Fracture 1977, Volume 2, ICF4, Waterloo, Canada, June 19 - 24, 1977

DISLOCATION STRUCTURES AROUND THE CRACK TIPS IN THE EARLY STAGE IN FATIGUE OF α-BRASS

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INTRODUCTION

There is a lot of experimental evidence showing that fatigue microcracks are formed within persistent slip bands. Dislocation arrangements in these slip bands have been eagerly studied, especially in copper [1~5]. Now, it seems very important to obtain direct evidence concerning the correlation between these dislocation structures and microcracks. Recently, by means of a high voltage electron microscope, the authors have succeeded in observing the dislocation structure just around the crack tip in the early stages of fatigue in copper bulk specimens [6]. This paper describes the results of the examination of the dislocation structure formed around the crack in the early stage of fatigue in α-brass. The mechanism of fatigue crack initiation and growth in this material is also discussed briefly in connection with that in copper.

EXPERIMENTAL PROCEDURE

The material used was polycrystalline 70/30 brass (Cu 69.1, Fe 0.01 wt%) extruded bar with a diameter of 10 mm. After being rolled to 3 mm thickness, it was machined into specimens having a cross section 3 mm x 10 mm. The specimens were then annealed at 873K for 1 h in Ar, and finally electro-polished to remove a surface layer of about 30 μm before testing. The grain size was about 100 μm in diameter.

Using a Shimazu UF-15 fatigue testing machine operating in plane bending at 33 Hz, the specimens were cycled with constant stress amplitude of 118 MPa (life time $N_f = 2.5 \times 10^6$ cycles). Tests were stopped at 0.8 N_f and observations were made for small cracks formed in the specimens.

Thin foils containing small cracks could be prepared as follows: First, fatigued specimens were electroplated with copper to ≈ 2 mm thickness and were mechanically cut longitudinally into slices perpendicular to the specimen surface. The slice (2.5 mm thick) was then thinned initially by abrasive papers, followed by electropolishing in order to remove the damaged layer. Then, the region in which a small crack is observed is thinned by a jet electropolishing technique. Finally, this area was electrolytically polished until perforation was witnessed by an optical microscope.

The foils were examined by a high voltage electron microscope operating at 2000 kV.

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OBSERVATIONS AND DISCUSSION

Dislocation Structure around the Crack in the Early Stage of Fatigue

Dislocation configurations and small cracks just beneath the specimen surface are shown in Figure 1. Here, slip bands on the specimen surface are clearly seen as irregularities of the surface contour consisting of an interface between the electrodeposited layer and the specimen. The height of steps reaches ≅ 0.5 μm. Though active slip planes containing dislocation lines and loops of high density correspond to the surface irregularities, no special dislocation structure associating with them is observed, as has been reported [7]. Several cracks not longer than 1 μm are initiated within some of these active slip planes. Three cracks have grown to the length of 10~20 μm. The route of the crack growth partially lies either on the trace of primary slip planes or on that of the secondary slip planes. Dislocation structures ahead of the cracks as well as just beside them were not different from the ones in the matrix.* And, therefore, it appears that no distinct plastic zone is associated with cracks in this stage. Contrast of the dislocations surrounding the crack as well as ones in the matrix disappeared by tilting the foil (Figure 1(b)). This suggests that all dislocations around the crack have the same burgers vector as in the case of the ones in the slip planes [8].

Cracks growing along the twin boundary were also observed. These kind of cracks, however, did not always grow exactly along the twin boundary (Figure 2) as the transcrystalline cracks along the slip planes.

Dislocation configurations near the sides of the transcrystalline crack which had grown up to the length ≅ 150 μm (Figure 3) were thought to be essentially the same as those in the initial stage, though some entanglement is observed.

Some evidence was obtained which showed that the grain boundary was acting as a barrier to crack growth. The transcrystalline crack shown in Figure 4 seems to have changed its direction of growth near the grain boundary. Figure 5 shows a crack which is growing in the second grain after crossing the grain boundary. Though a marked tortuosity of the crack route accompanied with dislocation entanglements of high density were observed just before and after crossing the grain boundary, most of the crack lies within the matrix structure without any special accompanying dislocation structures. Thus a change in the mode of cracking [9] does not always take place after the crack penetrates into the second grain [8]. Extinction of contrast of dislocations in the vicinity of this crack was again observed by tilting the foil.

No manifest void formation as reported in Al [10] was observed just ahead of the cracks growing along the slip planes in this material prepared from bulk specimen.

Nucleation and Growth of Crack

The dislocation configuration around the crack in the early stage of fatigue in α-brass mentioned above is fairly different from that in copper as shown in Figure 6. In copper, cracks are initiated and grow

*Dislocation configuration of the matrix in this specimen was those as shown in Figure 1 and highly tangled one or cell structure were not observed.

along the "ladder" structures accompanying no appreciable plastic zone ahead of their tips. Therefore, the "ladder" structure is thought to act as a channel for mobile dislocations contributing to the crack initiation and growth. Fractographic observations*, however, revealed that the morphology of the fracture facets of the cracks within the surface slip bands was essentially the same in both α-brass and copper specimens (Figures 7(a) and (b); steps in Figure 7(a) are considered to correspond to such parts of the crack route as shown in Figure 1, which lie in the trace of the secondary slip plane). Considering these circumstances, the following mechanism of the crack nucleation and growth in α-brass is plausible. Formation of a free surface (crack) as a product of dislocations escaping to the specimen surface is probable in the active slip planes on which a high density of dislocation lines are present because of difficulty in cross slip, and continued growth is also possible with the aid of the stress concentration of the crack itself. This process can also be the case with copper; cracks are frequently observed to grow within the mid-width of "ladders" in which no tangled dislocations are present. Channels corresponding to the persistent slip bands have also been reported in Al [11,12] and Al alloys [13~15].

However, the possibility still remains that cracking may occur by a reduction in cohesive force of slip planes due to increased density of dislocation and debris on them associated with concentrated normal stress of the crack [16,17]

ACKNOWLEDGEMENT

The authors would like to thank Professor H. Fujita and the members of HVEM Centre in Osaka University for giving them facilities to use the high voltage electron microscope.

REFERENCES

1. LAUFER, E. E. and ROBERTS, W. N., Phil. Mag., 10, 1964, 883; ibid., 14, 1966, 65.
2. LUKAS, P., KLESNIL, M., KREJCI, J. and RYS, P., Phys. Stat. Sol., 15, 1966, 71.
3. LUKAS, P., KLESNIL, M. and KREJCI, J., Phys. Stat, Sol. 27, 1968, 545.
4. WOODS, P. J., Phil. Mag., 28, 1973, 155.
5. FINNEY, J. M. and LAIRD, C., Phil. Mag., 31, 1975, 339.
6. KATAGIRI, K., OMURA, A., KOYANAGI, K., AWATANI, J., SHIRAISHI, T. and KANESHIRO, H., unpublished results.
7. LUKAS, P. and KLESNIL, M., Phys. Stat. Sol., 37, 1970, 833.
8. THOMPSON, A. W. and BACKOFEN, W. A., Acta Met., 19, 1971, 597.
9. FORREST, P. G. and TATE, A. E. L., J. Inst. Metals, 93, 1964-65, 438.
10. OGURA, T. and KARASHIMA, S., Trans JIM, 15, 1974, 324.
11. LEVINE, E. and WEISSMANM, S., Trans ASM, 61, 1968, 128.
12. MITCHELL, A. B. and TEER, D. G., Phil. Mag., 19, 1969, 609.
13. FORSYTH, P. J. E., Acta Met., 11, 1963, 703.
14. McGRATH, J. T. and WALDON, G. W. J., Phil. Mag., 9, 1964, 249.
15. CLARK, J. B. and McEVILY, A. J., Acta Met., 12, 1964, 1359.

*The details are to be published elsewhere.

16. GROSSKREUTZ, J. C., REIMANN, W. H. and WOOD, W. A., Acta Met., 14, 1966, 1549.
17. GELL, M. and LEVERANT, G. R., Fracture, 2nd Intern. Conf. on Fracture, Brigton, 1969, 565.

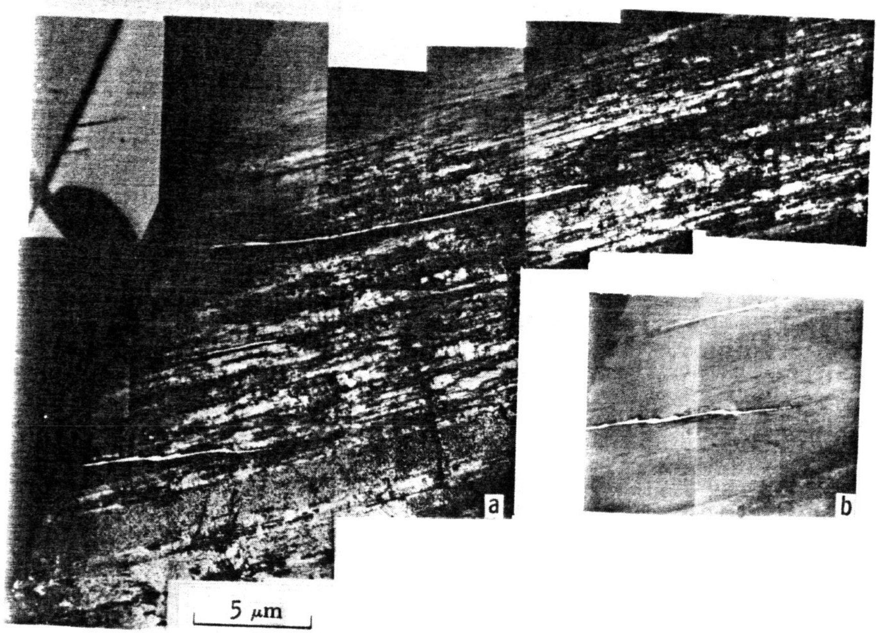

Figure 1 Dislocation configuration and short cracks just beneath the specimen surface [plane of the foil (1$\bar{2}$1)].
(a) dislocation lines and loops on the primary slip plane
(b) bottom of (a) tilted.

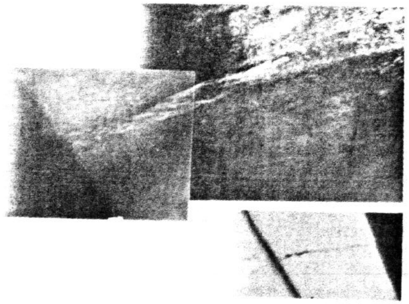

Figure 2 A crack growing along the twin boundary

Figure 3 Dislocation structure near the sides of the crack which have grown up to the length ≅ 150 μm.

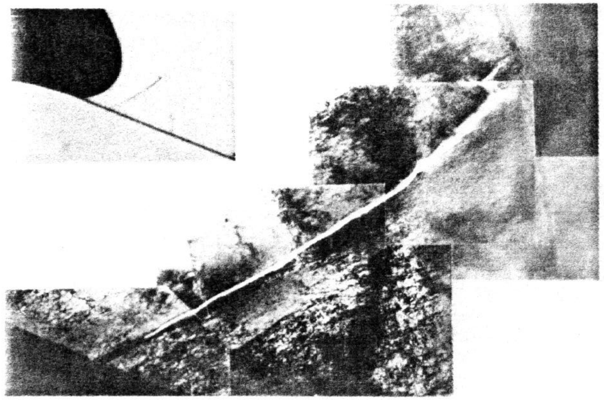

Figure 4 A crack changing its growth route near the grain boundary

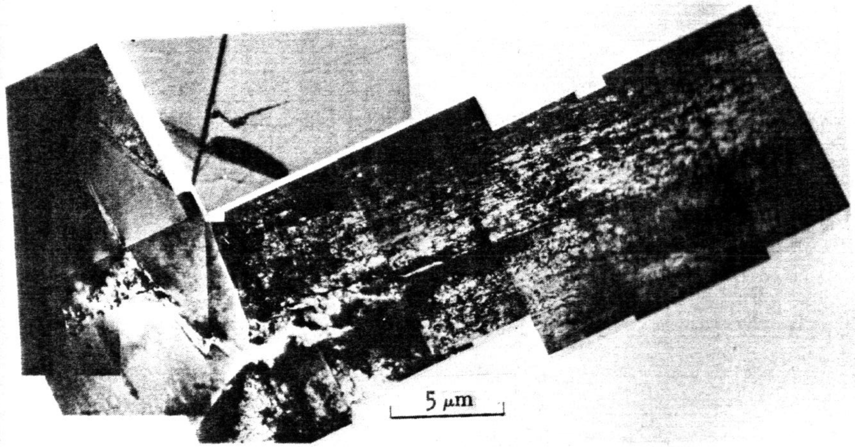

Figure 5 Dislocation structure around the crack in the second grain after crossing the grain boundary

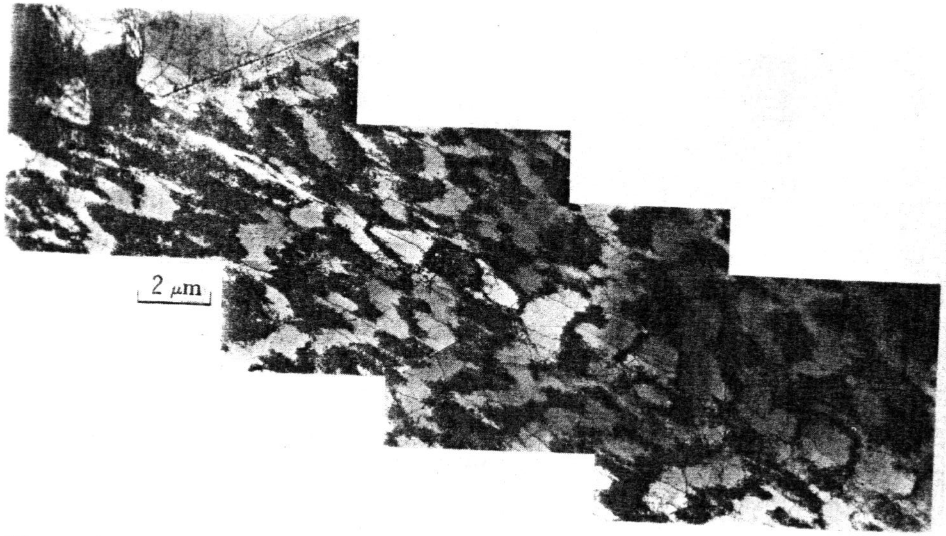

Figure 6 Dislocation configuration around the crack in the early stage of fatigue in copper (σ = 108 MPa, N_f = 1.0 × 10^6 cycles)

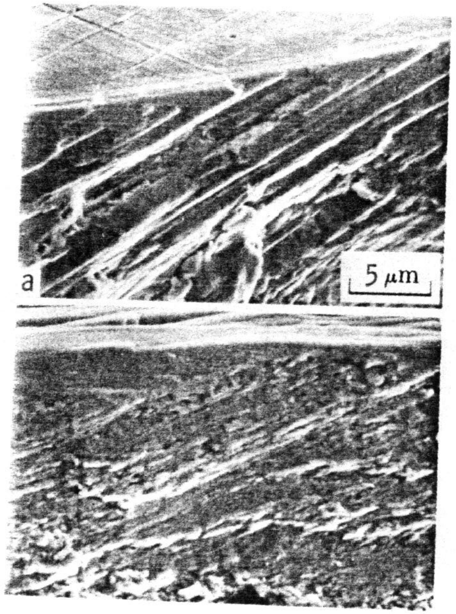

Figure 7 Facets on fracture surfaces of the cracks within the slip bands. (a) α-brass (σ = 118 MPa, N_f = 8.0 × 10^6 cycles)
(b) copper (σ = 88 MPa, N_f = 2.5 × 10^6 cycles)