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IRRADIATION DEFORMATION DUE TO SIPA INDUCED DISLOCATION ANISOTROPY

DEFORMATION PAR IRRADIATION PRODUITE PAR L'ANISOTROPIE DE DISLOCATION PAR SIPA

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RESUME

On examine la contribution que joue l'effet d'adsorption préférée produit par contraintes (SIPA) dans la déformation par irradiation. SIPA fait varier les vitesses de croissance des boucles de dislocation produites par irradiation suivant l'alignement de leurs vecteurs de Burgers, en fonction de la contrainte exercée. La prolongation sous une contrainte appliquée produit alors une structure anisotropique de dislocation dans laquelle la majorité des boucles de dislocations ont leurs vecteurs de Burgers dans le même plan que la contrainte. En présence de zones "neutres", la structure anisotropique de dislocation qui en résulte produit une déformation plastique comparable à la manière dont la croissance par irradiation a lieu dans le zirconium. On appelle ce méchanisme: la croissance produite par SIPA (SIG). Nous avons montré que la croissance SIG est très importante par rapport à l'effet SIPA, sauf si la croissance des boucles a été faible ou nulle pendant la période d'application de la contrainte.

Le présent rapport comporte la formulation détaillée et la dérivation des formules utilisées pour évaluer le rôle que joue la croissance produite par SIPA (SIG).

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ABSTRACT

A contribution to irradiation deformation resulting from the stress-induced preferred adsorption (SIPA) effect is considered. SIPA causes a variation of the growth rates of irradiation-generated dislocation loops, according to the alignment of their Burgers vectors with respect to the applied stress. A prolonged period under an applied stress then creates an anisotropic dislocation structure in which the majority of dislocations have their Burgers vectors in alignment with the stress. In the presence of "neutral" sinks, the resulting anisotropic dislocation structure causes plastic deformation similar to the way in which irradiation growth occurs in zirconium. This mechanism is called SIPA-induced growth (SIG). We have shown that SIG is very significant in comparison to SIPA, except when little or no loop growth has occurred during the period the stress is applied.

This report contains the detailed formulation and derivation of the formulae for the evaluation of the contribution due to SIG.

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CONTENTS

1. INTRODUCTION

Irradiation creep is a major cause of dimensional instability in reactor structural materials. It has been investigated by several authors and a number of mechanisms have been proposed. A summary of the most favoured ones was presented and discussed recently by Bullough and Wood[1]. Among these is the Stress-Induced Preferred Adsorption (SIPA) mechanism, originally proposed by Heald and Speight[2] and Wolfer and Ashkin[3].

Direct experimental evidence of SIPA has been obtained recently by Garner, Wolfer and Brager[4] through observations of the effect of externally applied stresses on radiation-induced evolution of dislocation microstructure. They studied the effects of stress on the concurrent development of voids, Frank interstitial loops and the dislocation network in 20% cold-worked M316 stainless steels, by transmission electron microscopy. They found that the loop density that developed on each of the various (111) planes showed a clear and direct dependence on the resolved normal stress component on each plane. At a hoop stress of 327 MPa, the loop number density and the corresponding loop line density of the aligned loops were observed to be up to eight times higher than the same parameters in non-aligned loops. This strong anisotropy of the dislocation structure was initially attributed to the stress-induced preferred nucleation of loops[5]. However, it was later found[4] that SIPA provides a more satisfactory explanation.

A second contribution to dimensional change in reactor components is that arising from irradiation growth. This occurs under zero stress and, in contrast to the stress-induced dislocation anisotropy introduced by SIPA, discussed above, is a consequence of an intrinsic dislocation anisotropy in the material. For example, in a crystallographically anisotropic material, such as zirconium, dislocation Burgers vectors lie predominantly along the three <11$\bar{2}$0> directions and rarely have any c-component[6]. In this case, irradiation (stress-free) causes growth in the form of elongation along the a-direction and contraction along the c-direction[7]. Similar behaviour has been observed in crystallographically isotropic stainless steel when an anisotropic dislocation structure is introduced prior to irradiation by cold-working produced by rolling. Thus, whereas no irradiation growth occurs in well-annealed (isotropic) material, the cold-worked material elongates in the rolling direction[8]. This type of deformation can be explained as follows. Interstitials have a larger size effect (independent of SIPA), and hence, have a stronger interaction with edge dislocations than vacancies. In the presence of neutral sinks, such as voids or grain boundaries, irradiation therefore produces a net flux of interstitials to the edge dislocations, causing them to climb. Thus, atoms are deposited on crystal planes normal to the Burgers vectors of the edge dislocations. The rate at which atoms deposit on any given crystal plane is directly proportional to the line density of edge dislocations with Burgers vectors normal to that plane. Consequently, an anisotropic dislocation structure causes different rates of deposition of atoms on different planes, thus producing a time-dependent deviatoric strain.

It is evident that the primary requirement for the operation of the irradiation growth mechanisms is the existence of an anisotropic dislocation structure, together with suitable neutral sinks. The origin of the dislocation anisotropy can be varied; we have presented several examples above. In particular, we have pointed out that not only can such anisotropy be initially present in pre-irradiated material (intrinsic anisotropy), but it can be induced by the combined action of irradiation and stress through SIPA (extrinsic anisotropy). Such a dynamic extrinsic anisotropy allows some hybridization of the SIPA- and irradiation-growth mechanisms. Essentially, SIPA, through its very nature, provides a continuous mechanism for maintaining an anisotropic dislocation structure, which lends itself (in the presence of neutral sinks) to irradiation growth, thus enhancing the elongation rate. This is SIPA-induced growth: SIG. Both SIPA and SIG produce elongation under a tensile stress by causing atoms to be deposited, by dislocation climb, on planes normal to the stress axis.

In this report, we shall derive analytically, using a rate-theory approach, a general expression for the total deformation rate due to the climb of edge dislocations in a system with neutral sinks and identify the various components of the resulting strain rate. From this, we shall extract the contribution due to SIG, compare it with the contribution due to SIPA climb alone and show that the former can be more than double the latter. Based on the calculated results, we shall discuss the possibility of swelling-enhanced creep.

2. FORMULATION

We consider a single crystal subject to a radiation field producing a displacement rate of K dpa/s and acted upon by a uniaxial stress $\underset{\sim}{\sigma}$. To avoid the complication of grain anisotropy effects[7] we assume that the single crystal is spherical and has a grain diameter d. The entire population of edge dislocations is supposed to have a line density ρ and is made up of discrete loops generated by irradiation, plus a continuous network due to pre-irradiation deformation. This dislocation population is divided into N classes, each with a unique Burgers vector $\underset{\sim}{b}_n$, making an angle ϕ_n with $\hat{\sigma}$ (the stress direction) and a unique line direction, making an angle λ_n with $\hat{\sigma}$. We denote the line density of the nth class by ρ_n.

To avoid unnecessary complications, we neglect, in this report, (1) the effects of impurity; (2) any thermal emission from the sinks; and (3) the continuous loss of vacancies through the collapse of vacancy-rich cascade cores.

We consider only the plastic deformation of the single crystal caused by the climb cf the edge dislocation (loops plus network). A rate-theory approach is used. Within this approach, we consider the generation of point defects in an effec~ive medium, representing the irradiated crystal, and the loss of these defects to the various sinks, giving rise to the deformation of the crystal. We distinguish between edge dislocation sinks (ED's) and non-edge-dislocation sinks (NED's). There are N classes of ED's, each class having a different bias factor, which is a function of the orientation of its Burgers vector and line direction. The collection of the NED's generally consists of grain boundaries, screw dislocations, cavities and impurity traps. For a complicated sink situation such as this, one has to choose between proceeding numerically, with some loss in the transparancy of the physics involved, or, analytically, accepting the penalty of algebraic profusion. By separating the effects of the NED's and ED's, we can reduce the algebra to a manageable level. The analytic approach is therefore adopted.

3. DISLOCATION CLIMB VELOCITY

Let us first derive an expression for dislocation climb velocity. Suppose k_j^2 and k_{Nj}^2 denote the total sink strengths due to the ED's and NED's, respectively, for the interstitials (j = i) and vacancies (j = v). Solving the pair of steady-state rate equations for vacancies and interstitials, respectively, we obtain, in the absence of the NED's

$$D_j C_j = \frac{aK}{k_j^2} \tag{1}$$

and in the presence of the NED's

$$D_jC_j' = \frac{a'K}{k_j^2(1+R_j)} \tag{2}$$

with

$$R_j = \frac{k_{Nj}^2}{k_j^2} \tag{3}$$

In equations (1) and (2)

$$a = 2\eta^{-1}[(1+\eta)^{1/2} - 1] \tag{4}$$

where

$$\eta = 4\alpha K/D_iD_vk_i^2k_v^2 \tag{5}$$

In equation (5), α is the recombination coefficient, D_j and C_j, respectively, are the diffusion coefficients and atomic concentration for interstitials ($j = i$) and vacancies ($j = v$). The quantity a is the fraction of point defects that survives the bulk recombination in the absence of the NED's, a' is the equivalent quantity in the presence of the NED's, with k_i^2 and k_v^2 replaced by $k_i^2 + k_{Ni}^2$ and $k_v^2 + k_{Nv}^2$, respectively. The change in climb velocity of the nth class of ED's due to the presence of the NED's is then given by

$$\frac{a}{a'} V_n^{c'} - V_n^c = \frac{1}{b}[Z_i^n D_iC_i(\frac{1}{1+R_i} - 1) - Z_v^n D_vC_v(\frac{1}{1+R_v} - 1)] \tag{6}$$

where $V_n^{c'}$ and V_n^c are the climb velocities of the nth class of dislocations in the presence and absence of the NED's, respectively. After some manipulation, we obtain

$$V_n^{c'} = \frac{1}{b\rho}\dot{S} + \frac{a'}{a}\frac{k_i^2}{k_i^2+k_{Ni}^2} V_n^c \tag{7}$$

Here $\dot{S}$ is the rate at which vacancies are annihilated at the NED's.

$$\dot{S} = k_{Nv}^{2} D_v C_v' - k_{Ni}^{2} D_i C_i' \tag{8}$$

Equation (7) states that the dislocation climb velocity has two components: The first is the climb velocity due to the net interstitial flux to the ED's (= net vacancy flux to NED's) caused by the larger size-effect of the interstitials. This component exists only in the presence of the NED's. The second component is the additional climb velocity resulting from mechanisms, such as SIPA, which causes an uneven absorption of point defects among different classes of ED's. This component exists, independent of whether the NED's are present or not, and is different for different classes.

We note that equation (7) has been derived without reference to any specific mechanism that produces the point-defect segregation which causes dislocation climb.

We now consider the calculation of V_n^c, i.e., the dislocation climb velocity in the absence of the NED's. Suppose we denote the stress-free bias factor of the nth class of edge dislocation by $Z°$ (the subscript i or v denoting the kind of defect is dropped to simplify the notation). In the presence of a stress, the bias factor Z^n has been calculated by Woo[9] [Eqs. (6) and (7)]. It is given by

$$Z^n = Z°[1 - \frac{Z°\sigma}{\mu e°} (Q + \nu\Delta Q \cos^2\lambda_n + \Delta Q \cos^2\phi_n)] \tag{9}$$

where μ is the shear modulus of the matrix, $e°$ is the relaxation strain of the point defect in the absence of an applied field, Q and ΔQ are given by

$$Q = \frac{(1-2\nu)\Delta\kappa}{6(1-\nu) + 2(1+\nu)\Delta\kappa} - \frac{5(1+\nu)\Delta\mu}{30(1-\nu) + 4(4-5\nu)\Delta\mu} \tag{10}$$

and

$$\Delta Q = \frac{15\Delta\mu}{30(1-\nu) + 4(4-5\nu)\Delta\mu} \quad (11)$$

where

ν = Poisson ratio

$\mu(1+\Delta\mu)$ = shear modulus of the point defect

κ = bulk modulus

$\kappa(1+\Delta\kappa)$ = bulk modulus of the point defect

Equation (9) expresses the bias factor of the nth class as an explicit function of the line orientation λ_n, and the Burgers vector orientation ϕ_n of the class with respect to the stress direction. The differential of the bias factors between the ED's in different classes causes point-defect segregation and hence deformation in general. In particular, the differential due to different ϕ_k gives rise to the SIPA mechanism of irradiation creep[9].

Both the line orientation differential and the Burgers vector orientation differential have the same origin, namely the effect of the applied stress on the inhomogeniety interaction between the point defect and the edge dislocation[10]. Both therefore vanish in the absence of an applied stress.

In the case of a pure edge dislocation loop, it can be shown that equation (9) reduces to (Appendix A)

$$Z^n = Z^\circ[1 - \frac{Z^\circ\sigma}{\mu e^\circ} (Q' + (1 - \frac{\nu}{2})\Delta Q\cos^2\phi_n)] \quad (12)$$

where

$$Q' = Q + \frac{\nu}{2} \Delta Q \quad (13)$$

when the line direction is averaged over the loop. Equation (12) shows that in the case of loops, the effect of the line direction is to reduce the SIPA by about 15%. A similar reduction was reported[9] when both the line direction and the Burgers vector a ·andomly distributed. To simplify the analysis in the following, we shall use equation (12) instead of (9).

The total sink strength in the absence of the NED's is given by

$$k^2 = \Sigma_n Z^n \rho_n \tag{14}$$

Solving the rate equations, we obtain

$$Z^n DC = Z^n \frac{aK}{k^2} \tag{15}$$

The climb velocity of the nth class of dislocations can be written as

$$V_n^c = b^{-1}(Z_i^n D_i C_i - Z_v^n D_v C_v) \tag{16}$$

Using equations (14) and (15), substituting for Z^n through equation (12) and assuming that the interstitial is soft in shear and the vacancy does not change the shear modulus of the matrix, i.e., $\Delta m_i = -1$, $\Delta m = 0$, we obtain

$$V_n^c = \frac{aK}{b\rho} \frac{Z_i \sigma}{\mu e_i^{\circ}} \frac{45}{32} (1-\frac{\nu}{2})(\cos^2\phi_n - \Sigma_K f_k \cos^2\phi_k) \tag{17}$$

Equation (17) is a generalization of the climb velocity due to SIPA alone, derived by Woo[9], to include the line direction effect. In this equation, f_k is equal to the ratio ρ_k/ρ.

4. STRAIN CONTRIBUTION DUE TO DISLOCATION CLIMB

We now consider plastic deformation of the single crystal caused by climb of the edge dislocations with a velocity as expressed in equation (7). The deviatoric strain rate in the stress direction due to dislocation climb is given by

$$'\dot{\varepsilon}_c = \frac{2}{3} \Sigma_n \rho_n V_n^{c'} \, b(\cos^2\phi_n - \frac{1}{2}\sin^2\phi_n) \quad (18)$$

Using equations (7) and (17)

$$'\dot{\varepsilon}_c = \dot{S}(\Sigma_n f_n \cos^2\phi_n - \frac{1}{3}) + \frac{45}{32}\frac{\sigma}{\mu}\frac{a'KZ_i q}{e_i^{\circ}}(1-\frac{\nu}{2})\,\Sigma_n f_n \cos^2\phi_n$$

$$(\cos^2\phi_n - \Sigma_k f_k \cos^2\phi_k) \quad (19)$$

where $q = [k_i^2/(k_i^2 + k_{Ni}^2)$ is the fractional reduction in vacancy-interstitial segregation among the "aligned" and "non-aligned" dislocations as a result of a net point-defect flux into the NED's.

The corresponding non-deviatoric strain rate is given by

$$\dot{\varepsilon}_c^{iso} = \dot{S} + \frac{a'}{a} q \, \Sigma_n \rho f_n V_n^c \quad (20)$$

As the last term of equation (20) vanishes, $\dot{\varepsilon}_c^{iso}$ is given by $\dot{S}$.

It is interesting to note that the climb component of the deviatoric strain rate has two contributions, corresponding, respectively, to the two terms on the right side of equation (19). The first term originates from the dislocation climb due to vacancy-interstitial segregation as a result of the larger size effect of the interstitials. This term can be non-zero even in the absence of an applied stress.

Moreover, it vanishes for a material in which the Burgers vectors are randomly oriented, so that $\Sigma_n f_n \cos^2\phi_n = 1/3$. This term therefore arises from the anisotropy of the dislocation structure and contributes to irradiation growth[7]. Now, the anisotropy of the dislocation structure may arise from three sources:

(1) inherent dislocation anisotropy due to crystallographic factors;

(2) induced anisotropic strain due to pre-irradiation deformation; and

(3) SIPA induced during irradiation.

The last effect arises because SIPA allows aligned interstitial loops to grow faster than the non-aligned ones. Direct experimental observation of SIPA-induced anisotropy in the dislocation structure in M316 steels has been made by Garner et al.[4], as mentioned in the introduction. Thus, in general, f_n is a function of the creep stress so that this term also contributes to irradiation creep. The effect of stress on this term may be described as SIPA-induced growth. We shall consider this in more detail in the next section.

The second term of equation (19) originates from the dislocation climb due to SIPA, modified to take into account the line direction and the presence of the NED's. This term vanishes in the absence of a stress and therefore is a true creep component, which can be examined further, as follows.

Firstly, the presence of the NED's generally reduces the magnitude of this component (by the factor q) by reducing the total number of point defects migrating to the dislocations. In materials where void swelling occurs, k_{Ni}^2 is typically $\sim$ 0.2 k_i^2, so that q is about 0.8. This represents a reduction of $\sim$ 20% in the SIPA contribution to the total creep. This result corroborates that of Bullough and Hayns[11] who claimed that void swelling causes a reduction in irradiation creep due to SIPA. However, as we shall see in the next section, this reduction is more than compensated by the swelling enhancement of the creep contribution due to SIG arising from the first term in equation (19). Of course, the total effect of swelling on irradiation creep depends on the balance between these two terms. The quantitative aspects of this will be considered in more detail as we proceed.

Secondly, we note that the factor $(1-\nu/2)$ due to the line-direction effect reduces the second term by $\sim$ 15%. However, this can be more than compensated by the effect of the skewness of the dislocation structure introduced by SIPA, which causes f_n to deviate from that typical of a random distribution of Burgers vector orientations. In fact, it has been shown previously by Woo[9] that an increase of 60% to 80% in the SIPA creep rate would result from a 20% to 40% skewness in the dislocation structure. These estimations may be useful in discussing the various effects on the creep contribution due to SIPA climb.

5. SIPA-INDUCED GROWTH

We now consider in detail the strain contribution due to the first terms of equation (19). We first calculate the anisotropy of the dislocation structure as expressed by the fraction of dislocation line, f_n, in various classes.

The rate of change of the line density of loops of class n, $\dot{\rho}_n$, is proportional to the climb velocity of the class $V_n^{c'}$ in equation (7). We assume that such a change is effected through a change in the average radius of the class. If we denote the number of loops in the nth class by N_L^n, then, using equations (7) and (17), $\dot{\rho}_n$ is given by

$$\dot{\rho}_n = \frac{2\pi N_L^{\ n}}{b\rho} \{\frac{45}{32} (1-\frac{\nu}{2}) \frac{a'K\sigma Z_i q}{e_i{}^{\circ}\mu} [\cos^2\phi_n - \Sigma_k f_k \cos^2\phi_k] + \dot{S}\} \qquad (21)$$

If we denote the total number of loops by N_L, then from equation (21), we obtain the rate of change of the difference in loop line densities between the classes n and m,

$$\frac{d}{dt}(\rho_n - \rho_m) = \frac{2\pi N_L a' Kq}{b\rho} \frac{45}{32} (1-\frac{\nu}{2}) \frac{\sigma Z_i}{e_i{}^{\circ}\mu} (g_n \cos^2\phi_n - g_m \cos^2\phi_m) \qquad (22)$$

where

$$g_n = N_L{}^n / N_L$$

Equation (22) expresses the difference in loop line densities in the various classes, as a function of their orientation.

From equation (22), $\rho_n - \rho_m$ can be calculated and from the result, we can obtain an expression for f_n:

$$f_n = \frac{1}{L} \frac{\rho_L}{\rho} + \frac{\rho_N{}^n}{\rho} + \frac{2\pi N_L}{b} (\frac{Z_i}{\rho} \int \frac{dt}{k_i{}^2 + k_{Ni}{}^2}) V_n{}^{\circ} \qquad (23)$$

where L is the total number of classes of loops, ρ_L is the total dislocation loop line density, $\rho_N{}^n$ is the network dislocation density of the nth class and $V_n{}^{\circ}$ is given by

$$V_n{}^{\circ} = \frac{45}{32} (1-\frac{\nu}{2}) \frac{\sigma Z_i a' K}{e_i{}^{\circ}\mu} (\cos^2\phi_n - \frac{1}{L} \Sigma_m \cos^2\phi_m) \qquad (24)$$

In deriving equation (23), we have neglected the effects of stress-induced preferred nucleation and assumed $N_L^n = N_L/L$, i.e., the number of loops is the same in all classes.

Equation (23) expresses the fraction of dislocation line due to the nth class as a sum of three terms. The first term is the contribution from loops belonging to the nth class in the absence of SIPA. This term accounts for the inherent anisotropy arising from crystal structure. The second term is the contribution from the network dislocations of the nth class. This term can be zero for the nth class and depends on pre-irradiation deformation. The third term vanishes when the stress is zero and is the contribution from SIPA-induced anisotropic loop growth.

Substituting equation (23) into equation (19) gives the total deviatoric strain rate due to dislocation climb as

$$'\dot{\varepsilon}_c = \dot{S}(\Sigma_n f_n^\circ \cos^2\phi_n - \frac{1}{3}) + \dot{\varepsilon}_{SIG} + \dot{\varepsilon}_{SIPA} \tag{25}$$

where f_n° is the fraction of dislocation lines in the class n in the absence of stress. It is equal to the sum of the first two terms on the right side of equation (23). The first term, therefore, contributes exclusively to irradiation growth. It originates from the anisotropy in the dislocation structure resulting from the crystal structure and pre-irradiation deformation. Experimental evidence of the former is abundant in the growth of zirconium(7), where dislocation loops are found predominantly in the three <11$\bar{2}$0> directions. Experimental evidence of the latter is also found in neutron irradiation of stainless steel and by neutron, fission fragment or proton irradiation of bcc and fcc materials, including nickel(8).

The second term is the creep rate due to SIG. It is given by

$$\dot{\varepsilon}_{SIG} = \frac{2\pi N_L}{b} \left(\frac{Z_i \dot{S}}{\rho} \int \frac{dt}{k_i^2 + k_{Ni}^2}\right) \dot{\varepsilon}^\circ_{SIPA} \tag{26}$$

where $\dot{\varepsilon}^\circ_{SIPA}$ is the creep rate due to SIPA climb, neglecting any anisotropy of the dislocation structure due to SIPA or pre-irradiation deformation and neglecting the factor q due to the presence of the NED's. $\dot{\varepsilon}^\circ_{SIPA}$ is given by

$$\dot{\varepsilon}^{\circ}_{SIPA} = \frac{45}{32}\frac{\sigma Z_i a' K}{e_i{}^{\circ}\mu}\left[\Sigma_n \cos^4\phi_n - \frac{1}{L}\Sigma_m \cos^2\phi_m\right] \quad (27)$$

We now calculate the factor in front of $\dot{\varepsilon}^{\circ}_{SIPA}$ in equation (26). To avoid complexity, we consider only the case $k_i^2 >> k_{Ni}^2$ such as in cold-worked materials or in cases where a considerable number of loops have already developed at the time the stress is applied. Then

$$\frac{Z\dot{S}}{\rho}\int\frac{dt}{k_i^2+k_{Ni}^2} \approx \frac{\dot{S}}{\rho}\int\frac{dt}{\rho(t)} \quad (28)$$

6. SIG IN THE PRESENCE OF VOID SWELLING

We first consider the case when swelling occurs so that the dominating NED's are the voids. Let us donate the void concentration by C_c and the average void radius by r_c. Then the swelling rate is equal to the rate at which a net vacancy flux arrives at the voids and we have the following equations:

$$aK\rho\ (Z_i-Z_v)\ \frac{4\pi C_c r_c}{(Z\rho+4\pi C_c r_c)^2} = \dot{S} = 4\pi C_c r_c^2 \dot{r}_c \quad (29)$$

where Z is the dislocation bias factor, neglecting whether it is for interstitials or vacancies. Assuming that $Z\rho >> 4\pi C_c r_c$, we obtain

$$r_c \dot{r}_c = \frac{aK(Z_i-Z_v)}{Z_i^2\rho} \quad (30)$$

From equation (30), we can calculate the right hand side of equation (28), and the result is

$$\frac{\dot{S}}{\rho}\int\frac{dt}{\rho(t)} = \frac{3}{2}(S-S_o) \quad (31)$$

where S_o is the swelling at the time the stress is applied. Now the rate of change of the total loop line density is related to the swelling rate through the rate of increase of the average loop radius:

$$\dot{\rho}_L = 2\pi N_L \dot{r}_L = 2\pi N_L \frac{\dot{S}}{b\rho} \frac{\rho_L}{\rho}$$

Remembering $\rho = \rho_L + \rho_N$ and assuming that ρ_N is not a function of time (specimen not too heavily cold-worked), we can integrate equation (29) and obtain

$$\frac{4\pi N_L}{b} (S-S_o) = (\rho^2-\rho_o^2) + 2\rho_N(\rho-\rho_o) + 2\rho_N^2 \ln \left(\frac{\rho_L}{\rho_{Lo}}\right) \qquad (33)$$

where ρ_o is the total dislocation density at $t = 0$, at which time the stress is applied, and ρ_{Lo} is the corresponding loop dislocation line density. The terms ρ_o and ρ_N are related by

$$\rho_o = \rho_{Lo} + \rho_N \qquad (34)$$

Substituting into equations (31) and (26), we obtain the strain rate due to SIG

$$\dot{\varepsilon}_{SIG}(\text{swelling}) = \frac{3}{4} \left\{ \left(1-\frac{\rho_o^2}{\rho^2}\right) + \frac{2\rho_N}{\rho}\left(1 - \frac{\rho_N}{\rho}\right) + 2\left(\frac{\rho_N}{\rho}\right)^2 \ln\left(\frac{\rho_L}{\rho_{Lo}}\right) \right\} \dot{\varepsilon}^o_{SIPA} \qquad (35)$$

7. SIG IN THE ABSENCE OF SWELLING

We will now consider the case without swelling. In this case, the grain boundaries are supposed to be the dominating NED's.

The rate of change of the total loop line density is related to the net flux of vacancies to the grain boundaries. For this case, equation (32) also applies and $\dot{S}$ is given by:

$$\dot{S} = \frac{aK\, d\sqrt{p}\ (\sqrt{Z_i} - \sqrt{Z_v})}{6(1 + 1/6\ d\sqrt{\rho})^2} \tag{36}$$

From equation (32) we can calculate the right side of equation (28) using equation (36). The result is

$$\frac{2\pi N_L}{b}\frac{\dot{S}}{\rho}\int\frac{dt}{\rho(t)} = \frac{2}{3}\left[1 - \left(\frac{\rho_o}{\rho}\right)^{3/2}\right] + 2\,\frac{\rho_N}{\rho}\left[1 - \left(\frac{\rho_o}{\rho}\right)^{1/2}\right] + \left(\frac{\rho_N}{\rho}\right)^{3/2} \ln\left(\frac{\rho_L}{\rho_{L^o}}\right) \tag{37}$$

where, we have neglected a term proportional to ln $[(\rho_o^{1/2} + \rho_N^{1/2})/(\rho^{1/2} + \rho_N^{1/2})]$.

Substituting into equation (26), we obtain the strain rate due to SIG

$$\dot{\varepsilon}_{SIG}(\text{no swelling}) = \left\{\frac{2}{3}\left[1 - \left(\frac{\rho_o}{\rho}\right)^{3/2}\right] + 2\frac{\rho_N}{\rho}\left[1 - \left(\frac{\rho_o}{\rho}\right)^{1/2}\right] + \left(\frac{\rho_N}{\rho}\right)^{3/2} \ln\left(\frac{\rho_L}{\rho_L^o}\right)\right\} \dot{\varepsilon}^{\circ}_{SIPA} \tag{38}$$

To obtain an estimate of the magnitude of $\dot{\varepsilon}_{SIG}$, compared to that of $\dot{\varepsilon}^{\circ}_{SIPA}$, which is approximately equal to the direct contribution due to SIPA climb, we plot the ratio $\dot{\varepsilon}_{SIG}/\dot{\varepsilon}^{\circ}_{SIPA}$ according to equations (36) and (38) in Figure 1, for different values of ρ_N/ρ and $\rho_L/\rho_{Lo}(= r_L/r_{Lo})$. At this point, we should be aware of the fact that ρ_L/ρ_{Lo} and ρ_N/ρ are not independent of each other. For large values of

ρ_N/ρ little loop growth has occurred and small values ($\sim$ 1) of ρ_L/ρ_{Lo} are appropriate. The correct values to be used should be obtained from experimental measurements or a full rate-theory calculation. Nevertheless, for our present purpose, the information provided by Figure 1 suffices.

From Figure 1, the following general conclusions can be drawn. Firstly, unless ρ_L/ρ_{Lo} approaches 1 (little or no loop growth since the application of the stress), the creep contribution due to SIG is very significant compared to that due to SIPA climb. Since one component does not occur without the other, this means that SIPA-induced growth may increase the total contribution due to SIPA by doubling or even tripling the direct contribution due to the climb component. Therefore, in cases where SIPA climb constitutes a major creep component, the effect of SIG creep should also be taken into account.

Secondly, for the same dislocation structure, the SIG creep rate is higher in the presence of swelling than in its absence. This shows the possibility of some enhancement of radiation creep due to swelling. However, the enhancement is small, only about 10 to 20%, and is balanced to some extent by a reduction of the creep contribution due to SIPA climb resulting from an increase in the NED sink strength (k_{Ni}^2) in the factor $k_i^2/(k_i^2 + k_{Ni}^2)$ when void swelling occurs. The present result that swelling may not enhance creep very significantly contradicts those of Boltax, et al.[12]. They concluded that swelling enhances creep significantly. However, their results for the non-swelling case were derived neglecting the net flow of vacancies to the grain boundaries. If the same assumption is made here, the right side of equation (38) would vanish, resulting in a large swelling enhancement of the creep. The magnitude of this enhancement is about 100%, which agrees with the results given by Boltax et al.[12]. We feel that the effectiveness of the grain boundary as a sink for the supersaturated vacancies should be taken into account. Therefore, we believe that their calculated creep enhancement may be very much overestimated.

Although creep due to SIPA climb and SIG occur simultaneously and are directly proportional to each other, there are differences between the two components, which makes them distinguishable. One of the differences is that SIG creep retains a memory of the stress, while SIPA climb does not. Thus, if a stress has been applied for enough time for the anisotropic dislocation structure to develop, the material should continue to elongate (although at a reduced rate) in the direction of the applied stress, even after the stress has been removed. This elongation is due to irradiation growth as a result of the SIPA-induced dislocation anisotropy. Mathematically, this history dependence of the SIG creep arises through the integral in equation (26). In this respect, the behaviour of SIG is similar to the stress-induced preferred nucleation (SIPN) for irradiation creep, first introduced by Hesketh[13]. However, unlike SIPN, the anisotropy of the dislocation structure in SIG is not necessarily induced during nucleation. It can be induced whenever a stress is applied for a prolonged period. Since the creep components due to SIPA climb and to SIG are concomitant, experiments to investigate such predicted behaviour may serve to test whether, and under what circumstances, SIPA is the dominating irradiation creep mechanism in a particular material.

Another difference between the SIG and SIPA climb-creep contributions is the response time. For the latter, the response to the application of the stress is immediate. For the former, enough time must be allowed before the induced anisotropy of the dislocation structure can be built up sufficiently for significant growth to occur. This implies that the SIG creep rate increases with time after the stress has been applied.

Thus, SIG will generate a significant contribution to irradiation creep, but only under conditions where the growth of loops after application of the stress has already led to $\rho_L/\rho_{Lo} > 2$. Its contribution will increase as ρ_L/ρ_{Lo} increases.

8. SUMMARY AND CONCLUSIONS

SIPA causes the growth rates of dislocation loops generated by irradiation to vary according to the alignment of their Burgers vectors with respect to the applied stress. A prolonged period under applied stress creates an anisotropic dislocation structure in which the majority of dislocations have their Burgers vectors in alignment with the stress. As a consequence, more interstitials are deposited at these aligned dislocations (during irradiation) as the edge dislocations climb in the presence of non-edge dislocation sinks (i.e., sinks other than edge dislocations). In turn, this develops a time-dependent deviatoric strain which contributes to irradiation creep. This is the mechanism we have called SIPA-induced growth or SIG. We have developed an expression for its contribution to irradiation creep, and we have shown that it is very significant in comparison to SIPA, except when little or no loop growth has occurred in response to the applied stress. Furthermore, we have shown that enhancement of creep by swelling may occur, but that it is not as large as predicted by Boltax et al.[12]. This discrepancy is attributed to the fact that grain boundary effects were ignored in their model.

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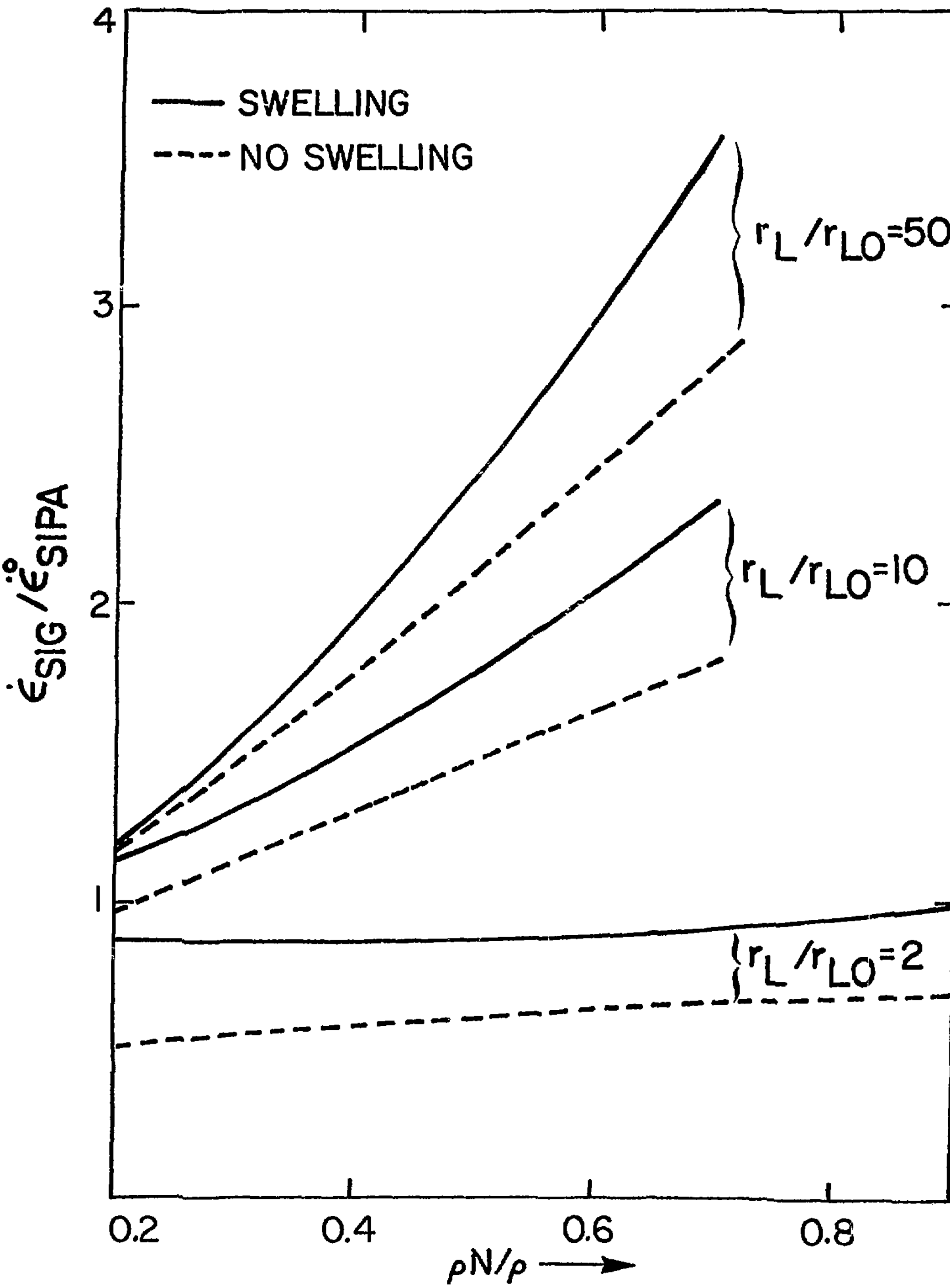

FIGURE 1: The Ratios $\dot{\varepsilon}_{SIG}/\dot{\varepsilon}_{SIPA}$ in the Presence of Swelling (———) and in the Absence of Swelling (-----) Plotted Against ρ_L/ρ_{Lo} for Different Values of ρ_L/ρ_{Lo} (= r_L/r_{Lo}).

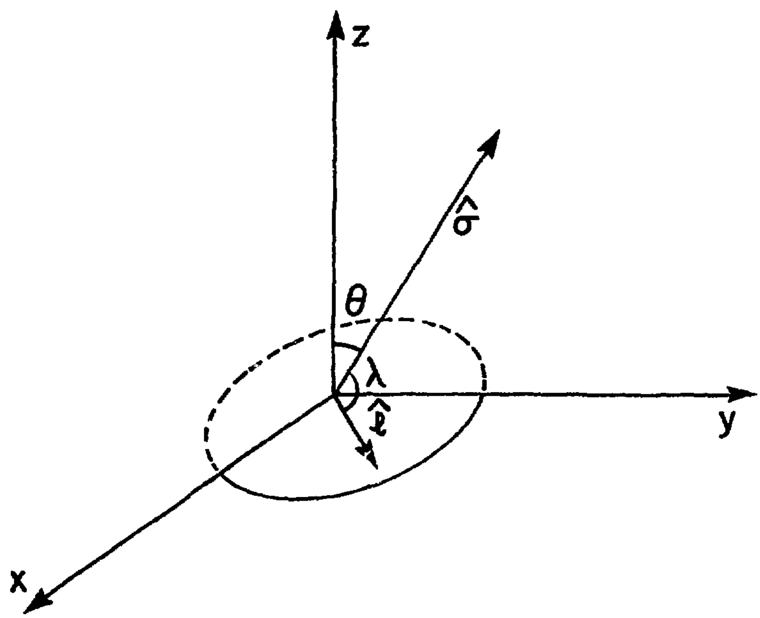

FIGURE 2: The Co-ordinate Systems Used in the Calculation of the Average Bias Factor of a Dislocation Loop.

APPENDIX A

BIAS FACTOR OF A PURE EDGE-DISLOCATION LOOP

The co-ordinate system is defined in Figure 2. The loop-plane normal and the Burgers vector lie on the z-axis. The stress axis σ lies on the y-z plane making an angle θ with the z-axis. The line direction of an element along the loop is denoted by the unit vector $\hat{\ell}$ which makes an angle χ with the x-axis. In this system, we may write

$$\hat{\ell} = (\cos\chi,\ \sin\chi,\ 0) \tag{A1}$$

and

$$\hat{\sigma} = (0,\ \sin\theta,\ \cos\theta) \tag{A2}$$

For the element $\hat{\ell}$ under consideration, we have

$$\cos\chi = \sin\theta\sin\chi \tag{A3}$$

Averaging over all directions of $\hat{\ell}$ as we sample over the entire loop

$$\overline{\cos^2\lambda} = \frac{1}{2\pi}\int_o^{2n} \sin^{-2}\theta\ \sin^{-2}\chi\ d\chi$$

$$= \frac{1}{2}\,(1 - \cos^2\theta) \tag{A4}$$

Substituting into equation (9), equation (12) is obtained.

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