Optimal growth and thermal stability of crystalline $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ alloy films on $\text{Al}_2\text{O}_3(0001)$

Dae-Sung Park,¹ A. Krupski,¹ A. M. Sanchez,¹ Chel-Jong Choi,² Min-Su Yi,³ Hyun-Hwi Lee,⁴ S. R. C. McMitchell,¹ and C. F. McConville^{1,a)}

¹Department of Physics, University of Warwick, Coventry CV4 7AL, United Kingdom

²School of Semiconductor and Chemical Engineering, Semiconductor Physics Research Center (SPRC), Chonbuk National University, Jeonju 561-756, South Korea

³Department of Materials Science and Engineering, Kyungpook National University, Sangju 742-711, South Korea

⁴Pohang Accelerator Laboratory, POSTECH, Pohang 790-784, South Korea

(Received 14 February 2014; accepted 25 March 2014; published online 7 April 2014)

The influence of growth temperature on the synthesis of $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy films, grown on highly-mismatched $\text{Al}_2\text{O}_3(0001)$ substrates, was studied by synchrotron x-ray scattering, high-resolution transmission electron microscopy and photoluminescence measurements. A single-phase $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy with a Be concentration of $x=0.25$, was obtained at the growth temperature, $T_g=400^\circ\text{C}$, and verified by high-resolution transmission electron microscopy. It was found that high-temperature growth, $T_g \geq 600^\circ\text{C}$, caused phase separation, resulting in a random distribution of intermixed alloy phases. The inhomogeneity and structural fluctuations observed in the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown at high temperatures are attributed to a variation in Be composition and mosaic distribution via atomic displacement and strain relaxation. © 2014 AIP Publishing LLC. [<http://dx.doi.org/10.1063/1.4870533>]

Band gap engineering of oxides has attracted a great deal of interest towards realizing oxide quantum heterostructures, such as multi-quantum wells (MQWs) and quantum Hall interfaces, for the advancement of oxide optoelectronics.^{1–5} Among oxide materials, $\text{Be}_x\text{Zn}_{1-x}\text{O}$ ternary alloys have been regarded as a possible alternative to achieve high efficiency ZnO-based optoelectronic devices such as ultra-violet (UV) light emitting diodes (LEDs), laser diodes (LDs), and UV sensors.⁶ The main advantage of using $\text{Be}_x\text{Zn}_{1-x}\text{O}$ ternary alloys for these applications is that ZnO and BeO have the same wurtzite crystal structure with a large difference in band gap energy.⁷ It is expected that by substituting Be into the Zn site in the host ZnO, the band gap energy of ZnO can be modulated from $\approx 3.3\text{ eV}$ (ZnO) through to $\approx 10.6\text{ eV}$ (BeO) without significant phase segregation in the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloys. This is unlike $\text{Mg}_x\text{Zn}_{1-x}\text{O}$ alloys which are limited by a Mg mole fraction of $x=0.38$.^{3,8} However, in principle, the phase instability of $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloys could occur due to: (i) internal strain effect from different bond lengths between Zn-O (a - and c -axis bond lengths: $1.97\text{ \AA}$ and $1.99\text{ \AA}$) and Be-O (a - and c -axis bond lengths: $1.65\text{ \AA}$ and $1.66\text{ \AA}$) and the subsequent lattice deformation of wurtzite structure;⁹ (ii) a large difference in thermal expansion coefficients [2.49 and $3.98 \times 10^{-6}\text{ K}^{-1}$ (5.35 and $7.18 \times 10^{-6}\text{ K}^{-1}$) at 300 K and 700 K for c -axis ZnO (c -axis BeO), respectively];¹⁰ and (iii) substantial strain from the lattice mismatch between the alloy film and the underlying substrate, e.g., Al_2O_3 ($\approx 18\%$) and Si (111) ($\approx 15\%$).¹¹ These can readily induce phase separation and compositional splitting in the ternary alloy system.¹² Hence, in order to obtain a homogeneous and stable alloy phase, it is essential to investigate the synthesis of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloys under various conditions. A few such studies have reported

the growth and properties of the alloy films in highly-mismatched system.^{13–15}

In this study, we report the effect of growth temperature on $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown on $\text{Al}_2\text{O}_3(0001)$. A single-phase hexagonal $\text{Be}_x\text{Zn}_{1-x}\text{O}$ film was obtained with the optimal Be content, $x=0.25$, into the host at $T_g=400^\circ\text{C}$. For higher growth temperatures, $T_g \geq 600^\circ\text{C}$, phase segregation and compositional fluctuation were observed by x-ray diffraction (XRD) peak splitting and broad multi-photoluminescence emission bands. The structural fluctuations following high-temperature synthesis of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloys were addressed by Be concentration variations under strain relaxation and thermally-driven Be displacement during the alloy film growth.

The $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy films were grown on c -axis Al_2O_3 substrates by radio-frequency (13.56 MHz) magnetron co-sputtering. The oxygen partial pressure (P_{O_2}) was 2×10^{-2} millibar and growth temperatures were in the range of $T_g=200\text{--}800^\circ\text{C}$. A low-temperature ($T=200^\circ\text{C}$) ZnO seed layer was employed prior to growth of the main alloy film. Film thicknesses in the range of $\approx 210\text{--}280\text{ nm}$ were determined by cross-sectional scanning electron microscopy (X-SEM) as shown in Fig. 1(a). Synchrotron radiation x-ray scattering measurements ($\lambda=1.069\text{ \AA}$) were performed at beamline 5 A of the Pohang Light Source II (PLS-II) to investigate the structural variation of the alloy films grown at different T_g . The variation in the 0002 Bragg reflections of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy films was evaluated using $q_z=4\pi \sin \theta/\lambda$, where q_z and θ are the peak position in the reciprocal space and incident photon angle, respectively. An undoped ZnO film prepared at $T_g=800^\circ\text{C}$ was measured under the same experimental conditions as a reference sample. High-resolution transmission electron microscopy (HRTEM) was carried out using a Tecnai G² F30 S-Twin (FEI) with an acceleration voltage of

^{a)}Electronic mail: c.f.mcconville@warwick.ac.uk

300 kV to characterize the microstructure of the $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ alloy film grown at $T_g = 400^\circ\text{C}$. Photoluminescence (PL) spectra of the alloy films were measured using a He-Cd laser ($\lambda = 325\text{ nm}$) at 5 K.

Figure 1(b) shows the synchrotron x-ray θ - 2θ diffraction profiles for the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown at different growth temperatures from 200 to 800°C . The 0002 peak position of the undoped ZnO thin film, grown at $T_g = 800^\circ\text{C}$, was at $q_z = 2.40\text{ \AA}^{-1}$. For the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ film grown at $T_g = 200^\circ\text{C}$, three distinct diffraction peaks were found at $q_z = 2.26, 2.46$, and $2.66\text{ \AA}^{-1}$, respectively, which correspond to the $10\bar{1}0$, 0002, and $10\bar{1}1$ plane reflections. However, the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ film grown at $T_g = 400^\circ\text{C}$ indicated a single 0002 peak with the largest shift to $q_z = 2.49\text{ \AA}^{-1}$ with respect to that of the undoped ZnO 0002 reflection. This results from a single wurtzite-phase and the highest possible Be composition in the alloy film. For $T_g \geq 600^\circ\text{C}$, significant phase separation occurred as the 0002 peak splits into Zn-rich and Be-rich phases. Additional broadening of the 0002 peak was also observed in these films.

In order to further examine the structural variation of the alloy films as a function of T_g , ω -rocking curves of the 0002 Bragg reflection were measured as shown in Fig. 1(c). A non-smooth rocking curve with broken lines and a large full-width at half maximum (FWHM) was observed for the $\text{Be}_{0.14}\text{Zn}_{0.86}\text{O}$ film at $T_g = 200^\circ\text{C}$, implying textured nucleation and misaligned crystallites on the underlying substrate.

This is usually caused by insufficient energy for atomic movement during low-temperature film growth.¹⁶ By contrast, a symmetric rocking curve was found with the smallest FWHM of 4.66° for the alloy film grown at $T_g = 400^\circ\text{C}$ [Fig. 1(c)]. However, for $T_g \geq 600^\circ\text{C}$, a superposition of both sharp and broader components in the rocking curves was observed with an increase in FWHM up to 11.16° . In previous studies,^{16,17} it was reported that such a narrow component generally reflects a two-dimensional (2D) coherent layer well-aligned with the underlying substrate in the early stages of growth. On the other hand, the broad component represents the mosaicity of a film with textured grains with respect to the substrate normal. Since a large lattice mismatch at the $\text{Be}_x\text{Zn}_{1-x}\text{O}/\text{Al}_2\text{O}_3$ interface occurs, it is expected that the formation of a three-dimensional (3D) columnar structure of the films is favorable during the growth to minimize lattice strain. The narrow component could be associated with the partial distribution of well-aligned alloy domains along the substrate normal (c -axis). As can be seen in the inset of Fig. 1(c), multiple narrow components in the rocking curve were observed at $T_g \geq 600^\circ\text{C}$. This is suggestive of the partial distribution of c -axis alloy domains with multiple in-plane alignments, as well as, different Be compositions. Furthermore, if the narrow component originates from the low temperature ZnO seed layer in a two-step growth mode (2D-3D transition),¹⁶ all of the diffraction peaks should involve ZnO-related diffraction, while these

FIG. 1. (a) Cross-sectional SEM images for the undoped ZnO ($T_g = 800^\circ\text{C}$) and the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films ($T_g = 200, 400, 600$, and 800°C) with thicknesses of ≈ 206 – 277 nm , grown on Al_2O_3 substrates. (b) Synchrotron radiation XRD θ - 2θ patterns of $\text{Be}_x\text{Zn}_{1-x}\text{O}$ 0002 are compared with undoped ZnO 0002 by the vertical red-dash line, (c) the ω -rocking curves of $\text{Be}_x\text{Zn}_{1-x}\text{O}$ 0002 Bragg reflection, and (d) two-dimensional (2D) x-ray scattering images for the ZnO grown at $T_g = 800^\circ\text{C}$ and the alloy films grown at different T_g (200– 800°C). The red-dashed horizontal lines referred to in (d) represent the 2θ position of the undoped ZnO film in the 2D x-ray reflection area.

peaks were not observed in films grown at $T_g \leq 400^\circ\text{C}$. Hence, the sharp components are believed to be due to well-aligned c -axis alloy domains in the alloy films. Note the broad component in the rocking curve becomes wider as T_g increased to 800°C . This results from further mosaic spread meaning the film crystallinity has deteriorated through the temperature-driven relaxation process. Therefore, thermodynamics plays a crucial role in the mobility of Be and strain relaxation. This initiates both phase separation and compositional fluctuation in the synthesis of $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy on a highly-mismatched substrate.

Two-dimensional XRD (2D-XRD) images of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films as a function of T_g are illustrated in Fig. 1(d). The 2D-XRD image of the undoped ZnO film shows a focused spot, indicating a uniform in-plane lattice arrangement and isotropic crystallinity. Diffraction rings were observed for the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown at $T_g = 200$ and 600°C due to textured in-plane lattices and low crystal quality for the alloy films. For $T_g \geq 600^\circ\text{C}$, the 2D x-ray reflection area was remarkably spread to include the undoped ZnO-related area (denoted by the horizontal red-dash lines). This is indicative of phase separation to Be-depleted and Be-rich phases and further distribution of misaligned lattices along the out-of-plane (2θ) and in-plane (χ) directions, causing the compositional inhomogeneity and structural deterioration of the alloy films. As a result, the most narrow 2D spot was found at $T_g = 400^\circ\text{C}$, corroborating the single crystalline nature with the most uniform Be composition in the alloy film.

HRTEM was performed to elucidate the microstructural properties of the $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ film grown on $\text{Al}_2\text{O}_3(0001)$ substrate at $T_g = 400^\circ\text{C}$. The cross-sectional TEM image of the alloy film showed a 3D columnar structure [as illustrated in Fig. 2(a)]. It is known that the columnar structure typically arises from the nucleation and coalescence of islands accompanied by lattice strain during the growth process.^{12,18} Figures 2(b) and 2(c) show the cross-sectional HRTEM of the wurtzite $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ along the $\langle 11\bar{2}0 \rangle$ direction and corresponding Fast Fourier transform (FFT) for the red-dash square in the HRTEM image. The FFT displayed typical spots of ZnO and Al_2O_3 for the $\langle 11\bar{2}0 \rangle$ and $\langle 10\bar{1}0 \rangle$ zone axes, respectively. This reveals that the prime domains of the alloy film are well-aligned along the substrate with a typical epitaxial relationship of $[0001]\text{Be}_x\text{Zn}_{1-x}\text{O} \parallel [0001]\text{Al}_2\text{O}_3$ and $[10\bar{1}0]\text{Be}_x\text{Zn}_{1-x}\text{O} \parallel [11\bar{2}0]\text{Al}_2\text{O}_3$, consistent with our previous study.¹⁹

From the plane spacings (d) of the $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ 0001 and $01\bar{1}0$ planes in the FFT, the lattice parameters of the in-plane and out-of-plane of the film were determined to be $a = 3.10 \text{ \AA}$ and $c = 5.00 \text{ \AA}$, respectively. This corresponds to a Be concentration of $x = 0.25$ in the alloy film calculated by Vegard's law.²⁰ The lattice constants remained unchanged with almost identical d in the entire film. This is in good agreement with the lattice constant, $c = 5.04 \text{ \AA}$ ($d = 2\pi/q_z$), of the thin film, determined by XRD. The extra spots depicted by green- and red-circle pairs in the FFT were observed in addition to the main diffraction spots of the alloy lattice in Fig. 2(c). Figure 2(d) displays corresponding areas in green and red at the interface region, which are selectively obtained by masking each extra spot-pair in the inverse FFT. Those

FIG. 2. (a) Cross-sectional HRTEM image of the $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ film grown on $\text{Al}_2\text{O}_3(0001)$ substrate at $T_g = 400^\circ\text{C}$. (b) HRTEM image along the $\langle 11\bar{2}0 \rangle$ zone axis of the wurtzite alloy film and (c) FFT image (corresponding to the area inside the red-dash square) of the interface region between the alloy film and substrate. (d) A superposition of the HRTEM image with the masked red and blue areas corresponding to the extra-spots (denoted by red- and blue-opened circles) in the FFT.

extra spots represent misaligned domains (tilted and/or rotated) with respect to the underlying substrate due to the large lattice mismatch ($\approx 13\%$) at the interface.²¹ The misoriented domains were predominantly distributed at the interface region compared to the upper layer region, and the alloy film was initially relaxed by a number of misoriented grains at the near interface region. The absence of a 2D ZnO seed layer at the interface region of the film was discussed earlier. Hence, it was assumed that either there is interlayer mixing between the seed and alloy layer or it forms 3D islands in the interface region during the early stage of growth.

The PL spectra of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown at different T_g are shown in Fig. 3. All of the spectra indicate excitonic emission features at $T = 5 \text{ K}$. The band emission energy region was blue-shifted compared to that of undoped ZnO normally in the range of around $3.2\text{--}3.4 \text{ eV}$.²² This is due to the incorporation of Be into the host ZnO, which results in the widening of the band gap. No specific origin of each exciton-related recombination center in $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloys has yet been determined. This is due to the change in the electronic band structure and the following ionized energy levels related to defects, e.g., Zn_i and hydrogen impurities.^{9,23} Therefore, the PL spectra were sub-divided into distinct regions: (i) the free-exciton emission (FXE), $3.795\text{--}3.754 \text{ eV}$; (ii) the different bound exciton emissions (BXE) containing many types of donor/acceptor-bound excitons and the correlated donor-acceptor pairs (DAP), $3.754\text{--}3.615 \text{ eV}$; and (iii) the DAP longitudinal optical (LO)-phonon replicas by an energy separation of $\approx 75 \text{ meV}$ with respect to the DAP energy, $3.615\text{--}3.488 \text{ eV}$.^{22,24–26} In addition, an emission band at a photon energy of around 3.3 eV was observed in all films, which is close to the energy region of the near-band-edge emission (NBE) for undoped ZnO.

FIG. 3. Photoluminescence spectra of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown at different temperatures (T_g : (A) 200 °C, (B) 400 °C, (C) 600 °C, and (D) 800 °C). All measurements were performed at $T = 5$ K.

The alloy films grown at relatively low temperature ($T_g \leq 400$ °C) showed distinct band-edge emission lines. The strongest excitonic emission bands with the highest energy ≈ 3.76 eV (FXE) was seen in the $\text{Be}_{0.25}\text{Zn}_{0.75}\text{O}$ film at $T_g = 400$ °C. This is consistent with previous results, indicating high crystallinity of the alloy film grown at this temperature. In contrast, for the high-temperature ($T_g \geq 600$ °C)-grown films, additional large broad emission bands were found near the NBE emission bands of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films. Broad emission bands attributed to compositional fluctuation under strain relaxation have been reported in other ternary systems, e.g., $\text{In}_x\text{Ga}_{1-x}\text{N}$ and $\text{Mg}_x\text{Zn}_{1-x}\text{O}$.^{27,28} In the above synchrotron XRD measurements, randomly-distributed intermixing of alloy phases separated into Zn-rich and Be-rich alloy phases was also observed. Hence, the multi-emission band (MEB) in Fig. 3 is due to the presence of localized energy states originating from a broad distribution of the local composition of Be in the alloy films. The intensity of undoped ZnO-related emission for the ternary films grown at $T_g \geq 600$ °C increased compared to that from the alloy films grown at low T_g , which is consistent with the XRD results. Therefore, the PL results shown in Fig. 3 provide evidence for the compositional fluctuation/splitting of Be in the relatively high-temperature-grown $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films.

The experimental observations clearly indicate that high-temperature growth of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films on a highly-mismatched substrate causes: (i) Be compositional fluctuation; (ii) phase separation; and (iii) increased film mosaicity. These factors all contribute to lowering the strain energy associated with incorporating the smaller Be ions into the ZnO lattices through Be segregation, made possible by atomic mobility at higher temperatures, exceeding the solubility limits. Here, different strains are induced by differing degrees of lattice mismatch between the different local compositions in the alloy film and the underlying substrate. Following the segregation process, the increased mismatch of the Be-rich alloy with the substrate drives Be transport away from the interface region, which has more elastic strain energy than the upper layers of

the film. With such an elastic strain distribution near the interface, Be atoms tend to move away from the substrate into more relaxed regions by a compositional pulling effect.²⁹ Subsequently, lower Be compositions can be present in the interface region, giving rise to the formation of undoped ZnO-like interface region. As T_g increases, this relaxation process is promoted by providing more energy for Be movement. As a result, the Be composition in the film is divided roughly into Zn-rich and Be-rich regions in the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy films grown at $T_g \geq 600$ °C. Further relaxation at higher T_g can lead to more mosaic spread in the alloy films. In addition, thermally-driven Be atoms moving to more relaxed areas are randomly dissociated from the alloy lattices. This causes an increasing structural deterioration of the alloy film. In our previous study,¹⁹ it was found that the dissociation/diffusion of Be in $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy films was induced by thermal annealing over 600 °C. Klingshirn *et al.* predicted the diffusion of Be and the resulted segregation in local areas of $\text{Be}_x\text{Zn}_{1-x}\text{O}/\text{ZnO}$ -based QW structures.^{6,30} Hence, we suggest that the compositional inhomogeneity and phase separation of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloys originate from the coupling of strain relaxation and displacement of Be during the high-temperature synthesis.

In this study, the influence of the growth temperature on the formation of crystalline $\text{Be}_x\text{Zn}_{1-x}\text{O}$ films grown on *c*-axis Al_2O_3 substrates has been investigated by using synchrotron XRD, TEM, and PL measurements. The XRD and HRTEM results showed that a single-phase $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy film with the optimal Be composition of $x = 0.25$ obtained at $T_g = 400$ °C. For $T_g \geq 600$ °C, phase separation into Zn-rich and a Be-rich phases were observed together with compositional inhomogeneity of Be in the alloy films. This corresponds to the XRD $\text{Be}_x\text{Zn}_{1-x}\text{O}$ 0002 peak splitting and broad PL emission bands. These results are attributed to a compositional pulling effect and wider distribution of mosaic structure in the high-temperature-grown films. The structural fluctuation is caused by the strain relaxation process in conjunction with thermal displacement of Be at the high-temperature during growth of the $\text{Be}_x\text{Zn}_{1-x}\text{O}$ alloy films on the highly-mismatched substrate.

The authors acknowledge technical support from R. I. Johnston in the Surface, Interface, and Thin Films Group in the Department of Physics at the University of Warwick. Access to equipment used in this study was provided through the Science City Research Alliance (SCRA) with capital equipment funding from the European Research and Development Fund (ERDF). Dae-Sung Park acknowledges funding from the University of Warwick for a Chancellor's scholarship.

¹A. Tsukazaki, A. Ohtomo, T. Kita, Y. Ohno, H. Ohno, and M. Kawasaki, *Science* **315**, 1388 (2007).

²A. Ohtomo and H. Y. Hwang, *Nature* **427**, 423 (2004).

³A. Ohtomo, M. Kawasaki, T. Koida, K. Masubuchi, H. Koinuma, Y. Sakurai, Y. Yoshida, T. Yasuda, and Y. Segawa, *Appl. Phys. Lett.* **72**, 2466 (1998).

⁴J. D. Ye, S. Pannirselvam, S. T. Lim, J. F. Bi, X. W. Sun, G. Q. Lo, and K. L. Teo, *Appl. Phys. Lett.* **97**, 111908 (2010).

⁵H. D. Sun, T. Makino, Y. Segawa, M. Kawasaki, A. Ohtomo, K. Tamura, and H. Koinuma, *J. Appl. Phys.* **91**, 1993 (2002).

- ⁶Y. R. Ryu, T. S. Lee, J. A. Lubguban, H. W. White, B. J. Kim, Y. S. Park, and C. J. Youn, *Appl. Phys. Lett.* **88**, 241108 (2006).
- ⁷Y. R. Ryu, T. S. Lee, J. A. Lubguban, A. B. Corman, H. W. White, J. H. Leem, M. S. Han, Y. S. Park, C. J. Youn, and W. J. Kim, *Appl. Phys. Lett.* **88**, 052103 (2006).
- ⁸W. J. Kim, J. H. Leem, M. S. Han, I.-W. Park, Y. R. Ryu, and T. S. Lee, *J. Appl. Phys.* **99**, 096104 (2006).
- ⁹L. Dong and S. P. Alpay, *Phys. Rev. B* **84**, 035315 (2011).
- ¹⁰T. Yao and S.-K. Hong, *Oxide and Nitride Semiconductors: Processing, Properties, and Applications* (Springer, 2010).
- ¹¹H. Iwanaga, A. Kunishige, and S. Takeuchi, *J. Mater. Sci.* **35**, 2451 (2000).
- ¹²D. S. Park, J. J. Mudd, M. Walker, D. Seghier, A. Krupski, N. F. Saniee, C.-J. Choi, C. J. Youn, S. R. C. McMitchell, and C. F. McConville, *CrystEngComm* **16**, 2136 (2014).
- ¹³J. M. Khoshman, D. C. Ingram, and M. E. Kordes, *Appl. Phys. Lett.* **92**, 091902 (2008).
- ¹⁴M. S. Han, J. H. Kim, T. S. Jeong, J. M. Park, C. J. Youn, J. H. Leem, and Y. R. Ryu, *J. Cryst. Growth* **303**, 506 (2007).
- ¹⁵C. Yang, X. M. Li, Y. F. Gu, W. D. Yu, X. D. Gao, and Y. W. Zhang, *Appl. Phys. Lett.* **93**, 112114 (2008).
- ¹⁶I. W. Kim and K. M. Lee, *Nanotechnology* **19**, 355709 (2008).
- ¹⁷S. I. Park, T. S. Cho, S. J. Doh, J. L. Lee, and J. H. Je, *Appl. Phys. Lett.* **77**, 349 (2000).
- ¹⁸J. M. Yuk, J. Y. Lee, Y. S. No, T. W. Kim, and W. K. Choi, *Appl. Phys. Lett.* **93**, 021904 (2008).
- ¹⁹D. S. Park, S. K. V. Farahani, M. Walker, A. Krupski, E. B. Thorsteinsson, D. Seghier, C.-J. Choi, C. J. Youn, S. R. C. McMichell, and C. F. McConville, "Thermally-induced recrystallization of highly-mismatched Be_xZn_{1-x}O alloys: The formation of a degenerate interface," *Appl. Mater. Interfaces* (submitted).
- ²⁰A. R. Denton and N. W. Ashcroft, *Phys. Rev. A* **43**, 3161 (1991).
- ²¹V. Srikant, J. S. Speck, and D. R. Clarke, *J. Appl. Phys.* **82**, 4286 (1997).
- ²²B. K. Meyer, H. Alves, D. M. Hofmann, W. Kriegseis, D. Forster, F. Bertram, J. Christen, A. Hoffmann, M. Straburg, M. Dworzak, U. Haboeck, and A. V. Rodina, *Phys. Status Solidi B* **241**, 231 (2004).
- ²³J. Sann, J. Stehr, A. Hofstaetter, D. M. Hofmann, A. Neumann, M. Lerch, U. Haboeck, A. Hoffmann, and C. Thomsen, *Phys. Rev. B* **76**, 195203 (2007).
- ²⁴M. Brandt, H. Von Wenckstern, G. Benndorf, M. Lange, C. P. Dietrich, C. Kranert, C. Sturm, R. SchmidtGrund, H. Hochmuth, M. Lorenz, M. Grundmann, M. R. Wagner, M. Alic, C. Nenstiel, and A. Hoffmann, *Phys. Rev. B* **81**, 073306 (2010).
- ²⁵C. Lin and A. Z. Li, *J. Cryst. Growth* **203**, 511 (1999).
- ²⁶A. Yamamoto, Y. Moriwaki, K. Hattori, and H. Yanagi, *Appl. Phys. Lett.* **98**, 061907 (2011).
- ²⁷S. Pereira, M. R. Von Wenckstern, E. Pereira, C. Trager-Cowan, F. Sweeney, K. P. O'Donnell, E. Alves, N. Franco, and A. D. Sequeira, *Appl. Phys. Lett.* **81**, 1207 (2002).
- ²⁸H. Matsui, H. Tabata, N. Hasuike, and H. Harima, *J. Appl. Phys.* **99**, 024902 (2006).
- ²⁹S. Pereira, M. R. Correia, E. Pereira, K. P. O'Donnell, C. Trager-Cowan, F. Sweeney, and E. Alves, *Phys. Rev. B* **64**, 205311 (2001).
- ³⁰C. Klingshirn, J. Fallert, H. Zhou, and H. Kalt, *Appl. Phys. Lett.* **91**, 126101 (2007).