THE ROUTE OF FATIGUE CRACKS AS DEDUCED FROM DISLOCATION PATTERNS

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ABSTRACT

An ultrahigh voltage electron microscope was used to observe dislocation structures formed near the fatigue crack tip. In the case of 70/30 brass, extremely dense dislocations were found on the primary and the conjugate slip planes just ahead of the tip of the tensile mode crack, and the crack was expected, under such a dislocation array, to grow by slip occurring alternately on these planes. On the other hand, in copper and iron, a well-defined cell structure was developed in front of the crack tip. However, there were observed a considerable number of cracks crossing the cell boundaries, and fatigue striations formed independently of the dislocation structure. Therefore, the cell structure is not necessarily associated with the propagation mechanism of the tensile mode crack in copper and iron. Some models proposed for fatigue crack propagation and the effect of stacking-fault energy on crack growth rates are briefly discussed in the light of the new evidence .

KEYWORDS

Fatigue; crack tip; crack route; slip plane; dislocation structure; cell boundary; stacking-fault energy; transmission electron microscope.

INTRODUCTION

To understand the mechanisms of fatigue crack propagation, it is essential to gain information on the crack route. For this purpose, considerable efforts have been made to observe the substructures developed very near the crack. Ogura and Karashima (1974) used specimens having a thinned portion and examined dislocation structures near the crack formed there, assuming that these structures were the same with those formed in fatigued bulk specimens. On the other hand, Wilkins and Smith (1970), and Grosskreutz and Shaw (1972) directly observed those developed in fatigued bulk specimens of Al-Mg alloy and several Al alloys, respectively. Such observations, however, are made through a transmission electron microscope, so that thin foils transparent to electron beams must be prepared from the bulk specimens, without losing regions near the crack tip. Because of this difficulty, electron micrographs supplying sufficient information about the immediate vicinity of the crack tip are scarce. Using both a jet-polishing technique similar to that adopted by Unwin and Wilkins (1969) and an ultrahigh voltage electron microscope, the present authors (Awatani and others, 1976a, 1976b, 1978a, 1978b, 1979; Katagiri and others, 1977a, 1977b, 1978) succeeded in observing the crack tip dislocation morphology in several metals and alloys. To get more information about the route of fatigue crack, further experiments were made using some improved thinning technique for electron microscopy foils. The results are given in this paper.

EXPERIMENTAL PROCEDURE

The materials examined were polycrystalline iron (C 0.01, Mn 0.28, Cu 0.02, Si 0.01 and Fe remainder in Wt pct), Copper (99.98 pct) and 70/30 brass (Cu 70.24, Zn 29.62 pct). The stacking-fault energy (γ) of copper and 70/30 brass are approximately 40 erg/cm^2 and 7 erg/cm^2, respectively; a considerably high value is expected for iron, because the distance between nearest neighboring atom-layers becomes small in bcc metals. These materials were annealed under the same conditions as those reported

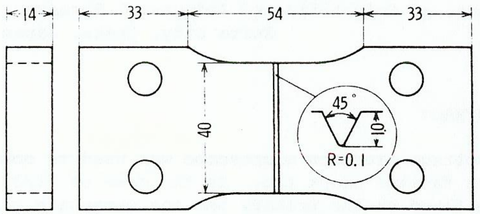

Fig. 1. Test piece used. (All dimensions in mm.)

in the references (Awatani and others, 1978a, 1979), and then machined to the test piece as shown in Fig. 1. Fatigue tests were done under completely reversed plane bending of a constant moment.

To examine the correlation between the crack tip dislocation pattern and the stress intensity factor (ΔK), crack length observations were made for the crack appearing on the sides of the test piece. When the crack grew to a required length, the test piece was removed from the machine and mechanically cut in slices perpendicularly to the surface and along the longitudinal direction of the test piece at intervals of about 4 mm, and then thinned from the front and the rear surfaces to the thickness about 0.4 mm using abrasive papers. These slices were subjected to a jet electro-polishing followed by a final electropolishing. Both the site and the depth of the indent formed by the jet polishing were suitably chosen, so that the crack was attacked just when the vicinity of the crack tip was perforated in the final electro-polishing. A good electron microscopy foils about 2 μm in thickness were able to be obtained with the probability of 20% or so, and observed through an electron micro-scope operating at 2000 kV.

EXPERIMENTAL RESULTS AND DISCUSSION

As mentioned later, the propagation rate of shear mode crack in the initial stage of fatigue seems to be insensitive to fatigue-induced substructures. In this paper, observation results of the route of tensile mode crack and the dislocation structures relevant to it are given, excepting only one shear mode crack.

Figure 2 shows the typical dislocation pattern just ahead of the crack tip in copper, which reveals a well-defined cell structure with cells somewhat elongated in the radial directions from the tip. The cell size seems to increase with increasing distance from the tip. The structure is essentially the same with that formed very

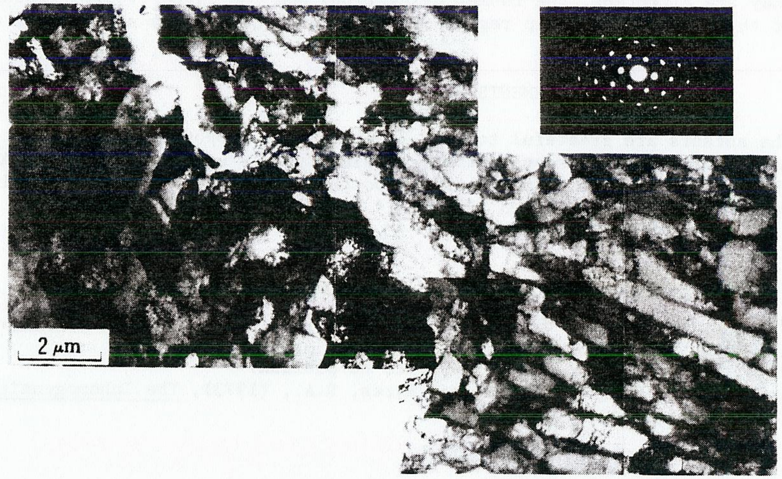

Fig. 2. Dislocation structure representative of the area just ahead of the crack tip in copper (ΔK = 206 N/mm$^{3/2}$). [By courtesy of ASM and AIME, from Awatani and others, 1979.]

near the crack tip in fatigued iron specimen (Awatani and others, 1978a). The fatigue-induced substructures near the tensile mode crack do not vary so much with a difference in crystal structure fcc or bcc, although they depend on the value of stacking-fault energy.

It has been considered that these cells, especially their boundaries, entail some significant consequence for the growth rate of the tensile mode crack in materials with high stacking-fault energy. In the present experiment, however, there were observed a considerable number of cracks growing across cell boundaries. Some of them are demonstrated in (a) and (b) of Fig.3; notice the areas indicated by the

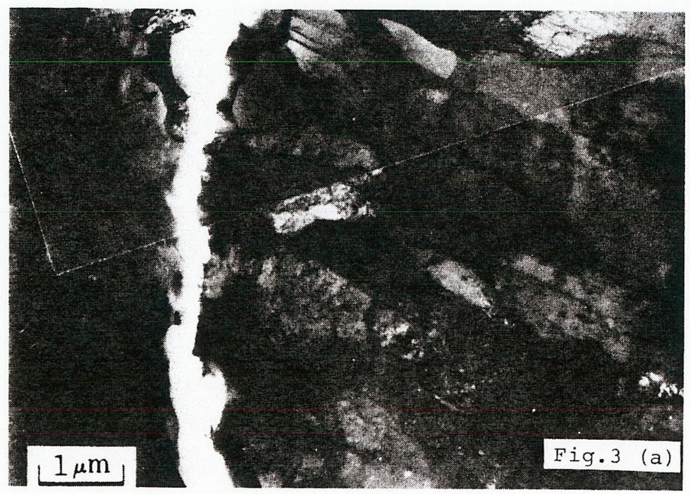

Fig.3 (a)

Fig. 3. Transmission electron micrographs showing that the crack propagates across a cell boundary. $\Delta K = 420\ N/mm^{3/2}$ in (a) and $830\ N/mm^{3/2}$ in (b). (Iron.)

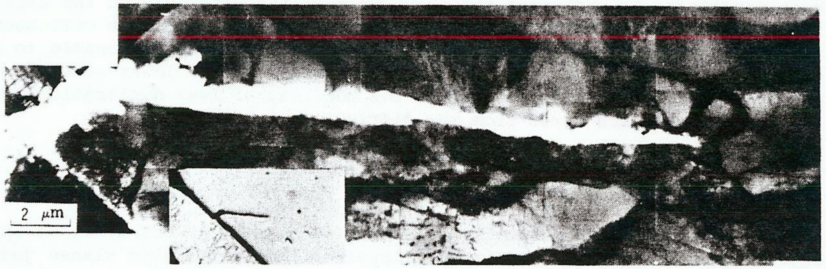

Fig. 4. Transmission electron micrograph showing that the shear mode crack runs the tip into a cell after passing the boundary. (Iron.)

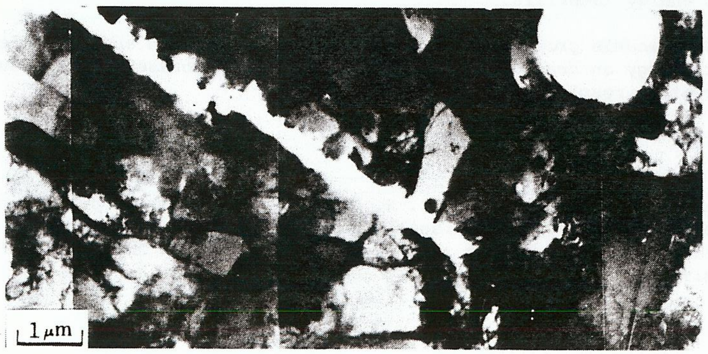

Fig. 5. Crack route with a serrated profile; notice that the profile is formed independently of the cell structure around there. (Iron.)

arrows. Also, such a crack was frequently found at the early stage of fatigue in iron: One of them is given in Fig. 4, which shows the crack tip running into a cell after passing through the boundary. Some more micrographs of cracks presenting the similar behavior are given in the references (Awatani and others, 1976a, 1978a, 1979; Katagiri and others, 1979). Besides these, there were sometimes observed the crack route with a serrated profile corresponding to striations of the fracture surface (Fig. 5). However, it seems to be formed independently of the cell structure around there. These results suggest that a cell structure as well as a cell boundary is not necessarily associated with the propagation rate of the tensile mode fatigue crack in materials with high stacking fault energy.

In the case of 70/30 brass, slip lines appeared more or less near the tip of the tensile mode crack during thinning of foil specimens. One of the transmission

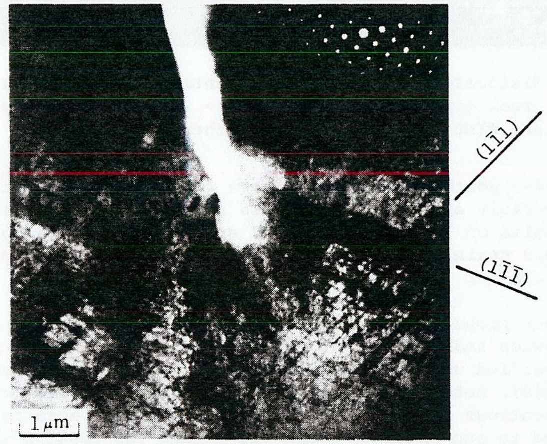

Fig. 6 Transmission elecron micrograph of the area where numerous slip lines appeared during thinning (70/30 brass). Traces of the active slip planes are $(1\bar{1}1)$ and $(1\bar{1}\bar{1})$; foil plane (110). [By courtesy of ASM and AIME, from Awatani and others, 1979.]

electron micrographs of areas developing numerous slip lines is shown in Fig. 6. Electron diffraction patterns taken from there reveal that two slip systems $(1\bar{1}1)$ and $(1\bar{1}\bar{1})$ are operative, and the dislocation density there is rather low. Such a slip seems to occur as a result of polishing away barriers to dislocation motion like a grain boundary in the course of thinning, and the escape of a number of dislocations to foil surfaces.

In areas where relatively few slip lines appeared, however, dislocation structures as shown in Fig. 7 were observed in front of the crack tip. It can be seen that the dislocation density is considerably high just ahead of the crack tip and reduced with increasing distance from the tip. Because of high density, the details of the structure near the tip are somewhat obscured, but from the inset electron diffraction patterns, most of dislocations there seem to be accumulated on the planes $(1\bar{1}1)$ and $(1\bar{1}\bar{1})$, instead of forming cells. Before thinning, much more dislocations are supposed to be closely packed in these planes in front of the crack tip.

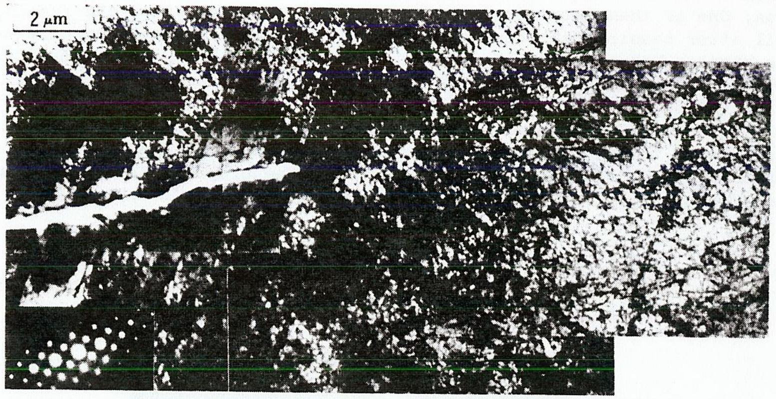

Fig. 7. Dislocation structure representative of the area just ahead of the crack tip in 70/30 brass (ΔK = 225 N/mm$^{3/2}$). [By courtesy of ASM and AIME, from Awatani and others, 1979.]

As is well known, whether or not cells are formed by cyclic stress depends on the value of stacking-fault energy (γ). In this point, it is interesting to compare the present results on 70/30 brass ($\gamma \cong 7$ erg/cm^2), in which no cell was found, with those on 18/8 stainless steel ($\gamma \cong 10$ erg/cm^2), in which rough cells were developed near the crack tip (Awatani and others, 1976b).

Fatigue crack propagation models with specification of the slip processes were proposed by Bowles and Broek (1972), Neumann (1969, 1974) and Pelloux (1969). These may be classified under the so-called plastic blunting process (Laird, 1962, 1969; Tomkins, 1968), but they can explicitly explain the crack extension per cycle. Very dense dislocations on the planes $(1\bar{1}1)$ and $(1\bar{1}\bar{1})$ as seen in fatigued brass specimens are expected to run into the crack tip during the tension stroke of stress cycle. If so, the crack will grow in the manner shown by these models.

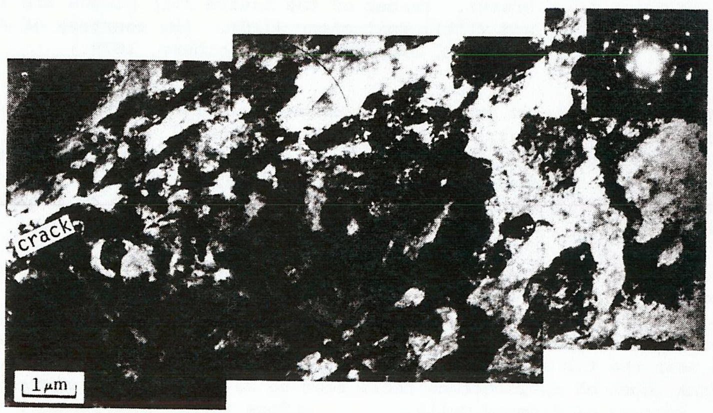

Fig. 8. Transmission electron micrograph around the unusual fatigue crack observed in 70/30 brass.

In the cases of copper and iron, it is doubtful whether most dislocations are lying on the specific slip planes as $(1\bar{1}1)$ and $(1\bar{1}\bar{1})$. (Refer to the electron diffraction patterns inset in Fig. 1.) Also, a few cracks surrounded by highly distorted zones as shown in Fig. 8 were found even in 70/30 brass. In such cases, it seems difficult to judge these models from the dislocation patterns observed.

Next, voids as observed in thin foils subjected to cyclic stress were not found in the present experiments. A void-linking mechanism seems unrealistic in fatigue of bulk specimens. A severely work-hardened zone was not observed, either, which was presumed to be formed just ahead of the crack tip and closely associated with the crack extension: If formed, the zone width should be equivalent to the crack extension per cycle.

Finally, the effect of stacking-fault energy on the crack growth rate will briefly be discussed. Avery and Backofen (1962) revealed that the cycling period necessary for the development of first crack about 12 μm in depth in Cu-Al alloys cycled with the same plastic strain amplitude was independent of stacking-fault energy. Total life, however, was markedly increased as the stacking-fault energy was lowered. This fact suggests that the growth rate of the tensile mode crack is sensitive to the substructures developed, even though insensitive in the early stage of fatigue, in which the crack is in shear mode. To explain this, the stacking-fault energy has been considered to influence the crack propagation rate through the effect of cell formation. The present observations, however, reveals that the cell boundaries do not necessarily constitute a plane of weakness. It seems reasonable to consider that the lower propagation rate of tensile mode crack in lower stacking-fault energy materials are due to the decrease of the mobility of free dislocations in front of the crack tip, which comes from the difficulty of cross slip.

CONCLUSIONS

1. In low stacking-fault energy materials such as 70/30 brass, it is expected that extremely dense dislocations are developed on the active slip planes just ahead of the fatigue crack tip, and the crack grows by slip occurring alternately on these planes

2. The cell structure formed in front of the crack tip does not seem to be associated with the propagation mechanisms of the tensile mode fatigue crack in high stacking-fault energy materials.

3. Some models proposed for fatigue crack propagation and the effect of stacking-fault energy on crack growth rates are briefly discussed on the basis of the observation results.

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