MIXED-MODE FRACTURE MECHANISMS NEAR THE FATIGUE THRESHOLD OF AISI 316 STAINLESS STEEL

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Abstract—Near threshold, mixed mode (I and II), fatigue crack growth occurs mainly by two mechanisms, coplanar (or shear) mode and branch (or tensile) mode. For a constant ratio of $\Delta K_I/\Delta K_{II}$ the shear mode growth shows a self-arrest character and it would only start again when ΔK_I and ΔK_{II} are increased. Both shear crack growth and the early stages of tensile crack growth, are of a crystallographic nature; the fatigue crack proceeds along slip planes or grain boundaries. The appearance of the fracture surfaces suggest that the mechanism of crack extension is by developing slip band microcracks which join up to form a macrocrack. This process is thought to be assisted by the nature of the plastic deformation within the reversed plastic zone where high back stresses are set up by dislocation pile-ups against grain boundaries. The interaction of the crack tip stress field with that of the dislocation pile-ups leads to the formation of slip band microcracks and subsequent crack extension. The change from shear mode to tensile mode growth probably occurs when the maximum tensile stress and the microcrack density in the maximum tensile plane direction attain critical values.

INTRODUCTION

Although fatigue crack growth threshold behaviour of long cracks can be described by continuum mechanics (i.e. LEFM), it also requires a detailed appreciation in microstructural terms particularly under conditions of mixed mode crack extension. This is because many engineering components and structures are subjected to various combinations of axial, bending and torsional loading which may involve two or three modes of crack growth [1-4]. Even under push-pull conditions alone at stress levels close to the fatigue limit the greatest proportion of lifetime is spent in stage I crack propagation [5] with the crack extending by both mode I and mode II mechanisms.

Finally it has been shown that the stage I to stage II transition, which will necessarily involve changes in interactions between modes I and II crack growth, is dependent on the three dimensional stress state [6]. Additional complications thereafter will arise when components are subjected to non-proportional loading [7, 8]. It follows that both metallurgical and mechanical analyses are required to further our understanding of crack growth close to threshold conditions where crack extension per cycle is of the same order of magnitude as microstructural features. Such analyses will help ascertain the conditions for both infinite fatigue life and the size of defects that can be tolerated in engineering situations.

Fig. 1. The upper and lower bound curves of fatigue thresholds under mixed mode I and mode II loading; S = specimen number. (a) Pre-fatigue-cracked specimens (crack tip radius $\rho \rightarrow 0$). (b) Slit specimens (slit root radius $\rho = 0.08$ mm).

Only a few studies on fracture mechanisms have been related to mixed mode loading [9–11] especially in the near threshold region. In one such study [12] combined mode I and Mode II loads were applied under plane strain conditions in laboratory air at room temperature. Figure 1 shows the threshold conditions for (a) coplanar crack growth in pre-fatigue cracked specimens and (b) tensile mode growth of a branch crack; these two possibilities giving a lower and an upper bound threshold condition respectively.

This paper presents optical, transmission and scanning electron microscopy results to assist our understanding of fatigue crack growth under mixed mode loading with particular reference to (i) the two bounding conditions mentioned previously, (ii) the degree of plastic deformation and (iii) dislocation density and morphology, under mixed mode loading.

EXPERIMENT DETAILS

Near-threshold fatigue tests were carried out on AISI 316 austenitic stainless steel in laboratory air at room temperature. The specimens were cut from a 16 mm thick solution treated, water quenched plate with a chemical composition (wt%) of 0.049 C, 0.54 Si, 1.36 Mn, 17.26 Cr, 11.20 Ni, 2.15 Mo, 0.018 S and 0.023 P. The average austenite grain diameter was 40 μm . Test details are given elsewhere [12].

Metallographic observations on mid-thickness fracture surfaces were carried out either on mechanically polished or electrolytically polished and etched specimens (see Table 1).

Fractographic observations were carried out on a Philips P.S.E.M. 500 $\times$ scanning electron microscope (SEM) operated at 25 KV with a spot size of 64 nm. Thin slices which incorporated the fracture surface were cut by a low speed diamond cutter as a first step in the preparation of foils for transmission electron microscopy (TEM). Discs of 3 mm diameter were cut by spark machining from these slices and then ground within the range 50–100 μm mainly on the opposite side to the fracture surface before finally being electro-polished by standard twin jet techniques. The electrolyte for TEM specimens was a 5% perchloric acid in acetic acid at 10–20°C. A Philips 310 electron microscope, operated at 100 KV and equipped with a tilting specimen stage, was used for the examination of the foils. Dislocations were observed directly in the thin area surrounding the perforation by means of diffraction contrast. Dislocation density determinations were

Table 1. Preparation details for metallographic optical examinations

	Electrolytic polishing	Electrolytic etching
Solution	10% HClO_4 in ethanol	10 g CrO_3 in 100 ml of H_2O
Voltage (V)	20	6
Current (A)	2	
Temperature ($^{\circ}\text{C}$)	5	5
Time (min)	2	1.5–2

always on the (111) plane. The average dislocation density (ρ) was measured by the method of Ham [13], i.e.

$$\rho = \frac{2NM}{Lt} \quad (1)$$

where N is the number of intercepts, L is the total length of the test line, t is the thickness of the foil and M is the magnification factor. The foil thickness was determined by counting extinction fringes either at grain boundaries or at the edge of the foil.

Optical microscopy

In the near-threshold region, irrespective of what combined loading mode is employed or what mode of crack growth predominates, the crack appeared to take the most favourable fracture path offered by local crystallographic conditions. Let us first consider a mode I pre-cracked specimen for which the crack was stopped at the threshold of the

Fig. 2. Crack growth under various loading conditions. (a) (A) end of mode I growth $\Delta K_I = 4.34 \text{ MN m}^{-3/2}$ (threshold); (B) end of mixed mode (I and II) coplanar growth $\Delta K_I/\Delta K_{II} = 1.43/3.52$; (C) mixed mode loading $\Delta K_I/\Delta K_{II} = 1.7/4.22$, branch crack growth perpendicular to maximum normal stress. (b) Fractograph showing the same three stages as in (a) but in this case for specimen S6 with $\Delta K_I = 1.12 \text{ MN m}^{-3/2}$, $\Delta K_{II} = 5.53 \text{ MN m}^{-3/2}$.

material $\Delta K_{thI} = 4.34 \text{ MN m}^{-3/2}$. Figure 2 shows the crack in Specimens 5 and 6 (S5 and S6) at three stages in its growth. Point A is the termination of mode I growth at $\Delta K = 4.34 \text{ MN m}^{-3/2}$. From A to B mixed mode growth occurs at a $\Delta K_I/\Delta K_{II}$ ratio equal to 1.43/3.52. From B to C the same ratio is employed but $\Delta K_I = 1.7$ and $\Delta K_{II} = 4.22 \text{ MN m}^{-3/2}$. Zone A to B in specimen 5 (S5), for which a coplanar shear-mode growth occurred, gave a crack growth increment of 0.11 mm. Finally, as stress intensity factors were increased above the upper bound to give crack growth B to C, a branch crack in the direction of the maximum normal stress plane appeared and this eventually propagated to failure.

Figures 2 and 1(a) illustrate that between the upper and lower boundary conditions, coplanar shear mode crack growth can be arrested at grain boundaries. At the upper boundary condition of higher stress intensity levels a branch crack of a tensile mode will develop at a grain boundary. This crack may grow in either a transgranular or intergranular manner, depending on whether the crystallographic fatigue slip bands or grain boundaries with their attendant stress concentration provide an easier path for crack growth. Similar crack path deviations were reported by Pickard *et al.* [14] except that no evidence of intergranular fracture was observed.

Electron fractography

For conditions of near-threshold growth the fracture surface is mainly crystallographic [see Fig. 3(a)–(d)]. At higher stress intensities the fracture surface is not crystallographic but has an appearance more akin to a low deformation tensile fracture.

A crystallographic fracture proceeds in its growth by developing microcracks along crystallographic planes, probably assisted by the nature of the deformation process which concentrates the slip within the slip bands. These slip band microcracks are then joined up to form the macrocrack. Evidence of microcrack formation and the linking-up of microcracks to produce crack growth has been shown for aluminium alloys [15]. There is, nevertheless, a difference between cracks growing in the direction of maximum shear and those growing perpendicular to the maximum tensile stress. Figure 3(a) and 3(b) illustrate this difference. For cracks growing along maximum shear planes [see Fig. 3(b)], the direction of the slip band microcracks are approximately identical to the macrocrack orientation which is conducive to forming a relatively smooth fracture surface. Fracture surfaces normal to the maximum tensile stress on the other hand, e.g. Fig. 3(a) are much more rugged in appearance and the slip band microcrack pattern in Fig. 3(c) and (d) also shows that conditions of multiple slip prevail in these cases. Priddle *et al.* [16] have also observed “cleavage” facets in 316 stainless steel tested in air at room temperature. The facets were $\{111\}$ planes and were associated with a reduced plastic deformation in certain grains. They also showed that for ΔK values which corresponded to a plastic zone size equal to the grain size, the percentage of “cleavage” increased. We equate these observations to slip band microcrack growth within a grain that gives the appearance of a cleavage facet [see Fig. 3(a) and 3(b)].

Figure 3(c) and 3(d) provide further photographic evidence of multiple slip (arrowed) and facetted crack growth at the start of a branch crack in specimen S2. Further observations made on the branch crack surface but now advanced 3 mm away from its beginning no longer provide evidence of crystallographic effects and the surface is flat and featureless. As the branch crack increases in length under load control testing conditions, its stress intensity factor range and its crack tip reversed plastic zone size (r_p)

Fig. 3. SEM micrographs of fatigue specimens fractured under different combined mode loading conditions. (a) Mode I near-threshold growth (S1: $\Delta K_I = 4.34 \text{ MN m}^{-3/2}$) showing slip band cracking (x) and more than one operated slip system in a single grain (y). (b) Shear mode growth starting from a slit (0.08 mm root radius) $\Delta K_{II} = 4.75 \text{ MN m}^{-3/2}$ (S7). The fractograph shows slip band cracks corresponding essentially to a single slip system. (c)–(d) Observations at the beginning of the branch crack showing slip band cracks inclined to the fracture surface and more than one slip system operating in each grain.

both increase. When r_p is much larger than the grain size, reversed plastic deformation at the crack tip becomes more typical of continuous (macro) plasticity.

Dislocation morphology

The crystallographic nature of the fracture surface, especially when slip band cracking is a contributing factor, may indicate a certain link between crack growth and the process of slip within the reversed plastic zone. With these considerations in mind a study of the dislocation structure formed beneath the fracture surface was undertaken.

The dislocation structures observed are shown in Fig. 4, which correspond to the specific zones of crack growth.

The main points arising are:

- (i) The original dislocation density of uncycled material was low (see Table 2).
- (ii) One of the configurations frequently encountered within the crack tip reversed plastic zone was dislocation pile-ups in front of lattice obstacles, see arrows in Fig. 4(a).

Fig. 4. Dislocation morphology formed under different cyclic loading conditions. (a) Dislocation pile-ups (arrowed) in front of lattice obstacle $35\ \mu\text{m}$ below fracture surface. (S1: $\Delta K_I = 4.22\ \text{MN m}^{-3/2}$). (b) Dislocations mostly arranged in a single direction $25\ \mu\text{m}$ below branch crack fracture surface (S3: $\Delta K_I = 2.75\ \text{MN m}^{-3/2}$, $\Delta K_{II} = 3.28\ \text{MN m}^{-3/2}$). (c) Dislocations on different slip planes, $25\ \mu\text{m}$ below branch crack fracture surface (S3: $\Delta K_I = 2.75\ \text{MN m}^{-3/2}$, $\Delta K_{II} = 3.28\ \text{MN m}^{-3/2}$). (d) Some grouping of dislocations into dense patches but still showing appreciable orientation at $25\ \mu\text{m}$ below coplanar growth fracture surface (S4: $\Delta K_I = 2.41\ \text{MN m}^{-3/2}$, $\Delta K_{II} = 3.9\ \text{MN m}^{-3/2}$).

(iii) In all cases, planar arrays of dislocations were seen to have developed either in a single direction [Fig. 4(b)] or in two or more directions [Fig. 4(c)] depending on the number of slip systems operating. The dislocation morphology which characterized the beginning of branching is illustrated by Fig. 4(c). For the branch crack, a few grains showed dislocations on only one plane, similar to Fig. 4(b), but multiple slip predominated.

(iv) If dislocation density increased further, a localized concentration of dislocations was visible. However no rearrangement of dislocations to form three dimensional cells had taken place, the dislocations are still keeping to their slip planes as shown in Fig. 4(d), which corresponds to a foil taken near the coplanar growth crack surface.

Dislocation density

The determination of the dislocation densities in areas close to the fracture surface

Table 2. Dislocation densities

	Distance below fracture surface (μm)	Average dislocation density (mm^{-2})
Mode I near-threshold growth	25	0.65×10^8
	50	0.16×10^8
Branch crack growth	25	0.54×10^8
Coplanar growth	25	1.14×10^8
	50	0.30×10^8
Uncycled	—	0.17×10^8

serves two purposes. One, it gives an indication of the extension of the plastic zone, and secondly, it allows for the calculation of stress and strains associated with crack propagation if a suitable dislocation model can be adopted. Dislocation densities for various combined loading conditions are listed in Table 2.

The dislocation density of uncycled material, with a $40 \mu\text{m}$ grain diameter, is compared in Table 2 with that observed at 25 and $50 \mu\text{m}$ below the fracture surface of the fatigued specimen. Dislocation density of material some $50 \mu\text{m}$ below the fracture surface is very similar to that of the uncycled materials, while material at $25 \mu\text{m}$ below the fracture surface has a much higher dislocation density. This may be taken as evidence that in the near-threshold region crack tip reversed plastic zone size is of the same order as grain size.

Metals with low stacking fault energy, like 316 stainless steel [14] or 70–30 brass [17], show the clear effect of grain size on both the fatigue limit and near-threshold crack propagation. There is no effect in metals with high stacking fault energy like copper [17, 18]. This effect may be related to the type of dislocation structures encountered (Fig. 4). For 316 stainless steel tested at room temperature, planar arrays of dislocations form and the slip band extends right across the grain (i.e. equal to the grain size). However, high stacking fault energy materials are known to form more complex arrangements of dislocations. They are collected in veins or cells so that the effective slip band length may be shorter than the grain diameter. Cross slip in these materials would also reduce the intensity of slip band stress fields.

A simple argument allowed Meakin and Petch [19] to show that the dislocation density ρ and the strain γ are related by the equation

$$\gamma = \rho \frac{bd}{4} \quad (2)$$

where b is the Burgers vector ($2.5 \times 10^{-7} \text{ m}$) and d the grain size ($40 \mu\text{m}$).

For the conditions listed in Table 2 the plastic strain, calculated using the above expression, ranges from 13 to 28%, which shows the high level of the plastic strains developed in the plastic zone even under near-threshold conditions. Equation (2) also indicates that the reversed plastic deformation associated with the crack tip is higher for coplanar growth than for branch or mode I near-threshold growth.

DISCUSSION

Although no direct evidence of microcrack formation and linking-up is presented here, the Kung and Fine work [15] and the type of fracture surfaces observed here suggests that in the near-threshold region fatigue crack growth is assisted by microcracks forming ahead of the crack front. In the present case, the microcracks are mainly of two kinds, slip band microcracks and grain boundary microcracks. The linking up of these microcracks with the main crack constitutes macrocrack growth, which would then be either a transgranular or an intergranular process.

Slip band microcracks are a feature common to all the cases of slow macrocrack growth, i.e. near ΔK_{th} mode I growth [Fig. 3(a)], coplanar shear mode growth [Fig. 3(b)] and the beginning of branch crack growth [Fig. 3(c) and (d)]. The formation of microcracks must be related to the kind of dislocation structures observed. As shown earlier, the dislocations keep to their slip planes, thus forming pile-ups on reaching the grain boundaries.

High stresses, in the form of back stresses, are developed by these pile-ups which, when combined with the crack tip stresses, are sufficiently high to form the slip band microcracks. These back stresses can be calculated approximately by the method proposed by Petch [19] based on the work by Li [20]. The parallel pile-up of dislocations are modelled by a wall of superdislocations of strength nb , where n is the number of dislocations in the pile-up.

The stress resolved along the slip plane is then

$$\tau = \frac{\mu \rho b d}{20\pi} \quad (3)$$

where μ is the shear modulus $= 7.6 \times 10^4 \text{ N/m}^2$.

For the conditions listed in Table 2 the calculated back stresses range from 660 to 1400 N/mm^2 . These are large enough stresses in themselves to produce cracking, probably too large to be realistic, and therefore in a real situation some relaxation of the pile-ups must take place, possibly by initiating slip in the neighbouring grain.

Nevertheless it clearly illustrates the feasibility of slip band cracking being initiated by the interaction of the back stress and the crack tip stress itself. If only a single slip system operates, the arrangement of slip band microcracks must be restricted to one direction, while if multiple operated slip systems cross each other, slip band microcracks may then be formed in different directions, not only by the process mentioned above but also by the process in which dislocations in two intersecting slip bands interact with each other.

For the formation of grain boundary cracks the pile-ups again play an important part. Here the wedge cracks associated with two slip bands of opposite sign may interact at the boundary to produce an intergranular microcrack; or the combined effort of the main crack stress field and an extra stress concentration at the front of a pile-up may fracture a weakened grain boundary.

The mechanism of microcrack formation in the near-threshold region seems to be the same, no matter whether the fatigue crack extends in a shear mode or in a tensile mode. However, the macrocrack growth direction for the former is approximately in the maximum shear plane, while for the latter, it is nearly perpendicular to the maximum normal stress.

Table 2 showed that the dislocation density associated with coplanar growth is higher than that related to branching and near ΔK_{th} mode I growth. This may indicate an easier process for microcrack linkage under a tensile stress rather than under a shear stress. Thus it appears that only when the maximum tensile stress and the microcrack density in the maximum tensile stress plane direction both attain critical values, can a tensile mode growth occur. If the two conditions cannot be satisfied simultaneously, or in other words, if either the maximum tensile stress or the microcrack density in the maximum tensile plane direction is too low, a tensile mode growth would not be possible.

In the case of stage I fatigue crack growth, where slip is almost on a single system, microcracks caused by slip band dislocation movement are restricted to fewer directions and consequently the microcrack density in the maximum tensile plane direction is not satisfied and only shear mode growth occurs.

In the present case, as shown in Fig. 4(c) and 4(d), for both coplanar growth and branch crack growth, there are two or more slip systems operating. However, a branch crack could only form, or in other words the tensile mode growth could only occur, when the apparent values of ΔK_I and ΔK_{II} corresponds to the upper bound curve of Fig. 1(a). The view put forward here is that a combined mode loading introduces via the mode II component an extra effect associated with crack closure, crack surface contact and rubbing. These make the crack surface no longer stress free, so that the effective stress intensity factors (ΔK_{Ieff} and ΔK_{IIeff}) of the crack are lower than the apparent stress intensity factors (ΔK_I and ΔK_{II}). Only if ΔK_{Ieff} and ΔK_{IIeff} attain critical values will a tensile mode branch form; see also Ref. [12].

It was also shown in previous work [12] that by using slit specimens instead of pre-fatigue-cracked specimens, so as to minimize crack closure, crack surface contact and rubbing, then the upper boundary curve would decrease to a minimum value and become approximately coincident with the lower bound curve [Fig. 1(b)]. In such a case, the crack propagated in the vicinity of the notch tip by a shear mode with only a single slip system operating [Fig. 3(b)]. After growing a few grain diameters it then changed into a tensile mode growth with the new branch in the maximum tensile plane direction. This was when ΔK_I and ΔK_{II} reached their critical values.

CONCLUSIONS

- (1) The fatigue crack growth for mixed-mode near-threshold conditions is highly crystallographic, the crack invariably following the weaker path, i.e. either in a transgranular or an intergranular manner.
- (2) For both tensile-mode and shear-mode crack growth, microcrack formation is closely related to dislocation interactions within the slip bands.
- (3) Macrocrack growth involves the joining up of microcracks into a main crack, either by the shear stress (coplanar growth) or by the tensile stress (branch growth). Tensile growth becomes dominant probably when the maximum tensile stress and the microcrack density in the maximum tensile plane direction attain critical values.

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