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A HIGH VOLTAGE ELECTRON MICROSCOPY STUDY OF SLIP BAND GROWTH IN COPPER

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Introduction

Slip tends to concentrate into clearly defined slip bands in pure FCC metal single crystals after prestraining, quench-hardening, or radiation hardening (1,2). On a macroscopic scale the deformation is often in the form of a Lüders band which spreads progressively along the length of the crystal. The stress strain curves have an initial stage of deformation during which little hardening occurs. For radiation hardened and quench-hardened crystals it is clear that the dislocation loop substructure created by condensation of vacancies and/or interstitials is unstable in the presence of long dislocations (3,4). Clear channels are created by expansion of the first few glide loops (5) resulting in strain softening in the layers of crystal immediately adjacent. Therefore, deformation proceeds by growth of slip bands within which the local shear strain can be very large. Within the volume where a slip band has formed, the initially high dislocation density in the form of isolated closed loops has been replaced by more stable dislocation tangles. The present electron microscope observations were undertaken to see if a similar explanation can explain heavy slip band formation in copper which has been prestrained in multiple slip. The experiments are similar to those previously reported by Sharp and Makin (6) but are more extensive and thicker foils could be examined at 500 kV.

Experimental Procedure and Results

Large single crystals (3.5×1×15 cm) were prestrained in tension along [111] to a shear stress of 2.5 kg/mm^2. From these were cut smaller crystals which were given a second deformation by compression in a single slip orientation in which the active slip plane was the (111) plane, which was inactive during the prestrain. The geometry of the crystals was the same as was used by Murty and Washburn (1). After the second deformation thin foils were obtained by acid sawing, followed by chemical thinning and jet electropolishing. Foils were examined using a 650 kV Hitachi electron microscope, operated at 500 kV to avoid radiation damage, and also in a 125 kV Hitachi electron microscope. The active slip system of the second deformation was defined so that the primary Burgers vector was along [$10\bar{1}$] and the active glide plane was (111). The dislocation arrangement after the second deformation was studied using ($1\bar{2}1$) and ($10\bar{1}$) foils (perpendicular to the primary glide plane) and (hkl) foils, in which both [$10\bar{1}$] and the primary glide plane were at 45° to the plane of the foil. In sections perpendicular to the primary glide plane traces of the well-defined slip bands which occur in such crystals were a very prominent feature (Fig. 2,3,5,6). Through the observation of several hundred micrographs it is possible to fairly accurately describe a typical heavy slip band as follows: The slipped volume was elongated in the edge direction, being from 10 to 100μ in length along [$10\bar{1}$] and about half this wide along [$1\bar{2}1$]. The dimensions and relative spacings of the slip traces agreed well with previous surface replica studies of slip bands (1). Dense mats of dislocations, lying approximately parallel to the primarily slip plane, often outline a slip band over parts of its length. Dislocations of all Burgers vectors are present in the mats, with the three Burgers vectors of the cross-slip plane accounting for somewhat more than half the dislocations in the mats.

Mats frequently occur in pairs. The volume of crystal between paired mats was usually rotated by a small amount (less than 1^{o}) with respect to the crystal above and below the mats. The axis of this rotation varied, having in general a component along [111] and a component along [$10\bar{1}$]. The amount and sense of the rotation also varied slightly along the length of a given pair of mats. The mats were also sometimes seen in groups of three. In parts of the slip band where dense mats did not occur, slip appeared to spread over a distance as great as 2μ normal to the primary glide plane (Fig. 2). In some areas of (hkl) and [$10\bar{1}$] foils there was a notable tendancy for tangles and individual dislocations to lie on the cross-slip plane (Figs. 4,5). No significant differences were found between foils taken within 100μ of the surface and foils taken from the interior of the crystals, either in the (hkl) or the ($1\bar{2}1$) foils.

Discussion

Prior to the second deformation in a single slip orientation the crystals contained the typical multiple slip substructure. Regions of relatively perfect crystal one to several microns in size were roughly separated by dislocation tangles of varying density and thickness (Fig.1). There was very little misorientation between neighboring cells. Therefore the dislocation tangles consisted mostly of paired dislocations in the sense that there was very little excess of one sign for any Burgers vector. This makes the tangles similar to the dislocation loop substructure of a radiation hardened crystal since both have a high dislocation density, with the net Burgers vector taken over a large circuit being small. Even for relatively thick foils of the order of 1μ, cells seldom appeared to be completely surrounded by dense tangles on all sides. Therefore for many glide layers it is likely that an expanding dislocation loop can propagate into most of the cells without cutting through the dense thick parts of the tangles. The following is a model for the formation of a slip band which is consistent with all the observations: The clustering of slip into bands and the formation of new bands, mostly in the regions of stress concentration surrounding a previously formed slip band (1), suggest that good sources are rare. Initiation of a slip band probably consists of the bowing out of a dislocation segment near the edge of one of the cell wall tangels. It must have the primary Burgers vector and the right sign to move away from the dense parts of the cell wall under the action of the applied stress. However, it does not seem likely that bowing out of a single segment of primary dislocation so that it expands over the area of one cell always results in growth of a slip band. More stringent requirements must apply to a source. We propose that it is also necessary that appropriate local conditions exist for multiplication by the double cross-slip mechanism. Intersecting dislocations in the cross-slip plane having any of the three Burgers vectors that lie in that plane promote easy cross-slip and dislocations on the primary plane with any of the three Burgers vectors in that plane provide easy access back into the primary plane (7). A group of primary dislocations on a set of nearby parallel glide layers is probably required for propagation of the slip band over areas which are large compared to the cell size. A group of primary dislocations of the same sign coming against the less dense parts of a cell wall tangle is more likely to promote annihilation and rearrangements of the dislocations in the tangle than is a single primary dislocation. It is likely that the critical number of dislocations in such a group increases with increasing prestrain. These groups of primary dislocations will be able to pass into neighboring cells through the less dense parts of the cell walls, but will be stopped by the most dense cell wall tangles. In areas where groups of primary dislocations are held up there will be stress concentrations which can cause any nearby sources of primary or secondary dislocations to operate (8). The attractive interactions between the secondary dislocations and the primary dislocations leaves a dense mat of dislocations along the path traveled by the group, thereby eventually stopping multiplication of primary dislocations within the slip band and immobilizing the groups of primary dislocations. At this stage a stress concentration exists all around the perimeter of the slipped region which increases the probability of nucleating a new slip band on a glide layer not too much above or below the just formed one. The perimeter of a slipped region is probably very irregular and the new band is likely to overlap the previously formed one in many places. This, we propose, is the origin of the frequently observed pairing of dislocation mats. It also explains the slight rotations of the material between mat pairs because the primary dislocation groups which produced the two mats of a pair will have had opposite Burgers vectors. Therefore the excess of one sign present in the two mats for both primary and secondary dislocations would be expected to be of opposite sign. The variability of the axis of rotation of the material between mat pairs should also be expected because different combinations of secondary sources will come into operation along the path of a group of primary dislocations. After extensive deformation in the single slip orientation the initially irregularly shaped cells are completely replaced by dislocation mats parallel to the primary plane and roughly parallel to the cross slip plane (Fig. 5). However, it is difficult in the case of an individual slip band to get clear evidence for the annihilation and rearrangement that takes place in the original cell structure because the observations cannot be made during the formation of a band. Unlike the case of radiation hardened crystals where loops are replaced by dislocation tangles, in prestrained crystals tangled dislocations are replaced by new tangles which at any given point are impossible to distinguish.

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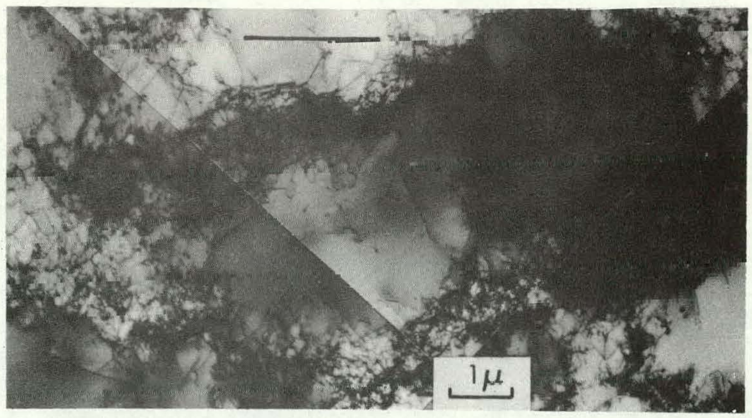

Fig. 1. $(1\bar{2}1)$ foil of a crystal not given a second deformation. The trace of the (111) plane is shown.

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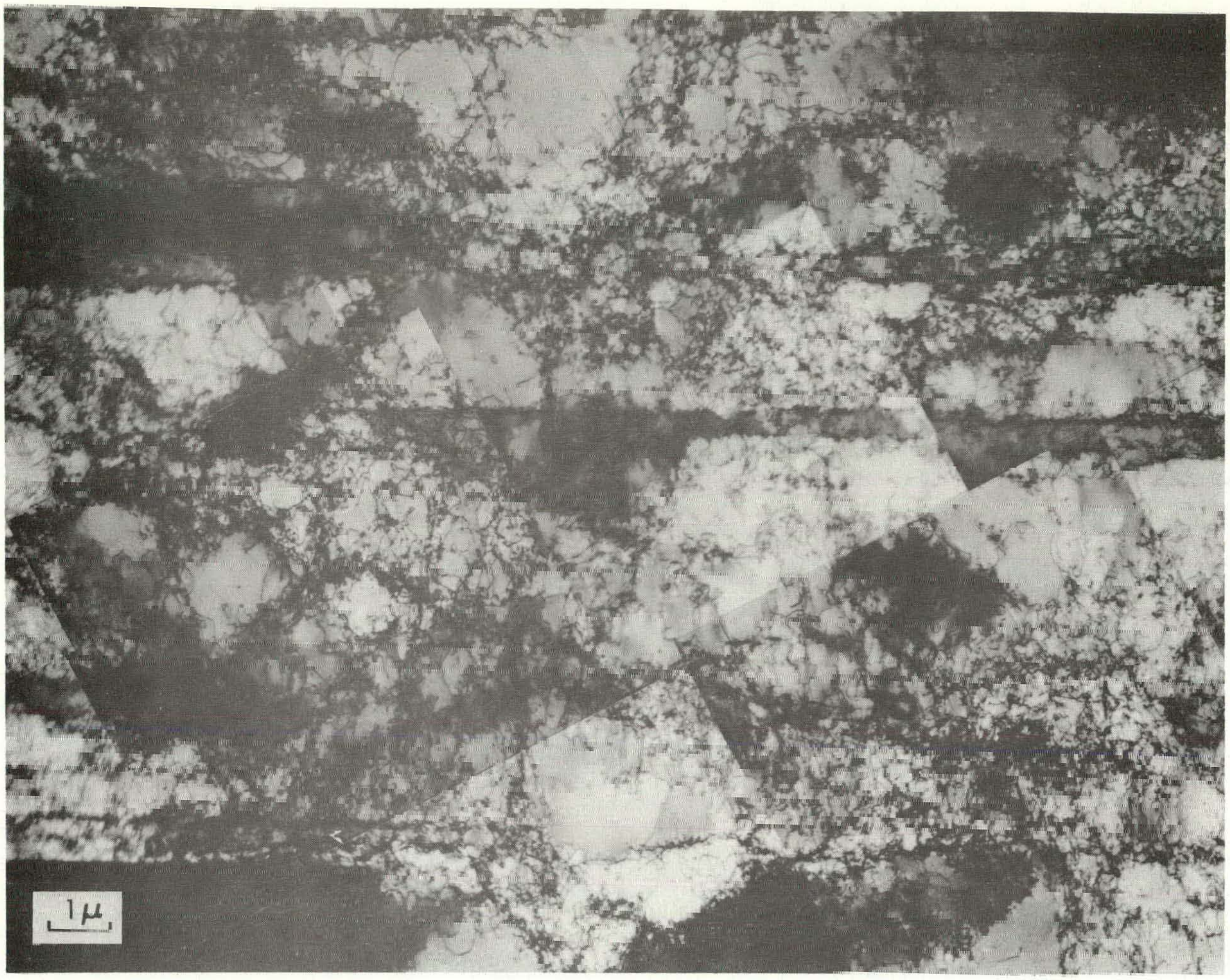

Fig. 2. ($1\bar{2}1$) foil. The primary Burgers vector (b_p) is in the plane of the foil; primary dislocations are out of contrast.

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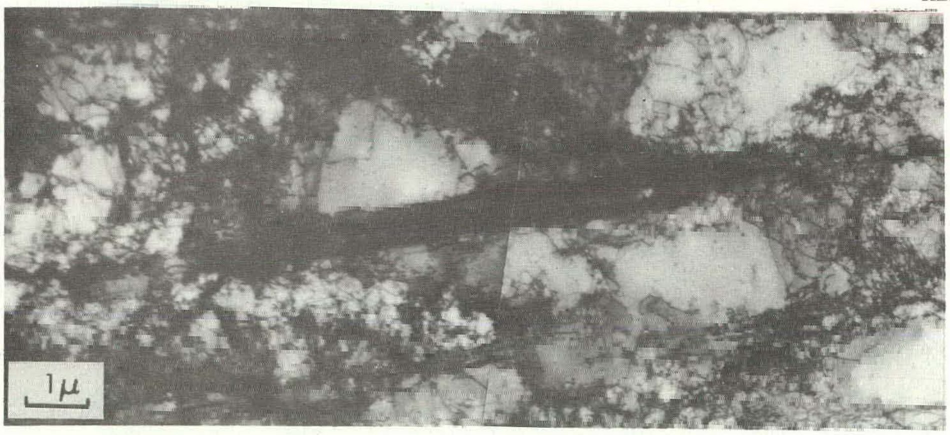

Fig. 3. ($10\bar{1}$) foil. b_p is perpendicular to the plane of the foil; primary dislocations are out of contrast.

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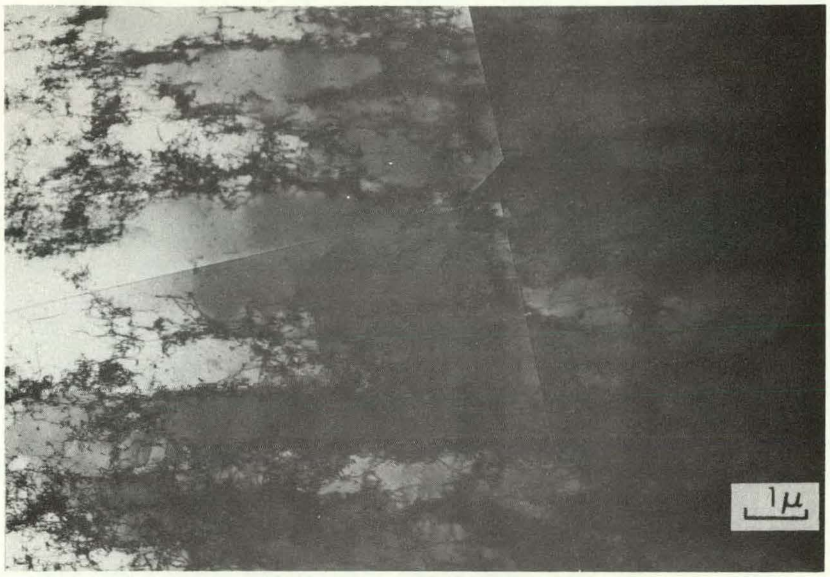

Fig. 4. (hkl) foil. Both b_p and (111) are at 45^o to the plane of the foil. Primary dislocations are in contrast.

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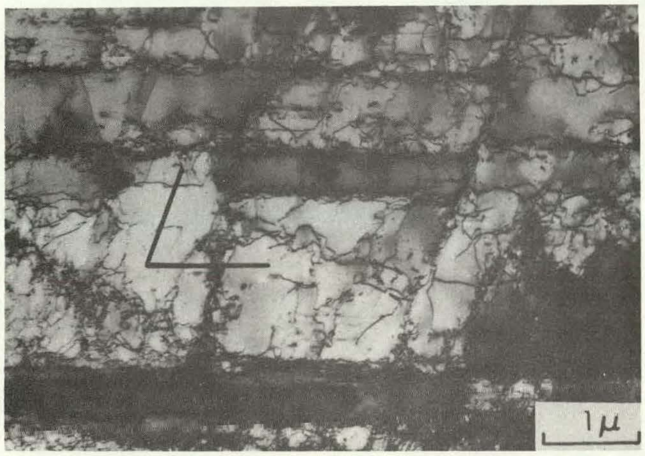

Fig. 5. $(10\bar{1})$ foil. The traces of the primary and cross-slip planes are indicated. Primary dislocations are out of contrast.

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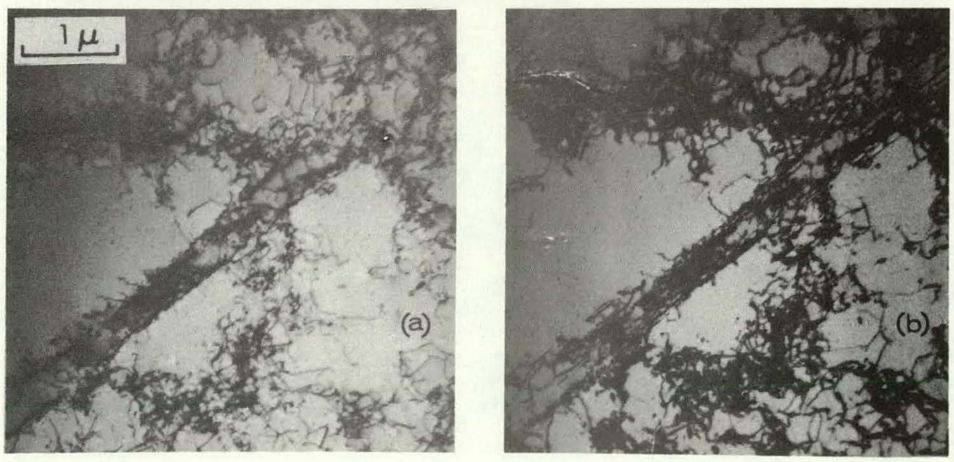

Fig. 6. $(1\bar{2}1)$ foil.

a) Primary dislocations out of contrast.

b) Same area. Primary dislocations are in contrast.

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