Advances in InN epitaxy and its material control by MBE towards novel InN-based QWs

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Abstracts:

Initial stage of MBE epitaxy processes of InN on **c-plane** GaN template and successful p-type doping were studied. It was found that InN epitaxy proceeds through SK-mode and 1 ML-thick InN can be coherently grown on GaN. Successful p-type doping of InN was achieved by Mg doping levels from 10^{18} to about $3 \times 10^{19} \text{ cm}^{-3}$. Overdoped Mg's introduced new donors in InN resulting in n-type conduction.

Novel structure InN-based QWs consisting of coherent 1ML-thick InN wells embedded in GaN matrix were proposed and fabricated. It was confirmed that extremely fine structure InN QWs **with quite sharp and flat hetero-interface** were fabricated by self-limiting growth mode under In-polarity growth regime at remarkably higher growth temperatures up to 650 °C.

Keywords: A1. Doping, Nanostructures, A3. Molecular beam epitaxy, Quantum wells, B1. Nitrides, B2. Semiconducting indium compounds

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1. Introduction

InN is the least studied material among III-nitrides and there are still several problems to be overcome for better understanding of its material properties and also for its device application [1-3]. Bonding between In and N is weaker and longer than those for GaN and/or AlN, and this results in large mismatches for InN in both lattice parameters and optimum epitaxy temperatures among III-nitrides [4]. Another important issue for InN epitaxy is a big difficulty in achieving p-type conduction, because the Fermi level stabilized energy is located inside the conduction band, resulting in easy introduction of donor levels together with high density surface electrons [5]. In particular, the presence of such high density electrons as $3\text{-}5 \times 10^{13} \text{ cm}^{-2}$ located in a few nm-thick surface regions together with those at the heterointerface between the InN epilayer and the substrate is a serious problem not only for achieving but also for detecting the p-type conduction for InN epilayers [5, 8-15].

We have been studying for a long time on the epitaxy and material control of InN and related alloys under paying attention on the crystalline polarity, i.e. controlled MBE processes for better quality InN crystals under In-polarity and N-polarity growth regimes [6,7]. On the basis of understanding on the residual donors in unintentionally n-type doped InN, we have very recently achieved successful p-type doping of InN after a systematic and very wide range [Mg]s doping study [8-13]. This is a very important mile stone when considering the application of InN for functionality photonic and electronic devices [14, 15].

Further, in order to find out one of solutions for minimizing the effects arising from large mismatch for InN-based III-N heterostructures, we have been also studying fabrication and characterization of InN-based nanostructures including quantum dots, nano columns, and conventional two dimensional multiple quantum wells (MQWs) [16-21]. Recently we have proposed and demonstrated novel structure InN/GaN MQWs consisting of very thin atomically flat and sharp InN wells, i.e., one monolayer (1 ML) and/or fractional ML (FML)

and also sometimes 2 MLs-thick InN wells embedded in the GaN matrix [22-24]. Since the critical thickness for the growth of InN on c-plane GaN is above 1 ML as discussed in the following section, 1ML-thick InN/GaN matrix MQWs are free of misfit dislocation in principle. Further, the growth temperature for 1 ML InN epitaxy on GaN can be remarkably increased, because of increased effective bonding strength between In and N in this case [22]. Therefore the novel structure InN/GaN MQWs possess potentially high quality nature and are considered as promising new-structure active layers for high efficiency laser diodes working in visible wavelengths for blue-green region, and remarkable improvements in lasing performances are expected [24].

In this paper, we report present status and future prospects of InN epitaxy including very recent study on p-type doping of InN as well as recent advances in nano-processes in InN-based low dimensional nanostructures, i.e., 1-ML InN wells/GaN matrix MQWs.

2. Epitaxy of InN on GaN by MBE

MBE is considered suitable for InN epitaxy process compared to MOVPE, because smaller epitaxy-temperature mismatch between InN and other nitrides is expected for MBE [4, 25, 26]. **Fig. 1** shows an MBE system specially designed for III-nitrides epitaxy in which several in-situ and real-time monitoring systems for epitaxy processes are attached. CAICISS (Coaxial Impact Collision Ion-Scattering Spectroscopy) is useful for detecting atomic species in a few atomic-layers surface regions which is applicable for determining the polarity of asymmetric crystalline structure epilayers such as III-nitrides. SE (Spectroscopic Ellipsometry) is useful for monitoring growth evolution in atomic level as well as precise real-time control of surface stoichiometry. Simultaneous observations of growth front using SE and RHEED are essential for understanding the coherent epitaxy process of InN on GaN with such a large lattice mismatch as 11 %.

One of the most important findings for InN epitaxy is the big difference in the highest epitaxy temperatures of about 500 °C and 600 °C for In-polarity and N-polarity growth regimes, respectively [6, 7]. These temperatures are critical ones where no film deposition is possible and only In droplets appears on the surface above them. Table I summarizes some properties of typical high-quality / nominally undoped n-type InN epilayers grown under In- and N-polarity. Reflecting higher growth temperature in N-polarity growth regime, the properties shown in this table look better for N-polarity InN epilayers than those for In-polarity. N-polarity growth is preferable in the point of better matching in epitaxy temperatures among other nitrides. As for the surface morphology and/or surface flatness, however, it was found that atomically flat surface having monolayer steps is much more easily obtained in In-polarity growth regime as shown by an AFM image and corresponding height line-profile in Fig. 2 (a). Fig. 2 (b) shows the dependence of RMS surface roughness on the growth temperature, indicating that higher growth temperatures are preferable for achieving atomically flat surface. It was found that the macroscopic surface-defects density appearing as growth spirals is following that of screw-component threading dislocations in GaN template [27]. Then in order to achieve fine InN-based nanostructures with atomically flat and sharp interface as in the case of ultimately thin 1ML-thick InN well insertion in GaN matrix, In-polarity or +c polarity growth regime is necessary as discussed in the following section.

The other important finding in InN epitaxy is that for the possibility of coherent epitaxy of InN on c-plane GaN template. It was found that the critical thickness of InN epitaxy on GaN is above 1ML and the deposition process follows Stranski-Krastanov (SK) growth mode [16]. Figs. 3 (a) and (b) show the dependences of in-plane lattice constant of InN and corresponding change in the real part of pseudo-dielectric constant during the growth, respectively, on the nominal InN coverage on GaN template. It was found from these

simultaneous observations in the very initial stage for the surface processes of InN epitaxy on GaN that InN deposition starts immediately after the InN supply but the in-plane lattice constant of epilayer does not change till the nominal coverage exceeds critical thickness of about 1ML.

Figs. 4 (a) and (b) show the cross-sectional dark field TEM images for InN epilayer grown on MOCVD-grown c-plane GaN template. It was found that screw- and edge-component threading dislocations in typical high quality InN epilayers grown on GaN are low 10^8 cm^{-2} and also low 10^{10} cm^{-2} , respectively, where the former is originating from the screw-component threading dislocations in GaN template while the latter is arising from the large lattice mismatch between InN and GaN [9].

3. Achieving p-type InN epilayers with Mg-doping

For the p-type doping of InN, remarkable reduction of unintentionally doped donors including both impurities and native defects in InN epilayers is necessary first. Residual electron concentrations in recently reported high purity InN epilayers are less than about $1 \times 10^{18} \text{ cm}^{-3}$ and they are approaching to low and mid 10^{17} cm^{-3} orders in the highest purity epilayers as shown in Table I. It is known that the edge-type threading dislocations in GaN introduce deep levels inside its forbidden band. It was theoretically confirmed, however, that they introduce donor levels above the conduction band bottom in InN though their exact levels are slightly different depending on their detailed core structures [28]. Then it was considered that the edge-type threading dislocation is one of residual donor species in InN when the donor impurities are reduced enough below low 10^{17} cm^{-3} levels. [9] On the basis of very simple assumption that each dangling bond in edge-type dislocation induces one such donor level in InN, electron density for the InN epilayer with $1 \times 10^{10} \text{ cm}^{-2}$ edge-type dislocations is roughly estimated to be about $3.5 \times 10^{17} \text{ cm}^{-3}$. This means that extrinsic donor

impurities in these InN epilayers are reduced enough and probably ready for p-type doping, and native defects such as edge-type dislocations may contribute as dominant residual donors in those high quality InN epilayers.

On the basis of these understandings stated above, p-type doping of InN with Mg acceptors was investigated. The growth temperatures for Mg-doped InN layers were 480 °C and 580 °C for In-polarity and N-polarity samples, respectively. It was found first that the sticking coefficient of Mg in both In- and N-polarity InN growth modes is almost unity. This is because their epitaxy temperatures are low and Mg's are effectively adsorbed and incorporated into InN. It was found, however, the polarity inversion from In-polarity to N-polarity takes place when [Mg] exceeds above about $3 \times 10^{19} \text{ cm}^{-3}$ [11]. This means that Mg doping of InN under keeping the crystalline polarity in In-polarity during growth is difficult for [Mg]s above $3 \times 10^{19} \text{ cm}^{-3}$. As already stated above, due to the effect of high density surface electrons it is difficult to characterize and detect net acceptors in these Mg-doped InN epilayers by conventional Hall measurements/analyses. That is, the surface/interface electron density of $3\text{-}5 \times 10^{13} \text{ cm}^{-2}$ corresponds to the bulk electron concentration of $3\text{-}5 \times 10^{17} \text{ cm}^{-3}$ for 1 μm -thick sample. Taking into account of the donors arising from edge-type threading dislocations together with the surface/interface electrons, these “residual” electrons correspond to about $1 \times 10^{18} \text{ cm}^{-3}$ electron density in 1 μm -thick InN layer. Further, since the electron mobility in InN is about 20-30 times higher than that of hole, about 40-50 μm -thick layer is necessary for p-type InN samples with the hole concentrations of low 10^{18} cm^{-3} in order to observe and characterize their p-type conduction by conventional Hall measurements.

Then ECV (electrolyte capacitance-voltage) analyses are performed to detect embedded net acceptors below high density surface electrons in Mg-doped InN epilayers [14, 15]. If the surface electrons can be successfully depleted by the applied electric field through the Helmholtz layer in the electrolyte, the Fermi levels at InN surface can be modulated by the

applied voltage as in the case of conventional MIS (metal-insulator-semiconductor) diodes. For achieving successful modulation of surface Fermi level, the electric field intensity in the Helmholtz layer must be very high and/or the layer must be strong enough even for so high electric field as that the $3\text{-}5 \times 10^{13} \text{ cm}^{-2}$ electrons can be depleted. Fig. 5 shows the principle/idea how to detect by ECV the embedded p-type region under high density surface electrons. Fig. 5(a) shows an energy-band diagram under zero applied bias-voltage for typical n-type InN sample with high density surface electrons. The surface Fermi level is located inside the conduction band due to high surface electrons. Since the InN is highly conductive in this case, the situation for measuring the capacitance value is almost the same as that for electrical double layer-capacitor. Here the measured capacitance value corresponds only for the Helmholtz layer itself and becomes the largest. Fig. 5(b) shows the situation where the surface Fermi level can be effectively modified/moved into the inside forbidden gap by applying high electric field. The measured capacitance value in this case is of course the synthetic capacitance value for serial connection of the Helmholtz layer and the depletion layer in InN. With further increasing the applied bias voltage the measured capacitance value finally approaches to that for the much smaller side capacitance, i.e. the depletion layer in InN. The net donor (or acceptor in p-type sample) concentration can be estimated by the minimum capacitance value or the slope for the relationship between the capacitance and applied voltage. Fig. 5(c) and 5(d) show the similar surface Fermi level modulation for p-type InN samples where the surface is also covered by high density electron, i.e. Fig. 5(c) and 5(d) are for under zero bias voltage and for the surface Fermi level modified into the forbidden gap, respectively. As stated above the minimum capacitance value reflects the net donor/acceptor concentration inside the InN layer.

The important point for detecting the conduction type of the semiconductor sample by ECV is as following; The applied voltage values necessary for modulating the surface

Fermi level so as to detect the minimum capacitance value are different between n-type and p-type samples as shown in Figs. 5(a) and 5(c). Because the Fermi level position inside the layer is different for n-type and p-type samples. Actually to detect the capacitance minimum value as well as the corresponding applied voltage, Mott-Schottky plots are prepared in which the $1/C^2$ vs. the applied voltage is plotted. Thus the presence of net acceptors below the surface electrons can be detected and estimated by observing the peak-shift and corresponding minimum capacitance value in the Mott-Schottky plots for ECV measurements [12-14].

A series of Mg-doped InN epilayers with very wide range of [Mg]s varied from low 10^{16} cm^{-3} to about $4 \times 10^{21} \text{ cm}^{-3}$ were prepared for both In- and N-polarities at 480 °C and 600 °C, respectively, and also corresponding Mott-Schottky plots are shown in Figs. 6 (a) and (b), respectively. As it is summarized for In-polarity InN:Mg samples in Fig. 7, it was found that p-type doping was successfully achieved in [Mg] region from $1 \times 10^{18} \text{ cm}^{-3}$ to about $3 \times 10^{19} \text{ cm}^{-3}$ in both polarity samples. Net acceptor (donor) concentrations in p-type (n-type) samples were estimated by taking into account of the effects of measuring frequency of ECV and/or midgap states at the interface of electrolyte and InN [12]. It was found that 1) the effective residual donor concentrations in these samples are about 10^{18} cm^{-3} and the presence of net acceptors is detected when [Mg] exceeds it, and 2) the overdoped Mg's above about $3 \times 10^{19} \text{ cm}^{-3}$ are not acceptors any more but they induce new donors in InN. Details of Mg-overdoping effects are under investigation now with using positron annihilation analyses etc.

Activation energy of Mg acceptors was estimated to be about 61 meV by the analyses of excitation-power dependent photoluminescence properties of Mg-doped InN samples [9, 10]. Hole mobility was successfully estimated to be about 17 – 36 cm^2/Vs at room temperature by the analyses of thickness-dependent p-type sheet-conductivities of Mg-doped InN samples [13].

4. Proposal and features of novel InN-based QWs embedded in GaN

As one of the ideas to solve the problems when fabricating fine structure InN-based nano-heterostructures **with quite sharp and flat hetero-interface** in spite of highly lattice- and epitaxy-temperature mismatched system, we have proposed a novel structure InN/GaN QWs in which the basic structure is consisting of “one monolayer-thick” InN wells embedded in GaN matrix formed under +c-polarity growth regime. There are several features expected for the proposed novel InN-based QWs:

- 1) *Coherent growth of InN wells on GaN*: Fabrication of high quality QWs can be expected in principle.
- 2) *GaN matrix effect for In-polarity InN deposition*: Growth temperature for 1ML InN on GaN can be extremely higher than that for thick In-polarity InN deposition. **The critical epitaxy temperature for In-polarity InN is 500 °C, but such thin InN layer as about 1 ML can be deposited even at temperatures up to 650 °C.**
- 3) *Immiscibility between InN and GaN*: Atomically flat and sharp 1 ML InN wells can be deposited on Ga-polarity GaN template by self-ordering and self-limiting processes under wider growth windows. **The self-limiting growth is possible only for about 1 ML-thick InN deposition when the deposition temperatures are higher than the critical epitaxy temperature, i.e. 500 °C for the In-polarity InN deposition.**

We first examined to fabricate a series of InN/GaN QWs at different growth temperatures from 500 °C to 650 °C with 50 °C step. As stated above, the highest growth temperature for “thick” In-polarity InN was about 500 °C, and the temperatures examined here were much higher than this critical temperature. It was found, however, that quite thin around 1 ML-thick InN deposition on Ga-polarity GaN template was possible even for just 1ML InN supply when the growth temperatures were between 500 °C and 600 °C. **Here the**

“1ML InN supply” means that the In and N species corresponding to 1ML-thick InN deposition for the case of lower growth temperatures than 500 °C were supplied. When the growth temperature was further highered up to 650 °C, about 3 ML InN total supply under increased supply rate was necessary to assure the InN deposition. Further, it was confirmed that the structural quality of obtained InN/GaN QWs was significantly improved with increasing growth temperature as shown below.

Fig. 8 shows extremely fine XRD 2θ - ω diffraction patterns around GaN (0002) planes for the sample grown at 650 °C. Fig. 9 shows a cross sectional TEM image of the sample grown at 600 °C. These XRD data and XTEM image also indicate that very clear and sharp MQW structure is fabricated. As for the thickness control of InN wells, it was confirmed that the deposition of InN wells up to 2MLs was possible. When the InN wells were deposited at temperatures above 650 °C, fabrication of fractional ML InN wells was also possible, i.e., InN tended to be not a continuous layer because of lowered sticking coefficient for In and N species at such high temperature as 650 °C and/or enhanced dissociation and re-evaporation of InN at this high temperature.

Fig. 10 shows a sub-Å resolution Z-contrast image obtained by scanning TEM for the sample grown under 3ML InN supply at 650 °C, in which the fractional ML InN wells were partly observed. First, it has been confirmed in the Z-contrast image that the QW is coherently embedded and no defects are generated at the hetero-interface. Further, the localization of the extra intensity in the Z-contrast image appears to indicate the presence of a single layer of InN in the GaN matrix, although additional analysis is required for such thin layers as this 1 ML QW. This analysis and detailed simulations are underway now.

Fig. 11 shows PL (photoluminescence) spectra at room temperature for the samples grown at 650 °C with different InN well-thicknesses, i.e., without InN layer, and 1 and 2 ML-thick InN wells. It was confirmed that the PL emission around ~390 nm can be observed

for the samples with 1 ML-thick QWs as indicated by the Z-contrast image in Fig. 10. Luminescence properties of the proposed novel structure InN/GaN QWs are basically following the behavior/properties of GaN-based excitons localized at very thin and deep InN potential wells. Considering these together with discretely changing PL-peak behaviors, we have concluded at this moment that the longer wavelength side PL emissions around 420-430 nm are originating from 2 ML-thick InN QWs in a GaN matrix. Moreover, due to the effect of increased oscillator strength for such excitons effectively localized at ultimately thin InN wells, strong luminescence nature is expected for the proposed QWs, and detailed study for analyzing emission mechanism is underway now.

Experimental LED structure was fabricated and its fundamental performance was characterized to confirm the potential application of the proposed novel InN-based QWs for blue-green color light emitting devices. Since the well layer is ultimately thin in the proposed QWs, the QCSE (quantum confined Stark effect) is expected to be also ultimately minimized. Experimental LED structures were fabricated by both MBE and MOCVD. We have extended our understanding for the MBE-epitaxy processes on 1ML InN-based nanostructures to MOCVD processes, and we have confirmed/observed basically the same growth behaviors for the self-limiting 1ML-thick InN well deposition by the in-situ MOCVD-process monitoring also using the spectroscopic ellipsometry. Significant difference between MBE and MOCVD is that the suitable deposition temperature region for MOVPE was about 100 °C higher than that for MBE, and this resulted in brighter LEDs. Fig. 12 shows EL (electro-luminescence) spectra at room temperature of LED structure consisting of 2ML-thick InN wells. Very bright EL emission was observed at around 418 nm and only small blue shift from 418 nm to 416 nm was observed for two orders different current injection levels varied from 1 mA to 100 mA. Thus it was confirmed that the QCSE was remarkably reduced with using ultimately thin InN wells as active layers. Detailed characterization for the performance of the LEDs is also

underway now.

5. Summary

Initial stage of MBE epitaxy processes of InN on c-plane GaN template and successful p-type doping were studied. It was found that InN epitaxy proceeds through SK-mode and 1 ML-thick InN can be coherently grown on GaN. Further, atomically flat surface with single ML steps is easily obtained under In-polarity growth regime though the epitaxy temperature for In-polarity is about 100 °C lower than that for N-polarity case. Screw-component threading dislocations are the causes for the macroscopic surface defects appearing as growth spirals, and they basically follow the screw-component dislocations in GaN template.

Edge-component threading dislocation density in typical InN epilayers grown on GaN is low 10^{10} cm^{-2} which is originating from their large lattice mismatch. Edge-type dislocations induces donor levels above the conduction band bottom in InN and one of the residual donors in the orders of $3.5 \times 10^{17} \text{ cm}^{-3}$ density. Successful p-type doping of InN was achieved by Mg doping levels from 10^{18} to about $3 \times 10^{19} \text{ cm}^{-3}$. It was found that overdoped Mg's introduce new donors in InN resulting in n-type conduction.

Novel structure InN-based QWs consisting of coherent 1ML-thick InN wells embedded in GaN matrix were proposed and fabricated. It was confirmed that fine structure InN QWs with quite sharp and flat hetero-interface were fabricated by self-limiting growth mode under In-polarity growth regime at remarkably higher growth temperatures up to 650 °C. Their optical properties are determined by GaN-excitons localized at InN wells. Experimental LED structure was fabricated and it was confirmed that QCSE was ultimately reduced due to the effect of ultimately thin 1ML InN wells.

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References

- [1] V. Yu. Davydov, A. A. Klochikhin, R. P. Seisyan, V. V. Emtsev, S. V. Ivanov, F. Bechstedt, J. Furthmüller, H. Harima, A. V. Mudryi, J. Aderhold, O. Semchinova, and J. Graul, *Phys. Status Solidi B* 229, (2002) r1.
- [2] J. Wu, W. Walukiewicz, K. M. Yu, J. W. Ager III, E. E. Haller, H. Lu, W. J. Schaff, Y. Saito, and Y. Nanishi, *Appl. Phys. Lett.* 80 (2002) 3967.
- [3] K. Xu, W. Terashima, T. Hata, N. Hashimoto. M. Yoshitani, B. Cao, Y. Ishitani, and A. Yoshikawa, *Phys. Status Solidi C* 0 (2003) 2814.
- [4] X. Q. Wang, A. Yoshikawa, *Prog. Cryst. Growth Charactr. Mat.* 48/49 (2004) 42.
- [5] W. Walukiewicz, S.X. Li, J. Wu, K.M. Yu, J.W. Ager, E.E. Haller, H. Lu and W.J. Schaff, *J. Cryst. Growth* 269 (2004) 119.
- [6] K. Xu, A. Yoshikawa, *Appl. Phys. Lett.* 83 (2003) 251.
- [7] A. Yoshikawa, S. B. Che, N. Hashimoto, H. Saito, Y. Ishitani, and X. Q. Wang, *J. Vac. Sci. Technol. B* 26 (2008).
- [8] X. Wang, S. B. Che, Y. Ishitani, and A. Yoshikawa, *J. Appl. Phys.* 99 (2006) 073582.
- [9] X. Wang, S.B. Che, Y. Ishitani, A.Yoshikawa, *Appl. Phys. Lett.* 90 (2007) 151901.
- [10] X. Wang, S. B. Che, Y. Ishitani, and A. Yoshikawa, *Appl. Phys. Lett.* 90 (2007) 201903.

- [11] X. Wang, S. B. Che, Y. Ishitani, A. Yoshikawa, H. Sasaki, T. Shinagawa, and S. Yoshida, Appl. Phys. Lett. 91 (2007) 081912.
- [12] X. Wang, S. B. Che, Y. Ishitani, and A. Yoshikawa, Appl. Phys. Lett. 91 (2007) 242111.
- [13] X. Wang, S. B. Che, Y. Ishitani, and A. Yoshikawa, Appl. Phys. Lett. 92 (2008) 132108.
- [14] W. L. Yim, R. E. Jones, K. M. Yu, J. W. Ager III, W. Walukiewicz, W. J. Schaff, J. Wu, Phys. Rev. B 76 (2007) 041303(R).
- [15] J. W. Ager III, R. E. Jones, D. M. Yamaguchi, K. M. Yu, W. Walukiewicz, S. X. Li, E. E. Haller, H. Lu, and W. J. Schaff, Phys. Status Solidi B 244 (2007) 1820.
- [16] A. Yoshikawa, N. Hashimoto, N. Kikukawa, S. B. Che, Y. Ishitani, Appl. Phys. Lett. 86 (2005) 153115.
- [17] X. Wang, S.B. Che, Y. Ishitani, A. Yoshikawa, Phys. Status Solidi C 3 (2006) 1561.
- [18] X. Wang, S.B. Che, Y. Ishitani, A. Yoshikawa, J. Cryst. Growth 301/302 (2007) 496.
- [19] S.B. Che, W. Terashima, Y. Ishitani, T. Matsuda, H. Ishii, S. Yoshida A. Yoshikawa, Appl. Phys. Lett. 86 (2005) 261903.
- [20] W. Terashima, S.B. Che, Y. Ishitani, A. Yoshikawa, Jpn. J. Appl. Phys. 45 (2006) L539.
- [21] S.B. Che, T. Mizuno, X. Wang, Y. Ishitani, A. Yoshikawa, J. Appl. Phys. 102 (2007) 083539.
- [22] A. Yoshikawa, S. B. Che, W. Yamaguchi, H. Saito, X. Q. Wang, Y. Ishitani, and E. S. Hwang, Appl. Phys. Lett. 90 (2007) 073101.
- [23] E. S. Hwang, S. B. Che, H. Saito, X. Q. Wang, Y. Ishitani, A. Yoshikawa, J. Electronic Materials 37 (2008) 597.
- [24] A. Yoshikawa and S. B. Che, Int. J. High Speed Electr. Syst. (2008) (in print).
- [25] A. Yoshikawa, K. Xu, Y. Taniyasu, K. Takahashi, Phys. Status Solidi A 190 (2002) 33.
- [26] K. Xu, N. Yano, A.W. Jia, A. Yoshikawa, K. Takahashi, J. Cryst. Growth 237-239 (2002) 998.

- [27] X. Wang, S.B. Che, Y. Ishitani, A. Yoshikawa, Jpn. J. Appl. Phys. 45 (2006) L730.
- [28] A. Yoshikawa, “Final report of InN-project as a CREST-program of JST” (in Japanese), JST (2008).

Table I. Comparison of MBE-grown In-polarity and N-polarity InN epilayers.

	In-polarity InN	N-polarity InN
Growth temperature	< ~500°C	< ~600°C
Surface flatness/morphology	Atomically flat, Monolayer steps	Columnar/grain-like rough, Bilayer steps
XRD-FWHM (002)	200~540 arcsec	200~250 arcsec
(102)	(800)~1100 arcsec	~650 arcsec
Residual electron density <i>a</i> RT	$2 \times 10^{17} \sim 3 \times 10^{18} \text{ cm}^{-3}$	$4 \times 10^{17} \sim 5 \times 10^{18} \text{ cm}^{-3}$
Electron mobility <i>a</i> RT	~1500-2150 cm^2/Vs	~1500-2200 cm^2/Vs

Figure Captions

Fig. 1. Schematic diagram MBE system with several in-situ and real time monitoring systems developed for III-nitrides epitaxy.

Fig. 2. AFM surface image of In-polarity InN epilayer (a), and dependence of rms roughness on growth temperature (b).

Fig. 3. Dependence on nominal InN supply of in-plane lattice constant of InN deposited at 500 °C on GaN (a), and corresponding change in pseudo-dielectric constant (b).

Fig. 4. XTEM images of InN epilayers grown on MOCVD-grown GaN template.

Fig. 5. Mott-Schottky plots for Mg-doped InN epilayers taking [Mg] as a parameter, (a): In-polarity and (b):N-polarity.

Fig. 6. Energy band diagram for ECV characterization on detecting the conduction type of bulk InN region located below high density surface electrons. The vertical thick lines indicate the Helmholtz layer in the electrolyte. (a) and (c) are for the zero-bias voltage for n-type and p-type samples, respectively. Correspondingly, (b) and (d) are for those under voltage-biased for detecting the minimum capacitance.

Fig. 7. Dependence of net acceptor/donor concentration in In-polarity InN epilayer on Mg concentration. Almost the same dependency was obtained for N-polarity samples.

Fig. 8. XRD 2θ - ω scan around GaN (0002) for the InN/GaN QWs grown at 650 °C.

Fig. 9. XTEM dark field image ($g=0002$) for the InN/GaN QWs grown at 600 °C.

Fig. 10. Sub-Å resolution Z-contrast image obtained by STEM for the 1 ML-thick ~InN embedded in GaN matrix grown at 650 °C.

Fig. 11. Room temperature PL spectra for InN/GaN QWs with different well thicknesses.

Fig. 12. EL spectra at RT of InN/GaN 5-QW LED under 2 % duty for 20 μ sec pulsed operation.

FIGURE 1

EIKO-MBE for InN/GaN/AlN epitaxy

FIGURE 2

FIGURE 3

FIGURE 4

FIGURE 5

FIGURE 6

FIGURE 7

FIGURE 8

FIGURE 9

FIGURE 10

FIGURE 11

FIGURE 12