SLIP PLANE FACETS IN FATIGUED ALUMINIUM ALLOYS

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ABSTRACT

Stage I fatigue crack growth in age hardened f.c.c. alloys occurs predominantly by cracking along well defined persistent slip bands, PSB, producing characteristic (111) slip plane facets on the resulting fracture surfaces. In aluminium alloys such facets display a loss of solute atoms from the surfaces indicating solute migration from the original PSB. In this paper the implications of solute migration are considered further and some evidence that the process is a necessary prelude to facet formation is presented.

KEYWORDS

Stage I fatigue, aluminium alloys, slip plane facets, solute migration, matrix : PSB hardness ratio.

INTRODUCTION

A renewed interest has been shown recently in Stage I fatique crack growth. A characteristic feature of such fatigue behaviour, particularly in age hardened f.c.c. materials, is that cracking occurs predominantly along persistent slip bands (PSB) resulting in well defined crystallographic facets on the fracture surfaces. (Gell and Leverant, 1968: Stubbington and Forsyth, 1966). Faceting of this type is enhanced by low applied loads, a large grain size and a microstructure with shearable precipitates (Nageswararao, Gerold and Kralik, 1975). Under these conditions quite complex faceting has been observed in aluminium alloys, with more than one plane contributing to the fracture surface even within single grains. An example is shown in Fig. 1. The facets are usually found to be of (111) orientation although other orientations may occur (for example Brett, Cantor and Doherty, 1977).

Electron microprobe analysis at glancing angles of incidence of facets from fatigued specimens of Al4%Cu and Al4.5%Zn1. 5%Mg alloys has revealed a general, though varying, loss of solute, either copper or zinc, from the surfaces and, by implication, from the original PSB (Brett and Doherty, 1978). Two facets from Al4%Cu specimens were subsequently investigated by Auger electron spectroscopy to provide a more detailed picture of the solute movement (Lea, Brett and Doherty, 1979). The solute profile with depth is reproduced in Fig.2.

Fig. 1 Faceted fatigue crack growth in an Al4%Cu specimen
Magnification : 120

The second facet displayed a similar profile but it should be pointed out that both the facets were chosen because they showed the largest depletions under microprobe analysis. It is possible therefore that Fig.2 represents a limiting case with facets in general exhibiting a range of depletions. Alternatively, the variations in solute loss found by microprobe analysis could be due to equivalent degrees of depletion in PSB of differing width. A combination of both effects is of course also possible.

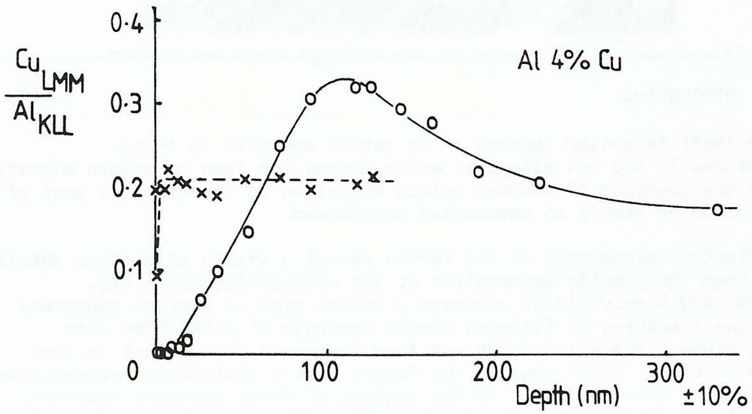

Fig. 2 Auger peak height ratio as a function of depth of metal removed by argon ion bombardment.
O (111) slip plane facet
X polished control
(Lea, Brett and Doherty, 1979).

Various theories of microstructural damage within PSB have been proposed and these are discussed elsewhere (e.g. Brett and Doherty, 1978; Calabrese and Laird, 1974). In general such localised damage requires the shearing of precipitates within the PSB and if this occurs movement of solute into the matrix may then follow by a process of Ostwald ripening after the model of Sargent and Purdy (1974), or by the change of solubility caused by precipitate damage such as disordering (Calabrese and Laird, 1974, Stoltz and Pineau, 1978). In order to investigate the effect of solute migration on subsequent faceting the two aluminium alloys were aged to contain microstructures with a range of shearability and, hopefully, a variation in the driving force for solute movement.

EXPERIMENTAL DETAILS

Specimens were prepared in the form of pin-loaded, 60mm x 24mm x 3.2mm coupons single edge notched to a depth of 2mm. They were broken in a tension-zero mode at a frequency of 30Hz. The Al4%Cu specimens were solution treated at 540°C and quenched in iced brine; the Al4.5%Zn1.5%Mg specimens were solution treated at 465°C and water quenched.

Previous work by the authors (1977) had shown that in the case of AlCu alloys as quenched specimens broken under the present loading conditions produced predominantly faceted fracture surfaces, while those peak-aged at 190°C showed mainly intergranular failures. As quenched specimens were therefore progressively aged at 190°C to explore the fractography of intermediate microstructures. It should be noted that in this particular alloy the quenched condition is not a solid solution. The Vickers Pyramid Number (VPN) is approximately 90 while that of a reverted solid solution is 66, indicating that some precipitation occurs during the quench.

A reversion heat treatment of 10 minutes at 200°C produces a hardness of 66 which remains constant indefinitely, presumably because the equilibrium vacancy concentration obtained at this temperature is not high enough to allow natural ageing in this alloy. Reverted AlCu specimens were used to represent the stable solid solution state. The precipitates in the peak aged state are non shearable while the as quenched GP zones are sheared by dislocations.

In contrast the AlZnMg alloy is easily quenched to the solid solution (VPN = 55) and specimens of this alloy were given various ageing times at 90°C to investigate the effect of the gradual development of a shearable GP zone microstructure. A minimum ageing time of 3 hours was chosen in order to avoid complications caused by natural ageing between the artificial ageing treatment and the test. In fact, because of the marked natural ageing of this alloy it was not possible to test a specimen containing a true solid solution.

The test details are shown in tables 1 and 2.

RESULTS

It was found that under the loading conditions employed faceting died out rapidly with ageing at 190°C for the AlCu alloy. Figure 3 shows the approximate percentages of the fracture surfaces occupied by facets and by intergranular cracking. The two modes of failure appear to be mutually exclusive. The areas of fracture surface unaccounted for were in all cases transgranular but unfaceted. The transition is therefore:

transgranular faceted → transgranular unfaceted → intergranular

TABLE 1

Specimen Number	Ageing time at 190°C (Hours)	Maximum load (MPa)	Cycles to Failure
AlCu7	1	64.6	8.31×10^5
AlCu8	1	57.8	8.01×10^5
AlCu9	1	57.8	7.94×10^5
AlCu1	2	54.1	7.62×10^5
AlCu2	2	53.6	9.89×10^5
AlCu3	4	65.1	1.93×10^5
AlCu4	4	57.8	3.18×10^6
AlCu5	8	62.8	1.5×10^6 (unbroken)
AlCu6	8	57.8	1.18×10^6

TABLE 2

Specimen Number	Ageing time at 90°C (Hours)	Maximum load (MPa)	Cycles to Failure
AlZnMg5	3	56.8	1.04×10^5
AlZnMg7	7	56.2	1.53×10^5
AlZnMg13	7	58.2	1.49×10^5
AlZnMg9	18	58.2	1.28×10^5
AlZnMg10	18	60.0	1.38×10^5
AlZnMg14	24	59.1	5.06×10^5

as the microstructure develops from the highly shearable GP zones to the non shearable peak-aged state containing θ" and θ'. The percentage values used for these two extreme cases are typical values taken from previous specimens aged to contain these microstructures.

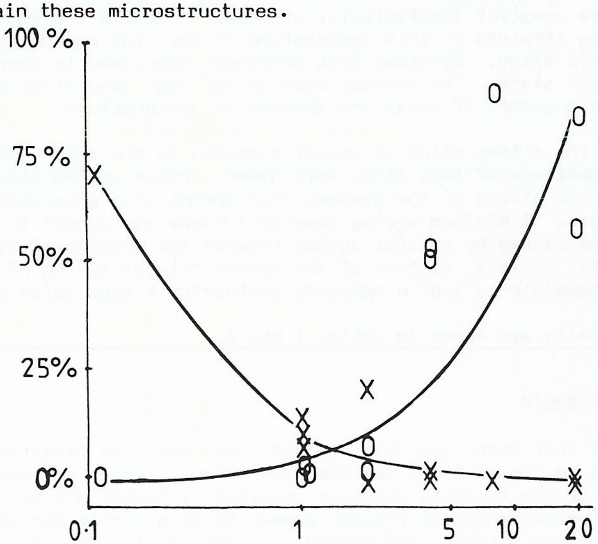

Fig.3 Percentages of fracture surface displaying faceting and intergranular failure for Al4%Cu specimens as a function of ageing time in hours at 190°C. X = faceted failure 0 = intergranular failure

Reverted specimens showed no faceting whatsoever. The failure points were found to fit reasonably well on the S-N curve for as quenched specimens indicating that the applied loads were not abnormally high in these cases.

All the AlZnMg specimens except AlZnMg5 displayed extensive faceting. AlZnMg5 had a generally transgranular fracture surface but the extremely shiny facets found in the other specimens were absent except in one localised area. A single grain in the centre of the fracture surface several millimetres from the notch had failed on conjugate crystallographic planes to form structures once referred to by Stubbington and Forsyth as 'Lamellae', Fig. 4. The significance of such structures will be discussed in the next section.

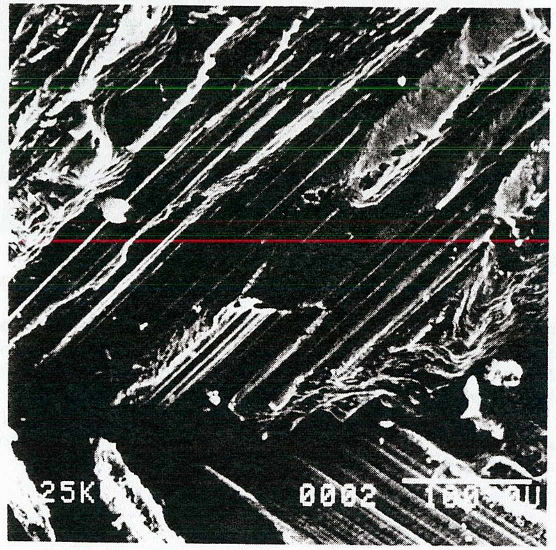

Fig. 4 'Lamellae' at the centre of the AlZnMg5 fracture surface.

DISCUSSION

Slip plane facet formation appears to be rather specific to those microstructures in the two alloys in which damage can lead to solute migration. This poses the question of whether solute migration is an essential part of facet formation or simply an associated phenomenon.

Scanning electron microscopy of the facets reveal a wealth of surface detail resulting from the plastic deformation at the propagating crack tip. Nageswararao and Gerold (1976) observed a steady rise in surface roughness of slip plane fractures in fatigued single crystals of Al5%Zn1%Mg with increasing crack propagation rates and then discussed this effect in some detail. On a visual level however the facets have a distinctive mirror-like appearance which may persist up to the regions of final unstable fracture. This indicates a considerable degree of confinement of crack tip processes to the PSB and this can only occur where the strength or hardness difference between the PSB and matrix material is sufficiently high. Clearly, solute migration will have a marked effect on this difference.

Some estimate of the strength difference can be made by considering those grains which fail on more than one set of crystallographic planes. The simplest case involves cracking on two sets of (111) planes to form lamellae and such structures were quite common on the fracture surfaces studied here. More complex structures incorporating three and even all four sets of (111) planes were also seen, although more rarely. The structures require a surface area of fracture greater than that generated by the minimum crack path through the faceted regions. Economies of fracture surface could be achieved if the fatigue crack took a non-crystallographic path through the matrix. Simple geometric considerations show that in the case of symmetric lamellae 76% extra surface area is required, while for a regular tetrahedral structure of three (111) planes the crack surface becomes three times the minimum necessary. The excess surface areas would involve a corresponding increase in the total plastic work of fracture unless the strength of the PSB material is proportionally lower than that of the matrix.

It is convenient in the present case to use hardness values rather than strength values. This is not unreasonable since the alloys under consideration are microstructurally very similar and their yielding behaviour, on which the correlation between strength and hardness depends, is also similar.

The ratio of the hardness of the matrix to that of the PSB can be denoted by P and this parameter can be calculated provided certain assumptions are made:

(i) The matrix hardness is taken to be the bulk hardness of specimens given the ageing treatment specified. Contributions from excess solute adjacent to the PSB are ignored.

(ii) Hardness differences created by dislocation arrangements alone are ignored It is assumed that such differences will be small compared to those generated by solute differences.

(iii) In the fully depleted conditions the PSB hardness is assumed to be that of annealed pure aluminium.

Table 3 shows P calculated for several microstructures which exhibit faceting for the limiting cases of complete depletion and no depletion (i.e. the PSB microstructure is destroyed but the solute remains in the band).

TABLE 3 VPN Values

Specimen	Full Solute Migration			PSB retains solid solution strengthening		
	Matrix	PSB	P	Matrix	PSB	P
AlCu as quenched	90	20	4.5	90	66	1.36
AlCu 18 hrs at 130°C	110	20	5.5	110	66	1.67
AlZnMg14	110	20	5.5	110	55	2
AlZnMg5	74	20	3.7	74	55	1.35

It can be seen that the hardness differences in the case of complete depletion are an order of magnitude greater than those where the solute is assumed to remain in the PSB. In terms of hardness ratio, if the minimum values of P can be approximately equated with surface areas of fracture then the minimum P for lamellae formation is about 1.76 and that for a tetrahedral structure about 3. On this admittedly crude basis only AlZnMg14 can form lamellae and none of the specimens can form tetrahedra unless solute migration takes place.

In fact if solute does not migrate the matrix hardness of the AlCu alloy would have to exceed 116 to obtain a P value greater than 1.76. Such a bulk hardness requires a heat treatment of more than 5 hours at 190°C which has been shown to take the alloy out of the faceting regime under the applied loading conditions. For the AlZnMg alloy an experimental threshold seems to have been found in Specimen 5. It is not clear however whether the threshold has been reached solely by underageing or by some other means. The specimen broke in the smallest number of cycles and it is possible that depletion was limited by the amount of time available for migration.

Specimens should in general exhibit a range of PSB development as the time for migration between the first effects of plastic deformation and the arrival of the fatigue crack varies from grain to grain across the specimen width. In specimen AlZnMg5 a single grain appeared to have PSB developed to the degree necessary for facet formation. Since lamellae were found in this grain while no clear facets were found in any other grains the PSB development here must have been relatively well advanced.

It has been suggested by Wilhelm, Nageswararao and Meyer (1979) that a period of time is required for PSB to become suitable crack paths and it is possible that this initial period is associated with solute migration. It is significant that in several AlCu specimens as well as AlZnMg5 the first areas of crack surface adjacent to the notch are transgranular but not faceted. An example is shown in Fig. 5. Crack initiation enhanced by the notch may result in the fatigue crack travelling through the first few grains of the specimen before time for sufficient PSB depletion has transpired. The transition from transgranular unfaceted to transgranular faceted cracking, which generally occurs at grain boundaries, will depend on the applied load. It may not occur at all where the load is so high that crack propagation is too rapid to allow the necessary development time in any of the grains crossed.

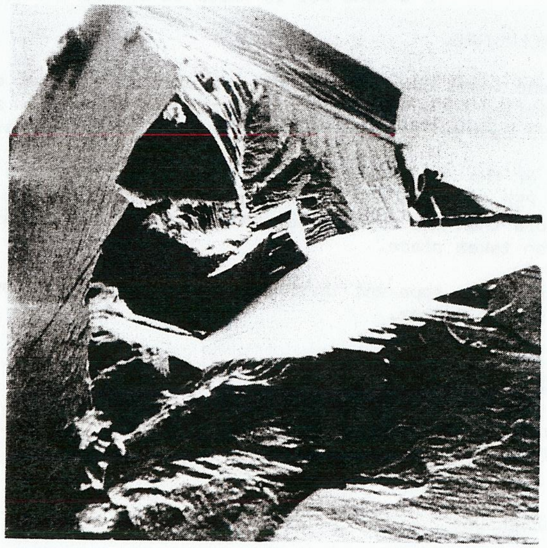

Fig. 5 Fracture surface of a large grain Al4%Cu specimen. The grains closest to the starter notch, shown at the top of the micrograph, have failed in a transgranular but unfaceted mode.

An indication of the possible existence of a threshold time for faceting is shown in Fig. 6 for two varieties of AlCu specimens. No facets at all were observed on the fracture surfaces of specimens broken in under 5 x 10^4 cycles. The threshold is relatively independent of load.

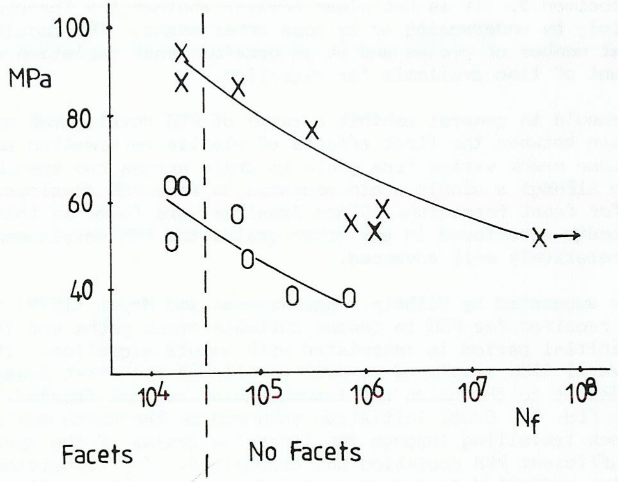

Fig. 6 Maximum stress against number of cycles to failure for two types of Al4%Cu specimens with shearable precipitates.
X = as quenched, 2 mm notch
0 = aged for 18 hours at 130°C, 5mm notch.

CONCLUSIONS

(i) Slip plane facet formation in the two aluminium alloys studied appears to be limited to those microstructures that are easily sheared and in which PSB damage can lead to solute migration to the matrix.

(ii) In view of the calculated possible hardness differences between the matrix and the PSB in these cases, it is unlikely that complex faceting incorporating more than one crystallographic plane can occur unless solute migration takes place.

(iii) There exists an apparent threshold in fatigue life below which slip plane facets do not appear.

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