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Electric resistivity of aluminum due to the structure state evolution under active loading near the yield limit

V.N.Nikiforenko, M.E.Bosin, Z.A.Krasnikova, S.A.Solodilov

"Cyclone" Measurement Equipment Institute,
126 Plekhanovskaya St., 61001 Kharkiv, Ukraine

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The structure state evolution near the yield limit of polycrystalline aluminum of various purity grades under active loading has been studied by electric resistance method. Characteristic parameters of the structure state have been determined, namely, the effective dipole height and coefficients of dislocation interaction. It has been shown that, no matter what is the material purity grade, the onset of the plastic flow, the deforming stress, and the electric resistance change are defined by the dipole structure formation.

Вблизи предела текучести методом электросопротивления изучена эволюция структурного состояния при активном нагружении поликристаллического алюминия разной степени чистоты. Определены характерные параметры структурного состояния: эффективная высота диполей, коэффициенты взаимодействия дислокаций. Показано, что независимо от степени чистоты деформируемого металла, начало пластического течения, деформирующее напряжение и изменение электросопротивления определяются формированием дипольной структуры.

To solve the problem of structure state control in a material under deforming, its dislocation structure evolution is to be described, namely, the density increase of dislocations and their arrangement changes under increasing external stressing force [1]. The use of resistometry in situ [2] to study the regularities of the deforming stress changes [3–5] makes it possible to determine the parameters of structure state forming during the material plastic flow. It has been shown in [3–9] that the material deformation near the yield limit results mainly from the interaction and accumulation of dislocations in the sample under study. The dislocation interaction is characterized by coefficient α that is the measure of the energy gain at the paired dislocation interaction [10]. The structure imperfection accumulation near the yield limit is associated first with formation of dislocation dipoles [3, 5, 6]. The coefficient of dislocation interaction was determined for various materials [11] under accumulation of dislocations in stationary mode. Determination of α using resistometry under active deforming was used only in [3, 5].

In this connection, it is of interest to determine the dislocation structure parameters (effective dipole height h_d and dislocation interaction coefficient α) under active deforming. The purpose of this work is to establish the dislocation structure parameters and their changes in aluminum under active loading as functions of increasing internal pre-strain fields.

The formation of dislocation accumulations near the yield limit under external stress depends on the initial internal stress fields [6]. In this connection, to discriminate the dislocation barriers in contrast to extrinsic ones, we have formed the internal stress fields in high-purity aluminum polycrystals under pre-straining near the yield limit by alternating bending with restoring sample shape. The dislocation accumulation in the sample due to pre-straining was intended to provide an increased internal stress level and thus influence the effective parameters of dislocation structure during the consequent tensile loading.

Accordingly, polycrystalline aluminum samples of 0.1 m length and 10^{-3} m diameter were subdivided into two groups depending on the purity grade characterized by $\delta = R(300\ \mathrm{K})/R\ (4.2\ \mathrm{K})$ value: 1st group with $\delta = 10^3$, 2nd group, $\delta = 10^4$ and pre-straining by alternating bending. Prior to tests, the samples were annealed at pre-melting temperature for 1 day and have the grain diameter of 350 μm. The samples of both groups were placed in clamps made of dielectric material and strained by tension at 300 K at a speed of $\dot{\varepsilon} = 2.6\cdot10^{-5}\ \mathrm{s}^{-1}$. The electric resistance changes associated with the straining force increase were recorded using a compensation scheme and recorder of 10^{-12} Ω·m sensitivity. The measurements being done in the course of active loading were controlled also using potentiometry in the relaxation regime. The $\Delta\rho$ was found to do not return to its initial state when being recorded in the relaxation regime and do not differ essentially from the values determined using electric resistance recording under active loading. Thus, $\Delta\rho$ is associated mainly with the structure state changes.

Typical curves of the additional electric resistivity $\Delta\rho$ at increasing straining stress τ ($\tau = m\sigma$, $m = 1/3$) for the Groups 1 and 2 of samples are presented in Fig. 1 as curves 1 and 2, respectively. These $\Delta\rho(\tau)$ dependences are seen to be substantially different. Although the 2nd group samples are of one order higher purity grade as compared to the 1st ones, the former show a twice higher yield stress τ_0. In this connection, it is just the structure state formed by pre-straining but not the impurities that must influence mainly the internal stress rise under subsequent tensing loading.

Near the yield limit, the $\Delta\rho$ dependences on the straining stress are approximated by the following relationships:

$$\tau_1 = \tau_{01} + A_{11}\Delta\rho_{11} + A_{12}(\Delta\rho_{12} - \Delta\rho_{11})^{12}, \quad \tau_1 < 3.7 MPa, \tag{1}$$

$$\tau_2 = \tau_{02} + A_{21}\Delta\rho_{21} + A_{22}(\Delta\rho_{22} - \Delta\rho_{21})^{12}, \quad \tau_2 < 8.5 MPa, \tag{2}$$

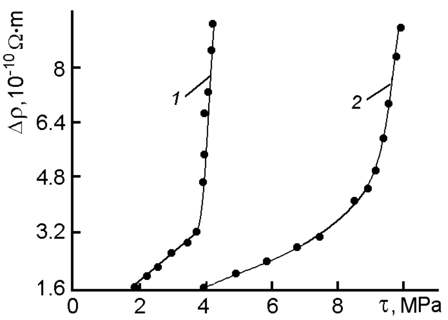

Fig. 1. $\Delta\rho(\tau)$ dependences for aluminum at different purity grade: $\delta = 10^3$ (*1*) and $\delta = 10^4$ (*2*).

where $\tau_{01,2}$ are the straining stresses corresponding to the plastic flow onset for the 1st and 2ng group, respecttively; A_{11}, A_{12}, phenomenological parameters for the 1st group; A_{21}, A_{22}, those for the 2nd one; $\Delta\rho_{11}$, $\Delta\rho_{12}$, resistivity values for the 1st group corresponding to the linear and transition curve sections, respectively, at $\tau_1 < 3.7$ MPa; $\Delta\rho_{21}$, $\Delta\rho_{22}$, the same for the 2nd group at $\tau_2 <$ 8.5 MPa. The values of parameters included in Eqs.(1) and (2) are presented in the Table.

MPa		MPa/(Ω·m)		MPa/(Ω·m)$^{1/2}$	
τ_{01}	τ_{02}	A_{11}	A_{21}	A_{12}	A_{22}
1.9	4	$1.1\cdot10^{10}$	$2.33\cdot10^{10}$	$0.75\cdot10^5$	$1.24\cdot10^5$

The relationships (1) and (2) relating the straining stress and the resistivity increase are necessary to consider the internal stress evolution basing on the resistometry data.

It has been shown according to the empirical Wiedersich model [12] and analytical studies using computer simulation [6, 13] that at the initial deforming stage, $\tau = AN$ where N is the dislocation density; A, the proportionality factor. It is to note that, according to resistometric investigations [3, 5], the linear relationship between τ and N is inherent mainly in dipoles. Therefore, it is the stress field of dipoles accumulated in the sliding lines that must be the main barrier in the motion path of dislocations in the 2nd group of samples. This results in the mentioned considerable difference in the $\Delta\rho(\tau)$ dependences.

Blewitt, Coltman and Redman [14] compared first the straining stress curve with the resistivity change. They supposed that the increase in both characteristics is due to the increasing density of dislocations generated according to Franck-Reed mechanism. The further studies done using sufficiently pure polycrystalline samples of copper and aluminum [7, 8], iron [9], and zirconium [5] revealed a relationship between the straining stress, $\Delta\rho$, and dislocation density. It was noted that contribution of vacancies to $\Delta\rho$ due to straining is negligible and thus, $\Delta\rho$ is defined mainly by the increased number of dislocations.

If the resistivity change is due mainly to dislocations, then [7, 9]

$$\Delta\rho = \rho_d\,(N - N_0), \tag{3}$$

where ρ_d is resistivity of individual dislocation; $N - N_0$, the dislocation density increase resulting from straining. In the case under consideration, taking into account Eq.(1), the Eqs.(1) and (2) can be transformed to obtain

$$\tau_1 = \tau_{01} + A_{11}\rho_d(N_{11} - N_0) + A_{12}\rho_d^{1/2}(N_{12} - N_{11})^{1/2},\ \tau_1 < 3.7 MPa, \tag{4}$$

$$\tau_2 = \tau_{02} + A_{21}\rho_d(N_{21} - N_0) + A_{22}\rho_d^{1/2}(N_{22} - N_{21})^{1/2},\quad \tau_2 < 8.5 MPa \tag{5}$$

At the dislocation density near the yield limit 10^{12} m^{-2}, $\rho_d = 13.6\cdot10^{-23}\Omega\cdot$m^3. It follows from (4) and (5) that near the plastic flow onset (τ_{01}, τ_{02}), the relationship between the straining stress and dislocation density (2nd item) is linear. This linearity allows to suppose that a structure containing dipoles is formed at the initial stage of aluminum straining [15]. It is to note that the dipoles have been found to be an immanent feature of copper and aluminum revealed in studies of strain hardening using transmission electron microscopy [16]. Assuming that changes in the straining stress and resistivity are due to increased dipole density $N_d \sim N$, let the effective dipole height up to yield limit be estimated at $K_1 = \rho_d A_{11}$, $K_2 = \rho_d A_{21}$ [15]:

$$h_{12} \sim K_{1,2}\,\pi(1 - \mu)/2Gb. \tag{6}$$

At $G = 2.65\cdot10^4$ MPa, $b = 4.05\cdot10^{-10}$ m, $\sqrt{\mu} = 0.3$, $K_1 = 15\cdot10^{-13}$ MPa$\cdot$m^2, $K_2 = 31.7\cdot10^{-13}$ MPa$\cdot$m^2, we obtain $h_1 = 0.15$ μm and $h_2 = 0.32$ μm, respectively.

The further increase of straining stress at $\tau_1 > 3$ MPa and $\tau_2 > 7.5$ MPa results in that the linear $\tau(N)$ dependence changes into power one, $\tau(N^{1/2})$. Using the symbol $\tau_{P1,2}$ to denote the yield limit for the corresponding sample group, let the further increase of straining stress at $\tau > \tau_{P1,2}$ be presented as

$$\tau = \tau_{P1,2} + \alpha_{12} GbN_{1,2}^{1/2}. \tag{7}$$

Perhaps the above-mentioned linear-to-power change near the yield limit is due to a change in the character of paired interaction of dislocations. The available experimental data make it possible to evaluate the dislocation interaction coefficient from Eqs.(4), (5) and (7) as

$$\alpha_{12} = A_{1,2}\rho_d^{1/2}/Gb. \tag{8}$$

Substituting successively the A_{12} and A_{22} values into (8), we obtain $\alpha_1 = 0.08$, $\alpha_2 = 0.14$. The small α values correspond to a small energy gain due to the paired interaction of dislocations. The dislocations involved in the interaction form configurations in the metal under straining that favor a considerable compensation of the internal stress fields. Those configurations include no doubt the dipoles. The dislocation interaction coefficient α determined from electric resistance data is small, perhaps due to a considerable compensation of the elastic field of dipoles being the main barriers for moving dislocations.

Thus, the resistometry in situ makes it possible to determine effective parameters of dislocation ensembles formed near the yield limit and to reveal the controlling mechanisms of plastic flow in polycrystalline aluminum samples.

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Електроопір алюмінію, обумовлений еволюцією структурного стану при активному навантажуванні поблизу границі текучості

В.Н.Нікіфоренко, М.Є.Босін, З.А.Краснікова, С.А.Солодилов

Поблизу границі текучості методом електроопіру досліджено еволюцію структурного стану шляхом активного навантажування полікристалічного алюмінію різного ступеня чистоти. Визначено характерні параметри структурного стану: ефективна висота диполів, коефіцієнти взаємодії дислокацій. Показано, що, незалежно від ступеню чистоти металу, що деформується, початок пластичної плинності, деформуюча напруга та зміна електроопіру визначаються формуванням дипольної структури.