

We shall conclude by pointing out that the phenomenon described above is typical only of compensated metals because the term $\pm 2h$ in α_{\pm} , which results in a difference between the amplitudes of the GK oscillations, appears due to the addition $\sigma_{\pm}^{(1)} + \sigma_{\pm}^{(2)}$ (the upper index of the conductivity denotes the edge along which the required function is calculated). In the case of uncompensated metals this difference vanishes and for compensated metals it is governed by the Hall conductivity of a local group of carriers.

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Investigation of the mechanism of the motion of charged dislocations in ZnSe

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The motion of electrically charged dislocations in ZnSe is investigated by measuring the dislocation currents. It is shown that the kinetics of the motion of the dislocations is determined by thermally activated surmounting of Peierls barriers, whose heights are found to be directly proportional to the linear density of the dislocation charge. A physical model that accounts for the appearance of the Peierls barriers and for a number of plastic effects is proposed.

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INTRODUCTION

There are two main reasons why it is of interest to study the regularities in the motion of dislocations in $A^{II}B^{VI}$ compounds (CdS, ZnS, ZnSe, CdSe, ZnO, ZnTe, and others). First, knowledge of the laws of motion of these dislocations may provide a key to the understanding of such phenomena observed in these compounds as the photoplastic effect (PhPE) and the electroplastic effect (EPE).^[1-3] Second, as has been shown in a number of studies,^[4-8] the dislocations in these compounds bear a considerable electric charge (of the order of one electron per interatomic distance), and this, it may be supposed, will lead to features of the motion of dislocations that are not observed in other materials.

Unfortunately, certain characteristics of $A^{II}B^{VI}$ compounds have prevented the use of familiar methods to study the laws of motion of dislocations in them. For example, the tendency of freshly introduced dislocations to remain fixed for times of the order of several minutes makes it impossible to use the etch pit method for recording the motion of individual dislocations. On the other hand, methods involving macroscopic deformation suffer from the disadvantage that in $A^{II}B^{VI}$ compounds the large plastic deformation that is necessarily accumulated during the course of such experiments alters the parameters describing the motion of the dislocations.

We have recently developed a method for recording plastic deformations by measuring the dislocation currents.^[9] We feel that this method will make it possible to overcome the difficulties mentioned above; in addition, this method makes it possible continuously to follow such an important parameter of the dislocations in $A^{II}B^{VI}$ compounds as their electric charge during the course of the deformation.

TECHNIQUE

In this study we used ZnSe single crystals with the sphalerite structure, grown from the melt and having a dark resistivity of 10^8 to 10^{13} $\Omega \cdot \text{cm}$ and a dislocation-nucleus density of $\sim 10^5$ cm^{-2} . Deformation specimens measuring $6 \times 4 \times 1.5$ mm were cut from ingots with a diamond saw, were ground smooth with abrasive powders, and were polished with diamond paste. Then the specimens were etched with $\text{CrO}_3 + \text{HCl}$ to remove the work hardened layer some 50–100 μm thick from the surfaces. Either liquid (In+Hg) or indium electrodes were fixed to the large (4×6 mm) surfaces of the specimens by ultrasonic soldering in order to record the dislocation currents. The specimens were deformed by compression along the long (6 mm) dimension. The 6×1.5 mm face was a $(1\bar{1}0)$ plane, and the (111) plane, in which the dislocations move, is perpendicular to this plane and makes an angle of 45° with the compression axis. As was previously shown,^[8] with such a deforma-

ation scheme the dislocation currents in ZnSe are due to the motion of $60^\circ \alpha$ dislocations, so the experimental results presented below pertain to the motion of just these dislocations.

Figure 1 is a schematic diagram of the apparatus. The essence of the method for measuring the plastic deformation ϵ_p from the magnitude of the dislocation currents I_d is the use of the relation^[1,6,9] $\dot{\epsilon}_p \propto I_d$. The small deviations from this proportionality discovered by Kirichenko *et al.*^[8] were taken into account in processing the data. The apparatus diagramed in Fig. 1 makes it possible to register deformation rates $\dot{\epsilon}_p$ and deformation rate differences $\Delta \dot{\epsilon}_p$ down to $\sim 10^{-8} \text{ sec}^{-1}$. The measuring procedure was as follows: The specimen was first deformed at a constant low rate (usually $10 \mu\text{m}/\text{min}$) until $\epsilon_p \approx 0.5\text{--}10\%$ in order to produce the dislocation structure to be investigated. During this stage the dual trace x - y plotter 7 recorded the $\sigma(\epsilon)$ and $I_d(\epsilon)$ curves, from which the linear charge density q of the dislocations could be determined.^[1,6,8] Then the specimen was subjected to cyclic loads σ varying between the limits 0 and σ_M , where σ_M was above the yield point σ_y . The cyclic loads were imposed at higher velocities ($100\text{--}1000 \mu\text{m}/\text{min}$) of the compressing plungers. The plastic deformation accumulated during a single cycle was only $\sim 10^{-2}\%$ and thus was much smaller than the preliminary deformation ($0.5\text{--}10\%$). In recording the dislocation currents during the cyclic loading the piezocurrents $I_p \propto \dot{\sigma}$ were compensated. During the cyclic loading, the $\dot{\epsilon}_p(\sigma)$ curve was recorded on the second plotter. The absence of a hysteresis loop on the $\epsilon_p(\sigma)$ curve as σ increases and decreases permits us to suppose that the extremely small plastic deformation ($10^{-2}\%$) accumulated in a single cycle does not alter the dislocation structure prepared in the preliminary deformation in any significant manner. Hence, in using the well known relation

$$\dot{\epsilon}_p \propto \rho_m \bar{v}_d b, \quad (1)$$

in which ρ_m is the density of mobile dislocations, b is their Burgers vector, and \bar{v}_d is their average velocity, we shall assume that

$$\rho_m = \text{const}, \quad \dot{\epsilon}_p \propto \bar{v}_d. \quad (2)$$

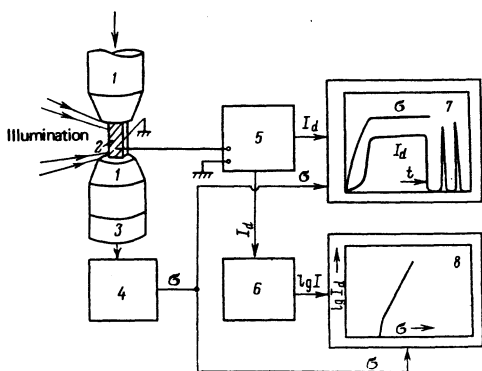


FIG. 1. Diagram of the experimental setup: 1—quartz plungers, 2—specimen with electrodes, 3—stress transducer, 4—amplifier, 5—electrometer, 6—logarithmic amplifier, 7—two-channel x - y plotter, 8— x - y plotter.

3. EXPERIMENTAL RESULTS

To clarify the mechanism of the dislocations' motion we must know the form of the function giving the velocity v_d of the dislocations in terms of the temperature T , the applied stress τ , and the electric charge density on the dislocation. Thus, we need the explicit form of the function $v_d(T, \tau, q)$ or, since $v_d \propto \dot{\epsilon}_p$ under our experimental conditions, the function $\dot{\epsilon}_p(T, \tau, q)$. This function was constructed from families of the following experimental curves

$$\dot{\epsilon}_p(\tau)|_{T, q=\text{const}}, \quad \tau(q)|_{T, \dot{\epsilon}_p=\text{const}}, \quad \tau[T, q(T)]|_{\dot{\epsilon}_p=\text{const}}.$$

As examples, we show in Fig. 2 some experimental $\dot{\epsilon}_p(\tau)|_{T, q=\text{const}}$ curves obtained by deforming ZnSe (curves 1 and 2) and ZnS (3 and 4) in darkness (1 and 3) and under illumination (2 and 4). As is evident from the figure,

$$\ln(\dot{\epsilon}_p/\dot{\epsilon}_0) = \alpha(T)\tau + f(T, q), \quad (3)$$

over a fairly wide range of deformation rates ($\dot{\epsilon}_p$ between 10^{-4} and 10^{-6} sec^{-1}), and $\alpha(T) = (\partial \ln \dot{\epsilon}_p / \partial \tau)_T$ turned out to be equal to γ/T . (That α is independent of q is illustrated in Fig. 2 by the fact that the curves recorded in darkness and under illumination are parallel, even though the values of q for the two cases are substantially different.)

The form of the function $f(T, q)$ in formula (3) was found from families of $\tau(q)|_{T, \dot{\epsilon}_p=\text{const}}$ curves, examples of which are shown in Fig. 3. As is evident from Fig. 3,

$$-aq + \gamma\tau = \text{const at } \ln(\dot{\epsilon}_p/\dot{\epsilon}_0) = \text{const}, \quad T = \text{const}. \quad (4)$$

Comparing formulas (3) and (4), we have

$$\ln(\dot{\epsilon}_p/\dot{\epsilon}_0) = (\gamma\tau - aq + E_0)/T \quad (5)$$

or

$$\dot{\epsilon}_p = \dot{\epsilon}_0 \exp [(-aq + \gamma\tau + E_0)/T]. \quad (5')$$

The coefficient γ , which has the dimensions of volume, is equal to $(20 \pm 2)b^3$ for ZnSe, $(25 \pm 4)b^3$ for ZnS, and $(50 \pm 10)b^3$ for CdS. Following established terminology, we shall call this coefficient the "activation volume"; it does not change for any of the listed materials when the specimen is illuminated and the PhPE is excited in it. In ZnS and ZnSe the coefficient γ also remains constant (within the 10–15% accuracy of its determination)

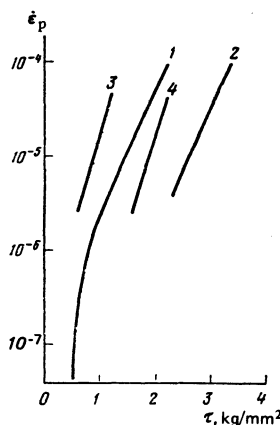


FIG. 2. Plastic deformation rate $\dot{\epsilon}_p$ vs shearing stress τ in the slip plane recorded at $T = 295^\circ \text{K}$ for ZnSe (curves 1 and 2) and ZnS (3 and 4) in darkness (1 and 3) and under illumination with light to excite the PhPE (2 and 4).

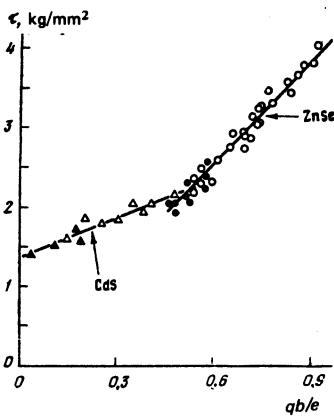


FIG. 3. Shearing stress τ for flow at $\dot{\epsilon}_p = 10 \mu\text{m}/\text{min}$ vs dislocation charge recorded at $T = 295 \text{ }^\circ\text{K}$ for ZnSe (circles) and CdS (triangles) in darkness (black symbols) and under illumination (open symbols).

when the deformation ϵ_p varies (within the range 0.5–10%). In CdS, γ decreases slowly from $\sim 50b^3$ to $\sim 30b^3$ as ϵ_p increases from 0.5% to 10%. For all three of these materials the activation volume γ begins to increase monotonically (when $\dot{\epsilon}_p < 10^{-1}$) as τ decreases (see Fig. 2).

The value of a that we obtained for ZnSe turned out to be $5.1 \times 10^{-3}(e/b)$ erg. Finally, we note that the functional dependence $\dot{\epsilon}_p(\tau, q, T)$ found in this way (formula (5)) correctly represents the experimentally observed $\tau(T)$ dependence at $\dot{\epsilon}_p = \text{const}$ (Fig. 4). The full $\tau(T)$ curves on Fig. 4 were calculated with formula (5), using the earlier determined values of a and γ and the measured values of $q(T)$. E_0 was treated as an adjustable parameter; its value for ZnSe turned out to be $0.0 \pm 0.03 \text{ eV}$.

Thus, the plastic deformation rate $\dot{\epsilon}_p$ in ZnSe is described by the usual expression for thermally activated processes (formula (5)) with the activation energy $E = aq - \gamma\tau$ and the very small activation volume $\gamma = (20-2)b^3$; moreover, γ is independent of the deformation and begins to increase at small values of τ . The importance of these facts in determining the mechanism for the motion of dislocations in ZnSe will be discussed in the following section.

Figure 5 shows the spectral dependences of the dislocation charge in ZnSe and of the magnitude $\Delta\sigma = \sigma_L - \sigma_D$

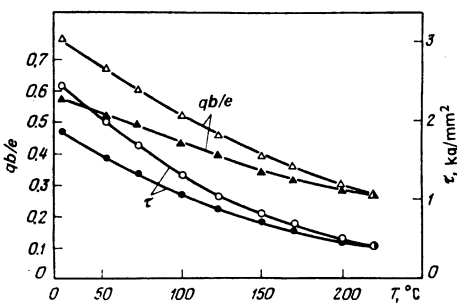


FIG. 4. Shearing stress τ for plastic flow at $\dot{\epsilon}_p = 10 \mu\text{m}/\text{min}$ (circles) and dislocation charge (triangles) for ZnSe vs temperature, recorded in darkness (black symbols) and under illumination (open symbols).

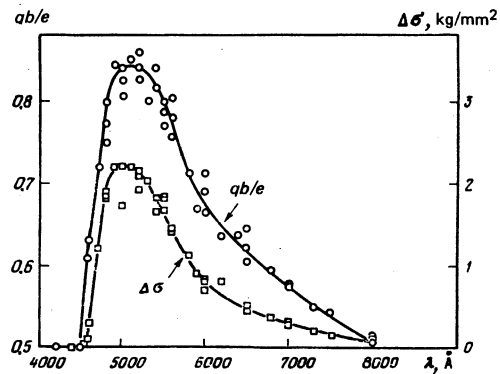


FIG. 5. Spectral dependences of the change $\Delta\sigma$ in the deformation stress and the dislocation charge in ZnSe at $\dot{\epsilon}_p = 10 \mu\text{m}/\text{min}$ and $T = 295 \text{ }^\circ\text{K}$.

of the PhPE (here σ_L and σ_D are the flow stresses in light and in darkness; $\tau = \sigma/2$ for our geometry). The identity of these spectra is striking. We feel that this identity and the identical dependences of Δq and $\Delta\tau$ on temperature (Fig. 4) and on the illumination intensity (Fig. 3) are far from accidental, but rather indicate that it is the change in the dislocation charge that causes the increase in the deformation stress (at $\dot{\epsilon}_p = \text{const}$) under illumination (the PhPE). It is precisely on this that the PhPE mechanism proposed in the following section is based.

DISCUSSION

We shall discuss the experimental results in two stages. First we shall discuss the mechanism of the motion of dislocations in $A^{II}B^{VI}$ compounds which, as we saw in the preceding section, has the same character for deformations both in light and in darkness. Then we shall consider the reasons for the changes in the quantitative parameters of the motion of dislocations under illumination, i.e., the nature of the PhPE.

Since in our case the motion of the dislocations turned out to be a thermally activated process (formula (5)), we shall not consider the viscous motion mechanisms but shall rather analyze the known thermal activation processes of dislocation slip: the Peierls mechanism, the overcoming of point defects, and the intersection of dislocations. As is well known, the primary parameter that is usually used to distinguish between these mechanisms is the size of the activation volume.^[10,11] For the Peierls mechanism γ is usually a few dozen times b^3 or even smaller; for the mechanism involving the overcoming of a "dislocation forest" it is hundreds of times b^3 ; and for the overcoming of point defects it is hundreds and thousands of times b^3 .^[11] Thus, the activation volumes observed in the present experiment correspond best to the Peierls mechanism. The independence of γ from the degree of deformation, which was observed in ZnSe and ZnS, is also characteristic of this mechanism. Also, the observed increase of γ at small values of τ is very characteristic of this mechanism.^[10]

Finally, let us make a few quantitative estimates that are usually cited in an analysis of the Peierls mechanism.

ism. Assuming the Peierls mechanism for the motion and determining the activation energy $E = aq$ ($\tau = 0$)—the energy for the formation of a double kink ($\tau = 0$)—we may estimate the Peierls stress τ_P —the stress for the unactivated motion. For a sinusoidal Peierls contour we have

$$\tau_P = \pi^2 E^2 / 16 G a_1^3 b^3 = 4.3 \text{ kg/mm}^2, \quad (6)$$

and for a quasiparabolic contour,

$$\tau_P = 16 E^2 / \pi G a_1^3 b^3 = 3.6 \text{ kg/mm}^2, \quad (7)$$

where G is the shear modulus ($G = 1.6 \times 10^{11}$ dyne/cm² for ZnSe) and a_1 is the distance between neighboring troughs of the Peierls contour. These values are close to those obtained by extrapolating the temperature dependences of τ to $T = 0$ (see Fig. 4, $q = q(300 \text{ }^\circ\text{K})$), and this, too, speaks in favor of the Peierls mechanism for the dislocation motion.

Finally, the observed long (up to 0.5 mm) straight segments of dislocations lying along low-index crystallographic directions can be regarded as an indirect indication of the presence of high Peierls barriers in $A^{II}B^{VI}$ compounds.^[12] Taking the Peierls mechanism of the motion of dislocations in ZnSe as established, let us consider the problem of the observed linear dependence of the height of the Peierls barriers (5) on the dislocation charge density q . In considering models for the surmounting of Peierls barriers by dislocations, let us first examine two published ones^[13,14] that involve the dislocation charge. Haasen^[13] assumes that the energy required for the production of a double kink on a charged dislocation is reduced by the electrostatic repulsion between the kink and the rest of the dislocation. According to this model, however, an increase of the dislocation charge under illumination would lead to softening of the crystal, i.e. to a decrease in the deformation stresses, whereas experimentally we observe hardening (the PhPE). In addition, the activation energy varies quadratically with the dislocation charge density according to this model. Pokrovskii and Petukhov^[14] assume that the formation of a double kink in a charged dislocation is facilitated by the electrostatic interaction of the dislocation with ionized impurities. For the following reasons, this model does not seem to us to be basic (decisive) for the PhPE: According to this hypothesis an increase in the dislocation charge should lead to softening, rather than to the observed hardening. One could obtain hardening under illumination within the limitations of this model by assuming all the charged impurities near the dislocation to be almost completely neutralized as a result of capturing minority carriers (electrons and holes). In that case the correspondence between Δq and $\Delta \tau$, which is always observed, would have to be regarded as accidental. The conclusion that the activation energy for the motion of the dislocations decreases when ionized point defects are introduced, which follows from this model, is also in direct contradiction with known data on $A^{II}B^{VI}$ compounds.^[15,16]

We can see still another possible way to relate the number of electrons captured on broken dislocation bonds to the height of the periodic potential contour of the lattice along which the dislocation is moving. In-

deed, since a large number of bonds must be disrupted and then joined to broken ones in order for a dislocation to move into the neighboring trough of the potential contour, the energy required for such motion may well depend on the state of the broken bonds, i.e. on whether they are held by one electron or by two. At present, however, we see no prospects for making quantitative calculations on the basis of this model.

Another possibility seems to us to be much simpler and more natural. Let us consider the atomic model of 60° dislocations in the sphalerite structure illustrated in Fig. 6. As is evident from the figure, when charged dislocations move in (11) planes they are moving parallel to planes consisting of rows of like atoms which, because of the partially ionic character of the bonds, are electrically charged. Because of the periodic disposition of the rows of like ions, the dislocation will move in a periodic electrostatic field. Let us estimate the amplitude of this spatially periodic field. First let us consider only one plane, consisting of parallel charged lines a distance a_1 apart and a charged dislocation moving at a distance a_2 from this plane. Let x be the coordinate specifying the position of the dislocation; then the electrostatic energy of the interaction between the dislocation and the plane will be

$$U = \frac{2qq'}{\epsilon} \sum_{n=-\infty}^{+\infty} \ln \left(\frac{R_{sc}}{[a_2^2 + (x + a_1 n)^2]^{1/2}} \right), \quad (8)$$

where q and q' are the linear charge densities on the dislocation and the rows of ions, respectively, ϵ is the dielectric constant, and R_{sc} is the screening radius ($R_{sc} \gg a_1, a_2$). The force acting on the dislocation will be

$$F_x = - \frac{\partial U}{\partial x} = \frac{2qq'}{\epsilon a_1} \sum_{n=-\infty}^{+\infty} \frac{n + x/a_1}{(n + x/a_1)^2 + (a_2/a_1)^2}. \quad (9)$$

Using the expansion of meromorphic functions in series of simple fractions, we sum the series in (9), obtaining

$$\begin{aligned} F_x &= \frac{q'q}{\epsilon a_1} \sum_{n=-\infty}^{+\infty} \left[\frac{1}{n + x/a_1 + ia_2/a_1} + \frac{1}{n + x/a_1 - ia_2/a_1} \right] \\ &= \frac{q'q\pi}{\epsilon a_1} \left[\text{ctg} \left\{ \left(\frac{x}{a_1} + i \frac{a_2}{a_1} \right) \pi \right\} + \text{ctg} \left\{ \pi \left(\frac{x}{a_1} - i \frac{a_2}{a_1} \right) \right\} \right] \\ &= \frac{2q'q\pi}{\epsilon a_1} \frac{\sin(2\pi x/a_1)}{\text{ch}(2\pi a_2/a_1) - \cos(2\pi x/a_1)}. \end{aligned} \quad (10)$$

In view of the fact that $a_2/a_1 \approx 1$, we shall have $|\cosh(2\pi a_2/a_1)| \gg |\cos(2\pi x/a_1)|$, and Eq. (10) will represent an almost sinusoidal contour. Since the amplitude of the contour falls off exponentially with increas-

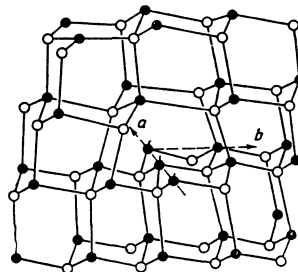


FIG. 6. 60° α dislocation in the sphalerite lattice: ●—cationic atoms, ○—anionic atoms, b —Burgers vector, a —direction of the dislocation line (Ref. 25).

ing distance of the dislocation from the plane of charged ions, the contribution from the next plane will be smaller by a factor of $\cosh 2\pi \approx 270$, and so on. Hence it is sufficient to take only the nearest plane into account.

Using the definition of the "Peierls stress" τ_P as the stress necessary for the motion of the entire dislocation as a whole, we obtain

$$\tau_P b = F_z |_{\max}, \quad (11)$$

and for ZnSe,

$$\tau_P = 2q' q \pi / \epsilon a, \quad \text{ch } 2\pi = 4.9 \text{ kg/mm}^2. \quad (12)$$

In this estimate we have used the value of Berlincourt *et al.*^[17] for the effective charge on the atoms in ZnSe and the typical value $qb/e = 0.6$ for the charge on a dislocation in the dark. We note the beautiful agreement between the estimate of Eq. (12) and the estimate obtained from the experimental data in Eq. (6).

Let us briefly set forth our entire idea. The motion of charged dislocations in ZnSe is well described by the Peierls mechanism, and in ZnSe virtually the entire height of the Peierls barrier is determined by the electrostatic interaction of the charged dislocation with rows of like charged lattice ions past which the dislocation moves. The dislocation charge density q is the most important parameter in this interaction, and any change in q brings about a change in the mobility of the dislocations and in the deforming stress τ . This mutual connection can be traced in the linear relationship between q and τ (Fig. 3) and in the identical temperature (Fig. 4) and spectral (Fig. 5) dependences of τ and q .

As was shown in Ref. 8, the charge on moving dislocations in ZnSe is made up of electrons collected from trapping levels by the dislocation as it moves. The more such electrons there are on trapping levels, the higher will be the charge, and the lower the mobility, of the dislocations. This accounts for the relation between the donor concentration in the crystal and the observed flow stresses.^[15,16] Irradiating the crystal with interband light increases the concentration of electrons on trapping levels and thus also increases the dislocation charge q . Incidentally, since electrons and holes are produced in equal numbers under excitation by light and the electrons are almost all captured on trapping levels and the holes on r centers, so that $n_t \approx p_r$, the relation between p_r and the flow stress that we reported earlier^[18] turns out to be understandable.

The above considerations are equally applicable to observations of the PhPE in other $A^{II}B^{VI}$ compounds with the sphalerite (ZnS) and wurtzite (CdS and CdSe) structures since the geometry of the (0001) slip plane in wurtzite is similar to that of the (111) plane in sphalerite. A relation between q and τ is also observed in all these compounds.

Let us see how the proposed model deals with other known plastic effects observed in $A^{II}B^{VI}$ compounds: the negative photoplastic effect (NPhPE),^[3] the after effect (AE),^[19] and the electroplastic effect (EPE).^[2]

The negative photoplastic effect. Single crystals with the wurtzite structure can be so oriented that plastic

deformation will take place by slipping of dislocations in the (10 $\bar{1}$ 0) prismatic plane of the first kind.^[20,21,3,7] The edge dislocations lying along the [0001] sixfold axis in this plane are charged.^[7] As was shown in Ref. 22, however, there are chains of ions of alternating sign (+--+-. . .) parallel to these dislocations, and the periodic electrostatic field described above does not arise in connection with their motion. Hence one cannot expect hardening to result from an increase in the dislocation charge under the action of light. In fact, the first studies of the effect of illumination on the mobility of dislocations in this slip plane showed no PhPE (ZnO,^[21] and CdS^[20]). Later^[3] it was found that the mobility of dislocations in these slip planes in certain CdS single crystals can be increased by illumination, i.e., one obtains softening (the NPhPE). In the absence of electrostatic Peierls barriers, an increase in the dislocation charge can lead to such softening according to models proposed by Haasen^[13] and Pokrovskii.^[14]

The electroplastic effect. Considerable mechanical hardening (the EPE) is observed when strong electric fields are applied to ZnSe specimens through current carrying electrodes.^[2] As further studies show,^[23,24] this hardening is due to the injection of minority carriers into the specimen through the electrodes. Further, the EPE also arises under monopolar injection of electrons and becomes especially marked under dual injection, when the limitation imposed by the space charge on the concentration of injected electrons is lifted.^[24] In this case virtually all the injected electrons—or all but a very small fraction (10^{-6} – 10^{-4}) of them—turn out to be captured on trapping centers,^[24] and this, in turn, leads to an increase of the dislocation charge^[8] and of the electrostatic Peierls barriers.

The after effect. Ermakov *et al.*^[19] reported that illuminating ZnSe single crystals already containing dislocations results in hardening, which persists for a long time (tens of seconds) after the light is turned off. The activation energy of the memory process was determined as $E \approx 0.2$ eV in the same study. Since the PhPE is due to excitation of the electronic subsystem of the crystal, it remained unclear for a long time how the thermally activated electronic process could have such a long waiting time (10–100 sec at room temperature). For the waiting time we have

$$t \sim \omega^{-1} e^{E/T}. \quad (13)$$

Hence for $t \sim 10^2$ – 10^1 , $E = 0.2$ eV, and $T = 0.0257$ eV we obtain $\omega \approx 10^1$ – 10^2 sec⁻¹ from Eq. (13), whereas the trial frequency ω for electronic transitions usually lies between 10^{10} and 10^{15} sec⁻¹. The energy 0.2 eV agrees well with the activation energy

$$E = aq = 0.19 \text{ eV}.$$

for the motion of dislocations having a typical charge value of $qb/e = 0.6$.

Thus, the memory following illumination takes place as a result of the capture of electrons on a dislocation level of energy $E_d \approx 1.2$ – 1.3 eV^[8] with times for ejection into the conduction band of ~ 10 – 100 sec, while the "de-

cay" after the light is turned off takes place as a result of a process having a lower activation energy: 0.2 eV—the motion of dislocations. As follows from results obtained in Ref. 8, if the dislocation charge is larger than the stationary value the dislocation will rapidly give off the excess electrons to the trapping centers that it encounters.

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Acoustic oscillations of superfluid solutions in narrow channels

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The linearized hydrodynamic equations for He³–He⁴ solutions are considered in the phonon temperature region $T \ll 0.7$ K under conditions of complete stopping of only the phonon part of the normal component. The velocities of two types of sound oscillations are calculated.

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A solution of He³ in superfluid He⁴ is an assembly of elementary excitations—phonons, rotons, and impurity quasiparticles—contained in a superfluid background. At temperatures $T \ll 0.7$ K the contribution of the rotons is exponentially small, and will hereafter be neglected. Under ordinary conditions the phonon and impurity subsystems are strongly coupled with each other, manifest, in particular, by the fact that the macroscopic motion of either the phonon gas or of the impurity-quasiparticle gas is characterized by the same local-equilibrium velocity of the normal motion. Under certain conditions, however, which will be spelled out below, these two systems can be decoupled and the contribution of the phonon part of the normal component can thus be separated.

Interactions of excitations in a solution are completely described by the corresponding scattering cross sections, which were calculated by Landau and Khalatnikov^[1] and by Khalatnikov and Zharkov,^[2] and are of the following order of magnitude:

$$\sigma_{i-i} \approx 2.8 \cdot 10^{-15} \text{ cm}^2, \quad \sigma_{ph-ph} \approx 6 \cdot 10^{-19} (xT)^4, \\ \sigma_{i-ph} \approx 4.7 \cdot 10^{-20} (yT) (xT)^4 \delta',$$

respectively for impurity-impurity scattering, for phonon-phonon interaction, and for scattering of a phonon by an impurity quasiparticle. Here x is the phonon in units of T/s (s is the speed of sound), y is the impurity energy in units of $3T/2$, and δ' is a function of the order of unity.