THRESHOLD ΔK VALUES AND NON-CLOSURE OF FATIGUE CRACKS

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ABSTRACT

The results from investigations of the fatigue crack growth behaviour and ΔK threshold levels in α-titanium, α-β titanium, α-brass, nickel base alloys and low carbon ferritic steels are presented. The observations show a substantial influence of microstructure on the mechanisms and modes of fatigue crack growth. The role played by transgranular and intergranular modes of failure in the intermediate and low ΔK regions are considered. The development of transgranular fracture and the out of plane crack trajectories with the associated "non-closure" of the fatigue cracks is suggested as a substantial contributor to the factors determining ΔK threshold levels.

KEYWORDS

Fatigue crack growth; ΔK threshold; non-closure; microstructure; environment.

INTRODUCTION

The modes of fatigue crack growth at low stress intensities and in the threshold region have been studied and reported upon previously (Newman, Fuhrot and Ve Hoft, 1979; Gerberich and Moody, 1979; Kirby and Beevers, 1979; Walker and Beevers, 1979; Beevers, 1977). The aim of this paper is to examine the factors which influence the mode of fatigue crack growth and to consider the resultant effect upon ΔK threshold levels. The results presented were obtained from tests on α-titanium, α-β titanium, α-brass, nickel base alloys and low carbon ferritic steels. The modes of crack growth are identified and in some cases mechanisms for crack extension are proposed. The effects of the crack face profile on the closure or more specifically on "non-closure" of fatigue cracks is considered in relation to observed ΔK threshold levels.

RESULTS

Single edge notch tension - tension test process 10 mm x 35 mm x 140 mm were tested at an R ratio (K_{min}/K_{max}) of 0.35. The test pieces had a grain size of ∿ 0.25 mm and were tested in laboratory air at a frequency of ∿ 100 Hz over a ΔK range of

$\sim$ 6 to 20 MN m$^{-3/2}$. Crack lengths were measured by optical and d.c. potential difference methods. The specimens were finally fractured and examined in the scanning electron microscope (SEM). The orientation of the grains at the specimen surfaces were obtained by the use of selected area channelling patterns. The orientations of the grains with respect to the principal stress axis and the major growth direction were obtained using an optical method external to the SEM (Ward-Close, 1977).

From a combined study of the grain orientations and fatigue fracture surface morphologies the following observation can be drawn; where the orientation of the grain with respect to the principal stress axis and crack growth direction was that illustrated in Fig. 1(a) striations were observed on the fracture surface (Fig. 2). The grain orientation indicated in Fig. 1(a), would permit slip activity on two sets of $\{10\bar{1}0\}$ type plans. In grains where the slip planes were not symmetrical about the crack growth plane "fissure" striations were observed. These observations favour an alternating slip mechanism for the formation of striations.

A second distinctive mode of crack growth was that of "furrow" formation as illustrated in Fig. 3 for grains with orientations similar to Fig. 1(b). The line of the furrows, two sets in adjacent grains in Fig. 3, were always the c direction. This mode of growth is envisaged as resulting from asymmetrical slip on two $\{10\bar{1}0\}$ type planes forming a series of tongue like cracks which subsequently join together by a ductile tearing process.

The third mode of fatigue crack growth illustrated in Fig. 4 occurred in grains with orientations similar to those in Fig. 1(c). The facets were observed to form on planes $\pm$ 1° to the (0001) plane. The facets developed as a consequence of repeated load cycling and formed over the whole ΔK range investigated.

The results presented in this section illustrated the dominant role that grain orientation can have on the mechanism of fatigue crack growth. This results from the anisotropic nature of plastic flow in α-titanium, which is reflected in the growth modes when the crack tip plasticity is contained within individual grains.

The out of place displacements which resulted from the variable crack growth modes (Figs. 2-4) led to point contrast between two crack faces. The crack faces opening was nominally mode I but clearly some mode II or mode III displacement would be required to produce the offsets of the fracture planes to result in the point contacts.

Fatigue crack growth studies were carried out on single edge notch three point bend specimens of α-brass 10 mm by 35 mm by 140 mm tested in laboratory air at R = 0.35 at a frequency of 90 Hz. The log ΔK versus log da/dN curves for specimens with grain sizes of 51µm to 714µm are presented in Fig. 5. The results show that the threshold ΔK increased with increasing grain size. The yield strength of the brass increased (75-120 MNm^{-2}) with the decreasing grain size. The macroscopic fracture surfaces are presented in Fig. 6, the coarser the grain size the rougher the fracture surface. SEM studies revealed a combination of intergranular and transgranular fracture modes in the intermediate and low ΔK region.

The major differences between the specimens were the threshold ΔK level and the macroscopic fracture surface appearance. The higher ΔK threshold at the larger grain size is attributed to the wedging open of the crack by the asperitues created on the fracture surface during crack growth.

Tests on compact tension specimens (62.5 mm x 50 mm x 25.4 mm) of a commercial mild

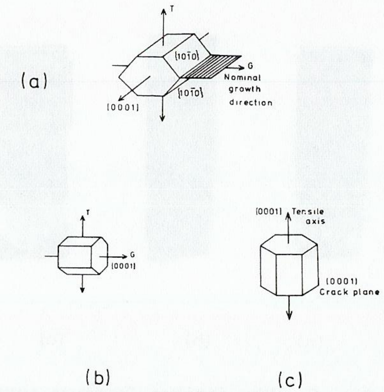

Fig. 1. Grain orientations at the crack tip with respect to the principal stress axis T and general crack growth direction G.

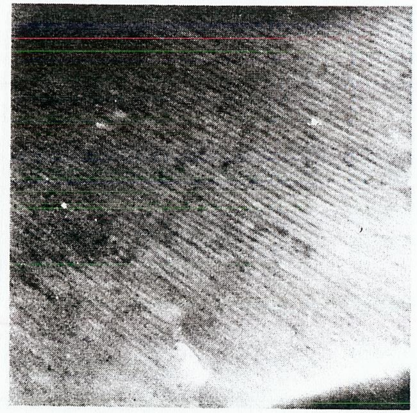

Fig. 2. Striations formed during a test in laboratory air $\Delta K \sim$ 16 MN m$^{-3/2}$. X 6,600

steel[1] and a vacuum melted laboratory low carbon steel[2] were carried out in laboratory air and in a vacuum of < 10^{-5} torr. The specimens were tested at R ratios in the range 0.1 to 0.7 and over the growth rate range 10^{-4} to 10^{-7} mm/cycle.

The commercial mild steel BS970 contained small amounts of pearlite and the vacuum melted laboratory steel had carbide films on some of the ferrite grain boundaries.

[1]Carbon content $\sim$ 0.09 wt%

[2]Carbon content 0.01 - 0.02 wt%

Fig. 3. Furrows formed on the fatigue fracture surface of a specimen tested in air at $\Delta K \sim$16 MN m$^{-3/2}$. X 6,400

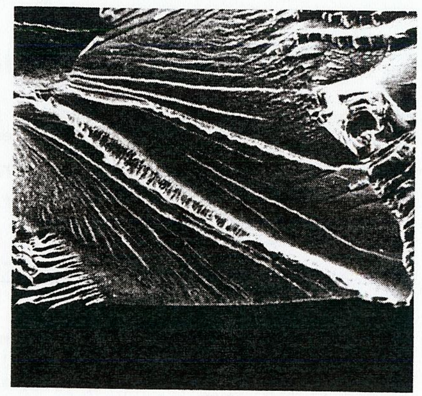

Fig. 4. Fatigue facet formed in air at $\Delta K \sim$8 MN m$^{-3/2}$. X 240

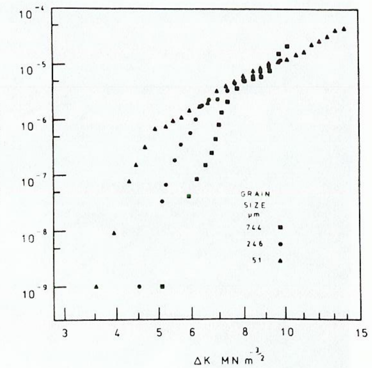

Fig. 5. Fatigue crack growth curves for α-brass tested in laboratory air, at R = 0.35.

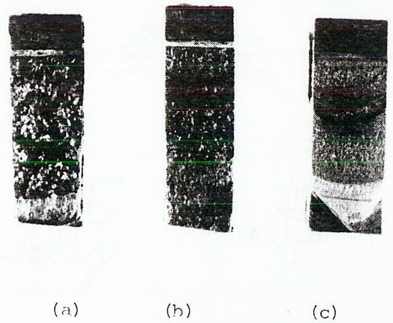

Fig. 6. The fatigue fracture surfaces of α-brass specimens with grain sizes a) 714 μm, b) 246 μm and c) 51 μm.

The mode of fatigue crack growth was a combination of trans and inter granular separation. The intergranular separation was along the ferrite grain boundaries and no significant observations of separation at the ferrite grains/carbon film interfaces in the laboratory steel were observed.

The extent of the intergranular mode of separation showed no unique dependence upon either ΔK or K_{max}. However, the occurrence of intergranular separation was dependent upon the stress state at the crack tip. Figure 7 shows the variation in % area of integranular fracture as a function of specimen thickness. For σ_y = 220 MNm^{-2} and K_{max} = 23 $MNm^{-3/4}$ the plane stress plastic zone Rp approximate to 1.6 mm for

$$R_p = \frac{1}{2\pi}\left(\frac{K_{max}}{\sigma_y}\right)^2$$

The maximum extent of intergranular fracture is achieved at a depth of ∿ 5 mm.

The results in Fig. 10 are a plot of percentage intergranular fracture as a function of K_{max} normalised by the stress intensity required generally to yield the ligament. The value of K_L was obtained by incorporating the appropriate P_L value (Bucci, Paris and Landes, 1972) in the standard K formation where

$$P_L = \frac{1.44\ \sigma_y\ (W - a)^2\ B}{4W}$$

The data in Fig. 8 was obtained from the vacuum melted low carbon steel with a range of grain sizes (20 μm - 110 μm). The % intergranular showed no systematic variation with K_{max} or ΔK. However Fig. 8 illustrates that for a range of grain sizes and yield stresses the occurrence of intergranular fracture is severely reduced by the onset of general yield. The results in Figs. 7 and 8 support the view that the occurrence of intergranular fatigue fracture in these steels is enhanced by plane strain conditions. The influence of grain size on the fatigue

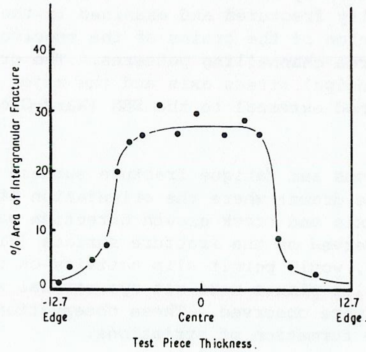

Fig. 7. Variation of intergranular fracture against test specimen thickness for a low carbon ferritic steel ΔK ∿ 16.3 $MNm^{-3/2}$.

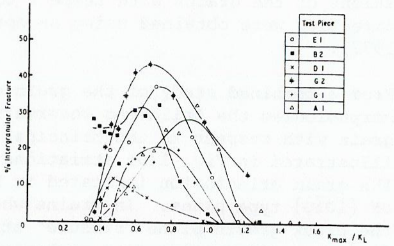

Fig. 8. Variation of percentage area of intergranular fracture with K_{max}/K_L for a low carbon ferritic steel tested at R = 0.35 in air.

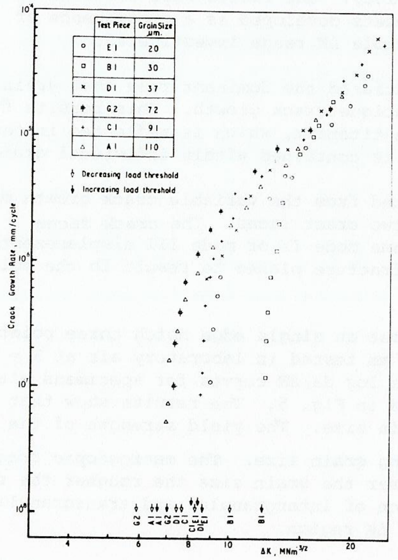

Fig. 9. The log ΔK versus log da/dN curves for a vacuum melted low carbon ferritic steel tested in air at ∿ 100 Hz and R = 0.35.

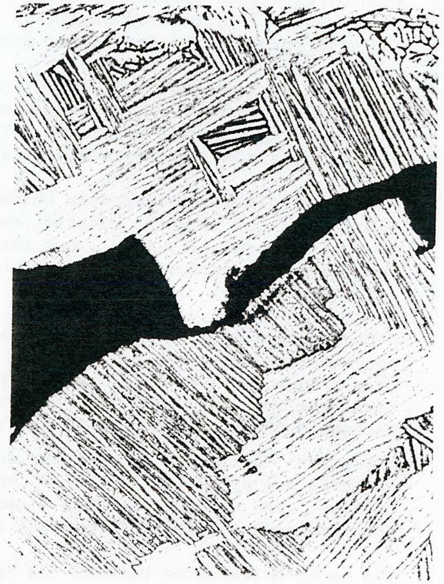

Fig. 10. Section through the fatigue crack faces of an α-β titanium alloy IMI 685 where plastic flow has occurred and bent the α-platelets in the protruding section. X 400

crack growth ratios and threshold ΔK levels for a vacuum melted low carbon steel are presented in Fig. 9.

DISCUSSION

The magnitude of ΔK threshold and associated fatigue crack growth rates are influenced by factors such as elastic modulus, yield strength, microstructural parameters, environment, load ratio, stress history and fatigue crack closure. In this paper the discussion will be limited to an assessment of the role of "non-closure" of fatigue cracks on crack growth rates and ΔK threshold levels.

In the first part of the discussion fatigue crack growth where the transgranular mode of separation dominates will be considered. A Ti-6Al-4V alloy (Halliday and Beevers, 1979) in a mill annealed and β heat treated condition was tested over the ΔK range 15 to 40 $MNm^{-3/2}$. Load versus crack mouth opening curves showed that in the β heat treated alloy with the coarser microstructure substantial load transfer occurred across the crack faces above minimum load. The β-heat treated alloy exhibited greater resistance to crack growth than the mill annealed alloy and thus was attributed to "non-closure" of the faces of the fatigue crack.

The wedging open of fatigue crack faces has been observed in other metals and alloys Fig. 10 illustrates this feature in an α/β titanium alloy IMI 685. The α-platelets are bent in the protruding section indicating high stresses in this region. Nickel base alloys Ni 901 and Astralloy exhibit similar features, point contact between the faces of a fatigue crack in an Astralloy specimen is presented in Fig. 11.

The wedging open of the crack faces can be detected from compliance measurements. In, sinusoidally loaded aluminium alloys, α-titanium and nickel base alloys the dynamic back face strain and COD time signals are truncated in the compressive half of the load cycle. An example of this behaviour is presented in Fig. 12 from tests on an aluminium alloy.

A recent paper (Walker and Beevers, 1979) on fatigue crack growth in α-titanium which reported the results of both optical and compliance measurements came to the following conclusions. Optical and replica studies of surface cracks showed that crack closure was caused by deviations in the crack trajectory associated with a transgranular mode of growth. As the crack faces approached one another over the reducing half of the load cycle small amounts of in-plane shear produced points of contact which tended to wedge the cracks open and prevent the stresses intensity from falling to that associated with the minimum load. Thus, the wedging open of crack reduces the range of crack tip opening and hence the driving force at the crack tip. The out of plane deviation of the crack would be dependent upon crystallographic texture and grain size. For random grain orientation the out of plane trajectories would be expected to increase with increasing grain size. The fatigue crack growth rates are reduced and the threshold ΔK values are increased in α-titanium (Robinson and Beevers, 1973) and α/β titanium (Halliday and Beevers, 1979). These observations are supported by similar effects in mild steel (Masovnave and Bailon, 1976) and stainless steel (Priddle, 1978) where the fatigue crack growth mode is dominantly transgranular and the ΔK threshold is increased in the coarser grained materials.

In the second part of the discussion the influence of the introduction of intergranular failure modes on fatigue crack growth rates and ΔK threshold levels will be considered.

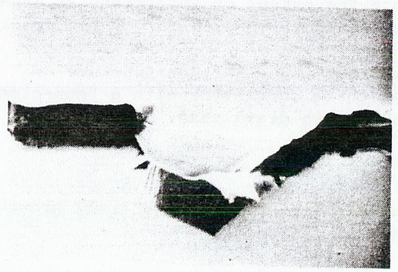

Fig. 11. A surface replica showing point contact of fatigue cracks in Astralloy. X 400

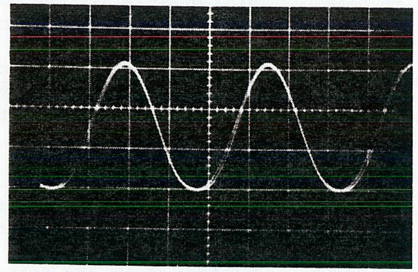

Fig. 12. The superimposed crack mouth opening and inverted back face strain signals from a sinusoidally load CT specimen. The signal is truncated in the compressive half of the cycle.

If an intergranular mode of separation were dominant in the crack extension process and lead to extensive out of plane crack trajectories a similar argument concerning grain size dependence and the influence of "non closure" on ΔK threshold should apply as in the case of transgranular failure modes. This however, is not the case and the influence of intergranular fracture modes on ΔK threshold cannot be readily predicted at the present time. In quenched and tempered 300M steel (Ritchie, 1977; Ritchie, 1977; Ritchie, 1977) the ΔK threshold was insensitive to changes in the prior austenite grain size. On the other hand in high strength as quenched 4% Cr 0.35% C steel (Carlson and Ritchie, 1977) the ΔK threshold increased with increasing prior austenite grain size. For a low carbon, low yield strength vacuum melted steel the results in Fig. 9 show no general trend of ΔK threshold with ferrite grain size.

The high principal stresses sustained at the crack tip by plane strain conditions encourages failure by intergranular modes of separation. Intergranular separation is also encouraged by the presence of water vapour and diminished by an inert environment such as a vacuum of $< 10^{-5}$ torr (Irving and Kurzfeld, 1978) in a quenched and tempered En 24 steel. A model based on hydrogen access to the crack tip has been proposed to explain the extent of intergranular failure during fatigue of an En 24 steel (Irving and Jurzfeld, 1978).

Comparison of tests in air and vacuum of quenched and tempered steels (Irving and Kurzfeld, 1978) and low carbon ferritic steels (Druce, 1977) stress lower crack growth rates and higher ΔK threshold in vacuum. This probably reflects an environmental contribution to both the transgranular and intergranular failure modes in air. In both high and low yield strength steels the % intergranular failure decreases below a ΔK of 15 to 20 $MNm^{-3/2}$ and diminishes to very small values at ΔK's of 5 $MN\,m^{-3/2}$ and below as the mode of failure is not operative at the threshold ΔK levels. Thus in many quenched and tempered steels, particularly at intermediate and high R ratios no intergranular failure occurs in the threshold region and hence cannot play a significant role in determining the ΔK threshold level.

In the low carbon ferrite steels a slightly different result was obtained. In Fig. 9 the small grain size specimen B1 had the higher ΔK threshold with 25% intergranular fracture at $\Delta K \sim 10$ $MNm^{-3/2}$. For specimen A1 with the largest grain size the intergranular failure at $\Delta K \sim 10$ $MNm^{-3/2}$ was only 10%. Thus, difference in heat treatment to obtain varying grain sizes may influence the extent of intergranular separation and the consequent effects upon crack profile and "non-closure" of the crack faces.

In α-brass, intergranular failure occurred but the resultant fracture surface morphologies (Fig. 6) were such as to be consistent with a "non-closure" explanation for the increase in ΔK threshold with grain size.

CONCLUSION

The influence of intergranular modes of separation upon fatigue crack growth rates and ΔK threshold are difficult to predict.

In metals and alloys where the fatigue crack growth is microstructurally sensitive and failure is dominantly transgranular, substantial out of plane deviations of the fatigue crack may occur. These irregular fracture faces can lead to "non-closure" of the fatigue crack and a decrease in the range of crack tip opening. A consequence of this can be the decrease in fatigue crack growth rate and increase in ΔK threshold.

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