several specimens, two sets of slip lines have been observed intersecting at angles corresponding to 111 planes. The slip lines disappear when the specimen is etched. Laue patterns of the deformed regions of bent specimens show asterism, but the material retains its single crystal character over the range of strain observed. When larger specimens are bent drastically, audible clicks can be heard.

At the lower end of the temperature range, the deformation exhibits a time delay or induction period of the order of minutes. When a constant load is applied at constant temperature, the time delay decreases with increasing temperature, as shown in the curves of Fig. 2. Each curve represents the angular deflection *versus* time for a bar, 1 mm square, clamped at one end and loaded with 100 grams at a point 1 cm distant. Above 600°C, the time delay was too short to be observed with the loading method used.

The electrical behavior of the deformed material is of interest, particularly in view of results obtained by other workers on the effect of heat treating germa-

nium.[1,2] It has been found that *n*-type germanium heated above 550°C and then quenched to room temperature is left with a permanent increase in resistivity and a decrease in carrier lifetime, probably due to the introduction of acceptor-type defects. A similar effect is found with deformed germanium. For bent specimens, the resistivity is higher in the region of bending, and the lifetime of photo-injected carriers is considerably lower. The germanium, originally *n*-type, was still *n*-type after deformation. Work is in progress to determine possible relations of defects introduced by deformation to those introduced by heat treatment.

Silicon single crystals have also been plastically deformed at temperatures above 900°C with somewhat higher stresses than for germanium. Slip lines have been observed along [111] planes. The writer is indebted to F. H. Horn for furnishing silicon crystals.

[1] Fuller, Theurer, and van Roosbroeck, Phys. Rev. 85, 678 (1952).
[2] W. DeSorbo and W. C. Dunlap, Jr., Phys. Rev. 83, 879 (1951).

The Plasticity of Silicon and Germanium

FREDERICK SEITZ
University of Illinois, Urbana, Illinois
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A qualitative analysis of Gallagher's measurements of plastic flow in germanium and silicon is made. It is reasonable to suppose the slip occurs by motion in {111} planes of dislocations having a Burger's vector equal to an allowed translation in the (1$\bar{1}$0) direction. It is suggested, following current thought, that the observed temperature-dependent incubation time for plastic flow is related to the freeing by thermal fluctuations of locking points of the Cottrell type associated with the broken bonds which occur along nearly pure screw dislocations that have a small edge component. It is proposed that foreign atoms reside at the broken bonds in normal materials. A discrepancy in the coefficient of the Boltzmann factor indicates, however, that these concepts cannot be applied in the most simple form. The analogy between Gallagher's measurement and those of Kramer and Maddin in β-brass is pointed out.

I. DISLOCATIONS IN LATTICE

THE recent work of Gallagher[1] on plastic flow in germanium and silicon provides a remarkable insight into the ductility of materials in general, as well as of valence crystals. In brief Gallagher has found that germanium and silicon become ductile in the vicinity of 500°C and 900°C, respectively. Slip occurs in {111} planes with the production of well-developed slip bands for stresses somewhat above 10^8 dynes/cm². One of the most interesting features of the flow process is the occurrence of an incubation time. From Gallagher's data for germanium one concludes tentatively that this incubation time τ satisfies an equation of the following kind:

$$\tau = \tau_0 \exp(Q/RT), \qquad (1)$$

where $\tau_0 = 10^{-5}$ second and Q is about 28,000 cal/mole.

[1] C. J. Gallagher, Phys. Rev. 88, 721 (1952).

It is interesting to see to what extent Gallagher's observations can be explained in terms of the present concepts of dislocation theory. The simplest assumption to make is that the active dislocations move in slip planes of the {111} type and possess a Burger's vector equal to an allowed translational distance of the (1$\bar{1}$0) type in such planes. Neighboring slip planes of the {111} type, which are also common cleavage planes, are linked by valence bonds which run normal to the planes. Since these bonds would be broken and not reformed if a dislocation whose Burger's vector is not an allowed translation were to pass across the slip plane, it seems highly unlikely that the active dislocations break into partial components[2] as is possible in close-packed cubic or hexagonal lattices, where the

[2] R. D. Heidenreich and W. Shockley, *Conference on the Strength of Solids* (Physical Society, London, 1948).

energies of twin surfaces are relatively low. In this respect valence crystals probably resemble the simple salts, such as the alkali halides.[3]

A pure Taylor-Orowan dislocation of the type assumed above can be formed, geometrically, by removing two neighboring {1Ī0} half-planes which contain an equal mixture of the two types of atom in the unit cell, much as in the alkali halides. There will be a row of broken bonds at the edge of each of the half-planes removed. Screw, or Burger's dislocations, can be formed without generating broken bonds but by bending existing bonds instead. The screw axis of such dislocations will, in the ideal case, run through the familiar channels of hexagonal cross section characteristic of the diamond lattice. A mixed Burger's and Taylor-Orowan dislocation will have a linear density of broken bonds along the dislocation line proportional to the degree of Taylor-Orowan character.

II. LOCKING POINTS

We may expect the unbonded atoms associated with the Taylor-Orowan character to provide ideal condensation points for foreign atoms which like to form valence bonds. These regions should provide locking points of the type proposed by Cottrell and Koehler for many atoms, including those such as hydrogen, carbon, and nitrogen which may enter interstitially. Interstitial atoms or vacancies of the parent material may also be bonded at such locking points.

It is natural to suppose that the observed slip bands are produced by Frank-Read generators which are initially locked because foreign atoms are found at points along the generating dislocation, where broken bonds occur. The stress required to free the locking points without the aid of thermal fluctuations is so great that it can not be applied without rupturing the crystal, whence the material is brittle at low temperatures. At higher temperatures, however, thermal fluctuations may supply sufficient energy to break the dislocations free of the locked points and permit the generators to operate. According to this view, the activation energy Q in (1) is the energy required to free the dislocation from the locked point.

From the information available at the present time, it is not possible to say whether the incubation time for start of plastic flow is strongly dependent on the time the specimen is held at temperature without application of stress. For the purpose of the present discussion we shall assume that the incubation time is essentially independent of such holding time and shall conclude tentatively that the freeing of bonds requires the application of shear stress as well as temperature. This in turn suggests that the dislocations which can be set in motion are those for which the separation of locking points is sufficiently great that the dislocation is able to move at least one atom distance under the applied

stress when the lock is broken by thermal fluctuations. These presumably are dislocations in which the locked points are separated from one another by distances of the order of 100 atom spaces or more. This suggests further that the active generators are nearly pure screw dislocations with a small component of edge type. It is conceivable that a nearly perfect edge-type dislocation would not be locked if the material were very free of foreign atoms. However, it seems most reasonable, at present, to assume that there is a sufficient number of foreign atoms present to lock most of the unbonded ends.

Since pure screw dislocations have no unbonded atoms, one might ask why they cannot act as generators without the need of an incubation time. Three replies seem possible: (1) Such dislocations may be exceedingly rare. (2) Any stationary dislocation running perfectly parallel to a crystallographic direction may be difficult to free because of the high value of the Peierls-Nabarro stress. Some edge admixture may be a valuable addition to a screw dislocation because mixed dislocations are easier to move.[4] (3) Even pure screw dislocations may be locked by impurities, in spite of the absence of broken bonds.

It seems difficult to say whether or not the migration along the dislocation of the broken bonds associated with the edge component is a process requiring thermal activation energy. At present the writer is inclined to believe that this motion does not require thermal activation energy provided the dislocation is moving sufficiently rapidly. This does not mean that it is necessary for the dislocation line to move normal to itself with a velocity nearly equal to that of sound. The phase-like motion of a broken bond may be more rapid than the dislocation, in a nearly pure screw dislocation, since the bond moves essentially parallel to the dislocation line. It is possible that a thermal activation energy or a large stress concentration is needed to start the phase-like motion of the broken bond; however, this activation energy may be considerably smaller than the value Q in (1). Thus the time required to set the broken bond in motion would become the observable, limiting time for initiation of flow only in very pure materials.

If the typical Frank-Read generator is 10^4 atom distances long and contains 100 locking points, one might suppose that τ_0 in (1) would be of magnitude $100 \times 10^{-13} = 10^{-11}$ second. The much larger observed value, namely, 10^{-5} second, seems to propose an interesting problem. It may be noted that τ_0 would be about 10^{-8} second if the relaxation time involved were that required for locking atoms to diffuse out of the area, about 10^4 atoms on an edge, over which the generators presumably act. Q would then represent an activation energy for diffusion. In this case, however, as with other conceivable mechanisms, one might expect the incubation time to be dependent on the time the specimen is

[3] F. Seitz, Phys. Rev. **80**, 239 (1950).

[4] See, for example, Pittsburgh Symposium on Plastic Deformation (Government Printing Office PB104,604), page 13.

held at temperature before applying the load, in contradiction with the assumptions made previously.

Cottrell and Bilby[5] have called attention to similar delays in initiating flow in polycrystals of the iron-carbon system and have suggested that it may be necessary to await the formation of avalanches of moving dislocations before flow becomes macroscopically visible. One might hope to eliminate delays of this type by using single crystals with relatively free glide planes; however, this possibility remains for investigation.

III. EXPERIMENTS OF KRAMER AND MADDIN

The experiments of Kramer and Maddin[6] on time delay of plastic flow in β-brass seem to be at least qualitatively related to the foregoing work. These investigators have measured the stress required to initiate plastic flow in α-brass, aluminum, and β-brass when the load is applied for relatively short times varying from 8×10^{-5} second to 6×10^{-3} second, and for several temperatures ($25°C$, $-50°C$, $-190°C$). No measurable effects of the high rate of strain were observed in the face-centered cubic metals aluminum and α-brass. In these materials the results seem to conform closely to static measurements.

At the temperatures of $-50°C$ and $-190°C$ it was found necessary to increase the stress in the case of β-brass when the time for which it was applied was less than a critical value τ_c, the increase being a continuous function of time. Apparently if the load is applied for times longer than τ_c, the shear stress required to induce plastic flow is closely the same as the static value. Within the accuracy of the measurements of Kramer and Maddin, the critical stress for plastic flow does not appear to be strongly temperature dependent when the time of application is longer than the τ_c. This result seems to be in contrast with those for the face-centered metals, where there is a gradual rise of yield stress with decreasing temperature.

It is possible to compare the results obtained by Kramer and Maddin for β-brass with those obtained

by Gallagher by making the following assumptions. It will be assumed that the generators for plastic flow in β-brass are locked, as in the case for those of silicon and germanium. As long as the stress is applied for time greater than τ_c, thermal fluctuations are adequate to free the generating dislocations from the pinning points. An increase in applied stress is necessary to decrease the activation energy for freeing the generating dislocations from the pinning points when the time of application of the load is less than τ_c. With this assumption, τ_c becomes essentially equivalent to τ in Eq. (1), providing a measure of the time required for the dislocations to become free without the need for lowering the activation barrier. From the data given in Fig. 10 of the paper by Kramer and Maddin, one may conclude, in rather rough approximation, that $\tau_c = 8.0 \times 10^{-3}$ seconds at $-190°C$ and $\tau_c = 4.0 \times 10^{-3}$ at $-52°C$. Using an equation of the form

$$\tau_c = \tau_{0c} \exp(Q/RT), \qquad (2)$$

one finds Q to be of the order of 180 cal/mole and τ_{0c} 3×10^{-3} second. The relation (2) is apparently only roughly in agreement with the observations. For example, according to (2) the minimum possible value of τ_c would be about 3×10^{-3} second if the parameters have the values just given. Actually Kramer and Maddin did not seem to observe a critical time at room temperature even when the load was applied for timest as short as 6.2×10^{-4} second. Nevertheless, it is clear that Q is approximately 100 times smaller in the case of β-brass than in the cases of silicon or germanium. It is also clear that an appropriate value of τ_{0c} is much larger than 10^{-11} second and is actually comparable to the value found in (1). Conversely, the very large values of Q for silicon and germanium make it seem unlikely that the incubation time observed by Gallagher in these solids can be diminished measurably by the application of stresses only a factor two or three larger than those he employed.

The writer is indebted to Dr. J. D. Eshelby and Dr. William Heller for discussions of this topic. He is particularly obligated to C. T. Gallagher for the privilege of seeing his manuscript prior to publication.

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[5] A. H. Cottrell and B. H. Bilby, Proc. Phys. Soc. (London) A62, 49 (1949).
[6] I. R. Kramer and R. Maddin, J. Metals 4, 197 (1952).