



Vacancy-oxygen defects in p-type Si1xGex

E. N. Sgourou, C. A. Londos, and A. Chroneos

Citation: Journal of Applied Physics **116**, 133502 (2014); doi: 10.1063/1.4896728 View online: http://dx.doi.org/10.1063/1.4896728 View Table of Contents: http://scitation.aip.org/content/aip/journal/jap/116/13?ver=pdfcov Published by the AIP Publishing

Articles you may be interested in

Influence of composition and substrate miscut on the evolution of {105}-terminated in-plane Si1xGex quantum wires on Si(001) APL Mat. **2**, 076102 (2014); 10.1063/1.4886218

Production and evolution of A-centers in n-type Si1xGex J. Appl. Phys. **113**, 113507 (2013); 10.1063/1.4795812

Vacancy-type defects in strained-Si layers deposited on SiGeSi structures probed by using monoenergetic positron beams J. Appl. Phys. **97**, 023532 (2005); 10.1063/1.1830086

Diffusion of single quantum well Si 1x Ge x / Si layers under vacancy supersaturation J. Appl. Phys. **89**, 2904 (2001); 10.1063/1.1341208

Electronic defect levels in relaxed, epitaxial p-type Si 1x Ge x layers produced by MeV proton irradiation J. Appl. Phys. **81**, 1180 (1997); 10.1063/1.365554

AIP Chaos

CALL FOR APPLICANTS Seeking new Editor-in-Chief



Vacancy-oxygen defects in *p*-type Si_{1-x}Ge_x

E. N. Sgourou,¹ C. A. Londos,^{1,a)} and A. Chroneos^{2,3,a)}

¹Solid State Physics Section, University of Athens, Panepistimiopolis Zografos, Athens 157 84, Greece ²Faculty of Engineering and Computing, Coventry University, 3 Gulson Street, Coventry CV1 2JH, United Kingdom

³Department of Materials, Imperial College London, London SW7 2AZ, United Kingdom

(Received 3 August 2014; accepted 17 September 2014; published online 1 October 2014)

Oxygen-vacancy defects and, in particular, the VO pairs (known as A-centers) are common defects in silicon (Si) with a deleterious impact upon its properties. Although oxygen-vacancy defects have been extensively studied in Si there is far less information about their properties in *p*-type doped silicon germanium (Si_{1-x}Ge_x). Here, we use Fourier transform infrared spectroscopy to determine the production and evolution of oxygen-vacancy defects in *p*-type Si_{1-x}Ge_x. It was determined that the increase of Ge content affects the production and the annealing behavior of the VO defect as well as its conversion to the VO₂ defect. In particular, both the VO production and the VO annealing temperature are reduced with the increase of Ge. The conversion ratio $[VO_2]/[VO]$ also decreases with the increase of x, although the ratios $[VO_3]/[VO_2]$ and $[VO_4]/[VO_3]$ show a tendency to increase for larger Ge contents. The results are discussed in view of recent experimental and theoretical studies in Si and Si_{1-x}Ge_x. © 2014 AIP Publishing LLC. [http://dx.doi.org/10.1063/1.4896728]

I. INTRODUCTION

In the past few years, the requirement to substitute Si with higher carrier mobility substrates has led to the consideration of $Si_{1-x}Ge_x$ and germanium (Ge).¹⁻⁵ An enabling factor is the advent of high dielectric constant (high-k) dielectrics that allows the departure from the prerequisite to use native oxides such as SiO₂ in Si-based devices.⁶⁻⁸ $Si_{1-x}Ge_x$ is a random alloy with the diamond structure that has one lattice site that can be occupied by either Si or Ge. The variation of its properties (such as band gap, electron and hole mobility, and lattice parameter) with respect to the Ge content is particularly important for microelectronic and optoelectronic devices.^{9,10} Interestingly, in previous experimental and theoretical studies it was shown than the structure and defect processes do not exhibit a Vegard law type behavior.^{11,12} That is they cannot be determined by a simple linear interpolation of the respective properties of Si and Ge, which are the end members of $Si_{1-x}Ge_x$. The deviations are typically larger for intermediate Ge contents, however, they can appear to be non-existent for very small or very large Ge contents. Importantly, vacancies in $Si_{1-x}Ge_x$ preferentially reside near at least one Ge atom as was determined by the positron annihilation spectroscopy (PAS) study of Sihto et al.¹³ This has also been supported by recent density functional theory (DFT) studies.^{11,14}

The defect processes of oxygen-related defects in $Si_{1-x}Ge_x$ alloys are not understood or studied to the level they are in Si.^{15–17} The A-center is composed of a V and an O interstitial (O_i) atom therefore the energetics of its formation will be the balance of these two constituent point defects. Previous investigations have revealed that O_i

preferentially bonds with two Si atoms (and is repelled from Ge atoms), whereas V are attracted to Ge atoms.^{18–20} In Si_{1-x}Ge_x, the Ge content has an impact upon the diffusion of oxygen as well as the conversion of the VO to VO₂ clusters.^{21–23} The aim of the present contribution is to study, using Fourier transform infrared spectroscopy (FTIR), the importance of Ge content on the production, evolution, and annealing of oxygen-vacancy defects in *p*-type Si_{1-x}Ge_x (x = 0.015, 0.028, 0.056).

II. EXPERIMENTAL METHODOLOGY

The samples were cut from Czochralski grown *p*-type $Si_{1-x}Ge_x$ boron (B) doped wafers, acquired by the Institute of crystal growth in Berlin. The concentrations of B and Ge of the wafers were determined by the provider. The Ge concentration of the crystals varied from 1.5%-5.6%. B was introduced during growth at a concentration of 10^{15} cm⁻³.^{24,25} The oxygen and carbon concentrations of the samples were in the range of $(5-11) \times 10^{17} \text{ cm}^{-3}$ and (2-5) $\times 10^{16}$ cm⁻³, respectively. The samples were irradiated with 2 MeV electrons at a temperature of 95 °C and a fluence of 5 $\times 10^{17}$ cm⁻² at the Takasaki Jaeri Dynamitron Accelerator. Thereafter, the samples were subjected to 20 min isochronal anneals from 50 to 550 °C with steps of 10 °C. Following every annealing step, the IR spectra were recorded at room temperature using an FTIR spectrometer. The expected percentage of errors in estimating the absorption coefficients of the defects is less than 3%.

III. RESULTS

Figure 1 shows a characteristic segment of IR spectra of the *p*-type $Si_{1-x}Ge_x$ (x = 0.015, 0.028, 0.056) samples after irradiation and at selective temperatures (50 °C, 400 °C, 500 °C, and 560 °C) in the course of the 20 min isochronal

a)Electronic addresses: hlontos@phys.uoa.gr and alexander.chroneos@ imperial.ac.uk







FIG. 1. Segments of the IR spectra of the *p*-type Si_{1-x}Ge_x (x = 0.015 (a), 0.028 (b), and 0.056 (c)) samples after irradiation and at 50 °C, 400 °C, 500 °C, and 560 °C in the course of the 20 min isochronal anneals sequence.

anneals sequence. Immediately after irradiation, we can see the IR bands of the VO (830 cm^{-1}) peak and at higher temperatures the VO₂ (888 cm^{-1}), VO₃ (904, 967, 1000 cm⁻¹), and VO₄ (985, 1010 cm⁻¹) peaks. Notably, for x = 0.056 the signals of VO₄ bands are very weak (refer to Fig. 1(c)). In the latter figure the bands of C_iO_i (862 cm^{-1}) and C_iO_i(Si₁) (1020 cm^{-1}) are present. These defects are not in the interest of the present study.

Figure 2 presents the thermal evolution of the VO (830 cm^{-1}) , the VO₂ (888 cm^{-1}) , the VO₃ (904, 967, 967, 967, 967)



FIG. 2. The thermal evolution of the VO (830 cm^{-1}) , the VO₂ (888 cm^{-1}) , the VO₃ $(904, 967, 1000 \text{ cm}^{-1})$, and the VO₄ $(888, 1010 \text{ cm}^{-1})$ bands of the *p*-type Si_{1-x}Ge_x (x = 0.015 (a), 0.028 (b), and 0.056 (c)) samples, in the course of the 20 min isochronal anneals sequence.

 1000 cm^{-1}), and the VO_4 (985, 1010 cm^{-1}) defects of the *p*-type Si_{1-x}Ge_x (x = 0.015, 0.028, 0.056) samples. We note that in the x = 0.056 sample the two bands of the VO_4 defect are very weak and only the evolution of the 985 cm⁻¹ band can be shown.

Figure 3 shows the thermal evolution of the VO (830 cm^{-1}) and the VO₂ (888 cm^{-1}) bands for the *p*-type



FIG. 3. Comparison of the thermal evolution curves of the VO (830 cm^{-1}) and the VO₂ (888 cm^{-1}) bands between the x = 0.015, 0.028, and 0.056 samples, in the course of the 20 min isochronal anneal sequence.

Si_{1-x}Ge_x (x = 0.015, 0.028, 0.056) samples in a collective manner for comparison purposes. It is observed that the VO amplitudes for the x = 0.028 and 0.056 samples are almost equal but lower compared to x = 0.015. Additionally, the amplitudes of the VO₂ bands are decreasing with increasing x. We also observe that the annealing of VO defect occurs at around 180 °C.

By inspection of Fig. 1, it is observed that the full-width half maximum (FWHM) of the VO band increases with the increasing x. This is an indication of additional contributions to the shape of the VO band and necessitates Lorentzian profiling to investigate the contributing structures. Figure 4 depicts the Lorentzian deconvolution of the frequency range of the A-center band in the spectra for the x = 0.028 and 0.056 samples, respectively. Regarding the x = 0.056 sample (refer to Fig. 4(b)), besides the VO band at 829 cm⁻¹, two bands at 834 and 841 cm⁻¹ could be detected indicating the presence of additional structures in the frequency range around the VO band. Notably the 841 cm⁻¹ band appears at



FIG. 4. Deconvolution of the A-center region IR bands into Lorentzian profiles at selected temperatures of the 20 min isochronal anneals sequence for the x = 0.028 and 0.056 samples. Solid lines represent the experimental results; dotted lines the Lorentzian profiles and dashed lines the best fitting curves.



FIG. 5. The evolution of the VO (829 cm^{-1}) , $(VO-Ge)_1 (841 \text{ cm}^{-1})$, and $(VO-Ge)_2 (834 \text{ cm}^{-1})$ bands for the x = 0.028 and 0.056 samples. The evolution of the 822 cm^{-1} band is also shown.

temperatures ~ 220 °C for this sample, although for the case of the x = 0.028 sample it appears at even higher T. Additionally, a band at 822 cm⁻¹ is present in both samples.

Figure 5 shows the evolution of the 829, 834, and 841 cm^{-1} bands for the x = 0.028 and 0.056. The VO band begins to decay around 180 °C accompanied initially by the enhancement of the 834 and 841 cm⁻¹ bands. Finally, as the annealing temperature increases further, all the contributing bands decay around 250 °C disappearing from the spectra at 350 °C (refer to Fig. 5(b)). Simultaneously with their decay another band at 888 cm⁻¹ attributed to the VO₂ defect emerges.

Figure 6 shows the a_{VO} with respect to the Ge concentration as derived from the Lorentzian profiles. Clearly, the VO amplitude decreases when the x increases.

A thorough inspection of the received spectra revealed that the shape of the 888 cm^{-1} band is disturbed, a clear indication of the existence of additional components. Apparently, these bands originate from alternative structures which give rise to IR signals in the same spectral range. Figure 7 shows the Lorentzian deconvolution of VO_2 band. At least two additional components at 892 and 896 cm⁻¹



FIG. 6. The absorption coefficient of VO (a_{VO}) with respect to the Ge concentration in *p*-type Si_{1-x}Ge_x.

seem to contribute in the shape of the spectral range of the VO_2 defect.

Figure 8 represents the dependence of the ratio of the absorption coefficients of VO_{n+1} over VO_n with respect to the Ge concentration for the *p*-type $Si_{1-x}Ge_x$ samples. It is observed that the a_{VO2}/a_{VO} decreases with the increase of x, whereas the a_{VO3}/a_{VO2} and a_{VO4}/a_{VO3} ratios show the opposite trend.

IV. DISCUSSION

Lindström and Svensson¹⁵ have established that a large percentage of A-centers in Si are transformed into VO2 clusters. However, in *n*-type $Si_{1-x}Ge_x$ this fraction decreases²³ as the Ge-content increases. This is observed also here for the *p*-type $Si_{1-x}Ge_x$. It has been reported^{17,26} that with the increase of the Ge content of the Si1-xGex material the amplitude of the VO band decreases although the FWHM of the band increases. The increase of FWHM has been attributed to additional contributions to the shape of the VO band originating from (VO-Ge) structures with the Ge atoms in adjustment positions to a VO core. In particular, Lorentzian profiling 17,23,26 of the band has attributed two bands at ~ 834 and 839 cm^{-1} to a (VO-Ge)₁ and (VO-Ge)₂ structures with the Ge atom in the first and second neighbour position in relation with VO, correspondingly. On the other hand, a reduction in the amplitude of the VO band was also



FIG. 7. Deconvolution of the VO_2 region IR bands into Lorentzian profiles at a representative temperature of the 20 min isochronal anneals sequence for the x = 0.028 sample. Solid lines represent the experimental results; dotted lines the Lorentzian profiles and dashed lines the best fitting curves.



FIG. 8. The conversion ratio of the a_{VO2}/a_{VO} , a_{VO3}/a_{VO2} , and a_{VO4}/a_{VO3} with respect to the Ge concentration.

verified^{21,22,27} by deep level transient spectroscopy (DLTS) and Laplace DLTS spectroscopy studies. In our case of *p*-type Si_{1-x}Ge_x material the deconvolution procedure has shown (refer to Fig. 4) that besides the above two bands at 834 and 841 cm⁻¹, another band at 822 cm⁻¹ is also present. The latter band may be correlated with the V_2O_2 defect.²⁸

The observed decrease of VO concentration (refer to Fig. 6) was attributed 17,26 to the ability of Ge atoms to act as annihilation centers for primary defects, that is vacancies and self-interstitials (I), thus reducing the availability of vacancies to be trapped by oxygen atoms. In other words, it was suggested that as a result of the Ge presence, the annihilation rate of the components of Frenkel pairs produced by irradiation was increased, resulting in a decrease in the production of oxygen-vacancy pairs. Regarding low temperature irradiations (T < 100 K) of Ge-doped Si ([Ge] < 1 at%) and $Si_{1-x}Ge_x$ ([Ge] up to 15 at. %), it has been determined¹⁷ that the process of formation of radiation defects is similar in both cases. The formed GeV pairs anneal out at about 280 K and then VO centers emerge in the spectra. Thus, at irradiations performed at room temperatures the GeV pairs are unstable and are not expected to form. The experiments suggest^{17,26} that in both cases the suppressive formation of the VO defect can be attributed to the Ge atoms being the centers of annihilation of primary defects. In that case, the number of vacancies able to be trapped by oxygen atoms is smaller, leading to a decrease in the formation of VO defects.

For both *n*- and *p*-type Si_{1-x}Ge_x the *V*O pair is stable up to ~180 °C, whereas in *n*- and *p*-type Si it is stable up to about ~250 °C (refer to Fig. 2 and to Ref. 23 for *n*-type Si_{1-x}Ge_x). The common behavior exhibited suggests that the lowering of the annealing temperature of the *V*O pair can be attributed to the host materials and not the *p*-type doping. It has been previously established by Kube *et al.*¹² (and references therein) that the increase of the Ge concentration of the Si_{1-x}Ge_x alloys leads to the lowering of the activation enthalpy of diffusion of dopants migrating via vacancy-related mechanisms. In essence, the introduction of the Ge leads to the formation of Si-Ge and Ge-Ge bonds (the latter is more populous at high Ge concentrations) that are weaker compared to the Si-Si bonds. Also the Si-Si bond lengths in Si_{1-x}Ge_x are increased due to the presence of the Ge atoms.

These effects lower the annealing temperature of the VO pairs in $Si_{1-x}Ge_x$ as compared to Si. An additional argumentation on the lowering of the annealing temperature based on the reactions that participate in the annealing of VO is presented below.

The observed decrease in the VO_2 formation (refer to Fig. 3) as the Ge content of the $Si_{1-x}Ge_x$ material increases could be attributed to the following scenario:

- (i) Due to the formation of (VO-Ge) structures less VO are available²⁶ to be transformed in VO₂ centers and this is manifested by the reduction of the 888 cm^{-1} band amplitude in the spectra.
- The annealing of VO defects is mainly governed by (ii) two reactions: $VO + O_i \rightarrow VO_2$ and $VO + I \rightarrow O_i$. It has been suggested²⁹ that the strain fields induced in the $Si_{1-x}Ge_x$ material due to the presence of the larger Ge atoms affect the availability of self-interstitials. In effect this changes the balance between the above two reactions and their contribution in the VO annealing. Notably, for Ge-doped Si (Ref. 29) it was argued that the larger the Ge content, the stronger and more extensive the strain fields in the material. As a result, the binding energy of self-interstitials in the large self-interstitial clusters³⁰ (which are the source of self-interstitials) is reduced leading to an increase of the number of available self-interstitials with the increase of Ge. Thus the contribution of the reaction $VO + I \rightarrow O_i$ is enhanced and that of the reaction: $VO + O_i \rightarrow VO_2$ is reduced. In essence, the whole process leads to a reduction of the VO₂ formation as well as earlier lower annealing temperature of VO pairs due to the earlier activation of the $VO + I \rightarrow O_i$ reaction. Indeed, in the $Si_{1-x}Ge_x$ material VO signal begins to decay from the spectra at a lower temperature than in Si.
- In another approach, one could consider the ability of (iii) Ge atoms to capture self-interstitials besides vacancies. In a previous report³¹ regarding Ge-doped Si it was suggested that Ge atoms compete with C atoms in capturing self-interstitials. This competition becomes significant when [Ge]/[C] is larger than 1000. We extend this conclusion for $Si_{1-x}Ge_x$ and in our case where the [Ge]/[C] ratio is much larger than three orders of magnitude this competition is stronger. As a result Ge_i form (Ge + Si_i \rightarrow Ge_i), which being mobile are immediately trapped by oxygen atoms to form GeO_i pairs. Notably, the formation of such centers has been previously considered in the literature.^{32,33} DFT results indicate that Ge repels O_i.³² IR signals from such a center were not reported. This does not mean that the defect is not existent. Its signal may be very weak or the defect is IR inactive. These centers may dissociate above 200°C producing Gei (via the reaction $Ge_iO \rightarrow Ge_i + O_i$). Then the Ge_i atoms immediately convert to Ge atoms by releasing self-interstitials, which in turn react with VO centers $(VO + I \rightarrow O_i)$. Thus, less VO centers are available for the reaction $VO + O_i \rightarrow VO_2$ and therefore less

 VO_2 centers are formed. The annealing of VO at a lower temperature is also explained in this scenario by the same reasoning as in scenario (ii).

As we succinctly mentioned in the results section the shape of the VO_2 band is disturbed and Lorentzian profiling has shown the existence of two additional contributing bands at 892 and 896 cm⁻¹. These bands are similar to those observed in our previous work²³ in *n*-type Si_{1-x}Ge_x. They were attributed to $(VO_2$ -Ge)₁ and $(VO_2$ -Ge)₂ defects originating from structures with the Ge atoms in the first and the second neighbour position in relation with a VO_2 core.

Considering the formation of the VO_3 and VO_4 clusters, the species that needs to migrate is O_i (note that in the case of VO_2 formation, VO is the migrating species) and it was previously determined that Ge enhances the O_i diffusivity.³⁴ In addition, previous calculations³⁵ have deduced that the strain fields due to the larger Ge atoms lead to a reduction of the potential barrier height, thus enhancing the diffusion of oxygen. Therefore, the present results (refer to Fig. 8) are consistent as they represent an increase of the ratio of the absorption coefficients of VO_{n+1} over VO_n (n = 2, 3) that translates to a higher percentage of VO_3 and VO_4 clusters forming as the Ge content is increased.

V. SUMMARY

In the present study, we used FTIR spectroscopy to investigate the production and evolution of A-centers in *p*-type $Si_{1-x}Ge_x$. It was verified that the increase of Ge content results (i) in the decrease of VO production, (ii) in the decrease of VO annealing temperature, and (iii) in a decrease of the a_{VO2}/a_{VO} ratio. Additionally, Lorentzian profiling has detected bands related to (VO-Ge) structures as in the case of *n*-type $Si_{1-x}Ge_x$. Furthermore, the shape of the VO₂ band was found to be disturbed by components related to the (VO₂-Ge) structures. These issues will need to be systematically investigated from different perspectives (advanced computational modeling and further experiments over a wider range of Ge concentrations). The present study is part of a concerted effort to unravel the defect processes of group IV random alloys.

ACKNOWLEDGMENTS

The authors thank Professor Abrosimov for the provision of the samples.

- ¹R. Roucka, J. Tolle, C. Cook, A. V. G. Chizmeshya, J. Kouvetakis, V. D'Costa, J. Menendez, and Z. D. Chen, Appl. Phys. Lett. 86, 191912 (2005).
- ²R. Kube, H. Bracht, A. Chroneos, M. Posselt, and B. Schmidt, J. Appl. Phys. 106, 063534 (2009).
- ³A. Chroneos, H. Bracht, R. W. Grimes, and B. P. Uberuaga, Appl. Phys. Lett. **92**, 172103 (2008).

- ⁴E. Hüger, U. Tietze, D. Lott, H. Bracht, D. Bougeard, E. E. Haller, and H. Schmidt, Appl. Phys. Lett. **93**, 162104 (2008); A. Chroneos, J. Appl. Phys. **105**, 056101 (2009).
- ⁵S. Kilpeläinen, K. Kuitunen, F. Tuomisto, J. Slotte, H. H. Radamson, and A. Yu. Kuznetsov, Phys. Rev. B 81, 132103 (2010).
- ⁶H. Kim, C. Chui, K. C. Saraswat, and P. C. McIntyre, Appl. Phys. Lett. 83, 2647 (2003).
- ⁷K. Kita, K. Kyuno, and A. Toriumi, Appl. Phys. Lett. 85, 52 (2004).
- ⁸A. Dimoulas, G. Mavrou, G. Vellianitis, E. Evangelou, N. Boukos, M. Houssa, and M. Caymax, Appl. Phys. Lett. 86, 032908 (2005).
- ⁹E. Kasper, Curr. Opin. Solid State Mater. Sci. 2, 48 (1997).
- ¹⁰H. G. Grimeiss, Semiconductors **33**, 939 (1999).
- ¹¹A. Chroneos, H. Bracht, C. Jiang, B. P. Uberuaga, and R. W. Grimes, Phys. Rev. B 78, 195201 (2008).
- ¹²R. Kube, H. Bracht, J. Lundsgaard Hansen, A. Nylandsted Larsen, E. E. Haller, S. Paul, and W. Lerch, J. Appl. Phys. **107**, 073520 (2010).
- ¹³S. L. Sihto, J. Slotte, J. Lento, K. Saarinen, E. V. Monakhov, A. Yu. Kuznetsov, and B. G. Svensson, Phys. Rev. B 68, 115307 (2003).
- ¹⁴A. Chroneos, R. W. Grimes, and H. Bracht, J. Appl. Phys. **105**, 016102 (2009).
- ¹⁵A. Chroneos, H. Bracht, R. W. Grimes, and B. P. Uberuaga, Mater. Sci. Eng. B 154–155, 72 (2008); E. N. Sgourou, D. Timerkaeva, C. A. Londos, D. Aliprantis, A. Chroneos, D. Caliste, and P. Pochet, J. Appl. Phys. 113, 113506 (2013).
- ¹⁶E. Hild, P. Gaworzewski, M. Franz, and K. Pressel, Appl. Phys. Lett. 72, 1362 (1998).
- ¹⁷Y. V. Pomozov, M. G. Sosnin, L. I. Khirunenko, V. I. Yashnik, N. V. Abrosimov, W. Schröder, and M. Höhne, Semiconductors **34**, 989 (2000); L. I. Khirunenko, V. I. Shakhovtsov, V. K. Shinkarenko, L. I. Shpinar, and I. I Yaskovets, Sov. Phys. Semicond. **21**, 345 (1987).
- ¹⁸I. Yonenaga, M. Nonaka, and N. Fukata, Physica B **308–310**, 539 (2001).
- ¹⁹S. Hao, L. Kantorovich, and G. Davies, Phys. Rev. B 69, 155204 (2004).
- ²⁰A. Chroneos, E. N. Sgourou, and C. A. Londos, J. Mater. Sci.: Mater. Electron. 24, 2772 (2013).
- ²¹V. P. Markevich, A. R. Peaker, L. I. Murin, and N. V. Abrosimov, Appl. Phys. Lett. 82, 2652 (2003).
- ²²V. P. Markevich, A. R. Peaker, J. Coutinho, R. Jones, V. J. B. Torres, S. Öberg, P. R. Briddon, L. I. Murin, L. Dobaczewski, and N. V. Abrosimov, Phys. Rev. B 69, 125218 (2004).
- ²³E. N. Sgourou, A. Andrianakis, C. A. Londos, and A. Chroneos, J. Appl. Phys. 113, 113507 (2013).
- ²⁴N. V. Abrosimov, S. N. Rossolenko, V. Alex, A. Gerhardt, and W. Schröder, J. Cryst. Growth **166**, 657 (1996).
- ²⁵N. V. Abrosimov, S. N. Rossolenko, W. Thieme, A. Gerhardt, and W. Schröder, J. Cryst. Growth **174**, 182 (1997).
- ²⁶L. I. Khirunenko, Yu. V. Pomozov, M. G. Sosnin, N. V. Abrosimov, M. Hohne, and W. Schroder, Solid State Phenom. 69–70, 209 (1999).
- ²⁷V. P. Markevich, A. R. Peaker, L. I. Murin, and N. V. Abrosimov, J. Phys.: Condens. Matter 15, S2835 (2003).
- ²⁸N. V. Sarlis, C. A. Londos, and L. G. Fytros, J. Appl. Phys. 81, 1645 (1997).
- ²⁹C. A. Londos, A. Andrianakis, V. V. Emtsev, G. A. Oganesyan, and H. Ohyama, Physica B 404, 4693 (2009).
- ³⁰R. C. Newman and R. Jones, in *Semiconductors and Semimetals*, edited by F. Shimura (Academic Press, New York, 1994), Vol. 42, p. 289.
- ³¹V. V. Voronkov, R. Falster, C. A. Londos, E. N. Sgourou, A. Andrianakis, and H. Ohyama, J. Appl. Phys. **110**, 093510 (2011).
- ³²A. Chroneos and C. A. Londos, J. Appl. Phys. **107**, 093518 (2010).
- ³³X. Yu, J. Chen, X. Ma, and D. Yang, Mater. Sci. Eng., R 74, 1 (2013).
- ³⁴A. K. Tipping, R. C. Newman, D. C. Neuton, and J. H. Tucker, Mater. Sci. Forum **10–12**, 887 (1986).
- ³⁵L. I. Khirunenko, Yu. V. Pomozov, V. I. Shakhovtsov, and V. V. Shumov, Mater. Sci. Forum **258–263**, 1773 (1997).