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Instability of interstitial clusters under ion and electron irradiations in ceramic materials

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Abstract

One type of candidate materials for future fusion reactors is ceramics, which can be applied as radio frequency windows, toroidal insulating breaks and diagnostic probes. The degradation of physical and mechanical properties of these materials under neutron irradiation is determined by the kinetics of radiation defects including a point defect cluster formation (dislocation loops, voids and so on). The physical mechanisms of defect structure development in ceramic materials, where point defects and their clusters can have an effective charge, are completely different from those in metals. We have investigated the physical mechanisms of instability of extended interstitial defect clusters (charged dislocation loops), which were formed in stabilized cubic zirconia under electron irradiation with 100–1000 keV due to the selective displacement damage in oxygen sublattice. A new theoretical model is suggested for the explanation of the growth process and instability of the interstitial clusters. The suggested model takes into account an accumulation of effective charge on growing dislocation loops due to the trapping of electrons in dislocation cores. Our calculations show that the elastic stress and strain fields are much intense around charged dislocation loops than non-charged dislocation loops, due to an additional stress and strain fields driven by an electric field of accumulated charge. The stress induced by the charged dislocation loops with the density of trapped electrons per atom $n = 0.4$ is found to be comparable with the theoretical yield stress of zirconia, which explains the multiplication of dislocation network at a critical size of the defect clusters observed by experiments.
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1. Introduction

Ceramic materials are expected to be applied in the future fusion reactor as radio frequency windows, toroidal insulating breaks and diagnostic probes. The radiation resistance of these materials under neutron irradiation is determined by the kinetics of radiation-induced point defects accumulation in matrix and point defect cluster formation (dislocation loops, voids and so on). Under irradiation due to the ionization process, excitation of electronic subsystem and covalent type of interaction between atoms the point defects can have an effective charge of point defects. For the understanding of the nucleation-and-growth of point defect clusters in ceramic materials, it is essential to take into account the charge state effects into consideration.

Previous studies performed on stabilized cubic zirconia [1–3] have shown that this material is exceptionally radiation resistant, especially to amorphization and radiation swelling. The present authors have, however, found an anomalous formation of extended defect clusters under electron irradiation in energy range 100–1000 keV subsequent to ion irradiation, such as 100 keV He^{+} and 300 keV O^{+} ions, through transmission electron microscopy. It has been demonstrated that the defect clusters possess very strong stress and strain fields and grow up to a critical size (1.0–1.5 μm), and then

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became unstable resulting in the multiplication of dislocation density and the formation of dislocation network near the charged dislocation loops. The defect clusters were discussed to be oxygen platelets caused by the selective displacement damage of oxygen sublattice in yttria stabilized zirconia (YSZ) [4,5].

In the present paper, a new theoretical model is suggested for the explanation of the growth kinetics and instability of charged dislocation loops in ceramic material (cubic zirconia), taking into account an effective charge on dislocation loops through the trapping of electrons in dislocation cores. The strain and stress fields induced by internal electrical field near charged dislocation loops (oxygen platelets) are also presented. The theoretical calculations show that the induced stress around the charged dislocation loops can be comparable with the theoretical yield stress in YSZ. A critical condition for the beginning of plastic deformation near charged dislocation loop (multiplication of dislocation network) is determined, which can be used for the estimation of critical radius of charged dislocation loop to be unstable. The obtained theoretical results for the growth rate and critical radius of charged unstable dislocation loops are compared with the observed experimental data.

2. Experimental results

The experimental results presented here have been obtained on single crystals of stabilized cubic zirconia (YSZ), which contains 13-mol% yttria [4,5]. Fig. 1(a) shows a typical example of defect clusters formed under 200 keV electron irradiation at 470 K subsequent to 300 keV O^+ ion irradiation at 470 K, illustrating a large defect cluster with strong black/black lobes contrast. Similar defect clusters were formed under electron irradiation with 100–1000 keV in YSZ, which were originally irradiated with 300 keV O^+ ions at 470 K, 100 keV He^+ ions at 870 K and 4 keV Ar^+ ions at 300 K. The nucleation-and-growth process of the defect clusters is extremely different from that of neutral interstitial-type dislocation loops. Their characteristic features are summarized as follows [5]: (1) very large black/black lobes contrast, which indicates an existence of strong strain-field around the defect clusters, (2) very large size up to 1.0–1.5 μm in diameter with very rapid growth rate nearly 1.0 nm/s at an electron flux of 1.5×10^{23} e/m^2 s, (3) preferential formation at the periphery of the focused electron beam, and (4) the transformation from the black/black lobes contrast to dislocation network at a critical diameter of 1.0–1.5 μm (from Fig. 1(a) and (b)). Electron irradiation of the defect clusters after the transformation revealed the renucleation of the defect clusters near or at the generated dislocations. The processes of nucleation, growth and transformation are repeated under electron irradiation [1]. This is a completely new phenomenon related to a multiplication of dislocations in ceramic materials under irradiation. The defect clusters are discussed to be oxygen platelets having an accumulated electric charge [4,5]. For the explanation of this phenomenon the following theoretical model is suggested.

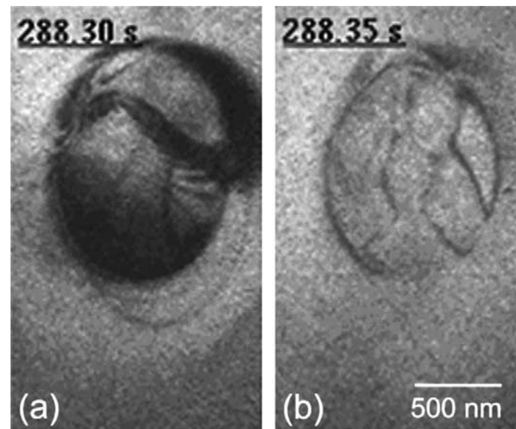

Fig. 1. The growth and multiplication of dislocations near charged dislocation loop at the critical radius $R = 600$ nm in yttrium stabilized cubic zirconia under 200 keV electron irradiation at 470 K. The samples were irradiated originally with 300 keV O^+ ions at 470 K to a fluence of 5.1×10^{17} O^+/m^2. Irradiation time from the upper-left micrograph is also shown at each micrograph.

3. Theoretical model

The microstructure change and point defect cluster formation in irradiated cubic zirconia are determined by the generation and kinetics of point defects (interstitials and vacancies). Due to the large mass difference between Zr and O atoms, the elastic displacement cross-section of O atoms σ_d^O is much larger than Zr atoms σ_d^{Zr} ($\sigma_d^O \gg \sigma_d^{Zr}$) under electron irradiation with 100–1000 keV. Therefore, radiation defects induced by fast electrons are considered to be mainly oxygen point defects (interstitials and vacancies) [4]. Under the irradiation during the production of displaced atoms, the interstitial oxygen atoms and vacancies in cubic zirconia can have an effective charge (O^{2-} or V^{2+} respectively) [6]. During the diffusion process, the charged interstitial atoms and vacancies can change the effective charge [7]. The changes of the effective charge of interstitial oxygen atoms and vacancies are determined by the kinetics of emission and absorption of electrons on these point defects. In this theoretical model, we assume that each charged interstitial oxygen atom loses electrons (charge) during diffusion process and migrates as a neutral interstitial

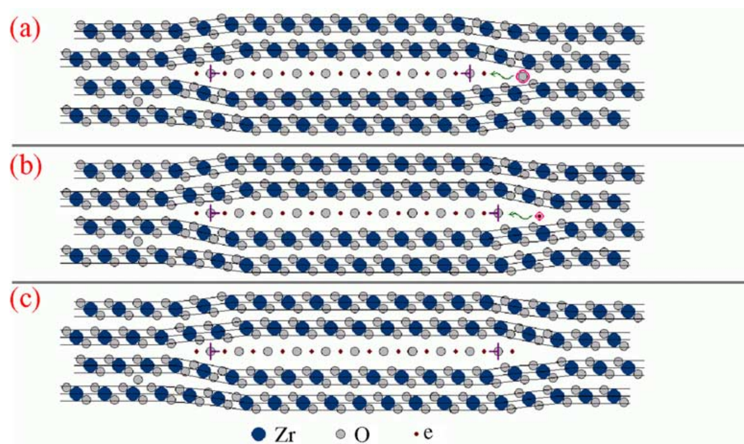

Fig. 2. The schematic model of charged dislocation loop formation in ceramic materials (ZrO_2) under electron irradiation due to absorption of neutral oxygen interstitial atoms and electrons. Free electrons (red points) produced under electron irradiation are trapped in dislocation core of dislocation loop (accumulation charge) and absorption of neutral oxygen interstitial atoms results in growth of dislocation loop.

atom. The energy migration barrier for neutral interstitial oxygen atoms is much less compared with that for charged oxygen vacancies and interstitial atoms [7]. It means that due to the high diffusivity of neutral interstitials, they can form interstitial oxygen clusters (dislocation loops) in the matrix. The growth kinetics of such interstitial dislocation loops is determined by the absorption of neutral interstitial oxygen atoms by dislocation loops. The emitted electrons from charged interstitial oxygen atoms can be consider as free and interstitial clusters, vacancies and dislocation lines can trap electrons. The free electrons in ceramic materials are also produced under electron irradiation due to elastic collisions between fast electrons and atomic electrons in the matrix when the kinetic energies higher than the binding energy (gap energy) are transferred. Previous experimental results in YSZ [8] confirms this point and demonstrate that under electron irradiation the strong strain contrast is formed along segments of edge dislocations, which is increased with the dose of irradiation. It suggests that free electrons produced under electron irradiation are trapped on dislocation lines. The trapping of electrons on dislocation lines should be more effective in dislocation core area, where the effective density of material is less and many vacant sites exist for the formation of additional electronic levels in the band gap. This process results in the accumulation of electrons and production an effective charge on dislocation lines. The formation of the strong strain contrast near dislocation lines is determined by the formation of an additional internal electrical field and polarization of dielectric material near charged dislocation line, which are increased with the accumulation of electrons in the dependence on the dose of irradiation. This electrical field produces in dielectric material additional strong stress and strain fields near charged dislocations.

In our theoretical model, the diffusion motion of two types of particles determines the process of dislocation loop growth and charge accumulation on dislocation loops: interstitial atoms (loop growth) and electrons (charge accumulation). Because the diffusivity of free electrons (D_e) in ceramic materials is higher compared with that of interstitial atoms (D_I), or $D_e \gg D_I$, the following stages can be considered (see Fig. 2):

(1) On the first stage the charged interstitial atoms during diffusion jumps lose electrons (charge) and form the nuclei of dislocation loops through the agglomeration of neutral interstitial atoms (see Fig. 2(a)).
(2) Free electrons produced under electron irradiation are trapped in dislocation cores of these dislocation loops. This process is determined by the diffusivity of free electrons (D_e) and it is very fast process (see Fig. 2(b) and (c)).
(3) The dislocation loops grow due to the absorption of neutral interstitial atoms. This process is determined by the diffusivity of neutral interstitial atoms (D_I) and it is very slow process ($D_e \gg D_I$). During this process the dislocation loop radius is increased and new places for the trapping of free electrons in dislocation core are created with a new size of dislocation loop. The accumulation of free electrons on dislocation loops results in the charge increase on these dislocation loops (see Fig. 2(a)).

Let us find the growth rate of interstitial dislocation loop. At the low temperatures ($T < 450$ K) only the interstitial atoms can move because the migration energy

of vacancies (E_V^m) is higher than interstitial migration energy (E_I^m) [7] and vacancies, which hardly migrate, can be considered only as recombination centers. In this case the defect clusters will grow only due to the migration and absorption of oxygen interstitial atoms on dislocation loops without taking into account on this process the vacancy diffusivity.

The growth rate of non-charged dislocation loops is determined by the agglomeration and absorption of neutral interstitial atoms and it is equal to

$$\frac{dR}{dt} = \frac{C_I D_I}{b}. \quad (1)$$

The changes of charged and non-charged interstitial (C_i, C_I) and vacancy (C_v, C_V) concentrations produced under irradiation in this case can be found from the following kinetic equations.

$$\begin{aligned} \frac{dC_i}{dt} &= G - \beta_i C_i = 0, \\ \frac{dC_I}{dt} &= \beta_i C_i - Z_I(\rho_d + 2\pi N_L R) D_I C_I \\ &\quad - \alpha D_I C_I C_V = 0, \\ \frac{dC_v}{dt} &= G - \beta_v C_v = 0, \\ \frac{dC_V}{dt} &= \beta_v C_v - \alpha_V D_I C_I C_V - \alpha_v D_I C_I C_v. \end{aligned} \quad (2)$$

Here G is the generation rate of charged point defects, produced under electron irradiation; β_I and β_v are the rates of the emission and absorption of electrons on charged interstitial atoms and vacancies respectively; ρ_d is the dislocation density; N_L is the density of dislocation loops; D_I is the diffusion coefficient of interstitial atoms; Z_I is the preference of absorption of interstitial atoms on dislocation lines; α_V and α_v are the recombination coefficients for neutral interstitial atoms on neutral and charged vacancies respectively ($\alpha_V = \frac{\alpha_{V0}}{b^2}$, $\alpha_v = \frac{\alpha_{v0}}{b^2}$, b is the lattice spacing).

The solution of system of kinetic equations (1) and (2) can be find very easy in two cases: for small and big size of dislocation loops.

(a) If the dislocation loop radius (R) is small ($R \ll R_c = \frac{\rho_d}{2\pi N_L}$) and the main sink of interstitial atoms is the dislocation lines, then the time dependence of dislocation loop radius is determined by the following relation:

$$R = \left(\frac{2Z_I}{\alpha_0 \rho_d}\right)^{1/2} (Gt)^{1/2}. \quad (3)$$

(b) For the large dislocation loops ($R \gg R_c = \frac{\rho_d}{2\pi N_L}$), when the interstitial atoms are absorbed mostly by dislocation loops the time dependence of dislocation loop radius is determined by the following relation:

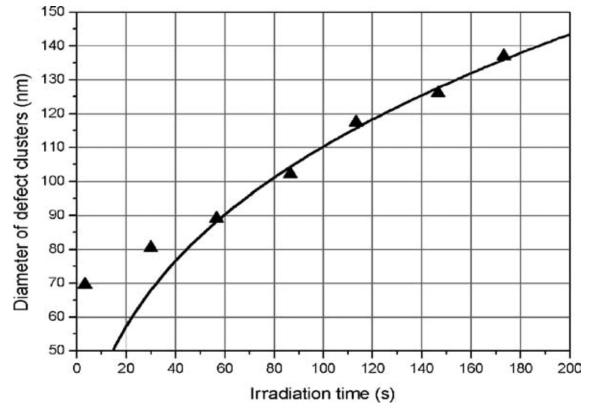

Fig. 3. Comparison of the experimental data (triangles) for diameter change of dislocation loop at 380 K as a function of irradiation time under 200 keV electron irradiation with a flux of 8×10^{21} e/m^2 s and theoretical calculations (solid line) obtained using the relation (4).

$$R = \left(\frac{3Z_I}{\pi\alpha_0 N_L}\right)^{1/3} (Gt)^{1/3}. \quad (4)$$

A comparison of the growth process of a dislocation loop at $T = 380$ K between experimental data and the theoretical result obtained by using Eq. (4) is presented in Fig. 3 as a function of irradiation time under 200 keV electron irradiation with a flux of 8×10^{21} e/m^2 s.

Fig. 3 comparison of the experimental data (triangles) for diameter change of dislocation loop at 380 K as a function of irradiation time under 200 keV electron irradiation with a flux of 8×10^{21} e/m^2 s and theoretical calculations (solid line) obtained using Eq. (4). Let us consider now the accumulation of effective charge on dislocation loops due to the trapping of free electrons by dislocation loops. The rate equations, which describe the diffusion of free electrons taking into account the effect of electrical driving force on diffusivity of free electrons near charged dislocation loop, can be written in the following form:

$$\frac{\partial C_e}{\partial t} + \mathrm{div}\vec{j}_e = 0, \quad \vec{j}_e = -D_e \nabla C_e - \frac{D_e}{kT} C_e q_e \nabla\varphi. \quad (5)$$

Here C_e, D_e, q_e are the concentration, diffusion coefficient and electrical charge of free electrons respectively ($q_e = e$); φ is the electrical potential near charged dislocation loops.

The change of electrical potential near charged defect clusters is determined by the following equation:

$$\Delta\varphi = -\frac{4\pi}{\varepsilon a^3}(q_i C_i + q_v C_v + q_h C_h + q_e C_e), \quad (6)$$

where C_h, q_h ($q_h = -e$) are the concentration and charge of holes in the matrix; q_i, q_v are the charges of

interstitials and vacancies respectively; ε is the dielectric permeability of ceramic material.

The role of boundary condition for Eq. (6) plays the condition of electrical neutrality – in the matrix, which takes place far from dislocation loop and this boundary condition has the following form:

$$q_i C_{i0} + q_v C_{v0} + q_h C_{h0} + q_e C_{e0} = 0. \quad (7)$$

Here C_{i0}, C_{v0}, C_{h0}, C_{e0} are the concentrations of charged interstitial atoms, vacancies, holes and electrons at positions far away from the charged dislocation loop. The concentration of vacancies and holes far from charged dislocation loops are considered to be constant in all volume of material. Follow assuming that the process of cluster growth is quazistationary process and taking into account Eqs. (5)–(7) we can obtain the following equation for the concentration of electrons:

$$D_e \Delta C_e + \frac{D_e}{kT} q_e \nabla C_e \nabla \varphi = \frac{D_e C_e}{kT} \frac{4\pi q_e}{\varepsilon a^3} \sum_{k=1,2} q_k (C_k - C_{k0}). \quad (8)$$

The right side in Eq. (8) is proportional to $\propto D_e C_e / r_0^2$, where r_0 is the screening radius of charged interstitial cluster for electrons. The screening radius is very large at the small concentration of point defects near charged dislocation loop. In this paper we will consider the case when the screening radius is very large ($r_0 \gg R$), where we can neglect the right side of Eq. (8). We can consider here also the situation that only one type of point defect (interstitial) can migrate in the matrix and the other type of point defects (vacancies) are not mobile.

Let us consider now the kinetics of defect cluster growth, which consists from trapped electrons and represents the conducting flattened out spheroid with the radius R, thickness $h = 2d$ and total charge Q. The following relation gives the electrical potential near such charged conducting spheroid in the cylindrical geometry:

$$\varphi = \frac{Q}{\varepsilon\sqrt{R^2 - d^2}} \operatorname{arctg} \sqrt{\frac{R^2 - d^2}{\xi + d^2}},$$
$$\xi = \frac{1}{2}\left[z^2 + \rho^2 - R^2 - d^2 + \sqrt{(R^2 - d^2 + \rho^2 + z^2)^2 - 4\rho^2(R^2 - d^2)}\right]. \quad (9)$$

The exact solution of Eq. (8) for the distribution of electrons near interstitial dislocation loop in this case is given by the following relation:

$$C_e = C_{e0} \frac{\exp(\lambda - q_e \varphi / kT) - 1}{\exp(\lambda) - 1},$$
$$\lambda = \frac{q_e Q}{kT\sqrt{R^2 - d^2}} \arccos \frac{d}{R}. \quad (10)$$

The total number of electrons trapped by the charged dislocation loop is determined by the total current of electrons on dislocation loop and it is expressed using the solution (10) by

$$\frac{dN_e}{dt} = 2\pi \frac{q_e Q C_{e0} D_e}{kT\omega} \frac{1}{\exp(\lambda) - 1}. \quad (11)$$

Let us consider the time dependence of the total charge accumulation on dislocation loop in two cases: small and big charge of dislocation loops.

(1) If the total charge Q of a dislocation loop is small ($Q \ll kTR/q_e$), then using the accumulation rate of the free electrons in the charged cluster (11) we can get the time dependence of the total charge of the dislocation loop for small and big dislocation loops.

(a) For the small dislocation loops ($R \ll R_c = \frac{\rho_d}{2\pi N_L}$), the time dependence of the total charge of the dislocation loop is expressed by using Eq. (3) in the following form:

$$Q \approx \frac{4\alpha_0 q \rho_d D_e C_e}{3\omega G Z_I} R^3 \approx \frac{4\alpha_0 q \rho_d D_e C_e}{3\omega G Z_I} \left(\frac{2Z_I}{\alpha_0 \rho_d}\right)^{3/2} (Gt)^{3/2}. \quad (12)$$

(b) For the big dislocation loops ($R \gg R_c = \frac{\rho_d}{2\pi N_L}$), the time dependence of the total charge of the dislocation loop can be found from the following relation by using Eq. (4):

$$Q \approx \frac{\pi\alpha_0 q N_L D_e C_e}{\omega G Z_I} R^4 \approx \frac{\pi\alpha_0 q N_L D_e C_e}{\omega G Z_I} \left(\frac{3Z_I}{\pi\alpha_0 N_L}\right)^{4/3} (Gt)^{4/3}. \quad (13)$$

(2) If the total charge Q of the charged spheroid is big ($Q \geqslant kTR/q_e$), then the accumulation rate of trapped electrons by the charged dislocation loop is exponentially small (see (11)) and the total charge Q of dislocation loop is determined in this case by the total number of trapped electrons N_e ($Q = q_e N_e$). Putting this relation in Eq. (11), we will get the equation for the total charge Q of interstitial cluster, which has the exact solution for small and big dislocation loops.

(a) This solution for the small dislocation loops ($R \ll R_c = \frac{\rho_d}{2\pi N_L}$) taking into account the relation (3) is equal to

$$Q \approx \frac{2kT_e}{\pi q} R \ln\left(\frac{\alpha_0 \rho_d}{G Z_I \tau_e} R^2\right) \approx \frac{2kT_e}{\pi q} \left(\frac{2Z_I}{\alpha_0 \rho_d}\right)^{1/2} (Gt)^{1/2} \ln\left(\frac{2t}{\tau_e}\right). \quad (14)$$

Here τ_e is the characteristic time of free electron relaxation ($\tau_e = \frac{\omega k T_e}{2\pi q^2 D_e C_e}$).

(b) The time dependence of the total charge Q of dislocation loop for the big dislocation loops ($R \gg R_c = \frac{\rho_d}{2\pi N_L}$) taking into account the relation (4) is equal to

$$Q \approx \frac{2kT_e}{\pi q} R \ln\left(\frac{\pi\alpha_0 N_L}{GZ_I\tau_e} R^3\right) \approx \frac{2kT_e}{\pi q}\left(\frac{3Z_I}{\pi\alpha_0 N_L}\right)^{1/3} (Gt)^{1/3} \ln\left(\frac{3t}{\tau_e}\right). \quad (15)$$

The charged dislocation loop due to the internal electric field $\vec{E}$ ($\vec{E} = -\nabla\varphi$) in ceramic (dielectric) materials produces near circular charged dislocation loop an additional stress field σ_{ik}^E ($\sigma^E \propto \varepsilon E^2$, see [4,9]). The shape and nature of total stress and strain fields near charged dislocation loop are completely different compared with the normal non-charged dislocation loop. The total stress field near charged dislocation loop (σ_{ik}^{tot}) is determined by the sum of normal elastic stress field (σ_{ik}^Y) and additional stress field induced by electrical field near charged dislocation loop (σ_{ik}^E). The elastic stress field σ_{ik}^Y near a pure (non-charged) prismatic dislocation loop is very well known from the dislocation theory [10,11]. The numerical calculation of total stress field (σ_{zz}^{tot}) near charged dislocation loop is shown in Fig. 4.

The numerical calculations for the total strain ε_{zz}^{tot} formed near charged dislocation loop are presented in Fig. 5. In the numerical calculations the following constants for cubic zirconia have been used [11]: $\mu = C_{44} = 60$ GPa, $\lambda = C_{12} = 100$ GPa, $v = 0.3$, $b = 0.51$ nm, $Q = qN_e = q\frac{\pi R^2}{a^2} n$ ($n = N_e/N_a$, n is the effective density of electrons per one oxygen atom (interstitial) in dislocation loop), $q = e$ (e is the electron charge), $\rho = 6$ (g/cm)^3, $\varsigma = 12.5$ (ς is the dielectric permeability of ZrO_2) [12,13]. The numerical calculations show that the distribution of modified stress and strain fields by electrical field near charged dislocation loops are stronger compared with non-charged dislocation loops and they are mostly controlled by electric field (see Figs. 4 and 5). The calculated shape and spatial distribution of total strain field near charged dislocation loop (Fig. 5) coincides with experimentally observed strain field contrast obtained by TEM near charged dislocation loop (see Fig. 1).

The critical conditions for the multiplication of dislocations (punching of dislocation lines) near charged dislocation loop (see Fig. 1) are determined by the conditions for the beginning of plastic deformation there. This process is started when the maximum value for the total stress tensor component is comparable or can exceed the yield stress value (σ_Y) ($\sigma_{ik}^{tot} \approx \sigma_Y$).

The theoretical value of shear strength is equal to $\sigma_Y^{th} = \mu/2\pi \approx 10^{-1}\mu$ [10]. The real experimental value of yield stress is less and equal to $\sigma_Y \approx 10^{-2}$–$10^{-3}\mu$ [10]. The critical radius (R_c) for the beginning of plastic deformation and multiplication of dislocation density near charged dislocation loop can be found from the following relation:

$$\sigma_{ik}^{tot}(R_c) = \sigma_Y^{th}. \quad (16)$$

Let us estimate the behavior of critical radius (R_c) as a function of concentration of trapped electrons (n) on

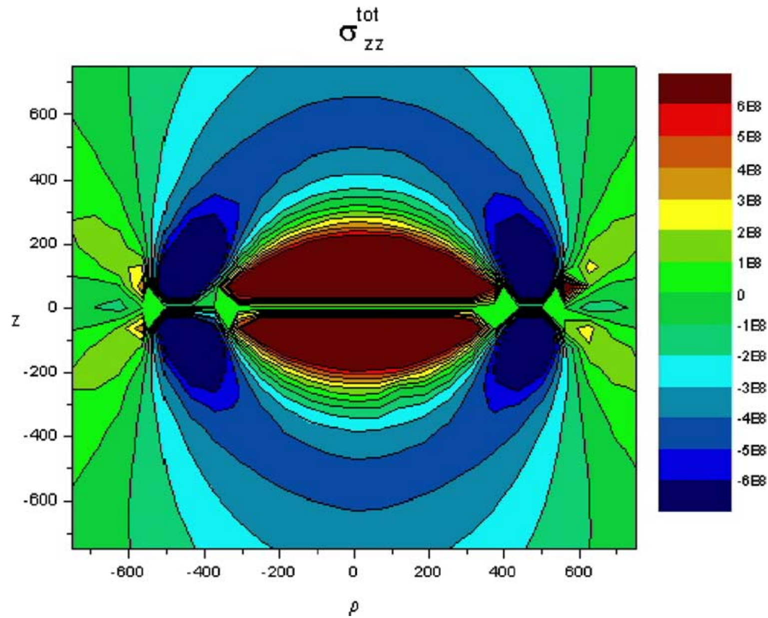

Fig. 4. Distribution of the total stress tensor component (σ_{zz}^{tot}) near a charged dislocation loop with radius $R = 2570$ Å (σ is in dyn/cm^2) with the density of trapped electrons per atom $n = 0.5$.

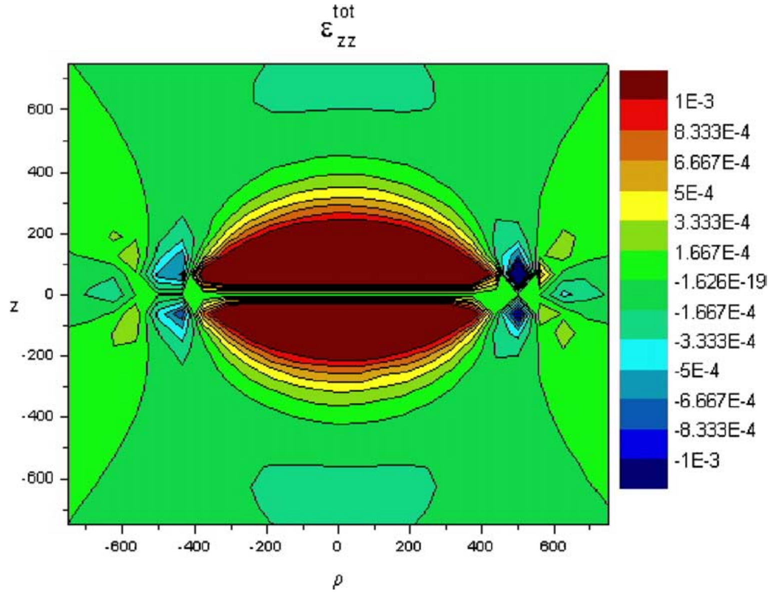

Fig. 5. Distribution of the total strain tensor component (ε_{zz}^{tot}) produced by the electrical field near a charged dislocation loop with radius $R = 2570$ Å with the density of trapped electrons per atom $n = 0.5$.

dislocation loop. Using the relation (9) we can obtain the values of electrical potential and electric field on the short distance ρ near edge of charged dislocation loop at $\rho \ll R$, which are given by the following relations:

$$\varphi = -\frac{Q}{\varepsilon R}\sqrt{\frac{2\rho}{R}}, \quad E = \frac{Q}{\varepsilon R}\sqrt{\frac{1}{2\rho R}}. \tag{17}$$

Using Eq. (17), we can obtain the total elastic stress field. The critical radius is also obtained through Eq. (16) and the relation of $Q = qN_e = q\frac{\pi R^2}{a^2}n$ as function of concentration of trapped electrons (n) on dislocation loop, which is equal to

$$R_c = \frac{16\varepsilon\kappa_0 a^4 \sigma_Y^{th}}{\pi q^2 n^2}. \tag{18}$$

The results of numerical calculations for the dependence of critical radius of a dislocation loop for the beginning of plastic deformation at a small distance ($\rho \approx a$) from the edge of a dislocation loop as a function of density of trapped electrons are shown in Fig. 6.

The performed numerical calculations for the total normal and shear stresses show the regions (see brown area in Fig. 4), where these values are comparable with the theoretical shear strength and where the plastic deformation can start ($\sigma^{tot} \geqslant \sigma_Y^{th}$). From the obtained numerical calculations we can see that the total stress (shear and normal components of stress tensor) on the small distances near charged dislocation loop for the some size of dislocation loop can exceed the theoretical yield stress and can be so high (in brown regions), that the plastic deformation (multiplication of dislocation density) can start here.

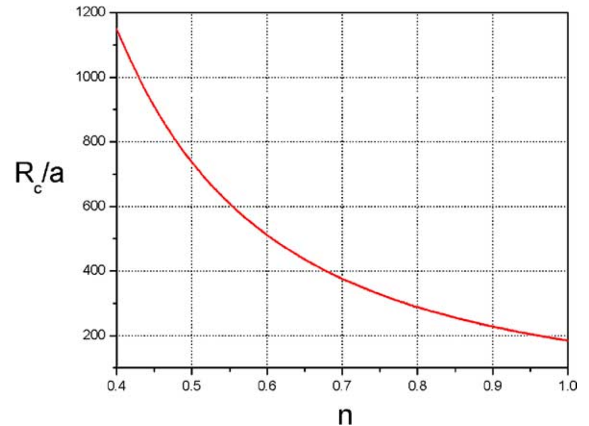

Fig. 6. The dependence of critical radius (R_c) as a function of density of trapped electrons (n).

4. Summary

(1) A new physical model for the explanation of growth and instability of anomalous large defect clusters (charged dislocation loops) with strong stress and strain fields, which have been experimentally observed in yttria-stabilized cubic zirconia (YSZ) under electron irradiation is suggested. The model

includes the formation and accumulation of an effective charge and absorption of neutral interstitial atoms on dislocation loops.

(2) Free electrons produced under electron irradiation due to the elastic collisions of fast electrons and atomic electrons are trapped at the edge of dislocation loops (dislocation cores of these dislocation loops) and this process results in an accumulation of effective charge on dislocation loops.

(3) The kinetics of dislocation loop growth is determined by the absorption of neutral interstitial atoms on dislocation loops. The obtained theoretical results for growth rate of dislocation loops have been compared with the experimental data.

(4) The charge accumulation on dislocation loops induces strong stress and strain fields (strong black/black lobes contrast), which are determined by the polarization of matrix and internal electric field formation near charged dislocation loops in dielectric (ceramic) material.

(5) The multiplication of dislocations near charged dislocation loops takes place due to the beginning of plastic deformation driven by the strong stress produced by internal electric field near such charged dislocation loops. The theoretically calculated total elastic stress near charged dislocation loops is comparable with the theoretical yield stress.

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