torque is then

$$L_4=\tfrac{1}{3}L_2+\tfrac{2}{3}L_{3a},$$

given in Eq. (5) in the text. In this case, for the first time, cosine terms are present and the calculated and observed torque curves are less symmetrical—in the 0° to 180° range in θ no symmetry is present. In general $L\neq 0$ at $\theta=0$.

Case 5: When no field is applied during cooling, the torque for the (100) plane is calculated in three parts, corresponding to three positions of the c axis: (1) regions having c axes parallel to [001], (2) regions having c axes parallel to [010], and (3) regions having c axes parallel to [100], perpendicular to the plane of measurement. The torque for the first 2 positions combined is described by L_1' (Case 1 of Appendix). For the third position there are two possible orientations of the a and b axes, 90° to each other in the (100) plane; for one of these $\theta_c=90°$, $\theta_b=\tfrac{1}{4}\pi+\theta$, $\theta_a=\tfrac{1}{4}\pi-\theta$, and the corresponding energy and torque are

$$E_{5a}=\tfrac{1}{2}(K_a+K_b+2K_c)+\tfrac{1}{8}(3K_a'+3K_b'+8Kc') +\tfrac{1}{2}(-K_a+K_b-K_a'+K_b')\sin 2\theta-\tfrac{1}{8}(K_a'+K_b')\cos 4\theta$$

and

$$L_{5a}=(K_a-K_b+K_a'-K_b')\cos 2\theta-\tfrac{1}{2}(K_a'+K_b')\sin 4\theta.$$

The energy, and torque L_{5b}, for the other orientation of position (3) are obtained by interchanging subscripts a and b in these expressions.

The result is then

$$L_{100}=\tfrac{1}{6}(4L_1'+L_{5a}+L_{5b})=-\tfrac{1}{12}(K_a'+K_b'-4K_c')\sin 4\theta.$$

Case 6: When no field is applied to the (110) specimen during cooling, $\tfrac{1}{3}$ of the specimen will have the torque L_2' (Case 2 above), $\tfrac{2}{3}$ will have L_3 (Case 3 above):

$$L_{110}=\tfrac{1}{3}(L_2+2L_3)=(-K_a'-K_b'+4K_c')(\sin 2\theta)/24 +(-3K_a'-3K_b'+8K_c')(\sin 4\theta)/48.$$

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Deformation of Copper Single Crystals at 300°K and 78°K

T. H. BLEWITT
Oak Ridge National Laboratory, Oak Ridge, Tennessee

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Single crystals of copper were deformed at 78°K and 300°K. There was slight temperature dependence of the critical shear stress in this range of temperature and for shear strains less than about 0.25 the stress-strain relationship was linear and independent of temperature. The linear relationship of the stress-strain curve persisted for higher strains at 78°K. The annealing at 300°K of crystals deformed at 78°K showed a decrease in electrical resistance accompanied by a phenomenon suggestive of strain aging. The magnitude of the recoverable resistance was dependent on the strain varying at a faster rate than a linear relationship. A measurable decrease in electrical resistance was not observed for strains less than about 0.30. These results are interpreted according to the theory of dislocations and indicate some modification to the theory of work hardening suggested by Mott.

I. INTRODUCTION

THERE has recently been considerable interest in the phenomena of low-temperature annealing of the electrical resistivity in cold-worked metals. The motivation for this work originates in an interesting paper presented by Seitz[1] at the Pittsburgh Conference on Plastic Deformation, in which Seitz pointed out that plastic deformation increased the number of vacant lattice sites in ionic crystals by about a hundredfold. He also pointed out that after about one hour at room temperature, the number of vacancies has decreased to the equilibrium number. Shortly thereafter a group of experimentalists at North American Aviation[2] found that metals, specifically copper and aluminum, irradiated by 36-Mev α particles at -160°C in the Berkeley cyclotron showed an apparent decrease in residual resistance on warming to room temperature. Since theoretical studies of Seitz[3] and others had indicated that vacant lattice sites and interstitial atoms should be formed by the bombardment of metals by high energy particles, the annealing effects observed by the North American Aviation, and more recently by a group at Illinois,[4] might be attributed to the movement of these defects. Thus, if metals showed a behavior similar to that observed by ionic crystals in that lattice vacancies or interstitial atoms were formed by plastic deformation then an annealing phenomenon in cold-worked metals similar to that observed in irradiated metals should be observed. With this in mind a group at Oak Ridge National Laboratory[5] made measurements on the effect of a room temperature warmup on the elec-

[1] F. Seitz, *Pittsburgh Conference on Plastic Deformation of Crystalline Solids, May, 1950*, U. S. Office of Naval Research Report NAVEXOS-P-834 (unpublished), p. 37
[2] Eggleston, Martin, and Tarpinian, Phys. Rev. **81**, 664 (1951).
[3] F. Seitz, Disc. Faraday Soc. No. 5, 271 (1949).
[4] Marx, Copper, and Henderson, Phys. Rev. **88**, 1006 (1952).
[5] Blewitt, Taylor, and Coltman, Phys. Rev. **82**, 769 (1951).

TABLE I. Qualitative analyses of impurities in copper metal.

	Sample 175-*B*	Sample 182-*B*	Sample 182-*A*
Group I as Ag	<1[a] ppm	<1[a] ppm	<1[a] ppm
Group IIa as As (As, Se, Te, Mo, Ge)	1.5	2.0	4.5
Group IIb as Pt (Pt, Au)	0.15	0.20	0.10
Group IIc as Sb (Sb, Sn)	<0.003	<0.003	0.01
Group III as Fe (Fe, P, Ga, Zr, Hf, W, Ta, Cr, In, RE's)	80	63	60
Group IV as Co (Co, Zn, Ni, Mn)	2.0	1.8	1.0
Group V as Ba (Ba, Sr, Ca, Mg)	<0.1	0.3	<0.1
Group VI as Na (K, Rb, Cs, Na)	200[b]	300[b]	200[b]

[a] Questionable—copper contaminated.
[b] Probably contamination external.

trical resistivity of polycrystalline copper and aluminum samples deformed at liquid nitrogen temperatures. A group at Delft[6] had simultaneously made similar and probably more extensive measurements. Considerable work has been done since these experiments from both a theoretical and experimental standpoint. In a recent paper Seitz[7] has described in some detail the manner by which vacancies are formed by moving dislocations and their possible contribution to work hardening. Measurements of the activation energy to recover the electrical resistivity at these relatively low temperatures have been made by the group at Delft[8] and by the group at North American Aviation.[9] Similar experiments are underway by a group at the General Electric Research Laboratory.[10]

The present paper describes the experiments which were performed by the Oak Ridge group using single crystals of copper, which seem to offer a more basic insight into the formation of mobile defects from deformation than do polycrystalline samples. The stress strain curves at 78°K and at 300°K are reported. Changes in the electrical resistivities and yield point at 78°K following a room temperature warmup are also reported and interpreted.

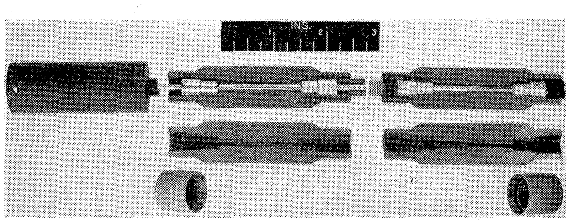

FIG. 1. The crucibles used in casting the samples showing the arrangement by which two crystals of the same orientation are grown.

[6] J. Molenaar and W. H. Aarts, Nature **166**, 690 (1950).
[7] F. Seitz, Advances in Physics **1**, 6.43 (1952).
[8] J. A. Manintveld, Nature **169**, 623 (1952).
[9] R. R. Eggleston, J. Appl. Phys. **23**, 1400 (1952).
[10] R. H. Pry, General Electric Research Laboratory (private communication).

II. EXPERIMENTAL PROCEDURE

A. Sample Preparation

The single crystals of copper were grown from high purity American Smelting and Refining copper by the Bridgman technique. Presumably the metal in its original condition was 99.999 percent copper. From an activation analysis,[11] the results of which are shown in Table I, it is clear that little if any impurity has been added during specimen preparation. The samples were grown in the shape of standard ASTM tensile specimen under high vacuum. Split AGOT graphite molds were used, and Fig. 1 shows the dimensions and physical arrangement of the molds and charging space. In addition to the ease in removing the samples, a split mold has the further advantage that repeated use of the same mold undoubtedly decreases the probability of introducing impurities from the mold. (In all cases two samples of the same orientation were prepared, by placing two crucibles in tandem with each other as is shown in Fig. 1.)

The potential probes used to measure the electrical resistivity consisted of No. 36 magnet wire sintered to the sample. The sintering was achieved by allowing the weight of crystals to be supported by the No. 36 copper wire and then annealing the assembly at a high temperature (1030°C) for 3 hours at a pressure of 10^{-5} mm or less. The sintering technique seemed to give a satisfactory bond, the cohesion being stronger than the wire. This technique has the advantage over welding or soldering in that it reduces the possibility of damaging the relatively fragile specimens and at the same time does not require the presence of any foreign material to secure a bond.

B. The Deformation Technique

The samples were deformed in an Instron Tensile Machine.[12] This is a screw machine with an amplidyne drive and a weighing system utilizing an SR-4 strain gauge to measure the elastic deformation of a stainless steel member. The machine was adapted for deformation at low temperatures.

The load was recorded directly and the elongation determined from the cross head motion. In all cases the cross head was driven at 0.020 in./min with the chart moving at 2 in./min so that each inch of chart represents 0.01 in. of cross head motion. Samples were gripped with split grips.

[11] G. W. Leddicotte and S. A. Reynolds, Nucleonics **8**, No. 3, 62 (1951).
[12] Instron Engineering Corporation, Quincy, Massachusetts.

In order to determine the reproducibility of this arrangement two samples of the same orientation were deformed at liquid nitrogen temperature. The resulting load-elongation curves were reproducible to $\pm\frac{1}{2}$ percent. While there is a very close agreement between the two curves it should be noted that this does not imply that absolute values of strain are this accurate. The assumption is made that there is no deformation in the shoulders of the specimens in using the cross head motion to determine the elongation. The determination of the elongation by direct means compared with the results from the above method indicates that shear strain is probably not in error by more than 0.04 anywhere on the stress strain curve.

C. The Measurement of Electrical Resistance

The electrical resistivity was determined by the potential drop method. A schematic drawing of the apparatus is shown in Fig. 2. Since the resistance of the samples was of the order of a few micro-ohms it was necessary to measure a relatively small potential drop. Consequently numerous precautions were taken to reduce thermal emf's to a minimum. In order to eliminate spurious potentials, readings were taken with the current flowing in both directions and the average of these conditions being taken as the real potential drop across the sample. In all cases potential drops were determined with two different currents flowing through the sample and the degree of reproducibility of these two determinations was used as a criterion of the reliability of the measurement.

The measurements were all made with the sample in the tensile machine. The following technique was used. The samples were loaded in the tensile machine and then cooled in a bath of boiling nitrogen and plastically deformed. The load was then released, 15 pounds being retained to insure good electrical contact. The resistance was then measured, the load released, and the sample allowed to warmup to about 300°K. Following a 16-hour anneal the samples were again cooled to 78°K and a tensile load of 15 pounds placed on the sample, and the resistance was again measured.

III. COMPUTATIONS

The computations of the shear stress and the shear strain were performed in the usual manner. For the case where the sample was in single slip, the resolved shear strain is given by[13]

$$S=\cos\lambda_i/\sin\chi_i-\cos\lambda_0/\sin\chi_0,$$

where λ_i is the angle between the tensile axis and the slip direction after the ith strain, χ_i is the angle between the tensile axis and the slip plane after the ith strain. The relation between χ_i, χ_0 (the original angle between the slip plane and tensile axis), and ϵ_i (the tensile elongation) is given by[13]

$$\epsilon_i=\sin\chi_0/\sin\chi_i.$$

A similar relationship is valid for λ_i, λ_0, and ϵ_i, namely

$$\epsilon_i=\sin\lambda_0/\sin\lambda_i.$$

It is assumed that duplex slip occurs when the angles λ_i and χ_i indicate that the specimen has rotated into the dodecahedral plane. The shear strain is then given by the following relationship:[14]

$$S=\left(\frac{\cos\lambda_D}{\sin\chi_D}-\frac{\cos\lambda_0}{\sin\chi_0}\right)+2.45\ln\left(\frac{1+\sqrt{2}\cot\delta_i}{1+\sqrt{2}\cot\delta_0}\right).$$

The angle δ_0 is the angle between the tensile axis at the point where it initially enters the dodecahedral plane and the (112) plane. For additional strains in a pure metal the tensile axis will remain in the dodecahedral plane and rotate towards the (112) plane. After the ith strain the angle δ_i is given by

$$\epsilon_i'=\sin\delta_0/\sin\delta_i,$$

where the tensile strain ϵ_i' is given by

$$\epsilon_i'=\epsilon_i/\epsilon_D,$$

ϵ_D being the tensile strain at which the tensile axis first enters the dodecahedral plane.

The shear stress in the region of single slip is given by[13]

$$\sigma_i=(F/A_0)\sin\chi_0\cos\lambda_i,$$

where F is the load and A_0 is the initial area.

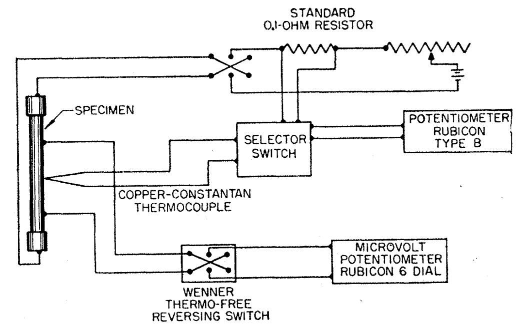

FIG. 2. Schematic diagram of the apparatus used to measure the electrical resistivity.

TABLE II. Effect of temperature on critical shear stress and orientation on the rate of work hardening.

Sample No.	Critical shear stress at 78°K kg/mm²	Critical shear stress at 300°K kg/mm²	Initial work hardening kg/mm²/unit strain	Ratio critical shear stress latent to active plane
142	0.219	0.218	12.84	0.554
158	0.276	0.278	15.4	0.808
175	0.254	0.257	18.86	0.904
182	0.200	0.325	14.3	0.806

[13] E. Schmid and W. Boas, *Plasticity of Crystals* (F. A. Hughes Company, London, 1950).

[14] R. Karnop and G. Sachs, Z. Physik **42**, 283 (1927).

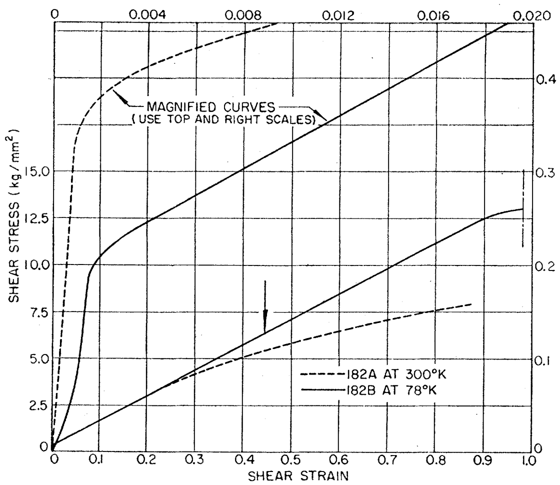

FIG. 3. Stress-strain curve of samples 182*A* and 182*B*. Sample 182*A* was deformed at 300°K and sample 182*B* at 78°K. The arrow denotes the strain at which the specimen axis enters the dodecahedral plane. The vertical dashed line denotes the strain at which a room temperature warmup was made and the resistance measured.

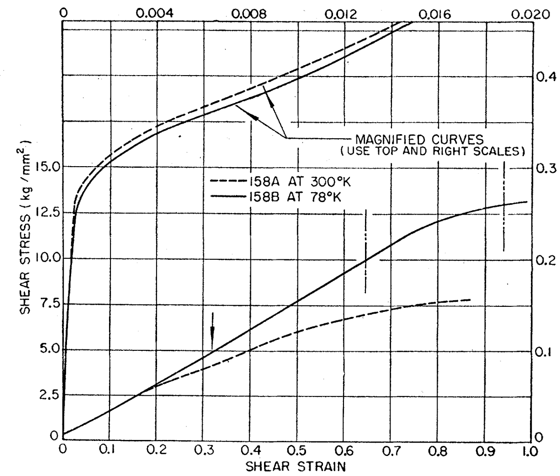

FIG. 5. Stress-strain curve of samples 158*A* and 158*B*. Sample 158*A* was deformed at 300°K and sample 158*B* at 78°K. The arrow denotes the strain at which the specimen axis enters the dodecahedral plane. The vertical dashed line denotes the strain at which a room temperature warmup was made and the resistance measured.

Upon the onset of double slip the equation must be modified and is given by

$$\sigma_i = (F/A_0)\epsilon_i \sin\chi' \cos\lambda',$$

where the angles χ' and λ' are given by

$$\sin\chi' = \cos 54°44' \cos(35°16' - \delta_i),$$

$$\cos\lambda' = \cos 30° \cos\delta_i.$$

IV. RESULTS

In addition to the data obtained on the low-temperature recovery of the electrical resistance, significant and somewhat surprising results were noted of the dependence of the rate of work hardening and the critical shear stress on the temperature. In Fig. 3 through Fig. 6 the stress-strain curves of various pairs of copper crystals are shown. In each figure the stress-strain relationships of two crystals of the same orientation are shown. One sample of each pair was deformed at 78°K, while the other was deformed at 300°K. In order to illustrate the transition from elastic to plastic strain the stress-strain curve for both samples has been shown on an expanded scale for small strains. The curve over the entire range to the ultimate critical shear stress is also shown in the same figure. In each of the figures the points where duplex slip should occur are indicated by vertical arrows and points where resistance measurements were made are indicated by dashed lines.

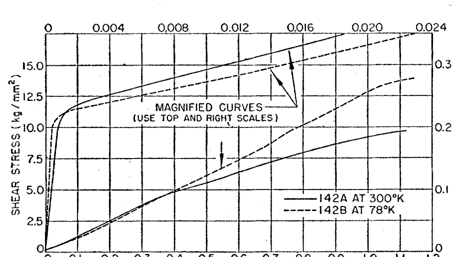

FIG. 4. Stress-strain curve of samples 142*A* and 142*B*. Sample 142*A* was deformed at 300°K and sample 142*B* at 78°K. The arrow denotes the strain at which the specimen axis enters the dodecahedral plane.

From these data (which is summarized in Table II) it would appear that the critical shear stress is independent of the temperature at least in the region from 78°K to 300°K, for with the exception of the pair of Fig. 3 (182*A* and *B*) each of the samples appears to

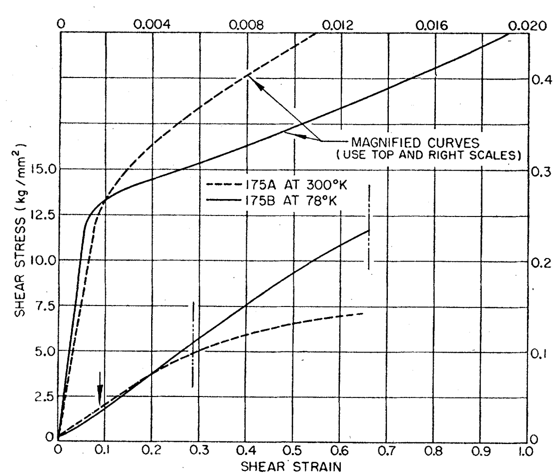

FIG. 6. Stress-strain curve of samples 175*A* and 175*B*. Sample 175*A* was deformed at 300°K and sample 175*B* at 78°K. The arrow denotes the strain at which the specimen axis enters the dodecahedral plane. The vertical dashed line denotes the strain at which a room temperature warmup was made and the resistance measured.

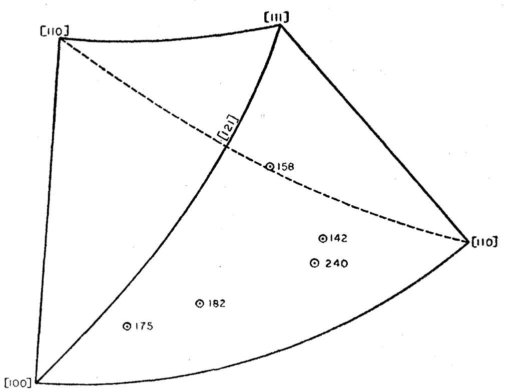

FIG. 7. Orientation of the copper crystals before deformation.

have, within experimental limits, the same critical shear stress. In the case of the 182 pair of samples, the one which was deformed at 78°K, 182*B*, has the lower critical shear stress. It should be noted that if an activation energy were associated with deformation that the converse should be expected. It is, however, not unreasonable that the apparent discrepancy arises as sample 182*A* was slightly deformed before the recorded deformation.

Evidence to support the viewpoint that the critical shear stress is independent of the temperature in the range from 78°K to 300°K, is apparent when the effect of temperature on the rate of work hardening is considered. From the data shown in Figs. 3 through 6 it is clear that for shear strains less than 0.25 the rate of work hardening is independent of temperature. For strains greater than 0.25 the rate of work hardening is less at 300°K than at 78°K. It is interesting to note that the elongation at which the stress-strain curves diverge is independent of the onset of duplex slip, as the point in question is observed to occur both prior to and after the onset of duplex slip. On the other hand there seems to be a decided orientation effect on the rate of work hardening. On consideration of the orientation of the crystals shown in Fig. 7 and the data of Table II, it seems that those crystals which have their specimen axis nearest the dodecahedral plane have the highest rate of work hardening and the lowest ultimate shear strain.[15]

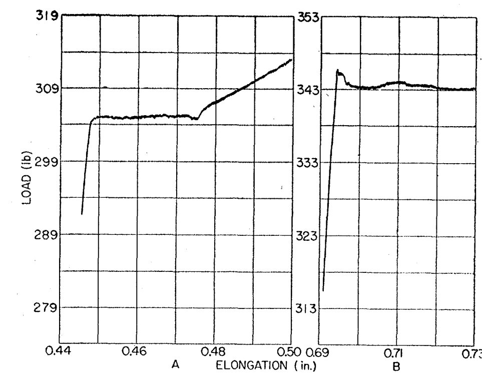

FIG. 8. The load-elongation curve of sample 158*B* after reloading. Curve *A* is that following the initial recovery in resistance. Curve *B* is that following the second recovery in resistance.

The data pertaining to the recovery of electrical resistance by annealing at 300°K are shown in Table III. The data of the table may be summarized as follows. Upon deformation at 78°K followed by a 16-hour anneal at 300°K, the electrical resistance at 78°K is reduced by an amount which is dependent on the amount of deformation. No change in resistance is apparent for strains less than 0.25. This recovery of the resistance is apparently not associated with the recovery of hardness as upon reloading the sample at 78° an increase rather than a decrease in the yield point is observed. A discontinuity of small, but measurable, magnitude is apparent when the yield point is examined on an expanded scale. Such results are illustrated in Fig. 8 and Fig. 9. It appears that a definite strain aging phenomena is observed. This phenomenon was observed only in those regions where a recovery of electrical resistance was observed. For example, stopping the test of a sample, releasing the load, warming up, and then reloading did not result in the small maximum and the "easy flow" region for a strain less than 0.25. Similarly in most cases following a rest period at 78°K the anomolous yielding phenomena was not observed. A yielding was observed for an annealing period at 78°K when the strain was very large and approaching the ultimate strain.

TABLE III. The effect of isothermal annealing at 300°K on the electrical resistance of copper crystals plastically deformed at 78°K.

Sample No.	Shear strain	Shear stress kg/mm²	Resistance before anneal ohms $\times 10^6$	Resistance after anneal ohms $\times 10^6$	Percent change in resistance at 78°K	Change in resistivity ohm-cm $\times 10^9$
49	0.20		7.99	8.00	0	0
182	0.98	14.1	24.44	22.35	8.6	17.2
175	0.28	5.5	8.51	8.48	0.35	0.7
	0.66	11.7	12.80	12.39	3.7	7.4
158	0.65	10.2	12.07	11.89	1.5	3.0
	0.94	13.4	17.43	16.48	5.5	11.0

[15] G. Masing and T. Raffelsieper, Z. Metallkunde 41, 65 (1950), also report an orientation dependence on the rate of work hardening in aluminum crystals.

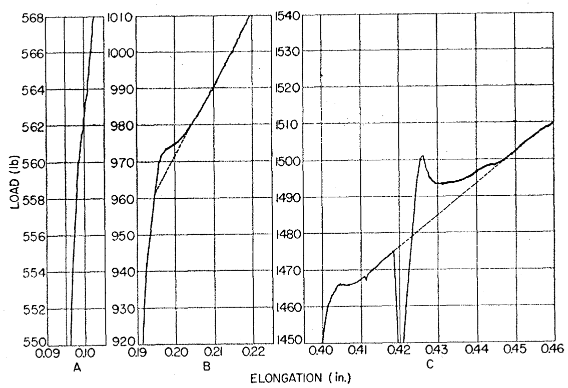

FIG. 9. The load-elongation curve of sample 200A ($\frac{7}{32}$-in. diameter) after reloading. Curve A is the reloading curve after a shear strain of 0.165 (shear stress 3.93 kg/mm²) and an anneal at 300°K for 1 hour. Curve B is that following a shear strain of 0.368 (shear stress 6.98 kg/mm²) following an anneal at 300°K for 1 hour. Curve C is that following a shear strain of 0.759 (shear stress 12.1). The first discontinuity arises from a one hour rest at 78°K. The load was then released and the sample annealed at 300°K for one hour and reloaded.

V. DISCUSSION

A. The Critical Shear Stress

It is reasonably clear from the data presented here that the critical shear stress is independent of temperature.[16] The results of other investigators[13,17] seem to indicate that the critical shear stress has a definite dependence on temperature for both hexagonal and cubic metals. In fact some effort has been made to explain this apparent temperature dependence by Andrade[17] and by Kochendörfer.[18]

More recently Mott,[19] using the concept of the Frank-Read[20] spinning dislocation, has pointed out that one should expect only a slight temperature effect which is directly proportional to the temperature dependence of the bulk modulus for pure and annealed metals.

He has further pointed out that impurity atoms would cause a hardening from the formation of a Cottrell atmosphere in the vicinity of Frank-Read generators. This latter hardening is strongly temperature-dependent,[21] and it would appear that the results of earlier investigators may be due to impurities. Experimentally a strong temperature effect has been observed on the critical shear stress of reactor-irradiated copper crystals and MS 80 α brass crystals,[22] which seem to support in some degree the above viewpoint.

B. The Stress Strain Curves

Koehler[23] and Mott[19] have recently given theoretical consideration to the mechanism of work hardening of metal crystals. Koehler's theory is primarily applicable for the region of small strains, where it has been assumed that hardening results from the fact that the Frank-Read generators lock and generators of shorter length must be used. Mott assumes that barriers to the motion of dislocations pile up a large number of dislocations and create a local stress, in a manner similar to that formulated in the original Taylor[24] theory, which locks the generator and causes work hardening. These barriers are assumed to be sessile dislocations when edge dislocations are generated, and lattice vacancies when screw dislocations are generated. Consider the data presented here in accordance with the theory suggested by Mott.

The relationship between stress and shear strain is interesting in that a linear relationship is observed during the first portion of the stress-strain curve at 300°K. The linear relationship seems to persist for higher strains at 78°K. It may be that at the zero point a linear relationship is preserved for the entire region of deformation by slip. In order to investigate this possibility, a pair of single crystals was deformed as described previously. In this case, one was deformed at 78°K and the other at 4.2°K. The results shown in Fig. 10, seem to substantiate the extrapolation to the zero point. The critical shear stress is nearly the same in both samples, and the stress strain curve at 4.2°K is linear as long as slip occurs. After an elongation of about 60 percent (shear strain of about 1.25) the sample deforms by a mechanism other than slip which has some of the characteristics of twinning, and it finally fractures by cleavage after an elongation of about 120 percent. There is a sharp transition between the region of slip and the region of apparent twinning. It would thus seem reasonable that the stress-strain relationship of copper is a linear one at the zero point.

Mott's treatment, on the other hand, indicates that the stress should be proportional to the square root of the strain. To obtain a linear relationship between stress and strain from this formulation, it is necessary that the length of the Frank-Read generator vary inversely with the strain. It would appear that this is a rather stringent requirement.

[16] The values reported in this paper are in close agreement with those reported by P. W. Neurath and J. S. Koehler, J. Appl. Phys. **22**, 621 (1951). It should be noted, however, that while the experimental technique was in general very different, specimens were prepared from similar materials in a similar manner to those described here.

[17] E. N. da C. Andrade, Phil. Mag. **43**, 1218 (1952).

[18] A. Kochendörfer, *Plastische Eigenschaften von Kristallen und metallischen Werkstoffen* (Verlag Julius Springer, Berlin, 1941).

[19] A. F. Mott, Phil. Mag. **43**, 1151 (1952).

[20] F. G. Frank and T. Read, *Pittsburgh Symposium on Plastic Deformation of Crystalline Solids, May, 1950*, U. S. Office of Naval Research Report NAVEXOS-P-834 (unpublished), p. 44.

[21] A. H. Cottrell and B. A. Bilby, Proc. Phys. Soc. (London) **A62**, 49 (1949).

[22] Unpublished data of Blewitt, Coltman, and Jamison give the following values of the critical shear stress of copper crystals irradiated for 6×10^{18} neutrons/cm²: 3.17 kg/mm at 300°K, 4.25 kg/mm² at 200°K, 6.30 kg/mm² at 78°K, and 7.78 kg/mm² at 4°K. R. E. Jamison also reports the following values for MS 80 brass crystals: 1.55 kg/mm² at 300°K and 3.07 kg/mm² at 78°K.

[23] J. S. Koehler, Phys. Rev. **86**, 52 (1952).

[24] G. I. Taylor, Proc. Phys. Soc. (London) **A145**, 362 (1934).

C. The Formation of Vacancies

It has been previously indicated that the low-temperature recovery of the electrical resistance has been interpreted to be the motion of vacant lattice sites formed by deformation.[25]

From the data of Table III and from the related data of strain aging, it appears as though very few defects are formed when the strain is small. Unfortunately, there is not sufficient evidence, due to the orientation dependence of the stress-strain curves, to decide the exact relationship between strain and the recoverable resistance. Since it has been proposed that those defects are formed by the interaction of a moving screw dislocation and a line dislocation lying in another plane, it would seem that either the number of moving screw dislocation or the number of dislocations with which it can interact must increase rapidly with the strain.

D. Temperature Dependence

It would further seem that the proposed dependence of the stress-strain relationship on the temperature

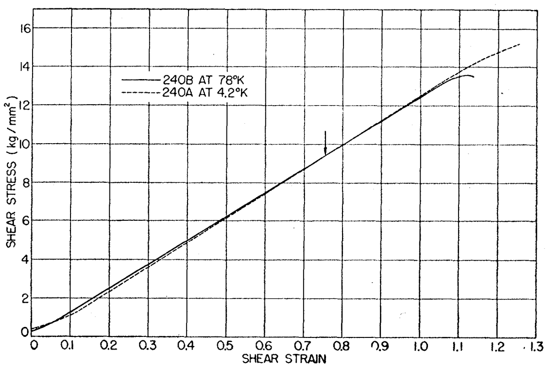

FIG. 10. Stress-strain curve of sample 240*A* and 240*B*. Sample 240*A* was deformed at 4.2°K and 240*B* at 78°K. The stress-strain curve of sample 240*A* considers only that part for which a slip mechanism is operative. A further tensile elongation of 0.60 was observed prior to brittle fracture.

[25] It is not clear that this interpretation should not be modified to some extent in view of recent unpublished data pertaining to the reactor bombardment of disordered Cu_3Au. It was found by Coltman and Blewitt, that the defects, apparently vacancies, formed by fast neutrons failed to induce diffusion and ordering at an appreciable rate below 0°C. It would appear that the activation energy for this process would certainly be no higher than for the motion of defects in copper. However, the defects moving at low temperatures in conformance with the earlier concepts will be arbitrarily termed vacancies with the reservation that further evidence is necessary to confirm this viewpoint.

might require some modification in the relatively low temperature regions where normal self-diffusion cannot occur, as it is quite apparent that the role attributed to vacant lattice sites in the hardening process must be modified. If these defects have been annihilated, as perhaps is suggested by the reduction in resistance, then following the annihilation or the motion of these defects to traps as the case may be, one should expect an accompanying reduction in hardness. Instead, a small but definite increase in resistance to flow is observed with the characteristics associated with strain aging. It seems that these vacant lattice sites have sufficient mobility at 300°K to move to dislocations, and hardening occurs in the manner suggested by Cottrell.[26] It is tempting to assume from the results in Figs. 3 through 6 that thermal fluctuations enable dislocations to break loose from the barriers restricting their motion. It appears that the activation energy for this process is dependent on the localized stress since a higher stress is required at lower temperatures. It may be that the mechanism proposed by Frank, which should be temperature-dependent, to explain the passage of slip lines through grain boundaries would be applicable here. That is, after Frank,[27] a stress concentration such as that in the vicinity of a dislocation "pile up" might result in the activation of a new generator with the help of thermal energy. As the temperature decreases, it is observed that the stress concentration must increase.

VI. ACKNOWLEDGMENT

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Note added in proof.—F. D. Rossi[28] has independently obtained results on the deformation of copper crystals at 300°K which show that the rate of work hardening is dependent on the crystal orientation.

[26] A. H. Cottrell, *Bristol Conference on the Strength of Solids* (Physical Society, London, 1948), p. 30.

[27] F. G. Frank, *Pittsburgh Conference on the Plastic Deformation of Crystalline Solids, May, 1950*, U. S. Office of Naval Research Report NAVEXOS-P-834 (unpublished), p. 89.

[28] F. D. Rossi, Sylvania Electric Company, private communication.

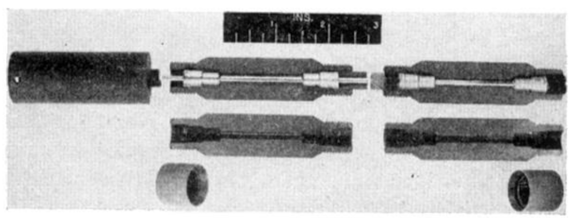

FIG. 1. The crucibles used in casting the samples showing the arrangement by which two crystals of the same orientation are grown.