Band structure properties of novel BxGa1-xP alloys for silicon integration

Nadir Hossain, T. J. C. Hosea, Stephen J. Sweeney, Sven Liebich, Martin Zimprich et al.

Citation: *J. Appl. Phys.* **110**, 063101 (2011); doi: 10.1063/1.3630018

View online: <http://dx.doi.org/10.1063/1.3630018>

View Table of Contents: <http://jap.aip.org/resource/1/JAPIAU/v110/i6>

Published by the [American Institute of Physics](#).

Related Articles

Cu-doping of ZnO by nuclear transmutation

Appl. Phys. Lett. **99**, 202109 (2011)

Microstructure, optical property, and electronic band structure of cuprous oxide thin films

J. Appl. Phys. **110**, 103503 (2011)

Electronic structure and optical properties of Ag₃PO₄ photocatalyst calculated by hybrid density functional method

Appl. Phys. Lett. **99**, 191903 (2011)

Dilute-nitride GaInAsN/GaAs site-controlled pyramidal quantum dots

Appl. Phys. Lett. **99**, 181113 (2011)

Modifications in structural and electronic properties of TiO₂ thin films using swift heavy ion irradiation

J. Appl. Phys. **110**, 083718 (2011)

Additional information on J. Appl. Phys.

Journal Homepage: <http://jap.aip.org/>

Journal Information: http://jap.aip.org/about/about_the_journal

Top downloads: http://jap.aip.org/features/most_downloaded

Information for Authors: <http://jap.aip.org/authors>

ADVERTISEMENT

AIPAdvances

Submit Now

**Explore AIP's new
open-access journal**

- **Article-level metrics
now available**
- **Join the conversation!
Rate & comment on articles**

Band structure properties of novel $B_xGa_{1-x}P$ alloys for silicon integration

Nadir Hossain,¹ T. J. C. Hosea,^{1,a)} Stephen J. Sweeney,^{1,b)} Sven Liebich,² Martin Zimprich,² Kerstin Volz,² Bernardette Kunert,³ and Wolfgang Stolz²

¹*Advanced Technology Institute and Department of Physics, University of Surrey, Guildford, Surrey GU2 7XH, United Kingdom*

²*Material Sciences Center and Faculty of Physics, Philipps-University, Marburg 35032, Germany*

³*NAsP III/V GmbH, Am Knechtacker 19, Marburg 35041, Germany*

(Received 12 April 2011; accepted 24 July 2011; published online 16 September 2011)

We have grown and investigated the band-structure properties of novel III-V alloys based upon $B_xGa_{1-x}P$. These layers are utilized as strain-compensating layers for the lattice-matched integration of novel direct bandgap Ga(NAsP) quantum well lasers on silicon. Experimental and theoretical studies reveal the dependence of the direct and indirect band gaps for strained $B_xGa_{1-x}P$ layers grown on silicon as a function of Boron composition from which we derive the properties of free-standing $B_xGa_{1-x}P$. For Boron fractions up to 6%, we find that the bowing parameter for the lowest (indirect) band gap is -6.2 ± 0.2 eV. High crystalline quality and promising optical material properties are demonstrated and applied to monolithically integrated Ga(NAsP)/(BGa)P multi-quantum well heterostructures on (001) silicon substrates. Our results show that novel (BGa)P layers are suitable for strain compensation purposes, which pave the way towards a commercial solution for the monolithic integration of long term stable laser diodes on silicon substrates. © 2011 American Institute of Physics. [doi:10.1063/1.3630018]

Silicon photonics has been a very active research area in recent years, with the ultimate goal of this rapidly emerging field being the integration of photonic and electronic devices.¹ The key challenge using silicon as a monolithic integration platform is the lack of a silicon-based laser and difficulties associated with integrating optically efficient III-V based laser sources on silicon substrates.² The large lattice mismatch between conventional III/V laser materials like GaAs or InP to silicon leads unavoidably to the formation of high densities of threading dislocations in the integrated III/V layer during the epitaxial deposition of laser material on silicon substrates, which is problematic for long-term lasing operation of the corresponding device structure. The novel quaternary dilute nitride Ga(NAsP)/GaP has been proven to be a promising laser material in recent years due to its direct electronic bandgap,³ efficient optical gain and lasing,^{4,5} and closely matched lattice constant to the silicon substrate. The epitaxial growth of the active material Ga(NAsP) on exactly oriented (001) silicon (Si) requires the deposition of a defect-free GaP nucleation layer beforehand. In order to prevent any defect formation in this nucleation layer, it is essential to guarantee charge neutrality along the III/V-silicon interface and avoid unwanted cross-doping caused by atom diffusion into the respective heterolayers as well as the formation of anti-phase domains.⁶ On the other hand, post-growth annealing procedures under high temperatures are needed in this laser material to improve its structural and optical properties,⁷ as in other dilute nitride laser materials.⁸ The small lattice mismatch between GaP and silicon becomes a crucial aspect in the deposition of thicker III/V

materials. The incorporation of small percentages of boron (B) or nitrogen (N) into GaP ensures the adjustment of the III/V lattice constant towards that of silicon.⁹ The incorporation of B into GaP using metal organic vapour phase epitaxy (MOVPE) can be easily controlled upon the gas phase ration of the applied precursors. In addition, the thick grown layers of (BGa)P on silicon show high crystalline quality. For this reason, we choose to incorporate boron to provide the lattice-matched deposition of thick III/V cladding layers and for the strain compensation of the compressively strained Ga(NAsP) QW material. However, (BGa)P is a little known material system with little published experimental data on its optical characteristics.¹⁰ In this work, we have grown novel $B_xGa_{1-x}P$ layers on silicon substrates and investigated their band-structure properties using surface photo-voltage (SPV) spectroscopy. Furthermore, to verify the suitability of this novel material for silicon integration, we have demonstrated the high crystal quality and promising optical material properties from high temperature annealed lattice matched monolithically integrated Ga(NAsP)/(BGa)P multi-quantum well heterostructures (MQWHs) on a (001) silicon substrate, where a novel (BGa)P layer is used for strain compensation purposes.

The epitaxial growth of $B_xGa_{1-x}P$ epi-layers and Ga(NAsP)/(BGa)P MQWHs was realised in a commercial Aixtron Aix-200-GFR-reactor system at a reduced pressure of 50 mbar using Pd-purified H_2 as carrier gas on (001) silicon substrate by MOVPE. The silicon substrate underwent a wet chemical etching procedure followed by a high-temperature treatment in the metalorganic chemical vapour decomposition reactor in order to get an oxide-free surface.¹¹ All wafers are nominally exact despite they have a small offset of $(0.1-0.2)^\circ$ in one of the $\langle 110 \rangle$ directions of the silicon crystal. “As-polished” wafers commonly have a mono-atomic stepped

^{a)}Present address: Ibnu Sina Institute for Fundamental Science Studies, Universiti Teknologi Malaysia, Johor Bahru 81310, Malaysia.

^{b)}Electronic mail: s.sweeney@surrey.ac.uk.

silicon surface. To reduce anti-phase domains in the overgrown GaP, a double-stepped silicon surface is necessary which is easier to achieve if a small offset is introduced. Nevertheless, exactly orientated silicon (001) substrates for CMOS process have an error bar of 0.5° . With our offset of $\sim 0.1^\circ$, we are within these specifications. More details, in particular, silicon surface preparation to reduce the anti-phase domain in these materials can be found elsewhere.^{12,13} Due to the metastable character of the Ga(NAsP) and (BGa)P specific low temperature MOVPE growth conditions at 575°C using the more efficiently decomposing group-V-sources tertiarybutyl arsine (TBAs), tertiarybutyl phosphine (TBP), and 1,1-dimethyl hydrazine (UDMH) in combination with the group-III-source triethyl gallium (TEGa) and triethyl boron (TEB) were applied. After the decomposition of the MQWHs, the post-growth annealing procedures under high temperatures and TBP stabilisation were undertaken. For structural investigations, high-resolution x-ray diffraction (HR XRD) measurements around the (004) reflection of Si using a X'Pert Pro MRD diffractometer from Panalytical and atomic force microscopy (AFM) imaging in tapping mode were performed. More details, in particular, about the MOVPE growth of these devices can be found elsewhere.⁹

We first discuss the surface photo-voltage results on $\text{B}_x\text{Ga}_{1-x}\text{P}$ epi-layers. These were of various boron concentrations and layer thicknesses, as shown in Table I. The boron content in $\text{B}_x\text{Ga}_{1-x}\text{P}$ epi-layers was determined by HR XRD measurements and using Vegards law. SPV has been proven to be a versatile spectroscopic characterisation technique and provides the means to reveal material band-structure information at room temperature (RT), without the need for cryogenics as is often necessary with other conventional spectroscopic measurements.^{14,15} SPV can be thought of as a type of absorption spectroscopy; for photons incident on the sample of energy $h\nu$, greater than the bandgap energy E_g , electrons and holes are generated which separate in the internal field and generate a measurable open-circuit voltage. In most semiconductors, there is a large increase in the absorption coefficient, α , near the bandgap energy. Therefore, a significant increase in SPV signal can often be observed at, and above, approximately E_g . The “raw” SPV spectrum after division by the system response yields a spectrum that is essentially proportional to the sample absorption spectrum $\alpha(\lambda)$. The SPV spectra of $\text{B}_x\text{Ga}_{1-x}\text{P}$ layers were measured in the standard way,¹⁶ in the range 475–875 nm at RT. We did not find a measurable signal below 475 nm. The energy bandgap (E_g) may be reliably obtained using the well known relations^{15,17}

$$\text{SPV} \propto \alpha \propto (h\nu - E_g)^2 \text{ for indirect gap semiconductors,} \quad (1)$$

$$\text{SPV} \propto \alpha \propto \sqrt{h\nu - E_g} \text{ for direct gap semiconductors.} \quad (2)$$

In Fig. 1(a), we show the dependence of $(\alpha)^{1/2}$ versus photon energy to distinguish the part of the spectrum where indirect transitions are dominant. Fig. 1(a) shows that the $(\alpha)^{1/2}$ response has two distinct spectral regions. The first region, between about 1.8 and 2.3 eV, is linear and can be evidently associated with indirect electron transitions in accordance with Eq. (1) above. The photo-threshold of this process is determined from the linear extrapolation and intersection of the fits shown in the figure as ~ 1.91 eV, according to Eq. (1). The second region in Fig. 1(a), above ~ 2.3 eV, where the α response increases abruptly, corresponds to the onset of direct electron transitions from the valence to the conduction band. The probability of this process is significantly larger than that of the indirect process since it does not require phonons to conserve momentum and energy.¹⁷ In order to reveal this more clearly, Fig. 1(b) shows $(\alpha)^2$ as a function of energy, which we would expect from Eq. (2) to show a linear contribution above the corresponding direct bandgap energy. In order to reveal and analyse this region, we have subtracted the linear $(\alpha)^{1/2}$ behaviour due to the indirect process, with the result as shown in Fig. 1(b). A linear fit to these data as shown in the figure reveals the photo-threshold in this case to be ~ 2.31 eV. The results of such analyses for all the samples are shown in Fig. 1(c). At the beginning of the SPV measurements on $\text{B}_x\text{Ga}_{1-x}\text{P}$ samples, the SPV of a pure GaP substrate was measured as a reference. The measured indirect bandgap of pure GaP was found to be ~ 2.23 eV, which is consistent with literature.¹⁵ Note that although it was possible to determine the indirect bandgap from SPV for all samples studied, the SPV signals could not provide unambiguous information on the direct bandgap for the two lowest-B samples and pure GaP substrate because we did not find a measurable signal below 475 nm. We note that a weak decrease in the direct energy gap with increasing Boron concentration is consistent with ellipsometric studies of the extinction coefficient in BGaP by Rogowsky *et al.*¹⁰

Now because the $\text{B}_x\text{Ga}_{1-x}\text{P}$ layers are grown on silicon, they will be under tensile strain for B concentration higher than 2%.¹⁰ Therefore, part of the observed decrease in the SPV-measured bandgaps with increasing B content in Fig. 1(c) is due to strain. In order to determine the change in bandgap due to B incorporation alone, it is necessary to account for the effect of strain. The change in lattice constant of the $\text{B}_x\text{Ga}_{1-x}\text{P}$ epi-layers can be calculated from the simple relation¹⁸

$$\frac{a_L - a_S}{a_S} = \left[\frac{C_{11}}{C_{11} + 2C_{12}} \cdot \left(\frac{\Delta a}{a} \right)^\perp \right], \quad (3)$$

where a_S and a_L are the unstrained lattice constants of silicon and bulk $\text{B}_x\text{Ga}_{1-x}\text{P}$ respectively, C_{11} and C_{12} the $\text{B}_x\text{Ga}_{1-x}\text{P}$ elastic moduli and $(\Delta a/a)^\perp$ the strained lattice mismatch of

TABLE I. The studied $\text{B}_x\text{Ga}_{1-x}\text{P}$ layers showing the different B concentrations, layer thicknesses, and deduced lattice mismatch, $(\Delta a/a)^\perp$ from XRD.

$\text{B}_x\text{Ga}_{1-x}\text{P}$ layer	Thickness (nm)	$(\Delta a/a)^\perp$ (%)
$\text{B}_{0.022}\text{Ga}_{0.978}\text{P}$	500	−0.04
$\text{B}_{0.032}\text{Ga}_{0.968}\text{P}$	4000	−0.31
$\text{B}_{0.037}\text{Ga}_{0.963}\text{P}$	1000	−0.47
$\text{B}_{0.046}\text{Ga}_{0.954}\text{P}$	250	−0.77
$\text{B}_{0.06}\text{Ga}_{0.94}\text{P}$	250	−1.2

FIG. 1. (Color online) (a) Determination of the indirect bandgap and (b) direct bandgap from the $\alpha(\lambda)$ measurement for B = 6.0%. The lines show fits with Eqs. (1) and (2). (c) The measured direct and indirect bandgaps from SPV, the calculated strain-induced direct and indirect bandgap change (from Eq. (3)), and the finally determined direct and indirect bandgaps with effects of strain removed. (d) The strain-free indirect bandgap deduced from SPV (inset) and a fit with Vegard's law (curve) giving the bowing parameter C , -6.2 ± 0.2 eV.

the $B_xGa_{1-x}P$ epilayer in the growth direction. The latter quantity can be measured by x-ray diffraction (XRD). The deduced results for $(\Delta a/a)^\perp$ are shown in Table I. The relative strain in the in-plane direction for the $B_xGa_{1-x}P$ epilayer can be calculated from the relation¹⁸

$$\varepsilon = \frac{a_L - a_S}{a_L}. \quad (4)$$

The change of E_g due to the effect of tensile strain on the $B_xGa_{1-x}P$ epi-layer can then be determined by the following equation:¹⁸

$$\Delta E_g = \left[-2p \left(\frac{C_{11} - C_{12}}{C_{11}} \right) - q \left(\frac{C_{11} + 2C_{12}}{C_{11}} \right) \right] \varepsilon, \quad (5)$$

where p and q are hydrostatic and shear deformation potentials, respectively. The values of C_{11} , C_{12} , p , and q for $B_xGa_{1-x}P$ are calculated from the corresponding values of GaP and BP (Refs. 19 and 20) and using Vegard's law for ternary alloys.²¹ The lattice constant of BP (4.5383 Å) and GaP (5.45 Å) required for this calculation are taken from lit-

erature.^{20,22} Figure 1(c) shows the resulting calculated change of E_g for $B_xGa_{1-x}P$ due to strain as a function of B concentration, for the direct and indirect bandgaps, respectively. The direct bandgap of GaP (~ 2.74 eV) required for this calculation is taken from literature.^{19,20} By subtracting the calculated strain-induced ΔE_g of $B_xGa_{1-x}P$ (from Eq. (5)) to the corresponding measured E_g (from SPV), we can determine the E_g values of $B_xGa_{1-x}P$ due to the B concentration alone, for direct and indirect bandgaps, respectively, as shown in Fig. 1(c). The resulting indirect bandgaps are reproduced in Fig. 1(d) (inset) where the E_g value of $B_xGa_{1-x}P$ for B = 100% (BP) is taken from literature.^{19,20} Figure 1(d) shows a fit of Vegard's law for ternary alloys to these indirect bandgap results giving a bowing parameter (C) value of -6.2 ± 0.2 eV for the $B_xGa_{1-x}P$ indirect bandgap. These experimental results agree well with recently published theoretical results on $B_xGa_{1-x}P$ material.²³

Now, we turn to the growth of MQW device structures based on Ga(NAsP)/(BGaP)/Si. A $B_{0.021}Ga_{0.979}P$ layer on silicon allows for the deposition of an unstrained III/V layer on silicon at RT. Nevertheless, at the growth temperature (GT) of 575 °C, $B_{0.021}Ga_{0.979}P$ is compressively strained

FIG. 2. (Color online) (a) Ga(NAsP)/(BGa)P/Si MQWHs on silicon. (b) High-resolution XRD pattern of a Ga(NAsP)/(BGa)P/Si MQWHs around the (0 0 4) reflection of Si. (c) AFM image of a Ga(NAsP)/(BGa)P/Si MQWHs. (d) PL spectra at RT of a Ga(NAsP)/(BGa)P/Si MQWHs.

(0.13%) on silicon. The deposition of lattice-matched material at GT requires a B_{0.03}Ga_{0.97}P layer, which leads again to tensile strain in the III/V layer at RT (Ref. 9). To improve the optical properties of this novel Ga(NAsP) material, a post-growth annealing is required at 850 °C (Ref. 7). If the integral strain of the entire III/V material system becomes too high during cooling down, the III/V film may crack. So far, we did not observe any cracking during cooling down after the deposition of 4 μ m B_{0.03}Ga_{0.97}P on silicon,⁹ which is sufficient to realize the first lattice matched monolithically integrated semiconductor lasers on silicon. Figure 2(a) schematically illustrates the design of the Ga(NAsP)/(BGa)P/Si MQWH. The III/V growth starts with a GaP nucleation layer

of approximately 110 nm thickness. This nucleation layer is followed by a 1000-nm-thick B_{0.033}Ga_{0.967}P layer. The MQW structure consists of three Ga(NAsP) QWs embedded in ~64-nm-thick B_{0.046}Ga_{0.954}P barriers. In this structure, the strain compensation of the QWs at growth temperature requires a B-concentration of about 4.6% in the barrier material. Furthermore, every QW was embedded in ~5-nm-thick GaP intermediate layers to avoid the formation of N–B bonds.²⁴ The complete MQW structure is then capped with a 50-nm-thick B_{0.033}Ga_{0.967}P layer. Figure 2(b) compares the XRD pattern of the Ga(NAsP)/(BGa)P/Si-multi-quantum well heterostructures (MQWHs) annealed at 850 °C (upper trace), as grown (middle trace), and a pattern obtained by

modeling using dynamical XRD theory (lower trace). A very good agreement between annealed, as grown, and theory is obtained. The narrow line width of the XRD satellite reflections, which remain constant up to the highest diffraction orders, proves that abrupt interfaces have been realized and underlines the high crystal quality for the annealed sample in the same way as for the as-grown sample. No indication of dislocation formation during epitaxial layer deposition and during annealing is observed as this would lead to a significant broadening of the entire experimental XRD pattern. Besides the appearance of the MQW satellite fringes, the corresponding peaks of the GaP nucleation layer as well as the thick (BGa)P can also be clearly distinguished close to the silicon substrate peak—see arrows in Fig. 2(b). The high crystalline quality is further evidenced by AFM imaging the surface of the MQWH as shown in Fig. 2(c). The surface morphology is flat within a height scale of 2 nm. There are no defects or cracks visible confirming the high quality crystal structure with smooth interfaces in this Ga(NAsP)/(BGa)P/Si MQW heterostructure.

RT PL measurements were carried out to verify the good optical quality of this material grown with the (BGa)P strain compensating separate confinement and barrier layers on silicon substrates. Figure 2(d) shows an RT luminescence spectrum of a Ga(NAsP)/(BGa)P/Si-MQWH annealed at 850 °C (same sample of Fig. 2(b)). The PL spectrum contains a luminescence peak coming from the Si substrate.⁹ A strong PL peak at 1.33 eV (930 nm) originating from the active III/V material is also observed at RT. Taking into account the fact that the QW PL corresponds mainly to the direct photon absorption in the active material, the absolute intensity compared to the substrate luminescence is a promising indication of good optical material quality.

In conclusion, we have grown novel $B_xGa_{1-x}P$ strain compensating layers on Si substrates to provide lattice-matched integration of III-V lasers on Si. The bandgap properties of $B_xGa_{1-x}P$ for $x \leq 6\%$ are investigated using a SPV technique and the bowing parameter for the indirect bandgap is found to be -6.2 ± 0.2 eV. A novel direct bandgap Ga(NAsP) MQW structure is monolithically integrated on a (001) Si substrate using a (BGa)P strain compensating layer. The high crystalline quality of a Ga(NAsP)/(BGa)P/Si-MQWHs (annealed at 850 °C) is demonstrated from the XRD pattern analysis and AFM imaging. Promising optical material quality of Ga(NAsP)/(BGa)P/Si-MQWHs has been

realized from PL measurements, which is a positive step towards a commercial solution for the monolithic integration of long term stable laser diodes on silicon substrates.

The authors gratefully acknowledge the EPSRC (UK), grant reference EP/H005587/1, the Royal Academy of Engineering (UK), the Kwan Trust Fund, the German Science Foundation (DFG), and the German Ministry for Education and Research (BMBF) for supporting this work.

- ¹M. Paniccia, *Nat. Photonics* **4**, 498 (2010).
- ²C. Karcher, K. Jandieri, B. Kunert, R. Fritz, M. Zimprich, K. Volz, W. Stolz, F. Gebhard, S. D. Baranovskii, and W. Heimbrodt, *Phys. Rev. B* **82**, 245309 (2010).
- ³B. Kunert, K. Volz, J. Koch, and W. Stolz, *Appl. Phys. Lett.* **88**, 182108 (2006).
- ⁴S. Borck, S. Chatterjee, B. Kunert, K. Volz, W. Stolz, J. Heber, and W. W. Ruhle, *Appl. Phys. Lett.* **89**, 031102 (2006).
- ⁵N. Hossain, S. R. Jin, S. J. Sweeney, S. Liebich, S. Reinhard, K. Volz, B. Kunert, and W. Stolz, in Proceedings of the IEEE Photonics Society 23rd Annual Meeting 2010 (IEEE, USA, 2010), pp. 65–66, <http://dx.doi.org/10.1109/PHOTONICS.2010.5698759>.
- ⁶H. Kroemer, *J. Cryst. Growth* **81**, 193 (1987).
- ⁷B. Kunert, D. Trusheim, V. Voßbüßer, K. Volz, and W. Stolz, *Phys. Status Solidi A* **205**, 114 (2008).
- ⁸K. Volz, J. Koch, B. Kunert, I. Németh, and W. Stolz, *J. Cryst. Growth* **298**, 126 (2007).
- ⁹B. Kunert, S. Zinnkann, K. Volz, and W. Stolz, *J. Cryst. Growth* **310**, 4776 (2008).
- ¹⁰S. Rogowsky, M. Baumler, M. Wolfer, L. Kirste, R. Ostendorf, J. Wagner, S. Liebich, W. Stolz, K. Volz, and B. Kunert, *J. Appl. Phys.* **109**, 053504 (2011).
- ¹¹K. Volz, A. Beyer, W. Witte, J. Ohlmann, I. Németh, B. Kunert, and W. Stolz, *J. Cryst. Growth* **315**, 37 (2011).
- ¹²B. Kunert, I. Németh, S. Reinhard, K. Volz, and W. Stolz, *Thin Solid Films* **517**, 140 (2008).
- ¹³I. Németh, B. Kunert, W. Stolz, and K. Volz, *J. Cryst. Growth* **310**, 1595 (2008).
- ¹⁴N. E. Fox, T. K. Sharma, S. J. Sweeney, and T. J. C. Hosea, *Phys. Status Solidi A* **206**, 796 (2009).
- ¹⁵L. Kronik and Y. Shapira, *Surf. Sci. Rep.* **37**, 89 (1999).
- ¹⁶S. Datta, S. Ghosh, and B. M. Arora, *Rev. Sci. Instrum.* **72**, 177 (2001).
- ¹⁷B. Adamowicz and J. Szuber, *Surf. Sci.* **247**, 94 (1991).
- ¹⁸H. Asai and K. Oe, *J. Appl. Phys.* **54**, 2052 (1983).
- ¹⁹A. Blacha, H. Presting, and M. Cardona, *Phys. Status Solidi B* **126**, 11 (1984).
- ²⁰I. Vurgaftman, J. R. Meyer, and L. R. Ram-mohan, *Appl. Phys. Rev.* **89**, 5815 (2001).
- ²¹S. Iyer, K. Eberl, M. Goorsky, F. Legoues, J. Tsang, and F. Cardone, *Appl. Phys. Lett.* **60**, 356 (1992).
- ²²G. A. Slack and S. F. Bartram, *J. Appl. Phys.* **46**, 89 (1975).
- ²³A. Jenichen and C. Engler, *Phys. Status Solidi B* **247**, 59 (2010).
- ²⁴B. Kunert, S. Liebich, A. Beyer, R. Fritz, S. Zinnkann, K. Volz, and W. Stolz, *J. Cryst. Growth* **315**, 28 (2010).