A WEAK BEAM TRANSMISSION ELECTRON MICROSCOPY STUDY OF A FATIGUE CRACK TIP IN A 2024 ALUMINUM ALLOY

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ABSTRACT

Weak-beam dark field transmission electron microscopy has been used to study the dislocation structure at a fatigue crack tip in 2024 aluminum. The region immediately ahead of the crack tip is characterized by a dislocation cell structure. Changes in the localized stress level at the crack tip are monitored by the measurement of the distribution of bowed out dislocation segments near cell walls.

KEYWORDS

Aluminum; 2024-T351; crack propagation; crack tips; dislocation distribution; dislocation structure; fatigue damage; plastic zone; transmission electron microscopy; weak beam microscopy.

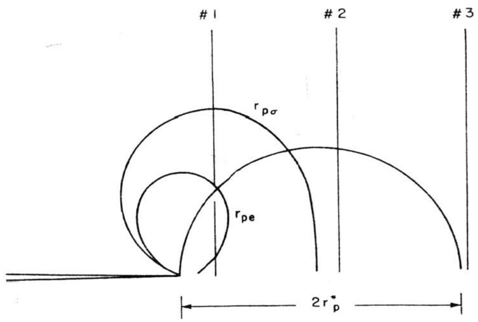
INTRODUCTION

Fatigue life prediction under variable amplitude loading is an essential part of modern aircraft design. It still remains a very imprecise art, however, and most models used for fatigue crack growth predictions are predominantly empirical. This is because there really is no basic understanding of fatigue crack growth and how the various loading variables can influence it. A particular fatigue crack growth model may give a good prediction for one loading spectrum, but changing the spectrum may result in large errors. Similarly the load interaction or retardation parameters that give good predictions for one material may not be suitable for another.

Some of these effects can be explained in terms of crack closure and consequent residual stresses around the crack tip; but some must be due to changes in the material itself in front of the crack. The purpose of this work therefore was to examine the material at the tip of a propagating fatigue crack at the dislocation level, to determine if there was some feature of the "structure" that might be used to characterize it. While dislocation structures at crack tips have been studied previously in pure metals and simple alloys, such studies on complex industrial alloys are limited (Grosskreutz and Shaw, 1967). In this paper we attempt to find some microstructural parameters that may be used to study crack tip plasticity in a 2024 aluminum alloy.

EXPERIMENTAL DETAILS

A compact tension fatigue crack growth rate specimen with width, W = 50 mm and thickness, B = 12.6 mm, was prepared from a 12.6 mm thick plate of 2024-T351 aluminum with the notch oriented in the long transverse direction (i.e. an L-T specimen). A pre-crack was initiated and grown under constant amplitude loading according to the ASTM standard for constant amplitude fatigue crack growth rate testing (E647) at a load ratio R of 0.1. Testing was stopped at a crack length of 20.7 mm when the stress intensity range was 17.3 MPa$\sqrt{m}$ which corresponds to a maximum stress intensity of 19.2 MPa$\sqrt{m}$. A 3 mm diameter core was then removed from the centre of the specimen (using electric discharge machining (EDM) techniques) such that it was parallel to and in the direction of crack propagation. This core, which then contained the centre portion of the crack, was cut into 0.25 mm thick slices for transmission electron microscopy (TEM). The cuts were made perpendicular to its length, again using EDM techniques, from a point near the end of the crack but on the cracked side. The distance between slices was 0.2 mm. The slices were then numbered from the first slice which did not contain the crack. Figure 1 shows the approximate position of the first three complete slices in relation to some simple plastic zone size estimates (Broek, 1982).



$$r_p^* = \frac{1}{2\pi}\left(\frac{K}{\sigma_{ys}}\right)^2$$

$$r_{pe}(\theta) = \frac{1}{4\pi}\left(\frac{K}{\sigma_{ys}}\right)^2 \left[\frac{3}{2}\sin^2\theta + (1-2\nu)^2(1+\cos\theta)\right] \qquad \text{(plane strain)}$$

$$r_{p\sigma}(\theta) = \frac{1}{4\pi}\left(\frac{K}{\sigma_{ys}}\right)^2 \left[1 + \frac{3}{2}\sin^2\theta + \cos\theta\right] \qquad \text{(plane stress)}$$

Fig. 1. Approximate positions of TEM slices in relation to the conventional plastic zone shapes.

Thin foils were prepared by conventional jet polishing techniques. Given the highly deformed state of the crack tip and hence high dislocation density, weak beam microscopy was employed along with conventional bright and dark field TEM techniques. While weak beam dark field microscopy has been applied to the examination of dense dislocation networks, relatively few studies have used this technique for the study of cyclically deformed structures (Rajan, Ramaswami, and Sastry, 1975; Antonopoulos, Brown, and Winter, 1976). In this paper we present the initial set of results in a program to apply this high resolution technique to the study of cyclic deformation at crack tips in commercial alloy systems.

The principal condition for obtaining weak-beam images from a reflection $\bar{g}$ is that the deviation parameter $|\omega| = |s_g \xi_g| > 5$, where s_g is the distance from the reciprocal lattice point $\bar{g}$ to the Ewald sphere and ξ_g is the extinction distance for the reflection $\bar{g}$. In the weak beam mode, if no reflection is strongly excited and $|s_g| > 2 \times 10^{-2}$ Å^{-1}, the image peak has a half-width of ≈15Å (Cockayne, 1970). In this study, $\{111\}$ type and $\{220\}$ type reflections were used to form weak beam images, with the $\bar{g}$-$3\bar{g}$ condition and s_{3g} slightly positive. Although the $\{111\}$ type reflections do not produce the optimum $|s_{3g}|$ value, they were used along with the $\{220\}$ reflections since they did provide good resolution.

EXPERIMENTAL RESULTS

The primary difference in structure between the area near the crack tip and further away from it was the presence of a dislocation cell structure. The formation of a cell structure in precipitation hardened alloys does not occur as readily as in pure metals. However, as noted by Stoltz and Pelloux (1977) the 2024-T4 alloy system contains precipitates shearable in fatigue deformation. This would suggest that dislocation rearrangement is in fact possible. Up to 1.0 mm from the crack tip a cell structure was observed and further away it was not apparent. Hence the existence of a cell structure was attributable to the fatigue crack. Figure 2 shows a comparison of bright field and weak beam dark field photographs at 1.0 mm from the crack showing the improvement in resolution with the latter technique. As can be seen in Figure 3, at 1.2 mm from the crack, even in the weak beam mode, a cell structure was not evident.

DISCUSSION

As has been shown above, the region immediately ahead of a fatigue crack tip in this alloy is characterized by a dislocation cell structure. The primary reason for dislocations to rearrange themselves into cell walls is to reduce the total elastic interaction energy. The presence of a cell structure therefore allows us to postulate on a mechanistic explanation of crack tip plasticity. The theoretical and experimental studies (especially on single crystals) of dislocation cell formation are extensive (Kuhlmann-Wilsdorf, 1977; Wilsdorf, 1983; Quesnel and Tsou, 1983). Such studies have demonstrated that an inverse relationship exists between the saturation flow stress obtained during fully reversed plastic strain controlled fatigue deformation and the average dislocation cell diameter. Empirical relationships between cell size and the localized flow stress may then be established. In a previous study for example Rajan, Ramaswami, and Sastry, (1959) used the cell diameter as a means of defining microstructurally the stress level of a fatigue crack tip.

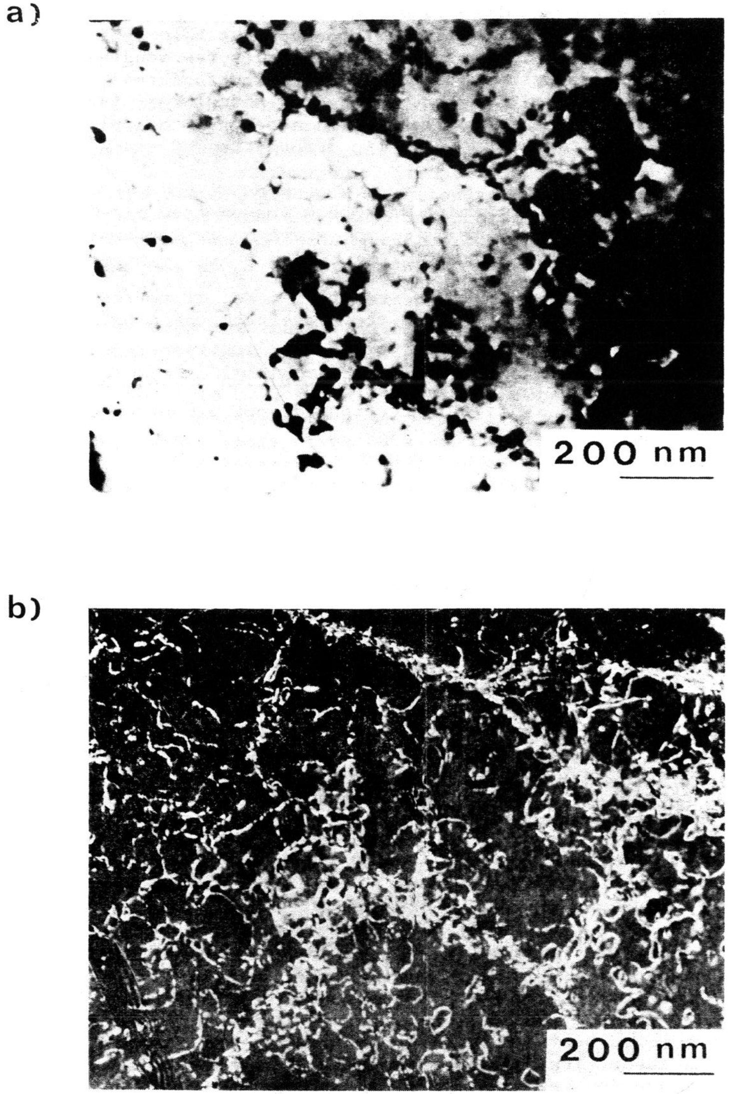


Fig. 2. Cell structure 1 mm from crack tip:
a) Bright-field
b) Weak-beam dark field
$\bar{g} = (111)$

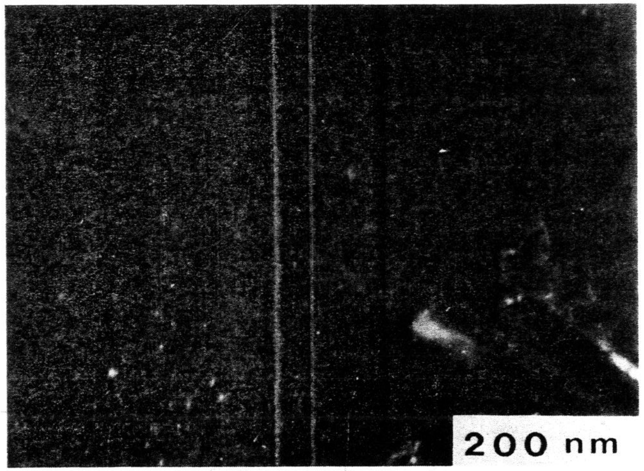


Fig. 3. Weak-beam photograph 1.2 mm from crack tip $\bar{g} = (220)$.

It should be remembered that in this study we are observing a dislocation cell structure in a precipitation hardened alloy and not in a pure metal. The dislocation rearrangement into cells does not occur as readily and the presence of a cell structure in itself indicates the very high stress immediately ahead of the crack tip in this material.

Although it has been established that higher stresses result in smaller cells, a fundamental question as noted by Prinz and Argon (1980) is the mechanism by which smaller cells are produced. The dislocation walls may be stabilized by reactions with secondary dislocations which produce nodes (Figure 4) whereas such nodes are not present in the cell interior. This suggests that the walls do not move during deformation. However, the refining of the cell size may be accomplished by the generation of dislocations originating from cell walls and then being trapped by their elastic interactions inside the cells. A major source for dislocation generation is the bowing out of dislocation segments from stable anchoring points (e.g. dipoles and multipoles) in the cell walls.

Therefore, in this study, we decided to examine more closely the region around cell walls for the presence of bowed out dislocation segments. Using weak beam electron microscopy allowed us to look for such details in a heavily deformed structure. In Figure 5 such bowed out dislocation segments are indicated. The presence of dislocation segments bowing out at cell walls, suggests that the stress level can be locally quantified by the Frank-Read stress; $\tau_{FB} \approx \frac{\mu b}{\ell}$, where μ is the shear modulus, b, the magnitude of the Burger's vector and ℓ is the spacing between anchoring points of the bowed dislocation segments. Figure 6 shows the frequency distribution of dislocation segments bowed out between anchoring points of spacing ℓ, at different distances from the crack tip. Also plotted are the corresponding values of the Frank-Read stress in

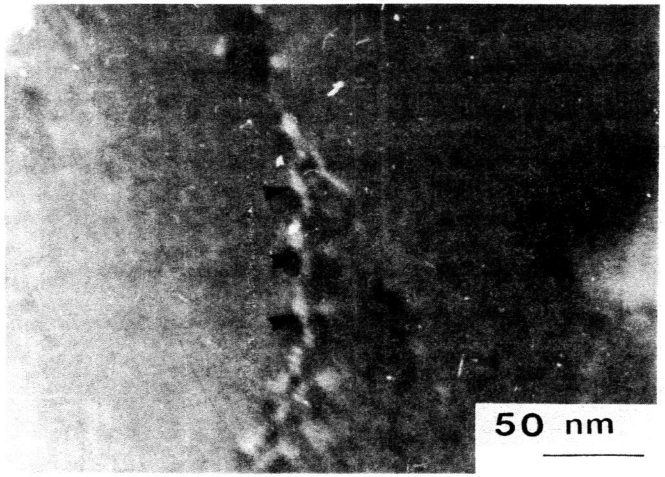


Fig. 4. Nodes at dislocation cell walls $\bar{g}$ = (220).

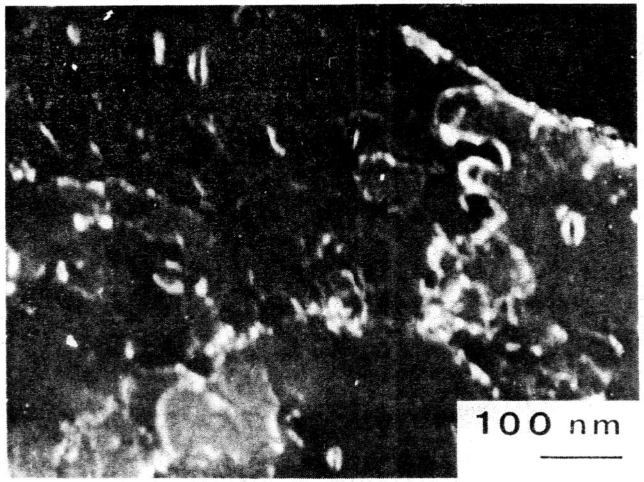


Fig. 5. Bowed out dislocation segments at cell walls $\bar{g}$ = (220).

terms of the shear stress for this polycrystalline alloy. It is clear from Fig. 6 that there is a definite shift to smaller dislocation segments closer to the crack tip. This indicates a shift to higher stresses as we approach the crack tip. The numerical values of the stresses are approximate since other stresses must be added to the Frank-Read stress to get a more quantitative assessment of the stress level (e.g. effect of shearable precipitates). Although the measurement of dislocation segment distributions has been described in the literature (Mughrabi, 1975; Grosskreutz and Mughrabi, 1975), this is the first time to our knowledge that these concepts have been applied to the study of fatigue crack tips (especially in complex alloy systems).

These bowed out segments may act as Frank-Read dislocation sources and thus increase the mobile dislocation density within the cell interior. Since the majority of dislocations in the walls are not available to accommodate the strain during cyclic deformation, it is the dislocations in the interior which are truly mobile and hence it is their density which is important for plasticity considerations. Thus we have, by using a strain sensitive microscopy technique, provided direct evidence of a possible micromechanism controlling crack tip plasticity in this aluminum alloy; namely, the expansion of dislocation loops from cell walls giving rise to a continuous increase in dislocation density.

The results of the present study therefore provide a microstructural basis for studying the response of 2024 aluminum to low cycle fatigue tests and crack propagation tests.

CONCLUSIONS

1. A dislocation cell structure is formed immediately ahead of a fatigue crack tip in 2024-T351 aluminum.
2. One of the mechanisms controlling work hardening (and hence plasticity) at the crack tip is the expansion of dislocation loops at cell walls.
3. Weak beam transmission electron microscopy is an invaluable tool for studying cyclically deformed structures at crack tips.

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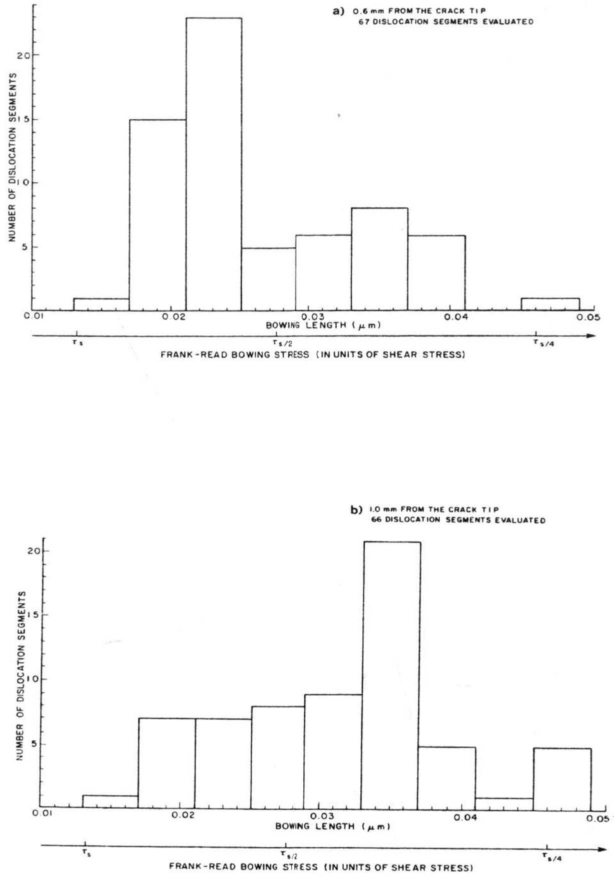

Fig. 6 Frequency distribution of bowed dislocation segments
a) 0.6 mm from crack tip (slice #2)
b) 1.0 mm from crack tip (slice #3)

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