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Investigations on morphology, growth mode and indium incorporation in MOCVD grown InGaN/n-GaN heterostructures

K. Prabakaran^{1*}, M. Jayasakthi¹, S. Surender¹, S. Pradeep¹, S. Sanjay¹, R. Ramesh³, M. Balaji⁴, K. Baskar^{1,2*}

¹Crystal Growth Centre, Anna University, Chennai, India

²Manonmaniam Sundaranar University, Tirunelveli, India

³Department of Micro and Nanosciences, Aalto University, Finland

⁴Department of Energy, University of Madras, Chennai, India

*Corresponding Author:

Dr. K. Prabakaran

Crystal Growth Centre

Anna University, Chennai-600 025

Tamilnadu, India

E-Mail address: karanphy07@gmail.com (K. Prabakaran)

drbaskar2009@gmail.com (K. Baskar)

Abstract

InGaN layers were grown on n-GaN/GaN/sapphire template using metal organic chemical vapor deposition system by varying Indium (In) flow rate as 11, 13 and 14 $\mu\text{mol/min}$. The nanoisland growth mode was observed by atomic force microscopy. The thickness and composition of In in the InGaN layers were determined by high resolution X-ray diffraction technique. The composition of In was found to be 15-17% depending on the In flow rate. Photoluminescence from the InGaN layers exhibit multiple peaks ranging between 479 to 657nm, originating from the distribution of In-rich InGaN nanoislands of various sizes giving rise to localized excitons due to the In composition fluctuations. It is interesting to note that, peak emissions at 539 nm, 549 nm and 543 nm showed red and blue shift with varying In flow rate of InGaN nanoisland structures which can be attributed to the quantum confined stark effect and

quantum confinement effect. Hall Effect measurement clearly indicate the variations in the semiconducting behavior of the InGaN nanoislands which influenced the composition fluctuation.

Keywords: Metalorganic chemical vapor deposition, Nitrides, High resolution X-ray diffraction, Atomic Force Microscopy, nanoislands

1. Introduction

InGaN based semiconductors and their nanostructures have attracted strong research due to their optoelectronic applications such as solar cells, light emitting diodes (LED), and laser diodes (LD) [1-3]. InGaN is a semiconductor material which covers a wide range of wavelength from infrared (0.7eV) to ultraviolet (3.42 eV) region by tuning the Indium (In) composition. InGaN has unique properties like wide direct band gap, high absorption coefficient ($\sim 10^5 \text{cm}^{-1}$), high carrier mobility, high drift velocity that can be used in high temperature and radiation [4] which is suitable for the InGaN-based solar cells, LEDs. In order to understand these applications, it is essential to study the high-quality low-defect crystalline layers and to investigate the difficulties in achieving a homogeneous In composition throughout the material. In the past years research groups have focused on InGaN quantum wells (QWs) [5, 6]. However, the efficiency of InGaN QWs decrease intensely towards longer wavelengths. The above problem may be overcome by using Quantum Dots (QDs) or nanoislands in the active region for high In content devices [7, 8]. High In compositions based QDs or nanoislands can minimize the overall strain without any dislocations. However, while increasing the InGaN critical thickness, it may increase the possibility of high dislocation densities [9]. In metal organic chemical vapor deposition (MOCVD), InGaN/GaN based QDs or nanoislands are formed due to the strain induced during Stranski–Krastanov (SK) growth mode. Therefore, the large lattice mismatch of the two

constituents InN and GaN layers play an important role in the growth of InGaN QD or nanoislands [10, 11]. The understanding of luminescence mechanism from the InGaN nanoisland has not been understood clearly. The present work focuses on the improvement of morphological, structural, and electrical properties of the MOCVD grown InGaN layers on n-GaN/GaN/sapphire template. The luminescence mechanism for different TMIn flows in the InGaN nanoisland have been investigated.

2. Experimental Details

InGaN layers were grown on n-GaN/GaN/sapphire template by MOCVD system (AIXTRON 200/4 RF-S). Trimethylgallium (TMGa) and Triethylgallium (TEGa) have been used as the gallium precursor in GaN and InGaN layers. Trimethylindium (TMIn) and ammonia (NH_3) were used as precursors for In and nitrogen. The sapphire substrates were initially heated to 1050°C in hydrogen (H_2) ambient for 10 min to remove the native oxide from the surface. A 25 nm thick GaN nucleation layer (NL) was grown at a temperature of 520°C with V/III ratio of 3453. The grown GaN–NL has been recrystallized at 1000°C for 1 min. The temperature was then ramped to 1010°C for growing n-GaN/GaN epilayer with 2 μm thickness using H_2 as a carrier gas. Silane flow rate for n-GaN layer is around 40 sccm with V/III ratio of 614. This was named as sample a. Further, nitrogen (N_2) as the carrier gas to grow thin InGaN layers of 5nm-6 nm was grown on 2 μm thick n-GaN/GaN/sapphire template at fixed growth temperature and TEGa flow rate for 720°C and 6.2 $\mu\text{mol/min}$ respectively. The TMIn flow rate was varied to 11, 13, 14 $\mu\text{mol/min}$ with V/III ratio of 13000, 11625 and 10078 respectively. Samples were subsequently named as b, c and d. The crystalline quality, composition and thickness of the grown InGaN layers were assessed by PANalytical X'Pert Pro MRD High resolution X-ray Diffractometer (HRXRD) and In content was examined (CasaXPS fitting software) using X-ray

photoelectron spectroscopy (XPS). The surface properties and roughness were studied by Park XE-100 atomic force microscopy (AFM). Photoluminescence (PL) of the samples was measured at room temperature under excitation by 244 nm Ar^+ laser radiation. The electrical properties of the samples were measured using the room temperature hall measurements (ECOPIA, HMS 5000).

3. Results and discussion

3.1 Atomic Force Microscopy

The AFM image shown in Figure. 1(a) represents the surface morphology of n-GaN layers along with its line profile. The terrace spaced steps are clearly observed with a surface roughness of around 0.55 nm. The line profile revealed that the width of the terrace step around 120 nm. It was also observed that the step–terrace formed terminate at the threading dislocations with pure or partial screw dislocations [12]. Figure. 1(b, c, d) represents the AFM images and line profile of InGaN layers grown at 720°C. The RMS roughness of the InGaN surfaces were found to be around 0.49 nm, 0.42 nm and 0.52 nm for samples b, c and d respectively. The InGaN showed flat and nanometer sized island with height fluctuations in all the samples. The width of the island structures observed on surface was found to be distributed randomly between 80 nm to 200 nm and height between 0.8 nm to 1.2 nm for sample b as shown in figure 1 (b). Figure 1(c) shows that there is a uniform island structures with the width and height around 120 nm to 250 nm and 0.6 nm to 1 nm for sample c. The width of sample d is 120 nm to 350 nm and height is ~0.9 nm to 1.6 nm is shown in figure 1 (d). The variations in the islands height can be attributed to the strong confinement and stark effect [13, 14]. The observed nanoislands width was found to be larger than the terrace steps and such type of islands were not observed on the grown n-GaN layer (Figure.1). From the AFM images it was found that the sample c possess uniformly smooth and atomically flat morphology compared to samples b and d. Similar kind of

morphologies were also observed by Kadir, A.et.al. [9], Pristovsek, M.et.al [15] respectively. In addition, pits were also observed in the InGaN layer. These surface pits are known as V-pits which is in the form of hexagon. Reports are available explaining the formation of V-pits on the InGaN surface like low In adatom mobility, indium segregation assembled in dislocation core and/or decreased Ga incorporation on the V-pits in comparison to the (0 0 0 1) plane InGaN surface [16]. It is suggested that, the origin of the InGaN nanoisland and surface pits formation is due to the low miscibility gap between InN, GaN and pre-existing threading dislocations from the underlying n-GaN/GaN layer. This indicates that, the width of nanoislands in the InGaN layer increases along with the indium content. This is attributed to the strain modifications on the InGaN nanoisland structure during the growth [17].

3.2 High Resolution X-ray Diffraction

Figure. 2 shows the ω -2 θ scans of (0 0 0 2) diffraction planes for three samples (b, c and d) obtained at different TMIn flow rate of InGaN/n-GaN heterostructures using HRXRD. The thickness and In composition in respect of InGaN layers were estimated by (0 0 0 2) plane of ω -2 θ scan. In principle, Vegard's law was used to estimate the composition of the elements with respect to the lattice parameter in III-nitride layer. It states that the lattice parameter of a solid solution alloy will have a linear dependence on composition. It is based on the dynamical X-ray diffraction theory. Using the epitaxy smooth fit software, the simulated peak was fitted with the obtained experimental peak along with the background intensity [18, 19]. The InGaN thickness and In compositions were found to be 5 nm - 6 nm and 15, 16, 17 % for samples b, c, d respectively. The diffraction peak of the n-GaN/GaN corresponding to (0 0 0 2) plane have been obtained for all samples. The high-intensity peak observed in figure.2 represents the n-GaN/GaN layers, whereas the InGaN layer was observed to be low-intense

peaks. The symmetric (0 0 0 2) and asymmetric (1 0 -1 5) planes are shown in Figure 3 (b, c, d). The full width at half maximum (FWHM) of (0 0 0 2) plane rocking curve of n-GaN/GaN are 414, 406, 395 arc-sec whereas (1 0 -1 5) are 266, 260, 263 arc-sec for samples b, c, d respectively. In the present study, the InGaN was sufficiently thin for the growth of the n-GaN/GaN/sapphire templates for all samples. The peak of the InGaN layer was quite broad. It is difficult to estimate the FWHM for InGaN layers due to lower diffraction in InGaN/n-GaN heterostructures.

The densities of edge dislocations (D_{edge}) and screw dislocations (D_{screw}) have been estimated from HRXRD rocking curve widths of the n-GaN/GaN layers. The threading dislocation (TD) density ($D = D_{edge} + D_{screw}$) is calculated using [20],

$$D_{edge} = \frac{\beta^2}{4.35|b_{edge}|^2} \quad D_{screw} = \frac{\alpha^2}{4.35|b_{screw}|^2} \quad (1)$$

Where D_{edge} is the edge dislocation density, D_{screw} is the screw dislocation density and are referred to FWHM value measured from rocking curve diffraction peak (0 0 0 2) and (1 0 -1 5) respectively. Where b_{edge} (a axis - 3.1878 Å) and b_{screw} (c axis - 5.185 Å) are the burger vector lengths of the edge and screw components respectively [21]. The calculated n-GaN/GaN layers edge, screw and threading dislocations density values are given in table 1. The majority of the TDs is edge-type dislocations that corresponds to (1 0 -1 5) plane. While increasing the In flow rate, there is a decrease in the order in n-GaN/GaN screw dislocation density for all samples. In particular sample c, edge dislocation density value of n-GaN/GaN layer is less in comparison with the other samples. This means that both the tilt and the twist components of the mosaicity gets varied.

The Asymmetric Reciprocal Space Mapping (RSM) of (1 0 -1 5) plane is shown in figure 4, clearly indicates that diffracted intensity of (1 0 -1 5) plane from InGa_N layer was less intense, and hence the analysis of RSM was very narrow in a thin InGa_N layer shown in samples b, c and d respectively. It is clearly seen that for all samples the InGa_N peaks are vertically aligned with the n-GaN/GaN peak. Upon increasing the In flow in InGa_N, the n-GaN/GaN layer contour shapes in the RSM are noted to be almost regular and broadened along the Q_x directions depending on the mosaic structure of the InGa_N layers. Hence, InGa_N layer is pseudomorphic with the n-GaN/GaN layers. Using the epitaxy smooth fit software, the Reciprocal Lattice Point (RLP) coordinates of Q_x values are estimated as 0.2770, 0.2783, 0.2800 and the Q_y values are found to be 0.7253, 0.7263, 0.7239 for InGa_N layers (sample b, c and d) respectively. In this RSM map, any pair of reciprocal lattice positions (Q_x , Q_y) in the map corresponds to a pair of lattice constants (a , c) of the wurtzite structure related by equation 2, [22]

$$Q_x = \left(\frac{\lambda}{a\sqrt{3}} \right) \quad Q_y = \left(\frac{5\lambda}{2c} \right) \quad (2)$$

Where Q_x and Q_y are parallel and perpendicular elements of reciprocal lattice points. In the case of InGa_N layers, derived the relaxed lattice constant using the equations 3, 4

$$c_0^{In_xGa_{1-x}N} = x.c^{InN} + (1-x).c^{GaN} \quad (3)$$

$$a_0^{In_xGa_{1-x}N} = x.a^{InN} + (1-x).a^{GaN} \quad (4)$$

Where $a^{GaN} = 0.31878$ nm, $c^{GaN} = 0.5185$ nm and $a^{InN} = 0.3537$ nm, $c^{InN} = 0.5706$ nm are the relaxed lattice constant values of GaN and InN respectively [21, 23, 24]. The relaxed values of lattice constant a_0 and c_0 values are 0.3241 nm, 0.3244 nm, 0.3248 nm and 0.5262 nm, 0.5267 nm, 0.5273 nm for various (11, 13 and 14 $\mu\text{mol/min}$) flow rates of InGa_N layer using Vegard's

law. The In-plane, out-of plane lattice constants and In-plane, out-of plane strains for InGaN layers were given in table. 2 [25].

$$\text{Out-of plane strain } (\varepsilon_{zz}) = \frac{c - c_0}{c_0} \quad (5)$$

$$\text{In-plane strain } (\varepsilon_{xx}) = \frac{a - a_0}{a_0} \quad (6)$$

Where c and a are the measured experimental out-of plane and In-plane lattice constants and a_0 , c_0 are the relaxed out-of plane and In-plane lattice constants respectively. The sample c has low out-of plane lattice parameter and the corresponding out-of plane tensile strain value is low (0.0066). This may be attributed to the distribution of the nanoislands which is rather homogeneous in comparison to other InGaN samples.

3.3 X-ray photoelectron spectroscopy

XPS was used to estimate the composition of the grown InGaN layers. Figure.5 shows the XPS survey spectrum of InGaN grown with TMIn flow of 13 $\mu\text{mol/min}$ (sample c). XPS analysis was carried out using CASA fitting software. From this spectrum, In4d, In3d, In3p, Ga3d and Ga3p, Ga2p, Ga3s, N (1s, 2s) were selected for the analysis. XPS quantitative analysis indicates the energy levels of all InGaN samples containing oxygen and carbon. In 4d, Ga 3d, and N 2s photoelectron peaks were selected to estimate the In composition in all InGaN samples. All the binding energies were C1s correction factor signal at 283.2 eV. The signal N1s represents the corresponding peaks of GaN (397.49eV). Additionally, energy levels corresponding to Ga3d/In4d/N2s were observed between 15 eV to 22 eV. The InGaN surface binds with oxygen and carbon at 529.7 eV and 283.2 eV [26, 27]. The intense Ga2p signals can be attributed to the two separate states of metallic “Ga” binding energy (BE) at 1144.99 eV and 1117.99 eV [28]. The surface oxygen concentration is to increase the ionic bonding nature of the metallic gallium

which represents the increase in the 2p binding energy [29]. The positions of the pure In3d doublet peaks with binding energies 444.99 eV, 452.49 eV and the difference in energy almost constant and equal to 7.5 eV which is good agreement with the literatures. The In3p peak can be categorized into peaks containing In₂O₃ and formation of InN_xO_y complex [30]. In context, indium surface contents are calculated to be only ~10, 12, 13 % for samples b, c, d respectively. Based on the above description, it is concluded that the In and Ga surface concentration in the grown InGaN samples explain the inhomogeneous distribution.

3.4 Photoluminescence

The optical properties of InGaN/n-GaN heterostructures were characterized using PL measurements by varying the TMIn flow rate. The emission spectrum is measured for each InGaN/n-GaN/GaN layer as shown in figure 6 (I) (b, c, d). The near band edge emission (NBE) peak occurs at 363 nm for n-GaN/GaN layers. In figure 6 (II) (b, c, d), the PL spectrum exhibits multiple peaks from 479 nm to 657 nm. In figure. 6 (II) more than five emission peaks were observed. In order to determine the peak position, each PL spectrum was fitted by a multi-gaussian distribution function and the spectrum was divided into individual wavelengths in all InGaN samples. The PL emission is shown in λ_1 notation for 479 nm (b), 480 nm (c), and 484 nm (d). In particular, sample c shows λ_2 to λ_8 emissions with almost equidistant interference fringes. It reflects almost uniform size and composition distribution. In the samples b, d interference fringes are not periodically formed. The inhomogeneous broadening of the peaks originates from different island size and composition distributions of the nanoislands. The beginning of the multi-peak emissions from distinct state may arise from the transitions concerning the ground and excited states of InGaN nanoisland structures. The lower and higher wavelength emissions originate from the transitions related to excited and ground states. It

accounts for the Boltzmann distribution function in InGaN nanoislands [31]. Also, it is noted that the high intensity emission peaks at 539 nm, 549 nm and 543 nm appear with varying TMIn flow of InGaN nanoisland structures. The red shift can be attributed to the quantum-confined Stark effect (QCSE) as a result of the change in polarization effect [32]. The blue shift results from the confinement effect inducing the shift of the electronic levels, whereas an increase in the exciton bohr radius is reported due to the reduced coulomb energy interaction between the electrons and holes in InGaN layer. It reveals different island sizes in InGaN samples and implies presence of the localized states [8, 33]. The bowing parameter of InGaN nanoislands was calculated using the following equation,

$$E_{g,InGaN}(X) = (1 - X)E_{g,GaN} + X E_{g,InN} - bx(1 - x) \quad (7)$$

Where $E_{g, InGaN}(x)$ represents the band gap energy of $In_xGa_{1-x}N$. $E_g(InN)$ and $E_g(GaN)$ represent the bandgap energies of InN (0.7 eV) and GaN (3.42 eV) respectively and b is the bowing parameter. The experimental data shows good fit with bowing curve b of 5 eV (Figure. 7). This anomalous bowing of the InGaN was found to be dependent on the In composition. Even though, there have been several reports for InGaN band gap information, there are significant scattering values for the bowing parameter ranging from 2.5 to 5.9 eV [34, 35]. The high bowing parameter of the InGaN bandgap represents the function of In composition fluctuations leading to a structural distortion and volume deformation due to different lattice constants of GaN and InN layers. Moreover, InN bandgap obtained between ~ 1.9 eV and ~ 0.7 eV still remains controversial [36, 37]. It is concluded that, green emissions originate from strong exciton localization effect [8] and the existence of the sub band luminescence in the strain induced nanoisland growth mode on InGaN layers are mandatory.

3.5 Hall Measurement

In order to investigate electrical properties, carrier concentration, sheet concentration and mobility, hall measurements were carried at on all samples. For the InGaN samples, hall measurement indicates n-type behavior, which can be attributed to the presence of nitrogen vacancies. From table 3, it was observed that the carrier concentration for samples b, c and d decreased. In the InGaN nanoisland samples, measured carrier concentrations is above the critical mott density ($n_c^{(1/3)} * a^*_H = 0.25$) where n_c and a^*_H are critical carrier concentrations and effective Bohr radius respectively. The values of these carrier concentration exhibit degenerate semiconducting behavior. The resistivity values are above the critical mott density for semiconductor behavior (8.68×10^{-3} , 8.89×10^{-3} and 8.78×10^{-3} ohm-cm for samples b, c, and d respectively). Consequently, the position of fermi level varies slightly and lies between the conduction band minimum and valence band maximum [34]. Sample c was obtained as low mobility of $291 \text{ cm}^2/\text{V.s}$ in comparison to the other two InGaN samples. The mobility of InGaN nanoisland is low in comparison to InN or GaN samples which is usually more than $400 \text{ cm}^2/\text{V.s}$ at room temperature. This performance suggests that as more In is getting incorporated into GaN, alloy scattering centers gets increased, which is good agreement to the Philips electronegative difference theory [38, 39].

4. Conclusion

In the present work, MOCVD was used to grow the strained InGaN layers on n-GaN/GaN/sapphire template. The influence of TMIn flow rate on the In incorporation, surface morphology, optical and electrical properties were analyzed using HRXRD, AFM, PL and hall

measurement studies. The InGaN thickness and In composition were estimated using HRXRD. The RSM was confirmed that InGaN layers grown on n-GaN/GaN layers have pseudomorphic growth. The InGaN layers grown with 13 $\mu\text{mol/min}$ TMIn flow exhibited a less out-of tensile strain and atomically flat (height less than 2 nm), smooth surface which was verified by HRXRD and AFM analysis. In addition, the origin of the InGaN nanoisland with V-pits formation was due to the low miscibility gap between InN, GaN and pre-existing threading dislocations from the underlying n-GaN/GaN layer. The Composition fluctuations was observed in the InGaN nanoisland structures and confirmed from the XPS analysis. PL result revealed the existence of the subband luminescence in strain induced nanoisland growth and also revealed the dominant green emission peaks from the strong localization effect. It showed that high bowing effect in InGaN/n-GaN heterostructures. Hall measurement analysis shows the changes in semiconducting behavior with variation of TMIn flow in addition to the influence of composition fluctuation. This work helps us to understand the growth of atomically flat and smooth surfaces to improve the morphology, optical and electrical properties. These InGaN nanoislands can be integrated into the active region of the InGaN/GaN MQW structures and suitable for the fabrication of high efficiency InGaN based devices.

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References

- [1]. Zhao, W., Wang, L., Wang, J., Hao, Z., & Luo, Y. J. Appl. Phys, 110(1) (2011) 014311
- [2] Yao, H. H., Lu, T. C., Huang, G. S., Chen, C. Y., Liang, W. D., Kuo, H. C., & Wang, S.C. Nanotechnology, 17(6) (2006) 1713
- [3] Zheng, X., Liang, H., Wang, P., Sun, X., Chen, Z., Wang, T., Sheng, B., Wang, Y., Chen, L., Wang, D. and Rong, X., Superlattices and Microstructures, 113 (2018) 650-656
- [4] Sang, L., Liao, M., Liang, Q., Takeguchi, M., Dierre, B., Shen, B., Takashi Sekiguchi, Yasuo Koide & Sumiya, M. Adv. Mater., 26(9) (2014) 1414-1420
- [5] Farrell, R. M., Neufeld, C. J., Cruz, S. C., Lang, J. R., Iza, M., Keller, S., S. Nakamura, S. P. DenBaars, U. K. Mishra and & Speck, J. S. Appl. Phys. Lett., 98(20) (2011) 201107
- [6] Seo, T. H., Shim, J. P., Chae, S. J., Shin, G., Kim, B. K., Lee, D. S., Young Hee Lee, & Suh, E. K. Appl. Phys. Lett., 102(3) (2013) 031116
- [7] Park, I. K., Kwon, M. K., Kim, J. O., Seo, S. B., Kim, J. Y., Lim, J. H., Seong-Bum Seo & Kim, Y. S. Appl. Phys. Lett., 91(13) (2007) 133105
- [8] Bayram, C., & Razeghi, M. Appl. Phys. A, 96(2) (2009) 403-408
- [9] Kadir, A., Meissner, C., Schwaner, T., Pristovsek, M., & Kneissl, M., J. Cryst. Growth 334(1) (2011) 40-45
- [10] Wang, T. C., Kuo, H. C., Lu, T. C., Tsai, C. E., Tsai, M. Y., Hsu, J. T., & Yang, J. R. Jpn. J. Appl. Phys., 45(4S) (2006) 3560
- [11] Lai, Y. L., Liu, C. P., Lin, Y. H., Hsueh, T. H., Lin, R. M., Lyu, D. Y., Zhao-Xiang Peng & Lin, T. Y. Nanotechnology, 17(15) (2006) 3734
- [12] Keller, S., Keller, B. P., Minsky, M. S., Bowers, J. E., Mishra, U. K., DenBaars, S. P., & Seifert, W. J. Cryst. Growth, 189 (1998) 29-32
- [13] Adelman, C., Simon, J., Pelekanos, N.T., Samson, Y., Feuillet, G. & Daudin, B., physica status solidi (a), 176 (1999) 639-642
- [14] Oliver, R.A., Kappers, M.J., Humphreys, C.J. and Briggs, G.A.D., Journal of applied physics, 97(1), (2005) p.013707
- [15] Pristovsek, M., Kadir, A., Meissner, C., Schwaner, T., Leyer, M., Stellmach, J., Michael Kneissl a, Francesco Ivaldi & Kret, S. J. Cryst. Growth, 372 (2013) 65-72
- [16] Liu, J., Jia, Z., Ma, S., Dong, H., Zhai, G. and Xu, B., Superlattices and Microstructures, 113 (2018) 497-501
- [17] Figge, S., Tessarek, C., Aschenbrenner, T. & Hommel, D., physica status solidi (b), 248 (2011) 1765-1776.

- [18] Carlo Lamberti, Characterization of semiconductor heterostructures and nanostructures, Elsevier, 2008, (Chapter 4), 93
- [19] Bowen, D. K., & Tanner, B. K, High resolution X-ray diffractometry and topography, CRC press, 2005, p.58
- [20] Lazarev, S., Bauer, S., Forghani, K., Barchuk, M., Scholz, F., & Baumbach, T, J. Cryst. Growth, 370 (2013) 51-56
- [21] Choi, S., Heller, E, Dorsey, D., Vetury, R & Graham, S, J. Appl. Phys, 113 (2013) 093510
- [22] Bae, SY, Song, YH, Jeon, SR, Kim, D M, Jho, Y D and Lee, D S, J. Cryst. Growth, 387 (2014) 23-28
- [23] Darakchieva, V., Monemar, B., & Usui, A, Appl. Phys. Lett, 91(3) (2007) 031911
- [24] Kitamura, S., Senshu, M., Katsuyama, T., Hino, Y., Ozaki, N., Ohkouchi, S., & Hogg, R. A. Nanoscale research letters, 10(1) (2015) 1-6
- [25] Moram, M A, & Vickers, M E, Reports on Progress in Physics, 72 (2009) 036502
- [26] Alamé, S., Quezada, A. N., Skuridina, D., Reich, C., Henning, D., Frentrup, M., Tim Wernicke, Ingrid Koslow, Michael Kneissl, Norbert Esser & Vogt, P. Materials Science in Semiconductor Processing, 55 (2016) 7-11
- [27] Cruz-Hernández, E., Ramirez-Lopez, M., Pérez-Caro, M., Mani-Gonzalez, P. G., Herrera Gómez, A., Gorbachev, A. Y., M. Lopez-Lopez & Méndez-García, V. H. Journal of Crystal Growth, 378 (2013) 295-298
- [28] Dasari, K., Thapa, B., Wang, J., Wright, J., Kaya, S., Jadwisieniczak, W. M., & Palai, R. Journal of Electronic Materials, 45(4) (2016) 2071-2077
- [29] Tran, N. H., Holzschuh, W. J., Lamb, R. N., Lai, L. J., & Yang, Y. W. The Journal of Physical Chemistry B, 107(35) (2003) 9256-9260
- [30] Chen, W. C., Wu, Y. H., Wang, W. L., Kuo, S. Y., Lai, F. I., & Chang, L, Influence of V/III Flow Ratio on Growth of InN on GaN by PA-MOMBE, ECS Journal of Solid State Science and Technology, 2(7) (2013) P305-P310
- [31] Kitamura, S., Senshu, M., Katsuyama, T., Hino, Y., Ozaki, N., Ohkouchi, S., & Hogg, R. A. Nanoscale research letters, 10(1) (2015) 1-6
- [32] Liu, W., Yang, J., Zhao, D., Jiang, D., Zhu, J., Chen, P., Liu, Z., Liang, F., Liu, S., Xing, Y. and Zhang, L., Applied Surface Science, 456, (2018), 487-492
- [33] Lee, C. K., Hsu, J. H., Wang, D. C., Chang, Y. H., Kuo, H. C., & Wang, S. C. Materials Science and Engineering: C, 26(5) (2006) 975-978

- [34] Yildiza, A., Dagdelenb, F., Acarc, S., Lisesivdinc, S. B., Kasapc, M., Aydogdub, Y., & Bosid, M. *Acta Physica Polonica A*, 113(2) (2008) 731-739
- [35] Sinha, N., Roul, B., Mukundan, S., Chandan, G., Mohan, L., Jali, V.M. and Krupanidhi, S.B., *Materials Research Bulletin*, 61, (2015) 539-543
- [36] Liu, J., Liang, H., Xia, X., Abbas, Q., Liu, Y., Luo, Y., Zhang, Y., Yan, L., Han, X. and Du, G., *Journal of Alloys and Compounds*, 735, (2018) pp.1239-1244
- [37] Lee, K.J., Oh, T.S., Kim, T.K., Yang, G.M. and Lim, K.Y., *Journal of the Korean Physical Society*, 47(5), (2005) pp.842-846
- [38] Yıldız, A., Kemal, Ö. M., Bosi, M., Özçelik, S., & Kasap, M, *Chinese Physics B*, 18(9) (2009) 4007
- [39] Papadomanolaki, E., Bazioti, C., Kazazis, S.A., Androulidaki, M., Dimitrakopulos, G.P. and Iliopoulos, E., *Journal of Crystal Growth*, 437, (2016) pp.20-25

Figures

Figure 1. AFM ($10\mu\text{m}^2$) images along with the line profile representing the InGaN/n-GaN heterostructures

Figure 2. HRXRD (0 0 0 2) profile of a InGaN/n-GaN Heterostructures (Blue Line) along with Simulation (Red line)

Figure 3. Symmetric (0 0 0 2) and asymmetric (1 0 -1 5) plane for n-GaN/GaN layer

Figure 4. Reciprocal Space Mapping corresponding to (1 0 -1 5) diffraction of samples b, c and d

Figure 5. XPS survey spectrum of the InGaN for In flow $13\mu\text{mol}/\text{min}$ along with its fitting (sample c)

Figure 6. (I) PL spectrum of InGaN /n-GaN Heterostructures (II) Multi-Gaussian fitting to the InGaN emission peaks

Figure 7. Optical band gap vs Indium content

Figure 1

Figure 2

Figure 3

Figure 4

Figure 5

Figure 6

Figure 7

Tables

Table 1. Dislocation densities of n-GaN/GaN layers

Table 1

Samples	Dislocations of n-GaN/GaN layers		
	Screw dislocations (10^8cm^{-2})	Edge dislocations (10^8cm^{-2})	Threading Dislocations (TD) (10^8cm^{-2})
b	3.46	3.76	7.22
c	3.31	3.59	6.9
d	3.17	3.68	6.85

Table 2. Lattice constants, In-plane and out-of plane strains for InGaN layers

Table

Samples	InGaN lattice constants and strains			Out –plane strain (ε_{zz})
	c_{InGaN} (Å)	a_{InGaN} (Å)	In-plane strain (ε_{xx})	
b	5.310	3.211	-0.0092	0.0091
c	5.302	3.196	-0.0147	0.0066
d	5.320	3.176	-0.0221	0.0089

Table 3. Room temperature electrical properties of InGaN layers

Table 3

Samples	Mobility $\text{cm}^2/\text{V.s}$	Carrier concentration cm^{-3} $\times 10^{18}$	Sheet concentration $\text{cm}^{-2} \times 10^{14}$
	RT	RT	RT
a	303	2.48	4.96
b	296	2.43	4.87
c	291	2.41	4.84
d	295	2.39	4.83