According to the theory of Dienes and Damask[6] the temperature independent diffusion effected by radiation corresponds to a linear annealing mechanism for the additional defects. These authors find under steady-state conditions for the diffusion coefficient due to radiation the proportionality

$$D' \approx \frac{k}{\alpha},$$

where thermal diffusion already has been neglected. The quantity k is the constant defect production rate caused by radiation, and α the proportionality constant for the annealing process. With the estimated value for k and the measured quantity D' α would be about 10^{10} cm/cm^3. In the model of Dienes and Damask α corresponds approximately to the dislocation density. Then the high value of 10^{10} cm/cm^3 for this quantity is improbable for the well annealed copper specimen. The assumed theory for the description of the annealing mechanism therefore seems to be insufficient for the interpretation of these experiments.

A similar behavior was also observed for the self-diffusion in lead subjected to α-radiation.[16]

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Slip in twinned copper crystals*

Since the time Blewitt *et al.*[1] first reported deformation twinning in copper single crystals, other workers have also reported deformation twinning in other f.c.c. metals.[2,3] However slip in twinned crystals, which is of interest in the study of ductile fracture in twinned crystals or in grains of polycrystalline metal with annealing twins, has not been considered. The purpose of this letter is to present a simple geometrical analysis of deformation in a twinned structure.

During the testing of crystals with the tensile axis parallel to $[\bar{1}11]$ bounded by faces parallel to (110) and $(1\bar{1}2)$, deformation twinning was observed at 78°, 20° and 4.2°K.[4] Twinning does not convert the whole crystal to the twin orientation, rather a laminated structure of alternating twinned and untwinned matrix results.

Deformation after Lüders-band type propagation through the gauge section was observed to proceed by slip on systems which did not violate strain compatibility across the twin boundary. Figure 1 shows two tetrahedra bounded by $\{111\}$ planes and $\langle 110\rangle$ directions, one on the left corresponds to the original matrix with the tensile axis parallel to $[\bar{1}11]$ and one on the right, the twinned orientation. Twinning in the $(\bar{1}\bar{1}1)$ $[\bar{1}21]$ as indicated results in an orientation which has the tensile axis parallel to a $\{110\}$ 19° 28′ toward $\langle 001\rangle$ from $\langle \bar{1}12\rangle$. The $\{111\}$ planes in the twinned orientation are labelled T_1, T_2, T_3 and T_4; the twinning plane is that plane common to both orientations and

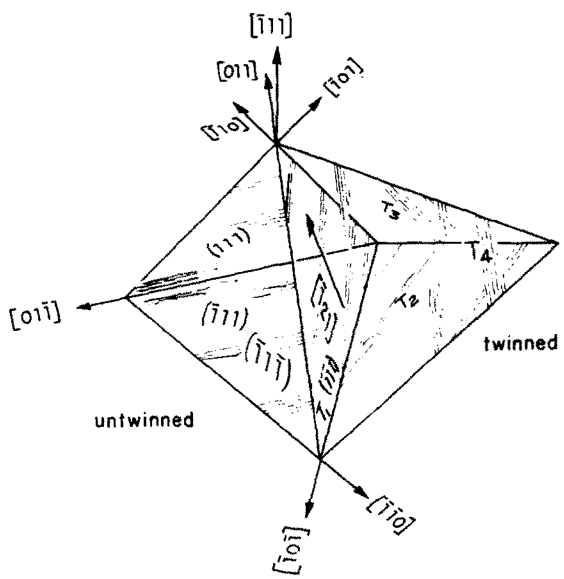

Fig. 1. Two tetrahedra with a common $\{111\}$ (twinning) plane; $(\bar{1}\bar{1}1)$ $[\bar{1}21]$, the twinning system. That on the left represents the original, untwinned orientation, that on the right the twinned orientation. T_1, T_2, T_3 and T_4, the octahedral slip planes in the twinned orientation. Tensile axis parallel to $[\bar{1}11]$.

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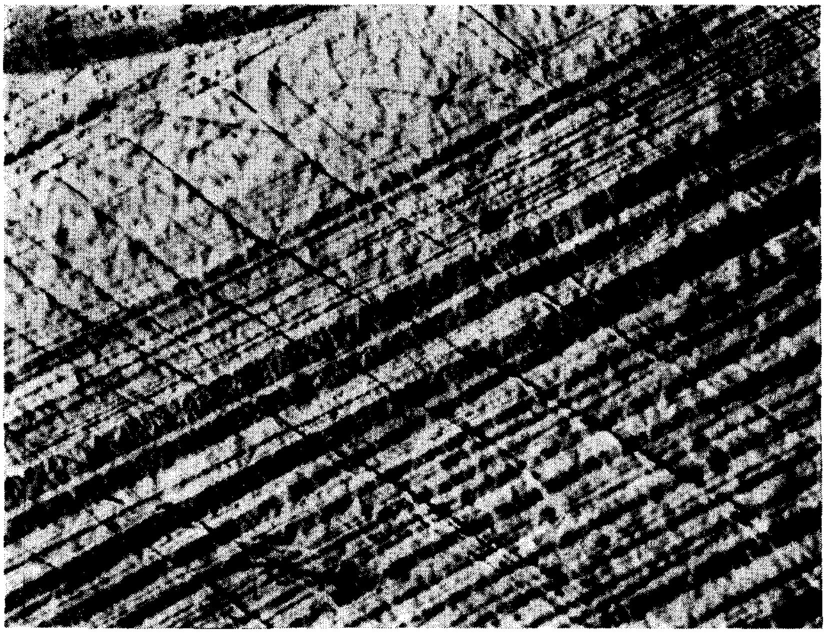
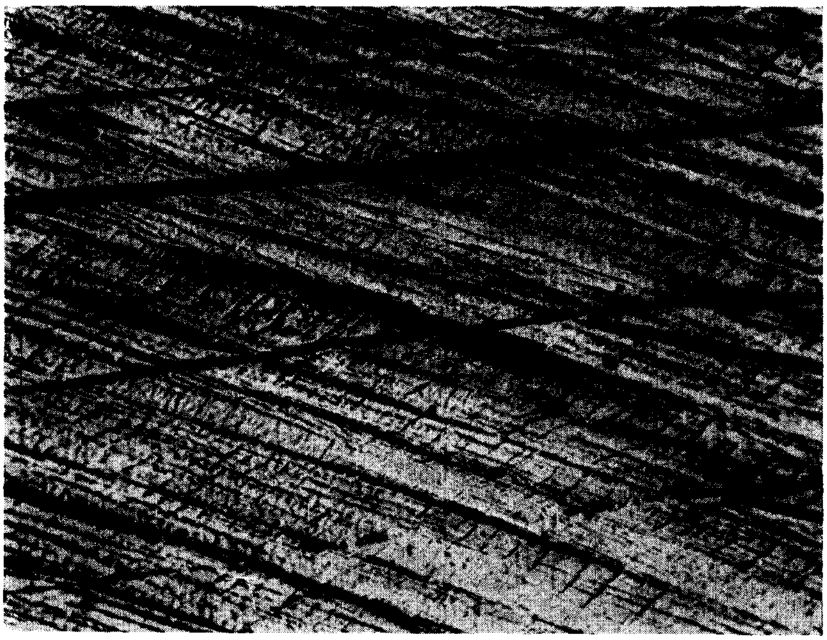


FIG. 2. (a) Slip through twinned [$\bar{1}$11] crystal. Twin lamellae are the white traces; tensile axis horizontal, (110) surface, × 500.



(b) Laminated structure of twinned and untwinned matrix. Traces of twin plane run diagonally down from left to right; tensile axis horizontal, (110) surface, × 250.

parallel to $(\bar{1}\bar{1}1)$ and T_1. From Fig. 1 it is evident that slip can take place on T_4 [011], T_3 $[\bar{1}10]$ and on T_1 in [011] and $[\bar{1}10]$ without violating strain compatibility. Furthermore T_4 [011] can criss-slip onto $(\bar{1}1\bar{1})$ and T_3 $[\bar{1}10]$ onto (111). It was possible to determine from the angular deflection as the slip trace crossed the twin boundary which {111} planes were continuous†. In every case it was found to be either T_4-$(\bar{1}1\bar{1})$ or T_3-(111).

Figure 2(a) clearly shows the continuity of the slip traces as slip proceeds across a twinned region. In highly twinned regions, short transverse traces become more predominant (Fig. 2(b)) and near the fracture these traces join up. Slipping off occurs by localized shear in a zone parallel to these traces.

Upon twinning the Schmid factor for the systems T_4 [011] and T_3 $[\bar{1}10]$ becomes 0.455 compared to 0.272 for all active slip systems in the original orientation. Hence should a shear stress concentration build up at the twin plane, Livingston and Chalmers' analysis(5) of computing resolved shear stresses on slip systems across a grain boundary shows that the most highly stressed systems in the original matrix are the $(\bar{1}1\bar{1})$ [011] and (111) $[\bar{1}10]$. These systems are precisely those which are favored from strain compatibility conditions.

From the symmetry of $(\bar{1}1\bar{1})$ [011] — T_4 [011] and (111) $[\bar{1}10]$ — T_3 [011] combinations with respect to the tensile axis, for a crystal which upon twinning still remains geometrically uniform, double shear is expected. A crystal displaying just this has been observed.(6) For twinned crystals originally oriented for single slip, one of the above combinations has a larger resolved shear stress and hence planar shear would be expected. The shear fracture in twinned crystals observed by Blewitt *et al.*(7) could be explained in this way.

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† This determination was facilitated by the fact that the lattice rotation in $[\bar{1}11]$ crystals was small making the surface analysis simple. Furthermore, the measurements were made on the thick end of slightly tapered crystals where the Lüders front had terminated.

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A note on the creep substructure of pure copper*

The purpose of this note is to report some observations made on the dislocation substructure present during steady state creep of 99.995% Cu. Creep samples were tested in a dry deoxidized hydrogen atmosphere under condition of constant stress on a creep unit equipped with a sliding furnace. The furnace assembly was such that following testing the entire furnace was lowered and the specimen cooled rapidly under load. The cooling rate above 100°C was approximately 300°C per min. Specimens were maintained in the hydrogen atmosphere during the entire cooling cycle. It was hoped that the rapid cool while under load would preserve the existing creep substructure.

Following testing, samples were chemically polished to a thickness of ~0.05 mm (initial thickness ~1.25 mm) using a solution of 50 ml HNO_3, 25 ml H_3PO_4 and 25 ml glacial acetic acid. Thin foils were then prepared by electro polishing in a solution of 67 ml methyl alcohol and 33 ml HNO_3 maintained at −30°C. All transmission microscopy was preformed with a Hitachi Hu 11 electron microscope operating at 100 kV.

All samples were fully recrystallized at 700°C prior to testing. This annealing treatment resulted in an average grain diameter of 0.22 mm. The as-annealed structure was free from substructure apart from a few randomly spaced dislocations.

A typical creep substructure for a sample strained 6.1% at 496°C and 2.07×10^8 dyn/cm^2 is shown in Fig. 1. The steady state creep rate for this sample was 3.6×10^{-7} sec^{-1}. The subgrain boundaries are generally well defined two-dimensional dislocation networks and do not resemble the tangled networks that characterise the low temperature deformation substructure of copper.(1) Rather the structure is more typical of a cold worked and recovered substructure, as might be expected from the high temperature creep conditions. The average subgrain diameter is 3.3 μ. Two examples