Dislocation Mobility in Copper*  

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ABSTRACT  

The velocity of dislocations of mixed edge-screw type in copper crystals of 99.999% purity has been measured as a function of stress at room temperature. Dislocation displacements produced by torsion stress pulses of microsecond duration were detected by etch pitting [100] surfaces. A nearly linear relationship between dislocation velocity and resolved shear stress was found. Stresses from $2.8 \times 10^6$ dyne/cm$^2$ to $23.1 \times 10^6$ dyne/cm$^2$ produced velocities from 160 cm/sec to 710 cm/sec. These data give a value of the damping constant for high velocity dislocations of  

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7 x 10^{-4} \text{ dyn sec cm}^{-2} , in good agreement with the values deduced from internal friction measurements. The results also agree, within experimental and theoretical uncertainties, with the phonon viscosity model for the mobility of dislocations.

INTRODUCTION

The velocities of dislocations have been determined from direct observations of the dislocation displacements produced by known stresses applied for known periods of time in lithium fluoride\textsuperscript{1}, silicon-iron\textsuperscript{2,3}, sodium chloride\textsuperscript{4}, tungsten\textsuperscript{5}, and various semiconductor crystals\textsuperscript{6,7}. The purpose of this paper is to present the results of the first such direct determination of dislocation velocity in a face-centered-cubic metal copper. The experiments were all performed at room temperature.

TEST SPECIMENS

Experiments were performed on single crystal copper specimens in the form of right circular cylinders 1.25 cm in diameter and from 1.48 to 1.90 cm long. The cylindrical surface was modified by four flat [100] observation surfaces, each about 0.3 cm wide, spaced at 90° intervals around the circumference and extending the full length of the specimen. The cylindrical axis was parallel to the [100] crystal axis to within \( \pm \frac{1}{2}^\circ \).

These test specimens were machined from single crystals about 2.5 cm diameter and 12.5 cm long. The crystals were grown in graphite crucibles from a charge of 99.999% copper by the modified Bridgman technique employed by Young and Savage\textsuperscript{8}. Some of the crystals were grown
in the authors' laboratory in a vacuum of $5 \times 10^{-4}$ mm Hg or less. Others, which were kindly supplied by the Sandia Corporation, Albuquerque, New Mexico, were grown in a helium atmosphere. Rough machining of specimens was performed by trepanning and wire slicing using spark erosion machines. Finish machining was accomplished by chemical lapping using a saturated solution of cupric chloride in concentrated hydrochloric acid and a rotating cloth-covered Lucite wheel. At least 0.04 cm thickness of material was removed by chemical lapping, thus removing the mechanical surface damage due to spark erosion machining. The test specimens were annealed for about 100h at 1030 ± 10°C in a hydrogen atmosphere after finish machining.

EXPERIMENTAL TECHNIQUES

Single torsional stress pulses of microsecond duration were applied to the specimens by means of the machine described by Pope et al. This machine utilizes zero-order mode torsional waves in cylindrical bars which are generated in the following manner. An initial static torque is applied to a section of a cylindrical rod which is a part of the torsion rod or load train shown schematically in Figure 1. The torque is applied to the top of the section by dead weight loading of a cranking disc attached to the rod through a rubber sleeve. A 0.015 cm thick glass disc cemented to the bottom of the section transmits this torque to a bakelite fixture which is attached to a fixed bearing tube surrounding the rod. A 0.005 cm thick aluminum foil is cemented between the glass disc and bakelite with Eastman 910 adhesive. The lower section of the torsion rod or load train is attached to the opposite side of the glass disc by means of a butt adhesive bond. This section of the torsion
rod is coaxial with the upper section but does not carry any static torque. The specimen is attached to the bottom end of this lower section of the torsion rod, and the bottom end of the specimen is the free end of the torsion rod system. The section of the rod above the cranking disc is coated with a viscoelastic material in order to attenuate the waves propagating away from the specimen.

The application of the stress pulse to the specimen is initiated by a high-voltage capacitor discharge through the aluminum foil. Explosion of the foil releases the static torque and results in elastic waves which propagate away from the glass disc interface of the torsion rod. The amplitude of the dynamic torque in the lower section of the torsion rod is one-half the initial static torque applied to the machine and, therefore, proportional to the weight hung on the cranking disc. The duration of the stress pulse at any point in the specimen is the time required for the wave to propagate from that point to the free end and return.

The stress wave generated by the release of the static torque propagates without dispersion through the isotropic elastic rod attached to the glass disk because the elastic zero-order mode torsional waves are non-dispersive. The propagation of zero-order mode torsional waves in the (100) oriented copper single crystal specimen is also non-dispersive and the distribution of shear stress over a cross-section of the specimen is the same as that in an isotropic material.

The materials and dimensions of the rod system of the torsion machine were chosen so as to make the acoustic impedance of all the rod sections equal to that of the specimen. Torsional wave reflections at the interfaces between different materials in the load train were
thereby avoided. The rods were made of steel 1.27 cm diameter. This provided an acoustic match with specimens of 1.25 cm diameter. A thermal buffer rod of polycrystalline copper, 2.3 cm long and 1.33 cm diameter, was cemented between the steel torsion rod and the copper specimen. This prevented spurious stresses in the specimens due to differential thermal expansion of the steel and copper while maintaining the acoustic impedance match. A conical joint of 30° included angle was made in the lower steel rod at a point 7.6 cm above the thermal buffer. This facilitated removal of the specimen for polishing and etching. These several portions of the specimen-torsion rod system were joined together by means of Eastman 910 adhesive. The conical joint in the steel rod was parted by the rapid application of heat from a gas torch when it was desired to remove the specimen from the machine in order to etch and observe the dislocations. A photograph of a test specimen cemented onto the thermal buffer and end section of the steel torsion rod is shown in Figure 2.

The amplitude and duration of the stress pulse were measured by means of silicon strain-gages, cemented to the steel torsion rod at a point 18.1 cm from the thermal buffer. The strain-gage output voltage was displayed on both beams of a Tektronic type 555 oscilloscope. The upper beam was triggered by the capacitor discharge employed to initiate the torsion wave and had a sweep rate of 200 μsec/cm. This provided information about any reflected stress pulses applied to the specimen at relatively long times after the primary stress pulse. The initiation of the lower trace was delayed about 116 μsec, and the sweep rate was 20 μsec/cm. This provided detailed information regarding the primary stress pulse.

Fresh dislocations were introduced into specimens by scratching the {100} observation surfaces in a controlled manner with a diamond phonograph stylus or an alumina whisker by means of a special scratching apparatus.
Scratching was performed after the surface had been chemically polished in the manner employed by Livingston. Two etches were used to reveal dislocations on the \{100\} observation surfaces of the specimens. Etching for 3 sec. in the solution used by Livingston produced about the same pit size at etching for 5 sec. in the solution used by Young. Immersion in either etchant was followed by a rinse in hydrobromic acid.

Specimens were etched prior to stress application to reveal the grown in dislocations and the fresh dislocations produced by scratching. A permanent record of this dislocation configuration was obtained by making a replica of the specimen surface on a cellulose acetate film.

Immediately after testing, the specimen was rinsed in distilled water, chemically polished for about 10 sec, etched, and rinsed in hydrobromic acid and distilled water. The ten-second chemical polish did not remove the prior dislocation etch pits. It prepared the \{100\} surfaces so that the subsequent application of etchant was less likely to produce general faceting which obscured the dislocation etch pits. After this second etching the observation surfaces were replicated again. When general faceting did occur during the second etching, the specimen was chemically polished long enough to remove the prior etch pattern and then re-etched and replicated. This latter procedure precluded use of the double etch technique of discerning dislocation displacement on the second replica alone. Dislocation displacement could be determined nevertheless by comparing the first and second replicas.

The blink microscope system developed by Pope et al. was employed to compare the first and second replicas in order to determine dislocation displacements. This system superimposes the images formed by two
microscope objective lenses in one field of view and displays the two images alternately in time at a frequency of about 1 cps. The new positions to which individual dislocations had moved during the stress pulse could then be determined by "blinking".

Considerable care was required to determine the original position of a given dislocation which had moved. It was assumed that the original position of a dislocation must have been somewhere along one of the two traces of \{111\} slip planes which pass through the final position of that dislocation. Use of the "double etch" technique, where possible, was also helpful in determining the original position of a displaced dislocation because the original etch pit is flat bottomed, since the dislocation was not there during the second etch. Use of a scratched specimen was also helpful because nearly all original dislocation positions are very near the scratch.

The distances between the original and final positions of individual dislocations were measured to within ± 4 microns using a filar micrometer eyepiece in the microscope. Dislocation displacements were measured at a minimum of three different stations along the axial length of each specimen. On the average, the displacements of 24 different dislocations were measured at each station along each specimen tested.

Four tests were performed with nominal resolved shear stresses at the outer radius of the specimen ranging from $2.5 \times 10^6$ to $25 \times 10^6$ dyne/cm$^2$. A different specimen was used in each test.

RESULTS

The photographic record of the oscilloscope traces of the torsion bar strain gage signal which was obtained during the test of specimen
number 6-2-1 is shown in Figure 3. The upper trace, at a sweep rate of 200 μsec/cm, shows that some waves reflected from the damping section of the testing machine reach the strain gages about 1,100 μsec after the primary wave front. These late arriving waves produce a stress in the test specimen of only about 3% of the stress produced by the primary wave because they rise in a time which is much longer (about 300 μsec) than the time for an elastic wave to make a round trip transit of the test specimen (10.0 μsec). Thus, the only significant stress acting on the test specimen is due to the primary incident torsion wave and the associated waves reflected from the specimen.

The lower oscilloscope trace in Figure 3, at a sweep rate of 20 μsec/cm, shows the details of the primary incident wave and the wave reflected from the test specimen. The rise time of the incident wave is seen to be about 2 μsec. About 60 μsec after the arrival of the incident wave a small dip in the trace may be seen. This is a small disturbance caused by the conical joint in the steel torsion rod. A second small dip in the record occurs about 115 μsec after the initial wave front. This is a small disturbance produced at the interface between the steel torsion rod and the thermal buffer.

The wave which is reflected from the specimen begins 135 μsec after the initial rise in the lower trace in Figure 3. At this point there is a small drop in torque which does not recover. This shows that the maximum torque transmitted into the specimen was a little less than the torque amplitude of the incident wave. Therefore, it must be concluded that a small plastic wave as well as an elastic wave was initiated in the specimen by the incident wave. At 146 μsec after the initial rise on the record, a large and rapid drop in torque occurs. This is the result of the
The elastic portion of the wave which is transmitted through the specimen to the free end and returns as an elastic unloading wave. The fact that this unloading wave does not reduce the torque completely to zero also shows that a portion of the wave in the specimen was a wave of plastic deformation. The final small portion of the unloading phase of the pulse shown in Figure 3 represents that portion of the wave in the specimen which propagated from the thermal buffer interface to the free end as a plastic wave and returned from the free end as an elastic unloading wave.

These results show that the major portion of the torsional wave in the specimen propagated in an elastic manner. Thus, to a first approximation, the magnitude of the stress is independent of position along the specimen and the duration of stress is linearly proportioned to distance from the free end of the specimen.

The maximum torque transmitted into each test specimen was determined from the portion of the strain gage record which began 135 μsec after the initial rise in the record as discussed above. The relationship between torque and the strain gage signal voltage is readily obtained from the fact that the signal voltage during the relatively long period of essentially constant trace deflection prior to the arrival of reflected waves at the strain gages corresponds to one half the initial static torque applied in the torsion machine. Values of the maximum torque transmitted to the specimen, $T_m$, obtained in this way for the four tests are given in the fourth column of Table I. The maximum shear stress, $\tau_m$, acting at the periphery of cross-sections normal to the specimen axis is obtained from the torque by the usual formula for torsion of cylindrical rods, namely
where $D$ is the diameter of the specimen, as listed in the second column of Table I.

\[ \tau_m = \frac{16 T_m}{\pi D^3}, \]  

(1)

\[ \tau_m = \frac{1}{\sqrt{6}} \tau_m. \]  

(2)

\[ \tau_m = \frac{16}{\pi \sqrt{6}} \frac{T_m}{D^3}. \]  

(3)

The maximum resolved shear stress, $\tau_m$, acting on (111) (110) slip systems can be shown to be

\[ \tau_m = \frac{1}{\sqrt{6}} \tau_m. \]  

(2)

Thus by combining equations 1 and 2, the maximum resolved shear stress is related to the maximum torque transmitted into the specimen by

\[ \tau_m = \frac{16}{\pi \sqrt{6}} \frac{T_m}{D^3}. \]  

(3)

Values of maximum resolved shear stress for each test performed are given in the second column of Table II.
TABLE II
CALCULATED RESULTS

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>$\tau_m \times 10^6$ dynes/cm²</th>
<th>$v$ cm/sec</th>
<th>Corrected Dislocation Velocity $v'$ cm/sec</th>
</tr>
</thead>
<tbody>
<tr>
<td>6-3-1</td>
<td>2.82</td>
<td>240</td>
<td>150</td>
</tr>
<tr>
<td>3-2-1</td>
<td>6.74</td>
<td>390</td>
<td>240</td>
</tr>
<tr>
<td>3-3-1</td>
<td>13.1</td>
<td>550</td>
<td>370</td>
</tr>
<tr>
<td>6-2-1</td>
<td>23.2</td>
<td>1040</td>
<td>670</td>
</tr>
</tbody>
</table>

There are three (110) slip directions on each {111} slip plane. One of these slip directions lies in the {100} observation surface. The resolved shear stress acting on this particular slip system is zero, however. The other two (110) slip directions are inclined to the observation surface and the resolved shear stress acting on these later two slip systems is the value given by equation 3.

The dislocation lines extending into the interior of the test specimen from the observed etch pits are probably oriented nearly at right angles to the (110) trace of the {111} slip planes on the {100} observation surface. The slip directions which are subjected to a non-zero resolved shear stress are inclined to this direction of the dislocation lines. Thus the dislocations which were observed to move in these tests were probably of mixed edge-screw type.

The torque versus time relation in the test specimen is known exactly only at the end of the specimen which is cemented to the thermal buffer rod (this will be designated as the "loaded end" hereafter). The reason
for this is the presence of a significant amount of plastic wave propagation in the specimen. The torque versus time relation at the loaded end of the specimen may be determined unambiguously from the test records because both the incident and reflected waves which propagate between the strain gauges and the loaded end of the specimen are purely elastic and propagate at known wave velocities. The torque versus time relation at the loaded end of specimen number 6-2-1 is shown by the solid line in Figure 4. The experimentally determined relation was approximated by the trapezoidal shaped relation shown by the dashed line, for the purpose of determining dislocation velocities in the manner to be explained. Similar torque versus time relations were determined for all tests. The times $t_1$, $t_2$, and $t_3$ from the beginning of these trapezoidal stress pulses to the beginning and end of the maximum stress period and the end of the pulse respectively are listed for each of the four tests in the fifth, sixth, and seventh columns of Table I.

Photomicrographs of the replicas of specimen number 3-3-1 taken before (a) and after (b) the application of the stress pulse are shown in Figure 5. The scratch (horizontal in the photomicrographs) is parallel to the longitudinal axis of the specimen. The same area of the specimen surface is identified by the segment of the substructure boundary in the lower right hand corner of these two photomicrographs. Two examples of dislocation displacement are indicated at A and B in Figure 5(b).

The points representing mean values of measured dislocation displacement are plotted as a function of distance from the free end of the specimen in Figure 6 for the four tests conducted. The number of individual dislocation displacements measured is indicated near each plotted point. The vertical lines through the plotted points represent the standard
deviation of the measurements. The maximum standard deviation is 13% of the mean dislocation displacement. The straight lines in Figure 6 were drawn to represent the data. This shows that the dislocation displacement is linearly proportional to distance from the free end of the specimen. The points marked "x" at the end of each line in Figure 6 represent the extrapolated value of dislocation displacement, \( d \), at the loaded end of each specimen. These values are given in the eighth column of Table I.

The observed linear relationship between dislocation displacement and distance from the free end of the specimen is consistent with the assumptions that the stress wave propagates through the specimen in an elastic manner and that the acceleration time for the dislocations is negligible. According to these approximations, the duration of stress, \( \Delta t \), is related to distance from the free end of the specimen, \( x \), by

\[
\Delta t = 2 \frac{x}{c},
\]

(4)

where \( c = 0.290 \times 10^6 \) cm/sec is the velocity of propagation of an elastic torsional wave along the \( \langle 100 \rangle \) specimen axis. Thus, if equation 4 is employed to convert the abscissa scale of Figure 6 to a scale of time duration of stress, then the slopes of the lines are equal to the dislocation velocity. First approximation values of dislocation velocity, \( v \), determined in this manner are given in the third column of Table II.

The first approximation values of dislocation velocity, \( v \), are found to be approximately proportional to the \( m = 0.7 \) power of the maximum resolved shear, \( \tau_m \), when these data are fitted to an equation of the form

\[
v = v_o \left( \frac{\tau_m}{\tau_o} \right)^m
\]

(5)
where \( m \) and \( \tau_0 \) are material constants and \( v_o \) is unit velocity. This result indicates, however, that a significant portion of the total dislocation displacement occurs during the rising and falling portions of the trapazoidal stress pulses such as that shown in Figure 4. For this reason a more refined method of analysis of the experimental measurements has been employed to obtained corrected values of dislocation velocity, \( v' \), as a function of resolved shear stress, \( \tau_m \).

The corrected dislocation velocity versus stress relation is obtained in the following manner: First, it is assumed that the relationship is of the form given by equation 5. Second, it is assumed that the torque versus time relations at the loaded ends of the specimens are the trapazoidal shaped relations exemplified by Figure 4 and defined by the values of \( \tau_m, t_1, t_2, \) and \( t_3 \) given for each test in Table I. Then it can be shown that the dislocation displacement at the loaded end of the specimen is given by

\[
d' = v_o \left( \frac{\tau_m}{\tau_0} \right)^m \left[ (t_2 - t_1) + \frac{(t_1 - t_2 + t_3)}{(m + 1)} \right],
\]

(6)

where \( \tau_m \) is the maximum stress during the stress pulse as given in Table II, and \( m \) and \( \tau_0 \) are the undetermined material constants in equation 5.

Various values of material constants, \( m \) and \( \tau_0 \), were chosen as to minimize the quantity

\[
e = \frac{1}{4} \sum_{n=1}^{4} \frac{d_n - d'_n}{d_n}
\]

(7)

where \( d_n \) and \( d'_n \) are the experimental and calculated values respectively of loaded end displacement for the \( n \)'th test. The "best fit" values of \( m \) and \( \tau_0 \) determined in this manner are \( m = 0.7 \) and \( \tau_0 = 0.25 \times 10^4 \) dyne/cm².
The corresponding mean error is \( e = 0.088 \). Thus, the "best fit" relation between the velocity of dislocations, \( v \), in copper and the applied resolved shear stress, \( \tau \), is

\[
v = v_0 \left( \frac{\tau}{0.25 \times 10^4} \right)^{0.7},
\]

(8)

where \( \tau \) is expressed in units of dyne/cm².

A convenient graphical representation of the "fit" between the experimental data and equation 8 may be obtained as follows: The dislocation velocity, \( v_m \), corresponding to the maximum resolved shear stress, \( \tau_m \), during the experimentally determined stress pulse is

\[
v_m = v_0 \left( \frac{\tau_m}{\tau_0} \right)^m.
\]

(9)

Thus the theoretical dislocation displacement at the loaded end, equation 6, may be expressed as

\[
d' = v_m \left[ t_2 - t_1 + \frac{t_1 - t_2 + t_3}{m + 1} \right].
\]

(10)

This shows that the effective duration of the trapezoidal stress pulse is

\[
\Delta t' = \left[ t_2 - t_1 + \frac{t_1 - t_2 + t_3}{m + 1} \right].
\]

(11)

Thus, a corrected dislocation velocity, \( v' \), may be defined for each value of actual dislocation displacement, \( d' \), at the loaded end by

\[
v' = \frac{d}{\Delta t'} = \frac{d}{\left[ t_2 - t_1 + \frac{t_1 - t_2 + t_3}{m + 1} \right]},
\]

(12)

where the "best fit" value of \( m \) is to be employed. Values of corrected dislocation velocity for each test, computed by means of equation 12
using \( m = 0.7 \) are given in the fourth column of Table II.

The values of corrected dislocation velocity, \( v' \), are plotted as a function of the experimentally determined maximum resolved shear stress during the stress pulse, \( \tau_m \), in Figure 7. The solid line in Figure 7 represents the "best fit" velocity versus stress relation given by equation 8. The vertical lines through the plotted points represent the uncertainty in dislocation displacement, \( d \), associated with the extrapolation procedure used to obtain \( d \) from the measurements in Figure 6.

The sensitivity of the "fit" between the experimental data and the choice of the velocity exponent \( m \) may be shown by arbitrarily assuming the attractive value \( m = 1 \). Then the other material constant, \( \tau_o \), may be chosen so as to minimize the mean displacement error as given by equation 7. When this is done it is found that \( \tau_o = 2.7 \times 10^4 \text{ dyn/cm}^2 \), and the mean displacement error is \( e = 0.209 \). Furthermore, new values of the corrected dislocation velocity, \( v' \), may be computed using the value \( m = 1 \) in equation 12. These values of corrected dislocation velocity corresponding to a velocity exponent of \( m = 1 \) are given in the fifth column of Table II and plotted versus maximum resolved shear stress in Figure 8. The corresponding dislocation velocity versus stress relation,

\[
v = v_o \left( \frac{\tau}{2.7 \times 10^4} \right) \text{ cm/sec,} \quad (13)
\]
is shown by the solid line in Figure 8.

**DISCUSSION**

The foregoing results show that when the data obtained in these experiments are fit to a power law functional relationship between dislocation velocity and resolved shear stress, equation 5, the "best
"fit" is obtained by employing a stress exponent, $m = 0.7$. The "fit" is not quite as good if it is assumed that the exponent is $m = 1$. However, the scatter and uncertainty in the experimental measurements are such that the value $m = 1$ may be the true value. Thus, the resistance to the motion of individual dislocations in 99.999% pure copper may be described approximately as a simple linear viscosity, at least in the dislocation velocity range of 100 to 1000 cm/sec.

Various investigators have suggested forms other than equation 5 for the dislocation velocity versus stress relationship. The theory of Gilman involving point defect drag and, consequently short-range interactions, is an attractive description for dislocation mobility in copper. This theory predicts a dislocation velocity versus stress relationship of the form

$$v = c_s \exp\left(-\frac{\text{constant}}{T}\right)$$

(14)

where $c_s$ is the velocity of propagation of elastic shear waves in the material. However, the limiting velocity, found by extrapolating a plot of the values of $\ln v$ versus $1/T$ obtained from this investigation, is $c_s = 10^3$ cm/sec rather than the actual shear wave velocity of $2.14 \times 10^5$ cm/sec.

The theory of Fleischer is based upon the idea that dislocation velocity is governed by the interaction between dislocations and point defects which produce large tetragonal lattice distortions such as result from the introduction of carbon into iron. The mobility data for copper does not extend over a sufficiently large range of velocity and stress to permit a good test of the functional dependence predicted by Fleischer's theory. The relative insensitivity of the flow stress of copper to
temperature would appear to rule out this theory however.

Another mechanism which can limit the velocity of dislocations is the interaction between the moving strain field and the thermal vibrations of the crystal lattice, commonly known as the phonon drag or phonon viscosity effect. Leibfried\textsuperscript{15} was the first to estimate the drag on a screw dislocation moving at a constant velocity caused by the scattering of phonons. Lothe\textsuperscript{16} reviewed and extended this estimate and concluded that for metals Leibfried's result is correct and should be about the same for edge and screw orientated dislocations. Lothe estimates that the effects of anharmonicity in the core region, the phonon viscosity, (lattice vibrations considered as a viscous phonon gas), and the scattering of phonons are each of the same magnitude. The total of these three effects gives a drag stress, \( \tau_d \), at ordinary temperatures, \( T > \Theta \), where \( \Theta \) is the Debye temperature, of

\[
\tau_d \approx \frac{3e}{10} \frac{v}{c_s}, \tag{15}
\]

where \( e \) is the thermal energy density, \( c_s \) is the shear wave velocity, and \( v \) is the dislocation velocity. Thus the phonon interaction theory predicts a linear stress dependence of dislocation velocity, in reasonable agreement with our experimental results.

The thermal energy density, \( e \) in equation 15 is assumed to be given by the relation

\[
e = \frac{3kT}{3} \frac{v}{a^3/z} \tag{16}
\]

where \( a = \) the lattice parameter = \( 3.6 \times 10^{-8} \) cm, and \( z = \) the number of atoms per unit cubic lattice cell = 4. The drag stress calculated from
equation 16 is about 55% of the measured values as shown by the dashed curve in Figure 8. Mason\textsuperscript{17} has also treated the effect of phonon viscosity on dislocation motion, and his theory predicts the same order of magnitude of the drag stress as does the theory of Leibfried, but gives a somewhat modified temperature dependence.

Those theories which depend upon thermally assisted dislocation motion, such as Fleischer's\textsuperscript{14}, predict that dislocation velocity decreases with decreasing temperature whereas those theories that depend upon the existence of a drag stress due to phonon scattering and viscosity predict that the dislocation velocity will increase with decreasing temperature. Stress pulse tests on copper specimens at low temperatures, which are in progress, should enable one better to select between the two types of theories of dislocation mobility.

An indirect method of obtaining the drag force on a moving dislocation is based on measurements of internal friction. The theory of Granato and Lucke, briefly reviewed in reference\textsuperscript{18}, has been used to obtain the damping constant $B$, which is defined as the drag force per unit length of dislocation per unit velocity. The damping constant, given by $B = b\tau_o/\nu_o$ in the notation used here, is $B = 7 \times 10^{-4}$ dyne sec cm$^{-2}$, when the value $\tau_o = 2.7 \times 10^4$ dyne/cm$^2$ from equation 13 and the known Burgers vector, $b = 2.55 \times 10^{-8}$ cm are employed. Alers and Thompson\textsuperscript{19}, Stern and Granato\textsuperscript{18} and Suzuki et al.\textsuperscript{20} have reported values of $B$ at room temperature based on internal friction measurements and an estimate of the dislocation density of their specimens. These values are listed in Table III together with the directly determined value from this investigation.
TABLE III

DISLOCATION DAMPING CONSTANT IN COPPER

AT ROOM TEMPERATURE, $10^{-4}$ dyne sec/cm$^2$

<table>
<thead>
<tr>
<th>Source</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>This study</td>
<td>7</td>
</tr>
<tr>
<td>Alerz and Thompson</td>
<td>8</td>
</tr>
<tr>
<td>Stern and Granato</td>
<td>6.5</td>
</tr>
<tr>
<td>Suzuki, et. al.</td>
<td>0.79</td>
</tr>
</tbody>
</table>

Thus the dislocation damping constant derived from the results of this investigation is in good agreement with most of the values determined from internal friction measurements. This agreement is especially interesting in view of the fact that the dislocation displacements in this investigation are several orders of magnitude greater than those in the internal friction experiments.

The mobility of dislocations at the yield stress in copper is markedly different from that in other materials for which direct data exists. Table IV lists the yield stress, dislocation velocity at the yield stress and mobility exponent of a number of materials. The relatively high velocity of dislocations in copper, combined with a low mobility exponent, $m$, is evident. The small rate sensitivity of the flow stress in copper must therefore be due to a strong stress dependence of the density of moving dislocations in copper. Adams suggested that a strong stress dependence of the moving dislocation density comes about when the internal stress amplitude is comparable in magnitude to the applied stress. The average velocity of a dislocation in copper should correspond to the applied stress and be unaffected by the internal stress, provided that the total stress does not drop to zero. This is a consequence of the linear relation between stress and velocity and the fact that the
TABLE IV

DISLOCATION VELOCITY AT THE YIELD STRESS

(at room temperature unless otherwise noted)

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield Stress $10^6$ dyne/cm$^2$</th>
<th>Velocity at Yield Stress cm/sec</th>
<th>Mobility Exponent $m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tungsten $^5$</td>
<td>730</td>
<td>0.001</td>
<td>4.8</td>
</tr>
<tr>
<td>Lithium Fluoride $^1$</td>
<td>88</td>
<td>0.015 (edge)</td>
<td>25</td>
</tr>
<tr>
<td>Sodium Chloride $^4$</td>
<td>17</td>
<td>0.000001</td>
<td>8</td>
</tr>
<tr>
<td>(High Purity)</td>
<td>2.9</td>
<td></td>
<td>17</td>
</tr>
<tr>
<td>Iron - 3% Silicon $^2$</td>
<td>1,200</td>
<td>0.0000015</td>
<td>35</td>
</tr>
<tr>
<td>Germanium (500° C) $^6$</td>
<td>280</td>
<td>0.0006</td>
<td>1.9</td>
</tr>
<tr>
<td>Copper</td>
<td>5</td>
<td>200</td>
<td>0.7</td>
</tr>
</tbody>
</table>
average stress on the slip plane is equal to the applied resolved shear stress. When the internal stress amplitude is comparable in magnitude to the applied stress, dislocations will be trapped whenever the total stress drops to zero, and very small changes in the applied stress can change the velocity of a significant number of dislocations from zero to the relatively high value corresponding to the applied stress. The rate sensitivity of the flow stress is thereby very small.

The theory of yielding of a single crystal proposed by Johnston states that for a small mobility exponent and a low density of moving dislocations, there will be a large yield drop when a stress-strain test is performed on the material. The density of mobile dislocations was assumed to be a function of strain but not explicitly a function of stress. Except in special cases, there is no yield drop in a stress-strain test of a copper crystal. Therefore, there must be a large density of moving dislocations at the flow stress, i.e., there must be weak pinning of aged dislocations compared to that provided by interstitials in body-centered cubic metals and divalent ions in lithium fluoride and sodium chloride.

SUMMARY AND CONCLUSION

Dislocations in copper move at high velocities at low stresses, and the velocity-stress relation is approximately linear. The low strain rate sensitivity of the flow stress is attributed to a strong stress dependence of the density of moving dislocations and to the high velocities at low stresses, rather than to a large mobility exponent.

The present theories of dislocation mobility may be divided into two categories. One which assumes that dislocation motion is thermally assisted; the other, that thermal assistance is negligible and that
interaction with lattice vibrations causes a major portion of the drag force on a moving dislocation. The form of the velocity-stress relationship observed in copper is in qualitative agreement with the latter theory. Further, this theory predicts drag stresses of about 55% of the measured values. This is considered to be good quantitative agreement between theory and experiment in view of the probable accuracies of both theory and experiment.

A more definitive differentiation between the thermally activated and the phonon viscosity mechanisms of dislocation mobility in copper may be obtained from measurements of the temperature dependence of dislocation velocity which are now in progress.
FIGURE CAPTIONS

Fig. 1  Schematic Diagram of Torsion Pulse Machine

Fig. 2  Test Specimen, Thermal Buffer, and Torsion Rod End Section Assembly

Fig. 3  Torsion Bar Strain Gage Signal versus Time for Specimen Number 6-2-1

Fig. 4  Torque versus Time at Loaded End of Specimen Number 6-2-1

Fig. 5  Etch Pits Near Scratch on Specimen Number 3-3-1, Before (a) and After (b) Testing

Fig. 6  Dislocation Displacement versus Distance from the Free End of the Specimen

Fig. 7  Corrected Dislocation Velocity versus Maximum Resolved Shear Stress, for m = 0.7

Fig. 8  Corrected Dislocation Velocity versus Maximum Resolved Shear Stress, for m = 1.0
FOOTNOTES


Fig. 1 Schematic Diagram of Torsion Pulse Machine
Fig. 2 Test Specimen, Thermal Buffer, and Torsion Rod End Section Assembly
Fig. 3  Torsion Bar Strain Gage Signal versus Time for Specimen Number 6-2-1
Fig. 5 Etch Pits Near Scratch on Specimen Number 3-3-1, Before (a) and After (b) Testing
Fig. 6 Dislocation Displacement versus Distance from the Free End of the Specimen
Fig. 7 Corrected Dislocation Velocity versus Maximum Resolved Shear Stress, for \( m = 0.7 \)
Fig. 8 Corrected dislocation Velocity versus Maximum Resolved Shear Stress, for $m = 1.0$