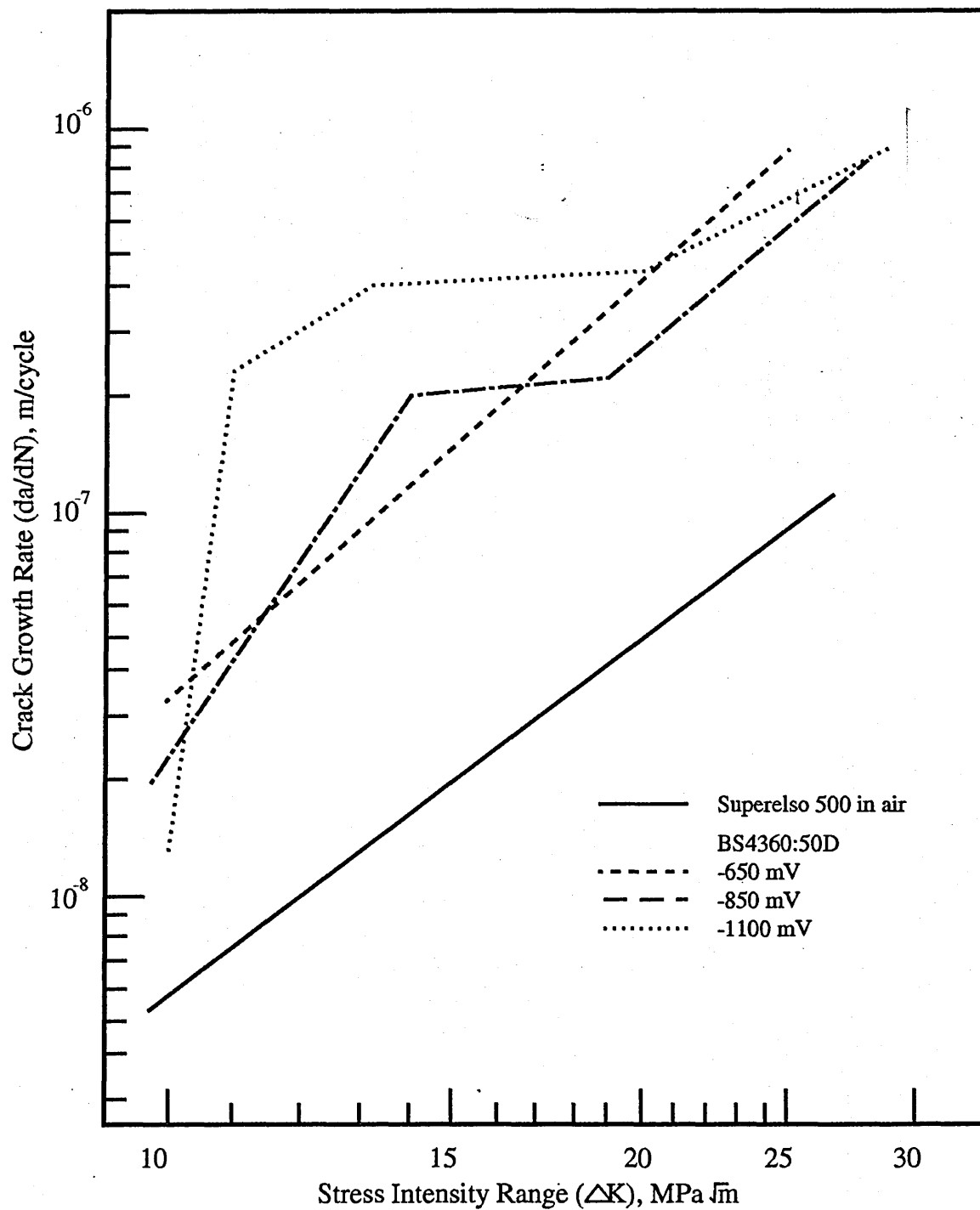


Figure 87 The effects on an environment on the fatigue crack growth rates of high strength steels compared to BS4360:50D.

Base, Active or Anodic end	Nickel (Active)
Magnesium	Brasses
Zinc	Copper
Aluminium (Commercial Purity)	Silver Solder
Cadmium	Nickel (Passive)
Duralumin	18/8 Stainless Steel (Passive)
Steel	Silver
Cast Iron	Titanium
High Nickel Cast Iron	Graphite
18/8 Stainless Steel Active	Gold
Lead Tin Solders	Platinum
Lead	Noble or Cathodic end

Table 1 *Galvanic series in sea water.*

Material	Consumption (kg A y ⁻¹)
Platinum & platinised metals	8 x 10 ⁻⁶
High silicon iron	0.25 - 1.0
Steel	6.8 - 9.1
Iron	9.5
Cast iron	4.5 - 6.8
Lead-platinum	0.09
Lead-silver	0.09
Graphite	0.1 - 1.0

Table 2 *Anode materials and their properties*

Location	Current Density mA/m ²		
	Initial	Mean	Final
Northern North Sea	160	120	100
Southern North Sea	130	100	90
Arabian Gulf	120	90	80
India	120	90	80
Australia	120	90	80
Brazil	120	90	80
Gulf of Mexico	100	80	70
West Africa	120	90	80
Indonesia	100	80	70
Buried Pipelines	50	40	30
Risers (Flowing Water)	180	140	120
Risers (Stagnant Water)	120	90	80
Saline Mud (Ambient Temp.)	25	20	15

Table 3 *Minimum design current densities for protection of fixed offshore structures*

<u>Materials</u>	<u>Relative Permeability</u>
Martensite	1.00
Lower Bainite	1.00
Globular Cementite	3.59
Troostite	3.91
Annealed Steel	4.78
Sorbite	7.00
Normalised Steel	7.43
Coarse Lamellar Pearlite + Ferrite	1.57
Very Fine Pearlite + Ferrite	2.79
Tempered Martensite @ 550°C	1.04
Fine Bainite	1.57
Coarse Bainite	1.95
Coarse Lamellar Pearlite	2.16
Fine Lamellar Pearlite	2.98

Table 4 *The relative permeability of hydrogen in different steel microstructures.*

	C	Mn	Si	S	P	Cr	Mo	Ni	Cu	V	Al
Manufacturer	.09	1.13	.31	.0005	.008	.18	.24	.48	.30	.05	.02
NEI	.08	1.35	.35	.001	.007	.35	.18	.35	.20	.06	.02

Table 5 *Manufacturer's analysis of Superelso 500*

	Y.S.		U.T.S.		A%	kVT - 50°C (J)
t	MPa	KSI	MPa	KSI		
40 mm	570	82	670	96	25	80-115-110
50 mm	560	80	655	94	24	70-110-90

Table 6 *Impact properties of Superelso 500*

	YS (MPa)	UTS (MPa)	Fracture	Distance from fusion line	KCv - 40°C	COD -20°C	Hv 10 max
SAW as welded	539	640	base material	+1 +3	200-177-193 105-110-117	0.39-0.47	280
SAW - 2 hrs at 580°C	540	635		+1	190-100-95		240

Table 7 *Impact and toughness properties of welded Superelso 500.*

	C	Mn	Si	S	P	Cr	Mo	Ni	Cu	V	Al	Nb	Ti
Manufacturer	.08	1.7	.43	.005	.017	.05	.04	.04	.06	.002	.05	.035	.01
NEI	.093	1.78	.418	.003	.014	.05	.04	.05	.07		.049	.035	.012

Table 8 *Chemical analysis of Mannesman X70*

	R _p 0.2	R _p 1.0	R _p 2.0	R _p 3.0	R _p 4.0	R _m	R _p /R _m	El %	A %
Longitudinal (Flat Bar)	493	491	506	533	553	616	80	27	
Transverse (Flat Bar)	514	518	532	560	581	644	80	25	
Through Thickness 1	474	502	541	576	596	629	75	49	74
(Round Bar) 2	482	504	553	586	606	632	76	51	75
3	479	499	550	576	591	623	77	47	76

Table 9 *Mechanical properties of Mannesman X70*

Test Temp °C	+100		+20		0		-20		-40		-60		-80		-100		
Energy/Area (shear)	J	%	J	%	J	%	J	%	J	%	J	%	J	%	J	%	
Longitudinal	1	232	100	214	100	250	100	232	100	231	100	200	100	196	100	53	5
	2	232	100	226	100	270	100	252	100	234	100	190	100	112	30	90	60
	3	242	100	240	100	264	100	262	100	232	100	222	100	206	100	114	70
Average	235		227		261		249		232		204		171		86		
Transverse	1	250	100	250	100	250	100	254	100	242	100	222	80	174	80	22	10
	2	229	100	240	100	234	100	232	100	210	100	180	70	174	70	6	5
	3	244	100	246	100	231	100	234	100	220	100	206	80	206	95	8	5
Average	241		245		242		240		224		203		185		12		
Thickness	1					112	25	39	2								
	2					90	20	42	2								
	3					205	75	47	2								
Average					136		43										

Table 10 *Impact properties of Mannesman X70*

Sodium Chloride	24.9 g/dm ³
Calcium Sulphide	1.50 g/dm ³
Magnesium Chloride	2.99 g/dm ³
Magnesium Sulphate	2.00 g/dm ³

Any other elements are only trace elements and are incidental.

Table 11 *Chemical composition of synthetic seawater.*

	Crack Growth Rates m/cycle		
	Weld Metal	HAZ	Parent Plate
1.5 kJ/mm	9×10^{-9}	2.2×10^{-8}	8.6×10^{-8}
3.1 kJ/mm	2.1×10^{-8}	4.2×10^{-8}	1.0×10^{-7}
6.1 kJ/mm	3.0×10^{-8}	1.3×10^{-7}	2.3×10^{-7}

Table 12 *Average crack growth rates for different weld regions of different heat input welds on Superelso 500.*

