

Growth of High Mobility InN Film on Ga-Polar GaN Substrate by Molecular Beam Epitaxy for Optoelectronic Device Applications

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The fabrication of high-speed electronic and communication devices has rapidly grown the demand for high mobility semiconductors. However, their high cost and complex fabrication process make them less attractive for the consumer market and industrial applications. Indium nitride (InN) can be a potential candidate to fulfill industrial requirements due to simple and low-cost fabrication process as well as unique electronic properties such as narrow direct bandgap and high electron mobility. In this work, 3 µm thick InN epilayer is grown on (0001) gallium nitride (GaN)/Sapphire template under Inrich conditions with different In/N flux ratios by molecular beam epitaxy. The sharp InN/GaN interface monolayers with the In-polar growth are observed, which assure the precise control of the growth parameters. The directly probed electron mobility of 3610 cm² V^{-1} s⁻¹ is measured with an unintentionally doped electron density of 2.24×10^{17} cm⁻³. The screw dislocation and edge dislocation densities are calculated to be 2.56×10^8 and 0.92×10^{10} cm⁻², respectively. The step-flow growth with the average surface roughness of 0.23 nm for $1 \times 1 \,\mu\text{m}^2$ is confirmed. The high quality and high mobility InN film make it a potential candidate for high-speed electronic/optoelectronic devices.

1. Introduction

The demand for high-speed communications has seen tremendous growth recently with networks growing from 4th generation (4G) to 5th generation (5G) and beyond. The next generation of the network is expected with operational frequencies in the Terahertz range.^[1] These devices require high switching speed, high output power, and operation at higher temperatures. The indium nitride (InN) has shown very promising properties due to its narrow and direct bandgap (0.64 eV), low electron effective mass $(0.04 m_0)$, high mobility and high steadystate drift velocity (5 \times 10⁷ cm s⁻¹).^[2-4] These unique properties make it attractive for high speed electronic and infrared optoelectronic device applications.^[5] The theoretical reported electron mobility of InN is 14000 cm² V⁻¹ s⁻¹ at 300 K, making

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it suitable for terahertz device applications.^[6] A lot of advancement have been made to improve the electron mobility through different approaches. However, the experimentally achieved mobility of InN is still far behind the theoretical estimation.^[7-10] The mobility is limited due to carrier scattering phenomena occurring at the scattering centers generated from the crystal imperfections in the hetero-epitaxial InN layer.^[11,12] The growth of high quality InN is challenging due to impurity incorporation, low dissociation temperature, narrow growth window and lack of suitable substrate.^[13-20] There are several methods to deposit InN epitaxial layer such as metal-organic chemical vapor deposition, pulsed vapor deposition, sputtering, hydride vapor phase epitaxy and molecular beam epitaxy (MBE).^[19,21-28] Among them, MBE is the most sophisticated growth technique, which can grow high quality InN structure under precise control. However, the InN films grown by MBE still have high electron concentration and low electron mobility. It has been investigated that the threading dislocations (TDs) are the major cause of limiting the low field mobility at room temperature and higher electron concentration.^[29-40] The nitrogen (N) and In-vacancies at the dislocation sites in InN create coulomb scattering centers, screening the carrier motion and ultimately decreasing the bulk electron mobility.^[41–43] Therefore, the reduction of the TD is crucial to achieve high electron mobility. Similarly, the impurities can be controlled by using MBE, which also greatly affect the electron mobility.^[44] Another important parameter effecting the dislocation density and electron mobility of InN is lattice mismatch with substrate. The growth phenomenon is severely affected by the choice of substrate. The lack of suitable substrate with lattice mismatch and thermal mismatch at the hetero-interface are the main factors of dislocations generation in InN-epilayer.^[45] The application of In-rich growth condition has been reported to improve the quality of InN due to the higher diffusion barrier of N-adatoms compared to In-adatoms. The growth of InN can also be improved by raising the substrate temperature, which enhances diffusion lengths and the two dimensional growth.^[46,47] However, it is difficult to achieve thick InN epilayer using this method as growth cannot proceed for a long time. The reason is that the In-atom accumulate and form droplets on the surface shortly after growth process starts, which cannot be evaporated under



growth temperature limits. These In-metallic droplets sequentially enlarge their size and absorb more heat due to low specific heat compared to InN epilayer.^[48] As the temperature of droplets exceeds the dissociation limit temperature of InN (500 °C), it leads to dissociation of InN under layers, hence reverses (decrease) the growth rate.^[49] One of the ways for continue growth is to apply N-radical beam.^[50] However it results in poor crystal quality and rough surface, which does not fulfill the requirement for device fabrication. Therefore, it is important to investigate new methodologies to limit the imperfections to get defect-free InN.

In this paper, we have optimized the In/N flux ratio for the growth of \approx 3 µm thick InN epilayer at \approx 500 °C by MBE with low TD, high electron mobility, and smooth surface morphology. gallium nitride (GaN)/Sapphire template has been chosen due to its low cost, large area, and lowest possible (≈11%) lattice mismatch for InN epitaxial layer. The sharp interface of InN/GaN and In-polarity observed by scanning transmission electron microscope (STEM) confirmed that the growth conditions were highly optimized. Low screw dislocation (SD) and edge dislocation (ED) density of 2.56×10^8 and 0.92×10^{10} cm⁻², respectively, were verified by high-resolution x-ray diffraction (HR-XRD). The step flow growth with root mean square (RMS) of 0.2 nm and average step height (~570 pm) equal to c lattice constant of InN was observed by atomic force microscope (AFM). High electron mobility of 3610 cm² V⁻¹ s⁻¹ was measured by directly probed hall characterization. The InN-epilayer also shows good optoelectronic behavior in dark as well as under the illumination of 532 nm laser with photo responsivity of 195 mA W⁻¹ and specific detectivity of 1.78×10^9 Jones.

2. Experimental Section

InN epilayer was grown by plasma-assisted-MBE. The schematic structure of the grown InN layer is given in **Figure 1a**. The (0001) GaN/sapphire template was used as a substrate with a 4.7 um GaN layer. After loading the substrate in the MBE growth chamber with a base pressure of 10^{-10} torr, the substrate surface was cleaned by degassing at 500 °C for 30 min. The reflection high energy electron diffraction (RHEED) pattern



Figure 1. a) Schematic structure of InN epilayer grown on GaN template. b) Growth condition for InN film. The growth starts with degas and regrowth of fresh GaN thin film over GaN template (GaN/Al₂O₃). The growth temperature for InN was ramped down (from boundary temperature) by 0.05 °C min⁻¹ till the end of growth. The short streaky RHEED pattern for GaN turned to long streaks during InN growth. The whole growth was performed under Radio Frequency Plasma power of 400 W with the Nitrogen flow of 1.2 sccm.



was observed to be dim streaky. The RHEED patterns along $[11\overline{2}0]$ azimuth is shown in Figure 1b. Later, the temperature was raised to 680 °C to regrow a fresh GaN laver. The nitrogen flow rate of 1.2 sccm with a forward power of 400 W was used to supply active nitrogen atoms (N^*) using a radio frequency plasma source. $A \approx 100 \text{ nm}$ GaN regrowth layer was deposited under Ga-rich growth condition, which was confirmed with a bright streaky pattern of RHEED. The temperature was ramped down to an optimized (thermocouple) temperature of 380 °C for InN growth, which is equal to InN boundary temperature (500 °C). It should be noted that the actual temperature of the growth surface is higher than the temperature measured by the control panel of MBE. There are different parameters, which may affect the actual temperature, for example, the distance of the thermocouple (used to measure the growth temperature) from the substrate, the heat absorbed by the sample holder, the specific heat of the sapphire, GaN layers, and the heat absorbed by the metallic In-droplets. The actual boundary temperature was measured by testing the InN dissociation temperature, which was observed by the change in the RHEED pattern while growing under In/N > 1. The growth temperature was increased step by step from 360 to 380 °C and a bright streaky RHEED pattern was observed. The pattern turned into bright spotty when the growth temperature was exceeded from 380 °C. This spotty RHEED pattern was the experimental evidence for the dissociation of grown InN, while the InN dissociation temperature has been reported to be 500 °C in literature.^[51] The InN epilayer was grown under 1.12, 1.20, 1.28, 1.36 In/N flux ratio with an average thickness of 3 µm by boundary temperaturecontrolled epitaxy method.^[7]

The beam equivalent pressure for the In-flux corresponding to different cell temperatures was confirmed by the residual gas analyzer, which is presented in Figure S1 (Supporting Information). This was measured by the difference in the flux with shutter open and shutter close conditions. Illustration of growth conditions for the degassing, GaN growth parameters, InN growth with growth chamber pressure, nitrogen flow, radiofrequency plasma power, the In/Ga Knudsen cell tip, and bottom temperatures are given in Figure S2 (Supporting Information).

After the growth of 100 nm InN layer, the substrate temperature was ramped down with the rate of 0.05 °C min⁻¹ till the end of growth to avoid InN dissociation as both the size and the temperature of the formed In-droplet increase with the growth time. The long and smooth bright streaks of RHEED were observed during the whole InN layer growth process given in Figure 1b, which confirmed the two dimensional layer-by-layer growth. The RHEED patterns of InN along $[11\overline{2}0]$ azimuth grew in +c direction under In-rich growth condition with In/N flux ratio of 1.12,1.20, 1.28, and 1.36 are presented in Figure S3 (Supporting Information). The streak brightness decreases as the metallic In-flux increases. The brightness was maximum at In/N flux ratio of 1.12, while almost dark for the ratio of 1.36. The samples were cleaned with hydrochloric acid to check the surface morphology. The sample preparation and important characterization steps are shown in Figures S4-S7 (Supporting Information). After the successful growth and characterization of InN epilayer, the Ag electrodes were deposited on the top surface to investigate the optoelectronic behavior of InN as an active layer of the device.



The Bruker high-resolution AFM was employed to investigate the RMS of surface roughness. The field emission FEI Nova Nano SEM 430 was used for cross-section imaging to measure the InN epilayer thickness. The FEI Titan Themis Cubid G2-300 aberration-corrected STEM was used for the high-angle annular dark-field (HAADF), integrated differential phase contrast (IDPC) imaging, and energy dispersive X-Ray analysis (EDX) of the grown samples. The samples for TEM were prepared by the conventional Ar-Ion milling process.^[52] For sample preparation, the InN samples were cut into small pieces and were pasted on Si substrate, which was subsequently bonded face to face (InN to InN). The sample was then mechanically ground and polished, which was finally thinned by using GATAN 691 precision ion polishing system. The beam energy was used to be 4.5 keV at the gun angle of ± 8 , which was finally reduced to 2 keV at the gun angle of ± 4 . The rotation speed was set to be 3 rpm. The sample preparation process is presented in Figure S7 (Supporting Information). The ED and SD were calculated indirectly using PANalytical X'Pert Pro MRD HR-XRD. The directly probed electron mobility was measured by the Vander-Pauw method using the Accent HL5500PC Hall Effect measurement system. Electrical and optical characteristics of the device were measured using a semiconductor parameter analyzing system (Keithley 4200-SCS) in dark and under the illumination of a 532 nm laser.

3. Results and Discussion

The growth control of the first few monlolayers at the InN/GaN interface is critical as this is the highest strain affected region due to lattice mismatch. The poor growth control will be detrimental to the subsequent epitaxial layer as most of the dislocations are generated from the interface. These dislocations not only degrade the crystal quality in the bulk but also act as doping constituents leading to high unintentional electron concentration. So the accurate growth condition, i.e., growth temperature, In/N flux ratio are very crucial parameters. The stability of plasma power and the amount of radical nitrogen beam are very critical according to which, the Influx should be optimized. This becomes more challenging for In/N > 1 condition, which is the key parameter for 2D growth. After the growth of multiple samples, we were able to optimize the best growth condition to achieve a high quality InN thick film with very low dislocation density.

The TEM was performed to directly investigate the dislocation density.^[53] It can be observed in **Figure 2**a that the EDs have a non-uniform distribution along the growth direction. The ED density is high near the InN/GaN interface due to a large lattice mismatch (≈11%) between the InN and GaN. These dislocations are annihilated by clustering, which leads to a reduction in EDs. These ED are driven to the sample surface having which can be directly observed in Figure 2a. The sharp InN/GaN interface can be seen in the STEM- HAADF image in Figure 2b. The compositional distribution from the top surface of the InN film (≈10 nm) was investigated by x-ray photoelectron spectroscopy (Figure S8, Supporting Information), which shows the binding energy peaks of In- and N-subshells.







Figure 2. The dislocations reduction in InN by precise growth condition a) TEM image of InN sample grown under In/N ratio of 1.36. The dislocation density reduces along the growth direction. b) The zoom-in information at the atomic resolution level by HAADF image of InN/GaN interface is confirming the good growth start over the GaN substrate. The inset with 1 nm IDPC image showing the hexagonal InN crystal symmetry with growth in Inpolar direction (In-atoms (pink color) and N-atoms (green color), c) EDS of InN/GaN Interface showing In, Ga, and d) N content in the grown sample.

The oxygen and carbon peaks are observed as these atoms are adsorbed on the InN surface from the ambient environment.

Further investigation of the interface was also performed with high resolution imaging at the atomic level. The wurtzite structure symmetry is obvious where the smaller dark Gaatomic layer is visibly distinct from the relatively larger and brighter In-atomic layer as given in Figure 2b. The reason for this highly compact atomic layer is the high growth temperature, which increases the diffusion length of incident Inadatoms and makes it well incorporated on the GaN surface. The IDPC imaging was performed to confirm the In-polar (0001) growth direction, the bright circles along with small dim tails are the high definition real-time images of In- and N-atoms respectively. The pink and green circles are used to easily distinguish the In- and N-atoms as shown in the inset image of Figure 2b. The EDX was performed to investigate the elemental distribution of In-, Ga-, and N-atoms. Figure 2c,d shows the EDX of the image given in Figure 2a. The HAADF, IDPC, and EDX confirmed that the high strcture quality InN monolayers were grown on GaN with a sharp interface in the In-polar direction.

The thickness and crystal quality (dislocation density) of the InN epilayer were investigated thoroughly. The SEM image of the 3 μ m thick InN layer with the 4.7 μ m thick GaN buffer layer is shown in **Figure 3**a. The crystal quality was investigated by HR- XRD. The HR-XRD patterns for the optimal sample including (002) and (102) peaks are shown in Figure S9a,b

(Supporting Information). The XRD pattern for the epitaxial layer of InN grown on Ga-polar GaN substrate is shown in Figure S10 (Supporting Information). The full width at half maximum (FWHM) of (002) and (102) rocking curves (RC) were measured to calculate SD and ED using the mosaic model.^[54] According to this model, it is assumed that the film is composed of blocks that are tilted or twisted. Based on this model, the density of the screw and edge dislocations can be calculated by classical equations. For screw dislocations, $\rho_{\text{screw}} = \beta_{\text{tilt}}^2/4.35 b_{\text{screw}}^2$ and for edge dislocations, $\rho_{\rm edge} = \beta_{\rm twist}^2/4.35 b_{\rm edge}^2$, can be used, where $eta_{
m screw}$ and $eta_{
m tilt}$ are the measurement of tilt and twist. The b_{screw} and b_{tilt} is the length of the corresponding Burgers vector, 0.57033 nm and 0.35378 nm, respectively.^[55] The value of $\beta_{\rm tilt}$ and β_{screw} can be obtained by $\beta_{\text{hkl}}^2 = (\beta_{\text{tilt}} \cos \chi)^2 + (\beta_{\text{twist}} \sin \chi)^2$, where, $\beta_{\rm hkl}$ is FWHM of the rocking curve is angle between (102) and(002)plane.

The measured values of (002) and (102) RC and corresponding SD and ED under different In/N flux (Figure S11, Supporting Information) are given in **Table 1**. It can be observed that the density of both types of dislocations decreases with the increase in the In/N ratio, which are important factors for limiting the electron mobility in the bulk InN structure. The higher In/N flux provides a higher number of In-atoms, which are more mobile on the growth surface with more atomic incorporation and covering more surface area, hence making more compact layers and leaving very few dislocation sites. Figure 3b provides a more clear understanding of the relation between







Figure 3. The bulk and structural investigation of InN sample grown under In/N ratio of 1.36. a) Cross-section scanning electron microscopy image of InN film showing 3 μ m InN film over 4.7 μ m GaN. b) The screw (002) and edge (102) dislocation densities for the samples grown under In/N flux ratios of 1.12, 1.20, 1.28, and 1.36, respectively.

Table 1. Electron mobility and dislocation density of samples grown under different In/N flux.

No.	In/N ratio	Temperature [°C]	FWHM		SD [cm ⁻²]	ED [cm ⁻²]	RMS [nm]	Electron concentration [cm ⁻³]	Mobility [cm ² V ⁻¹ s ⁻¹]
			002	102					
1	1.12	380–360	756	1656	9.31 × 10 ⁸	2.23×10^{10}	0.75	6.10×10^{17}	2680
2	1.20	380–360	684	1512	7.72×10^{8}	1.87×10^{10}	0.39	5.41×10^{17}	3060
3	1.28	380–360	540	1332	4.75×10^{8}	1.48×10^{10}	0.34	$3.74 imes 10^{17}$	3240
4	1.36	380–360	396	1044	2.56×10^{8}	0.92×10^{10}	0.23	2.24×10^{17}	3610

the In/N flux and dislocation densities. From these observations, it can be concluded here that high In/N flux is responsible to suppress the dislocation formation during the growth.

The surface morphology is the key requirement for the electronic and optoelectronic device fabrication as the interface of the subsequent layer for the device fabrication (i.e., InGaN/InN) will be formed above this surface. The details of the surface roughness of the InN epilayer grown under In/N ratio of 1.36 investigated by AFM are given in **Figure 4**a. The AFM images for the samples grown under different In/N ratio (Figure S12, Supporting Information) confirm that the surface smoothness increases with the increase in the In/N flux ratio. The small grain boundaries tend to grow larger with higher In-flux. A smooth surface was achieved without any grain boundaries in the case of the In/N flux ratio of 1.36. The clear terrace on the surface can be observed, which confirms the step flow growth with the RMS of 0.23 nm. To further investigate the growth step details, the linear graph between the surface and height is drawn, where the steps ending up at the sharp edges can be observed. The average height for each step can be seen in Figure 4b, which is \approx 570 pm equal to *c* lattice constant of InN. This shows that the entire InN layer was smoothly grown in the [0001] direction and can be used for device applications.

Finally, the electron mobility and concentration were measured by directly probed Hall Effect system using the van der Pauw technique operated at 0.5 T magnetic field. The



Figure 4. Surface morphology of InN film grown under In/N ratio of 1.36. a) Atomic force microscopy image of InN with the surface area of $1 \times 1 \mu m^2$ with the average surface roughness of 0.23 nm. b) Linear profile of surface roughness showing step height nearly equal to the unit cell of InN (0.571 nm).







Figure 5. Electronic properties measurements of InN films by van der paw method. a) Room temperature Hall mobility of InN films grown under In/N flux ratio of 1.12, 1.20, 1.28, and 1.36, respectively. b) Low-temperature (100–300 K) mobility measurement of InN film grown under 1.36 In/N growth conditions.

as-measured electron mobility of different samples with unintentional electron doping is shown in **Figure 5**a. The results show that the electron mobility is directly proportional to the In/N flux, i.e., the higher is the In/N flux, the more is the electron mobility up to a certain limit. Crossing this limit leads to a high accumulation of In-droplets, which first decreases the growth rate and later causes the dissociation of grown InN epilayer under these droplets due to heat accumulation. The cracks on the InN surface were observed and the layer peeled off for the samples grown beyond 1.36 In/N flux ratio (Figure S6, Supporting Information). The low-temperature electron mobility was also measured at a temperature from 100 to 300 K, which is presented in Figure 5b. The maximum electron mobility at 300K was measured to be 3610 cm² V⁻¹ s⁻¹ with unintentional electron doping of 2.24×10^{17} cm⁻³.

In order to investigate the reason for obtaining high mobility InN under the In-rich condition, we carried out XRD test on samples grown under different III/V and estimated the dislocation density by using the mosaic model. After using this model, we got the screw and edge dislocation densities of the samples grown under different III/V growth conditions. The details of the screw and edge dislocations with electron concentration and mobility under different growth conditions are presented in Table 1. Therefore, the relationship between III/V flux ratio with electron concentration and mobility under In-rich conditions is investigated, where we found that during InN growth under In-rich, the In-atoms can effectively block the incorporation of impurities, hence reducing the scattering effects from the ionized impurity centers.

It is also observed that surface-enriched with In-droplets can effectively hinder the incorporation of impurities, which can greatly reduce the effects of ionized impurity scattering, thereby contributing to an increase in the mobility of InN.^[14] When the InN surface is not covered with In-droplets, the Hydrogen and Oxygen impurities in the environment can easily enter inside the epitaxially grown InN film. However, with the surface accumulated with In-metallic droplets, most of the Hydrogen and Oxygen atoms are blocked by the In-metal, hence, their concentration inside the grown film is suppressed. Furthermore, the relationship between III/V and dislocation density reveals that the dislocations in InN are partially charged under the condition of In-rich growth. These partially charged dislocations greatly reduce the effect of electron scattering from dislocation centers on mobility and thus are more favorable for obtaining

high electron mobility InN epitaxial films under the In-rich growth conditions. So, a high mobility InN epitaxial film with electron mobility of 3610 cm² V⁻¹ s⁻¹ and an electron concentration of 2.24×10^{17} cm⁻³ was realized under the In/N condition of 1.36 at room temperature. The low-temperature Hall test shows that the sample can show maximum mobility up to 4180 cm² V⁻¹ s⁻¹ at 100 K, with the corresponding electron concentration of 5.78×10^{17} cm⁻³.

In order to investigate the optoelectronic properties of the InN film incorporated as an active layer of a photodetector, the I-V characteristics of the device (Ag/InN/Ag) are measured in the dark and under laser illumination of 532 nm. The schematic cross-sectional view and I-V characteristics of the device are shown in Figure 6a,b. It can be observed that the photocurrent is increased with an increase of illumination intensity, revealing the efficient photogenerated carriers through the InN layer. The transportation of efficient excitons will greatly be influenced by the difference between the electron affinity and potential energy of the photoactive material. The high electron mobility of the InN film can enhance the carrier transportation under an applied electric field. Therefore, the dissociation of photogenerated charge carriers and their transportation can easily be enhanced at low applied bias. The optoelectronic properties of the device can be measured by following the previously reported optoelectronic devices.^[56-58] Photoresponsivity (R) and specific detectivity (D*) are the major key factors to evaluate the performance of the photodetectors, and they are expressed as

 $R = \frac{I_{\text{photo}}}{P_{\text{ill}}}$, where P_{ill} is the incident illumination power intensity, and $D^* = R \times \sqrt{\frac{A}{2qI_{\text{dark}}}}$, where *q* denotes the electronic charge,

and *A* is the effective area of conductive channel for photodetector. As a result, *R* of 195 mA W⁻¹ and *D** of 1.78×10^9 Jones were obtained with 532 nm laser under illumination power of 1 mW cm⁻².

4. Conclusions

In summary, we have grown $\approx 3 \ \mu m$ thick InN epitaxial film by MBE under In-rich growth conditions. It is observed that the higher In/N growth condition leads to high quality InN crystal structure in a bulk with a smooth surfer over the growth termination. The higher In-atomic ratio not only improves the







Figure 6. Photoconduction measurement in InN film grown under In/N flux ratio of 1.36. a) Schematic diagram of the device structure with silver contacts (Ag/InN/Ag). b) Current–voltage characteristics under dark and 532 nm laser with different illumination powers.

crystal quality but also suppresses the impurity incorporation during the growth, which results in higher electron mobilities. The high quality InN film with very low SD and ED density $2.56\times10^8\,\text{cm}^{-2}$ and $0.92\times10^{10}\,\,\text{cm}^{-2}$ respectively is confirmed by HR-XRD. These low dislocation densities are responsible for the reduction in unintentional electron doping and hence lead to higher electron mobility. The Maximum mobility is measured to be 3610 cm² V⁻¹ s⁻¹ with an unintentional electron concentration of 2.24×10^{17} cm⁻³. The smooth step flow growth of the surface is confirmed by AFM with RMS of 0.2 nm. The growth conditions described in this research work are effective approaches to achieve low electron concentration, high electron mobility, and smooth surface InN thick films, which have the potential to be applied in the high speed communication devices, highly sensitive infrared detectors, and optoelectronic devices.

Supporting Information

Supporting Information is available from the Wiley Online Library or from the author.

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Conflict of Interest

The authors declare no conflict of interest.

Data Availability Statement

The data that support the findings of this study are available in the supplementary material of this article.

Keywords

high electron mobility, InN epilayer, low dislocation density, molecular beam epitaxy

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- J. Ben, X. Liu, C. Wang, Y. Zhang, Z. Shi, Y. Jia, S. Zhang, H. Zhang, W. Yu, D. Li, X. Sun, Adv. Mater. 2021, 33, 2006761.
- [2] J. Wu, W. Walukiewicz, W. Shan, K. M. Yu, J. W. A. Iii, S. X. Li, E. E. Haller, H. Lu, W. J. Schaff, J. Appl. Phys. 2003, 94, 4457.
- [3] I. Vurgaftman, J. R. Meyer, Nitride Semiconductor Devices: Principles and Simulation || Electron Bandstructure Parameters, 2007, WILEY-VCH Verlag GmbH & Co. KGaA, Weinheim 2007, pp. 13–48, Ch. 2.
- [4] H. Ünlü, N. J. M. Horing, J. Dabrowski, Low-Dimensional and Nanostructured Materials and Devices: Properties, Synthesis, Characterization, Modelling and Applications, Springer International Publishing, Switzerland 2016, pp. 1–47.
- [5] Timothy David Veal, Christopher F. McConville, W. J. Schaff, Indium Nitride and Related Alloys, 1st ed., Vol. 1, CRC Press, USA 2017, p. 645.
- [6] V. M. Polyakov, F. Schwierz, Appl. Phys. Lett. 2006, 88, 032101.
- [7] X. Wang, S. Liu, N. Ma, L. Feng, G. Chen, F. Xu, N. Tang, S. Huang, K. J. Chen, S. Zhou, B. Shen, *Appl. Phys. Express* **2012**, *5*, 015502.
- [8] H. Liu, X. Wang, Z. Chen, X. Zheng, P. Wang, B. Sheng, T. Wang, X. Rong, M. Li, J. Zhang, X. Yang, F. Xu, W. Ge, B. Shen, *Appl. Phys. Lett.* **2018**, *112*, 162102.
- [9] R. E. Jones, S. X. Li, E. E. Haller, H. C. M. van Genuchten, K. M. Yu, J. W. Ager, Z. Liliental-Weber, W. Walukiewicz, H. Lu, W. J. Schaff, *Appl. Phys. Lett.* **2007**, *90*, 162103.
- [10] L. Chen, S. Sheng, B. Sheng, T. Wang, L. Yang, B. Zhang, J. Yang, X. Zheng, Z. Chen, P. Wang, W. Ge, B. Shen, X. Wang, *Appl. Phys. Express* **2022**, *15*, 011004.
- [11] K. Wang, Y. Cao, J. Simon, J. Zhang, A. Mintairov, J. Merz, D. Hall, T. Kosel, D. Jena, *Appl. Phys. Lett.* **2006**, *89*, 162110.
- [12] Y. Zhang, X. Wang, X. Zheng, G. Chen, D. Ma, F. Xu, N. Tang, W. Ge, B. Shen, Appl. Phys. Express 2013, 6, 021001.
- [13] X. Wang, S.-B. Che, Y. Ishitani, A. Yoshikawa, Phys. Status Solidi C 2006, 3, 1561.
- [14] C. S. Gallinat, G. Koblmüller, J. S. Brown, J. S. Speck, *J. Appl. Phys.* **2007**, *102*, 064907.
- [15] W. C. Chen, S. Y. Kuo, F. I. Lai, W. T. Lin, C. N. Hsiao, *Thin Solid Films* 2013, 529, 169.

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- [16] H. J.-Y. Chen, Y.-Z. Su, D.-L. Yang, T.-W. Huang, I.-S. Yu, Surf Coat Technol. 2017, 324, 491.
- [17] K. Ishii, S. Yagi, H. Yaguchi, Phys. Status Solidi B 2017, 254, 1600542.
- [18] Y. H. Wang, W. L. Chen, M. F. Chen, Physica E 2009, 41, 1746.
- [19] U. Bashir, Z. Hassan, N. M. Ahmed, A. Oglat, A. S. Yusof, Mater. Sci. Semicond. Process. 2017, 71, 166.
- [20] P. Jantawongrit, S. Sanorpim, H. Yaguchi, M. Orihara, P. Limsuwan, J. Semicond. 2015, 36, 083002.
- [21] H. F. Liu, W. Liu, D. Z. Chi, CrystEngComm 2016, 18, 1871.
- [22] J. Mickevičius, D. Dobrovolskas, T. Steponavičius, T. Malinauskas, M. Kolenda, A. Kadys, G. Tamulaitis, *Appl. Surf. Sci.* 2018, 427, 1027.
- [23] J.-H. Ha, J. Wang, W.-J. Lee, Y.-J. Choi, H.-Y. Lee, J.-G. Kim, H. Harima, J. Korean Phys. Soc. 2015, 66, 994.
- [24] R. Togashi, Q. T. Thieu, H. Murakami, Y. Kumagai, Y. Ishitani, B. Monemar, A. Koukitu, J. Cryst. Growth **2015**, 422, 15.
- [25] U. Bashir, Z. Hassan, N. M. Ahmed, N. Afzal, J. Electron. Mater. 2018, 47, 4875.
- [26] A. Kobayashi, M. Oseki, J. Ohta, H. Fujioka, *Phys. Status Solidi B* 2018, 255, 1700320.
- [27] C. Kurşungöz, E. Uzcengiz Şimşek, R. Tuzaklı, B. Ortaç, Appl. Phys. A: Mater. Sci. Process. 2017, 123, 209.
- [28] R. D. Vispute, J. Narayan, H. Wu, K. Jagannadham, J. Appl. Phys. 1995, 77, 4724.
- [29] V. Darakchieva, T. Hofmann, M. Schubert, B. E. Sernelius,
 B. Monemar, P. O. Å. Persson, F. Giuliani, E. Alves, H. Lu,
 W. J. Schaff, *Appl. Phys. Lett.* 2009, *94*, 022109.
- [30] C. S. Gallinat, G. Koblmüller, J. S. Speck, Appl. Phys. Lett. 2009, 95, 022103.
- [31] L. F. J. Piper, T. D. Veal, C. F. McConville, H. Lu, W. J. Schaff, Appl. Phys. Lett. 2006, 88, 252109.
- [32] X. Wang, S.-B. Che, Y. Ishitani, A. Yoshikawa, Appl. Phys. Lett. 2007, 90, 151901.
- [33] V. Lebedev, V. Cimalla, T. Baumann, O. Ambacher, F. M. Morales, J. G. Lozano, D. González, J. Appl. Phys. 2006, 100, 094903.
- [34] M. Tangi, A. De, S. M. Shivaprasad, J. Appl. Phys. 2018, 123, 015701.
- [35] K. Viswanathan, S. Chandrasekar, J. Appl. Phys. 2014, 116, 245103.
- [36] Y. Takei, T. Nakayama, J. Cryst. Growth 2009, 311, 2767.
- [37] J. Kioseoglou, Phys. Status Solidi A 2013, 210, 204.
- [38] N. Miller, E. E. Haller, G. Koblmüller, C. Gallinat, J. S. Speck, W. J. Schaff, M. E. Hawkridge, K. M. Yu, J. W. Ager, *Phys. Rev. B* 2011, 84, 075315.

- [39] D. Cherns, C. G. Jiao, Phys. Rev. Lett. 2001, 87, 205504.
- [40] V. Cimalla, V. Lebedev, F. M. Morales, M. Niebelschütz, G. Ecke, R. Goldhahn, O. Ambacher, *Materialwiss. Werkstofftech.* 2006, 37, 924.
- [41] A. Janotti, C. G. Van de Walle, Appl. Phys. Lett. 2008, 92, 032104.
- [42] C. Stampfl, C. G. Van de Walle, D. Vogel, P. Krüger, J. Pollmann, *Phys. Rev. B* 2000, 61, R7846.
- [43] D. C. Look, H. Lu, W. J. Schaff, J. Jasinski, Z. Liliental-Weber, Appl. Phys. Lett. 2002, 80, 258.
- [44] X. Wang, A. Yoshikawa, Prog. Cryst. Growth Charact. Mater. 2004, 48–49, 42.
- [45] U. Bashir, Z. Hassan, N. M. Ahmed, J. Mater. Sci.: Mater. Electron. 2017, 28, 9228.
- [46] G. B. Cross, Z. Ahmad, D. Seidlitz, M. Vernon, N. Dietz, D. Deocampo, D. Gebregiorgis, S. Lei, A. Kozhanov, J. Cryst. Growth 2020, 536, 125574.
- [47] Z. Benzarti, T. Sekrafi, A. Khalfallah, Z. Bougrioua, D. Vignaud, M. Evaristo, A. Cavaleiro, J. Alloys Compd. 2021, 885, 160951.
- [48] M. Azadmand, L. Barabani, S. Bietti, D. Chrastina, E. Bonera, M. Acciarri, A. Fedorov, S. Tsukamoto, R. Nötzel, S. Sanguinetti, *Sci. Rep.* 2018, *8*, 11278.
- [49] X. Zheng, H. Liang, P. Wang, X. Sun, Z. Chen, T. Wang, B. Sheng, Y. Wang, L. Chen, D. Wang, X. Rong, M. Li, J. Zhang, X. Wang, Superlattices Microstruct. 2018, 113, 650.
- [50] T. Yamaguchi, Y. Nanishi, Appl. Phys. Express 2009, 2, 051001.
- [51] C. S. Gallinat, G. Koblmüller, J. S. Brown, S. Bernardis, J. S. Speck, G. D. Chern, E. D. Readinger, H. Shen, M. Wraback, *Appl. Phys. Lett.* 2006, 89, 032109.
- [52] Á. Barna, B. Pécz, J. Electron Microsc. Tech. 1991, 18, 325.
- [53] G. Capellini, M. D. Seta, Y. Busby, M. Pea, F. Evangelisti, G. Nicotra, C. Spinella, M. Nardone, C. Ferrari, J. Appl. Phys. 2010, 107, 063504.
- [54] V. Srikant, J. S. Speck, D. R. Clarke, J. Appl. Phys. 1997, 82, 4286.
- [55] I. Yonenaga, Y. Ohkubo, M. Deura, K. Kutsukake, Y. Tokumoto, Y. Ohno, A. Yoshikawa, X. Q. Wang, AIP Adv. 2015, 5, 077131.
- [56] M. Sulaman, S. Y. Yang, Y. R. Jiang, Y. Tang, B. S. Zou, Nanotechnology 2017, 28, 505501.
- [57] M. Sulaman, Y. Song, S. Y. Yang, Q. Hao, Y. Zhao, M. Y. Li, M. I. Saleem, P. V. Chandraseakar, Y. R. Jiang, Y. Tang, B. S. Zou, *Nanotechnology* **2019**, *30*, 465203.
- [58] M. Sulaman, Y. Song, S. Y. Yang, M. Y. Li, M. I. Saleem, P. V. Chandraseakar, Y. R. Jiang, Y. Cui, Q. Hao, B. S. Zou, ACS Appl. Mater. Interfaces 2020, 12, 26153.