23. Ductile Fracture of Single Crystals

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ABSTRACT

The ductile fracture behavior of single crystals is reviewed. Results of tensile tests at 77°, 293°, and 373°K, as well as metallographic studies, are reported. By using aged Al-5% Cu alloy crystals, which do not neck prior to failure, a quantitative approach is made possible. Fracture occurs on operative octahedral or cubic slip planes at coarse slip bands that form just prior to fracture. The test results suggest that the resolved shear stress is a suitable criterion for the onset of ductile fracture. Dislocation mechanisms of fracture are discussed.

Introduction

It is first necessary to define what is meant by ductile fracture in contrast to the phenomenon of brittle or cleavage fracture, which is familiar in crystals of zinc and iron. It is not sufficient to say that ductile fracture is preceded by substantial plastic deformation, for so also is cleavage fracture in many cases. In the extreme case, some single crystals and, indeed, polycrystalline metals draw down almost to a chisel edge or a point before breaking apart occurs; this can hardly be termed fracture in the normal sense and is usually referred to as rupture. This process arises from prolonged shear on slip planes within the necked region of the crystals, which at one point finally shears apart. Ductile fracture appears to lie between these two extremes in so far as deformation first takes place by shear, and then a crack propagates, but not usually as rapidly as in the case of cleavage. With cleavage, the only work needed is that to overcome the cohesive forces between atoms on each side of the crack, whereas with ductile fracture, this contribution appears to be small compared with the work necessary to extend the crack by further plastic deformation.

Survey of Single-Crystal Behavior

Many casual observations on the ductile fracture of single crystals have been made from time to time, but although cleavage fracture has received much attention in recent years, little systematic study has been made of ductile fracture.

At first sight, the hexagonal and tetragonal metals appear to behave in the simplest way because of the limited number of operative slip systems. For example, zinc crystals at −196°C show cleavage fracture with little or no prior plastic strain, while at 300°C, rupture occurs by marked coarse slip along a few bands in the crystal.[1] This high-temperature behavior occurs also in crystals of tin and cadmium; however, as room temperature is approached, these metals do not cleave but rupture in a ductile manner. Failure occurs in a necked region, and, while little attention has been given to the phenomenon, it seems clear that, in addition to the primary basal slip, further slip must occur on other systems. In some cases, tin crystals behave in a more complicated way: First, a primary necked region is formed by the simultaneous operation of two slip systems, and then one takes over while a further substantial elongation occurs. The crystal finally necks a second time, draws down almost to a point, and ruptures (Fig. 1).

Among the body-centered cubic metals, iron alone has had much attention. It is well known that at low temperatures (−196°C) iron crystals cleave on the (001) plane. Recent work by Allen, Hopkins, and McLennan [2] on iron crystals of 99.96% purity has shown that, above −196°C, crystals of all orientations are ductile and draw down to a chisel edge with 100% reduction in area when rupture occurs. However, at −196°C, they found crystals of some orientations (near [001]) to be completely brittle with no measurable plastic deformation, while others were fully ductile. These extremes are illustrated in Fig. 2; the behavior of other crystals lay between the two extremes. This behavior was also found by Cox, Horne, and Mehl [3] and more recently by Biggs and Pratt,[4] who showed that the ductile-brittle orientation dependence is influenced by the presence of impurities such as carbon or manganese.

Turning to face-centered cubic metals, which are the primary concern of this chapter, the usual mode of fracture is ductile preceded by necking. Cleavage fracture is not encountered. In pure metals, the fracture is usually preceded by marked necking, often to 100% reduction in area at room temperature. A ductile crack can, however, form at a late stage of the process. This behavior is more pronounced the lower the temperature; for example, Blewitt, Coltman, and Redman [5] have found that

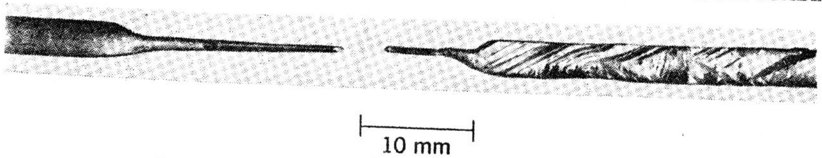


Fig. 1. A tin crystal showing double neck formation.

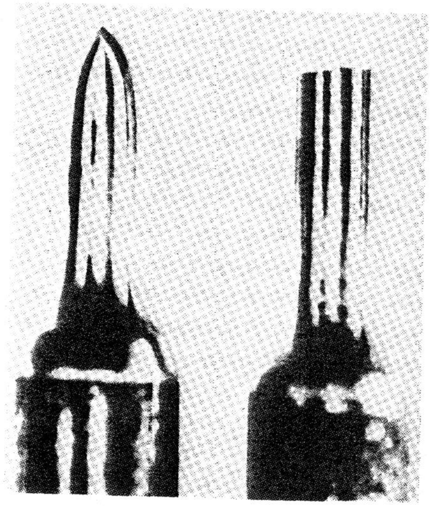


Fig. 2. Extremes of fracture behavior in iron crystals tested at −196°C. (*a*) Chisel edge fracture 100% reduction in area. (*b*) Cleavage fracture. (After Allen, Hopkins, and McLennan.[2])

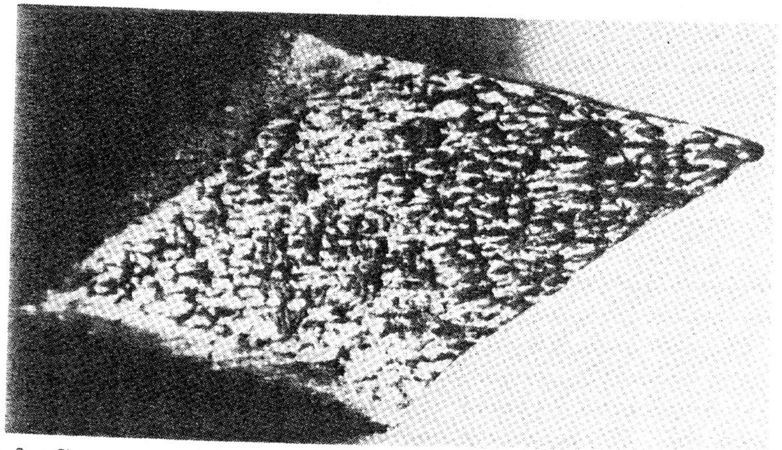


Fig. 3. Copper crystal fractured at 4.2°K. (After Blewitt, Coltman, and Redman.[5])

copper crystals deformed at 4.2°K, although they show some necking, do have a large fracture zone (Fig. 3).

While studying the deformation of alloy crystals, Elam[6] found that solid-solution crystals of aluminum-zinc with low zinc contents give wedge fractures typical of pure aluminum. However, with 10% zinc or more, the fracture becomes planar, inclined at approximately 45° to the tension axis. The fracture plane was shown to be (111). Similarly, Karnop and Sachs[7] found that aluminum-5% copper alloy crystals show plane fractures in the aged conditions, apparently on the operative slip plane.

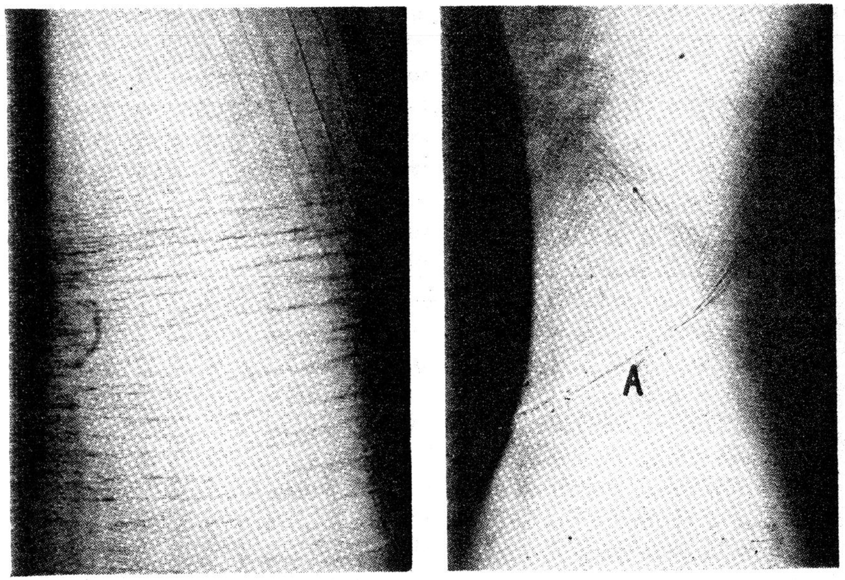
The Role of Necking in Ductile Fracture

It has already been pointed out that ductile fracture of pure metal crystals is usually preceded by the formation of a necked zone. The present authors have examined this zone metallographically by preparing a polished flat surface in the neck during final stages of deformation. In this way, the slip bands operative at this stage can be easily distinguished and studied.

Examination of a large number of aluminum crystals showed that two or more slip systems operate in the necked region. However, in many cases in which the neck is well developed, one system begins to predominate, and a narrow zone of extremely coarse slip bands is formed. These coarse slip bands are illustrated in Figs. 4*a* and *b*; in the latter case, a shear fracture occurred along the zone marked A, in the region of heavy slip. The macroscopic appearance of such a failure is shown in Figs. 5*a*, *b*, and *c*, where the gradual development of the shear failure is occurring parallel to a secondary set of slip bands. Some of the traces of the primary slip planes can be clearly seen. A crack propagates at the last stage of this process, if at all.

Crystals were carefully examined by sectioning in the last stages of deformation to determine whether any voids had occurred. Neither small nor large cavities were found. However, it should be mentioned that large voids have been observed by Orowan[8] in flat aluminum crystals and also in copper; and Greetham[9] has observed a hole in the neck of an aluminum-4% copper crystal which resulted in a crocodile type of fracture (Fig. 6).

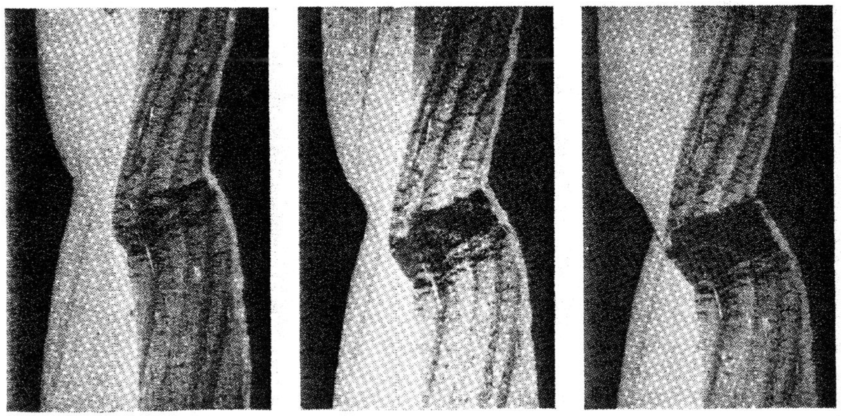
Several copper crystals have been examined during fracture at room temperature. Less neck formation occurs than in aluminum, and propagation of a ductile crack or tear occurs. As before, the first stage in fracture is the formation of a narrow zone of secondary slip (Fig. 7*a*). Deformation continues by shear on a restricted number of slip bands



(a) 1 mm (b)

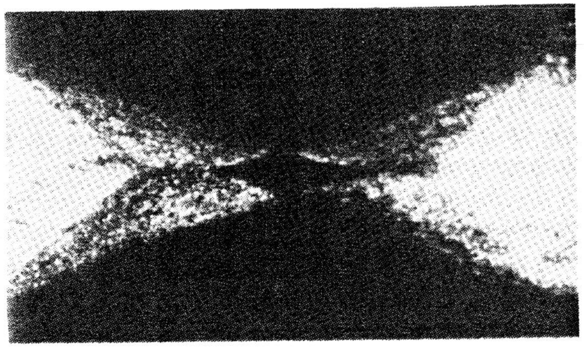
Fig. 4. Coarse slip bands in the necks of aluminum crystals. In (b), the final shear zone A is shown.

10 mm



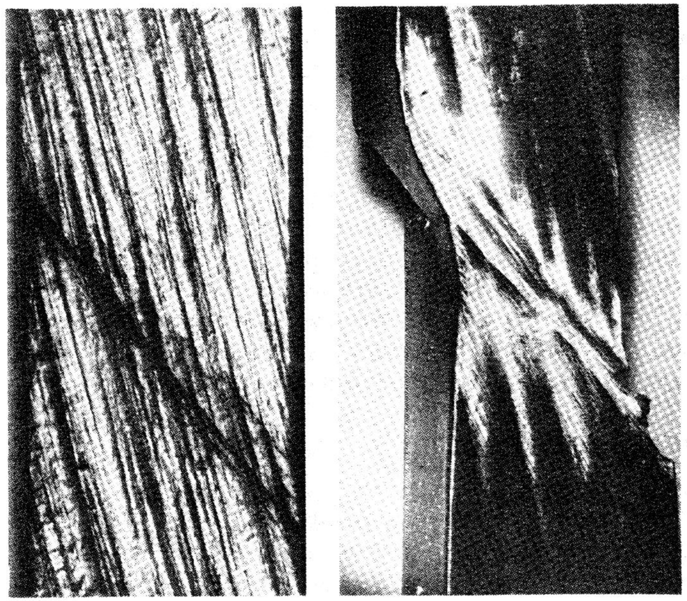
(a) (b) (c)

Fig. 5. Development of a ductile fracture by shear along a secondary slip plane in the neck of an aluminum crystal. Three stages.



1 mm

Fig. 6. Fracture of an aluminum-4% copper crystal showing hole formation.



(a) 10 mm (b)

Fig. 7. Fracture of a copper crystal at room temperature. (a) Early stage of neck formation. (b) Late stage showing crack formation.

that were initially parallel to the secondary slip system. Reduction in the section continues, and a ductile crack gradually develops (Fig. 7*b*).

In the light of such preliminary experiments, it was felt that a proper study of the formation of ductile cracks could be made only in the absence of the necking phenomenon. While slip on two or more systems is necessary for neck formation, the converse is not necessarily true, for substantial uniform deformation of a crystal can result from the simultaneous operation of two slip systems (for example, symmetrical orientations of face-centered cubic metals). Whether or not the localized severe deformation associated with neck formation occurs depends on the rate of work hardening of the material in the necked region. If the increase in stress resulting from the decrease in area is greater than the increase in stress needed to cause further deformation, the necking of the crystal will continue.

Pure metals normally possess high ductility, and fracture usually occurs at a point when the stress-strain curve shows a low rate of strain hardening. As we have seen, necking will occur under these circumstances; however, if the ductility of the metal is so reduced by alloying that fracture takes place at an earlier stage when the rate of strain hardening is still pronounced, then less neck formation should occur. Such behavior would be expected in some highly alloyed metals, such as the aluminum-zinc crystals investigated by Elam [6] and the aluminum-copper alloys (aged) investigated by Karnop and Sachs.[7]

To investigate ductile fracture in a more quantitative way, the present authors chose the aluminum-5% copper alloy because the deformation of similar alloys in various heat-treated conditions had been previously studied by Greetham and Honeycombe.[9] The initial intention was to determine, if possible, the macroscopic stress criterion for the initiation of a ductile crack and to study in detail the slip processes leading to fracture.

Experimental Details

The aluminum-copper alloys were prepared using commercial aluminum, as it was found easier to prepare long single crystals by strain-anneal methods in this material than in superpurity base material.

The work to be described is concerned primarily with two alloys of similar composition:

	Cu	Fe	Si	Al
No. 1	5.30	0.54	0.28	balance
No. 4	5.54	0.55	0.29	balance

Alloy 1 was cast in 1-in.-diam ingots, 10 in. long, which were homogenized at 475°C for seven days. The ingots were then hot-rolled to 0.35 in. round and cold-swaged to 0.214-in. diam. Intermediate annealing at 500°C was necessary. Finally the rods were cold-drawn to 0.125 in. square.

Rods 30 cm in length were annealed 4 min at 530°C and then strained to between 1.25 and 1.75% elongation. They were then slowly heated between 450° and 540°C at a rate of 5° to 10°C per day. Alternatively, the strained rods were passed through a gradient furnace (100°C/cm) at 0.65 cm/hr, the maximum temperature being 10° to 15°C below the solidus. By using either of these methods, crystals up to 30 cm long could be readily produced, but unfortunately, orientations tended to cluster towards the [111] corner of the stereographic triangle.

To eliminate these preferred orientations, Alloy 4 was prepared in slabs 10 in. × 6 in. × 1 in., cross-rolled hot to 0.5 in., and then cold-rolled to 0.125 in. with intermediate anneals at 300°C. This sheet was then cut into square section strips. Crystals grown from this material by the traveling-zone method showed a much wider spread of orientation; these were used for experiments in which a more complete study of the variation of the fracture stress with orientation was made.

In most experiments, the long crystals were cut into several identical specimens, which were electropolished in a perchloric acid-ethyl alcohol bath. To minimize the effects of the grips, gage lengths 2.5 cm long and 0.115 in. square were formed on the crystals by selective electropolishing. Satisfactory gripping of the crystals proved difficult in view of the high fracture stresses often encountered (~80,000 psi). Mounting in solder or aruldite proved inferior to friction grips; the effects of stress concentrations caused by the latter could be largely minimized by the use of electropolished gage lengths.

The testing machine used was a robust one of the Polanyi type designed to give a maximum load somewhat in excess of 1000 lb. The load was measured by deflection of a beam attached to the grips and specimen by a steel rod $\frac{3}{32}$-in. diam. The specimens tended to fracture abruptly at high stresses, so the deflection could not be easily measured by an optical system; however, a sensitive dial gage proved to have the necessary sensitivity and robustness. The machine was calibrated by dead loading to an accuracy of ± 2 lb.

Crystals were tested at three temperatures, 77°K in a liquid nitrogen bath, and 293° and 373°K in a silicone oil bath.

Experimental Results

By carrying out tensile tests on three crystals of identical orientation in the solution-treated, fully aged, and overaged conditions, it was found that necking was eliminated only in the case of the fully aged crystal. Consequently, it was decided to use this heat-treatment. The crystals were solution-treated at 540°C, cooled in a cold air blast, and then aged at 165°C to optimum hardness. For Alloy 1, 33 hr were needed, while for Alloy 4, with a slightly higher copper content, only 16 hr were required.

Stress-Strain Curves

Figures 8*a* and *b* show the nominal stress-strain curves for two fully aged crystals; specimens from these crystals were deformed at three different temperatures. The crystals represent hard (near [111]) and soft (on [110]–[111] boundary) orientations. At room temperature, there is no necking, and fracture occurs with negligible lowering of the stress-strain curve, while surprisingly, at both 77° and 373°K, a slight fall in the stress-strain curve shows that some necking has occurred.

A large number of tests has confirmed this general behavior. At room temperature, fracture occurs normally either on a rising stress-strain curve or a slightly falling one except when orientations are near [100]. Necking is most pronounced at 373°K, where the rate of strain hardening is much less than at the other two temperatures (Figs. 8*a* and *b*).

The fracture stress is quite sensitive to temperature, while the yield stress dependence is not so marked. Previous work by Greetham and Honeycombe [9] has shown that, in the optimum aged condition, the yield stress is relatively insensitive to temperature, but this behavior is markedly altered by changing the aging treatment. The fracture stress may also be very sensitive to the structural condition of the alloy, but this point has not yet been investigated.

The Nature of the Fracture

Crystals tested in the present investigation exhibited coarse slip bands just prior to fracture, and the fracture ran parallel to these (Fig. 9), indicating that fracture had occurred on an operative slip plane. This was normally a (111) plane but not exclusively. The behavior of these aged crystals, as typified by that shown in Fig. 9, gave some hope that the stress at fracture could be determined quite accurately and in conditions approximating uniaxial stressing.

Macroscopic observation of the fracture surface showed it to be rough and quite unlike a typical cleavage face. Carbon films were evaporated on the fracture surfaces and subsequently released by etching. These were then examined in an electron microscope and revealed the uni-

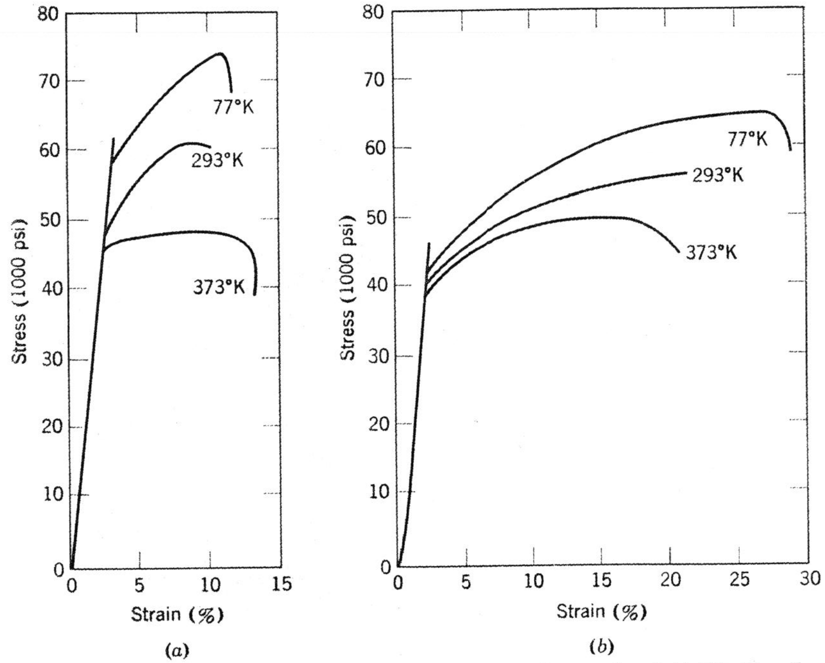


Fig. 8. Nominal stress-strain curves of two fully aged crystals of Al-5% Cu alloy. (*a*) Hard orientation, crystal C 77. (*b*) Soft orientation, crystal C 43. Curves are given for three temperatures.

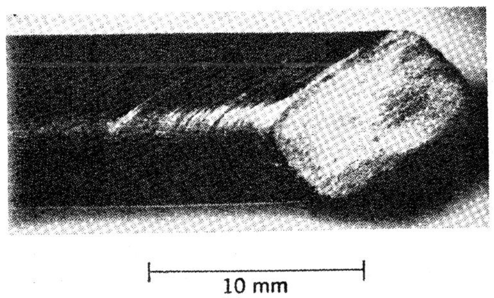


Fig. 9. Fracture of an aged aluminum-5% copper alloy crystal. Slip has taken place on one octahedral system.

directional cusps (Fig. 10) thought to be typical of ductile fracture. For example, electron micrographs of fractures taken by Crussard and co-workers [10] reveal this feature in material failing in a ductile manner.

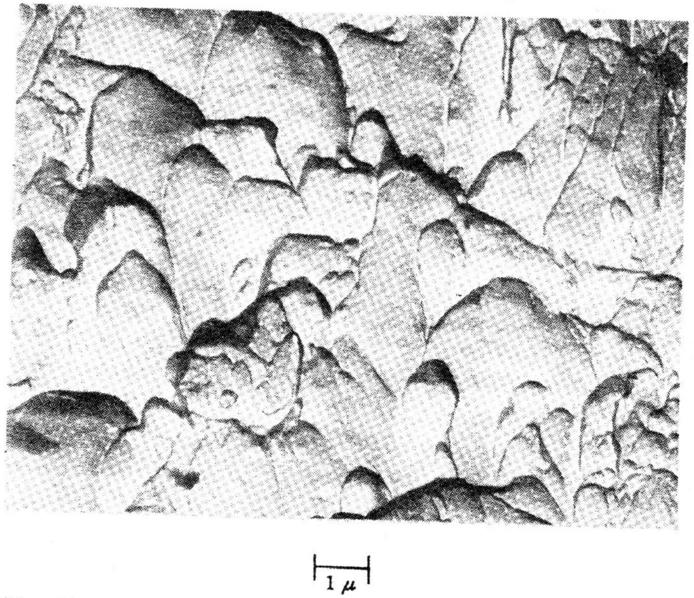


Fig. 10. Typical electron micrograph of a ductile fracture of an aluminum-copper crystal. Shadowed with Au-Pd alloy.

The Stress at Fracture

The fracture stresses were determined by dividing the load at fracture by the final cross-sectional area that was determined from micrometer measurements at the fracture face, the accuracy being $\pm 0.5\%$. It is considered that the errors involved in the measurement of the stresses are less than $\pm 1\%$.

The fracture stresses determined for crystals of Alloy 4 at room temperature are shown in Fig. 11. As might be expected, there is considerable variation with orientation. Values up to 73,600 psi were obtained with crystals of Alloy 1 near the [111] corner. That the fracture stresses are also very sensitive to temperature will be seen from the examples in Table 1. The whole range of orientations in Fig. 11 has not yet been investigated at all three temperatures, but similar temperature sensitivity would be expected.

To determine the normal stress and the resolved shear stress at fracture, it was necessary to determine the orientations of the crystals after

fracture and as near the fracture zone as possible. X-ray Laue photographs had diffuse spots, but the zones could be easily picked out even after 20 to 25% extension. (An error of 1° in the orientation causes an error of $\pm 2\%$ in the determination of the resolved shear stress.) The stresses were calculated using the usual relationships:

$$\tau = \frac{L}{A} \cos \phi \cos \lambda$$

$$N = \frac{L}{A} \cos^2 \phi$$

where L = load at fracture, A = final cross-sectional area, τ = resolved shear stress, N = normal stress, ϕ = angle between specimen axis and the normal to (111) fracture plane, and λ = angle between specimen axis and [110] slip direction on the slip plane. The particular (111) plane was of course normally defined by the fracture and was determined from the angles on two adjacent faces of the crystal, but the slip direction was not easily found experimentally. It was assumed to be the one on which the resolved shear stress was greatest.

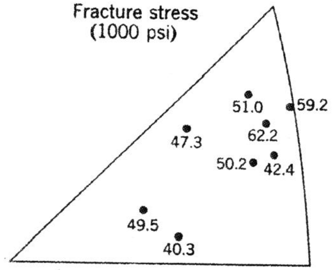


Fig. 11. Fracture stresses at room temperature (293°K) of aluminum-copper crystals of various orientations. (Alloy 4.)

TABLE 1. Fracture Stress at Various Temperatures

	Fracture Stress (psi)		
	77°K	293°K	373°K
Crystal C 20 (near [111])	88,500	73,600	54,500
Crystal C 79	84,900	64,700	53,500
Crystal C 70	—	65,600	50,900

The results were somewhat complicated by the incidence of fracture on (001) planes for orientations very near [111]; however, the complete results for crystals of Alloy 1 are shown in Figs. 12*a* to *c* and 13*a* to *c*. The results in Fig. 12 indicate that the normal stress varies widely, even

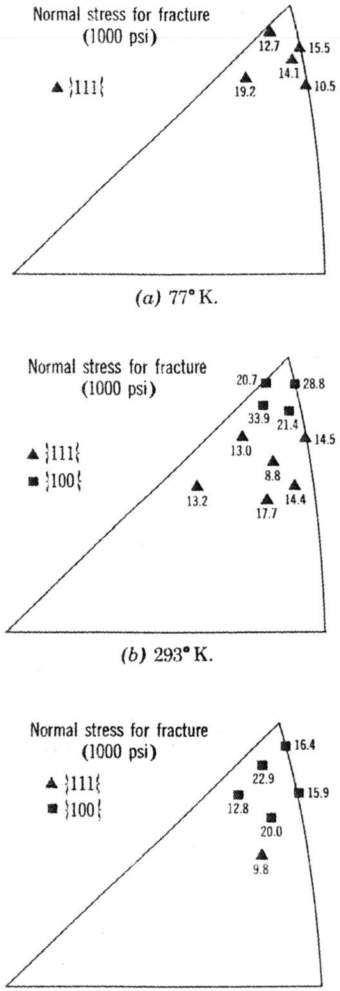


Fig. 12. Dependence of normal stress at fracture on orientation. (Alloy 1). (*a*) 77°K. (*b*) 293°K. (*c*) 373°K.

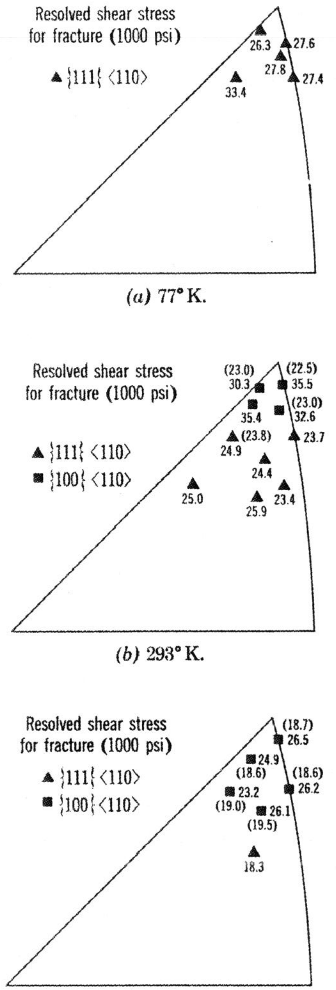


Fig. 13. Dependence of resolved shear stress at fracture on orientation. (Alloy 1). (*a*) 77°K. (*b*) 293°K. (*c*) 373°K.

when only those crystals that exhibit fracture on (111) planes are considered. On the other hand, the resolved shear stress results are much more constant. If we consider Figs. 13*a* and *b*, in which the majority of the fractures are on (111) planes, the resolved shear stresses at fracture are almost constant for a given temperature. At 293°K, the variation is

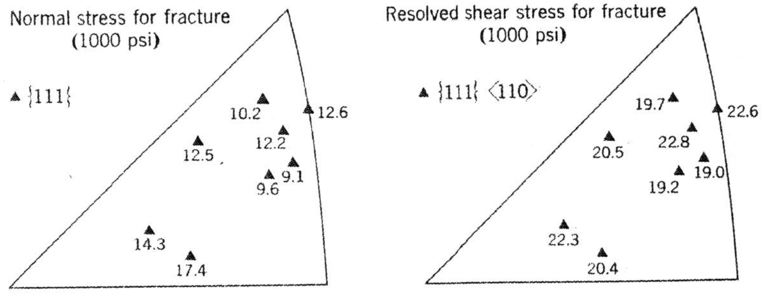


Fig. 14. Dependence of normal stress and resolved shear stress at 293°K on orientation. (*a*) Normal stress. (*b*) Resolved shear stress.

only between 23,400 and 25,900 psi for fractures on octahedral planes; however, it was obvious that more orientations should be studied.

Crystals prepared from Alloy 4 covered a wider range of orientation, and at room temperature all of these crystals broke along (111) planes. The normal stresses and the resolved shear stresses are shown in Figs. 14*a* and *b*. While again there is considerable variation in normal stress, the resolved shear stress varies only between 19,000 and 22,800 psi. These results suggest that the resolved shear stress is a suitable criterion for the onset of ductile fracture.

Metallography of Deformation and Fracture

In these investigations, the slip bands in most crystals became distinct only in the later stages of deformation. The distribution depended on the orientation; for example, crystals in the center of the stereographic triangle showed slip bands throughout the gage length, whereas hard orientations near [111] had visible slip bands only in the immediate vicinity of the fracture. All crystals resembled pure metal crystals in so far as fracture was preceded by very heavy slip-band formation in the fracture region.

Whether single or multiple slip occurred depended on the crystal orientation and the amount of rotation during deformation (Fig. 15). Some crystals (C 11, 43, and 70) deformed on only one octahedral plane, and fracture eventually took place on this plane before the specimen axis reached the [111]–[001] boundary. On the other hand, some crystals (C 79 and 58) exhibited octahedral slip on two systems when the symmetry boundary was reached. The duplex slip was normally restricted to the region near the fracture, and the fracture still occurred

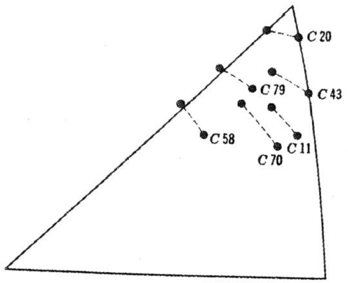


Fig. 15. Rotation of crystal axes during deformation to fracture at 293°K.

parallel to one of the operative planes, usually the secondary system (Fig. 16).

It has already been mentioned that as the orientations of the crystals approached the [111] pole, cubic slip made its appearance during the later stages of the deformation in the zone near the fracture. Figure 17 is a micrograph of such a crystal, showing two faint sets of octahedral slip bands and heavy horizontal traces which represent cubic slip. The crystal fractured parallel to these bands. In some crystals (for example, C 20), octahedral slip could not be detected, cubic slip occurring in a number of coarse bands in the immediate vicinity of the fracture, which ran closely parallel to them (Fig. 18).

The occurrence of cubic slip and fracture was markedly orientation and temperature dependent. In Fig. 13, the crystals that fractured on cubic planes are represented by square points. This does not necessarily define the incidence of cubic slip, but, when cubic slip was observed, fracture normally occurred parallel to it. No cubic fractures were observed at 77°K, while at 293°K they were restricted to the immediate vicinity of the [111] corner; at 373°K, the incidence of cubic fracture was more widespread. A fuller orientation survey at this temperature will probably reveal that octahedral fractures are more extensive.

An interesting feature of these results is that if the fracture stress is resolved on the cubic fracture plane the stresses obtained are much higher than those obtainable for crystals where octahedral fracture has taken place (Figs. 13*b* and *c*). Nevertheless, they are fairly constant for the crystal orientations available. However, if the stress at fracture is resolved on the most favored octahedral plane regardless of the fracture plane, then the values of this stress (indicated in brackets in Fig. 13) are

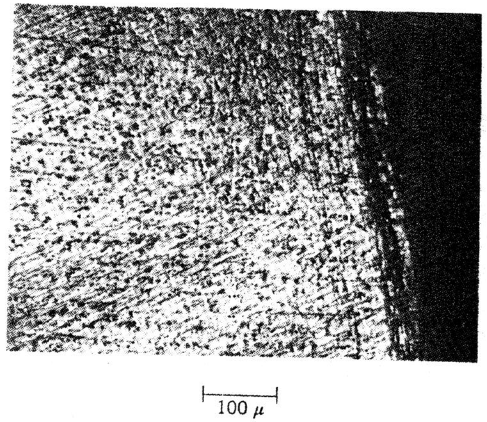


Fig. 16. Micrograph of fracture zone of crystal C 58 deformed at 293°K.

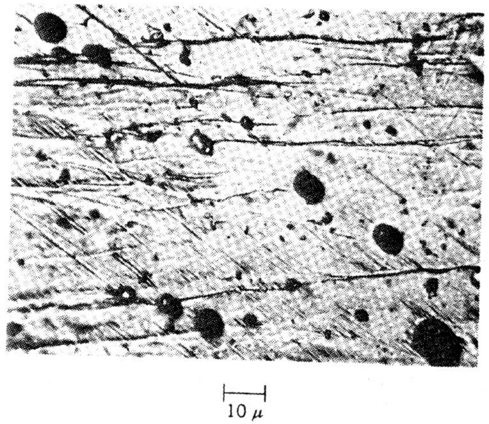


Fig. 17. Alloy crystal deformed at 293°K showing one cubic and two octahedral slip systems.

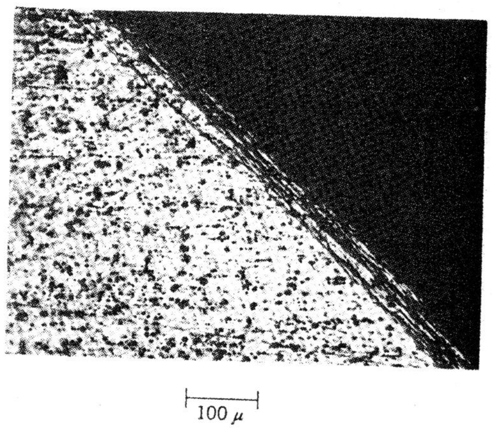


Fig. 18. Alloy crystal C 20 deformed at 293°K. Micrograph of fracture zone showing only cubic slip.

very similar to those obtained for crystals in which octahedral fracture occurs.

Discussion

The simplest mode of ductile failure appears to be that resulting from excessive shear along slip bands at high temperatures. The slip bands are very coarse because obstacles to dislocation movement are slight, and any tendency to form dislocation pile-ups would be minimized by climb. In these circumstances, a few dislocation sources will operate continuously, the large steps produced on the surface of the crystal acting as stress concentrators to keep the shear localized to the favored slip regions.

Deformation and fracture at room temperature is a more complicated process because, even in hexagonal metals, duplex slip takes place, and neck formation usually precedes fracture. However, even in these circumstances, a crack need not form, and the necked region can draw down to a chisel edge or a point by duplex or multiple slip. This behavior is typical of very ductile metals, which do not work harden rapidly, and, while dislocation pile-ups would be expected in the necked zone, the local stress concentrations so caused are not great enough to form a crack nucleus by destroying atomic cohesion. The high stress that builds up as the neck forms makes possible a breaking through of the barriers to slip (Lomer-Cottrell locks) so that further duplex slip occurs until the cross-sectional area approaches zero.

Alternatively, after the neck has formed by duplex slip, further slip predominates on one system probably because the changing stress system in the neck produces a higher resolved shear stress on this plane. This leads to the coarse bands observed both on aluminum and copper single crystals just prior to failure. These correspond to the coarse bands observed during the high-temperature rupture of hexagonal metals, except that they are much less straight because much cross slip is taking place (Fig. 17). Sufficient dislocation locking can often occur at this stage to induce the formation of a ductile crack, which, once formed, readily propagates because of the favorable stress conditions of the neck. Preliminary experiments suggest that ductile cracks form more readily in copper crystals than in aluminum crystals during deformation at room temperature. Copper work hardens more so than aluminum in such circumstances, and greater stress concentrations would thus be expected in copper at dislocation pile-ups in the necked regions.

An attempt has been made to simplify the problem by eliminating the neck formation. It should be emphasized that indiscriminate alloying does not have this effect; solid-solution alloys frequently do not work harden rapidly, and so neck formation occurs; for example, 70–30 α-brass crystals behave similarly to copper and aluminum-5% copper crystals, which neck when in the as-quenched condition. It is necessary to age these alloys to obtain a fine precipitate, thus not only raising the yield stress but raising the subsequent rate of work hardening, whereupon neck formation does not take place. Alternatively, with pure metals the temperature of deformation can be lowered, but work with copper [5] suggests that very low temperatures are needed, and in these circumstances the problem is complicated by the occurrence of twinning and discontinuous slip.

The quantitative results on aluminum-copper crystals show clearly that, for orientations over most of the stereographic triangle, the resolved shear stress on the fracture plane at fracture is constant within limits that are reasonable for single-crystal experiments. This could be explained by assuming that, when the dislocation locking is made adequately effective by use of precipitate particles as well as dislocation interactions, the pile-ups build up to a level at which the local stress is sufficient to cause a small crack rather than further glide. Fracture is always preceded by marked glide on a few slip bands, and the fact that the fracture follows the same path suggests that dislocation pile-ups large enough to initiate a crack occur only at this stage. This view is supported by the fact that the final slip bands can be revealed again after repolishing and etching (note the persistent slip bands produced during fatigue). The appearance of the fracture surfaces in the electron microscope shows that the fracture path is not along one plane but that it deviates frequently, producing the characteristic facets that are revealed as steps by shadowing and by using stereomicrography.

The change in the fracture plane from (111) to (001) as the [111] axis is approached and as the temperature of deformation is raised is an interesting phenomenon. Cubic slip is thus most likely to occur in crystals whose orientation would normally lead to slip on two or three octahedral planes. Face-centered cubic metal crystals with axes near the [111] pole strain harden most rapidly because of the ease with which Lomer-Cottrell locks can occur at an early stage of the deformation. In the highly alloyed crystals used in this investigation, the locking is probably so efficient that slip is forced on the alternative (001) plane, although (001) dislocations would normally be expected to dissociate into two partial (111) dislocations. The high degree of alloying may prevent this dissociation, although it may occur locally on a coarse band that is macroscopically parallel to (001).

The resolved shear stresses on those cube planes favorable for crystals near [111] are much higher than those on the most favorably oriented (111) planes, as is shown in Figs. 13*b* and *c*. Therefore, if it is assumed that the appropriate octahedral planes are locked, the cubic plane will be in a favored position. The fact that the shear stresses for crystals with cubic fractures, when resolved on the most favorable octahedral plane, are close to the values of those for crystals fracturing on octahedral planes suggests that crack initiation occurs at critical pile-ups on octahedral planes. However, with the near [111] orientations, the cracks propagate on a (001) plane because of the favorable macroscopic shear stress along this plane.

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DISCUSSION

W. D. Sylwestrowicz, *Bell Telephone Laboratories, Inc.* All three chapters in this volume dealing with ductile fracture are mainly concerned with the later stage of the phenomenon of fracture: the mechanism of the formation of cracks, their propagation, and the final break of the specimen. Only a small amount of information is available on the criteria leading to instability and consequent necking of a specimen. In the preliminary investigation described below, some observations have been made which appear to be pertinent to this problem.

It has been observed [1] that under certain conditions aluminum and

copper specimens strained at room temperature, after prestraining in liquid nitrogen, fracture after a considerably smaller amount of plastic deformation than if they were strained continuously at either room temperature or in liquid nitrogen. It was thought that a more detailed investigation of this phenomenon would give additional information on the mechanisms of fracture of metals and, in particular, on the mechanisms of necking.

In these experiments, high-purity (99.99% or better) copper and aluminum specimens were used. The polycrystalline copper specimens, 0.05-in. diam, were annealed at 260°C for $\frac{1}{2}$ hr. Two groups of aluminum specimens were used: those with a coarse-grained structure (3 to 4 grains across the section) produced by annealing at 600°C for 8 hr, and those of single crystals. All aluminum specimens were 0.1 in. in diam. The aluminum specimens were electropolished before testing. After deformation, the appearance of the slip bands was observed under the microscope and their crystallographic planes determined by X rays.

The following observations were made from these preliminary tests:

(*a*) The formation of the neck in a specimen is not related to the amount of strain. It is possible to start instability in the specimen over a large range of strain depending on the previous thermal history, as shown in Figs. D.1, D.2, and D.3. This observation applies to polycrystalline copper (Fig. D.1) and aluminum (Fig. D.2), as well as to single crystals of aluminum (Fig. D.3).

(*b*) It appears that at a given temperature the initiation of instability in a specimen is related to a specific configuration of the deformed structure. The value of the yield stress of this deformed structure is independent of previous changes of temperature during the test and is thus equal to the yield stress at which necking occurs when the whole test is conducted at this temperature. If, after deformation of the specimen at liquid nitrogen temperature, the yield stress at room temperature (curve 3 of Fig. D.1) is still below the value of the yield stress of necking that would be obtained if prior low-temperature deformation had not taken place, the specimen can be strained further until the yield stress reaches this value; on the other hand, if after deformation in liquid nitrogen the yield stress at room temperature is equal to or greater than the value obtained without prior low-temperature deformation of the specimen, instability is initiated immediately, and fracture occurs (curves 4 and 5). Since the ratio of the yield stress at room temperature δ_R to that at liquid nitrogen δ_N is constant and independent of strain,[1,2] then δ_R, after a given deformation of the specimen in liquid nitrogen, is determined by the value of δ_N alone and can be greater than the value of the yield stress of necking at room temperature (curve 5).

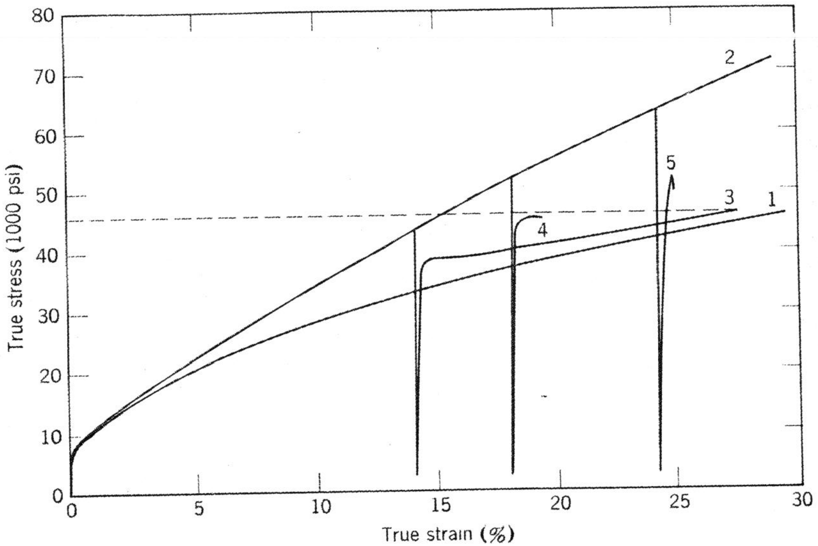

Fig. D.1. True stress — true strain curves for polycrystalline copper specimens, strained at 297°K after prestraining at 76°K. Curve No. 1 at 297°K; curve No. 2 at 76°K.

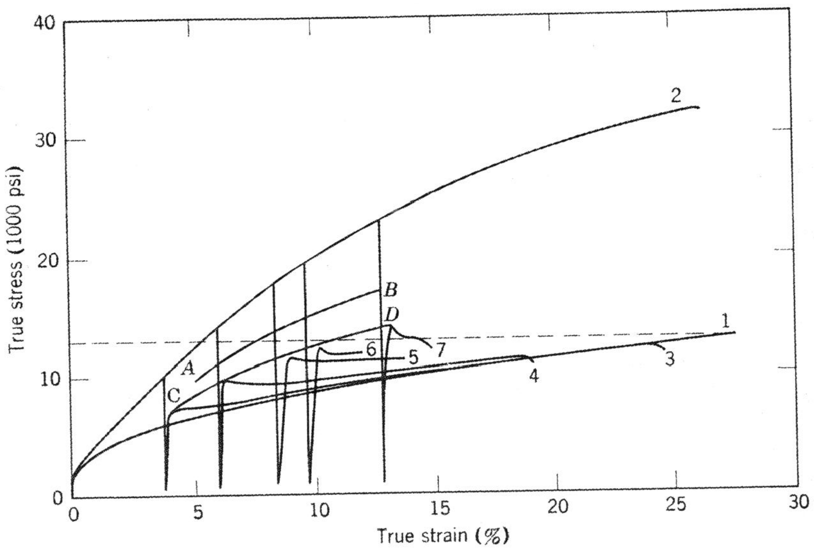

Fig. D.2. True stress — true strain curves for coarse-grained aluminum specimens, strained at 297°K after prestraining at 76°K. Curve No. 1 at 297°K; curve No. 2 at 76°K; and curve A–B, locus of the yield points of specimens strained at 297°K after prestraining at 76°K if there was no recovery.

(c) Experiments with high-purity aluminum specimens show another interesting feature. During heating of a specimen from liquid nitrogen temperature to room temperature, the drop of the yield stress is caused not only by the difference in temperature but also by recovery. In Fig. D.2, for specimens with a coarse-grained structure, curve A–B repre-

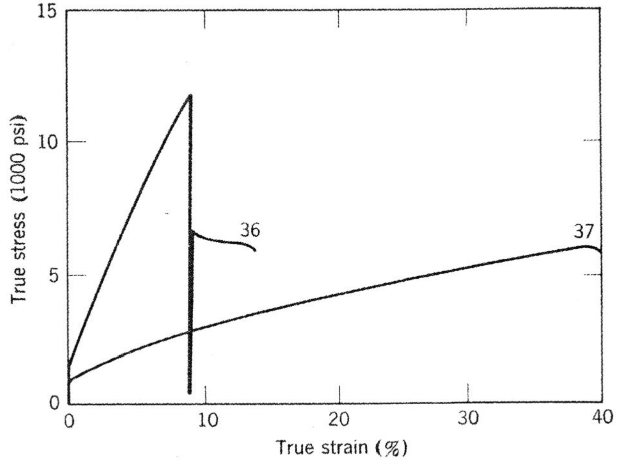

Fig. D.3. True stress — true strain curve for single crystals of aluminum. Single crystal No. 36 strained at 297°K after prestraining at 76°K. Single crystal No. 37 strained at 297°K.

sents the locus of the yield stresses after the specimens were heated from liquid nitrogen to room temperature, if there was no recovery. This curve is calculated from the ratio δ_R/δ_N, obtained from the test in which the specimen was cooled from room temperature to liquid nitrogen. Curve 6 in Fig. D.2 shows a specimen that failed after a small additional straining at room temperature, even though its yield stress was slightly below the yield stress of necking. However, it can be seen from the curve A–B that the yield stress at room temperature, if recovery had not occurred, is above the yield stress of necking.

This observation indicates that the yield stress in itself is not a principal factor in the initiation of the neck but appears to characterize only a state of plastic deformation, for example, a specific configuration of slip bands. Experiments with single crystals of aluminum (Fig. D.3) agree with the above observations.

Slip-band patterns on the surface of a fractured single-crystal specimen when combined with X-ray Laue photographs indicate that, in liquid nitrogen, slip takes place on {111} planes. When the test temperature is changed from liquid nitrogen to room temperature, a new cross slip appears, but only in the necked area. This new cross slip may be responsible for the formation of the neck. In contrast, a single crystal strained at room temperature to fracture shows traces of {111} slip planes. These surface traces are visible throughout the whole length of specimen, with the exception of the neck area, in which slip bands in one set of the {111} slip planes increase in density. A sample of coarse-grained aluminum shows a somewhat similar pattern of deformation. It should be mentioned here also that, while polycrystalline copper specimens fracture in "cup-and-cone" fashion, single crystals of aluminum neck to a knife edge.

The discusser wishes to express his thanks to Dr. W. C. Ellis and Dr. D. F. Gibbons for their valuable discussions.

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H. C. Rogers, *General Electric Company*. From the appearance of the ductile-fracture surface of an Al-Cu single crystal (authors' Fig. 10), it is obvious that the fracture takes place by the "void-sheet" mechanism.[1,2] The essence of this type of failure is the localization of the shear deformation. Because of the difference in the deformation characteristics of single and polycrystals, there is a difference in the gross appearance of the fracture, even though the basic fracture mechanism is identical. Polycrystals, by the very nature of the internal constraints imposed on them by variation in grain orientation and geometry, act much more like the homogeneous medium used by plasticians to study mechanical behavior, being much more sensitive to the geometry and the mechanics of the test conditions than are the single crystals in which crystal orientation predominates. In polycrystals, the deformation takes place in the region of maximum distortion energy, its direction and extent being determined by the crack size and shape, the shape of the neck, elastic energy of the system, and other mechanical variables, in addition to the metallurgical properties such as orientation, rate of strain hardening, and strength. On the other hand, in the case of the age-hardened alloy single crystals that the authors have studied, the reason for the localization of shear strain to a narrow region parallel to an operative slip system lies primarily in the deformation process and the effect of the metal properties on it. Whether the slip planes play any part in the fracture, other than that mentioned above, has not been resolved either by the authors or by the present writer.

It should be noted that the "crocodile type" of fracture observed by Greetham and referred to by the authors is the monocrystalline version of the polycrystalline "double-cup" fracture described previously.[1]

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C. J. Beevers and R. W. K. Honeycombe (Authors' Reply). We do not share the view of Rogers that fracture in our alloy crystals takes place by the "void-sheet" mechanism. No voids were observed to form prior to fracture. Dislocation pile-ups on the slip planes appear to play a predominant role in the fracture process, a view which is supported both by the metallographic evidence and the resolved shear stress criterion at fracture.

Fracture of polycrystalline specimens is much more difficult to interpret. It would appear that, apart from the role of inclusions in forming cavities, dislocation pile-ups at boundaries may also create crack nuclei in the form of small cavities.