THE METALLOGRAPHY OF FATIGUE IN THE HIGH STRENGTH ALUMINIUM ALLOY 7010

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ABSTRACT

A study has been made of the micromechanisms of fatigue crack propagation in the high strength aluminium alloy 7010. Use has been made of both scanning and transmission electron microscopy to investigate in great detail the surface topography and underlying substructure produced during stage II crack propagation in both moist and dry environments at a number of cyclic frequencies. In a moist environment a crystallographic crack plane of {110} was found with clearly marked fatigue striations which were of two extreme profiles. The thin foil studies revealed regularly spaced deformation bands separated by regions of low dislocation density; these bands were related directly to the surface striations. In a dry environment (Ar containing <10 ppm water vapour) no regular surface striations were found and the underlying substructure was of uniformly high dislocation density. The fracture plane was non-crystallographic. The results are discussed in relation to existing mechanisms of crack propagation and one is proposed for moist environments based on an initially brittle crack extension followed by plastic blunting at maximum load. The results are most consistent with a hydrogen embrittlement mechanism relying on dislocation transport of hydrogen. The localised constraints on crack plane and crack tip direction are thought to lead to different slip system orientations during blunting, giving rise to the two striation topographies.

KEYWORDS

Fatigue; aluminium alloys; electron microscopy; micromechanisms; crack propagation; hydrogen embrittlement.

INTRODUCTION

Many mechanisms have been proposed to describe fatigue crack propagation in metals on a microscopic scale and have recently been reviewed (Mughrabi 1980). These mechanisms differ, however,particularly in the field of stage II crack propagation and the influence of the environment thereon. It is well established that crack growth rates depend greatly on the environment, the magnitude of the effect depending on the material (Bradshaw and Wheeler, 1969). Water vapour is found to have the most pronounced effect on aluminium alloys although a dry oxygen environment leads to some increase in crack propagation rate over vacuum.

In the present work electron optical techniques have been used to study in detail surface topography and dislocation substructures, in order to obtain further information about the mechanism of stage II propagation in a high strength commercial aluminium alloy and the environmental effects. Several workers have attempted such studies, (Grosskreutz and Shaw 1966, Bowles and Broek 1972, Wanhill 1975, Johannsson 1977) but partially conflicting results were obtained and different conclusions reached. With the development of the TEMSCAN type of electron microscope, with its provision of high resolution scanning electron microscopy in combination with transmission electron microscopy, the scope of this type of study has been much extended since it is now possible to directly relate surface topography to underlying microstructure. It is also possible using this type of instrument to observe surface detail at much higher resolution than previously possible in a conventional SEM.

EXPERIMENTAL DETAILS

The alloy used for this study was the Al-Zn-Mg-Cu alloy 7010 in the T.76 condition. Fatigue testing was carried out in laboratory air (RH 50%) at frequencies of 0.1, 10 and 100 Hz with a stress ratio of 0.1. A resonance fatigue machine was used for the 100 Hz testing and servo-hydraulic machines for testing at lower frequencies. A simple programme of varying maximum and constant minimum load was also applied using a servo-hydraulic machine at 10 Hz. A less rigorous study was also made of crack propagation in a dry environment (Argon containing <10 ppm H_2O); a test frequency of 100 Hz and a stress ratio of 0.1 was used. Test specimens were then sectioned for fractographic observation, and two types of thin foil sample were prepared (Fig. 1), type I containing the fracture surface as one face of the foil and type II the crack tip itself.

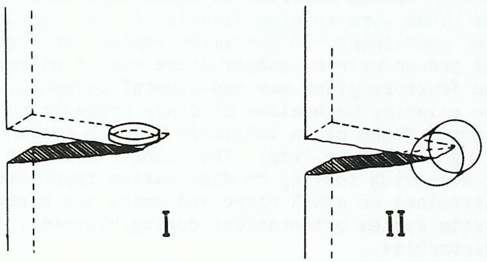

Fig. 1. Types of thin foil specimen.

Both were prepared in 3 mm disc form by spark machining followed by electropolishing using a jetting technique. Type I specimens were lacquered on the fracture surface and electropolished from the reverse face only in order to preserve the fracture surface. The lacquer was then dissolved off in acetone. Type I foils were observed in a JEOL 120 CX TEMSCAN electron microscope operating at 100 kV in both SEM and TEM modes. Bulk observations were also carried out in this machine. Type II foils were examined in an AEI EM7 high voltage electron microscope (HVEM) operating at 500 kV.

RESULTS AND DISCUSSION

Surface Topography

The use of the TEMSCAN in the SEM mode has permitted surface topography to be investigated in great detail. Observation of specimens tested in laboratory air has allowed some new information to be gained concerning the nature of fatigue

striations. Two types of striation have been observed within any single specimen tested at any of the three frequencies used. The first, type (A), were of saw-tooth profile with slip steps on the leading face, Fig. 2, and the second (B) consisted of deep slots traversing an essentially flat fracture surface, Fig. 3, with some evidence of slip steps directly ahead of the slots.

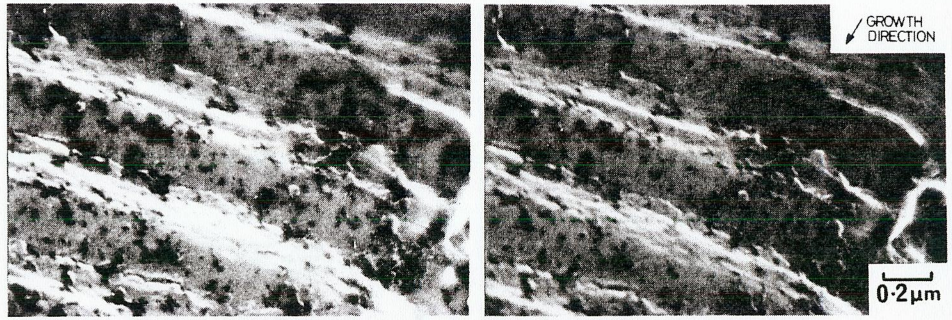

Fig. 2. Type A striations. SEM stereopair.

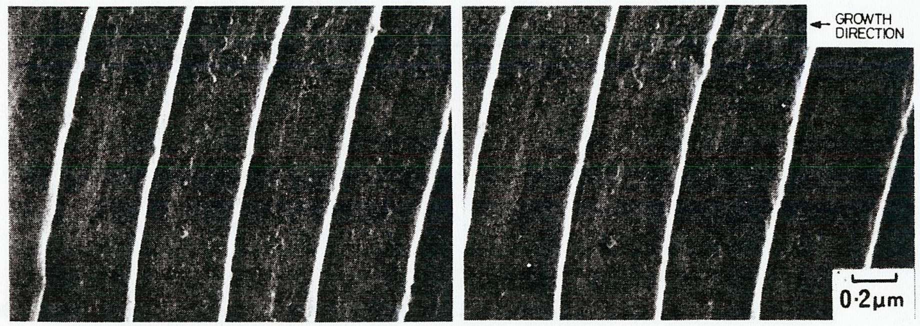

Fig. 3. Type B striations. SEM stereopair.

Extensive tilting experiments were routinely carried out in order to fully characterise the surface topography and clearly establish the co-existence of type A and B striations. Earlier attempts to rationalise the different striation appearances as merely a contrast effect were clearly unnecessary (Hertzberg 1966). Both types of striation have been observed previously (Stubbington 1963) but type A were termed 'ductile' and type B 'brittle', since the latter were more commonly observed on corrosion fatigue fractures. Here both types have been observed over a wide range of ΔK, (5 → 25 $MNm^{-3/2}$) at all test frequencies in the laboratory air environment, although type A may be more common at low ΔK where many apparently striation-free areas were found to be covered with type A striations when tilted to improve local contrast. Since both types of striation occur in adjacent areas the most likely factor determining the type is the local crystallographic orientation. Further evidence for this is provided by the fact that intermediate striation types have been found, Fig. 4, suggesting that the actual mode of propagation is similar in the two cases (further discussion of this appears in the following section).

The use of the simple programme of varying maximum stress and constant minimum stress, Fig. 5, has proven that crack propagation occurs cycle by cycle since the two lower peak stresses lead to smaller crack extensions.

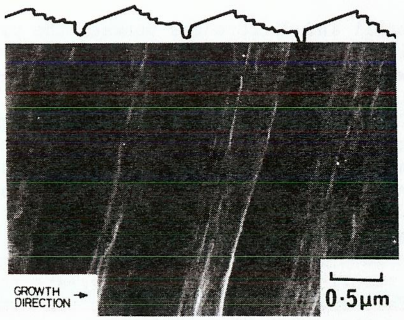

Fig. 4. Intermediate striation type.

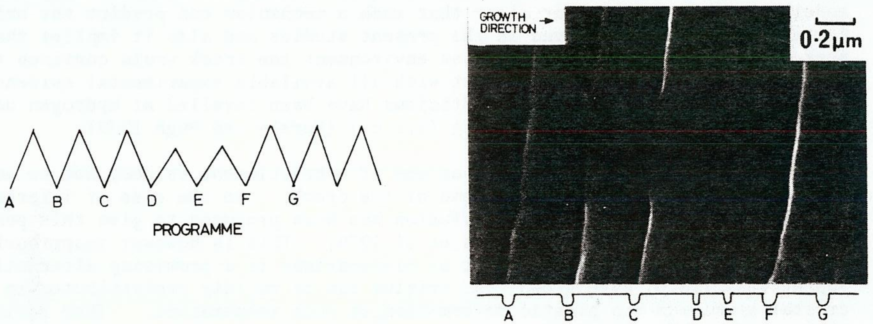

Fig. 5. Striation topography produced by load programme.

This evidence invalidates any theories to the contrary such as that proposed by Tomkins and Biggs (1969) which was applied to high strength aluminium alloys by Wanhill (1975). While a wealth of information exists to suggest that the slot in type B striation forms during crack closure (Laird and De la Veaux 1977), it is interesting to note that the actual width of the slot appears to be related to the magnitude of the following tensile cycle.

In the dry argon test environment the fracture surface appearance was found to be quite different. Generally a rumpled surface was found similar to that observed by Wanhill (1975) and Bowles (1978). Only at high stress intensities were some markings found which could be related to cycle by cycle growth, Fig. 6. These were similar in profile to type A striations but were far less regular and were rarely observed. The actual mode of crack propagation is obviously therefore quite different in this case and is related to the environment.

Microstructural Observations

In order to actually determine the relationship between surface topography and underlying microstructural features observation in both SEM and TEM of type I foils has been carried out. Typical results for both type A and B striations are shown in Figs.7 and 8. The crack growth direction is arrowed in each case. In both cases the most obvious substructural features are the heavy deformation bands of high dislocation density which exhibit a 1:1 correlation with the striations, lying beneath the leading edge of type A and ahead of the slot in

type B as shown in Fig. 9.

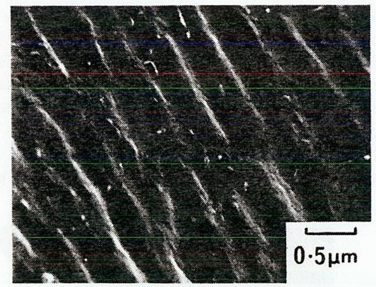

Fig. 6. Surface markings at high ΔK, dry argon test environment.

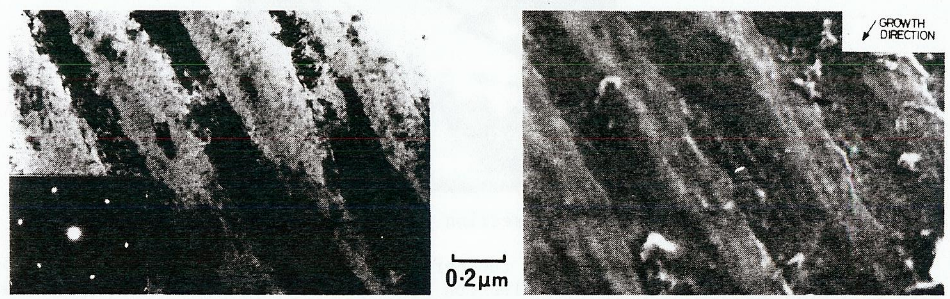

Fig. 7. Type A striation surface topography and underlying substructure TEM/SEM pair in identical areas. Type I foil.

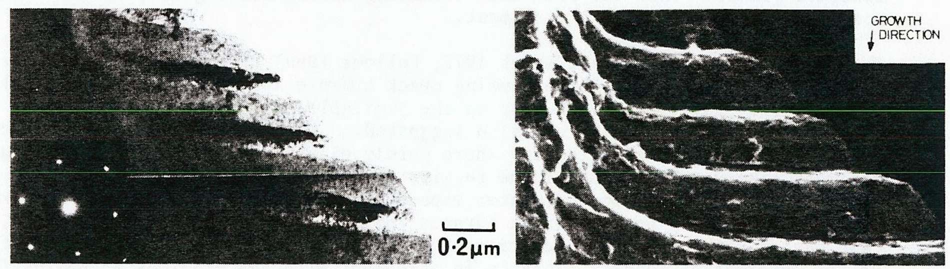

Fig. 8. Type B striation surface topography and corresponding substructure.

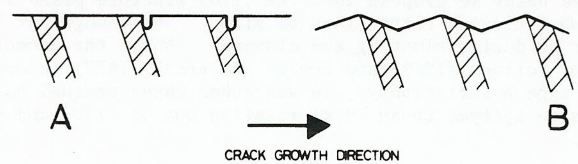

Fig. 9. Schematic representation of striation/deformation band relationship.

Evidence of subsurface deformation is seen on the surface in the form of slip lines, Fig. 2, Fig. 3, which correspond to the position of the deformation bands. Similar bands have been observed by Bowles and Broek (1972) who attributed them to deformation occurring during crack closure. It is felt that while this may be partly true, their formation must also be influenced by deformation processes occurring during blunting since it has been proven (Bowles 1978) that the crack is blunt at maximum load. Observation of type II foils in the HVEM operating at 500 kV has demonstrated the extent of the bands, Fig. 10, which run apparently only to a short distance below the fracture surface, demonstrating the localised nature of this heavy deformation.

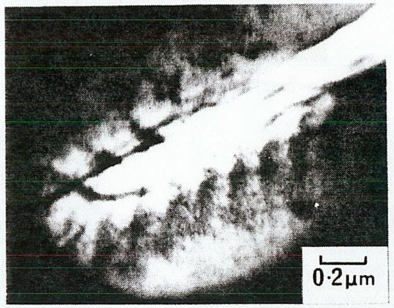

Fig. 10. Transverse section through crack walls. HVEM.

The intermediate regions between the deformation bands exhibit a low dislocation density and relatively perfect electron diffraction patterns. The latter also indicate that the fracture plane generally lies close to {110}. In complete contrast to this type I foils prepared from specimens tested in dry argon show a relatively uniform heavily dislocated substructure and absence of any preferred fracture plane. Thus the processes occurring during crack growth are considerably modified by the change in environment.

Previous studies (Bowles and Broek 1972, Pelloux 1969) have suggested that the deformation bands may be formed during crack closure and that a {110} crack plane could result from symmetrical slip on the available $\{111\}\langle 110\rangle$ slip systems. A $\langle 1\bar{1}0\rangle$ crack tip direction was also suggested. However, no such crystallography was obtained in dry argon testing where purely plastic crack advance must occur. The most likely explanation of the results obtained in moist environments is therefore that the presence of water vapour modifies the crack growth mechanism to one of initially brittle crack advance followed by plastic blunting, which, in combination with plasticity occurring during crack closure leads to the formation of the deformation bands. This is in agreement with the findings of Bowles (1978). However, contrary to earlier suggestions the crack front was not confined to $\langle 1\bar{1}0\rangle$. On the contrary it lay along a variety of directions including $\langle 1\bar{1}1\rangle$ as in Figs. 7 and 8 (this may be simply fortuitous since two such directions lie in a {110}). As discussed below we propose that the {110} fracture plane is determined by hydrogen embrittlement rather than by slip. The topography is determined by the slip occurring during blunting and closure. Where the geometry conforms to that discussed by Pelloux (1969) and Bowles and Broek (1972) incomplete slip reversal results in type B striations. In all other cases unequal amounts of slip on the operating slip systems cause local rotation out of {110} and give rise to type A striations.

The Embrittlement Mechanism

In the following discussion the mechanism proposed for crack growth will be

discussed in more detail and differences between the proposed model and that of Bowles (1978) explained. The basic concept is that hydrogen, produced as a result of the reaction between water and aluminium can give rise to embrittlement at the crack tip. There is substantial evidence for this in the literature, for example Bursle and Pugh (1977), and consequent transgranular embrittlement. In the case of hydrogen introduced by cathodic charging it has been shown (Bursle and Pugh 1977) that transgranular failure generally occurs on {110} planes, a fact which is in agreement with the present work. It is proposed that hydrogen embrittlement leads to a {110} fracture plane rather than any plastic process. However, Bowles (1978) has shown that, using an estimate of the bulk diffusivity of hydrogen in aluminium of $\sim 2 \times 10^{-9}$ mm^2/sec, the rate of hydrogen migration is insufficiently fast to explain his results. He therefore suggests that embrittlement occurs by the lowering of surface energy at the crack tip via hydrogen adsorption at the tip itself. Wei et al (1980) have shown that at low water vapour pressures the rates of supply and surface reaction are rate limiting, tending to support such a model. However, it is not clear that such a mechanism can predict the brittle/ductile transitions observed in the present studies and also it implies that if peak load were maintained in a moist environment the crack would continue to grow indefinitely. This is in conflict with all available experimental evidence (Barsom (1971). Very similar criticisms have been levelled at hydrogen adsorption models of stress corrosion cracking (s.c.c.) (Bursle and Pugh 1979).

These discrepancies can only be overcome if embrittlement is due, not to adsorption but to penetration of hydrogen ahead of the crack. In the case of intergranular s.c.c. grain boundary hydrogen diffusion has been proposed to give this penetration. (Montgrain and Swann 1974, Scamans et al 1976) This is however inappropriate in this case. Hydrogen transport by dislocations is a promising alternative. Foster etal (1970) have shown that tritium can be rapidly redistributed in single crystal aluminium via plastic deformation at room temperature. They indicate that an effective diffusivity of up to 10^5 greater than that due to bulk diffusion may result.

Significant, although less dramatic, enhancement of diffusion has also been predicted by Tien (1975) who calculated hydrogen penetration would be increased in depth by 240 x and 3 x under 1 Hz and 60 Hz cyclic loading respectively at room temperature. It is clear that brittle zones of the widths observed in the present work could be produced if a dislocation transport mechanism were involved. The mechanism is attractive since it can predict the observed brittle/ductile transitions; as the crack propagates beyond the hydrogen embrittled zone ductile blunting and closure will occur during which the moving dislocations introduce more hydrogen to cause the next step of brittle advance on the following half cycle. (Dislocations move throughout the plastic zone which is much larger than the extent of the deformation bands which contain those dislocations which were trapped at the crack tip.) The crack cannot grow under constant load since dislocations are not moving and hydrogen is thus not injected into the metal. Similarly hydrogen embrittlement will not occur if the supply of water vapour and hence hydrogen production rate is insufficient (Wei et al 1980) or if the frequency is so high that hydrogen transport via dislocations cannot occur. In the model of Tien (1975) the actual frequency will depend on the bulk diffusivity. If this is similar to that indicated by Eichennauer and Pebbler (1957) this frequency will be of the order of 10^3 Hz which is beyond the range of the present work.

CONCLUSIONS

1. The surface topography and underlying substructure produced during stage II crack propagation are clearly influenced by test environment.

2. In moist environments, hydrogen, produced by the water vapour/aluminium reaction is thought to lead to a brittle form of initial crack extension on {110} followed by plastic blunting and closure resulting in the deformation band structure and surface striations.

3. Two extreme types of striation have been observed,one of sawtooth profile and another of a slot type. It is believed that the different types are formed by the same process and their geometry is dictated by the orientation of {110} fracture plane and crack front orientation to the stress axis.

4. The results are best explained by a process based upon dislocation transport of hydrogen to a short distance ahead of the crack on a cycle by cycle basis.

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