

NONEQUILIBRIUM PHASE TRANSFORMATIONS INDUCED BY DISLOCATION GLIDE IN PLASTIC DEFORMATION OF ALLOYS

I. K. Razumov

UDC 669.112

It has been shown that in intense plastic deformation of alloys, dislocation glide gives rise to nonequilibrium phases or disperse structures. The phenomenon is due to the joint action of two factors: acceleration of diffusion and change in the chemical potential of the alloy components in the region of the dislocation core.

Keywords: *intense plastic deformation, spinodal decomposition, Cottrell atmospheres, diffusion cutting, dissipative structures.*

Introduction. Phase transformations in alloys subjected to intense plastic deformation have long attracted a great deal of researchers' attention [1, 2]. Experiments point to the development of nonequilibrium phase transformations during the action, such as stratification in miscible systems (Ni–Pd [3]), mechanical alloying in immiscible ones (Cu–Co and Fe–Cu [4, 5], and cyclic reactions (Fe–B [6]). Consequently, there are mechanisms converting mechanical energy to the alloy's internal energy and whose nature remains debatable.

A theory of stratification of an alloy on exposure to nonequilibrium-vacancy fluxes generated at the boundaries and junctions of grains in intense plastic deformation has been proposed in [7, 8]. It has been shown that regions adjacent to the vacancy source and sink are enriched with atoms with different diffusion mobilities, and the reached state is frozen after the exposure because of the sink of excess vacancies onto the grain boundaries. Since the effect is reversible, once the vacancy source and sink have exchanged places, its realization requires that the rotational mode of plasticity be restricted.

It has been hypothesized in [9, 10] that the nonequilibrium transformations observed in intense plastic deformation are induced by the local change in the alloy thermodynamics in the region of nonequilibrium grain boundaries [11] whose characteristic width is ~ 1 nm. However, the structure of nonequilibrium grain boundaries remains debatable; in particular, the density of edge dislocations in the region of nonequilibrium grain boundaries does not exceed 10^8 – 10^9 m^{-1} [11]. More powerful defects are declinations on triple grain joints, which lead to decomposition in nanograin alloys [12].

In the present work, we consider an additional mechanism of phase transformations in intense plastic deformation that requires no assumptions on the grain substructure. It is well known that plastic deformation is effected by dislocation glide, although the rotational mode additionally appears on attainment of a small grain size [13]. No dislocations are found in the volume after a long-duration action, but they are accumulated near the grain boundaries [11, 14]. The latter does not mean that the dislocation glide ceases; it rather demonstrates that the nonequilibrium grain boundaries act as barriers in gliding. In the dislocation tube whose radius is equal to several lattice parameters, the crystal lattice is distorted; therefore, the thermodynamic properties of the alloy locally change, which is confirmed by the formation of Cottrell impurity atmospheres [15, 16] and by the effect of diffusion cutting of phase boundaries [17, 18]. Furthermore, the tube diffusion is accelerated by two to three orders of magnitude compared to the volume one [19], which is due to both the reduction in the activation energy of the jump and the preferableness of cyclic atomic exchange [20, 21]. The role of these effects is increased by the random walk of the dislocation line under the action of thermal fluctuations. Concentration heterogeneities fix the dislocation, but it is capable of overcoming the obstacle under the applied stress. The nonequilibrium trail left behind the dislocation dissipates over periods determined by the coefficient of volume diffusion D_0 . The last effect has been found experimentally in [22], where long impurity trails leading to the anisotropy of

Institute of Metal Physics, Ural Branch of the Russian Academy of Sciences, 18 S. Kovalevskaya Str., Ekaterinburg, 620041, Russia; email: iraz@k66.ru. Translated from *Inzhenerno-Fizicheski Zhurnal*, Vol. 83, No. 3, pp. 439–446, May–June, 2010. Original article submitted April 22, 2009; revision submitted October 21, 2009.

the electrical properties of the material were observed behind the dislocations after the plastic deformation of silicon single crystals.

With allowance for what has been said above, we formulate the concept of nonequilibrium phase transformations induced by dislocation glide. Diffusion is accelerated in a dislocation tube where the equilibrium phase diagram is inapplicable; therefore, the nonequilibrium transformation is realized locally. The dislocation moves under stresses, leaving a nonequilibrium trail which is "frozen" because of the difference of the diffusion rates in the dislocation core and in the volume. The glide of a new dislocation through this region causes the nonequilibrium transformation to further develop. As a result the mechanical energy is permanently converted to the internal energy of the alloy, and the degree of decomposition (when the action is long) can be orders of magnitude higher than the segregation capacitance of immobile dislocations.

An analysis with a generalized Cahn–Hilliard spinodal-decomposition model [23] involving phenomenological description of the dislocation core enables us to classify nonequilibrium transformations and to formulate conditions for their realization. New effects such as the formation of disperse structures by the action of a dislocation flux are found. The results are compared to the predictions of other models and to experimental facts.

1. Formulation of the Model. We use the model of decomposition of an alloy with lattice deformations, which has been proposed earlier [9, 24]. In this model, the density of free mixing energy in the mean-field approximation is determined by the formula [25]

$$f(C_A) = \sum_{\sigma=A,B} (\Phi_{\sigma\sigma} - \Phi_{\sigma\sigma}^0) C_\sigma + (\Phi_{AB} - \Phi_{\sigma\sigma}) C_A C_B + kT (C_\sigma \ln C_\sigma), \quad (1)$$

where $C_A + C_B = 1$. In the case of heterogeneous deformations the energies $\Phi_{\sigma\sigma}$ are dependent on the coordinates, $\Phi_{\sigma\sigma} = \Phi_{\sigma\sigma}(\mathbf{r})$. The decomposition kinetics is determined by Fick's laws [25]:

$$\frac{\partial C_A}{\partial t} = -\nabla \mathbf{J}_A, \quad \mathbf{J}_A = -M \nabla \mu. \quad (2)$$

To refine mobility we use the Einstein relation: $M = DC_A C_B / (kT)$. The chemical potential μ is related to the density of free mixing energy by the equation [23]

$$\mu = \frac{\partial f}{\partial C_A} - \zeta R^2 \Delta C_A. \quad (3)$$

Substituting (1) and (3) into (2), we arrive at the expression for the atomic flux

$$\mathbf{J}_A = -D \left\{ (1 - 2\Psi C_A C_B) \nabla C_A + C_A C_B [\nabla \varphi + (1 - 2C_A) \nabla \Psi] - \frac{\zeta R^2}{kT} C_A C_B \nabla \Delta C_A \right\}, \quad (4)$$

where $\Psi = (2\Phi_{AB} - \Phi_{AA} - \Phi_{BB}) / (kT)$ and $\varphi = (\Phi_{AA} - \Phi_{BB}) / (kT)$. The first term is responsible for the ascending diffusion, the second term is due to lattice deformations and is nonzero in the dislocation core, and the last term is analogous to the contribution of the boundary to the Cahn equation for spinodal decomposition [23]. We restrict our consideration to homogeneous plastic deformation effected by the glide of edge dislocations [13]. We assume that the parameters of the alloy in the dislocation core are coordinate-dependent phenomenologically:

$$D = D_0 (1 + \delta_D \xi), \quad \varphi = \varphi_0 (1 + \delta_\varphi \xi), \quad \Psi = \Psi_0 (1 + \delta_\Psi \xi), \quad \xi = \left[1 + \left(\frac{\mathbf{r} - \mathbf{r}_0(t)}{L \delta_L} \right)^4 \right]^{-1}, \quad (5)$$

where δ_D , δ_φ , and δ_Ψ characterize the amplitude and $\xi(\mathbf{r})$ characterizes the form of disturbance of the alloy's parameters in the dislocation core, whereas the position of the dislocation is determined by the radius vector $\mathbf{r}_0(t)$. Thus, we disregard the difference of the values of the parameters ψ and Ψ in the regions of stretching and compression of the dislocation core, which is not universally true but enables us to investigate the basic qualitative features of transformations.

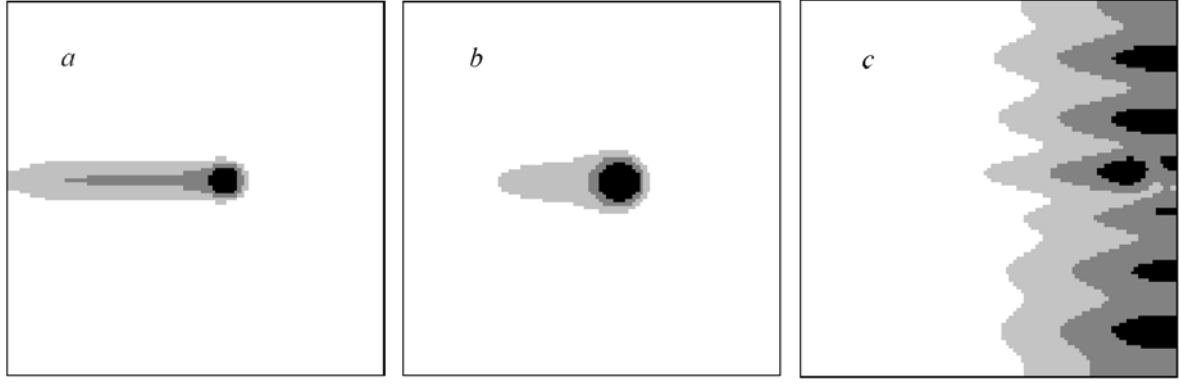


Fig. 1. Concentration distribution for $\Psi = 1.5$, $\phi_0 = 4$, $\langle C_A \rangle = 0.2$, and $\delta_\phi = -0.5$: a) isolated impurity trails in the cases $VL/(D_0\delta_D) = 10$, b) 1, and c) quasistationary distribution for $VL/(D_0\delta_D) = 10$.

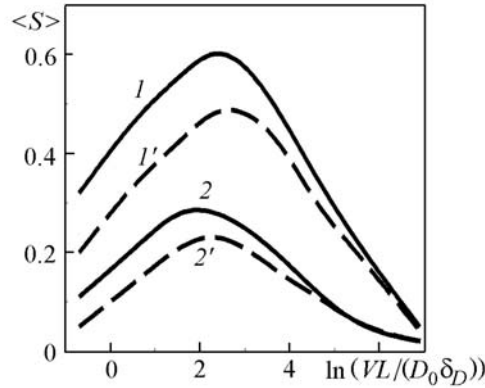


Fig. 2. Limiting degree of stratification vs. deformation rate for $\Psi = 1.5$, $\phi_0 = 4$, $\langle C_A \rangle = 0.2$, and $\delta_\phi = -0.5$ in the case of one (curves 1 and 2) and two glide systems: $\delta_\phi = -0.5$ (curves 1 and 1') and -0.2 (curves 2, 2').

According to Cottrell [15, 16], the impurity atoms are accumulated in the region of stretching or compression of the core depending on the relation of the atomic radii of the impurity and the matrix, with the result that the entire core is enriched with the impurity. According to Suzuki [26], for split dislocations, there is also the chemical interaction with the impurity. In the present model, the dislocation core is enriched with A or B atoms depending on the $\Delta\phi$ sign. Furthermore, local dissolution of precipitates can be realized in the presence of the mixing-energy gradient ($\nabla\Psi \neq 0$) [17, 18].

2. Modeling Results. Equations (2)–(5) are numerically solved by the standard Runge–Kutta method on an $L \times L$ square domain. The grain boundaries are assumed to be coincident with the square boundaries, and the atomic fluxes through the grain boundaries are assumed to be absent. For definiteness we take $\delta_D = 100$, $\delta_L = 0.05$, $R/L = 0.025$, and $\zeta/(kT) = 1$; different levels of concentration of component A in Figs. 1, 3, 5, and 7 are denoted as shades of gray. It is assumed that under constant-stress conditions, dislocations occur at the boundary $x = 0$ with a random coordinate y ; in the presence of the rotational mode of plasticity, they are generated along the perimeter of the square. The generated dislocation moves to the opposite boundary with a constant velocity V and disappears on it; at the same instant, a new dislocation is generated, so that the only dislocation is universally present in the sample. In analyzing the reached heterogeneous states, we use the quantities $\langle S \rangle$ and $\langle F \rangle$ characterizing the degree of decomposition and dispersion of an alloy and determined by the formulas

$$\langle S \rangle = \frac{1}{2L^2} \int \frac{|C_A(\mathbf{r}) - \langle C_A \rangle|}{\langle C_A \rangle \langle C_B \rangle} d\mathbf{r}, \quad \langle F \rangle = \frac{1}{L^2} \int |\nabla C_A(\mathbf{r})| d\mathbf{r}, \quad \langle C_\sigma \rangle = \frac{1}{L^2} \int C_\sigma(\mathbf{r}) d\mathbf{r}. \quad (6)$$

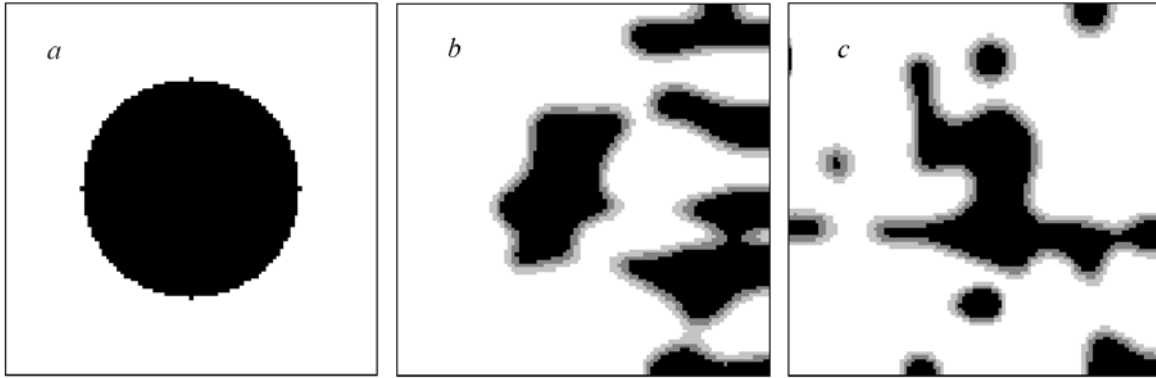


Fig. 3. Concentration distribution for $\Psi = 3$, $\varphi_0 = 4$, $\delta_\varphi = -0.5$, and $R^*/L = 0.3$: a) initial state; b) on traversal by 200 dislocations generated at the boundary $x = 0$; c) in the case of generation of the same dislocations along the perimeter of the square.

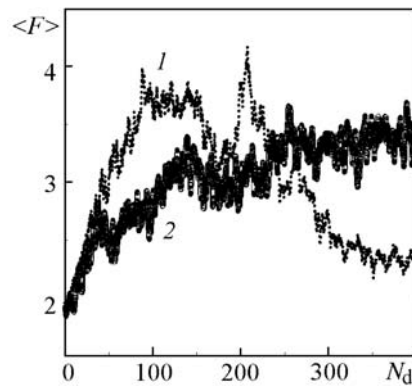


Fig. 4. Degree of dispersion of the alloy vs. number of dislocations that have traversed the sample for $\Psi = 3$, $\varphi_0 = 4$, $\delta_\varphi = -0.5$, and $R^*/L = 0.3$ in the case of generation of the dislocations at the boundary $x = 0$ (curve 1) and along the perimeter of the square (curve 2).

In most alloys, the dislocation core is preferable for segregation of atoms of one sort (Cottrell atmospheres [15, 16]), which is determined, in the present model, by the action of the factor $\delta_\varphi \neq 0$. For this case we first consider the influence of the dislocation flux on a homogeneous alloy that is stable in the absence of external actions. In dislocation glide, the impurity atoms float with the dislocations, which leads to a stratified alloy. However, for high V , no chemical equilibrium is established in the dislocation core and the energy supplied to the alloy is expended on forming extended impurity trails (Fig. 1a), which are compensated for when the action is long; therefore, the obtained stratification is slight. On the other hand, for low V , the impurity atmosphere is entrained by the dislocation (Fig. 1b), but the cumulative effect is suppressed, since the relaxation to an equilibrium is realized in the material's volume over the period of motion of the dislocation. Therefore, the stratification is maximum for the optimum deformation rate. The effect is reversible when the direction of the dislocation flux is reversed and consequently is suppressed by rotational modes. Figure 1c gives a typical quasistationary distribution of the concentrations on long-duration exposure to the dislocation flux. Figure 2 plots the limiting degrees of stratification $\langle S \rangle$ versus the deformation rate (curves 1 and 2); it is clear from the plots that the stratification is maximum for an optimum V . In the case of two glide systems (when dislocations are generated at $x = 0$ or $y = 0$ and the random second coordinate), we can assume that the dislocation is retarded on crossing the impurity trail, i.e., $V \sim (1 - C_A)$; however, $\langle S \rangle$ decreases (curves 1' and 2'). Thus, stratification of the alloy is attained by impurity transfer to the grain boundary, and nonequilibrium phases are not formed in the volume.

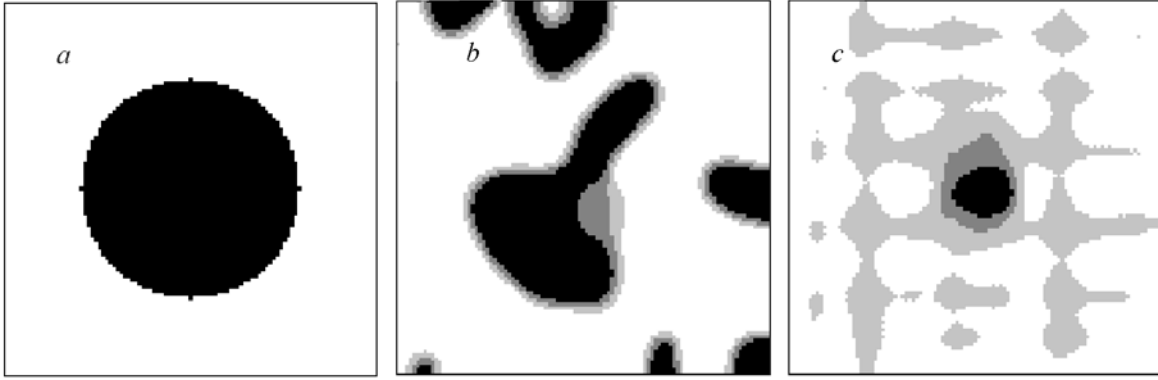


Fig. 5. Concentration distribution for $\Psi_0 = 3$, $\varphi = 0$, $\delta_\Psi = -1$, and $R^*/L = 0.3$: a) initial state; b) state reached by long-duration action in the case $VL/(D_0\delta_D) = 2$ and $N_d = 100$; and c) $VL/(D_0\delta_D) = 100$ and $N_d = 5000$.

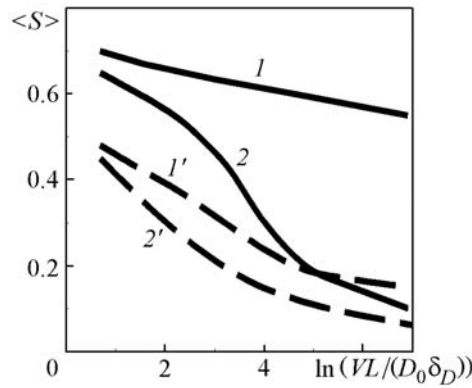


Fig. 6. Limiting degree of decomposition vs. deformation rate for $\Psi_0 = 3$, $\varphi = 0$, $\delta_\Psi = -0.5$ (curves 1 and 1') and $\delta_\Psi = -1$ (curves 2 and 2'), and the dimension of the initial precipitate of the equilibrium phase $R^*/L = 0.3$ (curves 1 and 2) and $R^*/L = 0.2$ (curves 1' and 2').

In the same materials, an equilibrium two-phase state whose boundary of existence in the model of a regular solid solution has been determined by the formula $\Psi(1 - 2C_A) = \ln(C_A/C_B)$ can be realized with reduction in the temperature [25]. Acting on the two-phase system, the dislocation flux causes the dispersion of precipitates to grow; the degree of decomposition remains constant. Figure 3 gives the initial equilibrium two-phase state, in which the precipitate of dimension R^* is at the center of the square in question (Fig. 3a), and typical concentration distributions: under the conditions of a constant dislocation flux (Fig. 3b) and in the presence of the rotational mode (Fig. 3c). Figure 4 plots the degrees of dispersion $\langle F \rangle$ versus the number of dislocations N_d that have traversed the sample. Under the conditions of a constant dislocation flux, the behavior of $\langle F \rangle$ is nonmonotonic (curve 1): first $\langle F \rangle$ grows due to the transfer of the impurity atoms from the grain volume to precipitates formed at the boundary $x = L$, but then it diminishes to the initial level as a result of the coalescence of the precipitates. Conversely, the rotational deformation model leads to the precipitate distribution throughout the volume, with the result that $\langle F \rangle$ grows to saturation (curve 2). As V grows, the attained degree of dispersion diminishes, since the dislocations have no time to entrain the impurity atoms into the matrix volume. Thus, for the rates $VL/(D_0\delta_D) = \{2, 5, 10, 15, \text{ and } 25\}$ and with the remaining parameters analogous to Fig. 3c, we have obtained the limiting values of $\langle F \rangle = \{4.5, 4.1, 3.6, 2.8, \text{ and } 2.1\}$ respectively.

Above, we have considered the action of the dislocation flux on the alloys with $\delta_\phi \neq 0$, where Cottrell impurity atmospheres are formed. If $\delta_D \cong 0$, the described effects are absent, but the nonequilibrium transformations can be due to the change in the energy Ψ in the dislocation core. We can expect the dissolution of equilibrium phases in the case $\Psi_0\delta_\Psi < 0$ and the decomposition of solid solutions stable in the absence of external action in the case $\Psi_0\delta_\Psi > 0$.

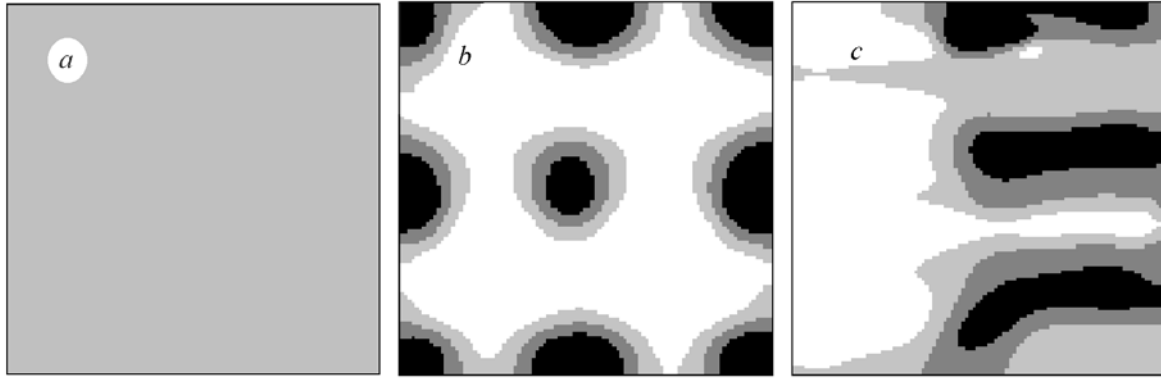


Fig. 7. Concentration distribution for $\Psi_0 = 1.5$, $\varphi = 0$, and $\langle C_A \rangle = 0.4$: a) initial state, b) state reached by long-duration action in the case $VL/(D_0\delta_D) = 1000$ and $N_d = 30,000$, and c) $VL/(D_0\delta_D) = 50$ and $N_d = 3000$.

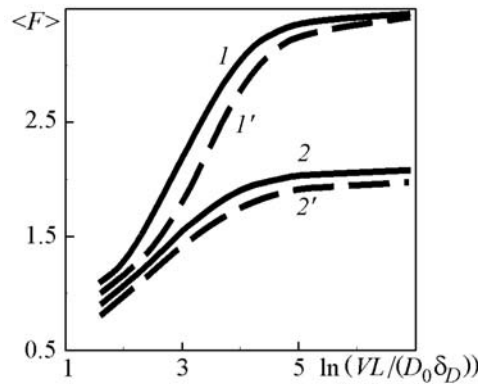


Fig. 8. Limiting degree of dispersion vs. deformation rate for $\Psi_0 = 1.5$, $\varphi = 0$, and $\langle C_A \rangle = 0.4$ in the cases $\delta\psi = 2$ (curves 1 and 1') and 1.5 (curves 2 and 2') and in generation of dislocations along the perimeter of the square (curves 1 and 2) or at the boundary $x = 0$ (curves 1' and 2').

Since the change in the conditions of ascending diffusion in the dislocation core, not the transfer of impurity atoms by the dislocations, is responsible for these transformations, we have the cumulative effect even in activation of the rotational plasticity modes; however, the disturbance amplitude must be higher than the critical. Figure 5 gives the initial equilibrium two-phase state (Fig. 5a) and typical concentration distributions in long-duration generation of dislocations along the perimeter of the square and for different deformation rates (Fig. 5b and c) under the assumption that $\Psi_0\delta\psi < 0$. It is seen that for low V , the dispersion of the alloy increases (Fig. 5b); for high V , partial dissolution of the equilibrium phase occurs (Fig. 5c). Figure 6 plots the limiting degrees of decomposition versus the deformation rate; it is clear from the plots that the solubility of precipitates increases with V and is dependent on their volume fraction. If the volume fraction of the precipitates is small, they are completely dissolved, so that $\langle S \rangle$ decreases to nearly zero.

In the case $\Psi_0\delta\psi < 0$ the dislocation flux can induce decomposition in miscible systems. The beginning of such a process is similar to the development of spinodal instability [23]. However, volume diffusion acts as the competing process, bringing the alloy back to the equilibrium homogeneous state. Therefore, the stage of coalescence of precipitates does not begin and the decomposition ends in the formation of a disperse dissipative structure (Fig. 7). The morphology of precipitates is dependent on the deformation rate, e.g., for a constant dislocation flux, the drop microstructure (Fig. 7b) gives way to a tape one (Fig. 7c). Figure 8 plots the limiting degrees of dispersion versus the deformation rate; it is clear from the plots that disperse structures are formed both under the conditions of a steady-state dislocation flux and in the presence of the rotational plasticity mode, and the degree of dispersion monotonically grows with V .

TABLE 1. Nonequilibrium Transformations Induced by Dislocation Glide

Properties of the dislocation core	Character of intense plastic deformation, initial state			
	constant dislocation flux, $V = \text{const}$		dislocation flux + rotational mode	
	I	II	I	II
$\delta_\varphi \neq 0$	Stratification for an optimum V , suppressed for alloys of the precipitated composition	Growth in dispersion for low V + coalescence of precipitates at the grain boundaries	Weak heterogeneity	Growth in dispersion for low V
$\Psi_0 \delta_\Psi < 0$	$\langle C_A \rangle = \frac{1}{2} \left(1 + \frac{\varphi_0 \delta_\varphi}{\Psi_0 \delta_\Psi} \right)$	Dissolution of precipitates for high V + segregations at the grain boundaries		Dissolution of precipitates for high V
$\Psi_0 \delta_\Psi > 0$	Disperse dissipative structures	The state remains constant	Disperse dissipative structures	The state remains constant

Note. I, homogeneous stable state; II, two-phase equilibrium state.

The considered transformations are controlled by kinetic mechanisms (dislocation glide and atomic diffusion) but have a thermodynamic motive force (energy gradients $\nabla\varphi$ and $\nabla\Psi$ in the dislocation core). Therefore, transformation tends to grow with reduction in T , whereas the kinetic capabilities are determined by the mode and rate of deformation and become weaker with reduction in T in actual systems. Consequently, nonequilibrium transformations in intense plastic deformation must be realized in the range of moderate temperatures. The alloys are broken into two classes depending on the motive force of nonequilibrium transformation: $\nabla\varphi$ or $\nabla\Psi$. In the first group, the transfer of impurity atoms by dislocations is realized; in the second group, the conditions of ascending diffusion change. We note that, according to (4), a more rigorous restriction on formation of Cottrell atmospheres is the condition $\nabla\varphi + (1 - 2C_A)\nabla\Psi = 0$ or alternatively, $\langle C_A \rangle = (1 + \varphi_0 \delta_\varphi / (\Psi_0 \delta_\Psi)) / 2$. If the value of $\langle C_A \rangle$ belongs to the permissible concentration range, the alloy of this composition is an ideal object of investigation of transformations of the second type.

3. Discussion of Results. It has been shown that nonequilibrium phase transformations due to the acceleration of diffusion and local change in the chemical potential in the region of dislocation cores are realized in the process of intense plastic deformation. In particular, a stable homogeneous alloy is stratified by the optimum-intensity action; the stratification requires that the rotational plasticity mode be suppressed. Conversely, activation of the rotational mode and increase in the deformation rate contribute to the dissolution of equilibrium phases and realization of disperse structures. The character of the reached states is dependent not only on the parameters of external action (mode and rate of deformation) but also on the parameters δ_φ , δ_Ψ , δ_D , and δ_L determining the properties of the dislocation core. Therefore, different scenarios of transformations are realized by the same action in different materials. The investigation results are generalized in Table 1.

We discuss the manner in which the presented ideas correspond to other models of phase transformations in intense plastic deformation. The decomposition of an alloy, which is induced by the nonequilibrium-vacancy flux [7, 8], is suppressed in the presence of rotational deformation modes, preserves a constant phase diagram in the grain volume, has a kinetic nature, and is realized even at high T . The action of the vacancy flux could be demonstrated by the extended nonequilibrium segregation at the grain boundaries, which occur in intense plastic deformations at temperatures much higher than those of the two-phase region or in miscible systems.

Transformations due to the disturbance of the crystal order in the region of nonequilibrium grain boundaries [9, 10] give rise to nonequilibrium grain-boundary phases in a ~ 1 -nm-thick surface layer, which is significant when a grain size of ~ 10 nm is attained. The motive force of the nonequilibrium transformation holds for a certain time after the action, since recrystallization of the lattice is time-consuming. The development of a nonequilibrium transformation in annealing after intense plastic deformation was observed, e.g., in Ni-Pd and Mu-Al alloys [3, 27].

Thus, comparing the results of Section 2, we can single out features characteristic only of transformations that are induced by dislocation glide: growth in the precipitates of nonequilibrium phases in the grain volume, mechanical alloying and decomposition of stable systems to form disperse structures in the presence of rotational plasticity modes, and existence of the optimum deformation rate for which the degree of decomposition is maximized. Transformations of a dislocation nature have, apparently, been found experimentally [4, 5, 28, 29]. In [4, 5], mechanical alloying in

intense plastic deformation in immiscible Cu–Co and Fe–Cu systems and the subsequent spinodal decomposition in annealing to form disperse structures have been observed. The latter means that the impurity atoms were distributed in the grain volume during the action and were not accumulated near the boundaries. The dissolution of intermetallides $Ni_3Me(Ti, Al, Si, \text{ and } Zr)$ in the face-centered cubic (fcc) Fe–Ni–Me matrix has been found in [28, 29], and it has been shown that the fraction of dissolved nickel atoms is linearly dependent on the degree of true deformation, which is proportional to the dislocation flux. This enabled the authors to state that the impurity atoms are entrained by dislocations to the grain volume. A comparison to the data of transmission electron microscopy with the example of an $FeNi_{35}Ti_3$ alloy has shown that the rotational deformation mode contributes to the dissolution of intermetallides. The competition between the nonequilibrium dissolution and the equilibrium precipitation of the intermetallide phase, which causes the dispersion of the precipitates to grow, has been noted. It has been noticed that homogenization of the alloy is stronger if the alloyed components have identical lattices, which confirms the dislocation nature of transformation.

Conclusions. Thus, the dislocation glide gives rise to nonequilibrium phase transformations, which are due to the acceleration of diffusion and to the change in thermodynamic properties in the region of the dislocation core. Depending on the material and the deformation mode, we have the stratification of hard solutions, the dissolution of equilibrium phases, and the formation of disperse structures. Rotational plasticity modes inhibit the stratification but contribute to the growth in the dispersion of precipitates and to the homogenization of the alloy.

The author expresses his thanks to Doctor in Physics and Mathematics Professor A. E. Ermakov and to Doctor in Physics and Mathematics Yu. N. Gornostyrev for substantive discussions.

NOTATION

C_σ , local concentration of σ atoms, $0 \leq C_\sigma \leq 1$; $\langle C_\sigma \rangle$, mean concentration of σ atoms, $0 \leq C_A \leq 1$; D and D_0 , diffusion coefficients in the dislocation core and in the material volume, m^2/sec ; f , density of free mixing energy, eV; $\langle F \rangle$, degree of dispersion of the alloy; \mathbf{J}_σ , flux of σ atoms, m/sec ; k , Boltzmann constant, $8.6 \cdot 10^{-5}$ eV/K; L , dimension of the sample, Å; M , mobility, $m^2/(sec \cdot eV)$; N_d , number of dislocations that have traversed the grain; \mathbf{r} , radius vector of a certain spatial point, Å; \mathbf{r}_0 , radius vector determining the position of the dislocation, Å; R , effective radius of atomic interaction, Å; R^* , dimension of the precipitate, Å; $\langle S \rangle$, degree of decomposition of the alloy, $0 \leq S \leq 1$; t , time, sec; T , temperature, K; V , velocity of motion of the dislocation, m/sec ; x, y , coordinates of a certain spatial point, Å; δ_D , relative change in the diffusion coefficient in the dislocation core; δ_φ and δ_Ψ , relative changes in φ and Ψ in the dislocation core; δ_L , dimensionless radius of the dislocation tube; ζ , density of surface energy, eV; μ , chemical potential, eV; ξ , shape function of disturbance of the alloy parameters in the dislocation core; φ and φ_0 , dimensionless "asymmetric" potential and its constant part; $\Phi_{\sigma\sigma}$ and $\Phi_{\sigma\sigma}^0$, energy of interaction of a σ atom with the lattice filled with σ' atoms in the alloy and undeformed pure substances, eV; Ψ and Ψ_0 , dimensionless total mixing energy and its constant part. Subscripts: i , lattice site, 1, ..., N ; σ and σ' , sorts of atoms A and B; d , dislocation.

REFERENCES

1. V. V. Neverov, V. A. Burov, and P. P. Zhitnikov, Formation of compounds and solid solutions in plastic deformation of binary mixtures of elements, *Izv. SO AN SSSR, Ser. Khim. Fiz. Nauk*, No. 12, Issue 5, 54–62 (1983).
2. A. E. Ermakov, Solid-phase reactions, nonequilibrium structures, and the magnetism of 3d-compounds with different types of chemical bond, *Fiz. Metal. Metalloved.*, No. 11, 4–45 (1991).
3. A. Ye. Yermakov, V. L. Gapontsev, V. V. Kondratyev, Yu. N. Gornostyrev, M. A. Uimin, and A. Yu. Korobeinikov, Phase instability of nanocrystalline driven alloys, *Mater. Sci. Forum*, **343**, 577–584 (2000).
4. V. A. Ivchenko, M. A. Uimin, A. Ye. Yermakov, and A. Yu. Korobeinikov, Atomic structure and magnetic properties of $Cu_{80}Co_{20}$ nanocrystalline compounds produced by mechanical alloying, *Surf. Sci.*, **40**, No. 3, 420–428 (1999).
5. U. Czubayko, N. Wanderka, V. Naundorf, V. A. Ivchenko, A. Ye. Yermakov, M. A. Uimin, and H. Wollenberg, Characterization of nanoscaled heterogeneities in mechanically alloyed and compacted Cu–Fe, *Mater. Sci. Forum*, **343–346**, 709–714 (2000).

6. V. A. Barinov, G. A. Dorofeev, L. V. Ovechkin, et al., Phase transformations in deformed Fe₂B powders, *Fiz. Metal. Metalloved.*, No. 1, 126–131 (1992).
7. A. E. Ermakov, V. L. Gapontsev, V. V. Kondrat'ev, and Yu. N. Gornostyrev, The phenomenon of deformation-stimulated phase instability of nanocrystalline alloys, *Fiz. Metal. Metalloved.*, **88**, 5–12 (1999).
8. V. L. Gapontsev, I. K. Razumov, Yu. N. Gornostyrev, A. E. Ermakov, and V. V. Kondrat'ev, Theory of diffusion phase transformations in nanocrystalline alloys in intense plastic deformation, *Fiz. Metal. Metalloved.*, **99**, No. 4, 26–37 (2005).
9. I. K. Razumov and Yu. N. Gornostyrev, Influence of the grain boundaries on the kinetics of decomposition of solid solutions, in: *Proc. of the 4th School-Seminar "Phase and Structural Transformations in Steels,"* Magnitogorsk (2006), pp. 99–112.
10. I. K. Razumov, Yu. N. Gornostyrev, and A. Ye. Yermakov, Kinetics of spinodal decomposition in driven nanocrystalline alloys, *J. Alloys and Compounds*, **434–435**, 535–539 (2007).
11. R. Z. Valiev and I. V. Aleksandrov, *Nanostructural Materials Obtained by Means of Intense Plastic Deformation* [in Russian], Logos, Moscow (2000).
12. N. M. Vlasov, A. S. Gontar', and V. A. Zaznoba, Decomposition of a solid solution on large plastic deformations of an alloy, *Fiz. Tverd. Tela*, **71**, No. 5, 63–66 (2001).
13. M. Yu. Gutkin and I. A. Ovid'ko, *Defects and Mechanisms of Plasticity in Nanostructural and Noncrystalline Materials* [in Russian], Yanus, St. Petersburg (2001).
14. K. Zhang, I. V. Alexandrov, and K. Lu, The X-ray diffraction study on a nanocrystalline Cu processed by equal — channel angular pressing, *NanoStruct. Mater.*, **9**, No. 1, 347–350 (1997).
15. A. H. Cottrell and M. A. Jaswon, Distribution of solute atoms round a slow dislocation, *Proc. Roy. Soc., London*, **199**, No. 189, 104–114 (1949).
16. A. H. Cottrell, *The Theory of Crystal Dislocation* [Russian translation], Metallurgizdat, Moscow (1958).
17. B. Ya. Lyubov, *Diffusion Processes in Inhomogeneous Solid Media* [in Russian], Nauka, Moscow (1981).
18. B. Ya. Lyubov and V. A. Shmakov, Theory of diffusion interaction of edge dislocations with new-phase precipitates, *Fiz. Metal. Metalloved.*, **29**, No. 5, 968–979 (1970).
19. M. Legros, G. Dehm, E. Arzt, et al., Giant diffusivity along dislocation cores, *Science*, **319**, No. 5870, 1646–1649 (2008).
20. R. Yu. Rakitin, G. M. Poletaev, M. S. Aksenov, and M. D. Starostenkov, Mechanisms of diffusion over grain boundaries in two-dimensional metals, *Pis'ma Zh. Tekh. Fiz.*, **31**, Issue 15, 44–48 (2005).
21. É. V. Dobrokhotov, Diffusion in a dislocation Ge and the model of a "liquid" dislocation core, *Fiz. Tverd. Tela*, **47**, Issue 12, 2166–2169 (2005).
22. V. G. Eremenko, V. I. Nikitenko, and E. B. Yakimov, Abnormal anisotropy of the mobility of electrons in plastically deformed silicon with a low density of dislocations, *Pis'ma Zh. Eksp. Teor. Fiz.*, **26**, Issue 2, 72–75 (1977).
23. J. W. Cahn, On spinodal decomposition, *Acta Metall.*, **9**, 795–806 (1961).
24. I. K. Razumov, Influence of lattice relaxation on the kinetics of spinodal decomposition of solid solutions, *Inzh.-Fiz. Zh.*, **82**, No. 4, 643–648 (2009).
25. J. Christian, *The Theory of Transformations in Metals and Alloys* [Russian translation], Mir, Moscow (1978).
26. H. Suzuki, *Dislocations and Mechanical Properties of Crystals*, New York (1957).
27. E. I. Teitel', M. A. Uimin, A. E. Ermakov, A. V. Shangurov, G. M. Makarova, V. A. Barinov, V. P. Pilyugin, and R. I. Kuznetsov, Change in the structural state and magnetic properties on annealing of a strongly deformed MnAl–C alloy, *Fiz. Metal. Metalloved.*, No. 8, 83–89 (1990).
28. V. V. Sagaradze, V. A. Shabashov, T. M. Lapina, N. L. Pecherkina, and V. P. Pilyugin, Low-temperature deformation dissolution of intermetallide Ni₃Al(Ti, Si, Zr) phases in Fe–Ni alloys with the fcc lattice, *Fiz. Metal. Metalloved.*, **78**, No. 6, 49–61 (1994).
29. A. G. Mukoseev, V. A. Shabashov, V. P. Pilyugin, and V. V. Sagaradze, Deformation-induced formation of a solid solution in the Fe–Ni system, *Fiz. Metal. Metalloved.*, **85**, 5–15 (1998).