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Mechanism of the Formation and Evolution of Periodic Surface Relief Nanostructures under the Scanning Laser-Induced Inelastic Photodeformation of Semiconductors

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Abstract—A defect-deformational (DD) mechanism is proposed for the self-organization of laser-induced point defects (vacancies and interstitials) under low-threshold (far from the melting point) local (10–100 μm) light-induced heating with the scanning periodic pulsed laser irradiation of a semiconductor resulting in an inelastic deformation of micron-sized regions of Ge. A linear theory of DD instability is developed within the model of a biaxially stressed defective film. This model describes the main experimental data on the formation of two- and one-dimensional periodic nanostructures on a semiconductor surface relief.

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INTRODUCTION

The laser annealing of defects in ion-implanted semiconductor layers by a pulsed laser light is well understood [1, 2]. The combined effect of the electronic excitation, heating, and deformation of surface layers of a semiconductor is essential for the inverse processes of laser-induced point-defect formation (and other atomic rearrangements) [3, 4]. Unlike the case of quasi-one-dimensional irradiation [5], when shear deformations inside the illuminated surface area are insignificant, defect formation and destruction in semiconductors (and metals) under local irradiations (with a laser spot size of $\omega \sim 10$–100 μm) are characterized by so-called dimensional effects [6–8] and by strengthening of the influence of shear deformations and stresses arising in a semiconductor [9–11].

The formation of point and nanodimensional defects under conditions of local repeated irradiations controlled by the number of irradiation pulses N and other parameters can develop intensely only in inelastically deformable semiconductor surface layers [6–11]. It has been shown previously [10, 11] that these inelastic processes are limited in the energy density of laser pulses W from below by the region of linear photoacoustics [12–16] (W_0 thresholds) and from above by uncontrolled destruction processes with a sharp reduction in the intensity of light specularly reflected from the semiconductor or metal (W_d thresholds) [17–21].

The accumulation of laser-induced point defects from pulse to pulse [9–11, 22] and their interaction with each other via the strain field of the elastic continuum [23] can result in the self-organization of defects, i.e., in the formation of their clusters and periodic defect-deformational nano- and microstructures [24].

Recently, the features of processes of self-organization of laser-induced point defects, which were found for the first time under conditions of scanning local irradiation just at the initial stages of the inelastic photodeformation of germanium, have been investigated by atomic-force microscopy (AFM) (Fig. 1) [25]. The conclusion about the inelastic nature of the processes instigated studies of germanium [6, 7, 10] by photoacoustic microscopy based on the laser beam deviation technique [26]. It turned out that the transition from repeated pseudoelastic deformations of micron-sized areas of a semiconductor to inelastic deformations occurs already at the level of rather insignificant shear deformations, $10^{-5} < \varphi_0(W_0) \equiv (dU_z/dr)_{\max} < 10^{-4}$, where U_z are effective normal quasi-elastic (i.e. reversible angstrom-sized [10, 26]) displacements of surface atoms during irradiation, r are micron-sized distances along the radius of the beam from its center, and $W_0 \sim 0.1$ J/cm^2 is the corresponding threshold of the incident energy density in the center of the laser spot in submicrosecond laser pulses [9–11, 25].

Defect formation in Ge near the W_0 thresholds was studied by the field effect technique [9], diffuse and Raman scattering of light [10], and by molecular luminescent probes [10]. The mode of laser scanning, shown schematically in Fig. 2a, was applied to the AFM [23] and to the electrophysical and optical techniques mentioned above. The action of a laser beam with the wave length $\lambda = 0.53$ μm and the depth of optical absorption $\sim 10^{-5}$ cm was carried out in several scanning areas (each 3×5 mm in size) [10, 23]. Fixed (with

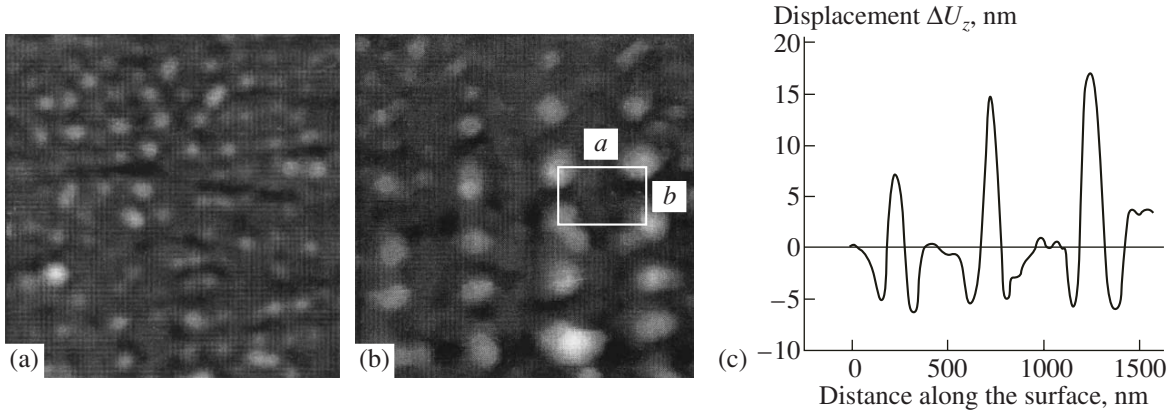

Fig. 1. (a) Nonoriented cluster formation, (b) generation of two-dimensional periodic lattices, and (c) generation of spatial profiles of irreversible normal displacements ΔU_z at initial stages of the inelastic ($W > W_0$) photodeformation of Ge according to the data taken from [25]. For both images (a) and (b), the image size is 2300 × 2300 nm. The lattice periods a and b are given in the text. The vertical and horizontal directions in panels (a) and (b) correspond to the y and x axes in Fig. 2a.

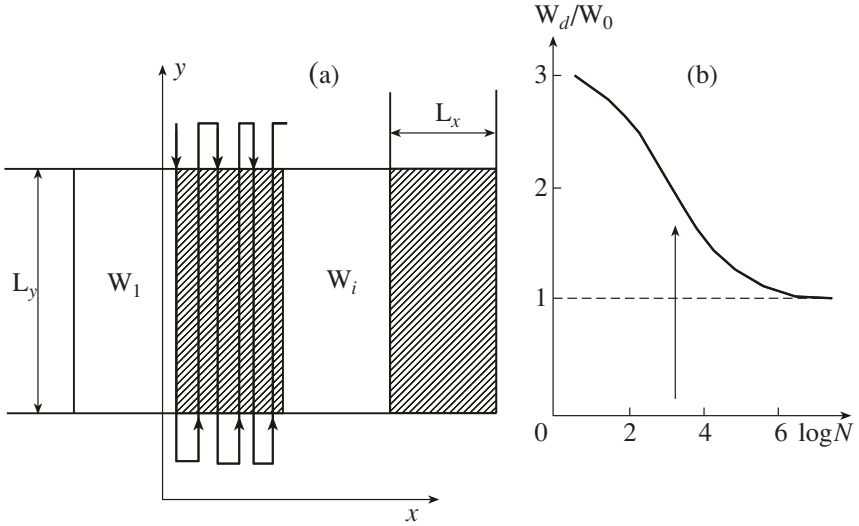

Fig. 2. (a) Schematic representation of the laser scanning of a semiconductor. Rectangular regions of $L_y \approx 5$ mm long (along the y axis) and of $L_x \approx 3$ mm wide (along the x axis) were scanned by a periodic pulsed laser beam with the size of a single-mode light spot $2\omega \approx 70$ μm on a sample. The scanning step along the x axis was $\delta \approx 5$–10 μm, and the beam scanning velocity along the y axis was $v_0 \approx 1$–5 mm/s. Pulses with the characteristic duration $\tau \approx 0.4$–0.5 μs followed with a repetition frequency $f_p \sim 10^4$ s^{-1} [25]. Designation: W_i is the incident energy density. (b) "Vertical" trajectory (arrow) of the variation of the incident energy density W at the center of a laser spot near thresholds $W_0 \approx 70$ mJ/cm^2 [6, 7, 10, 11] upon transition from one rectangular scanning area of Ge to another (i) against a typical kinetic curve of the thresholds of catastrophic destructions $W_d(N)$ of micron-size regions of a semiconductor (and other materials) [17–21, 27] presented in half-logarithmic coordinates of fatigue Wöhler curves [28, 29].

an accuracy of ~5–10%) incident energy densities W_i and N = const ~ 10^3 were used inside each scanning area. Only the energy density W_i (near the W_0 thresholds) changed in the transition from one scanning area to another (Fig. 2a), and the total number of irradiations and photodeformations arising in the semiconductor N remained constant.

A comparison of the results of studies of Ge by the specified experimental methods in different scanning areas (in the same semiconductor samples [10, 25]) makes it possible to establish the hierarchy of defect formation processes with increasing W and, hence, with growing amplitudes of shear deformations and corresponding subsurface stresses [7, 8] in the semiconductor. Thus, "vertical" transitions near W_0 were studied in [9–11, 25] in the half-logarithmic curves of thresholds of catastrophic destructions $W_d(N)$ [11], which, for semiconductors and metals, appeared [17–21, 27] quite similar in form to the known Wöhler curves [28, 29] (Fig. 2b).

EXPERIMENTAL NANOSTRUCTURES AND STATEMENT OF THE RESEARCH PROBLEM

It turned out that a rather weak (and consequently relatively safe for the semiconductor) point defect formation occurs in the pseudoelastic ($W < W_0$) conditions

of Ge photodeformations, which does not resulting its fatigue destruction down to $N \geq 10^6$–10^7 [6, 7, 9, 10]. At early stages of inelastic ($W_0 < W < 1.2$–$1.5W_0$) photodeformations of Ge ($N \sim 10^3$, Fig. 2b), more intense, but, as previously, hidden (latent) accumulation (from pulse to pulse, [22]) of point defects occurs in the majority of the scanning areas [9, 10, 25]. As well, the quasi-homogeneous distribution of their concentration is preserved as well as the character of the random nanorelief along the semiconductor surface [25].

According to the theory [23, 24], primarily, spatially unoriented nanosized defect clusters emerge with the further accumulation of defects in the inelastic regime of Ge photodeformations (Fig. 1a) [25]. Then, further self-organization of light-generated point defects (mostly, vacancies and interstitial atoms in a semiconductor, or deep electron traps and fluctuation defects in an oxide film) occurs at $W \geq W_l \approx (1.2$–$1.5)W_0$ [9, 10, 25]. These processes of self-organization of defects and their clusters result in the formation and evolution of the convexo-concave two-dimensional (and one-dimensional) periodic nanostructures (Figs. 1b, 1c) [25]. A qualitative consideration of the reasons of the formation of these structures is given in [25] on the basis of the defect-deformational approach [24].

In this work, a quantitative DD model [10, 25] is developed for the formation and evolution of periodic nanostructures of point defects and the surface relief of semiconductors, (Ge) which describes experiments on inelastic deformation under scanning multipulse irradiation. This model gives expressions for the period and the time of formation of a nanostructure, as well as for the critical concentration of defects exceeding which the formation of the structure occurs. The model also provides an interpretation of the following experimental facts that remained unexplained in [25]:

(1) at $W \geq 1.2W_0$, the period of structures $d_y \approx 350$–400 nm (size b in Fig. 1b) along the direction of beam scanning (y) was substantially smaller than the lattice period $d_x \approx 550$–600 nm (size a in Fig. 1b) across the scanning direction (x);

(2) at $W \geq 1.5$–$1.6W_0$, the individual convexities of the two-dimensional nanorelief merged along the direction of laser scanning (y) into parallel strips; i.e., quasi-one-dimensional (along x) submicron lattices of a relief with the period d_x were formed from the two-dimensional ones;

(3) along with the relatively small-scale periodic lattice of the nanorelief in Ge mentioned above, additional generation of formations with a large (several microns) spatial period of the relief was observed [25].

Let us also note that the role of interstitial defects in self-organization processes still remains to be clarified. These defects presumably participate in the formation of convexo-concave periodic nanostructures [25] along with vacancies, which are injected into the Ge depth by the "vacancy pump" mechanism [10, 29].

CLOSED SYSTEM OF EQUATIONS DESCRIBING DD INSTABILITY ON THE SURFACE

Let us assume that the most intense generation of point defects (vacancies and interstitials in Ge, or defects in an oxide film [9, 10, 25]) near W_0 thresholds, occurs mainly due to the induction of critical shear stresses under local laser irradiation [6–8]. Then, the thickness of the surface layer γ^{-1} enriched with defects by the vacancy pump mechanism [29] depends on the diffusion coefficient of defects and on the total (for N photodeformation cycles) duration of the stressed state of Ge [10] and amounts to about several micrometers [25]. Similar (the micron-level) estimation of γ^{-1} also gives a thermal length of $l_T = \sqrt{\chi\tau_p} \sim 3$ μm ($\chi \sim 0.33$ cm^2/s is the thermal diffusivity of Ge; $\tau_p \sim 0.4$ μs is the laser pulse duration). The second scaling parameter is the parameter h. The thickness of the defect-enriched surface layer initially presented in the sample due to technological pretreatment [3, 4] or the depth of the space-charge region [30] can serve as this parameter. In both cases, $h \sim 10^{-5}$ cm $\ll \gamma^{-1} \sim 10^{-4}$ cm. The two scaling parameters, h and γ^{-1}, determine two different (nanometer and micrometer) scales of surface DD structures. Since in this work we are basically interested in the formation of DD nanostructures, we will focus our attention on the scaling parameter h.

Let us consider a defect-enriched surface layer of thickness h (z direction), length L_y (y direction), and width L_x (x direction) as a "film" bonded to a substrate, which is the underlying part of a crystal. The z axis is directed inward the medium from the free surface of the film ($z = 0$). It is supposed that this layer differs in its elastic parameters from the underlying part of the crystal.

One may also assume that, because critical shear stresses generate microplastic deformations in the bulk of Ge [6–8] at thresholds W_0 [8–10], the near-surface (i.e., lying nearer to the surface) defective film of thickness h can shift rather easily in the horizontal direction; i.e., its bond with the substrate is weakened.

Let us also take into account that the lattice constant of a film saturated with dislocations is smaller than that in a dislocation-free crystal (subtraction dislocations). Then, because the deformation and the dislocation density are rather small outside the rectangular (Fig. 2) area of laser scanning, the film of thickness h would be under biaxial surface tension (along the x and y axes).

Because the linear dislocation densities (per unit length of a semiconductor) in the laser scanning direction (y) and in the perpendicular direction (x) are different as a result of the specificity of scanning conditions, the tensile stresses in the specified directions will also be different. Let us further characterize this defect-induced elastic anisotropy in a film h by the factor $k \equiv \sigma_x/\sigma_y$. The numerical value of k depends on the characteristics of laser scanning; its estimation for the experiments [9, 10, 25] is given below.

In an effort to describe the symmetry, period, and time of formation of a DD structure, we limit our consideration to only the initial (linear) mode of DD instability. Considering the redistribution of defects only along the surface, we obtain the equation for their concentration

$$n_d(x, y, z, t) \equiv N_d(x, y, t)\exp(-\gamma z), \quad (1)$$

where $N_d(x, y, t)$ is the defect concentration on the surface.

The surface flow of defects consists of the diffusion and deformation-induced parts. Surface diffusion and drift are considered to be isotropic in the model equations. Then, the influence of the stress-induced anisotropy of diffusion and drift is taken into account:

$$j_d = -D_d\nabla N_d + N_d D_d\frac{\theta_d}{k_B T}\nabla(\mathrm{div}\mathbf{u}_f)_{z=0}. \quad (2)$$

Here, D_d is the diffusion coefficient of defects (the diffusion is supposed to be isotropic), $dKa^3\mathrm{sign}\theta_d$ is the defect deformation potential (K is the modulus of elasticity, a is the unit cell size), $\mathbf{u}_f$ is the vector of the displacement in the film, k_B is the Boltzmann constant, the operator $\nabla = e_x\frac{\partial}{\partial x} + e_y\frac{\partial}{\partial y}$, and $\mathbf{e}_x$ and $\mathbf{e}_y$ are the unit vectors along the x- and y axis, respectively.

Using Eq. (2), from the equation of continuity, we obtain the diffusion equation for N_d with the drift taken into account:

$$\frac{\partial N_d}{\partial t} = D_d\Delta N_d - D_d\frac{\theta_d}{kT_B}\mathrm{div}[N_d\nabla(\mathrm{div}\mathbf{u}_f)_{z=0}], \quad (3)$$

where $\Delta = \frac{\partial^2}{\partial x^2} + \frac{\partial^2}{\partial y^2}$.

The deformation of the film $\mathrm{div}\mathbf{u}_f$ is expressed in terms of the film bending coordinate ζ, which measures the displacement of the film midplane points from their equilibrium positions along the z axis, by the equation [31]

$$\mathrm{div}\mathbf{u}_f = -\nu\left(z - \frac{h}{2}\right)\Delta\zeta, \quad (4)$$

where $\nu = (1 - 2\sigma)/(1 - \sigma)$, σ is the Poisson ratio of the film.

The coordinate ζ obeys the equation that results from the generalization of the conventional bending equation of a free film [31]:

$$\frac{\partial^2\zeta}{\partial t^2} + l_0^2c^2\Delta^2\zeta - \frac{1}{\rho}\left(\sigma_x\frac{\partial^2\zeta}{\partial x^2} + \sigma_y\frac{\partial^2\zeta}{\partial y^2}\right) = \frac{\theta_d}{\rho h}\int_0^h\frac{\partial n_d}{\partial z}dz, \quad (5)$$

where c (cm/s) is the coefficient of rigidity of the film, $l_0^2 = h^2/12$, and ρ is the density of the semiconductor. The terms σ_x and σ_y introduced into Eq. (5) take into account the influence of the anisotropic lateral tensile stress. In the second term in the left-hand side of Eq. (5), we neglect the mixed derivatives $\partial^4/\partial x^2\partial y^2$, which make a contribution on the order of $\sigma_i/\rho c^2 \ll 1$. The right-hand side of Eq. (5) takes into account the stress normal to the film surface, arising from the inhomogeneous distribution of defects along the z axis.

The set of Eqs. (3)–(5) subject to Eq. (1) is a closed system of equations describing the DD instability on a surface.

GROWTH INCREMENT AND THE PERIOD OF SURFACE DD LATTICES

Let us represent the concentration of defects on a surface as

$$N_d = N_{d0} + N_{d1},$$

where N_{d0} is the spatially homogeneous part, and $N_{d1} = N_{d1}(x, y, t)$ is the spatially inhomogeneous part of the surface concentration of defects. Assuming that the deformation is adiabatically adjusted to the defect subsystem ($\left(\frac{\partial^2\zeta}{\partial t^2} = 0\right)$), we obtain the following equation from Eq. (5) after integration under the condition $\gamma h \ll 1$:

$$\Delta^2\zeta - \frac{1}{l_x^2}\frac{\partial^2\zeta}{\partial x^2} - \frac{1}{l_y^2}\frac{\partial^2\zeta}{\partial y^2} = -AN_{d1}, \quad (6)$$

where $A = \frac{2\theta_d}{hl_0^2\rho c^2}$, and the characteristic scale parameters of DD lattices with the wave vectors along x and y:

$$l_{x,y} = h\left(\frac{\rho c^2}{12\sigma_{x,y}}\right)^{1/2}. \quad (7)$$

Linearizing Eq. (3), we obtain, in view of Eq. (4), the equation

$$\frac{\partial N_{d1}}{\partial t} = D_d\Delta N_{d1} - D_d BN_{d0}\Delta^2\zeta, \quad (8)$$

where $B = \frac{\nu\theta_d h}{2kT_B}$.

Assuming that the defect concentration and the film bending coordinate in the DD lattice are described by the equations:

$$N_{d1} = N_{\mathbf{q}}\exp(i\mathbf{qr} + \lambda_{\mathbf{q}}t) + \text{e.n.},$$
$$\zeta_{\mathbf{q}} = \zeta_{\mathbf{q}}\exp(i\mathbf{qr} + \lambda_{\mathbf{q}}t) + \text{e.n.},$$

where $\mathbf{r} = (x, y)$, we obtain from Eqs. (6) and (8) the set of two homogeneous equations for the Fourier amplitudes $N_{\mathbf{q}}$ and $\zeta_{\mathbf{q}}$, whence we have for the increment of instability

$$\lambda_{\mathbf{q}} = -D_d\mathbf{q}^2 + D_d\mathbf{q}^2N_{d0}\frac{AB}{\mathbf{q}^2 + l_x^{-2}\cos^2\varphi + l_y^{-2}\sin^2\varphi}, \quad (9)$$

where φ is the angle between the vector **q** and the x axis.

When the stresses along x and y are equal to each other ($l_x = l_y$), the dependence on the angle φ in Eq. (9) disappears. However, the renormalization of the diffusion coefficient $D_d = D_{d0}\exp[-(E - \nu\sigma)/k_BT]$ by the stress σ should be taken into account, where E is the diffusion activation energy, and ν is the activation volume. Then, the diffusion coefficient and, consequently, the increment (9) possess the maximum value for the directions along which the stress σ is applied, i.e., along the x and y axes.

In view of this, we find that the dependence $\lambda_{\mathbf{q}} = \lambda(\varphi, q)$ (9) selects the DD lattices by the angle φ and the wave number **q**. As functions of the angle, a lattice with $\mathbf{q}\|e_x(\cos\varphi = 1)$ and a lattice with $\mathbf{q}\|e_y(\sin\varphi = 1)$. In the first case, we obtain from Eq. (9) the following dependence of the increment of the lattice with $\mathbf{q}\|e_x$ on the wave number:

$$\lambda_q = -D_d q^2 + D_d q^2 N_{d0} l_x^2 \frac{AB}{1 + l_x^2 q^2}.$$

The dependence λ_q reaches a maximum value at $q = q_{mx}$, and

$$q_{mx} = \frac{1}{l_x}\left(\sqrt{\frac{N_{d0}}{N_{cx}}} - 1\right)^{1/2}. \quad (10)$$

The maximum value of the increment for the lattice with $\mathbf{q}\|e_x$ is

$$\lambda_{mx} = D_d q_{mx}^2(\sqrt{N_{d0}/N_{cx}} - 1) = \frac{D_d(\sqrt{N_{d0}/N_{cx}} - 1)^2}{l_x^2}. \quad (11)$$

Here, N_{cx} is the critical concentration of the formation of the DD lattice with the wave vector $\mathbf{q}_{mx}$ along the x axis,

$$N_{cx} = \sigma_x \frac{kT}{\nu\theta_d^2}. \quad (12)$$

The period of the dominant DD lattice with the wave vector directed along the x axis is $d_x = 2\pi/q_{mx}$. With an excess over the critical concentration N_{cy}, the expression for which is obtained from Eq. (12) by the replacement $x \longrightarrow y$, the DD lattice with the wave vector $\mathbf{q}_{my}$ directed along the y axis and the period $d_y = 2\pi/q_{my}$ is formed. The expressions for q_{my} and λ_{my} follow from Eqs. (10) and (11), respectively, upon replacement of $\sigma_x \longrightarrow \sigma_y$. Thus, the expression for the period of two lattices with the wave vectors $\mathbf{q}_{mx}$ and $\mathbf{q}_{my}$ can be represented in unified form:

$$d_i = \frac{2\pi l_i}{\left(\sqrt{\frac{N_{d0}}{N_{ci}}} - 1\right)^{1/2}}, \quad (13)$$

where l_i and N_{ci} are specified for the case $i = x$ by Eqs. (7) and (12), respectively, and are obtained for the case $i = y$ from Eqs. (7) and (12) by the replacement of $\sigma_x \longrightarrow \sigma_y$.

The resulting surface DD structure is obtained as a superposition of two surface DD lattices with the wave vectors directed along the x and y axes

$$N_{d1} = N_{q_{mx}}\exp(iq_{mx}x + \lambda_{mx}t) + N_{q_{my}}\exp(iq_{my}y + \lambda_{my}t) + \text{c.c.}, \quad (14)$$

$$\zeta = \zeta_{q_{mx}}\exp(iq_{mx}x + \lambda_{mx}t) + \zeta_{q_{my}}\exp(iq_{my}y + \lambda_{my}t) + \text{c.c.} \quad (15)$$

with the periods along x and y axes specified by Eq. (13).

The absolute extrema of the two-dimensional superposition DD structure specified by Eqs. (14) and (15) form a two-dimensional rectangular cellular structure on the surface.

DISCUSSION AND COMPARISON WITH EXPERIMENT

The prediction of the formation of the two-dimensional rectangular cellular structure of a surface relief (15) by the DD model under the conditions of experiments [25] corresponds to the results obtained in [25].

Let us estimate the critical concentration of defects at which a DD lattice is formed in Ge and the period of the lattice. Provided $k_BT = 5 \times 10^{-2}$ eV, ($T = 600$ K), $\sigma_x = 10^9$ dyne cm^{-2}, $\theta_d = 50$ eV, $h = 10^{-5}$ cm, we have $N_{cx} \sim 10^{16}$ cm^{-3} from Eq. (12). Provided $\rho \approx 5.3$ g cm^{-3}, $c = 10^5$ cm s^{-1}, and $N_{d0}/N_{cx} = 10^2$ ($N_d \sim 10^{18}$ cm^{-3}), we have $d_x \sim 4 \times 10^{-5}$ cm from Eq. (13) for the period of the DD lattice, which is close to the experimental value. The nonequilibrium diffusion coefficient of defects at which the formation of a DD structure can form during time t_{form} will be estimated from the condition $\lambda_{mx}t_{\text{form}} \sim 20$. Using Eq. (11) for the increment, we obtain $D_d = 20d_x^2/4\pi^2 t_{\text{form}}$. Using the estimate $t_{\text{form}} \sim N\tau_{\text{def}} \sim 10^{-2}$ s ($N = 10^3$ is the number of pulses, and $\tau_{\text{def}} = 10$–15 μs is the characteristic relaxation time of local quasi-static photodeformations of Ge after each pulse [10, 25]), we have $D_d \sim 10^{-7}$ cm^2/s. This means that a DD structure can form during the total time of existence of the stressed state of the semiconductor [10], which arises after each irradiation pulse ($\tau_p \geq 0.4$ μs [15, 16]) if the stress and temperature in Ge increase the diffusion coefficient of defects as compared to the equilibrium value. It is known that extrapolation of the diffusion coefficient of vacancies in Ge to room temperature $D_v \sim 4 \times 10^{-8}$ cm^2/s [29] is only slighty lower than the above estimation of D_d. This is indirect evidence of rather

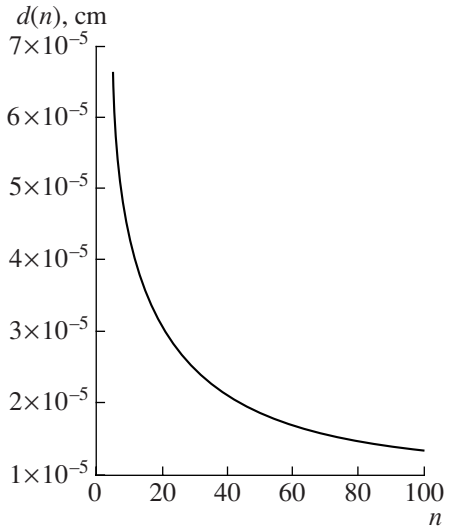

Fig. 3. Dependence of the period of a defect-deformational lattice $d_i(n)$ on the excess over the threshold of lattice formation $N_{d0}/N_{ci} = n$. The dependence is plotted using Eq. (13) at values of parameters specified in the text and $\sigma_i = 10^9$ dyn/cm^2.

low-temperature (i.e., far from the melting point [11]) character of the studied formation of DD lattices [25].

To obtain a quantitative assessment of the ratio of periods d_x/d_y, let us find the defect-induced anisotropy factor k for the scanning mode used in [25] (Fig. 2a). Let us consider the direction x in the rectangular scanning area. The tensile stress along the x axis $\sigma_{1x} = Ka/L_x$ arises when one vacancy disk (subtraction dislocation) whose plane is perpendicular both to the surface and x axis is located in this area. Let N_x (cm^{-1}) dislocation lines perpendicular to the x axis be formed on the surface per unit length along the x axis as a result of laser action (i.e., N_x vacancy disks are inserted per unit length perpendicular to the x axis). The total number of these dislocation lines at length L_x is N_xL_x, and the full tensile stress along the x axis is equal to $\sigma_x = \sigma_{1x}N_xL_x = KaN_x$. Given $K \sim 10^{12}$ dyn/cm^2, $a \sim 5 \times 10^{-8}$ cm, $L_i \sim 0.1$ cm, $N_x \sim 2 \times 10^6$ cm^{-1} (the density of laser-induced dislocations under multipulse irradiation is $\sim 10^{12}$ cm^{-2}), we have the estimate $\sigma_i \sim 10^{10}$ dyn/cm^2. In the same way, we obtain $\sigma_y = KaN_y$. Then, provided $N_{d0} \gg N_{ci}$, we obtain from [22] $d_x/d_y = (\sigma_y/\sigma_x)^{1/4} = (L_yN_y/L_xN_x)^{1/4}$. Since the same number of dislocation lines (both along y and along x) arise upon each pulse in the laser spot, the linear density of dislocation lines perpendicular to the x and y axes is proportional to the number of centers of laser spots fitting into length L_x and L_y upon scanning: $N_x = \text{const}(L_x/\Delta x)$ and $N_y = \text{const}(L_y/\Delta y)$, respectively. Here, $\Delta x = \delta$ and $\Delta y = (\nu_0/f_p)$ are the center distances between adjacent laser spots on the sample along the x and y axes, respectively ($\delta \approx 5$ μm is the scanning step along the x axis, $\nu_0 = 5$ mm/s is the beam scanning velocity along the y axis, and $f_p \sim 10^4$ s^{-1} is the laser pulse repetition frequency). Then, the anisotropy factor of the dislocation-induced stress is given by

$$k \equiv \sigma_x/\sigma_y = \frac{L_x^2\Delta y}{L_y^2\Delta x} = \frac{L_x^2}{L_y^2}\frac{\nu_0}{\delta f_p} \sim 4 \times 10^{-2}$$

at $L_y \approx 5$ mm and $L_x \approx 3$ mm. Using this estimation and Eqs. (13) and (7), we obtain $d_x/d_y = (\sigma_y/\sigma_x)^{1/4} \sim 2$, which is in satisfactory agreement with the experimental value $d_x/d_y \approx 1.5 \pm 0.1$.

With the growth of the energy density in the pulses W (mainly, at $W > W_0$ [6, 9, 10, 25]) and the corresponding growth of the defect concentration N_{d0}, the period of the DD lattice must decrease (Fig. 3) according to Eq. (13). This reduction in the period occurs via the rearrangement of the DD lattice by means of the escape of defects from the self-consistent deformation wells of the old lattice and the deformation-induced drift of defects into the deformation wells of the rearranged lattice. Because $\sigma_y > \sigma_x$ (by a factor of 10, according to our estimations), the stress-renormalized diffusion coefficient of defects [32] along the y axis can be greater than the analogous diffusion coefficient along the x axis. Therefore, the deformation-induced drift flow of defects along y exceeds that along x. This can explain the established fact [25] that only the DD lattice with the wave vector $\mathbf{q}_{my}$ is rearranged with the reduction in its period as the energy density W increases (at $N \sim 10^3$). This process results in the fact that only one DD lattice with the wave vector $\mathbf{q}_{mx}$ eventually remains.

Let us note that similar DD instability can also occur simultaneously and independently in a thicker defect layer ("film") $\gamma^{-1} \sim 10^{-4}$ cm thick. The arising modulation of the surface relief must have a micron-size period and can be responsible for the slow (micron-level) modulation of the nanometer relief of the surface observed in [25].

CONCLUSIONS

Thus, the linear theory of DD instability developed in this work in the model of a biaxially stressed defective film is able to describe the basic experimental data on the multipulse laser-induced formation and evolution of surface nanostructures in inelastically photodeformed Ge [25]. More detailed specification of the shape of DD structures, i.e., quantitative description of the spatial profiles of germanium "nanoasperities" and "nanovalleys" arising around them (Fig. 1c) [25], requires a nonlinear analysis of this model and detailed specification of the role of vacancies and interstitials in the formation of DD nanostructures on the Ge surface. In summary, note that the developed model of DD self-organization is also applicable for the description of the self-organization of nanostructures observed upon photodeformation of other semiconductors, in particular, GaAs [27] and Si.

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