Journal of Engineering Physics and Thermophysics, Vol. 65, No. 3, 1993

REGULARITIES OF A VARIATION IN ELECTRICAL RESISTANCE AND DEFORMING STRESS OF ABM-1 ALLOY AND ALUMINUM AT 77 K

V. N. Nikiforenko

UDC 620.18:620.17:537.312

A resistometric investigation of the onset of plastic flow and the regularities of strain hardening of ABM-1 alloy (Al-Be-Mg) and coarse-grained aluminum at liquid nitrogen temperature is performed. It is shown that plastic flow in the considered materials begins long before the appearance of the yield point. The established regularities of a variation in electrical resistance and deforming stress are explained by the evolution of a structural state characterized by a variation in the coefficient of dislocation interaction.

ABM-1 alloy, which, in the solid state, is a mixture of two phases: a β_{Be}-solid solution of up to 32% beryllium and an α_{Al}-solid solution of up to 5.5% magnesium in aluminum, is characterized by valuable physicomechanical properties [1, 2] that enable one to use them over a wide temperature interval in cryogenic engineering. To do this requires information on the variation in the internal stresses due to the evolution of the structural state under active deformation. The absence of such information for ABM-1 alloy limits its use. The currently available indirect methods for estimating the value of the internal stresses are based on measuring the velocity and temperature dependence of the yield point [3], the relaxation curve of deforming stress and electrical resistance [4, 5], and the stepwise change in the rate of deformation [6]. Determination of the onset of plastic flow from resistometry data makes it possible to obtain, owing to the high resolution of the method, pertinent information on weakly deformed specimens [7].

The aim of the present work is to establish the regularities of the variation in electrical resistance and deforming stress of ABM-1 alloy and coarse-grained ($d = 350$ μm) aliminum of purity $\delta = R_{300}/R_{4.2} = 10^3$ at 77 K, which would enable us to obtain data on the variation in the internal stresses and the structural state of these materials.

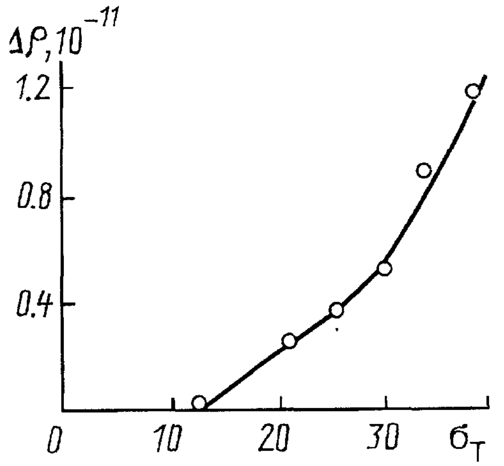
The experiment was performed in the regime of active extension of the alloy specimens, having cross section $4 \cdot 10^{-6}$ m^2, length 0.1 m, and average grain size $d = 22$ μm, with the rate $\varepsilon = 7.8 \cdot 10^{-5}$ sec^{-1} in liquid nitrogen. Deformation of aluminum polycrystals chracterized by length 0.1 m and cross section 10^{-3} m was performed with the rate $\varepsilon = 2.6 \cdot 10^{-5}$ sec^{-1}. Since the variation in the electrical resistance under active deformation could be accompanined by heat release the value of increment in electrical resistance recorded under active deformation was compared with $\Delta\rho$ in the unloaded state when heat release was excluded. The experiments performed with deformation of the specimens to 1% did not reveal a substantial difference in $\Delta\rho$ and, hence, the influence of the increased temperature on the additional electrical resistance under active deformation could be neglected.

Before considering the experimental results obtained on the alloy specimens we will turn to the results on the basic plasticizing component of ABM-1, i.e., aluminum. Figure 1 shows the characteristic curve of the variation in the deforming stress and electrical resistance of aluminum which is approximated by the expression

$$\tau_T = \tau_0 + A_1\Delta\rho + A_2\Delta\rho^{1/2}, \tag{1}$$

where $A_1 = 15.7 \cdot 10^{11}$ MPa $(\Omega \cdot m)^{-1}$, $A_2 = 3.23 \cdot 10^6$ MPa $(\Omega \cdot m)^{-1/2}$. Plastic flow begins at the stress $\tau_0 = 4$ MPa ($\tau_T = 1/3\, \sigma_T$) and persists with active increase in $\Delta\rho$ as the deforming stress increases. Since $\Delta\rho$ in the case

Physics and Technology Institute of Low Temperatures, Academy of Sciences of the Ukraine, Kharkov.
Translated from Inzhenerno-Fizicheskii Zhurnal, Vol. 65, No. 3, pp. 321-324, September, 1993. Original article submitted June 24, 1992.



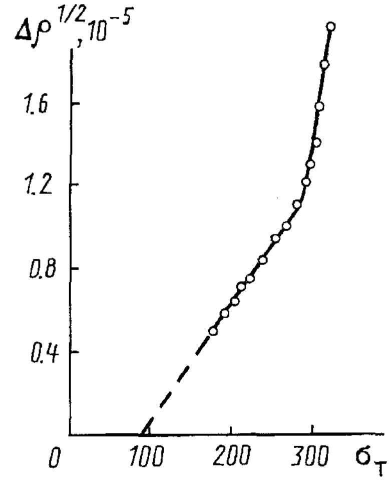


Fig. 1. Increment in the specific electrical resistance of aluminum $\Delta\rho$ ($\Omega\cdot$m) vs. applied stress σ_T (MPa) at 77 K.

Fig. 2. Additional electrical resistance of ABM-1 alloy $\Delta\rho^{1/2}$ $(\Omega\cdot\mathrm{m})^{1/2}$ as a square-root function of deforming stress σ_T (MPa).

considered is primarily due to a variation in the concentration of structural imperfections formed in the process of plastic flow, in the case of predominant contribution of dislocations with density N to the aluminum electrical resistance, relation (1) is representable as

$$\tau_T = \tau_0 + K_1 N + K_2 N^{1/2} \tag{2}$$

($K_1 = 2.2\cdot10^{-12}$ MPa$\cdot$m^2, $K_2 = 3.8\cdot10^{-6}$ MPa$\cdot$m) and is equivalent to the expression [6]:

$$\tau_T = \tau_0 + 2GbhN_d/\pi\,(1-\nu) + \alpha G b N^{1/2}. \tag{3}$$

Estimates show that in this case the effective height of dipoles varies within the interval $h_{1,2}$ = 20.4-58 nm and the coefficient of dislocation interaction α = 0.35 agrees with results on direct determination of this parameter from electron-microscopic data [8]. Since elastic interactions between the dislocations are determined by the coefficient α, a possible difference in the curves of strain hardening and $\Delta\rho$ of ABM-1 alloy compared to aluminum should be sought in the difference in the interactions of dislocations in the aluminum and beryllium components, since α for the interaction of dislocations in the magnesium component is extremely small: 0.05 [9].

Figure 2 gives the additional electrical resistance as a square-root function of deforming stress, obtained on ABM-1 alloy specimens. The plastic flow of the alloy begins at the stress σ_0 = 90 MPa, which is substantially smaller than the yield point σ_p = 275 MPa, at which mass multiplication and motion of dislocations occur. At a stress of 290 MPa on the curve $\Delta\rho^{1/2}$ (σ_T) an inflection characterized by increased intensity of the electrical resistance increment appears. The value of the alloy yield point is likely to be related primarily to the plastic flow in the aluminum component. Indeed when it is considered that σ_{pAl} = 30 MPa and for Be deformed at 77 K this value is equal to 310 MPa [10] it is not unlikely that active motion and multiplication of dislocations in beryllium grains begins at a stress close to the value of the yield point when the level of internal stresses aroun the beryllium grains becomes comparable with σ_p of the beryllium component of the alloy. As was noted in [1], aluminum, not coming into chemical interaction with Be, forms a plastic matrix around its grains. Hard beryllium particles having an armoring action take over most of the load applied to the specimen in the process of deformation. Comparable values of the yield point of Be and the stress at which we observe a sharp increase in $\Delta\rho$ enable us to establish the intervals of deforming stress with the most effective action of the plasticizing matrix and the β_{Be}-solid phase.

Taking in a first approximation the prevailing contribution to the variation in the dislocation electrical resistance and the orientation factor $\sim 1/3$, we represent the results of Fig. 2 as

$$\tau_T = \tau_0 + K_{1c} N^{1/2} + K_{2c} N^{1/2}, \tag{4}$$

where $\tau_0 = 30$ MPa is the value of the onset of the alloy plastic flow; $K_{1c} = 4.8 \cdot 10^{-5}$ MPa·m and $K_{2c} = 10^{-5}$ MPa·m depend on the specific resistance of the dislocations.

In connection with the fact that the elastic interactions between the dislocations are determined by a relation of the type [6]

$$\alpha_{1,2} = K_{1c,2c}/G_{1,2}b, \tag{5}$$

expression (4) is representable by the relation

$$\tau_T = \tau_0 + \alpha_1 G_1 b_1 N^{1/2} + \alpha_2 G_2 b_2 N^{1/2}, \tag{6}$$

where $\alpha_1 = 0.22$ and $\alpha_2 = 0.064$.

The decrease in the coefficient of dislocation interaction for ABM-1 alloy deformed near the yield point by a factor of 1.6 compared to aluminum is likely to be associated with unequal penetrability of obstacles by dislocations in the aluminum and the alloy. According to model representations [11], $\alpha \geq 0.32$ is observed in the case of penetration of dislocations through timber or other obstacles and in recombination of dislocations. The coefficient $\alpha = 0.22$, characterizing the energy gain in pair dislocation interaction at the yield point of the alloy, satisfies the model of movement of steps on the dislocations [11], for which $\alpha = 0.2$. Such a value of the coefficient is likely to be related to the intense field of internal stresses due to the high elastic constants of beryllium grains. The strong irregularity of the internal stress field due to the β_{Be}-solid phase apparently leads to the creation of vast regions in the specimens where the sliding dislocation is unable to penetrate even at a considerable level of external stress. The reason for that may be formation of a large number of dislocation loops in the aluminum matrix, which substantially hinder the motion of the subsequent train of sliding dislocations. Shielding the sliding dislocations on the side of the formed structural states and beryllium grains can evidently decrease the total penetrability of the dislocation ensemble at a stress of 290 MPa, which could be inferred by the value of the dislocation interaction coefficient. Starting with 290 MPa, the elastic interaction of dislocations decreases sharply, which is likely to be related both to shielding of the internal stresses accumulated near Be grains and to a quite probable plastic flow in the grains themselves. Considering beryllium grains as segregations with higher elastic constants than those of the matrix, it is pertinent to note that if stress concentrations occur near the segregations the shear in the matrix is maximal at the surface of the contact between the segregations and the plasticizer. Dislocation rings concentrated near a segregation create reverse stress, inhibiting the source action and hardening the material. The stress concentration due to accumulation of loops around inclusions can result in formation of fractures in the grains or near them. The aluminum matrix, possessing several times greater plasticity than the beryllium one removes the stress concentration in the β_{Be}-phase particles and hinders the formation of fractures. Fracture formation, the tendency for which shows up in comparing data on the ultimate strength of the alloy $\sigma_B = 350$ MPa and beryllium $\sigma_B = 340$ MPa, occurs basically in beryllium grains.

The investigation performed shows that plastic flow of ABM-1 alloy is controlled by the interaction of dislocations of the matrix with the β_{Be}-solid solution and is accompanied by the formation of stress concentrators near the boundaries of segregations, activating plastic flow in Be grains.

NOTATION

d, grain size; ε, rate of deformation; $\Delta\rho$, variation in specific electrical resistance; σ_T, deforming stress; τ_T, shear stress of flow; τ_0, stress of the onset of plastic flow; N, dislocation density; h, effective dipole height; G, shear modulus; b, Bürgers vector; ν, Poisson coefficient; α, coefficient of dislocation interaction; σ_p, yield point; N_d, dipole density.

REFERENCES

1. I. N. Fridlyander, N. M. Yatsenko, G. A. Nekrasova, et al., Metalloved. Term. Obrab. Met., No. 7, 50-55 (1970).
2. V. I. Grushko, F. F. Lavrentiev, V. N. Nikiforenko, et al., Strength of Materials and Structures at Low Temperatures [in Russian], Kiev (1984), pp. 74-78.

3. V. V. Pustovalov, Physics of Strain Hardening of Monocrystals [in Russian], Kiev (1972), pp. 128-171.
4. V. N. Nikiforenko, Strength of Materials and Structures, Abstracts of Papers at the 3rd All-Union Conference [in Russian], Vinnitsa (1991), pp. 53-57.
5. V. N. Nikiforenko, Inter. Cryog. Mat. Conf. ICEC-ICMC-14, MCGP26, Kiev (1992), pp. 125-128.
6. F. F. Lavrentiev and V. N. Nikiforenko, Dokl. Akad. Nauk SSSR, **315**, No. 5, 1127-1129 (1990).
7. V. N. Nikiforenko and F. F. Lavrentiev, Positive decision of VNIIGPE on inventor's application No. 4873373/28 (1991).
8. J.G. Rider and C. T. B. Foxon, Phil. Mag., **13**, No. 122, 289-303 (1966).
9. P. B. Hirsch and J. S. Lally, Phil. Mag., **12**, No. 117, 595-648 (1965).
10. F. F. Lavrentiev, O. P. Salita, V. V. Palei, et al., Fiz. Met. Metalloved., **64**, No. 6, 1221-1224 (1987).
11. F. R. N. Nabarro, Z. S. Bazinskii, and D. B. Holt, Plasticity of Pure Monocrystals [Russian translation], Moscow (1967).