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The influence of stacking fault energy on ductile fracture micromorphology

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The influence of stacking fault energy on microvoid coalescence in "pure" materials has been studied. It was shown that as a material's stacking fault energy (SFE) decreased, the extent of microvoid coalescence that occurred during ductile fracture also decreased. The decrease of microvoid coalescence in low SFE materials was attributed to a hindrance in the development of dislocation cells associated with the restricted motion of dislocations. In "pure" materials, microvoids are believed to initiate and grow along dislocation cell walls formed during deformation. As such, the absence or scarcity of cells in lower SFE materials limits the formation of these voids during ductile fracture.

1. Introduction

Microvoid coalescence is the common mode of ductile fracture in many engineering alloys and involves the nucleation, growth and eventual coalescence of many microvoids across the final fracture plane. The fracture surface resulting from these processes is characterized by the presence of nominally equiaxed (hemispherical) or elongated (parabolic) depressions. It is generally argued that microvoids initiate and grow from second phase particles in commercial materials. Void initiation can occur in these materials either by particle cracking [1] or by particle/matrix decohesion [2]. However, these mechanisms cannot explain the formation of microvoids in high-purity materials. In "pure" materials as well as in "clean" materials containing strongly bonded, ductile second phases, microvoids are believed to initiate and grow along the dislocation cell walls which form during the deformation process. Numerous studies during the past 15 years have used the high voltage electron microscope (HVEM) to directly observe the initiation of voids during *in situ* straining experiments (e.g. Wilsdorf and co-workers [3–6]). In each of these investigations, microcracks (or voids) were seen to initiate and propagate along dislocation cell walls although none of these investigations attempted to correlate the dislocation substructures to the resulting fracture surface morphology.

As microvoids in pure materials initiate and grow along dislocation cell walls, one might expect that the nature and extent of microvoid formation should be related in some manner to the dislocation cell size and the density of dislocations within the cell walls [7]. Specifically, it would be expected that with decreasing stacking fault energy (SFE) of a "pure" material, the size and extent of microvoids would decrease because SFE is a controlling parameter in the formation of dislocation cells. In low stacking fault energy materials, dislocation cross-slip is difficult because the partial dislocations are widely separated with their recombination onto the new slip plane requiring considerable energy. As a result, dislocations are restricted to planar arrays and dislocation cells do not readily form. With increasing SFE, the partial dislocations are closer together and may recombine readily to facilitate cross-slip. As such, dislocation cell formation is enhanced. Therefore, differences in microvoid morphology should be expected with changing SFE in association with void nucleation at dislocation cell walls. The objective of this study is to examine the effect of dislocation cell structures as influenced by SFE on the formation of microvoids.

2. Experimental procedure

Three high-purity ($> 99.999\%$ pure) materials were selected for this study: aluminium (SFE $\simeq 200\,\mathrm{mJ\,m^{-2}}$), copper (SFE $\simeq 50\,\mathrm{mJ\,m^{-2}}$), and copper–7 wt % aluminium (SFE $\simeq 3\,\mathrm{mJ\,m^{-2}}$). The alloy was induction melted under vacuum and all three materials were suitably cold-worked and annealed to produce samples of the same grain size. Following the annealing treatments, round bar tensile specimens from each material were machined in accordance with ASTM E8 specification [8] for subsize tensile bars. Axial tensile tests were conducted at ambient temperature on these samples using an Instron tensile machine at a strain rate of $2 \times 10^{-3}\,\mathrm{cm\,sec^{-1}}$. Additional tensile tests were conducted at 400° C with copper and at − 150° C with aluminium samples using an environmental chamber. The purpose of the high- and low-temperature tests was to alter the dislocation cell sizes for a given material by the enhancement (at high T) or the restriction (at low T) of dislocation climb and thermally activated cross-slip. Therefore, if the premise of this research is true, then the size and character of the microvoids for a given material should change with test temperature.

All fracture surfaces were examined in an Etec Autoscan scanning electron microscope (SEM) to determine the size and nature of the microvoids

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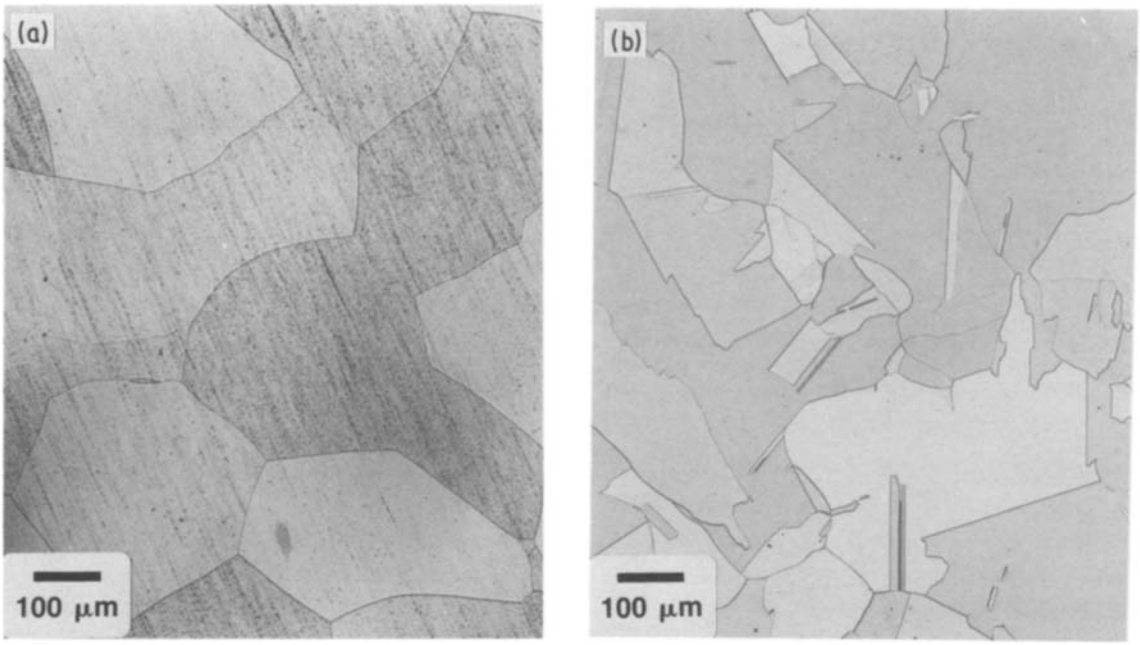


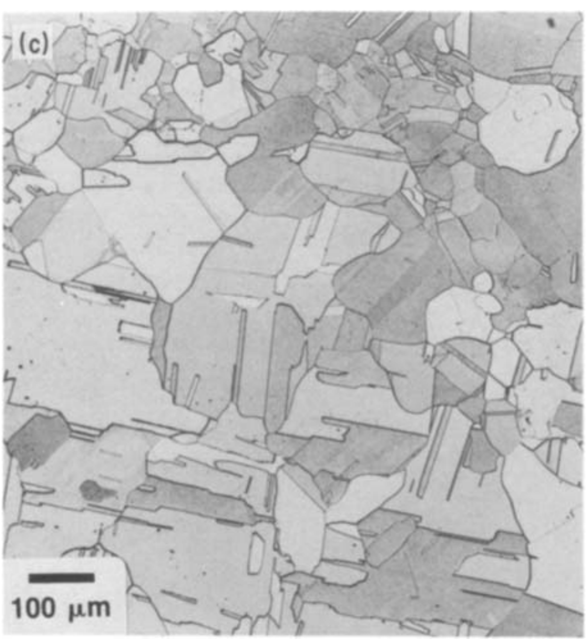


Figure 1 Optical micrographs of each material showing the initial grain sizes; (a) pure aluminium, (b) pure copper, and (c) Cu–7Al. Note the increasing number of annealing twins present in the microstructures as the SFE is decreased (a to c).

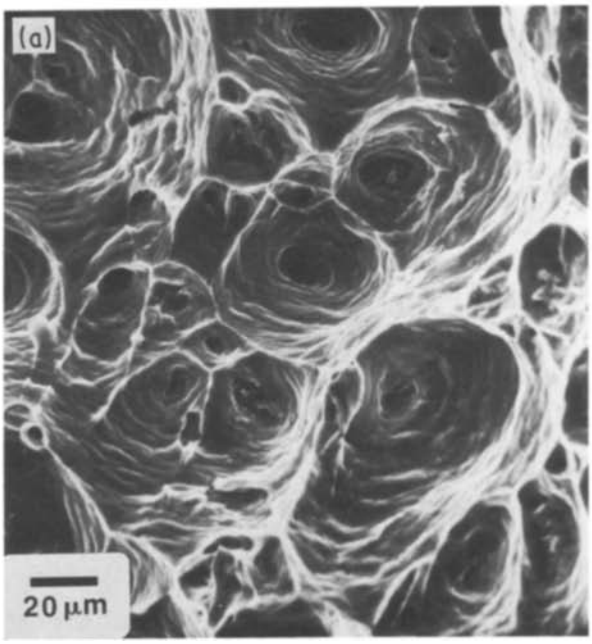
formed. Samples for TEM observation were sectioned from tensile specimens immediately adjacent to the fracture surfaces with all TEM foils being taken from necked regions where the local reduction in area exceeded 15%. Transmission electron microscopy was conducted in a Philips EM400T operated at 120 kV to determine the dislocation configurations resulting from the deformation. All TEM images were recorded under the same diffraction conditions close to an $\langle 0\,1\,1 \rangle$ zone.

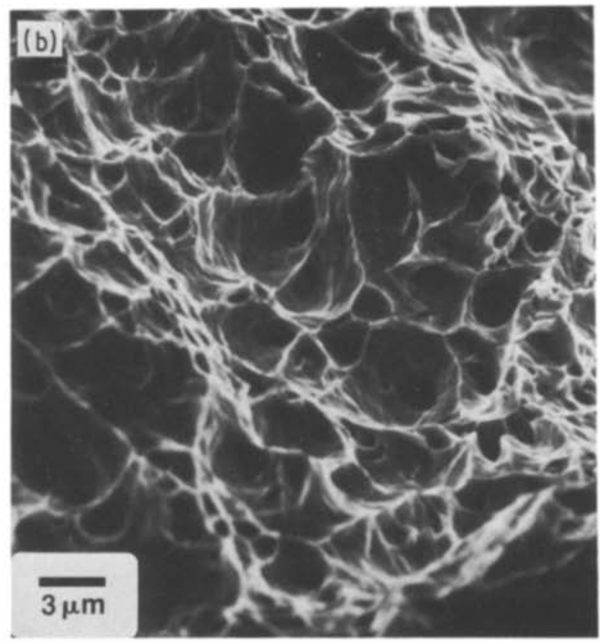
3. Experimental results and discussion

Prior to testing, the microstructure of each material was examined to determine the initial grain size. Fig. 1 shows optical micrographs of the microstructure of each material; the initial grain size in each material was between 200 and 300 μm. Annealing twins were observed in both the copper and Cu–7Al samples, in amounts consistent with the low stacking fault energy in each material.

Tensile tests revealed that both the pure aluminium and pure copper samples underwent greater than 90% reduction in area, whereas the Cu–7Al sample experienced $\sim 75\%$ reduction at fracture. Fig. 2 shows typical micrographs of the tensile fracture surfaces for each material generated at room temperature. In the pure aluminium sample (Fig. 2a) very large diameter ($> 20\,\mu$m) microvoids were formed and extended well below the fracture surface plane. Relatively few smaller diameter ($< 5\,\mu$m) microvoids were also observed on this fracture surface. Fig. 2b shows the microvoid fracture surface of the pure copper sample. In this case, the microvoids were 2 to 4 μm diameter and in most instances did not extend far below the fracture plane. The Cu–7Al fracture surface (Fig. 2c) reveals microvoids of 1 to 4 μm diameter which may be characterized as shallow essentially flat-bottomed voids which formed on shear walls oriented approximately 45° to the loading direction. As such, both the diameter and depth of microvoids formed in these metals decreased with decreasing stacking fault energy (Al $\rightarrow$ Cu $\rightarrow$ Cu–7Al).

Examination of thin foils taken immediately adjacent to these fracture surfaces revealed very different dislocation substructures. Fig. 3a shows the typical dislocation cell structure observed in the pure aluminium sample; well-defined narrow dislocation cell walls were developed with dislocation cell diameters found in the range 1 to 3 μm. The cell interiors are essentially free of dislocations and dislocation tangles. A large degree of misorientation between adjacent cells may be noted as evidenced by the large changes in diffraction contrast from one cell to another. Fig. 3b shows the dislocation substructure typical of the deformed pure copper sample. The dislocation cells are typically less than 1 μm diameter, and the cell walls are wider and less well-developed than in the





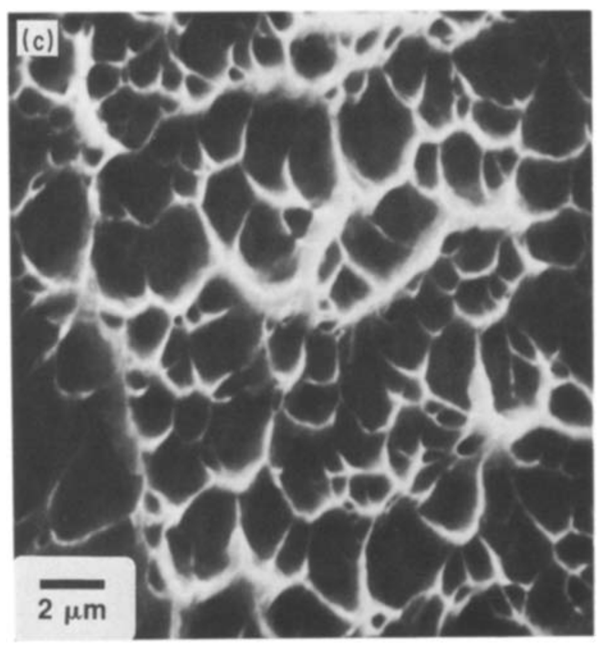


Figure 2 Scanning electron micrographs showing the typical fracture surface appearance of each material; (a) pure aluminium, (b) pure copper, and (c) Cu–7Al. Note the decrease in size and depth of the microvoids formed with the decrease in SFE from a to c.

aluminium sample. Also, the cell interiors contain numerous free dislocations and dislocation tangles.

In the Cu–7Al sample dislocation cells did not form during deformation (Fig. 3c). Rather, extensive planar slip occurred throughout the material with numerous stacking faults present in the microstructure. As no dislocation cell walls were observed in the microstructure, fracture is believed to occur by slip plane decohesion, with the flat voids forming as the crack propagates from one slip plane to an intersecting slip plane.

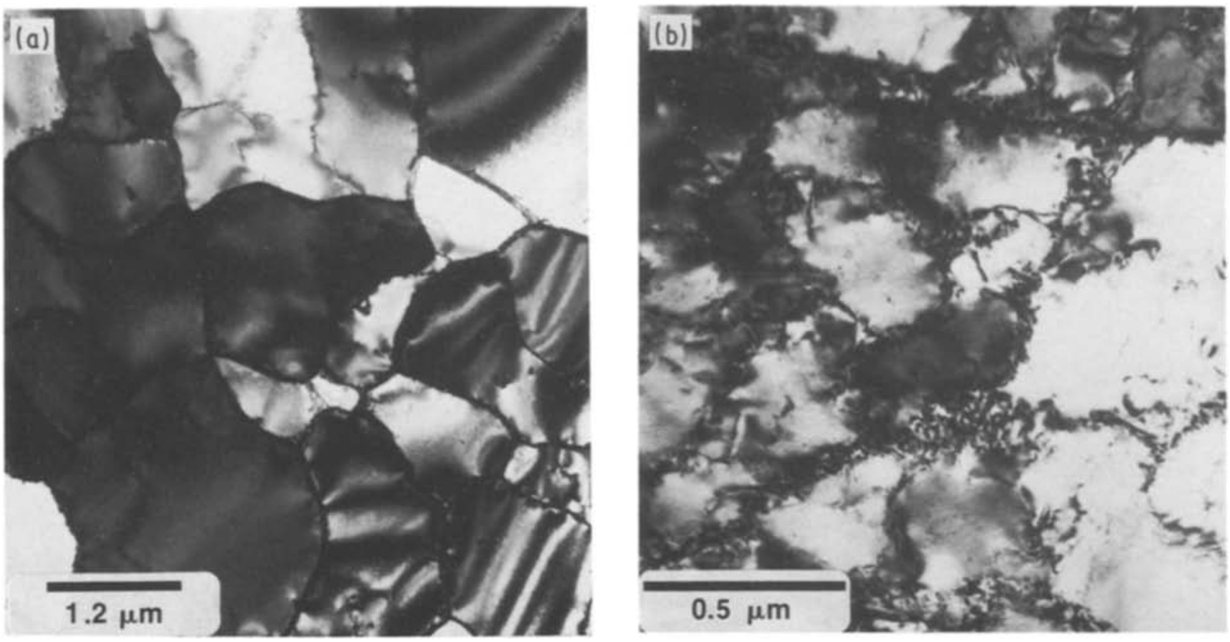
Although it appears that a one-to-one relationship between the size of the dislocation cell and microvoid does not exist, it is obvious that a relationship does exist between the extent of dislocation cell development and microvoid size and depth. Clearly, as the stacking fault energy of a material decreases, the size and extent of both microvoids and dislocation cells decreases.

Two additional experiments were conducted to investigate the relationship between dislocation substructure and microvoids. Fig. 4 shows the fracture surface and associated dislocation substructure from an aluminium tensile bar tested at $-150°$ C. By comparison to the aluminium sample tested at room temperature (Figs 2a, 3a), the microvoids in the low-temperature sample are smaller ($\sim 10\,\mu$m diameter), and are not as deep or as uniformly developed. Additionally, the dislocation substructure in the low-temperature sample contains a smaller, less uniform cell size (0.5 to $2\,\mu$m), with many free dislocations in the cell interiors. It is clear that dislocation kinetics and microvoid formation processes were attenuated under subambient testing conditions.

A tensile test of pure copper was also conducted at 400° C. At this temperature dislocation kinetics were enhanced compared to that associated with room temperature deformation processes. Fig. 5 shows the fracture surface and dislocation substructure in the pure copper sample tested at 400° C. By comparison with Fig. 2b and 3b of the room-temperature copper sample, the microvoids in the elevated temperature sample appear unchanged in size although they are more uniform and better developed (i.e. uniformly deep). Correspondingly, the dislocation substructure in the elevated temperature sample possesses more well-defined cell boundaries and slightly larger dislocation cells. The cell interiors still contain numerous free dislocations and dislocation tangles. Again, it appears that a relationship exists between the size and extent of microvoid formation and associated dislocation substructure resulting from changes in dislocation kinetics.

4. Conclusions

The morphology of microvoids formed during ductile



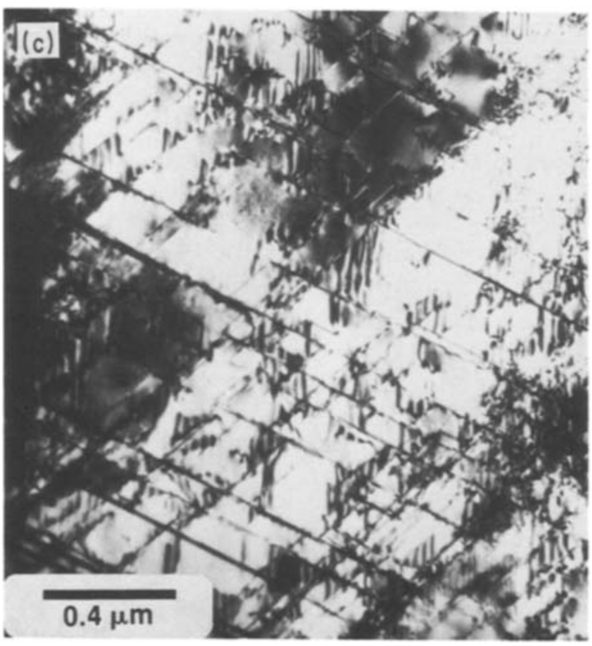


Figure 3 Transmission electron micrographs showing the typical dislocation substructures present immediately adjacent to the fractures shown in Fig. 2. (a) pure aluminium, (b) pure copper, and (c) Cu–7Al. Note the decrease in dislocation cell development and cell size with the decrease in SFE from a to c.

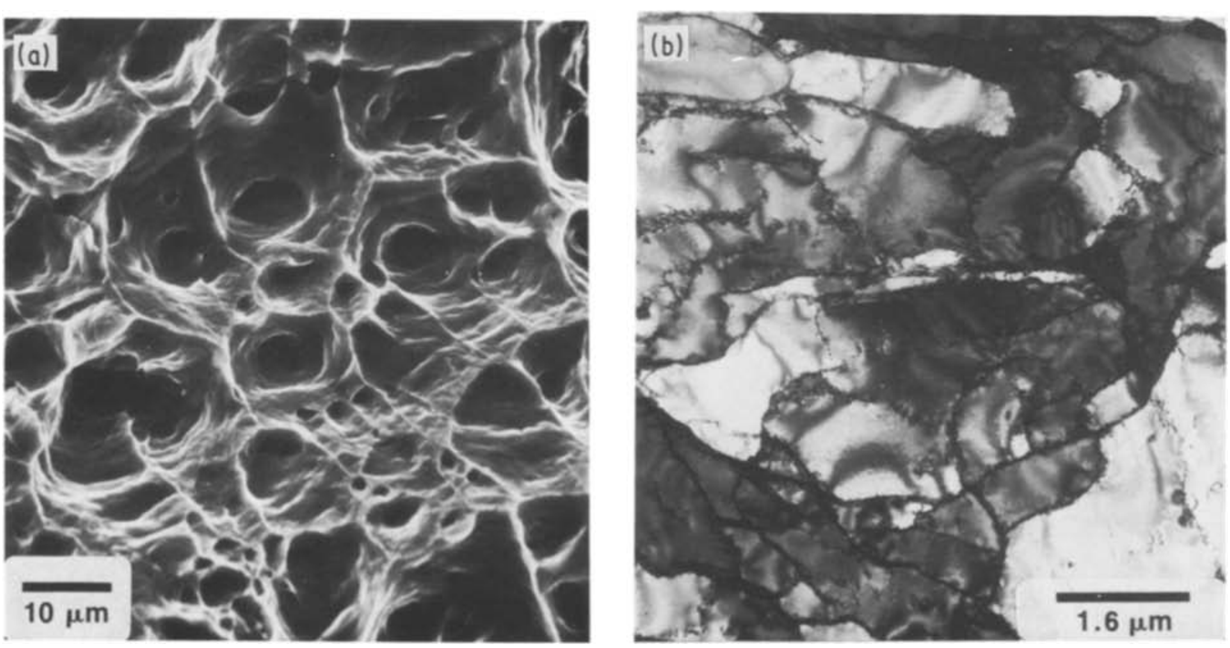


Figure 4 Pure aluminium specimen tested at -150° C. (a) Scanning electron micrograph of fracture surface; (b) Transmission electron micrograph of dislocation substructure immediately adjacent to the fracture surface. By comparison with the room-temperature pure aluminium sample (Figs 2a, 3a) both the microvoids and dislocation cells are smaller and less well developed at -150° C.

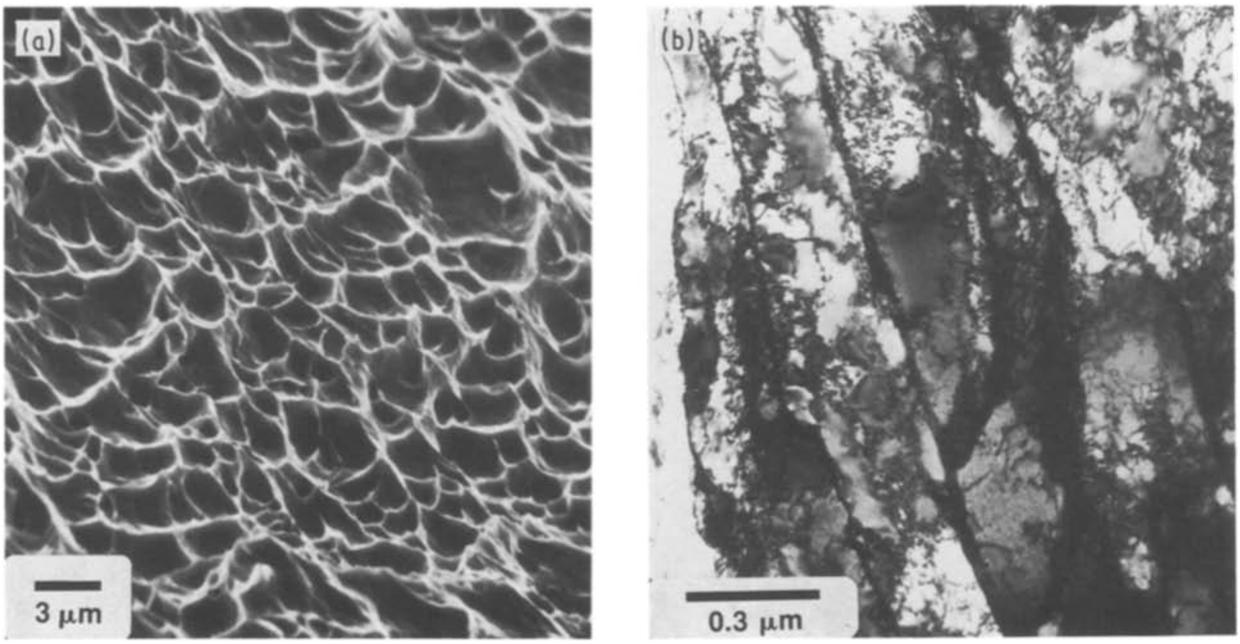


Figure 5 Pure copper specimen tested at 400° C. (a) Scanning electron micrograph of the fracture surface; (b) transmission electron micrograph of dislocation substructure immediately adjacent to the fracture surface. By comparison with the room-temperature pure copper sample (Figs 2b, 3b) the microvoids and dislocation cells appear more well developed at 400° C, although the absolute cell and microvoid sizes appear unchanged.

fracture in pure materials is influenced by the stacking fault energy of the material. With decreasing SFE, the size and extent of microvoids decreases due to the restriction of dislocation motion. Although a one-to-one correlation between dislocation cell size and microvoid size does not appear to exist, a clear relationship between the development of dislocation cells and the extent of microvoid coalescence is noted.

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