ture.[6] Thus a series of parallel metal platelets occurred in each crystal of V_2C. Since the bulk sample was made up of many of these crystals, oriented in a random way, a continuous interlacing thread of metal could be produced. This could shield the nonsuperconducting V_2C and cause the bulk effect. Niobium metal has been seen to deposit in the same way upon the cooling of Nb_2C. In addition, this type of behavior is consistent with the phase diagrams of these systems.[3,4] Ta_2C is presumed to behave in the same manner.

In view of these results, and the similarity between the transition temperature of the metal and the M-M_2C mixture, it appears that the materials used in the previous investigation contained a slight amount of the metal phase, which was mistaken for the total sample because of the bulk effect, and that pure Ta_2C, Nb_2C, and V_2C are not superconducting down to 1.98°K.

[6] C. S. Barrett, *Structure of Metals*, (McGraw-Hill Book Company, Inc., New York, 1952), Chap. 22, pp. 538–580.

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Modulus and Damping of Copper after Plastic Deformation at 4.2°K

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An investigation of the Bordoni dislocation relaxation peaks in copper has been carried out under experimental conditions which permit plastic deformation of specimens at 4.2°K, with measurement of their Young's modulus and internal friction upon subsequent warmup. Isochronal annealing at progressively higher temperatures in the interval from 100° to 360°K, taking data from 4.2°K after each anneal, shows (a) a pronounced reduction in height of both the major peak at 62°K and the minor peak at 28°K for annealing temperatures to 200°K, (b) a slow regrowth and shift to higher temperature of the major peak with annealing above 200°K, accompanied by continued diminution of the minor peak, and (c) a monotonic increase, through the full range of annealing temperatures, of the Young's modulus as measured at 4.2°K. A qualitative discussion of these results indicates that the interaction between point defects and dislocations is an essential feature of the dislocation relaxation process.

I. INTRODUCTION

THE initial observation by Bordoni[1] of low-temperature internal friction peaks in plastically deformed metals has led to extensive investigation. The work done in this area prior to 1960 has been reviewed by Niblett and Wilks.[2] It is generally accepted that the low-temperature anelasticity displayed by cold-worked metals arises from the thermally activated displacement of dislocations in the crystal lattice. Several markedly different models for the dislocation relaxation process have, however, been presented.

One of these, developed by one of the present authors,[3] is based upon the thermally activated motion of paired partial dislocations between vacancy pinning points. This model, involving an interaction between point defects and dislocations, is substantially different from those of Seeger[4] and of Brailsford.[5] Here the essential parameters are related to intrinsic properties of dislocations comprising a network with fixed pinning points, either dislocation nodes or point imperfections.

The present work has been undertaken in an effort to evaluate the relevance of point imperfections to the dislocation relaxation process. Our approach is based upon the assumption that it should be possible to exercise some control over the type and number of such imperfections at dislocation lines by initially cold working specimens at temperatures below which the imperfections can migrate. Then, upon warming in steps to progressively higher temperatures, the point defects present in the lattice migrate to sinks with the more mobile species moving first. It is presumed that dislocation lines constitute one type of available sink.

Interpretation of such an experiment rests first of all upon an understanding of the means by which point imperfections are generated in significant quantities by moving dislocations during cold work.[6] Furthermore, identification of the defect species migrating in a given temperature range depends upon the analysis of a wide variety of recovery experiments. These experiments record the irreversible variation of such parameters as electrical resistivity and release of stored energy; these changes occur upon annealing after low-temperature pretreatments including cold work, irradiation, and quenching. They appear as a series of steps or stages; the change associated with each extending over a relatively narrow range of annealing temperatures. The interpretation of these stages in terms of the

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[1] P. G. Bordoni, J. Acoust. Soc. Am. 26, 495 (1954).

[2] D. H. Niblett and J. Wilks, in *Advances in Physics*, edited by N. F. Mott (Taylor and Francis, Ltd., London, 1960), Vol. 9, p. 1.

[3] L. J. Bruner, Phys. Rev. 118, 399 (1960).

[4] A. Seeger, Phil. Mag. 1, 651 (1956).

[5] A. D. Brailsford, Phys. Rev. 122, 778 (1961).

[6] F. Seitz, in *Advances in Physics*, edited by N. F. Mott (Taylor and Francis, Ltd., London, 1952), Vol. 1, p. 43.

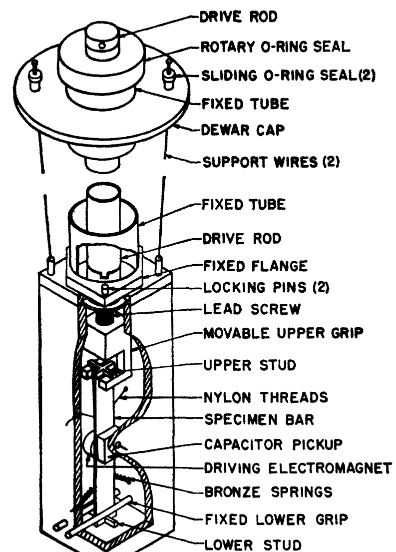


FIG. 1. Schematic diagram of the specimen holder and associated apparatus.

migration of specific point-defect configurations has been reviewed by a number of authors.[7–9]

II. EXPERIMENTAL METHOD

Internal fraction and Young's modulus are measured by a conventional dynamic method utilizing the flexural vibration of specimen bars. The bars are 8 cm long with a rectangular cross section of 1 cm by 1 mm. The rate of free-decay of specimen vibration, noted upon shutting off the external drive, provides a measure of the internal friction; Young's modulus is determined by measurement of the resonant frequency.

Figure 1 is a simplified sketch of the specimen holder and auxiliary apparatus. The specimen bar is suspended nodally using a combination of nylon threads and 0.003-in. phosphor-bronze spring wires as shown. Small stainless-steel studs are silver soldered to each end of the bar to facilitate its deformation. Deformation is accomplished by turning the drive rod from above and hence the lead screw to which it is coupled. This in turn raises the upper grip until it engages the upper pair of studs on the specimen bar. The bar then rises with the upper grip until its lower studs engage the fixed lower grip. At this point continued turning of the drive rod results in tensile deformation of the specimen. After deformation is completed the upper grip is lowered by reversing the direction of rotation of the drive rod; the specimen drops to an equilibrium position, determined by the length of the supporting threads, in which it hangs clear of both upper and lower grips. Finally the specimen holder is dropped clear of the fixed flange and tube as well as the drive rod by lowering its support wires from above. This serves to minimize mechanical and thermal contact with the external surroundings. The entire sequence of operations described above is performed with the specimen maintained at 4.2°K by suspension of the assembly from the Dewar cap into a liquid helium and concentric liquid nitrogen Dewar of conventional design. Flexible leads not shown in Fig. 1 maintain the necessary electrical connections to the specimen holder. A heater consisting of constantan wire wound on a copper shield, also not shown in Fig. 1, surrounds the holder and permits acceleration of the warm-up rate. Copper-Constantan thermocouples are used for temperature measurement and control.

A block diagram of the electrical circuitry employed is shown in Fig. 2. Excitation of specimen vibration is accomplished magnetically. For this purpose a 0.005-in. permalloy tape about 1 cm square is silver-soldered to the center of the specimen and faces a small electromagnet which is attached to the holder. Detection is by means of a condenser microphone-type pickup as illustrated. The output, after amplification and broadband filtering, is returned to the driving electromagnet through a phase shifter and limiter. With sufficient loop gain and appropriate adjustment of phase the specimen bar is driven spontaneously into oscillation at a frequency corresponding to excitation of its fundamental mode of flexural vibration. Standard electronic counting and timing equipment is used to monitor the resonant frequency of the self-excited specimen, and to measure the rate of decay of vibration when the loop is broken. Since the limiter output is constant the level of oscillation at its input or elsewhere in the loop is approximately proportional to the Q of the specimen. The level is recorded as a function of temperature as a supplement to the free-decay data. An exact propor-

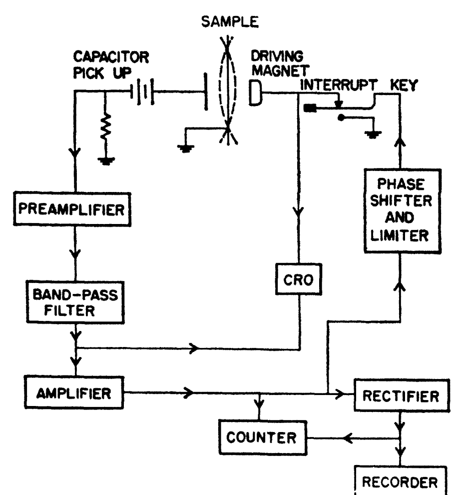


FIG. 2. Block diagram of the electrical circuitry.

[7] H. G. van Bueren, *Imperfections in Crystals* (North-Holland Publishing Company, Amsterdam, 1960), p. 283.
[8] A. Seeger, *Proceedings of the Second United Nations International Conference on the Peaceful Uses of Atomic Energy* (United Nations, Geneva, 1958), Vol. 6, p. 250.
[9] F. Seitz, Trans. Met. Soc., A.I.M.E. **215**, 354 (1959).

tionality is not maintained because of small temperature-dependent variations in drive efficiency and detector sensitivity.

III. RESULTS

All damping data are presented in terms of the inverse storage factor Q^{-1}. Modulus data are presented in terms of speciment resonant frequency ν, related to the Young's modulus E by the expression

$$\nu = (1/2L)(E/\rho)^{1/2}, \tag{1}$$

where ρ is the density of specimen material. The evaluation of L, a constant having the dimensions of length, is readily accomplished.[10] The result,

$$\Delta E/E = 2\Delta\nu/\nu, \tag{2}$$

obtained from Eq. (1) by differentiation, is used to present some of our results directly in terms of variation of the Young's modulus. Experimental points are not shown in the plots of Q^{-1} and resonant frequency vs temperature since in each case a number of curves are superposed. Q^{-1} has been measured with an accuracy of 3% or better; resonant frequency has been determined with a precision of 0.01%. Maintenance of specimens under helium gas at one-atmosphere pressure contributed a background component to Q^{-1} of less than 1×10^{-4}. No correction has been made for thermoelastic damping; this effect is completely negligible in the temperature range of the Bordoni peaks.

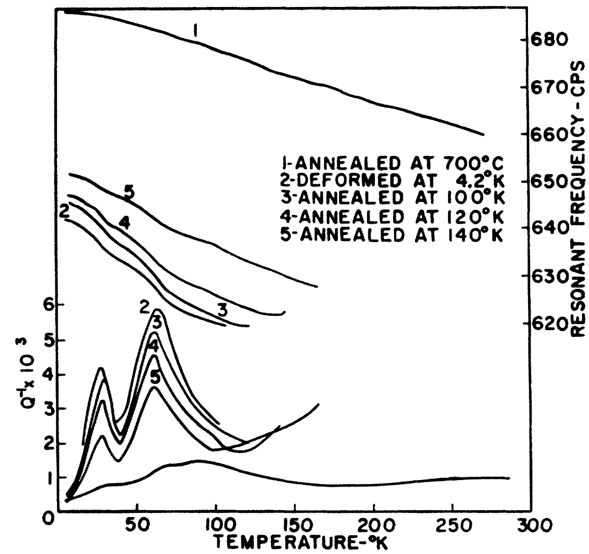
The specimens used in this study have been prepared from A. S. & R. spectrographic grade copper. After attachment of the stainless-steel studs and permalloy tape referred to above the samples were annealed and placed in the apparatus.

A complete set of annealing data on one specimen is exhibited in Figs. 3(a), (b), (c). This specimen, after preparation, was vacuum-annealed for 4 h at 700°C and showed a grain size of about 1 mm. The first run, prior to deformation, showed slight evidence of the Bordoni peaks [curves 1, Fig. 3(a)]. The consecutively numbered curves (2 through 12) indicate the marked increase in height of the Bordoni peaks immediately after 3% tensile deformation at 4.2°K, and the subsequent changes brought about by isochronal anneals at progressively higher temperatures in the range from 100° to 360°K. Anneals were for a period of 16 h. The last numbered curve in Fig. 3(a) and Fig. 3(b) is repeated in Fig. 3(b) and Fig. 3(c), respectively, to enhance continuity of presentation. The changes observed upon annealing may be summarized as follows:

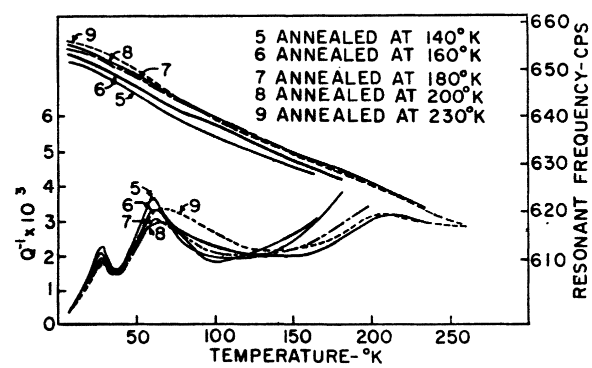
(a) The height of the minor peak at 28°K is steadily reduced throughout the full range of annealing temperatures.

(b) The major peak at 62°K is attenuated with annealing to 200°K; at higher annealing temperatures it grows and shifts to about 70°K.

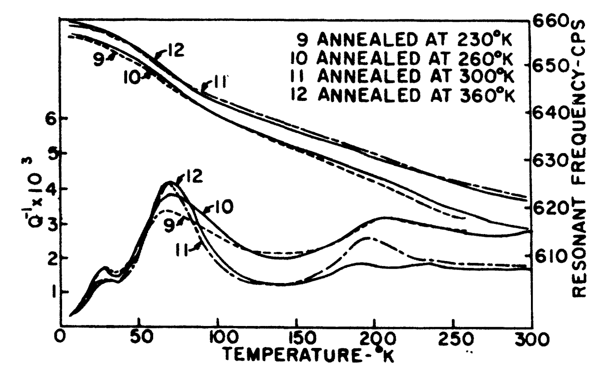
(c) The Young's modulus at 4.2°K increases monotonically throughout the full range of annealing temperatures.



(a)



(b)



(c)

FIGS. 3(a), (b), and (c). Internal friction and resonant frequency vs temperature for copper initially deformed at 4.2°K, then subjected to 16-h isochronal anneals at progressively higher temperatures in the interval from 100 to 360°K.

[10] P. M. Morse, *Vibration and Sound* (McGraw-Hill Book Company, Inc., New York, 1948), 2nd ed., p. 151.

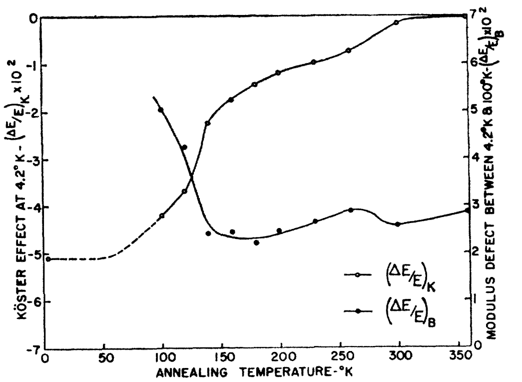


FIG. 4. Dependence upon annealing temperature of the 4.2°K modulus defect (open circles) and the total modulus defect due to relaxation occurring between 4.2° and 100°K (full circles). Note that the scale on the left, which pertains to the 4.2°K modulus defect, decreases from zero at the top of the graph.

It should be noted that no direct comparison can be made between resonant frequency curve 1 and the subsequent curves because of geometric changes associated with deformation of the sample. A precise geometric correction could not be made.

IV. DISCUSSION

In this section we relate our experimental results to the problem of dislocation relaxation, making specific reference to the models mentioned in the Introduction.

An essential part of our discussion will be a consideration of the recovery, originally observed by Köster,[11] of the Young's modulus and internal friction following plastic deformation of metals. This "Köster effect," observed in nonmetallic solids as well, is the subject of a review article by Nowick.[12] The recoverable modulus defect is explained by Mott[13] and Friedel[14] in terms of the stress-induced bowing out of dislocation segments between fixed pinning points. This defect is related to the properties of the dislocation network by the expression

$$(\Delta E/E)_K=\alpha N_0 l_0^3=\alpha N l_0^2, \tag{3}$$

where α is a dimensionless constant of order unity, N_0 is the number of active dislocation segments per unit volume, l_0 is an effective segment length, and $N=N_0 l_0$ is the density of dislocations contributing to the modulus defect. Recovery of the modulus can thus be explained in terms of a reduction of N or l_0 by redistribution or annihilation of dislocations, or in terms of a reduction of l_0 by migration of point defects to dislocation lines where they act as pinning points. Our data show that a modulus defect is present at 4.2°K immediately after plastic deformation, and that it is reduced by annealing. We consider this to be a manifestation of the Köster effect. In Fig. 4 we have plotted, using Eq. (2), the fractional increase of Young's modulus as a function of annealing temperature. The unpublished work of Dieckamp and Crittenden,[12] which was extended to higher annealing temperatures, indicates that the modulus defect was not entirely removed in our experiments.

It is important to point out that the Köster effect observed at 4.2°K arises from the stress-induced bowing out of dislocation segments, apparently without thermal activation; at any rate the importance of thermal energy to the process has not been established. As such, we should expect that the recoverable modulus defect at 4.2°K will constitute only a part of the defect observed at higher temperatures where more thermal energy is available. In particular, the Köster effect observed at temperatures where dislocation relaxation is complete would be expected to include variation of the full modulus defect associated with the Bordoni peaks when recovery processes affecting the amplitude of these peaks occur after plastic deformation.

The total modulus defect contributed by the Bordoni peaks has been plotted as a function of annealing temperature, also in Fig. 4, for comparison with the 4.2°K modulus defect annealing data. This dislocation relaxation contribution has been determined for each of the resonant frequency vs temperature curves numbered 3 through 12 of Figs. 3(a), (b), (c) by use of the relation

$$(\Delta E/E)_B=2[\nu(4.2°\text{K})-\nu(100°\text{K})-\Delta\nu]/\nu(4.2°\text{K}). \tag{4}$$

Here $\nu(T)$ is the specimen resonant frequency at temperature T and $\Delta\nu$ is a constant frequency difference which takes account of the normal temperature dependence of the Young's modulus in the absence of relaxation. The quantity $\Delta\nu$ is determined from data on the annealed sample and, since the same correction is used for all points on the plot of $(\Delta E/E)_B$ vs annealing temperature, any adjustment of its value would serve only to raise or lower the curve without changing its shape.

Comparison of the two curves in Fig. 4 indicates that the quantities $(\Delta E/E)_K$ and $(\Delta E/E)_B$ have closely related annealing characteristics; both display rather distinct step-like changes at about 130° and 280°K. At each step both quantities are reduced in magnitude. The reduction of $(\Delta E/E)_K$ continues monotonically between steps; $(\Delta E/E)_B$, on the other hand, shows a slight rise after each step. It is not possible to plot a point at 4.2°K for the $(\Delta E/E)_B$ curve since annealing effects were evident during the first warm up to 100°K following plastic deformation at 4.2°K [curves 2, Fig. 3(a)]. This annealing begins at about 50°K as indicated by comparison of resonant frequency vs temperature curves 2 and 3. The convergence of curve 2 toward curve 3 has been used as a rough guide in drawing the dashed portion of the plot of $(\Delta E/E)_K$ vs

[11] W. Köster, Z. Metallkunde 32, 282 (1940).
[12] A. S. Nowick, *Creep and Recovery* (American Society for Metals, Cleveland, Ohio, 1957), p. 146.
[13] N. F. Mott, Phil. Mag. 43, 1151 (1952).
[14] J. Friedel, Phil. Mag. 44, 444 (1953).

annealing temperature. Here a point is available at 4.2°K immediately after deformation, and since annealing does not begin until about 50°K we conclude that no significant change of Young's modulus would occur with prolonged aging at temperatures in the liquid helium range. If the $(\Delta E/E)_B$ curve could be extended to lower annealing temperatures (by making measurements between 50° and 100°K and returning to the desired annealing temperature so rapidly that negligible annealing could occur during measurement) it would be expected to bend over and assume a constant value below 50°K since no annealing effects are evident below this temperature. The level at which this saturation would occur cannot, however, be determined from our data.

We are now in a position to consider specific models for dislocation relaxation. The abrupt-kink model of Brailsford[5] supposes that dislocation segments, firmly pinned at each end, lie either parallel to or across close packed directions in the slip plane. Kinks form only by thermal activation on parallel aligned segments. Kinks, of sign and density determined by the average angle of deviation from parallel alignment, will be present at 0°K on those segments which lie across close-packed directions. It is proposed that the Bordoni peaks arise from stress-induced variation of the thermally activated diffusion rate of "built-in" kinks on dislocation segments of the latter class. Thermal generation of kinks is considered to be negligible at temperatures where the Bordoni peaks occur, hence parallel aligned segments do not contribute to dislocation relaxation according to Brailsford's model. This model indicates then that neither class of dislocations will respond to dynamic stresses at temperatures below those at which the Bordoni peaks are observed. Since all glissile dislocations must belong to one or the other of these classes, it appears to us that this conclusion cannot be reconciled with the observation of a modulus defect at 4.2°K. We assume this defect to be explained by the theory of Mott and Friedel discussed above.

The same classification of dislocations has been employed by Seeger[4] in his theoretical treatment of dislocation relaxation. Seeger concludes, however, that only parallel aligned dislocations contribute to relaxation. They do so by thermal activation of a loop or double kink in the otherwise straight dislocation line. The term "loop" as used here refers explicitly to a part of a dislocation segment which is displaced from its equilibrium position by thermal activation; this usage will be adhered to throughout our discussion. The loop formed must be of a critical length such that it can continue to expand under the influence of the external stress after formation. A shorter loop will collapse as a result of the attractive interaction between the pair of kinks of opposite sign of which it is composed. It is evident that, from this point of view, the same class of dislocations cannot be responsible for both the Bordoni peaks and the modulus defect at 4.2°K. We will therefore explore the consequences of the assumption that Seeger's mechanism is operative, and that the dislocations responsible for the 4.2°K modulus defect are among those which lie across close packed directions in the slip plane.

Seeger, Donth, and Pfaff[15] have shown that the maximum value of the internal friction, for this mechanism, is given by

$$Q^{-1}_{\max}=p/2(1+p)^{1/2}, \tag{5}$$

where

$$p=\beta N_0's. \tag{6}$$

Here N_0' is the number of active dislocation segments per unit volume and s is the mean area swept out by each loop after formation. The quantity β, evaluated explicitly in the reference cited, depends upon temperature, applied strain amplitude, and the Peierls stress as well as the elastic and crystallographic properties of the specimen material. A lower limit for $Q^{-1}_{\max}$ is obtained by setting $s=al_0'$, where a is the lattice spacing and l_0' is the average dislocation segment length. Then from Eqs. (5) and (6) we have

$$\begin{aligned} Q^{-1}_{\max}\ (\text{lower limit}) &\cong \tfrac{1}{2}\beta a N_0' l_0' \\ &\cong \tfrac{1}{2}\beta a N', \end{aligned} \tag{7}$$

where $N'=N_0'l_0'$ is the density of dislocations participating in the relaxation process. The primed quantities distinguish these dislocations from those that contribute to the 4.2°K modulus defect. Eq. (7) is derived assuming $p\ll 1$; this follows from Eq. (5) since $Q^{-1}_{\max}\ll 1$. Seeger *et al.* suggest that some dislocations may sweep out a larger area limited only by the applied stress and the dislocation line tension. They give as an upper limit for $Q^{-1}_{\max}$

$$Q^{-1}_{\max}\ (\text{upper limit})\cong N_0'(l_0')^3/24\cong N'(l_0')^2/24. \tag{8}$$

As suggested by the similarity between Eqs. (3) and (8), the dislocation motion here is identical in form to that postulated by Mott and Friedel. In this case, however, thermal energy is required; presumably the bowing of a dislocation in this limit proceeds by successive thermal generation of loops which are then spread out and stabilized by the external stress as described above. It is not clear that such motion could contribute to relaxation; it might rather be expected to produce plastic strain which is in phase with the applied stress at higher temperatures than that at which the internal friction maximum corresponding to the "one-step" process occurs. We, however, consider both limits.

When the lower limit for $Q^{-1}_{\max}$ applies, Eq. (7) indicates that recovery phenomena which change the height of the Bordoni peaks can do so only by altering

[15] A. Seeger, H. Donth, and P. Pfaff, Discussions Faraday Soc. 23, 19 (1957).

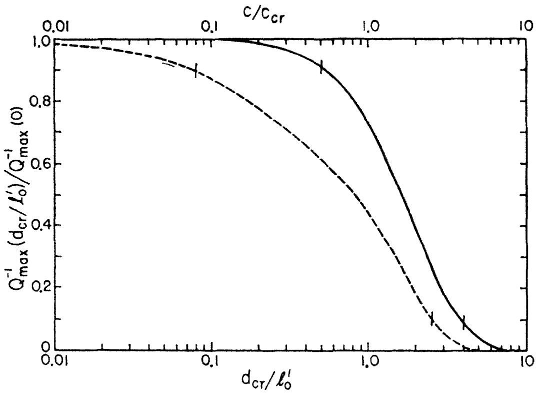


FIG. 5. Variation with pinning point density of the relative relaxation strength for the Seeger relaxation mechanism in the low damping limit. The solid curve illustrates the most rapid cutoff which could occur given a sufficiently high loop density. The dashed curve in the result of a numerical calculation based upon data appropriate to the major Bordoni peak in copper.

the density of active dislocations and not by changing the mean distance between pinning points. We expect this to be true only so long as l_0' exceeds d_{cr}, the critical loop length for expansion by the external stress as previously mentioned. Segments of length less than d_{cr} cannot be expected to contribute to relaxation. These remarks may be made more explicit by assuming a random distribution of pinning points and using the Koehler[16] segment length distribution to write

$$\frac{Q^{-1}{}_{\max}(d_{cr}/l_0')}{Q^{-1}{}_{\max}(0)} = \int_{d_{cr}}^{\infty} \frac{l}{(l_0')^2} e^{-(l/l_0')} dl = (1+d_{cr}/l_0')e^{-(d_{cr}/l_0')}. \quad (9)$$

The solid curve in Fig. 5 is a plot of $Q^{-1}{}_{\max}(d_{cr}/l_0')/Q^{-1}{}_{\max}(0)$ vs (d_{cr}/l_0') as given by Eq. (9). The concentration of pinning points may be taken as $c=(1/l_0')$, and, if we write $c_{cr}=(1/d_{cr})$, we may also plot the normalized relaxation strength as a function of (c/c_{cr}). This has been done in Fig. 5. As indicated by the vertical bars on the solid curve, the relative relaxation strength drops from 0.9 to 0.1 in a relatively narrow range of pinning point density extending from 0.5 c_{cr} to 4 c_{cr}. These results suggest that, when damping by the Seeger mechanism is governed by Eq. (7), any changes of the peak height should be attributed to a change of the density of active dislocations. It is unlikely that the pinning point density would lie within the narrow range over which it would be effective in altering the relaxation strength.

The calculation above assumes that no dislocation segments of length less than d_{cr} participate in relaxation, and that all segments of greater length do participate fully. A more rigorous result may be obtained by noting that, on parallel aligned dislocations, the equilibrium density of double kinks or loops having a length between λ and $\lambda+d\lambda$ is given[17] by $n(\lambda)d\lambda$, where

$$n(\lambda)=Ke^{(\gamma/\lambda)}. \quad (10)$$

The quantities K and γ depend upon temperature, the Peierls stress and the elastic and crystallographic properties of the specimen material. They are evaluated explicitly in reference 17. The density of loops effective in activating a segment of length l is then

$$\rho(l)=\int_{d_{cr}}^{l} n(\lambda)d\lambda, \quad l>d_{cr}$$
$$=0, \qquad l<d_{cr}. \quad (11)$$

The probability that such a segment will participate in relaxation is, assuming a random distribution of loops,

$$P(l)=1-\int_{l}^{\infty} q[\rho(l)]^2 e^{-q[\rho(l)]}dq, \quad (12)$$

where q is a variable of integration. Equation (9) then takes the modified form

$$\frac{Q^{-1}{}_{\max}(d_{cr}/l_0')}{Q^{-1}{}_{\max}(0)} = \int_{d_{cr}}^{\infty} \frac{l}{(l_0')^2} P(l) e^{-(l/l_0')} dl. \quad (13)$$

The integral in Eq. (13) has been evaluated numerically using data appropriate to the major Bordoni peak in copper[18] for the determination of K, γ, and d_{cr}. The result is the dashed curve in Fig. 5. Here the normalized damping varies between the limits 0.9 and 0.1 over the concentration range from 0.08 c_{cr} to 2.5 c_{cr}, representing a change of pinning point density by a factor of about 30. We conclude that changes in the height of the Bordoni peaks must probably still be attributed to an altered density of active dislocations if the Seeger relaxation mechanism as described by Eq. (7) is presumed to be operative.

Comparison of the plots of $(\Delta E/E)_B$ and $(\Delta E/E)_K$ vs annealing temperature suggests, as previously noted, that the step-like changes centered about 130° and 280°K are closely related. We assert further that these steps can probably be correlated with annealing stages II and III which are observed in the recovery of electrical resistivity of copper after cold work or irradiation at low temperatures. It is here assumed that, lacking specific evidence to the contrary, the irreversible change of different physical parameters occurring in the same temperature range represent different manifestations of the same recovery process. The occurrence of stages II and III after low-temperature irradiation of annealed copper would appear to rule out their explanation in terms of recovery processes involving dislocation re-

[16] J. S. Koehler, *Imperfections in Nearly Perfect Crystals* (John Wiley & Sons, Inc., New York, 1952), p. 197.

[17] A. Seeger and P. Schiller, Acta Met. **10**, 348 (1962).

[18] D. O. Thompson and D. K. Holmes, J. Appl. Phys. **30**, 525 (1959). See Table III, p. 537.

arrangement. Thus, in accord with the assumption set forth immediately above, we conclude that the annealing of the Bordoni peaks which we observe cannot be satisfactorily explained in these terms either. This picture is complicated by the fact that annealing stage II, as observed in irradiated copper, is very broad and extends from roughly 90° to 200°K. Recovery in this temperature range after cold work at lower temperatures is, on the other hand, much more abrupt. Further evidence against dislocation rearrangement in this temperature range is provided, however, by strain-aging experiments[19] in which the stress-strain curve after aging shows first a yield point, and then a smooth continuation of the curve prior to aging. This suggests that the dislocation network is unaltered by aging, but that point-defect pinning does occur and causes a yield point. It therefore appears unlikely that Seeger's "one-step" relaxation mechanism, in which the relaxation strength is proportional to the density of active dislocations, is operative.

When the upper limit for $Q^{-1}{}_{\max}$ according to Seeger's mechanism is applicable, we expect the relaxation strength to decrease with increasing pinning point concentration in all ranges through reduction of l_0'. This conclusion, valid for constant N', follows from Eq. (8). The coincident step-like decreases in magnitude of $(\Delta E/E)_K$ and $(\Delta E/E)_B$ might thus be plausibly explained in terms of simultaneous reduction of l_0 and l_0', for constant N and N'. This reduction could be associated with recovery processes in which the concentration of pinning points on both classes of dislocation lines is increased by migration of point defects thereto. This simple picture is difficult to accept, however, because of the contradictory behavior of $(\Delta E/E)_K$ and $(\Delta E/E)_B$ after each step. The continued decrease in magnitude of $(\Delta E/E)_K$ between steps implies a continued decrease of l_0. The increase of $(\Delta E/E)_B$ between steps, on the other hand, indicates an increase of l_0' in these annealing ranges. This difficulty is accentuated by observation of the regrowth of the major peak at all annealing temperatures above 200°K. The decrease of $(\Delta E/E)_B$ at 280°K is not accompanied by a reduction in the height of the major peak although it is narrowed. Thus, even within the class of dislocations responsible for the Bordoni peaks, the mechanics of pinning may differ. To remove these inconsistencies it appears necessary to suppose that clustering of point defects, with a resultant increase of l_0', could occur on some of the dislocations in the class contributing to relaxation. We must suppose further that such clustering does not occur on other dislocations within this class or on those responsible for the Köster effect at 4.2°K.

The preceeding discussion indicates that the explanation of our annealing data must be in terms of the migration of point defects to dislocations. It also suggests that the Seeger mechanism, if operative at all, cannot account satisfactorily for the entire dislocation relaxation spectrum but at best only for some portion thereof. We are therefore led to consider relaxation mechanisms in which the interaction between point defects and dislocations plays a more direct role. Consider, first without reference to a specific model, the thermally activated displacement of a dislocation pinning point between two or more minimum energy configurations. Binding occurs in the vicinity of the defect pin with displacement taken to be in a direction normal to the dislocation line. The height of the barrier separating the minima is governed by detailed properties of the defect-dislocation interaction. From this general description of the relaxation mechanism two conclusions emerge:

(1) It would not be necessary to assume that the dislocations responsible for relaxation belonged to a class different from those contributing to the Köster effect at 4.2°K. The plastic strain associated with athermal bowing of dislocation segments between stable pinning points at 4.2°K would be supplemented, at temperatures where relaxation could occur, by strain contributed by the superposed motion of the dislocation pinning point between minimum energy configurations. The situation described is illustrated in Fig. 6.

(2) An increase in the relaxation strength of the Bordoni peaks, accompanied by a decrease in magnitude of the modulus defect associated with the Köster effect, could result from the migration to dislocations of the particular point defect associated with relaxation. This would increase the density of stable pins at 4.2°K, thus reducing the magnitude of the Köster effect, but would also increase the density of sites at which relaxation could occur and hence the magnitude of the Bordoni peak.

We note, as a specific example, the model based upon the thermally activated motion of paired partial dislocations between vacancy pinning points[3] which was mentioned in the introduction. The attenuation of the Bordoni peaks could, for this mechanism, be

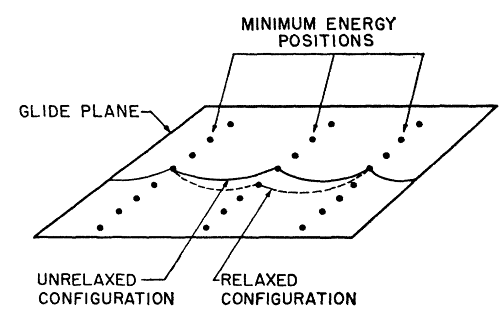


FIG. 6. Illustration of a relaxation mechanism based upon the interaction of point defects with dislocations, showing the simultaneous operation of the bowing mechanism to which the 4.2°K modulus defect is attributed.

[19] Howard K. Birnbaum and Floyd R. Tuler, J. Appl. Phys. 32, 1402 (1961).

attributed to the migration to dislocations of defects other than vacancies; attenuation would result from shortening the average free-segment length on either side of all preexisting vacancy pins. Subsequent regrowth of the peak could be attributed to the migration of more vacancies to dislocation lines.

As an alternative one might consider a mechanism of the type proposed by Kessler[20] to account for a relaxation peak in germanium. The peak was attributed to the "drag" of a dilute vacancy atmosphere by moving dislocations as a result of the elastic interaction between them. Presumably, in an interpretation of the Bordoni peaks in these terms, one would not be restricted to consideration of the vacancy as the active point defect.

Relaxation mechanisms of the type discussed above presuppose the existence of relatively stable point defect-dislocation configurations. Evidence indicating that such configurations are possible is provided by irradiation experiments on annealed copper. Thermal depinning of dislocations with the resultant restoration of the pre-irradiation modulus defect and internal friction is accomplished by annealing only when temperatures ranging from 200° to 500°C are employed.[21]

It is clear that interpretation of our data in terms of a particular model involving point defect-dislocation interaction depends upon a specification of the type and distribution of point defects on the dislocations immediately after low-temperature deformation, and particularly upon a reliable interpretation of all stages observed in recovery experiments. Since both are currently lacking, consideration of any specific model must be regarded as speculative. The two conclusions cited earlier, based upon general considerations of relaxation mechanisms involving point defect-dislocation interaction, are more firmly founded. Their application to our experimental results is thought to lend substantial support to the view that some form of interaction with point defects constitutes the basis of dislocation relaxation.

V. SUMMARY

Measurements of the internal friction and Young's modulus of copper, after plastic deformation at 4.2°K and subsequent isochromal anneals at temperatures between 100° and 360°K, have been correlated with recovery phenomena observed in other experiments. The relationship between dislocation relaxation phenomena, as illustrated by the Bordoni peaks, and the Köster effect or recoverable modulus defect at 4.2°K has been examined in detail; the recovery characteristics of each have been considered.

We assume initially that the 4.2°K modulus defect is attributable to the stress induced bowing out of dislocations between stable pinning points, without the aid of thermal energy. It is then pointed out that, if dislocation relaxation at higher temperatures is to be attributed to intrinsic properties of the dislocation network, it is necessary to ascribe the recoverable modulus defect at 4.2°K and the higher temperature relaxation to different classes of dislocations. The recovery data indicates in addition that these two classes differ in the sensitivity of attached point defects to clustering or to loss at other sinks.

It is suggested that both of the rather detailed assumptions which must accompany an explanation of dislocation relaxation in terms of properties of the network alone could be dropped if the interaction between point defects and dislocations is taken to be the basis of the relaxation mechanism. This simplification is thought not to depend upon the details of the mechanism.

Finally, specific models based upon this type of interaction are mentioned. It is concluded that no support for a detailed model can be derived from experiments of the type reported here without an unambiguous interpretation of recovery experiments generally. This work does, however, offer substantial support to the view that some form of interaction with point defects is an essential feature of dislocation relaxation.

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[20] J. O. Kessler, Phys. Rev. **106**, 654 (1957).
[21] A. Sosin, Acta Met. **10**, 390 (1962).