An inadvertent error has occurred in the derivation of the theoretical ratio of $m$- and $a$-plane segment lengths for the GaN equilibrium crystal shape using the Wulff-plot in projection along ⟨0001⟩ as shown in figure 7(b) of our paper (2013 New J. Phys. 15 053045). A value of $s = 2.4$ was reported, however, the correct value is $s = 1.8$. Within the accuracy of the measurement, the experimentally derived value of 2.2(4) is in agreement with this corrected value.

Therefore, this detail does not have any impact on the discussion and conclusions of our paper.
Ga-polar GaN nanocolumn arrays with semipolar faceted tips

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Abstract. Selective area growth of GaN nanocolumns (NCs) by molecular beam epitaxy on laser ablated pre-patterned GaN(0001) templates is shown to provide regular arrays of Ga-polar NCs. The Ga diffusion-assisted growth mechanism is analyzed and the experiments suggest that the effective growth conditions vary with the height of the NCs due to Ga diffusion on the mask and the NC sidewalls, ranging from N-rich up to stoichiometry. The obtained morphology with semipolar facets at the tip is discussed within the framework of equilibrium thermodynamics, which provides a consistent picture also for the growth of N-polar NCs with flat tips. The structural investigation reveals almost defect-free semipolar \{1102\} GaN facets at the top of the NCs, which is known to be a promising way of producing templates for nanoscale semipolar GaN-based heterostructures. Almost no polarization discontinuity is expected for In\textsubscript{x}Ga\textsubscript{1-x}N/GaN interfaces on such facets.

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1. Introduction

Group III nitride semiconductors grown along the polar [0001] c-direction are widely used for optoelectronic devices in the near ultraviolet and blue range of the electromagnetic spectrum [1, 2]. Polarization related problems present in these conventional GaN-based heterostructures have hampered the development of efficient GaN-based optoelectronic devices operating at longer wavelengths, into the green range of the visible spectrum. The heterostructure growth perpendicular to non-polar or semipolar GaN interfaces has been demonstrated to be a successful approach to increase the optical device efficiency [3, 4]. In fact the reduction of the internal electric fields which separate the electrons and holes via the quantum-confined Stark effect within the active region improves the radiative efficiency of semipolar GaN-based optoelectronic devices as compared with the polar ones. In the particular case of In$_x$Ga$_{1-x}$N/GaN semipolar interfaces the vanishing of the total polarization discontinuity is predicted for growth plane inclinations of about 45° with respect to the c-plane [5]. A straightforward procedure to obtain semipolar GaN layers is to cut mm-thick c-plane GaN wafers grown by hydride vapor phase epitaxy along the desired plane orientation with the obvious drawback of very small area substrate material. Alternatively, large area semipolar GaN layers can be grown from the c-plane-like sidewalls of trenches etched into r- or n-plane sapphire substrates [6–9]; however, this procedure requires several processing steps. A further approach is to grow GaN structures which form stable semipolar facets allowing for the growth of multi-quantum well structures onto them. This has been demonstrated by selective area metal–organic vapor phase epitaxy [10–12] as well as by molecular beam epitaxy (MBE) [13–17]. Depending on growth conditions and in particular on the Ga to N supply ratio in the MBE process GaN nanocolumns (NCs) with small area {1101} [13], r-plane {1102} [15], or {1103} [14] facets at the tip of the NCs have been obtained. The NCs grow along the c-axis which is the optimum growth direction for GaN, therefore resulting in low structural defect densities—in particular basal plane stacking faults (BSFs).

The selective area growth (SAG) process is characterized by a high control of both the size and the position of the grown NCs thus avoiding the significant dispersion in the optoelectronic properties intrinsic to the self-organized grown NC ensembles. The well-defined arrangement is also prerequisite for NC-based device processing [18]. Further it allows to characterize individual, virtually identical objects and provides the opportunity to control the growth rates.
among different arrays on the same sample in a decoupled manner, as they sensitively depend on the design of the array [19]. Setting the chemical environment of each column by the design of the array also helps to elucidate the growth mechanism which is not explained in full detail up to now [20].

In the present work, SAG of GaN by MBE on pre-patterned GaN(0001) templates is shown to provide regular arrays of Ga-polar GaN NCs. Laser ablation as a high-throughput technique was employed to pattern a thin Mo-mask, which acts as a functional layer. The observed morphology with semipolar facets at the tip of the Ga-polar NCs can be understood within the framework of equilibrium thermodynamics, which also provides a consistent picture of the N-polar GaN NCs with flat tips. The structural investigation reveals almost defect-free semipolar \{1\overline{1}02\} facets at the top of the NCs. Therefore this approach turns out to be a suitable method yielding a large substrate area for SAG by a few processing steps. In addition, due to the relatively large diameter of the NCs, the area of the semipolar planes at the tip can be increased as compared to the typically obtained NCs with diameters below 300 nm [13–15].

2. Experimental

The samples were grown on commercially available MOCVD GaN(0001)/Al₂O₃ templates in a Veeco GenII MBE system equipped with a radio frequency (RF)-plasma source and an effusion cell for gallium. The GaN templates were degreased in an ultrasonic bath using acetone, methanol and deionized water. Onto the bare substrate a 20 nm thin Mo layer was deposited by e-beam evaporation at a low deposition rate of 1 Å s⁻¹ which yields a smooth surface. A cubic array of apertures was patterned within a laser ablation process with 10 ps pulses at λ = 355 nm through interferometric projection of a cross grid onto the layer. The mask was analyzed by atomic force microscopy (AFM) and scanning electron microscopy (SEM).

SEM analysis after the laser ablation processing shows incomplete ablation in the aperture region and redeposition of Mo onto the mask (figure 1(a)). A higher ratio of ablated to non-ablated material can be obtained by increasing the laser power; however, due to the low ablation threshold of GaN also the substrate material is removed resulting in defects at the GaN template surface, which are then expected to propagate in the growing NCs. Therefore after laser ablation the Mo was etched down to 10 nm by a selective reactive ion etching (RIE) step. A mixture of SF₆ and Ar gases with an etch rate of about 3 Å s⁻¹ was used. Figure 1(b) shows that the RIE process provides apertures free of any Mo. The surface roughness as determined by AFM on a 1 × 1 µm² region yields a rms value of 0.3 and 0.7 nm for the as-deposited and etched Mo layers, respectively. The edge of the apertures is less sharp than obtained on similar structures by e-beam lithography. The final array dimensions are around 800 and 700 nm for aperture pitch and diameter, respectively (figure 1(c)). In the following, the pitch is defined as the nearest neighbor aperture edge interspacing.

In situ outgassing of the pre-structured samples took place at 600 °C shortly before growth in the MBE system. The substrate temperature for growth was Tₛ = 780 °C. The nitrogen plasma source was run with a nitrogen flux of 1 sccm and a plasma excitation power of 400 W. The N⁺/Ga supply ratio was about 9 as compared with stoichiometric growth conditions. The structural properties of the SAG NCs were investigated by SEM and transmission electron microscopy (TEM). The preparation of the cross-sectional TEM lamellae was performed using a dual beam focused ion beam (FIB).
3. Results and discussion

3.1. Selective area growth (SAG)

To investigate the morphology and the growth mechanism of the SAG GaN NCs, a series of MBE samples has been grown for increasing growth time. As already reported for SAG of GaN NCs by MBE using various types of mask materials [21–23] we also show that in the case of Mo the growth of the NCs is mainly driven by Ga adatom diffusion on the mask during the early stage of growth. Figure 2 shows the NC morphology evolution with growth time, the elaboration of the experiment by AFM is shown in figure 3.

Figure 2(a) shows that during the early stage of growth multiple GaN nuclei are formed in the apertures and coalesce. The coalescence is completed for growth durations longer than 20 min and a compact pedestal is formed (b). With increasing growth time semipolar facets are formed starting at the pedestal edges and grow from the outside of the NC inwards (c)–(e). Eventually the semipolar facets have coalesced and formed fully developed pyramidal structures on a pedestal with a hexagonal cross section (f). Due to the rough edges of the apertures (figure 1) also the sidewalls reveal a certain roughness.

Figure 3 shows the axial growth rate $w_{ax}$ of the NCs as a function of growth time as determined from quantitative analysis of the AFM data. An incubation time for nucleation of 5–7 min as observed by reflection high energy electron diffraction during growth has been taken into account. Furthermore the volume of the non-cylindrical tips was converted into an equivalent cylindrical volume.

Two different growth regimes can be distinguished: during regime I a constant axial growth rate is observed. This value is enhanced by a factor of about 5 as compared with the layer equivalent growth rate $w_{layer\,equiv}$ (dotted line), nevertheless it is still smaller than that of GaN MBE layers grown at the same substrate temperature under stoichiometric conditions using the same nitrogen supply ($w \approx 6 \text{ nm min}^{-1}$). The layer equivalent growth rate was extrapolated using the GaN layer growth rate at stoichiometry reduced by the ratio of the Ga fluxes. In regime II the growth rate decreases, yet yielding higher axial growth rates as compared with the layer equivalent growth rate. The radial growth rate $w_{rad}$ was deduced from SEM micrographs and is constant for both regimes within the accuracy of the measurements. It has to be noted that for longer growth times non-uniform radial overgrowth on the Mo mask occurs (figures 2(e) and (f)) and the radial growth rate is thus underestimated.
Figure 2. SEM top view (a)–(f) and birds view (a′)–(f′) micrographs of the SAG NCs. The growth times are \( t = 10, 20, 40, 60, 120 \) and 180 min for (a)–(f), respectively.

Under the assumption that the axial growth rate of the NCs is determined by both direct impingement on the aperture area (\( w_{\text{layer equiv}} \)) and a diffusion term of Ga atoms on the Mo mask (the circular collection area which does not overlap with the nearest neighbor area), one can estimate the Ga diffusion length on the Mo \( \lambda_{\text{Ga}} \) by simple geometrical arguments. Hence the diffusion enhanced growth rate reads \( w_{\text{ax}} = w_{\text{layer equiv}} (r + \lambda_{\text{Ga}})^2 / r^2 \), where \( r \) is the NC radius. From the average value of \( r = 245 \) nm and the value \( w_{\text{ax}} \) for regime I a Ga diffusion length of \( \lambda_{\text{Ga}} = 290 \pm 16 \) nm is estimated.

The decrease of the axial growth rate for regime II is explained by taking into account that the Ga adatoms supplied by diffusion on the Mo mask have to diffuse to the top of the NC to contribute to the axial growth. Once the NC height \( h \) exceeds the Ga diffusion length

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Figure 3. Axial and radial growth rates of the GaN NCs as a function of time. The dotted line shows the layer equivalent growth rate and the dashed line shows the enhanced axial growth rate due to Ga diffusion on the Mo mask. The grey shaded region represents the error of the axial growth rate due to the error of the Ga diffusion length. The red solid line is a guide for the eye.

on the NC sidewalls $\lambda_{Ga}^{sw}$, the reduced probability for the diffusing Ga adatoms to reach the top of the NC will lead to a decrease of the axial growth rate with increasing growth time. This explanation agrees well with the observation that for longer growth times either self-organized NCs nucleate directly on the Mo mask (re-evaporated Ga adatoms; see figures 2(e) and (f)), or radial overgrowth takes place. It is worth noting that we observe the onset of the reduced axial growth rate for regime II to go along with the formation of the semipolar facets, see figures 2(b) and (c). In crystal growth, the slowest growing facet will eventually determine the crystal shape. Thus, the overall axial growth rate will decrease due to the slower growth rate of the semipolar facets as compared with the axial growth rate perpendicular to the substrate. Therefore, we interpret the decrease of the growth rate to result from the reduced probability of Ga adatoms to reach the NC tip with increasing NC height and the formation of the semipolar facets.

We would like to emphasize that the consequences of two different Ga adatom diffusion lengths—$\lambda_{Ga}$ on the Mo mask and $\lambda_{Ga}^{sw}$ on the NC sidewalls—can have tremendous effects concerning the effective growth conditions of the NCs. The nominal N-rich growth conditions (supply ratio $N^+$/Ga = 9) effectively result in growth conditions being around stoichiometry (supply ratio equals 1) for the case $h < \lambda_{Ga}^{sw}$ in regime I due to the Ga diffusion from the Mo mask. For this estimation, we used the typical aperture and NC diameters from regime I and the value of $\lambda_{Ga} = 290$ nm. Contributions from the sidewalls have been neglected. If, however, the NC height exceeds the Ga diffusion length on the sidewalls only the direct impingement at the tip and the impinging flux at the NC sidewalls within a distance $\lambda_{Ga}^{sw}$ to the tip will contribute to axial growth. This may change the effective growth conditions at the tip toward the N-limited or Ga-limited regime again, depending on the value of $\lambda_{Ga}^{sw}$. The Ga adatom diffusion length on GaN $m$-plane sidewalls under similar growth conditions is reported in the literature to be around 40 nm [24, 25]. Estimating the effective supply ratio for this regime yields N-rich growth
3.2. Structural properties and polarity

In the following paragraph, the structural properties of the SAG NCs will be discussed. Thin NCs are often referred to as candidates for the growth of defect-free material due to their large aspect ratio and small contact area to the substrate. Although we observe a variety of defects with considerable densities in the SAG NCs—which have typically a much smaller aspect ratio than self-organized NCs—we will show that for a certain NC height the reduced defect density argument also applies in the SAG case due to effective dislocation bending. Figure 4 shows a bright field TEM image in [11\(\overline{2}0\)] zone axis geometry.

The tip facet of the NC shows an inclination of 43.9° with respect to the \(c\)-plane, which is in good agreement with the theoretical value of 43.2° for the \{1\(\overline{1}02\}\} surfaces. The base region of the NCs show a high defect density with the presence of threading dislocations (TDs) and BSFs.
as expected due to the coalescence of multiple GaN nuclei within the apertures (figures 2(a) and (b)). It is observed that the TDs either bend toward [1100] or appear to be terminated in the NC. As a termination of a TD inside any crystal is not possible [26], bending toward the [1102]-direction and subsequent termination at the surfaces of the NC is expected to occur. For GaN NCs on GaN template the [0001] length $l_{TD}$ of the TDs is reported to be a function of the radial distance $r_{TD}$ of the TD to the NC sidewalls [26]. For $r_{TD} = 200$ nm the corresponding value for $l_{TD}$ is about 400 nm. Thus, for NCs with a diameter of $d = 400$ nm and a centered TD, we therefore expect TD bending after the NC height exceeds about $h = l_{TD} = 400$ nm. In fact, for the grown NC in figure 4 ($d = 630$ nm) all the TDs are terminated at a height of $h = 475$ nm. TDs annihilate due to bending or interaction with BSFs [27]. Thus, the total defect density in the pyramidal region is reduced by up to one order of magnitude. Beyond a critical NC height no TDs are observed. The main type of defect in the tip region is BSFs with an average defect density of $\bar{\rho}_{BSF} = 5 \times 10^4 \text{ cm}^{-1}$. For some investigated NCs no BSFs at all are present in the pyramidal tip region.

For future optoelectronic devices not only the structural properties of the NCs but also the polarity is of interest. An influence on the incorporation behavior of In in GaN with different polarity is observed and higher achievable In concentrations are predicted for N-polar GaN as compared with Ga-polarity [28]. The polarity of GaN NCs and the corresponding morphology is frequently discussed in the literature [18]. For SAG GaN NCs grown by metalorganic vapor phase epitaxy the polarity is suggested to be determined by the substrate, namely Ga-polar NCs are grown on GaN(0001) template, whereas N-polar NCs are grown on GaN(0001) freestanding substrate.

A pyramidal morphology is observed for Ga-polar growth, whereas NCs with flat tips are related to N-polarity [29]. The latter morphology is also observed for self-organized N-polar NCs grown by MBE [30, 31]. However, the fact that MBE growth on non-polar substrates (for example Si(111)) also yields N-polar GaN NCs demonstrates that aside the polarity of the substrate further physical properties influence the polarity of the NCs.

Several experimental methods [30–33] can be used to determine the polarity of NCs; in this study we applied both convergent beam electron diffraction (CBED) and wet chemical etching. KOH is known to etch preferentially the N-face of bulk GaN [34] therefore the SAG NCs with fully developed pyramidal morphology were etched for 10 min in a saturated KOH solution at room temperature. The base region facing toward the Ga-polar substrate is etched and indicates the Ga-polarity of the NCs, at least in the bottom region. The resulting morphology is presented in figure 5(a). To corroborate these results and to extend the information to the tip region, CBED experiments were performed on the SAG NCs. The investigated sample shows slightly different tip facets ( {2203}); however, the overall morphology of these NCs is analogous to that of the NCs with {1102} tip facets. Due to the typically large aspect ratio of NCs the CBED experiment is not straightforward and the results have to be evaluated carefully as crystal edges are discussed to affect the appearance of the CBED disks [35]. The diameter of the SAG NCs is typically larger as compared with self-organized NCs, therefore polarity determination by CBED is more convenient in our case.

The measurements were done on cross-sectional TEM lamellae in [1100] zone axis geometry at 200 kV. The diameter of the investigated NCs was about 300 nm prior to FIB preparation, the thickness of the TEM lamella in the range of $d = 50$–150 nm. Therefore, the samples present a slice geometry for which edge effects due to the initial hexagonal cross section of the NCs have not to be taken into account. First, the CBED pattern of the Ga-polar
Figure 5. SEM micrograph of a SAG NC grown for \( t = 180 \text{ min} \) after KOH treatment for 10 min at room temperature (a). A simulated \((d = 102 \text{ nm})\) and experimental CBED pattern (base region) as well as intensity profiles (dotted lines) along the CBED disks are shown in (b) at the top and the bottom, respectively.

A peculiarity concerning polarity is found for the self-organized NCs which show flat \( c \)-plane tips and start to grow on the Mo mask after longer growth times. The polarity of these NCs was investigated by KOH etching using the same procedure as described above. The tips of the NCs are etched proving the N-polarity (not shown here). This result is consistent with KOH etching experiments performed by other groups [31]. Due to the polycrystallinity of the Mo—or of the thin MoN\(_x\) layer which forms at the interface—no information of the substrate’s polarity can be transferred to the NCs. These conditions therefore resemble the self-organized growth of N-polar NCs on SiN\(_x\)/Si(111) [30, 31].

3.3. Discussion

The characteristic morphologies which we observe for Ga- and N-polar NCs (pyramidal and flat tips, respectively) will be now discussed in a thermodynamic picture. It is worth noting that for a comprehensive understanding also adatom kinetics have to be taken into account [36–39]. However, the high growth temperature and low growth rates will bring the system closer to thermodynamic equilibrium [40]. The following section indeed will show that also in a solely thermodynamic framework one might be able to obtain a consistent picture.

In thermodynamic equilibrium the shape of a crystal will be given by the configuration, which minimizes the total surface energy while conserving the volume. This means that in principle we should calculate the surface energy for the different surfaces as described
in [40] and compare them quantitatively. These calculations also relate to specific growth conditions—by the difference of the chemical potentials $\mu_{\text{Ga}} - \mu_{\text{Ga}}^{\text{bulk}}$—which influence the atomic surface termination. Furthermore, surface reconstructions are included. However, experimentally the growth conditions in terms of the difference of the chemical potentials are difficult to determine, especially due to the deviation from the nominal supply because of the contribution by Ga diffusion. In addition, the absolute value of the surface energy can be calculated only when the two surfaces created by cutting a crystal are crystallographically identical [41]. This is the case for the non-polar GaN surfaces which have been theoretically investigated by Northrup and Neugebauer [42]. In contrast, only relative surface energy values for the investigated semipolar [43–46] and polar [40] GaN surfaces can be calculated and are thus not suited for direct quantitative comparison. Therefore, we will use a simple broken bond model in which the GaN crystal is cut at some arbitrary plane and the broken bond density $\rho_{\text{bb}}$ for the two created surfaces can be easily calculated. Since the presence of dangling bonds increases the surface energy [40], we will compare in the following their density at various GaN surfaces to estimate which surfaces are expected to be energetically more favorable. The specific chemical environment, surface reconstructions or edge effects are not taken into account.

Figure 6 shows the broken bond densities for various GaN surfaces in the case of Ga- and N-polar GaN. The non-polar $m$- and $a$-surfaces are stoichiometric and $\rho_{\text{bb}}$ is independent of the polarity. This is not the case for the polar and semipolar surfaces if a specific surface termination is assumed. The polar $\pm c$-surfaces were chosen to be N-terminated, namely kinetically stabilized due to the effective N-rich growth conditions [40]. The semipolar ($\overline{2}20n$) surfaces with $n$ ranging from 2 to 5 in the Ga-polar case and from $-2$ to $-5$ in the N-polar case show stoichiometric termination, mixed termination or complete coverage by one atomic species. The surface termination depends here on the exact cleavage position. Consistently, from the different N-terminated surface configurations the one with the lowest broken bond density was chosen for each value of $n$.

The $m$- and $a$-surfaces have the lowest broken bond densities with only small differences. For Ga-polar NCs, the semipolar surfaces show about 1.5 times higher density than the non-polar ones. The value for the N-terminated $c$-surface broken bond density is the largest one, $\rho_{\text{bb}}^c = 0.425 \text{ Å}^{-2}$. Thus, we find

$$\rho_{\text{bb}}^m \lesssim \rho_{\text{bb}}^a < \rho_{\text{bb}}^\text{semipolar} \ll \rho_{\text{bb}}^c.$$ 

For N-polar GaN NCs, the N-terminated (000$\overline{1}$)-surface has a broken bond density almost equal to that of the $a$-surface, the semipolar surfaces do not show significant changes with change of polarity. Therefore, the broken bond densities follow the relation:

$$\rho_{\text{bb}}^m \lesssim \rho_{\text{bb}}^a \approx \rho_{\text{bb}}^\overline{\text{c}} < \rho_{\text{bb}}^\text{c}.$$ 

In the following, a possible explanation for the observed NC morphologies will be discussed. For a growth direction parallel to [0001] (Ga-polar), the NC sidewalls will be stabilized by $m$- and $a$-surfaces with the lowest values of $\rho_{\text{bb}}$. The broken bond density of the semipolar planes $\rho_{\text{bb}}^\text{semipolar} \ll \rho_{\text{bb}}^c$, therefore the crystal tends to form the semipolar surfaces at the top which are then terminated at the sidewalls. This described morphology is experimentally observed for the Ga-polar SAG NCs (figure 2). Discussing the N-polar case requires the assumption that the growth direction of N-polar GaN NCs is [000$\overline{1}$]. This is not straightforward, since N-polar NCs are not only observed for epitaxial growth on N-polar substrates [29] but also for self-organized growth on polycrystalline substrates—the Mo mask layer in our case—for which no epitaxial
Figure 6. Calculated broken bond densities for various GaN surfaces. The N-terminated semipolar and polar surfaces are shown for the Ga-polar and N-polar case to the left and right, respectively. The values for the stoichiometric m- and a-surfaces are valid for both plots.

Figure 7. SEM top view image (a) of a SAG GaN NC with smooth segments of m- (green/light) and a-planes (blue/dark). The Wulff-construction in the GaN c-plane is shown in (b). The relative length of the vectors in the m- and a-direction is related to the corresponding surface energy densities. The equilibrium crystal shape cross section is deduced by the vectors normal intersections, which is given by the linear polygon.

relation is given [30, 31]. Without this assumption the growth of the NCs could proceed either in the a- or c-direction due to the very similar broken bond densities of the corresponding surfaces according to figure 6. The formation of semipolar surfaces can be excluded here due to their large broken bond densities compared with the a-, m- and c-surfaces. Therefore the crystal will form the non-polar sidewalls and the polar surface at the tip of the NC. Accordingly this morphology is experimentally observed for the case of N-polar self-organized NCs.
To further elucidate the description of the NC morphology within a thermodynamic framework, the ratio of the segment lengths $l$ of $m$- and $a$-plane sidewalls per NC was measured for several SAG NCs. This is exemplarily shown in figure 7(a). The expected ratio for GaN can be derived from the construction of the Wulff-plot [47] in the c-plane, as shown in figure 7(b). The theoretical surface energy densities $\sigma_{(1\overline{1}0\overline{0})} = 118$ and $\sigma_{(1\overline{1}2\overline{0})} = 123$ meV Å$^{-2}$ for $m$- and $a$-plane have been used [42]. From our experiments we estimate a value of the ratio $s = l_m/l_a = 2.2(4)$ which is in good agreement with the expected value of $s = 2.4$.

4. Conclusions

In conclusion the growth of ordered GaN NCs by SAG MBE on laser ablated pre-patterned GaN(0001) templates has been demonstrated. The obtained morphology with semipolar facets at the tip of Ga-polar GaN NCs has been discussed within the framework of equilibrium thermodynamics, which provides a consistent picture also for the growth of N-polar NCs. The structural investigation reveals almost defect-free semipolar $\{1\overline{1}02\}$ GaN facets at the top of the NCs, which are thus suited for the production of templates for nanoscale semipolar GaN-based heterostructures. Almost no polarization discontinuity is expected for such In$_x$Ga$_{1-x}$N/GaN interfaces. The SAG is shown to be Ga diffusion-assisted with contributions from the substrate and the NC sidewalls.

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