A new method of replacement atoms for the synthesis of epitaxial layers of SiC on Si: From theory to practice

S.A. Kukushkin1,2, A.V. Osipov1

1Institute of Problems of Mechanical Engineering RAS, V.O., Bolshoj pr. 61, St. Petersburg, 199178 Russia
2St. Petersburg State Polytechnical University, Polytechnicheskaya 29, 195251 St.Petersburg, Russia

E-mail: sergey.a.kukushkin@gmail.com

Abstract. The new method of epitaxy of low-defect unstressed nanoscaled silicon carbide (SiC) films on silicon (Si) substrates is theoretically developed and experimentally realized in the case of the big mismatch of lattices of a film a substrate (~20 %). The method consists in the replacement of a part of atoms of the Si matrix by atoms of C with the formation of SiC molecules. The laboratory model of the operating light-emitting diode is originally obtained based on these structures.

1. Introduction

The tremendous elastic stress appearing during the growth of single crystal films with the lattice parameter strongly differing from the lattice parameter of the substrate do not allow us to obtain high quality layers of new wide-bandgap semiconductors on silicon (Si) without misfit dislocations. It is assumed that the most promising of these are silicon carbide (SiC), gallium nitride (GaN), aluminum nitride (AlN), zinc oxide (ZnO), and some others. Integration of these materials into the silicon electronics plays a key role for the development of industrial technologies; therefore, it is very important to deposit epitaxial films of these materials just on silicon. However, due to the large difference in lattice parameters of Si and all the mentioned wide-bandgap semiconductors (~20%), during the ordered growth of these materials the stress energy rises. This leads to the appearance of a tremendous number of misfit dislocations in growing films and even to complete film cracking.

2. Theory

In this work, we suggest new relaxation mechanism of the elastic energy for growing dislocation-free heteroepitaxial films. The essence of this approach, which differs from all the existing methods of film growth, is based on the idea of preliminary incorporation of point defects into the crystal lattice of the silicon host. When growing the SiC film on the Si substrate, such defects are the carbon atom C placed in the Si interstitial position and the vacancy formed as a result of removal of one of the Si atoms. If these defects are attracted to each other by the elastic interaction in the Si matrix, the resulting elastic energy caused by their incorporation into the substrate host is considerably lower than the energy of no interacting defects. It is well known that the spherically symmetric dilatation centers do not interact with each other at all in an isotropic medium of infinite size. We show [1,2] that the dilatation centers can attract to each other in substantially anisotropic media, such as the crystal with...
the cubic lattice symmetry, thereby considerably decreasing the total elastic energy. Such attractive centers form stable objects of a new type, namely, the elastic dilatation dipoles. We calculated the elastic energy of the system using heteroepitaxy of the SiC film on the Si substrate (Si has a cubic lattice) as the example and showed that the elastic energy can relax completely only due to the ensemble of dilatation dipoles. It has been proven in [1,2] that the total elastic energy of a crystal during the formation of interacting defects inside its volume is

\[ W(\cos \varphi(x,y,z)) = E_C + E_v + E_{\text{int}}(\cos \varphi(x,y,z)). \] (1)

Here \( E_C \) and \( E_v \) are the elastic energies of the lattice distortion by the carbon atom and the vacancy

\[ E_{C,v} = \frac{2\mu^{Si}K^d}{3K^d + 4\mu^{Si}} \left( \nu^d - \nu^{Si,m} \right)^2, \] (2)

\( E_{\text{int}} \) denotes the energy of defects interaction

\[ E_{\text{int}}(\cos \varphi(x,y,z)) = -E_0 \left( \cos^4 \varphi_x + \cos^4 \varphi_y + \cos^4 \varphi_z - \frac{3}{5} \right), \] (3)

\[ E_0 = \frac{15K(3c_{11} - 4c_{44})\Omega^f\Omega^l}{8\pi(c_{12} + 2c_{44})r^3}, \] (4)

cos \( \varphi_i = x_i/r \) are the cosines between the axes \( x, y, z \) and a direction of the line connecting the centers of interacting defects. The expression \( \cos^4 \varphi_x + \cos^4 \varphi_y + \cos^4 \varphi_z - 3/5 \) reaches its minimum, equaled to -0.27, in the direction <111> corresponding to an attraction of similar defects. Its maximum, equaled to 0.4, is reached in the direction <100> corresponding to a repulsion of similar defects. Also it has a saddle point -0.1 in the direction <110>. Therefore, the direction <111> is the most energetically favorable for similar point defects. It corresponds to a situation when C atom is located along (110) plane and under (111) planes in an interstitial position, whereas a vacancy is located along <111> direction with respect to it. In this case, the attraction between the silicon vacancy and the carbon atom (both of them are compressing dilation centers) reaches its maximum. That is the considered dilatation dipole provides the maximal relaxation of the elastic energy [1,2]. Taking into account that \( n_1^2 + n_2^2 + n_3^2 = 1 \) \( (n_1 = \cos \varphi_x, n_2 = \cos \varphi_y, n_3 = \cos \varphi_z) \), Eq. (3) yields

\[ E_{\text{int}} = -E_0 \left[ \frac{2}{5} - 2(n_1^2n_2^2 + n_2^2n_3^2 + n_1^2n_3^2) \right] \approx 6 \cdot 10^{-19} \text{ J/atom}. \] (5)

This mechanism of elastic energy relaxation is extreme efficient. Indeed, if dilatation centers do not attract to each other, the total elastic energy per SiC molecule is \( W_{\text{dil}} = 0.22 \text{ eV} \). But if one part of the dilatation dipole is the carbon atom, the elastic energy per SiC molecule is \( W_{\text{dip}} = 0.13 \text{ eV} \). Every 8 atoms of silicon in the original matrix could be replaced by no more than 4 dilatation dipoles, therefore, the maximum relaxation of the elastic energy owing to dipoles is \( \Delta W = (W_{\text{dil}} - W_{\text{dip}})/2 = 0.2 \text{ eV} \) per one molecule of SiC. In other words, as shown in [35-37], the elastic energy can relax completely by dilatation dipoles alone, because \( W_{\text{dil}} - \Delta W \approx 0 \). And the lifetime of dilatation dipoles at the temperatures 1100-1250°C is about \( \sim 10^{-1} \text{ sec} \), that is three orders of magnitude greater than the formation time of the first SiC monolayer. So, the elastic dipoles play a role of ordering epitaxy centers. In the opposite case (when the point defects repel being located along
<100> direction), the dipoles will not arise and increase the elastic energy, that leads to layers disorientation.

3. The formation of transition state “elastic dipoles – silicon matrix” as a necessary condition for the synthesis of low-defective SiC layer on Si

In order to provide the effective usage of a new relaxation mechanism of elastic energy due to the interaction of point defects, we suggest the deposition process of SiC not from the vapor phase but immediately from the matrix of the single crystalline Si substrate due to the chemical reaction between the crystalline Si and gaseous carbon monoxide CO [3]

\[
2\text{Si (solid)} + \text{CO (gas)} = \text{SiC (solid)} + \text{SiO (gas)}.
\]  

(6)

We selected this reaction because of the fact that the forming gaseous silicon monoxide SiO partially carries the atoms from the Si matrix inducing vacancies in it, while gaseous carbon monoxide CO is the source of carbon atoms C arranged in atomic voids of the silicon lattice. Both Si vacancies and incorporated C atoms are the compressing dilatation centers in the cubic Si lattice and interact with each other [1] (see Fig.1).

![Fig. 1. Schematic illustration of elastic dipoles in the cubic lattice of Si. Large blue balls are Si atoms, small green ones are C atoms in the interstitial positions. The strings indicate the elastic interaction between dilatation centers (vacancies and C atoms) with dipole formation.](image)

Figure 2 represents the difference between the classical mechanism of thin film growth and the mechanism of epitaxy suggested here. In the first case the film grows on the substrate surface which results in enormous stress energy, whereas in the last case the film grows inside the substrate, and the attraction between point defects provide the complete relaxation of the stress energy. The merging process of vacancies after film nucleation results in the pore formation under the film [1,2].

![Fig. 2. Schematic pattern of regular thin film growth (a) and the suggested one (b). In the second case dumbbells represent the attraction between different point defects (dilatation dipoles).](image)
The chemical reaction (6) proceeds in two stages. On the first stage the intermediate compound consisting of activating complex, namely, Si vacancy – C atom – Si matrix is formed. C atoms are placed in interstitial positions of Si matrix. On the second one the activating complex transforms into SiC, and released Si vacancies merge into pores in Si under the layer of SiC. It is found that the activating complex consists of the ensemble of dilatation dipoles ordered by the original Si matrix. The transition of the activating complex into SiC and pores is the structural phase transition of shifting type. In this transition the ordering of SiC molecules is followed by the original crystalline Si matrix. The mechanism of heterogeneous chemical reaction between gas and solid is investigated on the example of the reaction (6).

New type of phase transitions, namely, the transition through the intermediate stage, when one of the phase stimulates the nucleation of another one, is described theoretically. Formation work and all the main thermodynamic and kinetic characteristics of two-phase phase transitions are calculated. Conditions are found at which the dislocation-free growth of SiC films on Si takes place [4].

Assembly of nano-layers of SiC on Si by the synthesis of dilatation dipoles is realised on practice for the first time and high quality templates SiC/Si is obtained [5].

The process of phase transformation suggested in this work is very similar to the process of full replacement of living cells of a tree by inorganic. Such processes happened to the trees which originally got to sea hot water and have lain under a layer of earth for one hundred millions years. The photo of such mineral tree of 250 million years is given in Fig. 3.

**Fig. 3.** The hardened tree which has lain in hot water saturated with mineral salts. Living cells of this tree were completely replaced with inorganic minerals as a result of slow exchange process

### 4. Experimental and Results

To confirm experimentally the suggested relaxation mechanism of elastic energy, Si (111) substrates 35 mm in diameter were held in a vacuum furnace at $T = 1100–1300^\circ$C in the atmosphere of CO at $p = 10–300$ Pa for 5–60 min [3]. A SiC film 50–200 nm thick grew inside the Si substrate during this time. The average value of tensile elastic stresses in SiC films measured by an FLX-2320-S thin film stress measurement system is 0.5 GPa in the absence of lattice misfit and cracks [1–4]. Microscopic analysis revealed an almost ideal conjugation of the lattices of silicon and silicon carbide [5]. Such a low measured elastic energy with the ideal conjugation of lattices in the absence of misfit dislocations and cracks can be interpreted by the relaxation of elastic stresses due to the ensemble of dilatation dipoles. Monocrystal nanoscaled layers of SiC of hexagonal polytypes 4H and 6H are obtained on Si for the first time. Best of them have FWHM $\omega-20$ in the range $0.4^0–0.65^0$ at the SiC thickness of 100 nm, and FWHM $\omega-20$ in the range $0.1^0–0.15^0$. At the present time SiC layers on Si substrates of 6 inch are available [6].
5. The growth of semiconductor films AlN, GaN and ZnO on SiC/Si(111) substrates

Nowadays III-nitride heterostructures are grown mainly on the substrates of sapphire, Al₂O₃ or SiC. However, the growth of such structures on the silicon substrates provokes the greatest interest due to the prospect of integration of the gallium nitride and silicon electronics, the possibility of exploiting of a large-diameter substrates, their low cost and good electrical and thermal conductivity.

The large difference of lattice constants (17%) and thermal expansion coefficients (33%) of GaN and Si is the main reason of deformation of GaN/Si heterostructures leading to a high defect density in GaN epitaxial layer directly grown on Si [7,8]. The essential difference in the lattice parameters of SiC and Si (20%) resulting in the formation of misfit dislocation arrays on their interface also affects GaN films grown on Si/SiC substrates. Even the best GaN/SiC/Si samples obtained by MOVPE on 1µm SiC layers contain ~10⁹/cm² of misfit dislocations. In [9] has been made an attempt to improve the layer structural quality by annealing and formation of helium bubbles in the silicon bulk. According to the authors the stress field around the bubbles redirects the misfit dislocations in the substrate, thereby improving the structural quality of the epitaxial layer. Most researchers try to reduce the defect density in GaN layer by adjusting the growth conditions and composing the buffer layers of different composition and thickness (AlₓGa₁₋ₓN, AlN, SiC or theirs combination ). GaN epitaxial layers with thickness up to 3µm have been successfully synthesized on Si by MOCVD using AlN intermediate layers [10], but the growth of thicker layers (over 10µm) of AlN or GaN by HVPE became very difficult due to the deformation of the structure. To suppress the process of defect formation and to reduce the deformation of GaN epitaxial layer the different intermediate layers are used (TiN, HfN, ZnO, SiNₓ, etc.). Recently, the attempts are made to use the intermediate layers of silicon carbide of cubic modification, β-SiC [11].

To answer the questions put in the beginning of this section we have performed a comprehensive analysis of structural, crystallographic, physicochemical, electro-optical and spectroscopic studies of AlN, GaN and ZnO films grown on the SiC/Si substrates [12-17]. Two methods, MOCVD and HVPE, are commonly used for the deposition of the films of AlN, GaN or AlₓGa₁₋ₓN solid solution. ZnO films are grown by the magnetron sputtering.

A distinctive feature of SiC films synthesized by the solid-phase epitaxy [1-4] is the ability to obtain in a single process the heterolayers of silicon carbide of both cubic β-SiC and hexagonal α-SiC polytypes [1-4]. Moreover, the inverted layers could be grown, i.e. β-SiC bottom layer (lying on the silicon substrate) and α-SiC upper one, and vice versa. Or even the layer containing the polype type periodically distributed over the substrate. And grown single-crystal defectless SiC layers ensure a substantial relaxation of the elastic thermal stresses covering the pores between the silicon carbide and the silicon, even if β-SiC films are used as buffer layers.

In Ref.[12] the results of electron microscopic analysis of GaN/AlN/nanoSiC/Si(111) have been presented. The layers of gallium nitride and aluminum chloride were grown using the hydrogen chloride gas epitaxy (HVPE) and the layers of nano silicon carbide – by the original method of the solid-phase epitaxy [1-4]. The grown GaN layer (of 20 µm thickness) had a good structural quality with no misfit dislocations parallel to the growth axis [0001] “c” (the overall dislocation density in GaN layer was less than 10⁷cm⁻²). In several samples cut from the different portions of the 3-inch diameter substrate studied in [4] these dislocations were absolutely absent. However, we do not insist that they do not present on the whole 3-inch surface without exceptions. Number of screw dislocations in the layer was about 4⋅10⁹cm⁻² and the total dislocation density of mixed type was 8⋅10⁹cm⁻². This value is significantly less than the density of dislocations in GaN layers grown on Si with an intermediate layer of AlN by MBE and it competes with the dislocation density in gallium nitride
layers grown on the sapphire by MOCVD. According [18] the density of dislocation oriented parallel to the growth in GaN films grown on the sapphire by MOCVD is \((3.4\pm0.5) \times 10^7\) cm\(^{-2}\), of the perpendicularly oriented \((4.5\pm0.5) \times 10^7\) cm\(^{-2}\) and the density of inclined dislocations is \((1.1\pm0.3) \times 10^6\) cm\(^{-2}\). It should be noted that the dislocations oriented parallel to the growth axis and inclined dislocation (to a lesser extent) have a significant effect on the electrical characteristics of the structure. These dislocations decrease the electrical conductivity of the layer and thus increase the heat generation reducing the efficiency of the LEDs fabricated on such structures. The samples of GaN obtained on nano-SiC/Si substrates these dislocations are practically not observed [12]. It should be noted that GaN films has been grown using a thin (<100nm) AlN nucleation layer on the pre-synthesized SiC/Si substrate. The combination of thin AlN layer and of the layer of SiC on the surface of the pores (Fig.4) results in substantially complete elastic relaxation in thick (over 2 µm) GaN layers.

![Fig. 4. SEM image of the light-emitting diode structure on SiC/Si template received by the method of solid state epitaxy.](image)

In Ref. [15] the LED structure grown by MOCVD on the basis of III-nitrides on SiC nano-layer (50-200nm) formed by the solid-phase epitaxy on Si(111) substrate has been reported. Firstly, the nucleating layer of AlN was deposited on SiC/Si template. Then the transition layer of Al\(_x\)Ga\(_{1-x}\)N solid solution was grown with Al content varying from \(x=1\) to \(x=0\) along the growth direction. After that the layer of GaN was formed in order to provide the planarity of structure surface. Finally, standard LED structure was grown containing n- and p-type GaN layers and the active region with five InGaN quantum wells separated by barrier GaN layers. To prevent cracks its total thickness was limited by 2.5µm. The density of misfit dislocations of the structure was less than \(10^8\) cm\(^{-2}\) and overall dislocation density \(~8\times10^8\) cm\(^{-2}\). In this case the thickness of continuous AlN layer was about 200 nm. The increase of AlN layer thickness results in the increase of the dislocation density in grown GaN film as compared to the dislocation density in the films of GaN grown on the nucleation layer of AlN. The number of dislocations in GaN films grown by MOCVD on 400 nm AlN layer was \(~5\times10^7\) cm\(^{-2}\) with no further increase with AlN layer thickness [6] Data on the growth of GaN films on SiC/Si by MBE are not accessible at the moment. But in this case one should expect a decrease in dislocation density till \(<10^7\) cm\(^{-2}\). Thus, a combination of SiC layer on Si (grown by the substitution of Si atoms in the substrate) and of a thin nucleation layer of AlN (deposited on SiC surface) can dramatically reduce the dislocation density in the grown layer of GaN. This method allows growing the structures with the dislocation density comparable with these grown by MOCVD and HVPE on sapphire and bulk crystals of SiC (<10^8 cm\(^{-2}\)) [18]. SEM photomicrographs of the working LED heterostructure and its ellipsometric spectrum are presented at Fig.5. The analysis of ellipsometric spectrum reveals that the basis of this structure is a single-crystal GaN layer with a band gap of about 3.4eV, the electroluminescence spectra has a maximum in the wavelength range 430-460nm.
Besides, in Ref. [14] the utilization of SiC layers (grown by the new method) as a buffer foundation for the growth of not strained ZnO films by the magnetron sputtering has been discussed.

Obviously, SiC nano-layers also can be used as a buffer layer for the growth of thick (over 1 μm) SiC films. It will get a possibility to obtain the hexagonal polytypes of the silicon carbide on the silicon. However, this problem is not yet investigated.

6. The growth of semi-polar layers of GaN(2023) on SiC/Si(100) substrates

One of the most important properties of nanoSiC synthesized by the solid-phase epitaxy method is a unique opportunity to obtain the semi-polar layers of hexagonal crystals and, in particular, of the semi-polar epitaxial layers of GaN. Wurtzite gallium nitride crystals possess a significant internal electric field in the direction of “C” axis due to the spontaneous polarization and associated piezoelectric effect. It plays a negative role in the utilization of such layers in light-emitting devices. However, as has been theoretically predicted in Ref. [19], in InGaN/GaN structures there are some directions where the piezoelectric polarization becomes zero. And this property is inherent not only for the nonpolar but also for the semi-polar orientation if the crystal plane is inclined to “C” plane on about 45°. It has stimulated a large number of studies dealing with the GaN growth in so-called nonpolar or semipolar directions.

In Ref. [20] the semipolar GaN has been grown on nanoSiC surface synthesized on Si (210). However, the structural perfection of GaN crystal films remains poor [20]. More perfect semipolar GaN films, e.g. GaN (1101) [20], could be grown by masking the surface of Si(100) and etching it by the etchant providing the different etch rates for the different crystallographic axis. In this case, it is possible to obtain Si(111) planes with an inclination of 55° to Si(100) and to grow GaN (1101) on these planes.

But a longitudinal wedge-shaped hump of silicon carbide can be formed by the method of solid-phase epitaxy [1-4] without any surface masking devices or chemical etching substances. As described in section 4, the method is based on a preliminary introduction of the carbide atoms (from carbon monoxide) into interstitial positions of silicon matrix crystal lattice and on a following transformation of the intermediate complex “dilatation diploes – silicon matrix” into the silicon carbide. According to the thermodynamics, if Si(100) plane will be inclined on 1°-10° from <100> towards <011> direction and then heated above 600°C then Si(100) plane will be covered by steps. Upper and lower planes will be (100) and the steps will be restricted by (011) planes. Due to the silicon lattice “looseness” along
<011> direction (determined by the specific crystallographic lattice structure of the silicon) the molecules of CO will rush along this direction deep into the silicon substrate (perpendicular to the steps). The silicon surface will become saturated with CO and the dipoles “silicon vacancy – carbon atom – silicon matrix” will be formed owing to Si–CO reaction. Since the attraction between the silicon vacancy and the carbon atom in the silicon matrix is maximum along the direction <011>, then Si(011) step becomes SiC(112). The angle between (112) and (100) planes in a cubic crystal with diamond lattice is about 35°. Therefore, a longitudinal wedge-shaped hump of silicon carbide will be formed having a top above the upper step and an inclined face (with 35° slope) extending to the terrace of the underlined step. In general, Si(100) surface will be composed by an array of the parallel steps having a shape of a triangular prism. The prismatic symmetry is typical for both cubic and hexagonal(2023) crystals. That is these surfaces can be used to grow both type of crystals depending on the growth conditions (not on the substrate orientation). It provides completely new opportunities for growth of hexagonal crystals of semipolar orientation. This method has been firstly implemented in Ref. [21] to obtain the semipolar 10µm GaN(2023) and (1120) layers with FWHM_{ωθ}=24 arcmin.

7. Summary

Thus, a new relaxation mechanism of elastic energy due to the formation of dilatation dipoles is put forward. Usage of this mechanism allowed us to originally grow the light-emitting diode epitaxial structure on the silicon substrate and fabricate the operating laboratory model of the light-emitting diode. The employment of this relaxation mechanism also permits us to obtain heteroepitaxial films of wide-bandgap semiconductors (such as SiC, AlN, GaN, AlGaN) on silicon without misfit dislocations and cracks having sufficient quality to fabricate many devices of microelectronics and optoelectronics.

The authors gratefully acknowledge the partial support of this work by the Russian Foundation for Basic Research (grants 11-02-12154-ofi-m and 12-02-00935), and by Program of the Presidium of the Russian Academy of Sciences (No 24 "Fundamentals of Technology of Nanostructures and Nanomaterials").

References

[1] Kukushkin S A, Osipov A V, 2012 Doklady Physics 57, 217
[2] Kukushkin S A, Osipov A V, 2013 J. Appl. Phys. 113, 4909
[3] Kukushkin S A, Osipov A V, 2008 Physics of the Solid State 50 1238
[4] Kukushkin S A, Osipov A V, 2014 Physics of the Solid State, 56, 792
[5] Sorokin L M, Veselov N V, Shcheglov M P, Kalmykov A E, Sitnikova A A, Feoktistov N A, Osipov A V, and Kukushkin S A 2008 Technical Physics Letters 34 992
[6] Kukushkin S A, Lukyanov A V, Osipov A V, and Feoktistov N A, 2014 Tech. Phys. Lett. 40, 36
[7] Kumagai Y, Nagashima T, Murakami H, Takada K, and Koukitu A 2008 Phys. Stat. Sol.(c) 5 1512
[8] Zhao Y, Kita K, Kyuno K, Toriumi A 2009 Appl. Phys. Lett. 94 042901
[9] Liliental-Weber Z, Maltez R.L., Xie J., Morkoc H 2008 J. Cryst. Growth 310 3917
[10] Dadgar A, Hums C, Diez A, Blasing J, Kroste A 2006 J. Cryst. Growth 297 279
[11] Severinoa A, Lockeb C, Anzalonea R, Camardaa M, Pilusoa N, Magnaa A La, Saddowb S E, Abbondanzac G, D’Arrigoca G, Viaa F La 2011 ECS Transactions 35 99
[12] Sorokin L M, Kalmykov A E, Bessolov V N, Feoktistov N A, Osipov A V, Kukushkin S A, and Veselov N V. 2011 Technical Physics Letters 37 326
[13] Kukushkin S A, Osipov A V, Osipova E V, Razumov S V, Kandakov, A V 2011 J. Opt. Technol. 78 440
[14] Osipov A V, Kukushkin S A, Feoktistov N A, Osipova E V, Venugopalb N, Vermab G D, Gupta Bipin Kumar, Mitra Anirban, 2012 *Thin Solid Films* **520** 6836.

[15] Kukushkin S A, Osipov A V, Zhukov S G, Zavarin E E and Lundin W V, et al. 2012 *Tech. Phys. Lett.* **38** 297.

[16] Sorokin L M, Kalmykov A E, Bessolov V N, Feoktistov N A and Osipov A V, et al, 2011 *Tech. Phys. Lett.* **37**, 326.

[17] Ratnikov V V, Kalmykov A E, Mysoedov A V, Kukushkin S A, Osipov A V, and Sorokin L M, 2013 *Technical Physics Letters* **39** 994.

[18] Kyutt R N 2010 *Technical Physics Letters* **36** 690.

[19] Romanov A E, Baker T J, Nakamura S, and Speck J S 2006 *J. Appl. Phys.* **100**, 023522.

[20] Bessolov V, Konenkova E, Shcheglov M, Sharofidinov Sh, Kukushkin S, Osipov A, and Nikolaev V 2013 *Phys. Status Solidi C* **10** 433.

[21] Bessolov V, Konenkova E, Kukushkin S A, Mysoedov A V, Osipov A V, Rodin S N, Shcheglov M P, Feoktistov N A, 2014 *Technical Physics Letters* **40**, 802.