On the possible mechanism(s) of fracture in commercial aluminium

Recently the deformation behaviour of a commercial grade of aluminium (analysis wt%: Fe 0.51; Si 0.10; Mn 0.06; Cu 0.05; B 0.007; Ti 0.003; V 0.003; Al balance) under different experimental conditions was reported [1, 2]. As the solubility limit of Fe in Al is only ~0.052 wt%, the as received material consisted of Al as 0.8–1.0 vol% of coarser Al_3Fe (1 to 5 μm) precipitates (which were formed during solidification).

Two different grain sizes were developed by soaking the cold-rolled tensile specimens at 773 and 873 K, respectively, for an hour and water quenching them. At 773 K all elements other than Fe would be fully in solution. If linear variation in the solubility of Fe in Al is assumed in the range 623 to 923 K then, in material of smaller grain size about 0.026 wt% Fe would be in solution. On the other hand, the coaser-grained material would contain ~0.043 wt% Fe in solution. Thus tensile testing of the as-quenched specimens in the range 300 to 700 K was equivalent to testing Al which contains very small but varying amounts of Si or Al_3Fe [1, 2].

Somewhat complicated elongation to fracture–test temperature curves resulted. The only effect of an increase in strain rate or a decrease in grain size was to shift the extrema to higher temperatures (Fig. 1). The three minima were respectively due to the precipitation of Si, Al_3Fe and an ultra-fine precipitate which was too small for structural identification. Between a minimum and the subsequent maximum the corresponding precipitate gradually re-dissolved. In its range of stability, ductility decreased with an increase in particle-size.

It was clear that all the three types of precipitates blocked the motion of dislocations and that precipitation occurred preferentially near the grain boundaries [2]. Thus fracture could have been initiated by any one of the following three mechanisms: (a) effective pinning of the boundaries by the precipitates which could lead to intergranular failure, (b) cracking of the precipitates during deformation which causes trans-granular fracture, or (c) decohesion along the precipitate-matrix interface which is responsible for subsequent intra-granular failure.

Both optical and scanning electron microscopy

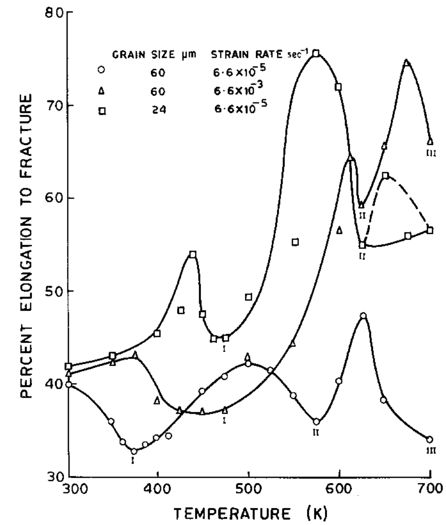

Figure 1 Elongation to fracture against test temperature curves as a function of strain rate and grain size. The initial strain rate–grain size combinations, and the positions of the three minima in ductility, have been indicated.

of fractured specimens were used for identifying the possible mechanism(s) of fracture. While preparing specimens for optical microscopy the usual mechanical grinding and polishing was supplemented by an electro-etch in a bath of 80% ethyl alcohol and 20% perchloric acid and a final chemical etch in a solution consisting of 1% HF, 1.5% HCl, 2.5% HNO_3 and 95% H_2O. Material of both grain sizes was examined for identifying the important features.

The untested material consisted of regular, polyhedral grains, typical of the annealed condition. At all temperatures grain boundary shear, which made the boundaries appear broad and distorted, was present (Fig. 2). No clear evidence

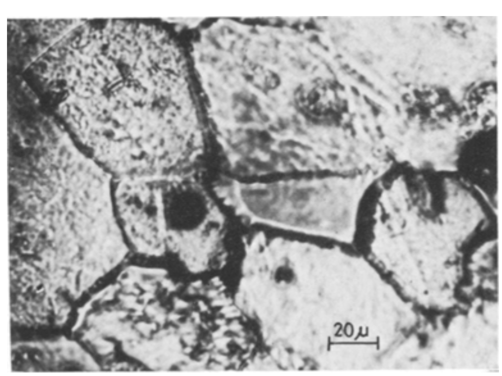

Figure 2 Broad and distorted grain boundaries, (evidence for grain boundary shear), seen at the temperature of the third maximum in ductility.

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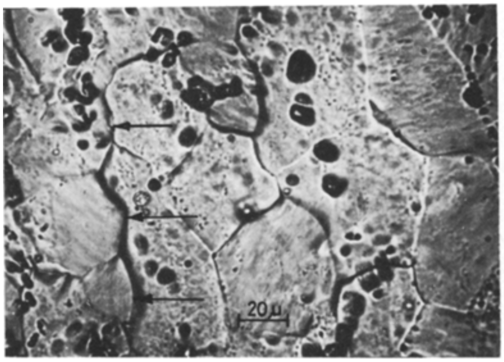

Figure 3 Evidence for the formation of cusps at the boundaries (shown by arrows) at the temperature of the first minimum in ductility. This indicates boundary pinning.

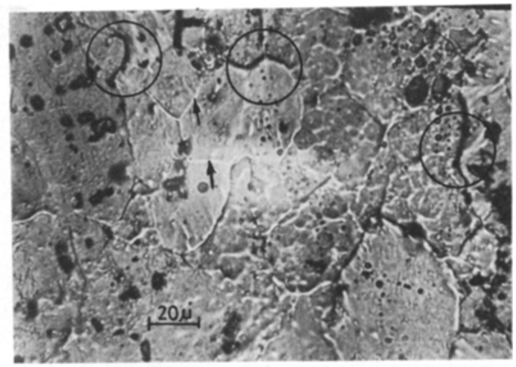

Figure 5 Evidence for the formation of grain boundary cusps (shown circled) and for boundary migration (shown arrowed), seen at the temperature of the second minimum in ductility.

for the pinning of the boundaries by the precipitates (consisting of both the coarser precipitates produced during solidification and the finer ones resulting from deformation) could be found *except* at the temperature of the first minimum in ductility (Fig. 3). However, even here the

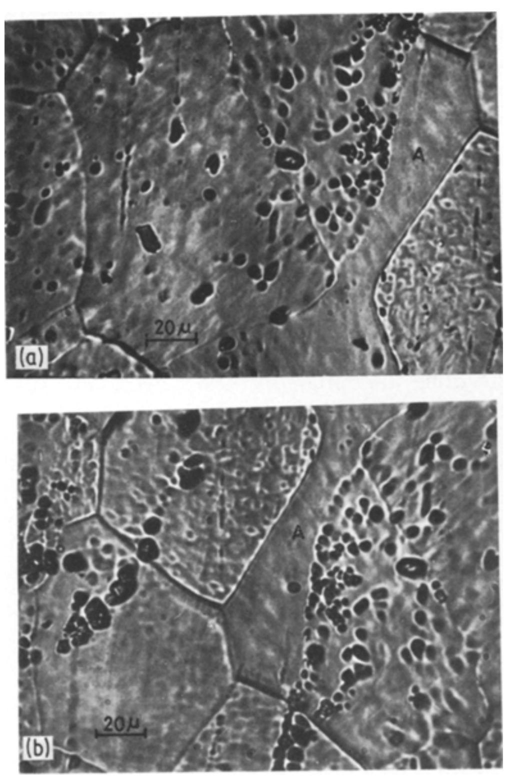

Figure 4 Evidence for boundary migration at the temperature of the first minimum in ductility. The swept regions have been marked "A". (a) and (b) are produced from different regions of the specimen.

mobility of boundaries was high and migration easy (Fig. 4). At the second minimum in ductility evidence for boundary migration was also present (indicated by arrows in Fig. 5) and perhaps some evidence for boundary pinning could be seen as well (circled in Fig. 5). No evidence for the pinning of the boundaries could, however, be obtained at the temperatures corresponding to the ductility maxima. This was reasonable because earlier work [1, 2] has shown that at the temperature corresponding to a maximum in ductility the precipitate responsible for the previous minimum had re-dissolved. Therefore on the basis of optical metallography it was concluded that even when the precipitates pinned the boundaries the latter

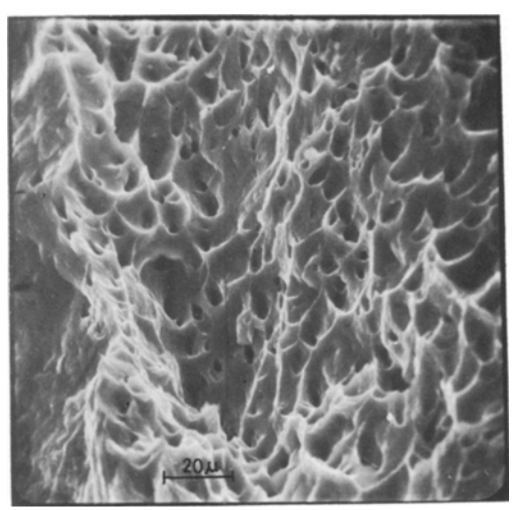

Figure 6 Scanning electron micrograph of a specimen tested at an initial strain rate of $6.6 \times 10^{-5}\ \mathrm{sec}^{-1}$, with initial grain size $60\,\mu\mathrm{m}$ and a test temperature of 398 K.

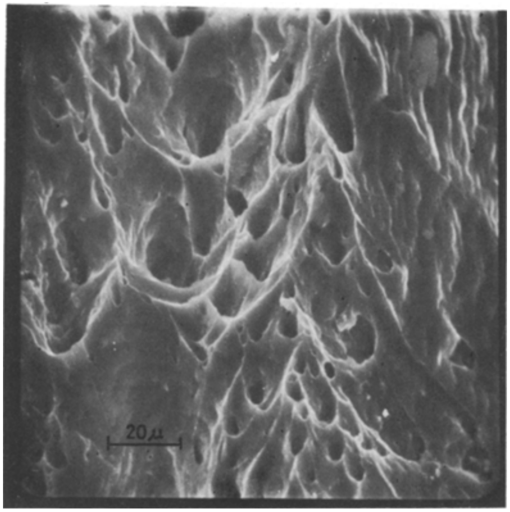

Figure 7 Scanning electron micrograph of a specimen tested at an initial strain rate of 6.6×10^{-5} sec^{-1}, with initial grain size 60 μm and a test temperature of 498 K.

could easily migrate. Therefore, in the entire range of the tests, fracture should be trans-crystalline.

This conclusion was confirmed by scanning electron microscopy. Three specimens corresponding to (a) the neighbourhood of the temperature of the first minimum, (b) the temperature of the second maximum and (c) a temperature mid-way between the third maximum and the third minimum in ductility were examined in detail (Figs. 6 to 8). It was clear that in all cases the fracture was trans-granular. Comparison of these figures with the ductility values reported in Fig. 1, on the other hand, indicated that ductility increased with the size of the dimples (which was only to be expected).

It has been reported [3, 4] that in aluminium alloys Si behaves in a brittle manner up to about 623 K and Al_3Fe is brittle up to between 723 and 773 K. As the ductility minima due to Si and Al_3Fe are encountered well within these specified ranges of temperature, initiation of fracture by the breaking of the precipitate particles is certainly a possible mechanism. Some evidence in support of this conclusion was perhaps present (a broken particle has been arrowed in Fig. 8). Nevertheless it is felt that further work in the form of interrupted tests (involving different strains) is necessary for unequivocal inference and for checking whether decohesion at the particle-matrix interface is also present or not.

An interesting but accidental observation encountered during the optical microscopic work concerns the formation of a "diamond" configuration (Fig. 9) which has earlier been the subject of comment in these columns [5 to 7]. On the basis of quantitative metallography Vakil Singh *et al.* [7] have concluded that the shape of the "diamond" grain is essentially the same as that of the "annealed" grain but in a distorted form. As in the earlier case [7], Fig. 9 also represents a surface observation (longitudinal section). Moreover, the following (qualitative) conclusions concerning the origin of the "diamond" shape

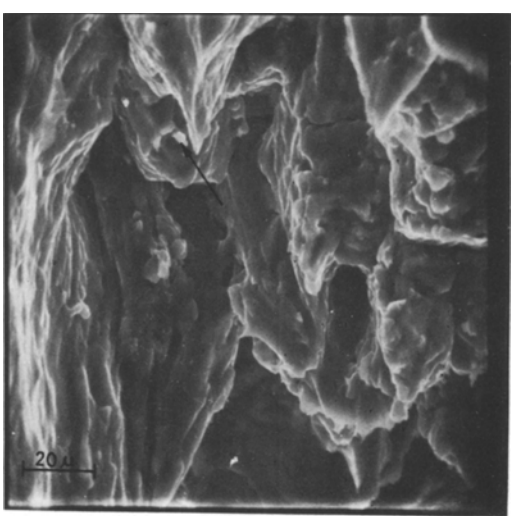

Figure 8 Scanning electron micrograph of a specimen tested at an initial strain rate of 6.6×10^{-5} sec^{-1}, with initial grain size 60 μm and a test temperature of 648 K. A broken particle is shown arrowed.

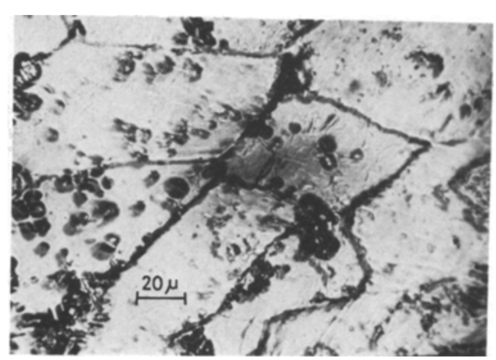

Figure 9 Illustration of the tendency for the formation of a "diamond" grain configuration on the surface of a tensile tested specimen at the temperature of the second maximum in ductility. The stress direction is horizontal.

[7, 8] are consistent with our results: (a) grain boundary sliding must be present; (b) boundary mobility, e.g. as conferred by migration, should be high; (c) defect imbalance on either side of a boundary is an important variable so that it is minimal when the boundaries have assumed a 45° position; and (d) grain boundary cavitation, if present, stabilizes the boundaries at a 45° orientation (see Fig. 9).

However, as noted by Vakil Singh *et al.* [7] themselves, the observation of a "diamond" pattern on the surface of deformed aluminium cannot be easily reconciled with their conclusion that during monotonic deformation the pattern will be seen only if the crystal structure is h c p.

It is also significant that Fig. 9 corresponds to a temperature of maximum ductility, i.e. a condition where grain boundary sliding occurs dominantly and boundary mobility is high. No "diamond" pattern could be seen at the temperatures of the minima in ductility. At these temperatures the precipitates pinned the boundaries which later migrated (Figs. 3 to 5). Under these conditions the boundary mobility should be less. It is pertinent that Vakil Singh *et al.* [5] also failed to observe a "diamond" pattern in a dilute zinc alloy in which the solute atoms segregated to the grain boundaries.

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Received 7 January
and accepted 30 January 1980

A. M. HAMMAD*
K. A. PADMANABHAN†
T. R. ANANTHARAMAN
Department of Metallurgical Engineering,
Banaras Hindu University,
Varanasi – 221005 (India)

*Present address: Atomic Energy Establishment, Cairo, Egypt.
†Present address: Department of Metallurgy, Indian Institute of Technology, Madras – 600 036 (India)