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Dislocation substructure associated with propagating fatigue crack

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Summary
Transmission electron microscopy was used to study the dislocation substructure in the vicinity of the fracture surface and in front of the crack tip during tensile mode crack growth in copper single crystals oriented for single slip and cycled at a constant stress amplitude leading to fracture after $\sim 10^6$ cycles.

Dislocation substructure near the fracture surface was characterized by a very fine, highly disoriented cell structure having a cell size of the order of 0·1 μm. A correlation between cell size and striation spacing was shown. The changes in the dislocation distribution in the plastic zone in front of the crack tip are also described. The experimental results threw doubts about the possibility of crack propagation occurring by some proposed simple dislocation mechanism and suggest that the strain relaxation mechanism requiring plastic deformation at the crack tip is more realistic.

Introduction

Earlier investigations [1, 2, 3, 4] have revealed two types of fatigue crack growth. The shear mode crack growth is characterized by propagation along persistent slip bands in the surface layer of the specimen and by a growth rate of the order of angstroms per cycle. After the crack has penetrated a certain depth, the shear mode growth is followed by tensile mode crack growth. The crack now spreads perpendicular to the maximum tensile stress both micro- and macroscopically. This paper gives the results of direct observation of dislocation configurations associated with crack growth during the tensile mode stage and considers the justification of some models explaining the mechanism of crack propagation.

Experimental procedure

Single crystals of copper (purity 99·99%) having a cross section 3 mm × 7 mm were subjected to push-pull loading at a resolved shear stress amplitude of $\pm 3{\cdot}25$ kg/mm^2 corresponding to the fatigue life 10^6 cycles. Specimens having $(1\bar{2}1)$ and (hkl) surfaces were oriented for single slip (Fig. 1). The orientation of the surface (hkl) is near to the orientation $(54\bar{1})$.

Optical microscopy revealed that a crack followed exactly a (111) plane to a length of about 0·4 mm after which it began to deviate continuously towards a direction perpendicular to the specimen axis.

After fracture the broken halves of a specimen were electrodeposited with copper. Thin sheets parallel to the surface ($1\bar{2}1$) or (hkl) were cut with an electric spark saw. From these sheets thin foils were prepared in the usual way. To study the dislocation structure in front of the crack tip, where the plastic zone is formed, the loading was interrupted at given crack lengths. Thin sheets parallel to the surface ($1\bar{2}1$) were cut off and ground to the usual thickness. The area just in front of the crack tip was thinned from both sides by a jet polishing technique, polishing then being contined by conventional means. In this way it was possible to get foils from a distance 0·1 mm and more in front of the crack tip.

Experimental results

Results of the transmission electron microscopy investigation of the dislocation substructure associated with the shear mode crack growth have been given previously (7); those relevant to the tensile mode crack growth are given in this paper.

Fig. 2 shows the interface between the electrodeposited copper layer and the specimen; the foil was parallel to the surface ($1\bar{2}1$). The observed area corresponds to a crack length of 3·5 mm. The micrograph shows the relief of the fracture surface in the direction of crack growth; the striation shape is in good agreement with that implied by the Laird relaxation mechanism [3]. The distance between striations corresponding to the crack growth during one cycle is visible; it represents dimension of several cells in length.

The tiny cell structure in the vicinity of the fracture surface is also shown in Fig. 3 which was obtained from the foil perpendicular to the direction of crack growth. In this (hkl) section the fracture surface is rather smooth.

With shorter cracks the lower rate of crack growth rate results in a smaller distance between striations. Fig. 4 shows the interface between the electrodeposited layer and the specimen in the area corresponding to a crack length of about 1·5 mm. The cells appear to be a little bit larger than those in Fig. 2, implying that the intensity of cumulative plastic deformation in the plastic zone decreases with decreasing crack length. In the direction perpendicular to the fracture surface the cell structure changes into a band structure in the distance of about 200 μm. To see whether the cell structure is a product of the rubbing together of the fracture surfaces, the dislocation structure in the plastic zone ahead of the crack tip was examined. The size of the plastic zone connected with a crack of the length of 1·5 mm long was 0·7 mm (measured in the direction of crack growth). The boundary between the plastic zone and the matrix is shown in Fig. 5; the transition from the band structure typical of the whole interior of the crystal after fatigue hardening to the cell structure is clearly seen. At a distance 0·4 mm in front of

the crack tip the cells are already well developed (Fig. 6) with ordered directions of their boundaries. At a distance of about 0·1 mm in front of the crack tip a disoriented cell structure exists (Fig. 7), the cell size being comparable to that observed at the same distance from the fracture surface.

These results demonstrate that the tiny cell structure is formed by cyclic plastic deformation in the plastic zone in front of the spreading crack; it is neither the product of the rubbing together of the fracture surfaces nor the immediate product of the deformation at the crack tip by its spreading during one cycle.

Discussion

Despite the fatigue process being composed of several stages (cyclic hardening, crack nucleation, crack growth) it is possible to state that a crack spreading from the crack length of several μm to final fracture represents, in ductile metals, the most important stage. The stage of hardening, which corresponds to the arrangement of lattice defects in the bulk of the specimen, is finished comparatively quickly in copper single crystals; in the present tests after about 10^4 cycles, which represents 1% of the whole fatigue life. The second stage is characterized by the formation of persistent slip bands in the surface layer; in the case of copper single crystals (oriented for single slip) persistent bands reach a depth of several hundreds of μm and some of them cross the entire section [5, 6].

As shown previously [7] the surface layer of the fracture surface created by shear mode crack growth is composed of small cells, whose diameter increases with depth below the fracture surface; beyond a depth of several μm the cell structure changes into a band structure. It is generally assumed and it has also been proved in previous experiments [7, 8, 9] that a shear mode crack propagates along a persistent slip band. It was concluded that at the crack tip of a shear crack spreading along a persistent slip band, multislip takes place which results in the formation of small cells. Because of the evidence indicating that persistent slip bands are softer than the surrounding metal, it can be assumed that the plastic deformation at the crack tip and also the crack spreading will be confined to the persistent slip band.

As a crack grows longer it devalues from a crystallographic phase and tends to follow the plane perpendicular to the maximum tensile stress; its rate of growth increases and striations become resolvable on the fracture surface. There is considerable experimental evidence showing a one-to-one correspondence between striation spacing and the stress cycle.

Several models of crack propagation have been proposed. Some of them (e.g. by Forsyth or Neumann [1, 10]) are based on a simple motion and interaction of dislocations at the crack tip during one stress cycle. Laird [3] proposed a mechanism supposing plastic relaxation at the crack tip during the tensile part of a fatigue cycle followed by resharpening of the crack in the compression part. This model which explains nearly all phenomena connected with crack growth in both metals and in plastics assumes, that the crack extension mechanism is not dependent on the details of dislocation movements, but is governed only by the average plastic deformation they produce.

Holden [11], and Grosskreutz and Waldow [12, 13] observed by means of X-ray analysis the highly disoriented cell structure near the fracture surface created by a fatigue crack. Grosskreutz and Shaw [14] also observed, by means of transmission electron microscopy, a cell structure several μm below the fracture surface but their experimental technique did not allow them to get micrographs from a smaller distance from the fracture surface.

The findings reported in this paper, especially the existence of a fine, highly disoriented cell structure both directly at the fracture surface and in front of the crack tip throws serious doubt on models assuming the formation of fracture surface (formation of striations) during one cycle through a simple movement of dislocations, this is especially so when it is realized that the fine cell structure is already formed in the plastic zone in front of the crack tip. The idea that the fracture surface follows the boundaries of cells, as was suggested by both Holden and Grosskreutz [11, 12, 13] was based on indirect evidence and present experiments do not support the validity of this mechanism. On the other hand they do justify the relaxation model, because the fragmented metal in the crack vicinity can behave as a quasihomogenious medium; in this case the movement of dislocations will produce only the plastic deformation necessary for the blunting of the crack tip.

Conclusions

Experimental results have shown that during tensile mode crack growth, a tiny cell structure is formed as the product of multislip in the plastic zone ahead of the crack tip; the size of the cells is of the order of tenths of μm. The striation spacing observed in the later stages of crack growth is much higher than the cell size. These findings suggest that the distance a crack grows during one cycle is not controlled by the simple movement and interaction of dislocations and a relaxation mechanism seems to be more realistic. The plastic zone in the second stage has macroscopic dimensions, whereas in the first stage multislip is probably confined to a persistent slip band, which represents a soft area on the microscopic scale, in the cyclically hardened matrix. It can be assumed, that shear mode crack growth is controlled by the same mechanism.

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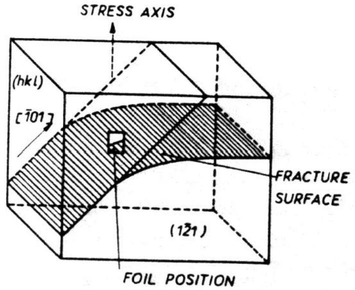

Fig. 1. Orientation of the specimen and the position of the foil.

Fig. 2. Dislocation structure associated with the fracture surface at a position corresponding to a crack length of about 3·5 mm; section ($1\bar{2}1$).

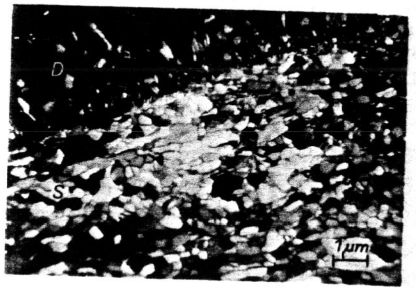

Fig. 3. As in Fig. 2; section (hkl).

Fig. 4. Dislocation structure associated with the fracture surface at the position corresponding to a crack length of about 1·5 mm; section ($1\bar{2}1$).

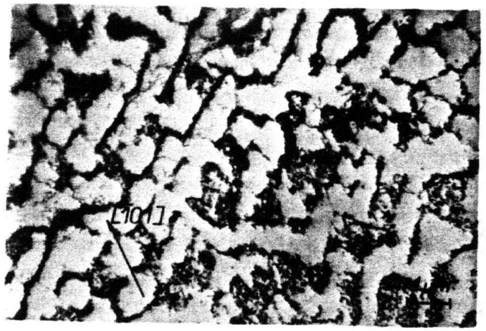

Fig. 5. Transition from the plastic zone (top left) to the matrix structure (bottom right); section $(1\bar{2}1)$.

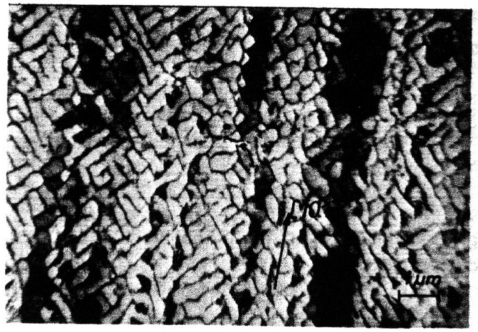

Fig. 6. Dislocation structure in the plastic zone ahead of the crack tip. Crack length of about 1·5 mm; distance from the crack tip of about 0·4 mm; section $(1\bar{2}1)$.

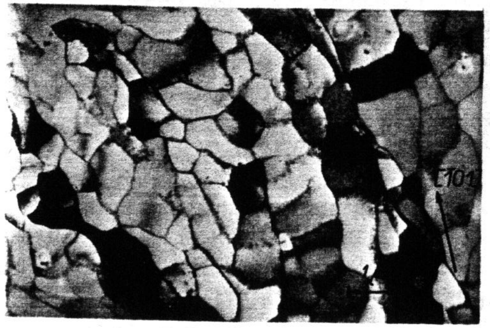

Fig. 7. As in Fig. 6, distance from the crack tip of about 0·1 mm.