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INVESTIGATION OF DISLOCATION MOTION BY ETCH PIT AND YIELD STRESS IN ELECTRON IRRADIATED COPPER*

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ABSTRACT. The dislocation motion determined by the etch pit technique and the macroscopic stress-strain relation are examined in electron irradiated copper single crystals in the temperature range from 4.2° to 100° K. A good correlation is found between these two observations and the dislocation kinetics near the yield point has been determined.

INTRODUCTION

The yield stress and the stress-strain behavior of crystalline materials have been interpreted in terms of dislocation motion and dislocation generation. However, the correlation of the macroscopic deformation with the dislocation behavior is often conjectural and a direct experimental correlation has rarely been achieved in metals. The present study is designed to examine this correlation in specimens in which the electron irradiation has generated suitable dislocation obstacles (1,2), particularly over an extended temperature range.

EXPERIMENTAL PROCEDURES

Single crystals were grown from 99.999% copper rods by the Bridgeman technique. Specimens of 10 x 3 x 3 mm were cut from the single crystals and annealed for 24 hrs at 1020° in a vacuum of 10^{-5} torr to obtain a dislocation density of $\sim 10^4$ disl/cm^2.

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DISCLAIMER

A pair of $(1\bar{1}1)$ surfaces were used to observe the dislocation etch pits. Specimens of two orientations were prepared: one with the long axis parallel to [011] for the etch pit study, and the other 8° away from [011] for the macroscopic deformation. Both $(1\bar{1}1)$ surfaces were irradiated with 2 MeV electrons (3×10^{16} electrons/cm^2) at 100° K. The deformation temperatures were 4.2°, 25°, 50°, and 100° K.

EXPERIMENTAL RESULTS AND DISCUSSION

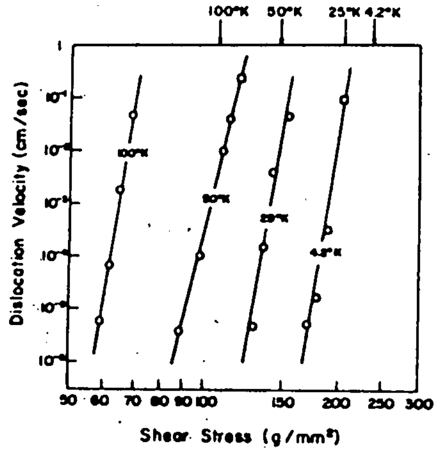

Fig. 1 - The dislocation velocity vs. the resolved shear stress plot. The arrows indicate the macroscopic yield stresses at the corresponding temperatures. The square symbols denote the cases where the dislocation multiplication had taken place and the dislocation velocities were estimated from the length of the slip lines.

The electron irradiation reduced the dislocation velocity drastically and increased the temperature and stress dependence of the dislocation velocity. At the same time, the yield stress was increased roughly four-fold. Figure 1 shows the dislocation velocity-stress relation at 4.2°, 25°, 50°, and 100° K. It can be seen in the figure that the power law, $v = v_0(\tau/\tau_0)^m$, can represent the experimental results well, where v_0 is taken to be 1 cm/sec. The experimental constants, m and τ_0, were determined by the least square method and are summarized in Table I. The parameters determined from macroscopic deformation, m* and τ_y^I, are also tabulated in Table I.

The power exponent, m, of the irradiated copper is far greater than that of the unirradiated copper ($m = 1 \sim 2$) (3-5). It should also be noted that the magnitudes of m are very close to those of m*, especially if one considers the discrepancy between these two parameters in the unirradiated copper. The temperature dependence of the dislocation velocity is reported to be negative in the unirradiated copper (2,3). In the irradiated

Table I: Microscopic and Macroscopic Parameters for Dislocation Motion.

	4.2° K	15° K	50° K	100° K
m	59	53	38	58
$\tau_0 (g/mm^2)$	213	161	125	73
m*	84	39	39	46
$\tau_y^I (g/mm^2)$	236	203	147	107

$m \equiv (d\ell nv/d\ell n\tau)_T$; $m^* \equiv (\Delta\ell n\dot{\epsilon}/\Delta\ell n\tau)$;

τ_y^I : macroscopic yield stress of electron irradiated copper.

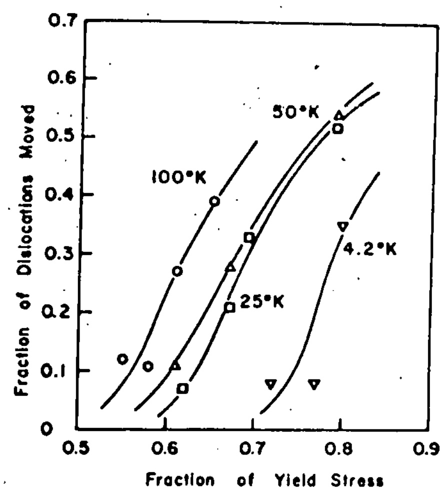

Fig. 2 - The fraction of dislocations moved as a function of applied stress. The applied stresses are normalized with respect to the yield stress of the corresponding temperature.

copper, the temperature dependence is positive and is consistent with the temperature dependence of the strain rate.

Figure 2 indicates the fraction of mobile dislocations as a function of the stress normalized to the yield stress at the corresponding temperatures, τ_y^I (indicated by arrows in Fig. 1). The dislocation multiplication took place above 80% of the yield stress and the dislocation density increased rapidly as the applied stress approached the macroscopic yield stress. Figure 2 also indicates that the dislocations became mobile only when the stress considerably exceeded the yield stress of the unirradiated specimens as the irradiation quadrupled the yield stress, indicating that the source pinning of the dislocations had taken place. On the other hand, this source pinning was considered to be inconsequential to the macroscopic yielding since nearly all the dislocations became mobile and the multiplication took place below the yield stress of the irradiated specimen.

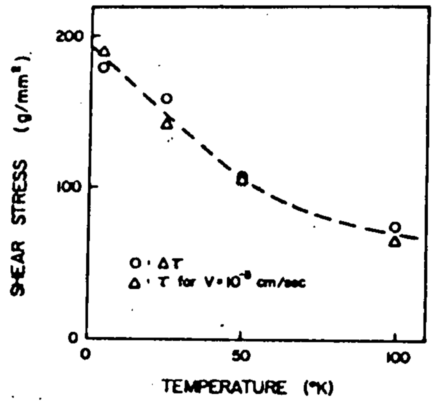

Fig. 3 - The increase of yield stress due to irradiation and the stress required to maintain the dislocation velocity at 10^{-3} cm/sec are plotted against temperature.

Figure 3 compares the stress required to maintain the dislocation velocity of 10^{-3} cm/sec with the yield stress increase due to the irradiation, $\Delta\tau$, at the test temperatures of 4.2°, 25°, 50°, and 100° K. A similar relationship can be found by plotting the estimated effective stress component of τ_y^I and τ_0 or τ_{-1}, the stress for the dislocation velocity of 10^{-1} cm/sec. These observations strongly suggest that the temperature dependence of these two stresses, the yield stress and the stress required to maintain a certain dislocation velocity, are the same. Although m and m* are not equal at every temperature, the deviation is not systematic; therefore one can conclude that the stress dependences of the dislocation velocity and the strain rate are also the same. An analysis of the present results was also performed by computing the activation volumes and activation energies of dislocation motion and plastic deformation, arriving at an essentially similar conclusion.

These observations encourage one to analyze the macroscopic deformation in terms of dislocation kinetics, assuming that the microscopic and macroscopic parameters determined in the present investigation are directly related. It should be noted, however, that the microscopic parameters were determined below 0.8 τ_y^I, while the macroscopic parameters were determined at or above τ_y^I. Extrapolating the dislocation velocities to the corresponding yield stress levels (indicated by the arrows in Fig. 1), one obtains the velocities of $10^3 \sim 10^5$ cm/sec. The corresponding mobile dislocation densities should then be $10 \sim 10^{-1}/cm^2$ for the strain rate used (3×10^{-4}/sec). Clearly, the extrapolation overestimates the dislocation velocity at the macroscopic yielding. On the other hand, the coincidence in stresses such as indicated in Fig. 3 strongly suggest that the velocity of the dislocations is in the range of $10^{-3} \sim 10^{-1}$ cm/sec at the yielding and thereafter. As the applied stress exceeds $0.8\ \tau_y^I$, the rapid multiplication of dislocations increases the dislocation density drastically. The dislocation density was estimated to be $\sim 10^7/cm^2$ at the yielding. The resulting internal stress reduces the dislocation velocities considerably from the extrapolated values. Based on this model, a calculation can predict reasonable yielding behavior along with internal stress, mobile dislocation density and dislocation velocity.

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