

- ¹The structure of the crystal in a strong magnetic field in the envelope of a neutron star has been considered by Kaplan and Glasser^[13] and by Ruderman,^[14] Lozovik and Yudson,^[15] and also Chaplik,^[16] have considered a Wigner crystal in inversion films in a magnetic field.
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Inelastic scattering of electrons by dislocations in copper- and aluminum-base alloys

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The scattering of electrons in metals by quasilocal vibration modes of linear defects excited by increase in temperature results in a characteristic temperature dependence of the dislocation-induced electrical resistivity $\rho_d(T)$. This dependence $\rho_d(T)$ is in the form of a step, whose relative height depends on the density of dislocation-pinning points. If the free segment of a dislocation is limited in length to about 30 atomic distances by adding impurities to the base metal, the rise of $\rho_d(T)$ is suppressed.

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The dislocation-induced electrical resistivity ρ_d of a number of metals (Cu, Ag, Au, Al, Zn, and Mo) rises very rapidly with increase in temperature in a narrow range of the latter.^[1-3] In some cases the cross section for the scattering of electrons by dislocations increases by a factor exceeding 2.^[1,2]

If the dependence $\rho_d(T)$ is represented by the dimensionless curve $r(T) = [\rho_d(T) - \rho_d(0)] / \rho_d(0)$, it is found that pure fcc metals are characterized by $r_{\max} \leq 2$ and the dependences $\rho_d(T)$ have a step whose height is a function of the density of the dislocation-pinning points. Such points can be dislocation line modes or impurity atoms. For example, in the case of copper samples of a given degree of purity and with the same dislocation density the step increases when dislocation lines have a certain preferred orientation.^[2] The addition of dislocation-pinning impurities to dislocations increases very greatly the height of the step.^[3]

The value of r_{\max} can be altered by changing the number of pinning points. Foreign atoms present in a low concentration in a homogeneous alloy form a relatively regular sublattice with which dislocations interact. At a low dislocation density the length of a dislocation segment free of pinning points L_f is governed by the distance between the impurity atoms. However, at higher dislocation densities we can expect more frequent intersections (nodes).

Our aim was to determine the influence of dislocation pinning on the temperature dependence of the electrical resistivity ρ_d . Since it was of interest to find the conditions for complete suppression of the rise of $\rho_d(T)$, similar to that observed by the author earlier,^[3] the impurity concentration in some of the alloys was selected to be $c \sim 0.1$ at.%. An important feature of the investigation was the determination of $\rho_d(T)$ for alloys with different ratios of the masses of the impurity and host (matrix) metal atoms.

EXPERIMENT

Aluminum-base alloy ingots were prepared by melting together Ag and Mg with Al of the A999 grade, containing less than 0.001 wt. % of impurities. The ratio of the room- and helium-temperature resistances of pure aluminum $\gamma = R(293^\circ\text{K})/R(4.2^\circ\text{K})$ was 6000. The charges were placed in an Alundum crucible. They were heated in an argon atmosphere by induction currents. The melt was kept for about an hour in the crucible.

Alloys of copper with $c \approx 0.1$ at. % Pt or Al were prepared from copper of the specially pure grade ($\gamma \approx 200$) containing 0.005 wt. % of impurities. Appropriate amounts of Cu and Pt or Al were melted together in 7×10^{-6} Torr vacuum in a graphite crucible and the melt was mixed for a few hours by a graphite paddle. An alloy of Cu with 0.01 at. % Pt was prepared from high-purity copper ($\gamma \approx 3000$) containing less than 0.001 wt. % of impurities.

The ingots were rolled down to a foil which was 0.5 mm thick. Next, pieces of the Al alloy foils were annealed in 1×10^{-5} Torr vacuum for an hour at 550°C and pieces of the Cu alloy foils were annealed in the same vacuum for an hour at 800°C . Such annealing resulted in complete recrystallization of the foils. The size of the resultant flat grains was usually greater than 1 mm.

Annealed foils were deformed to various degrees and this was done mainly by rolling or—in the case of small strains—by stretching; next, samples used in the measurements were cut from the foil.

The samples were rectangular polycrystalline plates, 90 mm long and of $(3-4) \times (0.1-0.5)$ mm cross section; they were kept at room temperature for 24 hours before the measurements in order to remove the vacancies formed in the course of deformation. After measurements the samples and standards were annealed under the conditions described above for an hour.

The electric resistivity was measured by the potentiometric method. The distance between the potential contacts was 40 mm.

The value of $\rho_d(T)$ was found from

$$\rho_d(T) = \left[\frac{R(T)}{R^*(273^\circ\text{K})} - \frac{R_{st}(T)}{R_{st}(273^\circ\text{K})} \right] \rho(273^\circ\text{K}), \quad (1)$$

where R and R_{st} are the resistances of the sample and standard, respectively; $\rho(273^\circ\text{K})$ is the electrical resistivity of the alloy; $R^*(273^\circ\text{K})$ is the resistance of the sample annealed after measurements in such a way as to complete recrystallization.

TABLE I.

Alloy composition		$\rho, 10^{-6} \Omega \cdot \text{cm}^*$		Alloy composition		$\rho, 10^{-6} \Omega \cdot \text{cm}^*$	
Base metal	Impurity, at. %	273°K	4.2°K	Base metal	Impurity, at. %	273°K	4.2°K
Al	0.048 Ag	2.49	0.055	Cu	0.11 Pt+0.023 Ir	1.97	0.41
»	0.05 Mg	2.45	0.02	»	0.01 Pt	1.582	0.0315
»	0.08 Mg	2.466	0.0325	»	0.17 Al	1.755	0.205
»	0.29 Mg	2.557	0.112	Cu OS4	0.005 wt. %	1.56	0.0078

*The value of ρ of the undeformed alloys is given.

The resistivities of the annealed alloys and their compositions are listed in Table I. The values of $\rho(273^\circ\text{K})$ of the aluminum alloys were taken from the paper by Seth and Woods.^[4] The resistivity $\rho(273^\circ\text{K})$ of the copper alloys was measured with a relative error of less than 2%. The error in $\rho(273^\circ\text{K})$ was eliminated by the use of the dimensionless dependence $r(T)$.

In Eq. (1) the result of the division of $\rho(273^\circ\text{K})$ by the corresponding resistance $R^*(273^\circ\text{K})$ of the sample or $R_{st}(273^\circ\text{K})$ of the standard is the factor which allows for the geometric dimensions. A reduction in the dimensions of a sample as a result of cooling gave rise to an error in the determination of $\rho_d(T)$ below 78°K amounting to less than 0.5% of $\rho_{d \max}$. Therefore, no correction was made to allow for the contraction of the sample at low temperatures.

In each experiment a thermostat contained six samples of which two or three acted as standards. The error in the determination of $\rho_d(T)$ was estimated by comparing the temperature dependences of the resistivity of various standards. Near 273°K this error was less than 10% of $\rho_{d \max}$ for samples with a low dislocation density and less than 5% of $\rho_{d \max}$ for the other samples. Below 78°K the error in $\rho_d(T)$ was less than 2% of $\rho_{d \max}$.

The results of our determinations of $\rho_d(T)$ are presented in Figs. 1-6.

Alloy of copper with 0.17 at. % Al (Fig. 1). The temperature interval of the main change in $\rho_d(T)$ is the same for all the seven investigated samples of this alloy. The height of the step r_{\max} also does not vary greatly. However, an increase in the degree of deformation causes the dislocation density to rise¹⁾ from about 1×10^{10} to $3 \times 10^{11} \text{ cm}^{-2}$ and $\rho_d(4.2^\circ\text{K})$ changes from 0.8 to 30% of $\rho(4.2^\circ\text{K})$ found for the undeformed alloy.

Alloys of copper with Pt (Fig. 2). Even very low con-

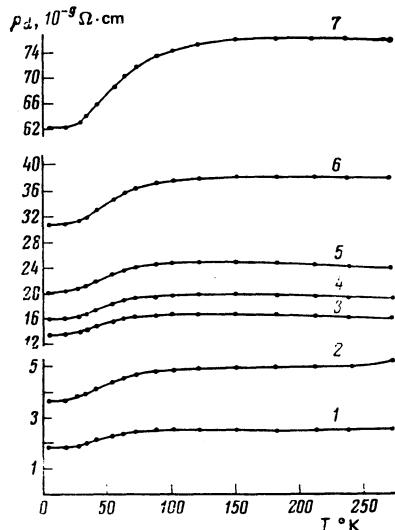


FIG. 1. Dependences $\rho_d(T)$ for the alloy of Cu with 0.17 at. % Al. The degree of deformation is as follows: 1) 8%, 2) 12% by stretching; 3) 18%, 4) 20%, 5) 35%, 6) 65%, 7) 90% by rolling. Value of r_{\max} : 1) 0.35; 2) 0.35; 3) 0.24; 4) 0.25; 5) 0.24; 6) 0.245; 7) 0.23 (the numbers of the curves correspond to the numbers of the samples of a given alloy).

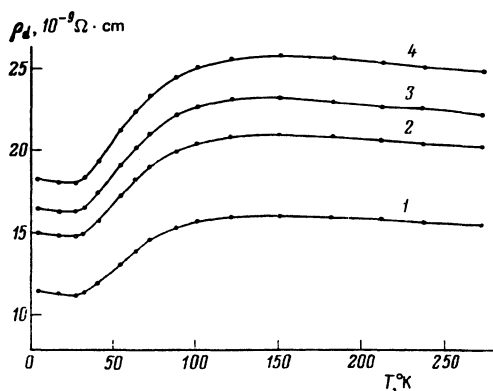


FIG. 2. Dependences $\rho_d(T)$ for the alloy of Cu with 0.01 at. % Pt. Degree of deformation by rolling: 1) 17%; 2) 23%; 3) 35%; 4) 40%. Value of r_{\max} : 1) 0.45; 2) 0.435; 3) 0.433; 4) 0.434.

centrations of Pt are sufficient for a considerable reduction in the height of the step. In the presence of 0.01 at. % Pt the step r_{\max} is half that for copper samples of "specially pure" grade (Fig. 3) in which the impurity content is only slightly less (by a factor of 2 or 3). This strong influence of a small amount of Pt on r_{\max} is evident also if we compare r_{\max} of this alloy with the height of the step observed for stretching-deformed samples of purer ($\gamma = 1900$) polycrystalline copper^[2] for which $r_{\max} \approx 1$.

We also determined $\rho_d(T)$ for alloys of copper with 0.11% Pt and 0.23% Rh; in this case the rise of $\rho_d(T)$ is completely suppressed.

Alloy of aluminum with 0.048 at. % Ag (Fig. 4). In contrast to the copper-base alloys, the temperature interval of the main change in $\rho_d(T)$ of the aluminum alloy increases considerably when the dislocation density is just doubled (see Footnote 1). The height of the step depends on the degree of deformation of a sample. For example, it is twice as high for sample No. 4 than for the weakly deformed sample No. 1 (curves 4 and 1).

A comparison of r_{\max} for pure Al and the alloy of Al with 0.048% Ag, both deformed to approximately the same degree, shows that, in spite of the difference in $\rho_d(4.2 \text{ K})$, an addition of 0.048% Ag to Al reduces strongly (by a factor of 5 or 6) the height of the step

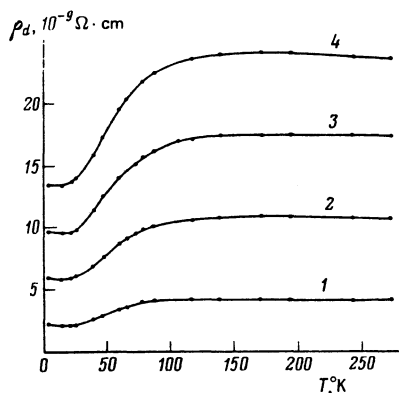


FIG. 3. Dependences $\rho_d(T)$ of "specially pure" copper. Degree of deformation by rolling: 1) 7%; 2) 14%; 3) 17%; 4) 28%. Value of r_{\max} : 1) 0.95; 2) 0.9; 3) 0.82; 4) 0.8.

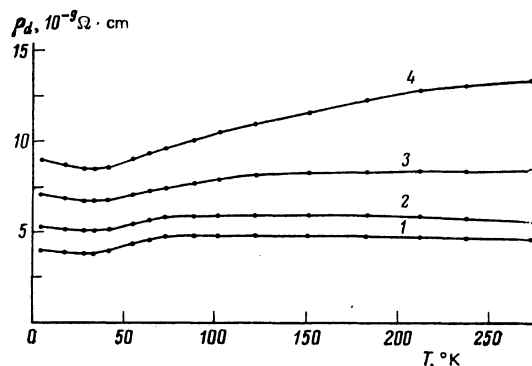


FIG. 4. Dependences $\rho_d(T)$ for the alloy of Al with 0.048 at. % Ag. Degree of deformation by rolling: 1) 35%; 2) 50%; 3) 70%; 4) 60%. Value of r_{\max} : 1) 0.29; 2) 0.2; 3) 0.275; 4) 0.63.

(compare Fig. 1 in the author's earlier paper^[3] and Fig. 4 in the present paper).

An increase in the degree of deformation δ of the samples of the Alloy Al with Ag results, as in the case of pure Al, in an initial rise of the dislocation density but at high values of δ there is a reduction in this density because of extensive regrouping of dislocations (in the course of deformation) in narrow cell walls whose structure resembles that of low-angle boundaries with a moderately well ordered distribution of dislocations.^[8] The interaction of dislocations with one another in such structures results in straightening of dislocation lines and, consequently, is the cause of the reduction in the number of Ag atoms in dislocation cores, especially as the attraction between Ag atoms and dislocations is weak since the atomic dimension of Al and Ag are almost identical.

Alloys of aluminum with Mg (Figs. 5 and 6). All the investigated Al alloys and particularly those with Mg are characterized by some reduction in $\rho_d(T)$ on increase of temperature in the low-temperature range. In contrast to the rise of $\rho_d(T)$ at higher temperatures, this "negative" deviation from the Matthiessen rule is observed in the presence of dislocations only for some of the alloys: for example, it is not exhibited by the alloy of Cu with Al. The investigated alloys of Al with Mg containing 0.05 and 0.08 at. % Mg have similar dependences $\rho_d(T)$ after moderate deformation (curve 1 in Fig. 5). The dislocation-induced rise of the resistivity is comparable in magnitude with the reduction in $\rho_d(T)$ at low temperatures.

The step is clearly visible only in the case of samples

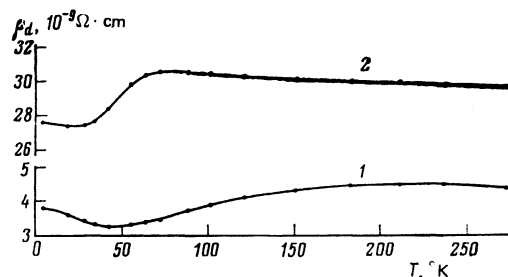


FIG. 5. Dependences $\rho_d(T)$ for the alloy of Al with 0.08 at. % Mg. Degree of deformation by rolling: 1) 35%; 2) 95%. Value of r_{\max} is 0.11 for curve 2.

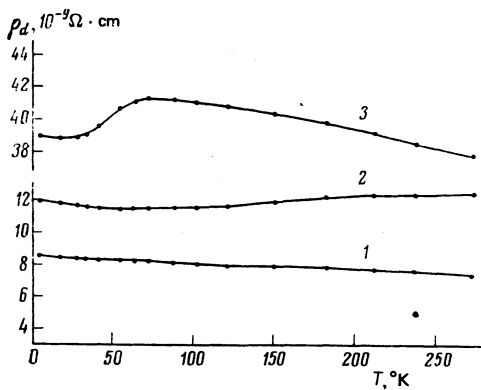


FIG. 6. Dependences $\rho_d(T)$ for the alloy of Al with 0.29 at. % Mg. Degree of deformation by rolling: 1) 35%; 2) 50%; 3) 95%. Value of r_{\max} is 0.06 for curve 3.

with a high dislocation density. The height of the step is always small. The presence of 0.29 at. % Mg is sufficient to suppress the rise of $\rho_d(T)$ —see Fig. 6. Only the strongly deformed sample No. 3 exhibits, above 30 °K, a rising part of the curve characterized by $r_{\max} \approx 0.06$.

DISCUSSION

It follows from our results that there is a considerable change in r_{\max} which is not associated with the usual mechanisms responsible for the deviation from the Matthiessen rule, i.e., it cannot be explained by the influence of nonadditivity of the scattering by centers of different nature, but is due to the action of impurities on the type of scattering caused by dislocations themselves.

The variable parameters in our experiments are the impurity concentration, dislocation density (degree of deformation), and the ratio of the atomic masses of the host metal and impurity. The dimensions of impurity atoms are in all cases, with the exception of the alloy of Al with Ag, considerably greater than the dimensions of the atoms of the host lattice, which ensures that the interaction of impurities with dislocations is strong.

A common feature of all the experimental results is that a reduction in the average length of a dislocation segment free of pinning points reduces r_{\max} .

At $T \approx 300$ °K in Al and particularly in Cu we can ignore the diffusion of atoms. Some changes in the impurity positions in the dislocation cores are possible.^[9] However, we shall estimate L_f on the assumption that there is no significant redistribution of impurities after formation of dislocations.

The determination of the concentration of a homogeneous solid solution shows that the average distance between impurity atoms in the bulk, on a selected plane, or along some line is proportional to $1/c^{1/3}$, $1/c^{1/2}$, and $1/c$, respectively. In plastically deformed fcc metals we can expect dislocations in {111} planes. If their attraction to impurities is strong, we can estimate L_f from $L_f \approx a/c^{1/2}$, where a is the atomic size, or from $L_f \approx a/c$, if the interaction of dislocations with impuri-

ties can be ignored. Moreover, we can expect nodes (dislocation interactions) along dislocation lines and the density of these nodes is proportional to $1/N^{1/2}$, where N is the dislocation density. The influence of the concentration of sites on r_{\max} is strong^[2] but in the case of weakly deformed alloys with $c \approx 0.1$ at. % it is sufficient to allow only for the presence of impurities.

The absence of a step in the dependences $\rho_d(T)$ is observed for the alloys of Cu with 0.11 at. % Pt and 0.023 at. % Rh and also, after relatively weak deformation (curve 2 in Fig. 6) of the alloy of Al containing 0.29 at. % Mg. Hence, the estimate of L_f corresponding to the absence of a step gives $L_f \approx a/c^{1/2} \approx (20-30)a$.

The influence of the mass of the dissolved impurity atoms can be established conveniently by comparing the temperature dependences $\rho_d(T)$ of the alloys of Cu with Al or Pt. The metallic radii of Al and Pt, which are 1.43 and 1.38 Å, respectively, are greater than the Cu radius (1.28 Å), i.e., both impurities attract dislocations strongly. In fact, in the case of the alloy of Cu with Al and also in the case of the alloy of Cu with just 0.01 at. % Pt the value of r_{\max} is practically independent of the degree of deformation. Nevertheless, it is quite clear that a higher concentration of Al than of Pt is necessary to ensure a comparable reduction in r_{\max} .

In the case of the alloy of Al with Ag a strong reduction in the step height compared with pure Al^[3] can again be achieved by adding even a small amount of the impurity, particularly if we bear in mind that the features of the interaction between the Ag atoms and dislocations in Al make the length of a free dislocation segment much greater than in the other investigated alloys so that in this case it is preferable to use the estimate $L_f \sim a/c$.

The results obtained suggest that the investigated effect clearly depends on the difference between the atomic masses of the impurity and the host, but this requires further experimental studies of a large number of alloys with different ratios of the masses of the impurity and host metal atoms. The existence of a relationship between r_{\max} and the mass of the atoms which pin dislocations would be regarded as confirmation that the mechanism responsible for the rise of $\rho_d(T)$ is associated with the excitation of vibrations in dislocation cores.

The possibility of existence of additional branches of the vibration spectrum, localized near linear defects, has been considered theoretically.^[10-12] Heavy impurities in dislocation cores may alter considerably the law of dispersion of local and quasilocal vibrations of a linear defect.^[13]

The low-temperature parts of the dependences $r(T)$ (for the majority of samples this means below 60–70 °K) can be approximated quite satisfactorily by the simple dependence

$$r(T) \propto A e^{-k/AT}, \quad (2)$$

where A is a constant factor and k is the Boltzmann constant. Even in those cases when the curves are

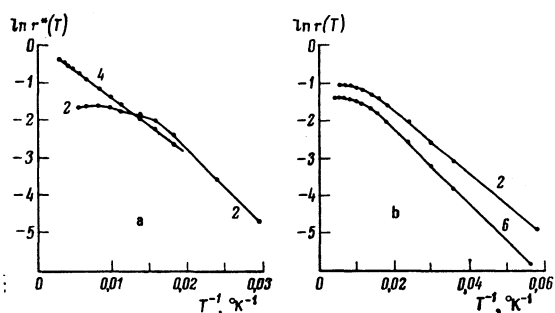


FIG. 7. Dependences of $\ln r^*(T)$ on T^{-1} for: a) Al+0.048 at.% Ag; b) Cu+0.17 at.% Al. In Fig. 7a we have $r^*(T) = [\rho_d(T) - \rho_{d\min}] / \rho_{d\min}$. The numbers alongside the curves give the numbers of the samples of the alloys (compare with Figs. 4 and 1).

slightly distorted and do not have an abrupt step, Eq. (2) describes the rise of $r(T)$ from the minimum to almost the point of transition to a plateau (case a in Fig. 7).

The energy ϵ calculated from Eq. (2) is within the range $(0.7-1.2) \times 10^{-2}$ eV for the various Cu alloys and it amounts to $(1.3-1.7) \times 10^{-2}$ eV for the Al alloys. In the case of samples with different dislocation structures the value of ϵ varies somewhat and clearly the alloys are characterized by higher values of ϵ than the weakly deformed pure metals.^[2-3]

Our experiments give information mainly on the scattering by high-frequency vibrations. Clearly, the experimental results give some particular value of ϵ governed by the position of the maximum of the spectral density of phonons in the short-wavelength region. In this case the value of ϵ is determined by the characteristic frequency ω of the quasilocal vibration modes of dislocations. The temperature position of the step suggests that the characteristic frequency of these vibrations should be between 2 and 3 times less than the

Debye frequency.

The results of our measurements show that if L_f is less than 30 lattice periods, achieved in our experiments by the addition of Mg to aluminum or of Pt and Rh to copper, the characteristic dislocation scattering channel is then suppressed.

¹Estimates are made on the basis of the value of $\rho_d(4.2^\circ\text{K})$. In the case of copper we find that $\rho_d(4.2^\circ\text{K})/N \approx 1.8 \times 10^{-19} \Omega \cdot \text{cm}^3$ (Ref. 5) and in the case of aluminum it is $\rho_d(4.2^\circ\text{K})/N \approx 1.8 \times 10^{-19} \Omega \cdot \text{cm}^3$ (Ref. 6) (N is the dislocation density). The presence of impurities should not alter significantly the value of this ratio.^[7]

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Passage of current through a Josephson barrier of finite dimensions

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The static solutions of the nonlinear differential equation describing the current distribution in a Josephson barrier of finite width are studied. All the possible types of distributions are described, and the stability of the obtained solutions is investigated. It is shown that the solutions corresponding to the presence of excited vortex states in junction with a transport current are unstable and, therefore, cannot be realized under ordinary conditions. Examples are given of self-oscillating solutions, which arise in a junction with a current whose strength exceeds the critical value. The dependence of the period of the self-oscillations on the parameters of the problem has been determined with the aid of a computer.

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1. The problem of the nature of the passage of a current through a Josephson barrier of finite width in the absence of an external (current-unrelated) magnetic field was considered earlier by Owen and Scalapino,^[1]

who, by numerically integrating the corresponding nonlinear differential equation, found, in particular, that in a wide barrier the transport current is largely concentrated near the barrier edges and does not penetrate