Available online at www.sciencedirect.com
ScienceDirect
Thin Solid Films 515 (2006) 786 – 789
thin solid films
www.elsevier.com/locate/tsf

Effect of growth temperature on the luminescent and structural properties of InGaAsSbN/GaAs quantum wells for 1.3 μm telecom application

L. Borkovska [a,*], O. Yefanov [a], O. Gudymenko [a], S. Johnson [b], V. Kladko [a], N. Korsunska [a], T. Kryshtab [c], Yu. Sadofyev [b], Y.-H. Zhang [b]

[a] *Institute of Semiconductor Physics NASU, Pr. Nauki, 45, Kyiv, 03028, Ukraine*
[b] *Arizona State University, Tempe, AZ 85287, USA*
[c] *Department of Material Sciences, ESFM-IPN, Av. IPN, Ed.9 U.P.A.L.M., 07738, Mexico*

Available online 7 February 2006

Abstract

Variations of the characteristics of Sb-surfactant assisted grown InGaAsN/GaAs single quantum wells (QWs) in dependence on QW growth temperature (T_{GR}=442–505 °C) are investigated by the photoluminescence (PL) and high-resolution X-ray diffraction (HRXRD) methods.

The QWs grown at ~480 °C demonstrated optimal PL characteristics, namely the highest PL intensity and small potential fluctuations. A good quality of heterointerfaces is proved by HRXRD. These structures emit at ~1.29 μm at 300 K and are promising for application in long wavelength opto-electronic devices. The good structural properties of these QWs are assigned to Sb surfactant effect that allows shifting of the T_{GR} to higher temperatures without significant alloy decomposition. The increase of T_{GR} in its turn results in the decrease of the density of nonradiative defects that are the specific feature of low temperature growth.
© 2005 Elsevier B.V. All rights reserved.

PACS: 68.35 Ct; 61.10 Nz; 78.55.Cr; 78.67.De
Keywords: Surfactant-assisted growth; III–V semiconductors; Quantum well; HRXRD; PL

1. Introduction

An interest to InGaNAs/GaAs quantum well (QW) structures is caused by their high potential application in telecom lasers operating in 1.3–1.5 μm wavelength range. Room-temperature 1.3 μm InGaNAs/GaAs vertical-cavity surface-emitting lasers have been successfully demonstrated by a number of research groups [1–4]. The long-wavelength emission for InGaNAs alloy system is based on its uniquely strong bowing parameter of the bandgap energy vs. composition diagram [2]. Besides, due to a smaller lattice constant of GaN compared to GaAs, a near-lattice matching to GaAs is possible to achieve, when the In concentration is approximately three times that of the N concentration [1]. However, mixed group-III nitride–arsenides are known to suffer from alloy disorder and even from phase separation, which results in optical property degradation [5,6]. A common way to prevent this process is to use the non-equilibrium low-temperature growth conditions. But, a low temperature growth of In(Ga)As(N) QW materials was found to result in the formation of nonradiative defects [7–9], which can be eliminated partially upon post-growth annealing [5,7,9]. These call forth much narrower temperature window resulting in good crystallinity for GaInNAs than for GaInAs [5].

To improve the optical characteristics of N-containing III–V alloys a Sb-surfactant assisted growth was proposed [10–12]. It has been demonstrated that Sb is beneficial to reduce the surface faceting as well as to delay the formation of dislocations and acts as an efficient surfactant in the growth of GaN [13], GaInP [14], GaInAs [15,16] and GaInNAs [17]. The segregation of Sb at the growing surface that suppresses the surface diffusion was assumed to be responsible for the surfactant effect [10,15]. However until now information about the dependence of the properties of Sb-surfactant assisted grown InGaNAs QWs on the growth temperature is scarce.

Here we present the results of our study of the optical and structural properties of InGaNAsSb/GaAs QWs with low Sb (~1%) and low N (up to 1%) contents by the photolumines-

* Corresponding author. Tel.: +38 044 525 72 34; fax: +38 044 525 83 44.
E-mail address: bork@isp.kiev.ua (L. Borkovska).

0040-6090/$ - see front matter © 2005 Elsevier B.V. All rights reserved.
doi:10.1016/j.tsf.2005.12.194

GaAs cap layer 40 nm
$Al_{0.4}Ga_{0.6}As$ layer 50 nm
GaAs layer 40 nm
InGaAsNSb well 6 nm
GaAs layer 100 nm
$Al_{0.4}Ga_{0.6}As$ layer 50 nm
GaAs buffer layer
n-GaAs substrate (100)

Fig. 1. Schematic structure of the samples under investigation.

cence (PL) and high-resolution X-ray diffraction (HRXRD) methods. By varying the growth temperature, we established the optimal growth conditions for obtaining high efficiency QW emitting at 1.29 μm at room temperature.

2. Experimental details

The samples were grown by molecular beam epitaxy (MBE) on (001) GaAs substrates using a V80H VG system equipped with In, Ga and Al effusion cells, As and Sb valved cracking cells, and an Addon radio-frequency plasma cell for N. Growth mode was controlled by in situ reflection high-energy electron diffraction (RHEED). Each structure was grown on GaAs buffer layer and contained sequentially 50 nm thick $Al_{0.4}Ga_{0.6}As$, 100 nm thick GaAs layers, single InGaAsNSb QW, 40 nm thick GaAs, 50 nm thick $Al_{0.4}Ga_{0.6}As$ and 40 nm thick GaAs cap layers. The thickness of InGaNAsSb QWs was 6 nm, the Sb content was ~1%, while the In content was 36% and 38% in different samples. A schematic structure of the samples studied is presented in Fig. 1. The N content estimated from the PL spectra varied from ~0.5% to ~1% in different structures. Both the AlGaAs and GaAs barriers were grown at 590 °C and at As/III equivalent beam pressure ratio of 1.25. The QW growth temperature T_{GR} varied from 442 to 505 °C. The reference InGaAsSb/GaAs QW samples were also grown with an identical structure and under the same growth conditions. All N-containing samples were annealed under As vapour excessive pressure of $2 \cdot 10^{-6}$ Torr for 30 min at 730 °C in the MBE chamber immediately after growth.

The PL spectroscopy was performed at low (T=5 K) and room temperatures. The PL was excited with 514.5 nm line of an Ar^+ laser with a ~5 W/cm^2 power density and detected by a cooled Ge detector with a standard lock-in technique. The HRXRD measurements were carried out using a high resolution X-ray diffractometer Philips MRD with a 4×Ge (220) monochromator and Cu anode.

3. Results

Fig. 2 shows the (004) diffraction pattern of a reference InGaAsSb QW and two InGaAsNSb QWs with the lowest and the highest nitrogen content studied. The structures were grown

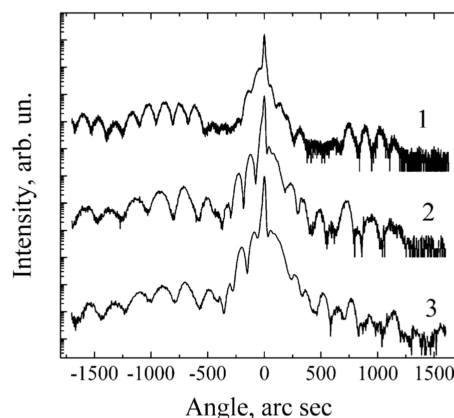

Fig. 2. HRXRD (004) rocking curves from the Sb-surfactant assisted grown $In_{0.36}Ga_{0.64}As$/GaAs (curve 1) and $In_{0.36}Ga_{0.64}AsN$/GaAs single QWs with N content of ~0.5% (curve 2) and ~1% (curve 3).

at 495–497 °C. The thicknesses of the barriers in the InGaAsSb QW structure were slightly larger than that in the N-containing samples. This is in agreement with smaller distance between the interference peaks in the InGaAsSb QW structure. Sharp and well-defined interference peaks indicate the abrupt interfaces in all samples. This means that adding N in small quantities to Sb-surfactant assisted grown InGaAs QWs does not change noticeably the interface sharpness. In the InGaAsNSb QWs the separation between the interference peaks remains the same independently of N content, while the first-order satellite peak corresponding to InGaAsNSb shifts towards the GaAs substrate peak when N composition increases. The latter testifies to the decrease of lattice mismatch strain in the structure as the N content increases.

Fig. 3 shows the low-temperature PL spectra of the samples whose diffraction patterns are presented in Fig. 2. As one can see, the adding of N to InGaAsSb QW has a strong influence on the PL spectrum, namely it leads to: (i) the red shift of the PL peak position, (ii) the decrease of the PL intensity, and (iii) the increase of the PL line full width at a half maximum (FWHM). These results are in agreement with other data [5,7,8]. If one supposes the shift of 150 meV per 1% of N for the structure with strain [2], the N content could be estimated to

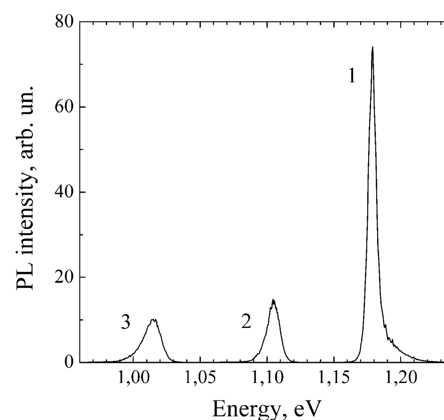

Fig. 3. PL spectra of the Sb-surfactant assisted grown $In_{0.36}Ga_{0.64}As$/GaAs (curve 1) and $In_{0.36}Ga_{0.64}AsN$/GaAs single QWs with N content of ~0.5% (curve 2) and ~1% (curve 3). T=5 K, E_{exc}=2.41 eV.

be ~0.5% and ~1.0% in the InGaAsNSb QWs investigated. The decrease of the PL efficiency in the N-containing samples indicates that introduction of N results in the formation of non-radiative defects that are not removed completely by post-growth annealing. These defects can be ion-damage defects or charged species originating from the N-plasma source [7], as well as background impurities (carbon- or oxygen-contaminations) [5,18] or N interstitials [3].

To study the influence of the growth temperature on the optical properties of InGaAsNSb QWs, the low-temperature PL spectra of two sets of the samples with nitrogen content of ~0.8% and 1% grown at 442–505 °C were analysed. It was found that the PL characteristics of the InGaAsSbN QWs, i.e., the PL intensity, FWHM and peak position changed non-monotonically with the T_{GR} (Fig. 4). When the T_{GR} rises from 442 to 480 °C, the PL intensity of the QWs with the nitrogen content of ~0.8% increases more than in order of value, while the FWHM remains basically constant (~11 meV) and the PL peak position shifts slightly to the high energy region. When the T_{GR} rises from 480 to 505 °C, the PL intensity and peak position do not change while the FWHM starts to increase. In the QWs with the highest nitrogen content the rise of the T_{GR} from 480 to 505 °C results in the saturation and following decrease of the PL intensity, significant increase of the FWHM and red shift of the PL peak position.

It should be noted, that no significant broadening of the interference fringes was observed in (004) symmetrical as well as in (224) asymmetrical reflections of all structures studied. This implies a good quality of heterointerfaces as well as that the layers are coherently strained and two-dimensional growth mode remains. The latter is in agreement with the RHEED observations during the growth.

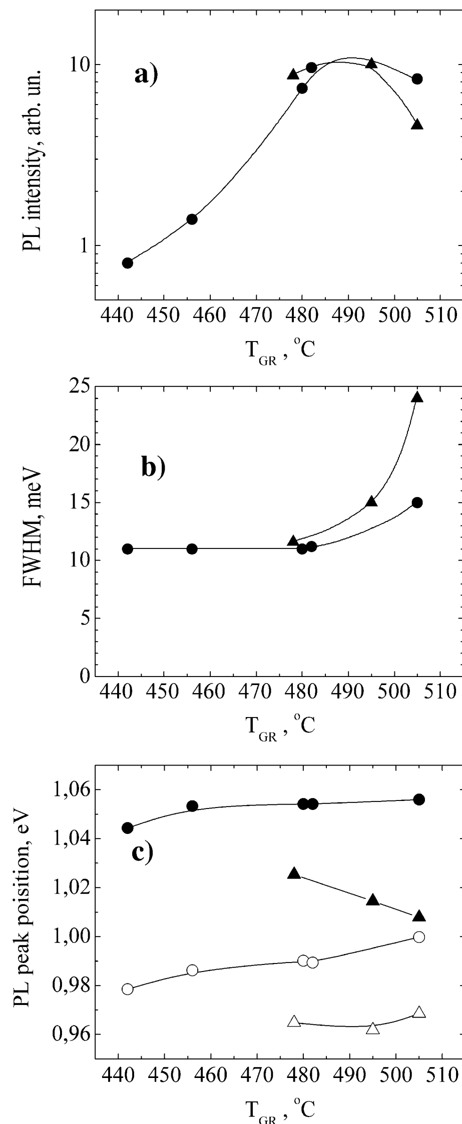

Fig. 4. The QW PL intensity, FWHM and peak position dependencies on the QW growth temperature for the Sb-surfactant assisted grown $In_{0.38}Ga_{0.62}AsN$/GaAs single QWs with N content of ~0.8% (circles) and $In_{0.36}Ga_{0.64}AsN$/GaAs single QWs with N content of ~1% (triangles) measured at T=5 K (solid symbols) and 300 K (open symbols). E_{exc}=2.41 eV.

4. Discussion

As Fig. 4a shows, the highest PL intensity is found in the QWs grown at 480–495 °C. Degradation of the PL efficiency observed both at lower and at higher T_{GR} is obviously due to the formation of non-radiative defects in the QWs because the intensity of the PL from the barriers (not shown in the Fig. 3) remains basically the same in all samples studied. Two processes dependent on the growth temperature can be responsible for non-radiative defect formation in the N-containing alloys [5]: the low temperature growth per se and alloy decomposition.

The former process was found to be common for In(Ga)As(N) QW materials [8,9], the lower the temperature, the higher defect density. These defects were thought to be native point defects (As_{Ga}-related complexes) [9] and background impurities [5,18]. It is obvious that this process is responsible for the PL intensity degradation when the T_{GR} decreases from 480 to 440 °C.

Alloy decomposition is supposed to be the reason of the increase of non-radiative defect density at T_{GR}>480 °C. This is in agreement with steady increase of the PL efficiency observed by us in the InGaAsSb QWs (not shown here) when the growth temperature rose from 465 to 525 °C. At the same time in N-containing samples the PL intensity saturated (when N content was ~0.8%) or decreased (when N content was ~1%) at T_{GR}>480 °C.

The conclusion made is also proved by the comparison of the PL peak positions measured at low and room temperatures (Fig. 4c). In all samples the PL peak position measured at room temperature (Fig. 4c, open symbols) shifts to the high energy region when the T_{GR} increases from 442 to 505 °C. This can be explained by the increase of In reevaporation as the T_{GR} rises. At the same time the PL peak position measured at low temperature (Fig. 4c, solid symbols) shows the same dependence only in the temperature range 442–480 °C. When the T_{GR} rises from 480 to 505 °C it remains constant (in the structures with N content ~0.8%) or red shifts (in the samples with N content ~1%). Several investigations [8,19–21] have shown that at low temperatures excitons in N-containing QWs are strongly

localized on potential fluctuation of the band edge caused by local compositional disorder or interface roughness. At room temperature excitons are delocalized. Therefore, change of room temperature PL peak position with the T_{GR} reflects the change of quantized level position in the QW. The facts that in the range T_{GR}=480–505 °C the PL peak position measured at low temperature does not show a blue shift or demonstrates a red shift can be assigned to the increase of potential fluctuations in the QW. It should be noted that potential fluctuations apparently do not change in the range T_{GR}=442–480 °C.

These conclusions are in agreement with the FWHM dependence on the T_{GR}. In fact, the increase of potential fluctuations should result in the inhomogeneous broadening of the localized exciton band [22]. As it follows from Fig. 4b, in the range of T_{GR}=442–480 °C the FWHM remains basically constant (~11 meV) and close to that observed by us in the InGaAsSb QW grown at 480 °C (~10 meV). This testifies to small and independent on T_{GR} potential fluctuations introduced by nitrogen in the QW at these T_{GR}. The rise of the FWHM value in the range of T_{GR}=480–505 °C indicates the increase of potential fluctuations in the QW, the higher N content, the deeper potential fluctuations. As it follows from the X-ray diffraction data, the QW interface remains abrupt at all T_{GR} used. This means that potential fluctuations in the QW are mainly due to composition fluctuations.

Therefore, the increase of non-radiative defect density al low T_{GR} is caused by the low-temperature growth conditions, while at T_{GR}>480 °C it is due to alloy decomposition.

The results obtained show that the decrease of the growth temperature usually used to prevent phase separation in the N-containing alloys improves optical characteristics of the QW studied only down to T_{GR} ~480 °C. Further temperature decrease does not reduce potential fluctuations in the QW, but results in additional non-radiative defect formation. The structures grown at ~480 °C demonstrated an optimal ratio of the PL intensity to the PL line half-width. In these samples the longest wavelength emission was observed at 1.29 μm at 300 K. This T_{GR} value is higher than that found by other investigators for InGaAsN QWs [4–6]. In Ref. [6] two-dimensional growth mode resulted in the best PL efficiency was achieved only at low growth temperatures (~400 °C). When the growth temperature was increased to 450 °C, the InGaAsN/GaAs QW interface became rough and at 475 °C the formation of an island-like structure was observed. In Ref. [4] the highest PL intensity was found in the samples grown at 450 °C.

We believe that the reason of improved PL characteristics observed in our samples is the Sb surfactant effect. Sb-surfactant assisted growth gives possibility to shift the T_{GR} to higher temperatures without significant alloy decomposition. The increase of T_{GR} in its turn results in the decrease of the density of non-radiative defects that are the specific feature of low temperature growth.

5. Conclusion

In the present paper the results of the PL and HRXRD studies of InGaAsSbN/GaAs single QWs with low nitrogen content (≤1%) grown by MBE in the temperature range 440–505 °C are presented.

The highest PL intensity is found in the QWs grown at 480–495 °C. Analysis of the QW PL spectra shows that degradation of the PL intensity al lower T_{GR} is caused by the increase of non-radiative defect density arising from the low-temperature growth conditions, while at T_{GR}>480 °C it is due to alloy decomposition. The latter process is accompanied not only by the decrease of the PL intensity, but also by the increase of potential fluctuations in the QW. It is found that at T_{GR}<480 °C the potential fluctuations do not depend on T_{GR}. Since HRHRD demonstrates a good quality of heterointerfaces in all samples studied, the potential fluctuations in the QW are supposed to be mainly due to composition fluctuations.

Thus, the use of Sb as a surfactant allows the temperature range shifting of QW alloy decomposition to higher temperatures and growing high efficiency InGaAsN QW with small potential fluctuations suitable for telecom applications.

References

[1] V.M. Ustinov, A.E. Zhukov, Semicond. Sci. Technol. 15 (2000) R41.
[2] M. Kondow, T. Kitatani, S. Nakatsuka, M.C. Larson, K. Nakahara, Yo. Yazawa, M. Okai, K. Uomi, IEEE J. Sel. Top. Quantum. Electron. 3 (1997) 719.
[3] S.G. Sprutte, M.C. Larson, W. Eampler, C.W. Colden, H.E. Petersen, J.S. Harris, J. Cryst. Growth 227 (2001) 506.
[4] M. Pessa, C.S. Peng, T. Jouhti, E.-M. Pavelescu, W. Li, S. Karirinne, H. Liu, O. Okhotnikov, Microel. Eng. 69 (2003) 195.
[5] M. Kondow, T. Kitatani, Semicond. Sci. Technol. 17 (2002) 746.
[6] J.-M. Chauveau, A. Trampert, M.-A. Pinaults, E. Tournie, K. Du, K.H. Ploog, J. Cryst. Growth 251 (2003) 383.
[7] Z. Pan, L.H. Li, W. Zhang, Y.W. Lin, R.H. Wu, Appl. Phys. Lett. 77 (2000) 1280.
[8] M.-A. Pinault, E. Tournie, Sol.-St. Electron. 47 (2003) 477.
[9] N.Q. Thinh, I.A. Buyanova, W.M. Chen, H.P. Xin, C.W. Tu, Appl. Phys. Lett. 79 (2001) 3089.
[10] X. Yang, M.J. Jurkovic, J.B. Heroux, W.I. Wang, Appl. Phys. Lett. 75 (1999) 178.
[11] H. Shimizu, K. Kumada, S. Uchiyama, A. Kasukawa, Electron. Lett. 36 (2000) 1379.
[12] T.S. Kim, T.V. Cuong, C.S. Park, J.Y. Park, H.J. Lee, E.-K. Suh, C.-H. Hong, J. Cryst. Growth 263 (2004) 156.
[13] L. Zhang, H.F. Tang, T.F. Kuech, Appl. Phys. Lett. 79 (2001) 3059.
[14] J.K. Shurtleff, R.T. Lee, C.M. Fetzer, G.B. Stringfellow, Appl. Phys. Lett. 75 (1999) 1914.
[15] J.C. Harmand, L.H. Li, G. Patriarche, L. Travers, Appl. Phys. Lett. 84 (2004) 3981.
[16] T. Kageyama, T. Miyamoto, M. Ohta, T. Matsuura, Ya. Matsui, T. Furuhata, F. Koyama, J. Appl. Phys. 96 (2004) 44.
[17] X. Yang, J.B. Heroux, L.F. Mei, W.I. Wang, Appl. Phys. Lett. 78 (2001) 4068.
[18] H. Carrere, A. Arnoult, A. Ricard, X. Marie, Th. Amand, E. Bedel-Pereira, Sol.-St. Electron. 47 (2003) 419.
[19] I.A. Buyanova, W.M. Chen, G. Pozina, P.N. Hai, B. Monemar, H.P. Xin, C.W. Tu, Mater. Sci. Eng., B 82 (2001) 143.
[20] M. Polimeni, M. Capizzi, M. Geddo, M. Fischer, Reinhardt, Forchel, Appl. Phys. Lett. 77 (2000) 2870.
[21] S. Mazzucato, A. Erol, R.J. Potter, N. Balkan, P.R. Chalker, S. Thomas, T.B. Joyce, T.J. Bullough, Sol.-St. Electron. 47 (2003) 483.
[22] K.K. Bajaj, Mater. Sci. Eng., R 34 (2001) 59.