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The influence of grain boundary movement on radiation-induced segregation in binary alloys

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Abstract

An analytical model of radiation-induced segregation (RIS) which takes into account absorption and desorption of the point defects at grain boundaries is developed. Using this model, interrelation between the grain boundary movement and RIS is investigated. It has been shown that fast diffusing alloy atoms are depleted at the grain boundary only when the boundary velocity does not exceed some critical value ($v < v_c$). In the opposite case ($v > v_c$) the fast atoms are accumulated at the grain boundary. These analytical results are compared to those obtained in the experiments. © 1998 Elsevier Science B.V. All rights reserved.

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1. Introduction

High levels of point defect supersaturations in irradiated alloys result in directed diffusion currents of these defects from the interior of a material to grain and interphase boundaries. These currents lead to diffusion of the alloy atoms [1]. Due to the different mobilities of the alloy components this causes the spatial segregation of the alloy atomic species during irradiation [2,3]. The radiation induced segregation (RIS) can stimulate unusual phase transitions and considerably change the material properties. The segregation also changes the rates of the defect annihilation at sinks and influences the radiation swelling of alloys [4].

The possible influence of RIS on material properties simulated numerous experimental and analytical investigations [2–6]. In spite of the advancements in the description of RIS, the analytical investigations usually neglected the evolutions of point defect sinks during interactions of these sinks with point defects. For example, to formulate an analytical description of the unusual accumulation of the fast diffusing alloy component at the moving grain boundary during RIS, which was observed in [5,6] when this boundary velocity was sufficiently high, one has to take accurately into account the kinetics of point defects in this boundary vicinity.

In this paper we formulate the RIS problem in a self-consistent manner, taking into account the microscopic processes at the surfaces of the point defect sinks. Using the developed approach, correlations between the grain boundary movement and RIS have been investigated.

In Section 2 we present a system of the diffusion equations and appropriate boundary conditions that describe the segregation of the alloy atomic species during irradiation and take into account evaporation and absorption of point defects at the grain boundary. In Section 3 we investigate the influence of the grain boundary movement on RIS and obtain a critical velocity of the grain boundary at which no segregation of the alloy species takes place.

2. Formulation of the problem

In this paper we consider a two-component alloy as an ideal solution with random distribution of A and B kinds of atoms. These atoms can occupy both the lattice sites and interstitial positions. Diffusion of the alloy

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atoms results from jumps of point defects (vacancies and interstitials) through the lattice. We describe here a semi-infinite irradiated solid with a flat free surface and assume that dislocations and voids constitute the volume sinks for point defects. To take into account the defect accumulations in the irradiated alloy we introduce spatially depending generation rates, both for vacancies, $G_v(\vec{r})$ and interstitials, $G_i(\vec{r}) = G_{Ai} + G_{Bi}$ (Where G_{Yi} are the partial generation rates for the alloy species, $Y = A, B$ and $G_v = G_i$).

2.1. Governing equations of the model

Defining the total volume density of the lattice sites, which can be occupied by atoms, vacancies and voids as $n = \omega^{-1}$ (where ω is the atomic volume), we set the following system of equations describing temporal and spatial variations of the atomic and point defect concentrations in the 'mean field' [7] approximation for the dilute ensemble of the volume sinks:

$$\frac{\partial C_v}{\partial t} = -\omega\nabla\overrightarrow{j_v} + \omega G_v - R_A C_v C_{Ai} - R_B C_v C_{Bi} - k_v^2 D_v\left(C_v - \bar{C}_v^{\mathrm{eq}}\right), \tag{1}$$

$$\frac{\partial C_Y}{\partial t} = -\omega\nabla\overrightarrow{j_Y} - \omega G_{Yi} + R_Y C_v C_{Yi} + (k_{Yi})^2 D_{Yi}\left(C_{Yi} - \bar{C}_{Yi}^{\mathrm{eq}}\right), \tag{2}$$

$$\frac{\partial C_{Yi}}{\partial t} = -\omega\nabla\overrightarrow{j_{Yi}} + \omega G_{Yi} - R_Y C_v C_{Yi} - (k_{Yi})^2 D_{Yi}\left(C_{Yi} - \bar{C}_{Yi}^{\mathrm{eq}}\right), \tag{3}$$

$$\frac{\partial C_s}{\partial t} = k_v^2 D_v\left(C_v - \bar{C}_v^{\mathrm{eq}}\right) - \sum_{Y=A,B} (k_{Yi})^2 D_{Yi}\left(C_{Yi} - \bar{C}_{Yi}^{\mathrm{eq}}\right). \tag{4}$$

Here $Y = A, B; C_v = \omega\langle n_v\rangle, C_Y = \omega\langle n_Y\rangle, C_{Yi} = \omega\langle n_{Yi}\rangle, C_s = \omega\langle n_s\rangle$; $\langle n_v\rangle$ is the volume density of the vacancies averaged over the ensemble of the volume sinks; $\langle n_Y\rangle$ and $\langle n_{Yi}\rangle$ are the same for Y-atoms occupying the substitutional and interstitial positions respectively; $\langle n_s\rangle$ is the averaged volume density of the lattice sites occupied by voids; R_Y is the recombination factor for the vacancies and Y-interstitials; $k_\beta^2 = Z_\beta^{\mathrm{d}}\rho_{\mathrm{d}} + 4\pi N_{\mathrm{V}}\langle Z_\beta^V R_{\mathrm{V}}\rangle(\beta = v, \mathrm{A}_i, \mathrm{B}_i)$; $Z_\beta^{d/v}$ are the bias factors for the absorption of β-defect at the dislocations/voids; N_{V} and R_{V} the void volume density and the void radius respectively; $\langle\ \rangle$ denotes the average over the ensemble of the volume sinks; D_β is the diffusion coefficient of the β-defects; ρ_{d} is the dislocation density; $\bar{C}_\beta^{\mathrm{eq}}$ are the equilibrium defect concentrations at the surfaces of the volume sinks averaged over the ensemble of the volume sinks. The equilibrium interstitial concentrations at the sink surfaces are usually neglected in the densely packing metals, but keeping in mind materials with low volume density of the lattice sites, we retain these concentrations in the 'sink' terms of the system (1)–(3).

The system (1)–(3) is correlated with obvious 'conservation law' for the volume density of the lattice sites

$$C_A + C_B + C_v + C_s = 1. \tag{5}$$

In the present treatment we consider separate equations for A and B interstitials, implying that the type of an interstitial is not changed during diffusion. This assumption is introduced in order to take into account the creation of new Y-atoms in the substitutional sites, which is sensitive to a particular interstitial type.

Diffusion currents are defined by the following equations [2,3,6]:

$$\omega\overrightarrow{j_Y} = -d_{Yv}C_v\nabla C_Y + d_{Yv}C_Y\nabla C_v, \tag{6}$$

$$\omega\overrightarrow{j_{Yi}} = -d_i\nabla C_{Yi}, \tag{7}$$

$$\omega\overrightarrow{j_v} = -D_v\nabla C_v + d_{Av}C_v\nabla C_A + d_{Bv}C_v\nabla C_B. \tag{8}$$

Here $D_v = d_{Av}C_A + d_{Bv}C_B; d_{Y\beta} = \frac{1}{6}\lambda_\beta^2 z_\beta v_{Y\beta}$ is the partial diffusion coefficient [2]; λ_β, z_β are, respectively, the diffusion jump length and the number of nearest neighbors for β-defect; $v_{Y\beta} = v\exp(-\Phi_{Y\beta}^m/T)$, v is the attempt frequency of atomic jumps; $\Phi_{Y\beta}^m$ is the free enthalpy of the Y-atom migration via the β-defect. For simplicity we suppose equal partial diffusion coefficients of the different alloy species via interstitials ($D_{Yi} = d_{Yi} = d_i$).

Substituting Eqs. (6)–(8) into Eqs. (1)–(3) we obtain

$$\frac{\partial C_v}{\partial t} = (d_{Av}C_A + d_{Bv}C_B)\nabla^2 C_v - \left(d_{Av}\nabla^2 C_A + d_{Bv}\nabla^2 C_B\right)C_v + \omega G_v - R_A C_v C_{Ai} - R_B C_v C_{Bi} - k_v^2 D_v\left(C_v - \bar{C}_v^{\mathrm{eq}}\right), \tag{9}$$

$$\frac{\partial C_Y}{\partial t} = d_{Yv}\left(C_v\nabla^2 C_Y - C_Y\nabla^2 C_v\right) - \omega G_{Yi} + R_Y C_v C_{Yi} + (k_{Yi})^2 d_i\left(C_{Yi} - \bar{C}_{Yi}^{\mathrm{eq}}\right), \tag{10}$$

$$\frac{\partial C_{Yi}}{\partial t} = d_i\nabla^2 C_{Yi} + \omega G_{Yi} - R_Y C_v C_{Yi} - (k_{Yi})^2 d_i\left(C_{Yi} - \bar{C}_{Yi}^{\mathrm{eq}}\right). \tag{11}$$

We assume an equilibrium and space uniform initial conditions for the concentrations of the defects and the alloy species:

$$C_\beta(x, 0) = C_{\beta,\mathrm{eq}}^0, \qquad \beta = v, Ai, Bi, \tag{12}$$

$$C_Y(x, 0) = C_Y^0 = \mathrm{const}, \qquad Y = A, B, \tag{13}$$

where x is the distance from the sample flat surface at the initial time and $C_{\beta,\mathrm{eq}}^0$ are initial equilibrium defect concentrations at the flat surface.

The evaporation and absorption of the defects at the sample surface result in the surface movement.

Therefore, in addition to the boundary conditions for the diffusion currents of point defects at the sample surface [8]:

$$j^{\beta}_{x=l(t)} = \left(j^{Q}_{\beta} - j^{P}_{\beta}\right), \tag{14}$$

we have to set an equation describing this surface movement

$$\frac{\mathrm{d}l}{\mathrm{d}t} = \omega(j_{Ai} + j_{Bi} - j_v)_{x=l(t)}. \tag{15}$$

Here $l(t)$ is the position of the sample surface at time t and j^{Q}_{β} and j^{P}_{β} are, respectively, the number of β-defects evaporated and absorbed by a unit surface per second.

We also demand that values of all concentrations are bounded above.

3. Influence of the grain boundary movement on RIS

Eqs. (9)–(11) are too complicated for an analytical solution and require numerical solution which will be presented elsewhere. Below, based on Eqs. (9)–(15) we establish a relation between the grain boundary migration and RIS.

3.1. Steady-state diffusion of point defects

Because of the flat surface of the sample, the characteristic lengths of our problem are the screening lengths for the diffusion of the defects. These lengths are defined by the strengths of the volume sinks of the point defects [7];

$$\lambda_{s\beta} = k_{\beta}^{-1}. \tag{16}$$

When the surface movement results from the diffusion of point defects it is easy to show that the characteristic times of the defect diffusion t^{d}_{β} are much smaller than that of the surface movement $t_l \sim \lambda_{s\beta}/|v|$ ($v = \mathrm{d}l/\mathrm{d}t$ is the surface velocity):

$$\left(t_l/t^{\mathrm{d}}_{\beta}\right) \geqslant C_{\beta}^{-1} \gg 1. \tag{17}$$

Moreover, the characteristic times t^{d}_{Y} of the diffusion of the atoms occupying the lattice sites are also much larger than those of the defect diffusion:

$$\left(t^{\mathrm{d}}_{Y}/t^{\mathrm{d}}_{\beta}\right) \sim C_{\beta}^{-1} \gg 1. \tag{18}$$

Eqs. (17) and (18) allow us to use below the coordinate system connected with the moving surface of the sample and assume that in this coordinate system the spatial profiles of the point defects are given by the steady-state equations:

$$\begin{aligned} &D_v\nabla^2 C_v - \left(d_{Av}\nabla^2 C_A + d_{Bv}\nabla^2 C_B\right)C_v + \omega G_v \\ &- R_A C_v C_{Ai} - R_B C_v C_{Bi} - k_v^2 D_v\left(C_v - \bar{C}_v^{\mathrm{eq}}\right) = 0, \end{aligned} \tag{19}$$

$$\begin{aligned} &d_i\nabla^2 C_{Yi} + \omega G_{Yi} - R_Y C_v C_{Yi} - (k_{Yi})^2 d_i\left(C_{Yi} - \bar{C}_{Yi}^{\mathrm{eq}}\right) = 0, \\ &Y = A, B. \end{aligned} \tag{20}$$

The steady-state spatial profiles of these concentrations in the moving coordinate system ($\tilde{x} = x - vt$) are monotonic, have positive slopes, saturate far from the surface to some radiation depended steady-state values and do not depend on the surface velocity:

$$\begin{aligned} &\nabla^2 C_{v/i} < 0, \qquad 0 < \tilde{x} < \infty, \\ &\nabla C_{v/i} \quad > 0, \qquad 0 < \tilde{x} < \infty, \\ &\nabla^2 C_{v/i}, \nabla C_{v/i} \to 0, \qquad \tilde{x} \gg \lambda_{sv/i}. \end{aligned} \tag{21}$$

3.2. An influence of the surface velocity on RIS

An accumulation of the fast alloy component at the moving grain boundary during RIS was observed in [5,6] when this boundary velocity was sufficiently high and an ordinary depletion of this component was obtained at lower boundary velocities. Based on the presented model we can describe this unusual effect.

Substituting Eqs. (7), (8) and (5) into Eq. (15) we obtain

$$\begin{aligned} v = {} & [D_v\nabla C_v - d_i\nabla C_i]_{\tilde{x}=0} - [(d_{Av} \\ & - d_{Bv})C_v\nabla C_A]_{\tilde{x}=0} + [d_{Bv}C_v(\nabla C_v + \nabla C_s)]_{\tilde{x}=0}, \end{aligned} \tag{22}$$

where $\tilde{x} = x - vt$ and $C_i = C_{Ai} + C_{Bi}$. Below we shall neglect the third term in Eq. (22) due to small vacancy and void concentrations and uniform spatial distribution of the voids

$$\begin{aligned} &d_{Bv}C_v \ll D_v = d_{Av}C_A + d_{Bv}C_B; \\ &C_v \ll C_Y; C_s \ll C_Y; \nabla C_s = 0. \end{aligned}$$

When $(D_v\nabla C_v)_{\tilde{x}=0} > (d_i\nabla C_i)_{\tilde{x}=0}$, one can see from Eq. (22) that for the fast diffusing A atoms ($d_{Av} > d_{Bv}$) (see Fig. 1) we obtain their accumulation at the sample surface ($\nabla C_A < 0$) when the surface velocity exceeds some critical value ($v > v_{\mathrm{c}}$) and have the opposite case when this velocity is less than the critical one ($v < v_{\mathrm{c}}$). We have a uniform distribution of the alloy species (C^{u}_{Y}) when the surface velocity coincides with its critical value:

$$\nabla C_A < 0, \quad v > v_{\mathrm{c}}, \qquad \nabla C_A > 0, \quad v < v_{\mathrm{c}}, \tag{23}$$

$$\nabla C^{\mathrm{u}}_{A} = 0, \quad v = v_c, \qquad v_{\mathrm{c}} = D^{\mathrm{u}}_{v}\nabla C^{\mathrm{u}}_{v} - d_i\nabla C^{\mathrm{u}}_{i}, \tag{24}$$

the diffusion coefficient $D^{\mathrm{u}}_{v} = d^{v}_{A}C^{\mathrm{u}}_{A} + d^{v}_{B}C^{\mathrm{u}}_{B}$ and concentrations $C^{\mathrm{u}}_{v}, C^{\mathrm{u}}_{i}$ are calculated using uniform steady-state spatial distribution of the species $C^{\mathrm{u}}_{A}, C^{\mathrm{u}}_{B}$.

3.3. Definition of the critical surface velocity

As follows from Eqs. (22)–(24), in order to define the critical velocity we can calculate the steady-state defect

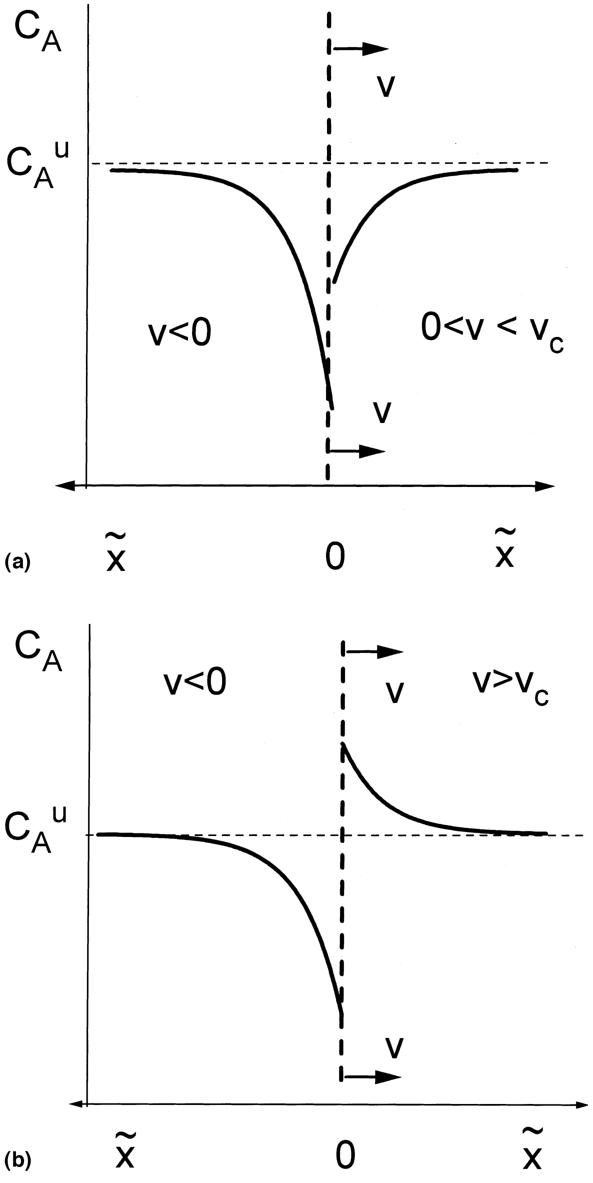

Fig. 1. Steady-state spatial distribution of the fast diffusing alloy atoms as a function of the velocity of the grain boundary movement: (a) $v < v_c$; (b) $v > v_c$. The current grain boundary position is located at $\tilde{x} = 0$. The left sides of these figures correspond to negative grain boundary velocities.

concentrations $C^{\mathrm{u}}_{\beta}(\infty)$ assuming the uniform spatial distribution of the alloy species $(C^{\mathrm{u}}_{A}, C^{\mathrm{u}}_{B})$. In this case boundary conditions (14) and (15) for Eqs. (19) and (20) are reduced to

$$\left.\begin{array}{l} C^{\mathrm{u}}_{\beta}(\infty) \text{ is bounded above} \\ j_{\beta\mathrm{u}}(0) = -(D^{\mathrm{u}}_{\beta}\nabla C^{\mathrm{u}}_{\beta}/\omega)_{\tilde{x}=0} = j^{Q}_{\beta} - j^{P}_{\beta} \end{array}\right\} \qquad (25)$$

For sufficiently high temperatures ($T > 0.4T_{\mathrm{m}}$, where T_{m} is the material melting temperature) which are interesting for RIS we can neglect recombination terms in Eqs. (19) and (20) [9,10]. Neglecting the spatial dependencies of the generation rates G_{β} and the sink strengths k^2_{β} we obtain the following solution of Eqs. (19) and (20)

$$C^{\mathrm{u}}_{\beta}(\tilde{x}) = \tilde{C}_{\beta}\,\exp(-k_{\beta}\tilde{x}) + g_{\beta}/(D^{\mathrm{u}}_{\beta}k_{\beta}), \qquad (26)$$

Where $\tilde{C}_{\beta}$ are constants which can be defined from boundary conditions and

$$g_v = \omega G_v - k^2_v D_v \bar{C}^{\mathrm{eq}}_v, \qquad (27)$$

$$g_{Yi} = \omega G_{Yi} - (k_{Yi})^2 d_i \bar{C}^{\mathrm{eq}}_{Yi}. \qquad (28)$$

From Eq. (26) we obtain the diffusion current of β-defects at the sample surface:

$$j_\beta^{\mathrm{u}}(0) = \tilde{C} k_\beta D_\beta^{\mathrm{u}}/\omega. \tag{29}$$

To define the probabilities of absorption and desorption of point defects at the sample surface we shall use a method developed in [8] for the description of the nucleation of gas bubbles in irradiated materials.

The number of the point defects (vacancies) absorbed by a unit surface per second (j_{vu}^{P}) is defined by the vacancy concentration in the subsurface layer whose thickness is of the order of the diffusion jump length of the vacancy, and the frequency of the vacancy exchanges with the alloy atoms

$$j_v^{P\mathrm{u}} = \frac{C_v^{\mathrm{u}}(\lambda_{dv})\lambda_{dv}\xi_v \nu}{6\omega}\{C_A^{\mathrm{u}} \exp[-(\varepsilon_{Av} + \Phi_{Av}^m)/T) \\ + C_B^{\mathrm{u}} \exp[-(\varepsilon_{Bv} + \Phi_{Bv}^m)/T)\}, \tag{30}$$

where λ_{dv} is the length of the diffusion jump of the vacancy, ξ_v the fraction of the surface sites which are available for the absorption and desorption of the vacancy, ε_{Yv} is an additional diffusion barrier of the vacancy jump due to distortion of the lattice near the sample surface.

In Eq. (26) the screening length of the defect diffusion is much larger than the characteristic diffusion jump length ($\lambda_{\beta s} = k_\beta^{-1} \sim 10^{-5}$ cm, $\lambda_{d\beta} \sim 3 \times 10^{-8}$ cm)

$$\lambda_{\beta s} \gg \lambda_{d\beta} \tag{31}$$

and we can set in Eq. (30)

$$C_v^{\mathrm{u}}(\lambda_{dv}) = C_v^{\mathrm{u}}(0). \tag{32}$$

Moreover, as described below (see Eq. (39)), in case (31) we can neglect differences between the additional diffusion barriers

$$\varepsilon_{Av} = \varepsilon_{Bv} = \varepsilon_v. \tag{33}$$

Substituting Eqs. (33) and (32) into Eq. (30) we obtain

$$j_v^{P\mathrm{u}} = \frac{C_v^{\mathrm{u}}(0)\xi_v}{\lambda_{dv}\omega} D_v^{\mathrm{u}} \exp(-\varepsilon_v/T), \tag{34}$$

where $D_v^{\mathrm{u}} = d_A^v C_A^{\mathrm{u}} + d_B^v C_B^{\mathrm{u}}$.

The number of the vacancies evaporated by a unit surface per second ($j_v^{Q\mathrm{u}}$) is defined by the equilibrium vacancy concentration $C_{v,\mathrm{eq}}$ at the flat surface and the frequency of their exchange with the alloy atoms in the subsurface layer

$$j_v^{Q\mathrm{u}} = \frac{C_{v,\mathrm{eq}}\xi_v}{\lambda_d^v \omega} D_v^{\mathrm{u}} \exp(-\varepsilon_v/T). \tag{35}$$

The probabilities of absorption and desorption of the interstitials at the sample surface can be obtained like in Eqs. (34) and (35) and the differences between the absorption and desorption probabilities of β-defects at this surface have the following form:

$$j_\beta^{Q\mathrm{u}} - j_\beta^{P\mathrm{u}} = \frac{D_\beta^{\mathrm{u}}}{\omega\lambda_{\beta d}} \xi_\beta \exp(-\varepsilon_\beta/T)[C_{\beta,\mathrm{eq}} - C_\beta(0)], \tag{36}$$

where $\lambda_{\beta d}$ is the length of the diffusion jump of the β-defect, ξ_β the fraction of the surface sites which are available for the absorption and desorption of the β-defects, ε_β is an additional barrier of the β-defect diffusion near the sample surface and $C_{\beta,\mathrm{eq}}$ is the equilibrium concentration of the β-defects at the flat surface.

Substituting Eq. (36) into Eqs. (25) and (29) we obtain

$$C_\beta^{\mathrm{u}}(\tilde{x}) = \frac{\omega G_\beta \lambda_{\beta s}^2}{D_\beta^{\mathrm{u}}} - \frac{\lambda_{\beta s}}{\lambda_{\beta s} + \tilde{\lambda}_{\beta d}} \\ \times \left[(\bar{C}_\beta^{\mathrm{eq}} - C_{\beta,\mathrm{eq}}) + \frac{\omega G_\beta \lambda_{\beta s}^2}{D_\beta^{\mathrm{u}}}\right] \exp(-k_\beta \tilde{x}), \tag{37}$$

$$j_\beta^{\mathrm{u}}(0) = -\frac{D_\beta^{\mathrm{u}}}{\omega(\lambda_{\beta s} + \tilde{\lambda}_{\beta d})}\left[\left(\bar{C}_\beta^{\mathrm{eq}} - C_{\beta,\mathrm{eq}}\right) + \frac{\omega G_\beta \lambda_{\beta s}^2}{D_\beta^{\mathrm{u}}}\right]. \tag{38}$$

Here $\tilde{\lambda}_{\beta d} = \lambda_{\beta d} \exp(\varepsilon_\beta/T)/\xi_\beta$. In the diffusion limited case ($\tilde{\lambda}_{\beta d} \ll \lambda_{\beta s}$) the informaion about the additional diffusion barriers ε_β and value ξ_β is lost in Eq. (38) and we obtain

$$j_\beta^{\mathrm{u}}(0) = -\frac{D_\beta^{\mathrm{u}}}{\omega\lambda_{\beta s}}\left[(\bar{C}_\beta^{\mathrm{eq}} - C_{\beta,\mathrm{eq}}) + \frac{\omega G_\beta \lambda_{\beta s}^2}{D_\beta^{\mathrm{u}}}\right]. \tag{39}$$

Using Eqs. (39) and (15) we obtain the critical value of the grain boundary velocity during RIS

$$v_c = \frac{D_v^{\mathrm{u}}}{\lambda_{vs}}(\bar{C}_v^{\mathrm{eq}} - C_{v\mathrm{eq}}) \\ + \sum_{Y=A,B}\left[-\frac{d_i}{\lambda_{Yis}}(\bar{C}_{Yi}^{\mathrm{eq}} - C_{Yi,\mathrm{eq}}) + \omega G_{Yi}|\delta\lambda_{Yis}|\right]. \tag{40}$$

We introduce here the values

$$\delta\lambda_{Yis} = \lambda_{vs} - \lambda_{Yis} \simeq \frac{1}{2k_v}\frac{k_{Yi}^2 - k_v^2}{k_v^2}, \tag{41}$$

which result from the different bias factors $Z_{d/b}^\beta$ (and, hence, different sink strengths k_β^2) for the vacancies and interstitials. When

$$\delta\lambda_{Ais} \approx \delta\lambda_{Bis} = \delta\lambda_s \\ = \frac{(Z_i^{\mathrm{d}} - Z_v^{\mathrm{d}})\rho_d + 4\pi N_{\mathrm{V}}\langle(Z_i^V - Z_v^V)R_{\mathrm{V}}\rangle}{2(Z_v^{\mathrm{d}}\rho_d + 4\pi N_{\mathrm{V}}\langle Z_v^V R_{\mathrm{V}}\rangle)^{3/2}},$$

we obtain

$$v_c = |\delta\lambda_s|\omega G_v + \frac{D_v^{\mathrm{u}}}{\lambda_{vs}}(\bar{C}_v^{\mathrm{eq}} - C_{v\mathrm{eq}}) \\ - \sum_{Y=A,B}\frac{d_i}{\lambda_{Yis}}\left(\bar{C}_{Yi}^{\mathrm{eq}} - C_{Yi,\mathrm{eq}}\right). \tag{42}$$

The value of the critical velocity can be separated to the ‘radiation’ and ‘thermal’ parts. The radiation terms are important when the dislocations are the dominant volume sinks ($\bar{C}_\beta^{\mathrm{eq}} = C_{\beta\mathrm{eq}}$) and/or the generation rate is high ($|\delta\lambda_s|\omega G_v \gg D_\beta^{\mathrm{u}}\bar{C}_\beta^{\mathrm{eq}}/\lambda_{\beta s}$). The value of this term depends on the difference between the bias factors $Z_i^{d/V} - Z_v^{d/V}$ of the volume sinks for the vacancies and interstitials. At the parameter values used in [5,6] ($k_v^2 = 10^{10}$ cm^{-2}, $k_i^2 = 1.3 \times 10^{10}$ cm^{-2}, $G\omega = 5 \times 10^{-4}$ dpa/s) we estimate 'radiation' term of the critical velocity

$$v_c^{(r)} \simeq 0.75 \times 10^{-9} \text{ cm/s} = 15 \text{ nm/dpa}. \tag{43}$$

Having in mind the results obtained in this approach, at the developed stage of the radiation swelling, when surfaces of large voids can be assumed as flat the depletion of the fast diffusing alloy atoms at the surface of these voids may change to the enrichment. Because RIS of alloy species at the surfaces of the point defect sinks influences the rate of the defect annihilation and bias factors [4], this change of RIS mode may result in a change of the radiation swelling rate. This problem will be investigated elsewhere.

4. Summary

In this paper the interrelation between the grain boundary velocity and RIS was investigated. It was shown that the fast diffusing alloy component is depleted at the grain boundary only when this boundary velocity does not exceed some critical value ($v < v_c$). In opposite case ($v > v_c$) the fast component is accumulated at the grain boundary. These analytical results are in a good agreement with those obtained in the experiments [5,6].

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