Creep Behavior of Ni₃Ge Single Crystals of the Deformation Axis Orientation [001]: Experiment and Modeling

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Abstract—The results from an experimental study of the creep of Ni₃Ge single crystals of orientation [001] are presented. Calculated creep curves of alloys with an $L1_2$ superstructure and high energies of antiphase boundaries (APBs) are also presented. The calculations are performed using a model of creep for single crystals of $L1_2$ alloys that is based on the concept of strengthening and recovery.

INTRODUCTION

The creep of alloys and intermetallic compounds with an $L1_2$ superstructure displays certain features that distinguish it from the creep of pure metals with an fcc structure. The main such features are low rates of creep at moderate and high temperatures, anomalous dependences of creep in some temperature intervals, and the occurrence of creep at a constant rate in the high-temperature range [1−4]. This work presents experimental results from studying the creep of single Ni₃Ge crystals with orientation [001] and calculated curves of creep for alloys with an $L1_2$ superstructure. Our calculations were performed using the model of creep for single crystals of $L1_2$ alloys [5]; this model is based on the concept of strengthening and recovery that was described in detail in [6].

EXPERIMENTAL

Alloy of composition 75 at % Ni + 25 at % Ge was prepared by melting in a resistance furnace under a vacuum of $1-2 \times 10^{-6}$ mm Hg using nickel of N-0 grade and high-purity Ge (99.999%). The single crystals were grown according to the Czochralski method. To study creep, we used samples in the shape of parallelepipeds with dimensions of $3.0 \times 3.0 \times 6.0$ mm, oriented such that compression would occur along the [001] axis. The accuracy of determining the orientation was ±2°. After electric-spark cutting, the surface of the samples was purified by chemical etching; the samples were then homogenized at $T = 1226$ K for 48 h and cooled in the furnace to room temperature. Creep was studied under a constant applied load.

RESULTS AND DISCUSSION

The creep curves of our Ni₃Ge single crystals of orientation [001] were recorded upon compression at different external loads (Fig. 1). The creep curves contained (Figs. 1a, 1b) two stages: that of the primary creep with a continuously slowing creep rate, and a secondary stage with a steady-state or stationary creep. We observed no stage where the creep accelerated due to fracturing.

Such results are associated with certain features of the loading scheme (since the creep occurred under compression), and possibly with an insufficient time of observation. One exception is the creep curves obtained at high temperatures ($T > 0.5T_m$) and high stresses ($\tau > 0.9\tau_0$). The multistage nature of these curves differs from the dependences obtained under other conditions. The length of the stage of primary creep grew shorter as the testing temperature rose. At high values of applied stress and high temperatures ($\sigma = 1040$ MPa, $T = 923, 973$ K), the stage of primary creep ended. An anomalous temperature dependence of the creep rate at the stage of primary creep was observed for the single crystals of Ni₃Ge alloy. The creep curves intersect in Fig. 1b, so the creep rate at the temperature of 923 K in the first stage was consid-
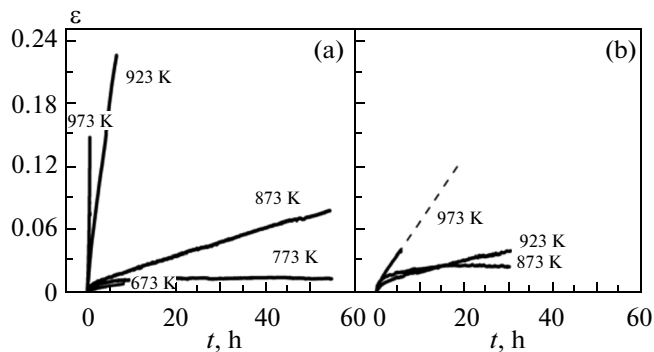
Fig. 1. Creep curves of single Ni₃Ge crystals at different temperatures of testing. Orientation, [001]; load values (a) $\sigma_1 = 1040$ MPa and $\sigma_2 = 370$ MPa.

Rates of steady-state creep, s⁻¹

Stress σ, MPa	Orientation of deformation axis [001]			
	temperature, K			
	773	873	923	973
1040	2×10^{-9} (8.3)	3.8×10^{-7}	9.5×10^{-6}	3×10^{-5} changes 7.7×10^{-5}
370		1.4×10^{-8}	2.7×10^{-7}	1.7×10^{-6}

erably less than at 873 K. Such an anomaly was observed for the first time. However, it should be noted that a similar effect was observed at relatively low applied stress that was equal to one third of the yield stress [2]. The second stage of stationary creep also displayed anomalous behavior in comparison to pure metals and disordered alloys. The rates of steady-state creep are given in the table. At a high value of shear stress (0.9 of the value of the yield stress) and a temperature of 773 K, we observed an anomalously low $(2 \times 10^{-9} \text{ s}^{-1})$ creep rate (Fig. 1a). Deformation in this case ceased at stage II; it grew primarily at the stage of primary creep. The catastrophic increase in the rate of stationary creep as the temperature rose from 773 K by 100, 150, and 200 K also proved to be anomalous (Fig. 1a). At a temperature of 923 K, the creep rate was comparable to the rates of active deformation (10^{-3} to 10^{-4} %/s).

Along with the usual creep curves characteristic of the different experimental regimes, at $T = 973$ K and $\sigma_1 = 1040$ MPa we observed a sharp increase in the creep rate (i.e., a so-called inverse creep). This type of creep, observed at temperatures that exceed the temperature of the peak of an anomaly) had not been sen earlier. In contrast to the one mentioned in the literature, such a creep is an *inverse high-temperature creep* [3, 4].

Results from a Mathematical Simulation of the Creep of Single Crystals of Alloys with an L1₂ Superstructure

The modern concept of strengthening and recovery lying at the heart of our model of creep is that the intensity of the multiplication of any product of deformation is a consequence of the superposition of the processes behind the generation (nucleation, production, multiplication) of the products of deformation and their relaxation (destruction, annihilation, transformation). In the matrix form, this can be written as [6]

$$\begin{cases} \bar{X} = \bar{G}(x_i \ y_i \ t) + \bar{R}(x_i \ y_i \ t) \\ \bar{Y}(x_i \ y_i \ t) = 0. \end{cases} \quad (1)$$

Here, $\bar{X}$ is the matrix that describes the accumulation of deformation products over time t; $\bar{G}$ and $\bar{R}$ are the

matrices of the generation and relaxation of deformation products, respectively; and $\bar{Y}$ is the matrix that describes the conditions under which plastic deformation occurs (in most simple cases, it describes the experimental conditions imposed by the experimenter).

In the concept of strengthening and recovery, the mathematical model of the creep of alloys and intermetallic compounds with an $L1_2$ superstructure [5] is the set of equations of the balance of deformation-induced defects (dislocations, interstitial atoms, and vacancies) plus an equation that determines the creep rate. To construction our model of creep, we used the idea of a superposition of anomalous (associated with the specific structure of dislocations in $L1_2$ alloys) and normal (characteristic of pure fcc metals) mechanisms of deformation. In this model, it is assumed that plastic deformation occurs via the emergence and expansion of superlattice dislocation loops. The equation of the balance of dislocations considers the formation and accumulation of peripheral (shear-forming) and intrazone dislocations (i.e., dislocation barriers—Kear–Wilsdorf locks and diffusional locks), along with the annihilation of shear-forming dislocations as a result of the climb of their edge components. The equations of the balance of deformation-induced point defects consider their generation as a result of the nonconservative dragging of jogs by dislocations of screw or close-to-screw orientations, their annihilation as a result of precipitation at edge dislocations, and their mutual annihilation. The creep rate is due to the thermoactivated motion of superdislocations and is determined by the density of mobile dislocations, which diminishes upon an increase in the deformation temperature. This is because in alloys with an $L1_2$ superstructure, some dislocations prove to be immobile since they blocked by dislocation barriers (Kear–Wilsdorf locks and diffusion locks), the intensity of formation and the accumulation of which increases as the temperature rises. The density of mobile dislocation thus diminishes with an increase in temperature. In addition, the equation for the creep rate considers that the presence of the Kear–Wilsdorf locks alters the
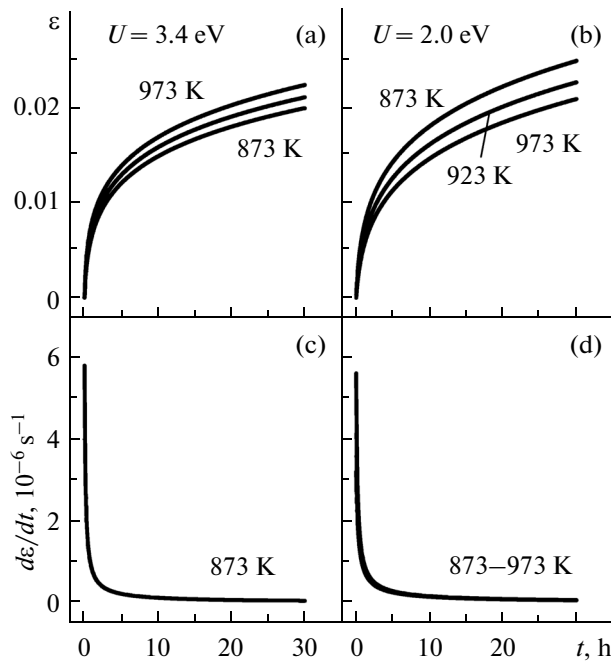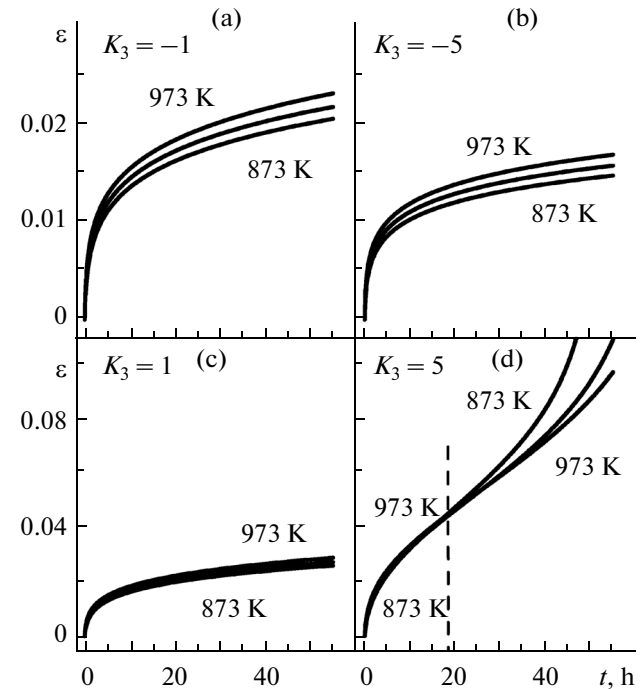
Fig. 2. (a, b) Creep curves and (c, d) time dependences of the creep rate for alloys with high energies of APBs under constant external stress. Load $\sigma = 370$ MPa.

Fig. 3. Creep curves for alloys with high energies of APBs, allowing for the localization of deformation: (a, b) compression and (c, d) tension. Activation energy for the thermoactivated motion of superdislocations $U = 3.4$ eV; stress $\sigma = 1000$ MPa.

length of the dislocation segments undergoing elementary acts of thermoactivated motion. One condition for deformation in the case of creep is assumed to be the constancy of stresses or the constancy of the load.

Figure 2 shows the calculated curves of creep and the time dependence of the creep rate for our $L1_2$ alloys with high energies of APBs under a constant external stress. Calculations are presented for two values of energy of activation U of the thermoactivated motion of superdislocations in the slip plane (3.4 and 2 eV) at different temperatures of deformation (from 873 to 973 K). The applied stress was assumed to be $\tau_0 = 166.5$ MPa ($\sigma = 370$ MPa).

The $\varepsilon(t)$ and $\dot{\varepsilon}(t)$ curves (Fig. 2) reveal two stages of creep: the first is that of primary creep with a continuously decreasing rate; the second is that of steady-state creep, where the rate remains virtually constant. Where the activation energy of creep is $U = 2.0$ eV, we observe an anomalous dependence of the creep rate on the temperature: the rise in temperature from 823 to 973 K led to a reduction in the creep rate. In practice, this situation is observed at the initial stage of creep for single crystals of the Ni_3Ge alloy under the same initial stress (Fig. 1b).

Allowing for the localization of deformation during creep influences the creep curves appreciably (in Fig. 3, $|K_3|$ is the fraction of the sample volume in which the creep is localized). With compression (Figs. 3a, 3b), the changes in the creep curves are only of a quantitative nature; with tension, qualitative changes are also observed (Figs. 3c, 3d): in place of the second stage of steady-state creep, a stage of inverse creep with a creep rate that grows over time emerges under certain conditions (Fig. 3d). This type of inverse creep was observed in Ni_3Al during tests for creep with tension under a constant load [1].

CONCLUSIONS

Our calculations showed that the creep characteristics of alloys with $L1_2$ superstructure that are observed experimentally [1–4] are satisfactorily described by a model based on the concept of a superposition of the mechanisms of deformation characteristic of pure fcc metals, and of the mechanisms of deformation characteristic only of ordered alloys with an $L1_2$ superstructure with a specific structure of their superdislocations.

ACKNOWLEDGMENTS

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