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Direct current magnetron sputtered ZrB₂ thin films on 4H-SiC(0001) and Si(100)

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Abstract

ZrB₂ thin films have been synthesized using direct current magnetron sputtering from a ZrB₂ compound target onto 4H-SiC(0001) and Si(100) substrates kept at different temperatures (no heating, 400 °C, and 550 °C), and substrate bias voltage (-20 V to -80 V). Time-of-flight energy elastic recoil detection analysis shows that all the films are near stoichiometric and have a low degree of contaminants, with O being the most abundant (< 1 at.%). The films are crystalline, and their crystallographic orientation changes from 0001 to a more random orientation with increased deposition temperature. X-ray diffraction pole figures and selected area electron diffraction patterns of the films deposited without heating reveal a fiber-texture growth. Four point probe measurements show typical resistivity values of the films ranging from ~95 to 200 μΩcm, decreasing with increased growth temperature and substrate bias.

Keywords

Zirconium diboride

Silicon carbide

Thin films

Compound target
Industrial scale deposition system
Crystalline films

1. Introduction

Zirconium diboride (ZrB_2) lends its high hardness, high melting point, and low resistivity from a hexagonal crystal structure, where the boron atoms form honeycombed, graphite-like sheets that are stacked between hexagonal close packed zirconium layers [1]. The ceramic properties, demonstrated by ZrB_2 , originate from covalent type of bonding between the metal and boron atoms as well as between the boron atoms in the sheet, whereas the electrical conductivity stem from metal-metal bonding and electron transfer from the metal to the boron sheet to yield graphite-like conduction. The unit cell of ZrB_2 exhibits an a -axis of 3.17 Å and a c -axis of 3.53 Å, with the zirconium atoms at the basis (0,0,0) and the boron atoms positioned in interstitials at $(\frac{1}{3}, \frac{2}{3}, \frac{1}{2})$ and $(\frac{2}{3}, \frac{1}{3}, \frac{1}{2})$. Furthermore, the lattice parameter of 3.17 Å in the a direction gives ZrB_2 a small lattice mismatch to technologically important semiconductor-materials such as silicon carbide (4H-SiC(0001)) and gallium nitride (GaN(0001)) with values of -3.8 % and 0.6 %, respectively. The prospect of growing epitaxial films with well-defined properties in combination with the ceramic metallic characteristics demonstrated by ZrB_2 makes it interesting as a thin film electrical contact material for demanding applications. Hitherto, growth on GaN(0001) has been investigated [2-6], whereas growth on 4H-SiC(0001) remains to be demonstrated.

The literature shows that ZrB_2 films have been deposited by physical vapor deposition (PVD) [7-10], as well as chemical vapor deposition (CVD). [11,12]. Films grown by sputtering have more or less exclusively been deposited from a ZrB_2 compound target or

from Zr-B composite targets, since reactive sputtering of ZrB_2 is complicated by the lack of a suitable boron source; the first natural choice of precursor, diborane (B_2H_6), being both highly toxic and explosive, as well as reported to cause substantial target poisoning during growth of the closely related phase TiB_2 , resulting in the sputtering of boron at low rates [13].

However, there are also difficulties in sputtering of ZrB_2 films using compound sources. Results on growth of films with composition deviating, sometimes substantially, from the composition of the utilized source, as well as growth of films with a high level of contaminants have been reported [2,4,6,10,14-18]. The presence of a foreign element such as oxygen during nucleation of ZrB_2 is likely to cause unwanted chemical reactions between the film constituents and the contaminant. Such conditions will affect the properties, including the electrical resistivity, [10] of the deposited films and can, for instance, hamper potential epitaxial growth to substrates with a close lattice match, such as 4H-SiC(0001). The necessity of a stoichiometric, growth flux during deposition of ZrB_2 films is further supported by thermodynamics, predicting ZrB_2 to be a line phase with a narrow homogeneity range ($<1\%$) [19], thus the composition of the films can only deviate very little to avoid unwanted phase separation. These properties, shown by ZrB_2 , are in contrast to, e.g., the transition metal carbides and nitrides that are characterized by large homogeneity ranges [20] as well as possible solid solution of oxygen [21]. As a consequence, any excess atoms of Zr or B in ZrB_2 will not be accommodated in the crystal structure, but instead nucleate as separate phases, thus disrupting sought epitaxial growth conditions. For the related boride TiB_2 , it has been shown that excess B segregates to the grain boundaries [22] and it is likely that this would happen in ZrB_2 as well. With respect to

the resulting properties of sputtered ZrB_2 films, it has been reported that B-depleted films exhibit higher resistivity values compared to that of films with a composition closer to stoichiometry [10]. The feasibility of overcoming the inherent drawbacks for such growth conditions in film synthesis during high vacuum (HV) conditions by combining substrate bias and heating should thus be explored.

In this study, we investigate the influence of growth temperature and substrate bias during synthesis of ZrB_2 thin films on 4H-SiC(0001) for potential epitaxial growth as well as on Si(100) substrates, using magnetron sputtering of a ZrB_2 compound target in an industrial scale HV deposition system at short deposition times (100 s) and at a high power density of the target (11.36 W/cm^2). The deposited films were characterized with respect to their structural properties, microstructure, growth relationship, and interfacial properties to the substrate material, as well as their electrical properties.

2. Experimental details

The films were deposited by direct current magnetron sputtering (DCMS) from a ZrB_2 compound target, with a purity of 99.5 %, excluding Hf, in a commercial industrial HV system (CemeCon CC800/9). The unit was held at a base pressure of $\sim 0.13 \text{ mPa} - 1.3 \text{ mPa}$, depending on the growth temperature with corresponding monolayer formation times (τ) of 0.2 s and 2 s, respectively. Prior to deposition, the power on the rectangular shaped cathode ($50 \times 8.8 \text{ cm}^2$) vertically mounted on the side door was ramped behind a shutter from 500 W to 5 000 W (power density 11.36 W/cm^2) and then clean sputtered for 210 s. During deposition, the target faced the substrate thus enabling the sputtered growth species to

arrive perpendicular to the substrate. As no rotation was used the target-to-substrate distance was constant at 7 cm. The process gas (Ar) pressure was set to 0.5 Pa by regulating the flow of Ar. The turbo molecular pumps were operated either at 66 % of full speed or at full speed to investigate the influence of the pumping speed on the load of residual gas during film growth. A range of dc bias voltages between -20 V and -80 V was applied to the substrate and the deposition time was 100 s, yielding ~300 nm thick films, corresponding to a growth rate of about 8 unit cells/s in the *c*-axis direction. The typical floating potential for a DCMS process is around -10 V in this deposition system [23]. The power of the heater was 0 W, 5 000 W or 10 000 W, corresponding to approximately room temperature (RT), 400 °C, and 550 °C respectively, as calibrated by thermocouple. Getter pumping was applied immediately before the deposition to further minimize the impact of the residual gas. This was achieved by sputtering of a separate titanium target for 600 s onto a shield that covered the magnetron, using a power of 1 500 W and at an Ar pressure of 0.42 Pa. The target was located on the opposite side in the chamber as seen from the ZrB₂ target. Two different substrates were deposited upon in each run, Si(100) and 4H-SiC(0001) with resistivity values of 10-20 Ωcm and 0.023 Ωcm respectively. The size of the Si-substrates was 20*25 mm² and the size of the SiC-substrates was 12,5*12,5 mm². The Si(100) substrates with the native oxide present were ultrasonically degreased in isopropanol prior to mounting. The 4H-SiC(0001) substrates were subjected to the following cleaning steps: 1) Warm trichloroethylene for 5 min, 2) acetone for 3 min, 3) ethanol for 3 min, 4) 1:1:5 solution of NH₄OH:H₂O₂:H₂O at 85 °C for 5 min, 5) 1:1:6 solution of HCl:H₂O₂:H₂O at 85 °C for 5 min, and 6) 1:3 solution of HF(48 %):H₂O for 1 min. After each step the sample was rinsed in ultra-pure water.

X-ray diffraction (XRD) measurements were performed in a Philips PW 1820 system equipped with a Cu K α source operated at 40 kV and 40 mA during $\theta/2\theta$ and ω -scans. XRD pole figures were recorded using a PANalytical EMPYREAN system operated at 45 kV and 40 mA. The film morphologies and thicknesses were investigated using cross section scanning electron microscopy (SEM, LEO 1550 Gemini) with an acceleration voltage of 5 kV or 10 kV. The cross-sectional TEM specimens were prepared by gluing two pieces of samples face to face together, polishing from both sides of the specimen down to 60 μm in thickness, and finally ion milling to electron transparency. Both HRTEM imaging and collection of selected area diffraction patterns (SAED) were carried out by using a FEI Tecnai G2 TF20 UT HRTEM with a field emission gun operated at 200 kV and with a point resolution of 1.9 Å. The elemental composition of the as-deposited films was obtained by time-of-flight energy elastic recoil detection analysis (ToF-E ERDA) at an accuracy of $\pm 5\%$ for Zr and $\pm 3\%$ for B. The measurements were carried out with a 36 MeV $^{127}\text{I}^{8+}$ ion beam using the set-up at Uppsala University [24,25]. The incident angle of primary ions and exit angle of recoils were both 67.5 ° to the sample surface normal giving a recoil angle of 45 °. The measured ToF-E ERDA spectra were converted into relative atomic concentration profiles using the CONTES code [26]. Room temperature four point probe measurements were performed using an Auto map system Model 280C from Four Dimensions, Inc. The in-plane resistivity was then calculated by multiplying the obtained sheet resistance with the film thickness obtained by SEM imaging of cross-sections of films deposited on Si(100) substrates.

3. Results and discussion

3.1. Composition

Table 1 shows the boron-to-zirconium ratio (B/Zr) determined by ToF-E ERDA for 8 selected films deposited on Si(100) substrates at different process conditions. As can be seen, the first 6 numbered films are nearly stoichiometric with $1.97 \leq \text{B/Zr} \leq 2.00$, while films 7 and 8 exhibit Zr-rich compositions with $\text{B/Zr} = 1.84$ and $\text{B/Zr} = 1.88$. The latter two films were deposited at a combination of high temperature and high bias and suggest that boron is abstracted from the growing film at such process conditions. Further, we note from films 5 and 6 in Table 1 that an increased substrate temperature is not sufficient to reduce the B-content. This result is in contrast to Mitterer et al. who found that the B/Zr ratio increased with increasing deposition temperature [9]. However, the condition that a high bias favors a Zr-rich composition is in accord with previous studies, as re-sputtering of boron during growth of ZrB_2 from a compound target have been proposed by Mitterer et al. [9] and Samuelsson et al. [7], using DCMS and high power impulse magnetron sputtering, respectively. It is likely that a preferential re-sputtering process at moderate bias voltage of -80 V, as applied in our experiments, is eased at elevated temperatures. Furthermore, given the fact that the ZrB_2 phase demonstrate a very narrow homogeneity range (< 1 at%) [19], it is plausible that the small surplus of Zr atoms present during growth of films at high temperature and high bias segregates to the grain boundaries where they are liable to chemically interact with the residual gas following a similar behavior as reported for B in TiB_2 [22]. Such chemical reactions are likely to disrupt the growth of films with well-defined properties, i.e. epitaxial films.

ToF-E ERDA shows that the level of contaminants, as constituted by the elements H, Ar, N, C and O is low. The measured amounts being typically below 0.2 at.% for H, Ar, and N and 0.4 at.% for C and 1 at.% for O. Such contents, particularly the oxygen content, must be considered to be low in PVD films compared to the literature [10,15-18], as the growth was carried out in a HV system and without external heating in some of the processes. The oxygen level is also low compared to thermally activated CVD films deposited at a temperature of 450 °C [11]. From our measurements, we note that the choice of speed applied to the turbo molecular pumps has no influence on the amount of contaminants in the films, thus indicating a low impact from the residual gas (mainly water vapor at HV conditions). Instead, we believe that the oxygen mainly stems from the compound target. This has also been suggested in a study by Takeyama et al [16] as well as by Shappirio et al [17]. While it is not possible to determine whether the oxygen originates from oxides in the target material or as oxygen containing species dissolved in pores present in the sintered material, we note that several studies in the literature have been performed, using target materials with densities less than the theoretical bulk value [9,10,17]. Their results indicate that a denser target material is likely to enable the deposition of films with a reduced level of contaminants. Thus it is likely that at least part of the oxygen originates from oxygen containing species dissolved in pores in the target material.

3.2. Structural properties

The XRD $\theta/2\theta$ diffractograms in Figure 1 that show relatively broad peaks $2\theta: \approx 25.0^\circ, \approx 32.6^\circ, \approx 41.5^\circ, \approx 51.4^\circ$, and $\approx 58^\circ$ were recorded from films deposited on 4H-SiC(0001) substrates at RT, 400 °C, and 550 °C and using a substrate bias of -80 V. The broad peaks can be assigned to ZrB_2 [27] and demonstrate that the phase has been

deposited. Visible is also the main 4H-SiC 0004 peak at $2\theta \approx 35.7^\circ$ as well as the 0003, 0005, 0006, and 0004 (K_β) peaks from the substrate. In addition, the diffractograms show that the film orientation is dependent on temperature. Films deposited without external heating only has two visible peaks from ZrB_2 ; the 0001 at $2\theta \approx 25.0^\circ$ and the 0002 at $2\theta \approx 51.4^\circ$. This indicates either a textured or an epitaxial growth to 4H-SiC(0001). A 0001 preferred orientation for ZrB_2 films deposited on substrate materials such as Si(100), molybdenum, and steel at low growth temperatures has been reported in the literature [7,8,28]. As the substrate temperature is increased the intensity of the 0001 peak reduces and the 0002 peak becomes absent. The ZrB_2 films now display a more random oriented growth behavior with the 10-11 peak at $2\theta \approx 41.5^\circ$ showing the highest intensity. Mitterer et al. also reported this growth behavior for ZrB_2 films when deposited on molybdenum and steel substrates [29].

The films deposited on Si(100) substrates (not shown) exhibit similar structural properties as those grown on 4H-SiC(0001), i.e. 0001 oriented at RT conditions and with a random orientation at elevated growth temperatures. The 10-10 peak of ZrB_2 is very weak or absent in all films, thus ruling out potential epitaxial growth on Si(100) [12]. This is probably due to the native oxide present on the substrate, obstructing a lattice-match growth between the film and the Si(100) surface. In general, the films deposited on Si(100) show higher intensities, in relation to the background, compared to those grown on 4H-SiC(0001). XRD ω -scans of the 0001 peak suggest a higher crystalline quality for the films deposited on Si(100) by showing higher intensities and lower full-width-at-half-maximum values for

films grown on Si(100) at RT conditions compared to those deposited on 4H-SiC(0001), with typical values of $\sim 8^\circ$ and $\sim 12^\circ$, respectively.

For the films deposited on Si(100) substrates without external heating at substrate bias voltages from -20 V to -80 V, we find that the 0001 peak is shifted compared to the unstrained bulk value [27]. Figure 2 shows that films deposited with a substrate bias voltage ≥ -40 V exhibit shifts of the 0001 peak towards lower diffraction angles that suggests that the films are in a state of compressive residual stress [30]. It is here suggested that the observed residual stress is due to energetic bombardment by Ar ions during synthesis. Such growth behavior is supported by the facts that the ZrB_2 phase permits very little deviation from stoichiometry that may cause deviation in peak position and the tensile contribution should be nearly constant as the film thicknesses are more or less constant according to SEM analysis, see section 3.3. Also, as the films were grown without intentional heating, we can disregard the stress component due to differences in thermal expansion of the substrate and film. An increasing level of compressive stress at increasing substrate bias is supported by initial depositions at even higher substrate bias ≥ -120 V, showing levels of stress high enough for spontaneous delamination of the film from the Si(100) substrate. Shifts of the $\text{ZrB}_2(0001)$ peak towards lower diffraction angles were also evident for films deposited on 4H-SiC(0001) at RT and bias voltages of -60 V and -80 V (not shown).

Figure 3 shows XRD pole figures of a ZrB_2 film grown at RT on 4H-SiC(0001) with a substrate bias of -80 V. For the monitored 0001 pole in Figure 3 (a) the film exhibits a

broad spot in the center of the projection. This is in agreement with an expected high intensity at $\psi \approx 0^\circ$ (cf. Figure 1). The 10-11 pole in Figure 3 (b) displays a ring at $\psi \approx 50^\circ$ of uniform intensity, which is in accord with the calculated value for the angle between the 0001 plane and the 10-11 plane in ZrB_2 , being 52° . The ring in the projection demonstrates that the film is textured and not epitaxial and the rotational symmetry supports a fiber-textured growth to the 4H-SiC(0001) substrate at RT deposition conditions. Further support is given from the monitored 11-21 pole (Figure 3 (c)) where a ring can be seen at $\psi \approx 70^\circ$, corresponding to the nominal value of 69° for the angle between the 0001-plane and the 11-21 plane in ZrB_2 . A textured growth behavior to 4H-SiC(0001) is also supported by HRTEM investigations below. A fiber texture growth behavior for ZrB_2 films deposited on Si(100) substrates at RT conditions was also confirmed by pole figure measurements (not shown).

Pole figures recorded from the films deposited on 4H-SiC(0001) at 550°C show scattered intensities for the monitored 0001, 10-11, and 11-21 poles, which is consistent with the randomly oriented growth determined from $\theta/2\theta$ diffraction.

There are several possible reasons for not achieving epitaxial growth on 4H-SiC(0001) substrates. The first limiting factor, particularly for films deposited at RT conditions, is insufficient mobility on the surface for the adatoms, which favors a large number of nucleation points during the initial growth of ZrB_2 . Here, we note that the temperature applied when depositing epitaxial films on Si(111) as well as on Si(100) using CVD and the precursor zirconium borohydride $\text{Zr}(\text{BH}_4)_4$ is substantially higher, typically 900°C [31], whereas a lower temperature of 450°C in CVD yields amorphous films [11]. A second

limiting factor seems to be the substrate preparation prior to growth as Si(111) and Si(100) surfaces applied for growth by CVD is firstly *ex-situ* etched by HF and then *in-situ* heated [12,31,32]. It is well-known that a thin oxide is formed on our 4H-SiC(0001) following exposure to the air. This oxide remains on the surface since the temperatures used in our experiments are not sufficient to remove it. The oxide layer as well as adsorbed water vapor and hydrocarbons, likely present on the substrates deposited at RT, will negatively affect the conditions for epitaxial growth. Finally, Tolle et al. [31] concluded that the epitaxial growth of ZrB₂ films by CVD on Si(111) is controlled by kinetics and therefore limited the growth rate to ~1.2 nm/min (at 900 °C). Comparing this value with our growth rate of ~180 nm/min (at ≤ 550 °C) we can conclude that this is too high to achieve epitaxial growth on 4H-SiC(0001) substrates.

3.3. Microstructure

Figure 4 shows HRTEM low magnification images from two films deposited on 4H-SiC(0001) at (a) RT and at (b) 550 °C, respectively. The images show that both films exhibit a columnar microstructure and a smooth surface structure. The SAED patterns recorded from both films that are displayed as insets in Figure 4, support that the film deposited at RT is textured and not epitaxial and that the film deposited at 550 °C is randomly oriented. From the low magnification images, it was possible to determine the film thickness, which are: ~307 nm for the film in (a) and ~302 nm for the film in (b).

Furthermore, the micrograph in (a) suggests that the size of the columns decreases towards the film substrate interface. This is supported by the high magnification image in

Figure 5 (a) revealing that there is a ~20 nm thick nanocrystalline or amorphous layer deposited closest to the substrate. Figure 5 (b) shows that such a layer is not present for the film deposited at elevated temperature. From these results, we note that the oxygen profiles recorded from the ToF-E ERDA investigations display higher concentration for oxygen at the substrate-film interface for films deposited at RT conditions than from those deposited at 550 °C. It is likely that a substantial part of this signal originates from other sources than the thin native oxide on the investigated Si(100) substrates, thus indicating the presence of adsorbed oxygen containing species on the substrate surfaces of the films deposited at RT conditions. Reactions between the film constituents and such remaining species will affect the crystalline quality of the film deposited closest to the substrate and may result in a nanocrystalline or amorphous growth. As the deposition process proceed the energetic flux will start to heat the substrate, thereby enabling both a higher probability for desorption of un-reacted oxygen containing species as well as an increased surface mobility for the film constituents. Such conditions will favor the growth of films with higher crystalline quality, which is reflected in the columnar growth behavior of the film further from the substrate and to the film-vacuum interface.

SEM analysis shows a columnar growth for ZrB₂ on Si(100), see Figure 6 for (a) a film deposited at RT and a substrate bias of -80 V and (b) at 550 °C and a substrate bias of -80 V. This behavior is in agreement with the results obtained for DCMS sputtered films in the study of Samuelsson et al. [7]. Furthermore, the surface of the deposited films is in general smooth, in particular for films deposited at RT and with a bias voltage > -40 V, which is in accord with the results from the HRTEM investigation. The images indicate that

an increased growth temperature in combination with a substrate bias < -40 V results in surface roughening.

From the recorded SEM images of films deposited on Si(100) substrates, it was not possible to discern the amorphous/nanocrystalline layer visible in the TEM images from films deposited on 4H-SiC(0001) and at RT conditions. However, TEM images obtained from ZrB₂ films deposited on Si(100) confirm an initial growth of a layer with such properties also on this substrate material. From the cross section images, it was possible to determine the thickness to 312 ± 48 nm for all films deposited on Si(100). These values are in good agreement with the thicknesses 307 nm and 302 nm obtained from the TEM images of films deposited on 4H-SiC(0001) at 550 °C and without external heating, respectively.

3.4. Resistivity

The resistivity of the deposited films range between 95 and 200 $\mu\Omega\text{cm}$ irrespective of the choice of substrate material, see Figure 7. However, there is a trend indicating lower resistivity for films deposited at higher temperature or with higher substrate bias voltages, ≥ 60 V, and with the lowest values obtained for films grown on Si(100). We further observe that films with the lowest resistivity also exhibit low oxygen content.

The obtained resistivity values are substantially higher than the bulk value for ZrB₂ (~ 10 $\mu\Omega\text{cm}$ [33]). Our obtained values are also higher than for films deposited using remote-plasma CVD (~ 45 $\mu\Omega\text{cm}$) [11], where the growth rate was very low (~ 10 nm/min) and for thermal CVD (~ 20 $\mu\Omega\text{cm}$), where the growth rate was even lower (~ 0.3 nm/min)

[34]. The high growth rate and insufficient adatom mobility in our experiments does not allow for the atoms to reach lattice equilibrium positions, influencing the conductivity. However, the resistivity values of our films are comparable to other magnetron sputtered films that show resistivity values ranging between 160 and 440 $\mu\Omega\text{cm}$ [7,10,16,17,35].

4. Conclusions

We show that ZrB_2 thin films can be deposited on 4H-SiC(0001) and Si(100) substrates by dc magnetron sputtering from a ZrB_2 compound target. The films are close to stoichiometric and exhibit a low degree of contaminants, albeit with a slight Zr-rich composition when deposited at a temperature of 550 °C and a substrate bias of -80 V. XRD and SAED patterns confirm that the films are 0001 textured on both type of substrates when grown without external heating, and with a transition to a random orientation at higher temperatures. 0001 oriented films are in a state of compressive stress when deposited at a substrate bias ≥ 60 V; HRTEM reveals that a thin ~ 20 nm amorphous/nanocrystalline layer is deposited closest to the substrate for such films, but not for those grown at a temperature of 550 °C. The electrical resistivity values of the films range between 95 and 200 $\mu\Omega\text{cm}$, which is similar to other as-deposited films grown by sputtering from a compound target.

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Table 1 Deposition parameters and the B/Zr ratio for 8 samples investigated by ToF-E ERDA.

Sample	Deposition Temp (°C)	Substrate bias (-V)	Speed of turbo molecular pumps (%)	B/Zr
1	RT	20	66	1.99
2	RT	40	66	1.98
3	RT	40	100	1.97
4	RT	60	66	1.98
5	400	20	66	2.00
6	550	20	66	1.98
7	550	80	66	1.84
8	550	80	100	1.88

Figure 1 $\theta/2\theta$ XRD diffractograms of coatings deposited on 4H-SiC(0001) with a substrate bias of -80 V at different temperatures. The peak seen at $\approx 30^\circ$ for the coating deposited at 400 °C is a detector artefact.

Figure 2 $\theta/2\theta$ XRD diffractograms, showing the ZrB₂ 0001-peak, of coatings deposited on Si(100) at RT and at different substrate bias voltages. The dashed line represents the literature value of $2\theta = 25.205$ for the ZrB₂ 0001-peak.

Figure 3 XRD pole figures for a film deposited on 4H-SiC(0001) with a substrate bias of -80 V at RT showing the a) 0001 pole, b) 10-11 pole and c) 11-21 pole. Color online.

Figure 4 Low magnification TEM micrographs and SAED patterns of films grown on 4H-SiC(0001) with a substrate bias of -80 V at a) RT and b) 550 °C

Figure 5 HRTEM micrographs of the substrate-film interface of films grown on 4H-SiC(0001) with a substrate bias of -80 V at a) RT and b) 550 °C

Figure 6 Cross sectional SEM micrographs of films grown on Si(100) with a substrate bias of -80 V at a) RT and b) 550 °C

Figure 7 Electrical resistivity values for films grown on 4H-SiC(0001) and Si(100) at substrate biases of -20, -40, -60, and -80 V and at RT, 400 °C, and 550 °C.