The Shear Mode of Ductile Fracture in Materials with Few Inclusions

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The mechanism of the shear mode of ductile fracture has been investigated in sheet tensile specimens of an α brass containing few inclusions. The techniques of optical metallography, scanning electron microscopy and transmission electron microscopy of both thin foils and replicas were all used in order to obtain as complete a picture of the fracture mechanism as possible. It was found that a zone of shear deformation, made up of many intense shear bands, developed across the specimen neck and fracture occurred within this zone along the shear bands. These shear bands consisted of material with a fine subgrain structure. No voids were found in the necked region of either deformed or fractured specimens yet the fracture surfaces were covered with elongated dimples. It is therefore suggested that the dimples on the fracture surface are not a precursor of the fracture but form only during the final process of separation.

DUCTILE fracture of commercial metals and alloys involves two basic modes, the *fibrous mode* and the shear mode. Fibrous fracture is characterized by the formation and growth of voids from nonmetallic inclusions or second phases leading to a fracture surface covered with equiaxed dimples. The mechanism of this type of fracture is fairly well understood.[1,2] The shear mode of ductile fracture, on the other hand, gives rise to fracture surfaces covered with strongly-oriented parabolic dimples. The mechanism by which this type of fracture occurs is not so well understood, particularly for some classes of materials.

The shear mode of ductile fracture occurs in a number of different situations including the outer portion of cup-cone fractures, the tensile fracture of sheet materials and the fracture of some materials under hydrostatic pressure. A general model for this type of fracture has been proposed[2] which involves the development of zones of heavy shear deformation within which sheets of voids form, followed by fracture due to the axial tensile stress. No specific mechanism for the formation of the large number of voids within the shear zone was suggested by the model.

The formation of these voids can be satisfactorily explained in terms of nucleation at inclusions for materials which contain many small inclusions, such as internally oxidized copper alloys[3] and aluminum alloys.[4] Additional features of this type of fracture which have become apparent from studies of these materials are that, where there is a strong inclusion-matrix interface, such as in aluminum alloys, voids develop only just prior to fracture and that, in any case, void coalescence occurs rapidly and only close to the fracture path.

The mechanism of formation of sheets of voids in inclusion-free material has not been studied in detail. A number of studies have been made of the problem of nucleation of voids in the absence of inclusions which suggest that cracks could form at grain boundaries due to dislocation pile-up[5] or by clustering of vacancies produced by moving dislocations.[6] Also, crack propagation in single crystals of high purity has been shown to develop by coalescence of crystallographic holes in the reoriented material formed in the necked region.[7] It seems unlikely however that any of these mechanisms could account for the formation of sheets of voids in polycrystalline material subjected to the very large strains which exist in a zone of heavy shear deformation.

The present paper describes an examination of the shear mode of ductile fracture in metals relatively free from fine inclusions. The material chosen for study was a commercial purity α brass. The techniques of optical metallography, scanning electron microscopy (SEM) and transmission electron microscopy (TEM) of both thin foils and replicas were used in order to obtain as complete as possible a picture of the shear fracture and of the processes leading to it.

MATERIALS AND METHODS

The material used was an α brass of composition 70.1 wt pct Cu and remainder Zn with trace impurities only, originally in the form of 13 mm thick hot rolled plate. This brass contained a few elliptical ZnS inclusions, about $30/mm^2$, which were visible under the optical microscope. The average size of these inclusions was 2.6 μm by 1.2 μm with the larger dimension parallel to the rolling plane. Flat tensile specimens with round threaded ends were machined from the midsection of the plate so that the gage cross-section of 10 mm $\times$ 0.65 mm lay parallel to the rolling plane and the 20 mm gage length was parallel to the rolling direction. Sheet tensile specimens were used because previous work[8] has shown that such specimens fracture entirely by the shear mode. Specimens of this geometry were also convenient for obtaining TEM specimens.

The specimens were deformed in an Instron tensile machine usually at a deformation rate of 0.5 mm/min although some specimens were deformed more slowly so that details of the load-extension curves could be studied. In some cases the deformation was continued until complete separation of the specimen had occurred whilst in others the deformation was stopped prior to this.

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Manuscript submitted May 3, 1976.

Detailed optical microscopy of some of these deformed specimens was carried out by examining longitudinal sections perpendicular to the wide faces at a number of positions across the specimen width. In this way the details of the macroscopic development of the deformation and fracture could be followed. The fracture surfaces of some specimens were examined using SEM and TEM of two-stage carbon replicas. The SEM studies were useful for gaining an overall picture of the topography of the fracture surfaces whilst replicas were useful for looking at fine details. TEM studies at 100 kV were made of thin foils taken from both fractured and highly deformed specimens. In the case of the highly deformed specimens 3 mm diam discs for thinning were spark-machined from the center of the necked region. Fractured specimens were first electroplated with a thick layer of copper then discs were spark-machined so that the fracture edge was approximately in the centre of the disc. Final thinning was done in both cases by electropolishing in the chromium trioxide/acetic acid solution described in Ref. 9. The distances of examined areas in fractured specimens from the fracture edge were ascertained from optical micrographs of the thinned discs. It was found that discs could be repeatedly electropolished until thin areas were obtained close to the fracture edge.

RESULTS

a) Load-Extension Data

The material was quite ductile, having an elongation to fracture of between 100 and 130 pct on the 20 mm gage length. Necking began at an elongation about 10 pct less than the fracture elongation. After necking had started the load began to decrease gradually but this was followed by a rapid drop in load when fracture started. The load-extension curve of a specimen strained at a lower rate than usual (0.05 mm/min) showed that the rapid load drop actually occurred in about six sharp steps with most of the elongation occurring between these steps under conditions of gradually decreasing load. At this late stage of the test the deformation could be stopped between any of the sharp steps in load but not during them.

b) Optical Metallography

Examination of longitudinal sections perpendicular to the wide faces of strained specimens showed that a broad neck began to develop after a uniform reduction in thickness of about 37 pct. This neck developed by general deformation until the specimen thickness was reduced by approximately 55 to 60 pct. At this stage the deformation became localized into a zone of shear deformation extending across the specimen neck. This zone contained many roughly parallel bands of shear deformation (see Fig. 1). The width of the individual dark etching bands was about 0.7 μm.

At positions within the specimens where the reduction in thickness exceeded 60 pct a well developed fracture had usually occurred. Such fractures first occurred near the center of the specimen width and invariably they occurred within the shear zone. Fig. 1 shows an early stage of the fracture. The mating

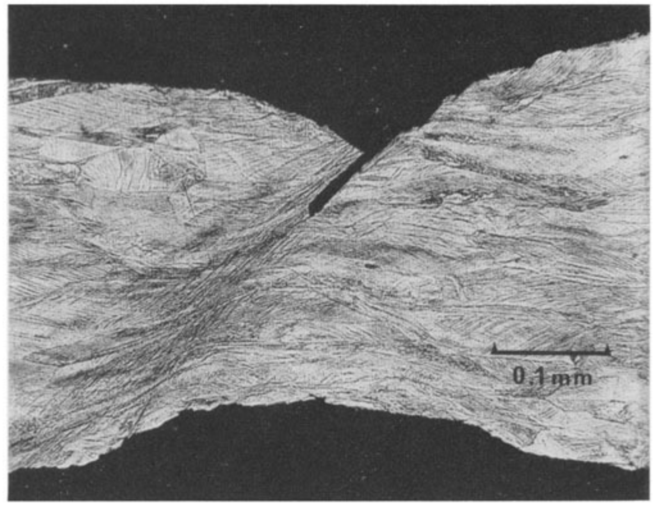


Fig. 1—Optical micrograph of longitudinal section perpendicular to the wide face of a deformed specimen showing the beginning of fracture.

surfaces were often quite wide apart indicating that some elongation had occurred after their formation. Longitudinal sections away from the center of the specimen width showed that the fracture consisted of two or more portions of differing crack width indicating that fracture spread toward the specimen edges in a number of discrete stages.

The zones of shear deformation and the subsequent fracture appear to occur independently of the inclusions in the material since it was rare for sections of zones to contain an inclusion. In the very few instances when a shear zone passed through an inclusion the inclusion was broken into a number of pieces and voids existed between these pieces. An important feature of the shear zones was that no optically visible voids were found in the zones in either deformed or fractured specimens, except at the rare inclusions. Any such voids must therefore form only along the fracture path very late in the deformation.

c) Studies of the Fracture Surfaces

Studies of carbon replicas of the fracture surfaces using the TEM showed that the surfaces were covered with shallow dimples. The majority of these dimples were elongated and had a parabolic outline as shown in Fig. 2. This appearance is typical of the shear mode of fracture. Occasionally the fracture surfaces contained isolated larger dimples which were more equiaxed and appeared to contain remnants of particles. These were probably the voids which formed at inclusions. An important feature of the fracture surfaces was that no imprints of particles could be found at the roots of the elongated dimples (see Fig. 2). This observation, together with the fact that the dimple density of approximately $5 \times 10^5/\text{mm}^2$ was several orders of magnitude higher than the inclusion density, show that the majority of elongated dimples were not nucleated at inclusion particles.

Stereo-pairs of SEM micrographs of the fracture surfaces proved very useful for studying fracture surface geometry. Such stereo micrographs revealed

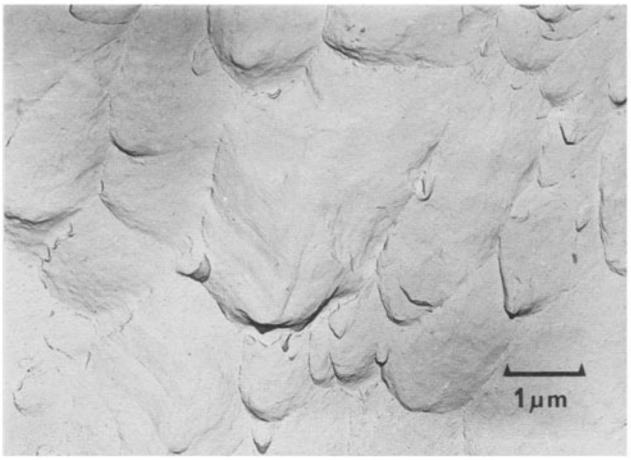


Fig. 2—Electron micrograph of two-stage carbon replica from fracture surface showing elongated dimples with no imprints of parent inclusions.

that, on a local scale, the fracture surfaces were made up of a number of roughly parallel planes a few dimples wide joined by short featureless regions at approximately right angles to these planes, as shown in Fig. 3. Also, stereo-pairs from matching regions of the fracture surfaces of the two halves of a fractured specimen showed that in general the dimples pointed in opposite directions indicating that these dimples formed in a shear manner.

d) Transmission Electron Microscopy

Thin foils were prepared from both deformed and fractured specimens from uniformly strained regions, necked regions and regions near the fracture surface. Micrographs from uniformly strained regions showed that this region contained many dislocations and stacking faults. Foils from closer to the necked region contained an increasing number of deformation twins.

Micrographs from regions within the neck of specimens were found to contain three types of deformation structure, all of which often occurred within close proximity of one another. A few areas contained a dense dislocation structure whilst other areas contained many closely spaced deformation twins. However, most of the thin areas from within the necked region had a structure consisting of fine subgrains of related orientation. Areas of this structure were most clearly recognized from their diffraction patterns which consisted of arced regions making up sections of a ring pattern as shown in the diffraction patterns of Fig. 4. The appearance of micrographs from these areas varied considerably, but two main types of structure were found. One consisted of irregular light and dark regions as shown in Fig. 4(*a*) whilst the other consisted of strongly banded regions as shown in Fig. 4(*b*). The darker areas of both of these types of appearance usually contained the fine fringe-like substructure visible in Fig. 4. Turley[9] has found similar structures in α brass rolled to reductions greater than 70 pct and has established that they are regions of fine subgrains.[10]

In fractured specimens the proportion of fine subgrain structure increased as the fracture surface was approached until, in regions less than about 0.05 mm from the fracture surface, virtually all of the material consisted of this fine subgrain structure.

These results suggest that the regions of subgrain structure observed by TEM might correspond to the dark etching shear bands observed optically since the subgrain structure, like the shear bands, occurred only within the necked region close to the fracture surface. Stronger evidence for this assertion was obtained when some large thin areas about 0.3 mm from the fracture surface were examined. These areas were made up of quite wide zones containing many parallel deformation twins bounded by quite

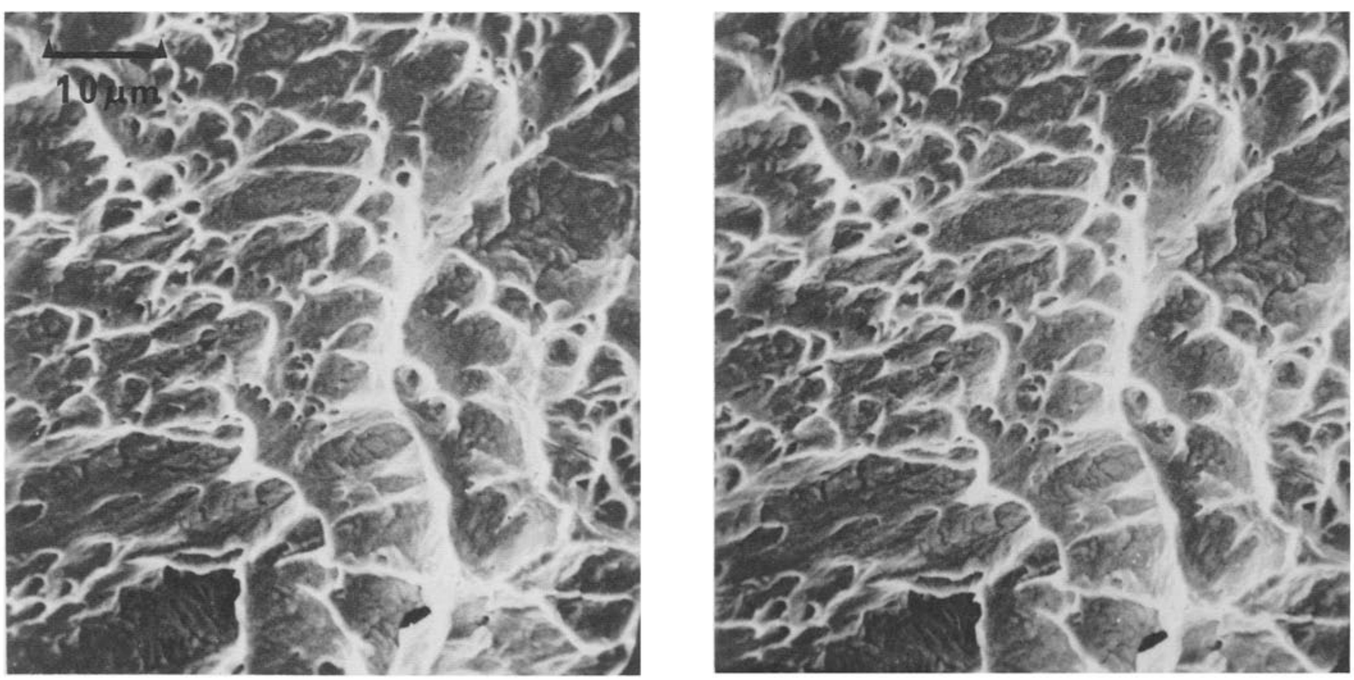
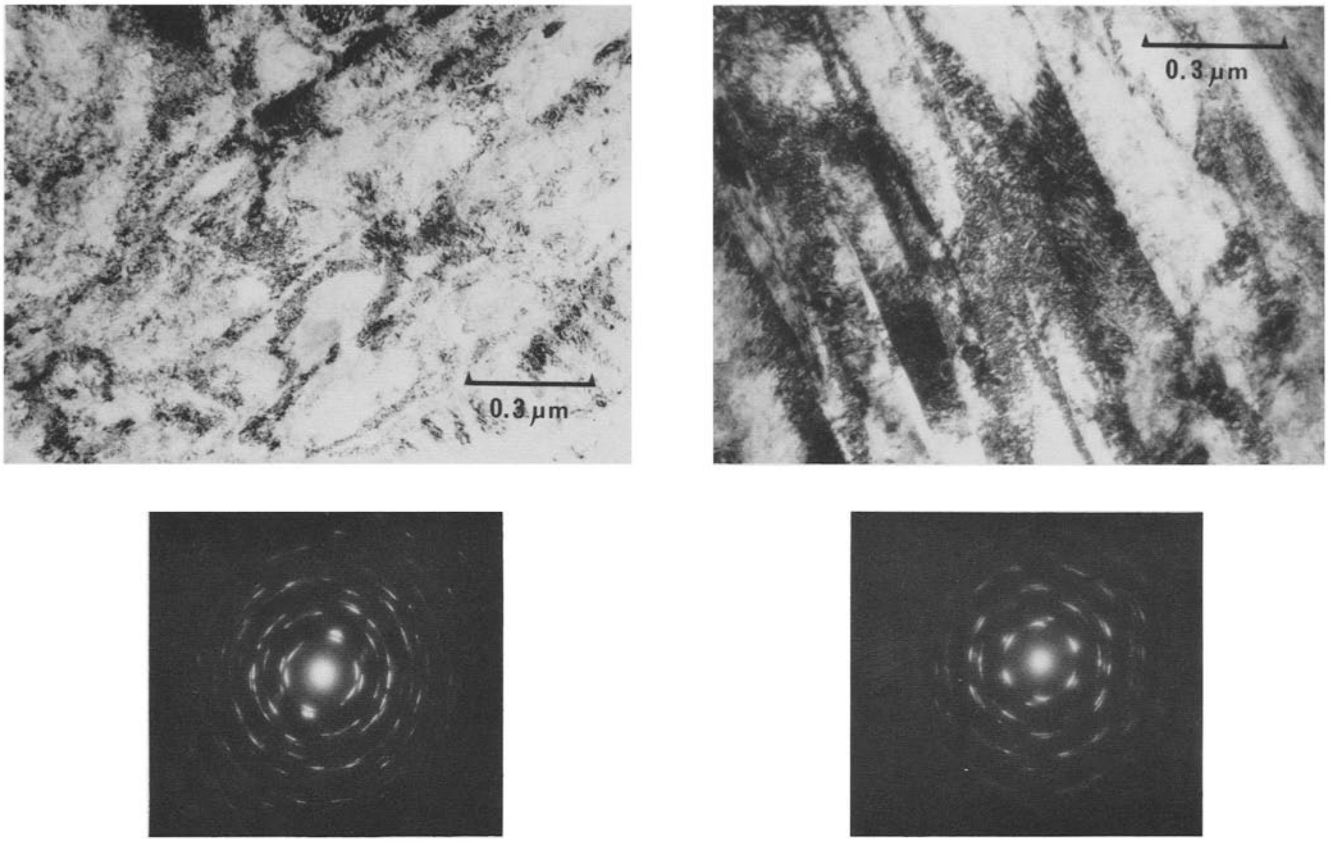


Fig. 3—Stereo pair of scanning electron micrographs of fracture surface showing that the dimples occur on many roughly parallel planes.



(a) (b)

Fig. 4—Transmission electron micrographs from within the necked region of a specimen showing the two types of appearance associated with regions of fine subgrain structure; (a) irregular light and dark regions, (b) strongly banded regions.

narrow zones (about 0.5 μm wide) with a less easily interpretable structure. Such an area is shown in Fig. 5. When the narrow zones were examined at higher magnification they were found to consist of diffuse lighter and darker regions typical of the subgrain structure. Selected area diffraction patterns from the narrow zones always consisted of arced sections of a ring pattern, similar to those in Fig. 4. The distance of these regions from the fracture surface was about that at which shear bands began to appear in optical micrographs. It therefore is very likely that the shear bands observed optically are regions of subgrain structure.

Although many areas were examined in transmission during the course of this work, no inclusions, other than the few large ones visible optically, were found. Similarly, no voids were found in any of the thin areas, even those areas which were within only 0.05 mm of the fracture surface.

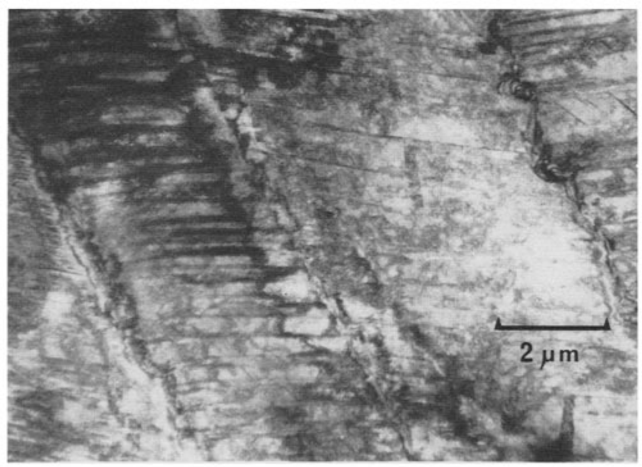


Fig. 5—Transmission electron micrograph of region away from fracture surface showing wider regions containing twins bounded by narrow regions of fine subgrain structure.

DISCUSSION

A model of the sequence of events leading to the shear mode of ductile fracture in this material can now be developed as follows. After the specimen necks a zone of shear deformation develops parallel to one of the two planes of maximum shear stress. The deformation in this shear zone proceeds by the formation of many narrow bands of intense shear. These bands are in fact regions of material with a fine subgrain structure which form separately and roughly parallel to one another. As the deformation proceeds, the material towards the center of the shear zone becomes almost totally made up of these bands. Evidently the amount of strain that can be accommodated by a single band is limited since many bands develop and the number increases with deformation.

Fracture occurs along the shear bands and the fracture surface is made up of elongated dimples. The fracture surface consists, on a local scale, of sections on a number of roughly parallel planes so that the fracture path must frequently move between shear bands. Since no voids were found except the elongated half-voids on the fracture surface, the formation of such voids cannot be a precursor to fracture but must be part of the fracture process. The shear mode of ductile fracture in α brass in sheet form therefore differs from that suggested by the void-sheet model[2] in that no voids are formed prior to the actual fracture. The fracture spreads across the specimen width in definite macroscopic stages as borne out by both the sharp drops in load observed and by the observation that the fracture contains portions of different widths.

The highly localized bands of shear deformation described above have been found to occur prior to tensile fracture in a number of materials and in a number of specimen geometries. They occur in round cross-section specimens of copper[11] and brass[12] tested under superimposed hydrostatic pressure and also in sheet specimens of copper and α-β brass[8] tested at both room pressure and high pressures. The present results suggest that these bands are regions of fine subgrain structure. Under confining pressures between room pressure and a critical pressure (which depended on the material) the formation of bands was followed, as in the present case, by an elongated dimple type fracture. At pressures greater than the critical one however all void formation was suppressed. In such cases many more shear bands developed and fracture finally occurred at a chisel-point.[8,11]

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