The annealing of freshly induced defects in ^{4}He crystals

A. A. Levchenko and L. P. Mezhov-Deglin

Institute of Solid-State Physics, Academy of Sciences of the USSR

(Submitted 17 June 1983; resubmitted 6 December 1983)
Zh. Eksp. Teor. Fiz. **86**, 2123–2133 (June 1984)

The dependence of the additional thermal resistivity W_d due to scattering of phonons by dislocations induced by bending, on the annealing time of deformed ^{4}He crystals has been investigated at fixed annealing temperatures $T_{\rm ann} \geqslant 0.45$ K. For weakly deformed crystals the $W_d(t)$ dependence is almost logarithmic, and the corresponding relaxation times τ which determine the rate of the recovery processes decrease exponentially with increasing temperature. The effective activation energies of the recovery process for crystals grown at pressures of 26, 31, 50 and 84 atm, calculated from the slopes of the $\tau(T)$ curves are almost the same as those for diffusion of ^{3}He impurity atoms and of positive charges in ^{4}He specimens of the same molar volume.

1. INTRODUCTION

This work is devoted to a study of the kinetics of recovery processes in hcp ^{4}He crystals deformed by bending. The helium crystals were plastically deformed at temperatures below 0.4 K (bending–unbending to the initial state, degree of deformation $\lesssim 1\%$) and were followed by the reduction in the magnitude of the additional thermal resistivity $W_d = \varkappa_d^{-1} - \varkappa_i^{-1}$ and ($\varkappa_i$ and $\varkappa_d$ are the thermal conductivities in the initial and deformed specimen) as a function of the annealing time at higher temperatures $T_{\rm ann} \geqslant 0.45$ K. The value of W_d is due to the scattering of phonons by the freshly introduced defects. We showed earlier[1] that the main contribution to the thermal resistivity W_d comes from the scattering of phonons by vibrating dislocations (the flutter effect). The contribution to W_d from scattering of phonons by additional vacancies can be neglected: the cross section for scattering by vacancies is much less than for dislocations and falls as T^4 with decreasing temperature, so that their concentration would have to be of the order of ten percent to produce a thermal resistivity comparable with that observed at 0.4 K. The reduction in W_d with time thus reflects structural changes, including the movement, redistribution and annihilation of dislocations. We stress the fact that we are always here just talking about recovery processes and not recrystallization, since judging from the thermal conductivity, the orientation and characteristic dimensions of the blocks of which the specimens consisted remained unchanged after prolonged anneal at temperatures above $0.8T_m$, which leads to practically complete recovery of the thermal conductivity. Out of several tens of specimens studied, recrystallization after annealing was only observed in four. These specimens were remelted and were replaced by new ones.

Much work in the last two decades has been devoted to studying mechanisms for diffusion of different kinds of defects in quantum crystals. The first experimental investigations in this direction were initiated by Shal'nikov,[2] who started a study of the motion of induced charges in solid helium. Present-day ideas about quantum crystals and the influence of quantum effects on the properties of defects in them were formulated by Andreev and Lifshitz.[3,4]

Pokrovskiĭ and Petukhov[5] reanalysed the features of the motion of single dislocations in a Peierls potential in a quantum crystal. However, the results of this work[5] are evidently unsuitable for describing the kinetics of recovery processes, the driving force for which is the reduction in elastic energy of distortions due to the rearrangement of the structure of extra dislocations, interacting among themselves. As was recently shown by Kaganov and Maksimov,[6] taking account of the interaction between defects in a quantum crystal can have an appreciable effect on the diffusion mechanism and in particular can lead to the localization of defects at absolute zero.

The most interesting results of our measurements on specimens grown from pure ^{4}He at pressures $P = 26$, 31, 50 and 84 atm are as follows:

1. At fixed annealing temperature $T_{\rm ann}$, the dependence of W_d on the annealing time t is close to logarithmic for all specimens, which is typical in studies of recovery in metals,[7,8] and also in solidified inert gases at temperatures above some critical value.[9]

2. For a single specimen the $W_d(t)$ dependences at different annealing temperatures and degrees of deformation can be described by a single expression of the form

$$W_d(t)/W_d(0) = 1 - A\ln(1 + t/\tau), \tag{1}$$

where A is coefficient, τ is a parameter which is strongly temperature dependent (the characteristic relaxation time), $W_d(0)$ is the additional resistivity at the initial time.

3. The $\tau(t)$ dependences are close to exponential. The activation energies for the recovery process, calculated from the slope of the $\tau(t)$ curves, are practically the same as the effective activation energies for diffusion of ^{3}He impurity atoms and positive charges in ^{4}He specimens with the same molar volume, calculated from the temperature dependences of the diffusion coefficient. Some results of measurements at 31 atm were published earlier.[10]

We also note two observations concerning the crystallization conditions of helium in a narrow capillary.

a) Superfluid pure ^{4}He liquid (less than $10^{-6}\%$ impurity) compressed to 26 atm can be supercooled by 0.6 K below the solidification temperature ($0.4T_m$) and maintained for tens of minutes until the moment of spontaneous appearance

0038-5646/84/061234-07$04.00

of a crystal. The addition of 0.005% ^{3}He impurity atoms reduced the supercooling.

b) Crystals of hcp ^{4}He are strongly anisotropic, so that the orientation of specimens can be judged from the temperature dependence of the conductivity $\varkappa(T)$ above the position of the maximum $T_{\max}$, where phonon mean free paths are limited by inelastic phonon–phonon scattering. We showed before[1] that in a number of cases the temperature dependence $\varkappa(T)$ above $T_{\max}$ could be explained on the assumption that the orientation of a crystal grown in a capillary bent into a semicircle was unchanged relative to the capillary walls (the direction of heat flow). Some specimens were also obtained in the present work with orientation, judged by the thermal conductivity, constant relative to the capillary walls. These features are evidently related to the strong effect of interaction of the helium with the surface of the container and also of the geometry of the container on the crystallization processes of the compressed liquid. Strong supercooling of compressed helium has been observed recently.[11,12]

2. EXPERIMENTAL METHOD

The helium crystals were grown from purified[13] ^{4}He containing $\leqslant 10^{-6}\%$ of ^{3}He impurity atoms, at constant pressures $P = 26$, 31, 50 and 84 atm. The container was a thin-walled stainless steel capillary of internal diameter $d = 1.8$ mm and length 15 cm. A tapered copper cold finger was soldered to the lower end of the container, connecting it to a ^{3}He liquid bath. The upper end of the tube ended in a conical copper lid to which was soldered a coiled nickel 0.1 mm wall thickness capillary, 0.3 mm in diameter, connecting the working container with the high-pressure line. Wire heaters for temperature control and for measuring thermal conductivity were positioned on the container lid and near the cold finger. Temperatures were measured by two carbon resistance thermometers which were soldered at distances of ~ 2 cm from the ends of the container. The aperture angle from the arc between the thermometers $\Delta\beta = 160°$. Typical values of the temperature difference when measuring conductivity were 2–3 mK, the uncertainty in measuring the temperature difference was not more than 2%.

The upper end of the container could be displaced vertically by a distance of 2 cm with the help of a drawbar. The degree of deformation of the helium crystal, calculated from geometrical considerations, was $\approx 1\%$ for such a displacement. For all the experiments discussed below, the rate of displacing the upper end of the container was 0.1 cm · min^{-1}. The total duration of the deformation was 40 min (20 min bending, 20 min unbending), the deformations were carried out at temperatures < 0.4 K. The specimen heated up to by not more than 0.01 K. By nearly doubling the capillary diameter compared with that used before,[1] the $W_d(t)$ dependence could be studied for smaller degrees of deformation.

The experimental method was as follows. The thermal conductivity of the initial specimen $\varkappa_i$ was first measured over the temperature range 0.37–1.3 K. The orientation and degree of perfection of the specimen could be gauged from this. The specimen was then deformed at a low temperature (bending–unbending at $T \leqslant T_0 = 0.403$ K for specimens grown at pressure $P \geqslant 31$ atm, and $T \leqslant T_0 = 0.384$ K for $P = 26$ atm) and the thermal conductivity of the specimen was measured at the temperature T_0. The temperature of the specimen was then raised at a rate ~ 1 K · min^{-1} to a given temperature T_{ann}, held at this temperature for a certain time t, the temperature was again lowered to T_0 and the measurement of conductivity repeated. The dependence of the additional thermal resistivity W_d on annealing time at a given T_{ann} was determined in this way.

The annealing temperature was changed in 0.05 K steps in the range 0.45–1.10 K. The time for remaining at a given T_{ann} between consecutive measurements varied from 1 to 30 min. In order to achieve the same initial density of freshly introduced defects, the specimens were annealed at $T \gtrsim 0.8\, T_m$ for 10 min and again deformed at $T \leqslant T_0$ after a cycle of measurements at fixed T_{ann}. The values of the additional thermal resistivity $W_d(0)$ at the point T_0 before the start of the measurements of the $W_d(t)$ dependence, for the same deformation of the initial and annealed specimen, agreed within an accuracy up to 30%, although unlike the results of measurements in a narrower capillary,[1] here maintenance for 10 min at temperatures $\approx 0.8\, T_m$ was not always sufficient for complete recovery of the thermal conductivity to the initial value $\varkappa_1$. We represent the results of the measurements in the form of graphs of $W_d(t)/W_d(0)$ in order to remove the effect of variations in the initial values of $W_d(0)$ for different T_{ann}. To a first approximation the ratio $W_d(t)/W_d(0)$ is proportional to $N(t)/N(0)$, where $N(0)$ is the density of defects (dislocations) introduced.

As was said before, specimens were grown both from normal ($P \geqslant 31$ atm, growth rate ~ 1 mm · min^{-1}) and from superfluid liquid ($P = 26$ atm). Strong supercooling of the liquid by ≈ 0.6 K below the melting temperature $T_m = 1.46$ K was, in general, observed on slow cooling of compressed superfluid liquid. The temperature of the liquid increased when crystal appeared, but a large temperature gradient was required for the crystal to grow, so that the temperature difference along the crystal reached 0.5 K at the moment when the solid-liquid helium boundary passed the upper thermometer.

3. RESULTS OF THE MEASUREMENTS AND DISCUSSION

Initial specimens

Judging the values of thermal conductivity below the temperature of the maximum, $T_{\max}$, where the phonon mean free path l_{eff} is limited by scattering by the surfaces of the container or by defects in the bulk of the specimen, the degree of perfection of the specimens could differ strongly from experiment to experiment. In the best of them, grown at low pressures, the effective phonon mean free path at $T \leqslant T_{\max}$ reached 10 mm ($l_{\text{eff}} \approx 5d$) as a result of the onset of Poiseuille flow of the phonon gas. The development of Poiseuille flow in the most perfect of the specimens grown not only at 26, as before,[1] but at 31 atm (Fig. 1, curve 1) could be observed on increasing the capillary diameter from 1 to 1.8 mm. In the least perfect specimens l_{eff} at $T \leqslant T_{\max}$ could correspond to

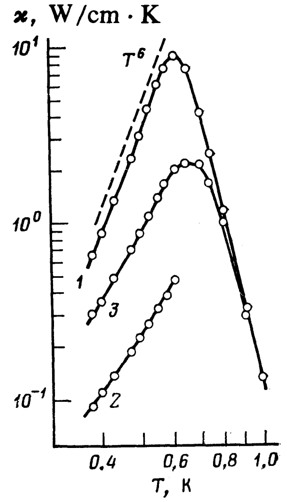

FIG. 1. Thermal conductivity of a specimen grown at 31 atm: curve 1) initial specimen, 2) after deformation and annealing for 1 h at $T = 0.616$ K, 3) after annealing for 10 h at 1.0 K.

$\sim 0.1\, d$. We were thus working with specimens in which the mean block dimensions varied from several centimeters to tenths of a millimeter.

The strong anisotropy of the properties of hcp ^{4}He crystals enables the dominant orientation of a crystal being studied to be judged from the slope of the high-temperature part of the $\varkappa(T)$ curve.[14–16] When the C_3 symmetry axis of the crystal is perpendicular to the heat flow, the thermal resistivity

$$W_\perp \equiv \varkappa_\perp^{-1} \sim \exp(-\Theta/2{,}6T).$$

If C_3 is parallel to the flow, then

$$W_\parallel = \varkappa_\parallel^{-1} \sim \exp(-\Theta/4{,}6T);$$

where Θ is the Debye temperature. For intermediate orientations, the thermal resistivity of the crystal

$$W(\beta) = W_\parallel \cos^2\beta + W_\perp \sin^2\beta = \frac{W_\parallel + W_\perp}{2} + \frac{W_\parallel - W_\perp}{2}\cos 2\beta, \quad (2)$$

where β is the angle between the C_3 axis and the direction of heat flow (in our case the capillary walls). For a polycrystal

$$W = {}^2/_3 W_\perp + {}^1/_3 W_\parallel.$$

It was difficult to predict in advance how the thermal conductivity of crystals grown in a severely bent capillary would behave. It is traditional to consider that the orientation of a crystal growing from a seed remains constant in space. If the C_3 axis is constant is space and lies in the vertical plane, then the angle β on passing from the upper to the lower end of the capillary, bent into a semicircle, changes by $\Delta\beta = 180°$ and according to Eq. (2) the minimum effective value of the thermal conductivity should be greater than $\varkappa_{||}$. However, we were previously (Fig. 6 of Ref. 1) able to observe specimens with slopes of the $\varkappa(T)$ curves at high temperatures close to the $\varkappa_{||}$ (the minimum conductivity, $\beta \cong 0°$) and to $\varkappa_\perp$ (maximum, $\beta \approx 90°$). The minimum conductivity corresponds to the case when the C_3 axis is at all points laong the capillary parallel to the heat flow, i.e., parallel to the tangent to the arc into which the capillary is bent, and lies in the same plane with it (a bent single crystal). We found this very interesting.

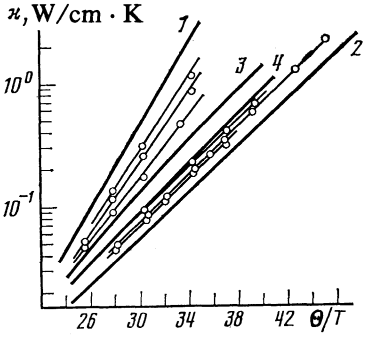

FIG. 2. Temperature dependences of the thermal conductivity of a series of specimens grown at 31 atm, at high temperatures $T \gg T_{\max}$. The points are experimental. Straight lines 1, 2 are for $\varkappa_\perp$ and $\varkappa_{||}$, 3) polycrystalline specimen, 4) calculated minimum value for the thermal conductivity of a single crystal with C_3 axis invariant in space. The aspect angle of the arc between the points where the thermometers are fixed is 160°.

In the present work we specially studied the temperature dependence of the conductivity above and below $T_{\max}$ of 14 initial specimens grown at a pressure of 31 atm. Some of the results of the measurements are shown in Fig. 2. (points along the curves). Curves 1 and 2 correspond to $\varkappa_\perp$ and $\varkappa_{||}$, curve 3 gives the behavior of the conductivity of a polycrystalline specimen, curve 4 is the calculated minimum conductivity of a specimen with C_3 axis invariant in space. In calculating curve 4 we assumed that the C_3 axis lies in the plane of the semicircle and consequently the angle between the heat flow and the C_3 axis as the flow moves through the capillary from the upper thermometer to the lower changes by 160°. It can be seen from Eq. (2) that for $\Delta\beta = 180°$ the thermal resistivity of a specimen should not depend on the orientation of the C_3 axis in the plane and should equal $(W_{||} + W_\perp)/2$. In our geometry the minimum thermal conductivity corresponds to the case when the C_3 axis is oriented vertically. It can be seen that the $\varkappa(T)$ curves of two specimens clearly lie below the minimum curve 4. This is evidence in favor of the assumption that the orientation of a single crystal growing under the action of a temperature gradient can remain invariant relative to the capillary walls. As in the case of supercooling of the liquid, this can be associated with the influence of the walls on the crystallization process for helium in a narrow capillary (crystallization starts from the walls).

Together with the bending, the degree of perfection of specimens can be changed by a sudden change in temperature by several tenths of one degree (thermal shock[17]). Unlike bending, thermal shock leads to a change in the specimen's conductivity over the whole temperature range. The most interesting case is shown in Fig. 3 where the conductivity of a specimen above $T_{\max}$ increased after thermal shock, i.e., not only the degree of perfection but the dominant orientation of the specimen (from a single crystal to polycrystalline) also changed. Prolonged annealing at 0.8 T_m after the thermal shock did not lead to recovery of the thermal conductivity to the initial value over the whole temperature range: the mobility of the defects arising on sharp temperature jumps, as of

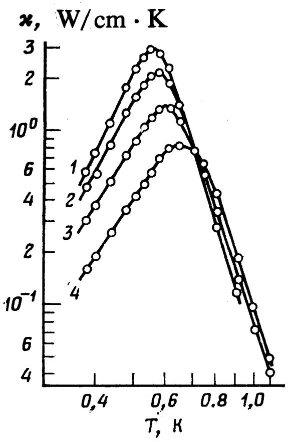

FIG. 3. The effect of thermal shock on the thermal conductivity of a specimen ($P = 31$ atm.). Curves: 1) thermal conductivity of the initial specimen, 2, 3, 4) thermal conductivity of the specimen after a series of consecutive thermal shocks (jumpwise increase in temperature from 0.4 to 1.1 K).

growth defects, is much less than that of freshly introduced dislocations.

The effect of plastic deformation on thermal conductivity

Deformation of specimens at $T < 0.4$ K leads to a reduction in thermal conductivity in the temperature range $T \leqslant T_{\max}$. An example of the change in conductivity of a specimen grown at 31 atm after $\sim 0.5\%$ deformation (capillary unbent at $T \leqslant 0.4$ K) was shown in Fig. 1. Curve 1 is the initial specimen, curve 2 the conductivity of the deformed specimen after annealing for 1 h at $T = 0.616$ K, curve 3 is after annealing at 1.0 K (0.55 T_m) for 10 h. Curves 2 and 3 were measured on lowering the temperature so that annealing of defects during the time of the measurements had little effect on the conductivity here.

The dislocation density introduced by bending can be estimated from the change in the radius of curvature of the capillary on deformation. It is of the order of

$$N \sim 1/bR,$$

where b is the Burger's vector and R the radius of curvature. On the Debye approximation, the magnitude of the additional thermal resistivity produced by scattering of phonons by the introduction of dislocations of density N, can be estimated from the expression

$$W_d = 4\cdot 10^{-6} V_m^{2/3} \Theta^2 T^{-3} N \mathcal{T}, \tag{3}$$

where $\mathcal{T}$ is the radius for scattering of phonons by dislocations, V_m is the molar volume. Substituting into Eq. (3) $T = 0.4$ K, $W_d \approx 10$ cm · K · W^{-1}, $N \lesssim 10^7$ cm^{-2}, for the specimen respresented in Fig. 1, we obtain $\mathcal{T} \gtrsim 5 \times 10^{-6}$ cm. This value is 500 times greater than the theoretical value of the radius for scattering of phonons by the strain fields around single dislocations:[18] $\mathcal{T}_{\rm th} = 0.4 \times 10^{10}\, b^2 T V_m^{-1/3} / \Theta \approx 10^{-8}$ cm for $T = T_0$. Furthermore, the $W_d(T) \sim T^{-n}$ dependence, where $\gtrsim 3$, observed experimentally, differs strongly from quadratic, corresponding to scattering of phonons by static linear defects. Consequently, as indicated earlier,[1] scattering of phonons by freely vibrating dislocations (flutter effect) must play the dominant part.

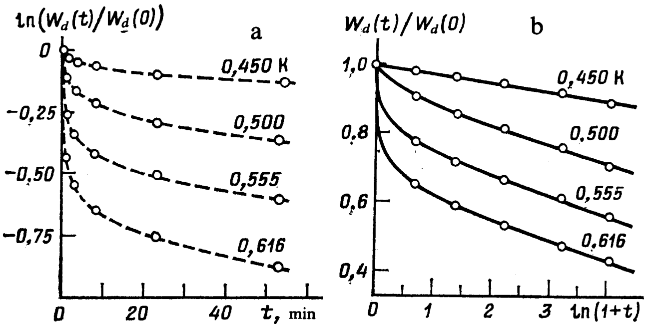

FIG. 4. Time dependence of the relative additional thermal resistivity $W_d(t)/W_d(0)$. $P = 31$ atm, specimen No. 6 in Table I (figures a and b differ in the choice of scales).

Typical dependences of the relative thermal resistivity $W_d(t)/W_d(0)$ on annealing time t at different temperatures $T_{\rm ann}$ for specimens deformed by 1%, are shown in Figs. 4 and 5. Since after measurement at a given $T_{\rm ann}$, the specimens were heated at $T \gtrsim 0.8 T_m$ and again deformed at low temperature, to every $T_{\rm ann}$ in Figs. 4 and 5 correspond their own $W_d(0)$ with values which differ from the arithmetic mean for a given specimen by less than 30%.

Figures 4a and 4b differ in the choice of coordinate. The dependence of ln $[W_d(t)/W_d(0)]$ on t is shown in Fig. 4a (an exponential law would correspond to straight lines), while Fig. 4b shows the ratio $W_d(t)/W_d(0)$ as a function of $\ln(1+t)$ (the time t is measured in minutes). The full curves in Fig. 4b and then in Fig. 5 are calculated from Eq. (1). The points are the results of the measurements. It can be ssen that the ex-

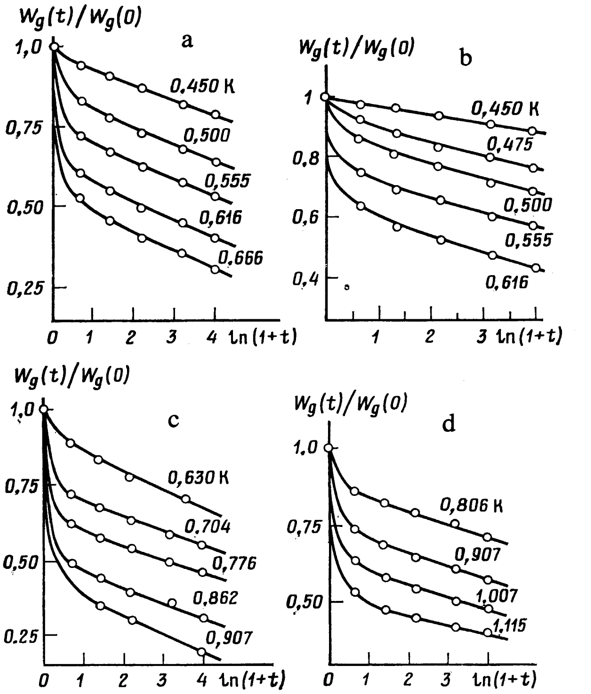

FIG. 5. The $W_d(t)/W_d(0)$ dependences for specimens grown at pressures, 26 (a), 31 (b), 50 (c) and 84 (d) atm (specimens No. 3, 5, 10 and 11 in Table I respectively). The full curves are calculated according to Eq. (1).

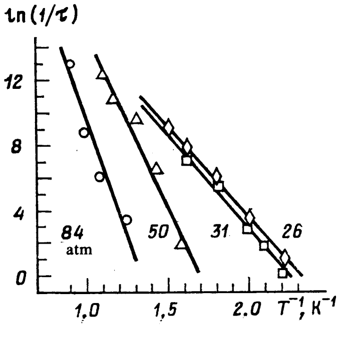

FIG. 6. The dependences of $\ln(1/\tau)$ on inverse temperature for the same specimens as in Fig. 5. Time is measured in minutes.

perimental $W_d(t)/W_d(0)$ dependences are close to logarithmic and not to exponential and can be described by an expression of the form of Eq. (1).

The values of A and τ were first calculated for each of the specimens from the results of measuring $W_d(t)/W_d(0)$ for given T_{ann}. It turned out that A is practically independent of temperature (the scatter in the values is less than 20%), while the value of τ decreases exponentially with increasing T_{ann}. The dependences of $\ln(1/\tau)$ on inverse temperature for the same specimens as in Fig. 5 are shown in Fig. 6. The straight lines in Fig. 6 correspond to the relations

$$\tau = C \exp(\Delta/T). \tag{4}$$

The mean values found for the parameters A, C, and Δ calculated for the specimens studied are shown in Table I. The values of $\varkappa_i$ at the point T_0 and the mean values of $W_d(0)$ are also shown there.

In order to estimate the influence of the degree of deformation, we compared the $W_d(t)/W_d(0)$ dependences at $T = 0.55$ K for a specimen grown at 26 atm for 0.1 and 1% deformations. On increasing the degree of deformation by an order of magnitude, the value of A increases 1.5-fold, the other parameters practically not changing. Control measurements were also carried out on one of the specimens grown at 31 atm from helium containing 0.05% ^{3}He (specimens 9 and 9^a in Table I. The degree of deformation, estimated from the period of the deformation at constant rate, differed by a factor 4 (5 and 20 min), while the initial thermal resistances at a temperature $T_0 = 0.403$ K differed by a factor 10. The ratios $W_d(t)/W_d(0)$ were measured at 6 different T_{ann} from 0.45 to 0.67 K. The results of calculations according to Eq. (1) are shown in Table I and in Fig. 7. It turned out that the values of A, C, and Δ practically did not change with an increase in the degree of deformation.

The dependence of the activation energy for recovery processes in ^{4}He on molar volume of the specimens, V_m, is shown in the composite graph of Fig. 8. For comparison, the activation energies of ^{3}He impurity atoms, negative charges and positive charges are also shown, taken from the composite graph of Efimov and Mezhov-Deglin.[19] It can be seen that the activation energies of recovery processes calculated from the temperature dependences $\tau(T)$ and the values of the diffusion coefficients of point defects, the diffusion of ^{3}He impurity atoms and of positive charges in crystals of the same molar volume are practically the same.

In the process of recovery, the magnitude of W_d can decrease from annihilation of dislocations of different sign in the bulk and on the surface, and as a result of rearrangement of the dislocation structure and pinning of mobile dislocations at defects, at grain boundaries and due to their mutual intersection. We must remember that solid helium only exists at elevated pressures and in our experiments the crystals fit into the capillary surface. The annihilation of dislocations at the surface of the specimen is, unlike in ordinary cyrstals, hindered. Measurements of thermal conductivity showed that the quality could be different form specimen to specimen: from a single crystal with characteristic block dimensions of several centimeters to a polycrystal with grain sizes of tenths of a millimeter. However, the $W_d(t)/W_d(0)$ dependences were the same in all cases. This indicates that volume effects play the main part in the recovery process. Since the scattering cross section for phonons of freely vibrating dislocations is much larger than of the strain fields around immobile dislocations, the pinning of dislocations can have as strong an influence on the magnitude of W_d as their annihilation.

Although the logarithmic dependence of measured characteristics on annealing time (shear stress, electrical re-

TABLE I. Mean values of the parameters A, C and Δ, of the thermal conductivity of the specimens, $\varkappa_i$ and the initial resistivity $W_d(0)$ at the point T_0.

P, atm	N_{cr}	$\varkappa_i$, W·cm^{-1}·K^{-1}	$W_d(0)$, cm·K·W^{-1}	$A\cdot 10^2$	C, min	Δ, K
26	1	0,24	8	5,0	$3{,}8\cdot10^{-12}$	11,6
	2	1,00	10	5,0	$8\cdot10^{-13}$	12,2
	3	1,12	7	4,8	$6{,}3\cdot10^{-12}$	11,2
	4	0,87	0,3	4.2	$3{,}7\cdot10^{-10}$	9.0
31	5	0,61	3	4,4	$4{,}6\cdot10^{-12}$	11.6
	6	0,88	1	5,0	$9\cdot10^{-11}$	10,7
	7	0,20	4	4,6	$7{,}5\cdot10^{-11}$	10,7
	8	0.84	3	3,3	$2.8\cdot10^{-15}$	14,1
	9	0,18	0,7	5,8	$3{,}1\cdot10^{-10}$	9,6
	9^a	0.18	7	5,4	$2{,}8\cdot10^{-10}$	9,6
50	10	0.44	2	4,6	$2{,}6\cdot10^{-16}$	21
84	11	0,12	11	4,0	$6{,}6\cdot10^{-17}$	27.7
	12	0,23	1,5	3,5	$1{,}7\cdot10^{-19}$	32,3

Note: The uncertainty in determining the activation energy Δ is ± 10%. Specimens No. 4 and 9 were deformed by 0.15 and 0.25% respectively, the rest by 1%.

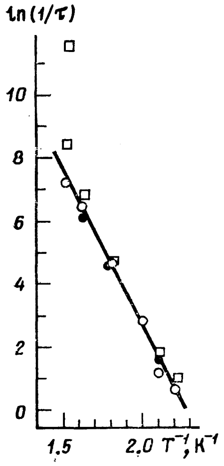

FIG. 7. The dependence of $\ln(1/\tau)$ on inverse temperature for different degrees of deformation of a specimen containing 0.05% ^{3}He, $P = 31$ atm (specimen 9, 9^a in Table I): □) deformation 1% ○), ●) deformation 0.25%, measurements carried out before and after 1% deformation.

sistivity, thermal resistivity in our case) has been observed several times.[7-9] there is no unique analysis of the rate of different mechanisms in the kinetics of recovery processes.

It is assumed in the simplest theoretical model[8] that the activation energy for revovery Δ is a decreasing function of strain. Recovery comes about by the motion of the extra dislocations overcoming barriers as a result of thermal fluctuations. Assuming that W_d is proportional to the density of vibrating dislocations N, while the internal stresses are only due to elastic interactions between them, we can write

$$\frac{1}{W_d(0)}\frac{dW}{dt}=\frac{1}{N(0)}\frac{dN}{dt}=-K\exp\left[-\frac{\Delta_0-BN}{T}\right], \qquad (5)$$

where K, Δ_0 and B are constants for a given T_{ann}. In writing this equation it is assumed that the activation energy Δ is a linear function of the stress acting, i.e., $\Delta = \Delta_0 - BN$, where $BN \ll \Delta_0$. The values of the constants entering Eq. (5) are connected with the quantities A, τ and Δ which are determined from the experiments, by the relations

$$\tau=\frac{T}{BN(0)}\exp\left[\frac{\Delta_0-BN(0)}{T}\right] \quad A=\frac{T}{BN(0)},$$

$$\Delta=\Delta_0-BN(0)=\Delta_0-\frac{T}{A}.$$

Substituting A and D from Table I into these relations, we find that, for example, at 0.5 K and 1% deformation, the value of $BN(0)\approx\Delta_0$, which contradicts the initial assumption. Furthermore, it follows from the control experiments that A and Δ are independent of the degree of deformation. We can thus conclude that the mechanism of a classical thermally activated diffusion of dislocations across the barrier is inapplicable here.$^{1)}$

The closeness of the activation energy of recovery processes in ^{4}He specimens and the activation energy of diffusion of ^{3}He impurity atoms could be evidence that the reduction in W_d comes about as a result of the pileup and capture by vibrating dislocations of ^{3}He impurity atoms migrating in the bulk (pinning). However, it would follow from the work of Iwasa and Suzuki[20] in measuring the attenuation coefficient of untrasonic waves in ^{4}He crystals containing small concentrations of ^{3}He ($\leqslant 3\times10^{-2}\%$), that at these concentrations the pinning of dislocations by impurity atoms is unimportant, since the binding energy of ^{3}He atoms and dislocations is 0.3 K, for $T \geqslant 0.4$ K the concentration of ^{3}He atoms near dislocations is practically equal to that in the bulk. We also came to this conclusion directly. An increase in the ^{3}He concentration compared with pure helium by several orders of magnitude up to $5\times10^{-2}\%$ had practically no effect on the magnitudes of the parameters A, τ and Δ.

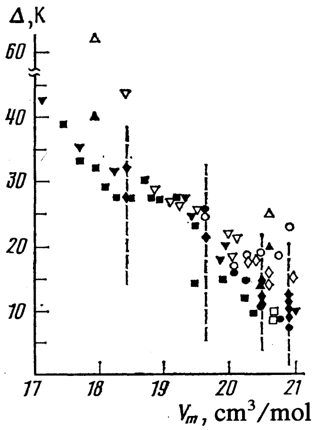

FIG. 8. The molar volume dependence of activation energy Δ for recovery processes (◆, present work), of positive charges (▲, ●, ■, ▼), of negative charges △, ○ and for ^{3}He impurity atoms (◇, □, ▽); the dashed lines are the values of V_m at which measurements were made.

It thus follows from the control experiments and from calculation that the observed recovery cannot be associated with a classical thermally activated diffusion of dislocations across a barrier or with pinning of freshly introduced dislocations as a results of their capture by impurities. We can therefore, assume that the movement of dislocations due to their interactions with point defects plays the main part in the recovery processes. This movement can be achieved by the transport of atoms to the dislocations or from them because of vacancy migration. Beamish and Frank[21] have recently studied recovery processes for the attenuation coefficient of ultrasonic waves in ^{3}He specimens. These authors also connect the agreement between the activation energies for recovery and the activation energies of vacancies with the determining role of the interaction of dislocations with vacancies in recovery processes. Mikheev *et al.*[22] have shown that ^{3}He atoms are delocalized at temperatures below 1.5 K and pressures $P \leqslant 50$ atm. It is natural to expect that at these temperatures vacancies are also delocalized (vacancions). One can therefore say that the interaction of dislocations and vacancions possesses a high mobility. In this case the agreement between the activation energies of recovery processes and a thermally activated motion of defects which we obtained, shows that the concentration of delocalized vacancies in crystals weakly deformed by bending is close to that in thermal equilibrium.

In conclusion, the authors express their thanks to V. N. Khlopinskiĭ for help in preparing and carrying out the experiments, to V. Ya. Kravchenko, V. B. Efimov, A. O. Fedotov, and A. I. Shal'ynin for discussion of the results obtained, and also to A. Ya. Parshin who made a number of critical comments on reading the original form of the article.

---

[1] A. Ya. Parshin drew our attention to the importance of these estimates.

[1] A. A. Levchenko and L. P. Mezhov-Deglin, Zh. Eksp. Teor. Fiz. **82**, 278 (1982) [Sov. Phys. JETP **55**, 166 (1982)].

[2] A. I. Shal'nikov, Zh. Eksp. Teor. Fiz. **47**, 1727 (1964) [Sov. Phys. JETP **20**, 1161 (1965)].

[3] A. F. Andreev and I. M. Lifshitz, Zh. Eksp. Teor. Fiz. **56**, 2057 (1969) [Sov. Phys. JETP **29**, 1107 (1969)].

[4] A. F. Andreev, Usp. Fiz. Nauk **118**, 251 (1976) [Sov. Phys. Usp. **19**, 137 (1976)].

[5] B. V. Petukhov and H. L. Pokrovskiĭ, Zh. Eksp. Teor. Fiz. **63** 634 (1972) [Sov. Phys. JETP **36**, 336 (1973)].

[6] Yu. Kagan and L. A. Maksimov, Zh. Eksp. Teor. Fiz. **84**, 792 (1983) [Sov. Phys. JETP **57**, 459 (1983)].

[7] R. W. K. Honeycombe, The Plastic Deformation of Metals, Arnold, London (1968).

[8] A. H. Cottrell, Dislocations and Plastic Flow in Crystals, Clarendon Press, Oxford, (1953).

[9] D. N. Bol'shutkin, N. F. Kulik, A. V. Leont'eva, A. V. Markov, and Yu. S. Stroilov, Plastic Deformation of Solidified Inert Gases. Physic of Low Temperatures, Kharkov (1972), Vol. 19, p. 9.

[10] A. A. Levchenko and L. P. Mezhov–Deglin, Pis'ma Zh. Eksp. Teor. Fiz. **37**, (1983) [JETP Lett. **37**, 205 (1983)].

[11] J. R. Beamish, A. Hikata, L. Tell, and C. Elbaum, Phys. Rev. Lett. **50**, 425 (1983).

[12] A. A. Golub, N. V. Zuev, and G. A. Mikhaĭlov, Fiz. Nizk. Temp. **9**, 453 (1983) [Sov. J. Low Temp. Phys. **9**, 229 (1983)].

[13] L. P. Mezhov-Deglin, Prib. Tekh. Eksp. No. 3, 217 (1971) [Instrum. Exp. Tech. **14**, 911 (1971)].

[14] L. P. Mezhov-Deglin, Zh. Eksp. Teor. Fiz. **49**, 66 (1965) [Sov. Phys. JETP **22**, 47 (1966)]; Zh. Eksp. Teor. Fiz. **52**, 866 (1967) [Sov. Phys. JETP **25**, 568 (196)].

[15] E. M. Hogan, R. A. Guyer, and H. A. Fairbank, Phys. Rev. **185**, 356 (1969).

[16] R. Berman, C. R. Day, D. P. Goulder, and J. E. Vos, J. Phys. C**6**, 2119 (1973).

[17] A. A. Golub and S. V. Svatko, Fiz. Nizk. Temp. **7**, 413 (1981) [Sov. J. Low Temp. Phys. **7**, 203 (1981)].

[18] J. Z. Ziman, Electrons and Phonons, Clarendon Press, Oxford (1960).

[19] V. B. Efimov and L. P. Mezhov-Deglin, Fiz. Nizk. Temp. **4**, 857 (1978) [Sov. J. Low Temp. Phys. **4**, 406 (1978)].

[20] I. Iwasa and H. Suzuki, J. Phys. Soc. Jpn. **49**, 1722 (1980).

[21] J. R. Beamish and J. P. Frank, Phys. Rev. B **28**, 1419 (1983).

[22] V. A. Mikheev, N. P. Mikhin, and V. A. Maĭdanov, Fiz. Nizk. Temp. **9**, 901 (1983) [Sov. J. Low Temp. Phys. **9**, 465 (1983)].

Translated by R. Berman