

Comparison of GaNAsSb and GaNAs as quantum-well barriers for GaInNAsSb optoelectronic devices operating at 1.3–1.55 μm

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GaNAsSb/GaAs quantum wells were grown by solid-source molecular-beam epitaxy utilizing a radio-frequency nitrogen plasma source. The GaNAsSb layers, originally the quantum well barrier materials for GaInNAs(Sb) devices, were studied for their general growth characteristics as well as their structural and optical properties, which give an indication of its quality as a quantum well barrier material. Reflection high-energy electron diffraction, high-resolution x-ray diffraction, secondary-ion mass spectroscopy, and photoluminescence (PL) measurements were used to study those properties. The growth parameters including arsenic overpressure and substrate temperature were changed systematically to determine the properties during deposition and to optimize these conditions. It was found that the addition of antimony to GaNAs did not improve the material as it did for GaInNAs. PL measurements indicated a decreasing optical quality with an increasing substrate temperature and no change with the arsenic overpressure. In addition, the addition of antimony had eliminated the lattice-strain compensation provided by the GaNAs barriers. Using GaNAs rather than GaNAsSb barriers in the GaInNAs(Sb) devices was found to be advantageous and has dramatically improved the performance of long-wavelength GaAs-based lasers. © 2004 American Institute of Physics. [DOI: 10.1063/1.1807028]

I. INTRODUCTION

The dilute-nitride GaInNAs has been found to optically emit at wavelengths longer than what previously thought possible for materials grown coherently on GaAs. Kondow *et al.* discovered that the addition of nitrogen to GaAs decreases both the overall lattice parameter and the band gap.¹ This has enabled the development of lasers at the important telecommunication wavelength of 1.3 μm .^{2–4} However, the attempts to push out to 1.55 μm have not been entirely successful with GaInNAs. The addition of further indium or nitrogen results in a severely degraded material quality due to phase segregation or relaxation.⁵

In an effort to improve the material quality, Shimizu *et al.* and Yang *et al.* discovered that antimony could be used as a surfactant in InGaAs and GaInNAs growth, respectively.^{6,7} However, it was discovered that antimony acted as both a surfactant and a constituent when used in GaInNAs, forming GaInNAsSb.^{5,8} Until recently, devices utilizing GaInNAsSb as the quantum well (QW) material used GaNAsSb as the barrier material.^{9,10} Antimony was used in the barriers because it was also thought to be a surfactant that would remain on the surface during the QW and barrier growth even if the antimony shutter was closed. It could also improve the quality of GaNAs and the interfaces between the QW and the barriers. Although GaInNAsSb has been extensively studied as the quantum well material, there have been only a few published studies of GaNAsSb by one research group.^{11,12}

In this paper, we investigate the effects of growth conditions on the material properties of GaNAsSb. Characteristics such as lattice strain and optical quality and intensity are studied in depth to examine the results of altering the material properties. We also discuss the potential effects of utilizing GaNAs barriers rather than GaNAsSb barriers for the GaInNAsSb QWs operating in the 1.3–1.55 μm range. Reflection high-energy electron diffraction (RHEED), high-resolution x-ray diffraction (HRXRD), photoluminescence (PL), and secondary-ion mass spectrometry (SIMS) measurements were obtained to analyze the characteristics of the material grown. The GaNAsSb compositions and thicknesses in this study were chosen to be similar to those utilized in the previous GaInNAsSb/GaNAsSb structures from this research group.^{5,9,13}

II. EXPERIMENTAL DETAILS

The GaNAsSb/GaAs QW samples prepared for this study were grown on *n*-type (100) GaAs substrates by solid-source molecular-beam epitaxy (MBE) in a Varian Mod Gen-II system. Nitrogen was supplied by a modified SVT Associates plasma operating at a rf of 13.56 MHz. A standard Knudsen effusion cell was used to supply gallium. A valved arsenic cracker supplied As_2 and an unvalved antimony cracker supplied $\sim 100\%$ Sb_1 .

To study the properties of the GaNAsSb material used in the barriers of previous 1.3 and 1.5 μm laser devices, the barriers themselves were “converted” into single-QW (SQW) samples. A series of samples of 20 nm GaNAs(Sb) QWs with GaAs barriers were grown to examine the different

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growth conditions, such as the substrate temperature and arsenic overpressure. The compositions of GaNAs(Sb) used in this study were chosen to be similar to those used in the 1.3 and 1.55 μm GaInNAsSb devices and are uniquely determined due to the restrictive nature of MBE. Since the MBE chamber only has one gallium cell, the flux is fixed during the entire growth and is preferentially set to obtain the proper GaInNAsSb QW composition. The nitrogen content is predetermined due to its inverse proportional relationship with the group-III flux and the antimony flux is unchangeable because it is supplied by an unvalved cracker.¹⁴ These combined conditions do not allow the barrier compositions to be arbitrarily changed.

The GaNAs(Sb) SQWs were all grown at 425 °C (except for the substrate temperature study) and the growth rate was either 0.45 $\mu\text{m}/\text{h}$ (to duplicate the 1.3- μm QW barriers) or 0.30 $\mu\text{m}/\text{h}$ (to duplicate the 1.5- μm QW barriers). The composition of nitrogen in GaNAs was determined by HRXRD, and nitrogen and antimony in GaNAsSb by SIMS and HRXRD. As mentioned earlier, the QW thicknesses were all 20 nm with 50 nm GaAs capping layers. An arsenic-to-gallium overpressure of 20 \times (except during the arsenic overpressure study) and an antimony flux of 0.8–1.0 $\times 10^{-7}$ Torr beam equivalent pressure were supplied during the QW growth. The HRXRD was obtained with a Philips X'Pert Pro-MRD four-crystal high-resolution x-ray diffractometer. The PL measurements were performed at room temperature with an argon-ion laser and an InGaAs detector. The SIMS analysis was outsourced to Charles Evans & Associates.

III. RESULTS AND DISCUSSION

A. Initial findings

RHEED is a good method to examine the quality and properties of a growth surface. During the growth of the GaNAs and GaNAsSb samples, RHEED patterns were recorded to examine any significant differences on the surface in the presence of antimony. Shown in Figs. 1(a) and 1(b) are the patterns obtained from the GaNAs and GaNAsSb samples, respectively. It is seen that, although very faint, the RHEED pattern from GaNAs is streaky. Although not shown in the figure, the orthogonal [110] directions for GaNAs grown under the growth conditions mentioned earlier showed a 2 \times 4 reconstruction. When antimony was applied to form GaNAsSb, the RHEED pattern changed significantly. Instead of a streaky pattern, a spotty pattern emerged, suggesting that the surface quality was not as good as GaNAs. This result was slightly surprising because previous studies have shown that antimony had the opposite effect with InGaAs and GaInNAs.^{6,7} Again, in examining the orthogonal [110] directions, the reconstruction here appeared to be 1 \times 4. From the RHEED, it appears that the growth surface of GaNAs is smoother than that of GaNAsSb.

Figure 2 shows the HRXRD (004) $\omega/2\theta$ scans from four different GaNAs(Sb) SQW samples. The nitrogen compositions of the GaNAs samples were determined by a dynamical simulation of the HRXRD spectra. For the GaNAsSb samples, simulations were also performed using the values

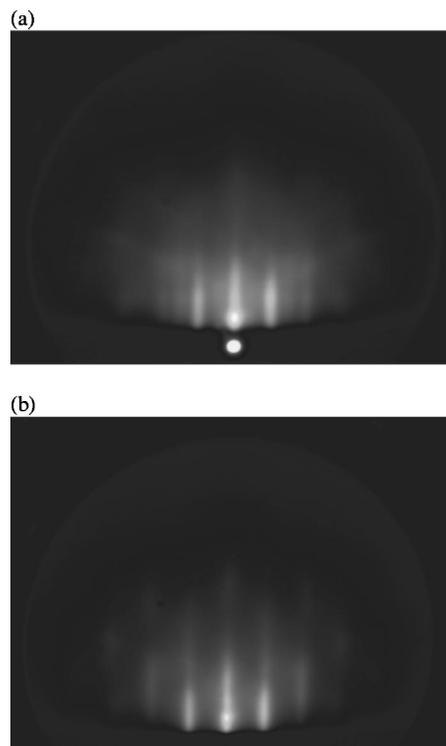


FIG. 1. RHEED pictures showing the streaky patterns from (a) GaNAs and the spotty patterns from (b) GaNAsSb.

obtained for the antimony and nitrogen compositions from the SIMS to confirm the validity of the SIMS measurements. The results for the 1.3 and 1.55 μm device growth conditions are shown in Table I. As expected in the 1.55 μm device growth conditions, there was more nitrogen found in both samples due to the slower growth rate. For a fixed nitrogen-to-arsenic flux ratio, varying the growth rate will alter the amount of nitrogen incorporated as mentioned earlier. However, for a fixed antimony-to-arsenic flux ratio, the variations in the growth rate did not change the antimony

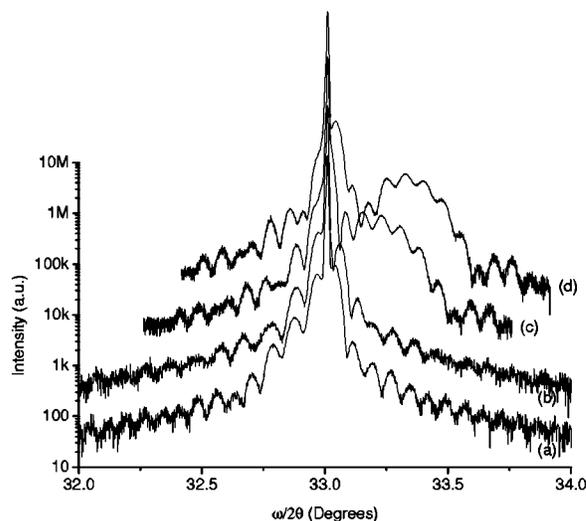


FIG. 2. (004) $\omega/2\theta$ XRD spectra showing the amount of strain in the samples. (a) GaN_{0.029}As_{0.873}Sb_{0.098}, (b) GaN_{0.034}As_{0.867}Sb_{0.099}, (c) GaN_{0.019}As_{0.981}, and (d) GaN_{0.027}As_{0.973}. (a) and (c) are grown under the 1.3 μm device growth conditions, whereas (b) and (d) are grown under the 1.55 μm device growth conditions.

TABLE I. XRD and SIMS compositional results of GaNAs and GaNAsSb grown under the normal 1.3 and 1.55 μm device growth conditions.

	GaNAs	GaNAsSb
1.3 μm growth conditions	1.9% N	2.9% N, 9.8% Sb
1.55 μm growth conditions	2.7% N	3.4% N, 9.9% Sb

composition. The antimony incorporation appeared to be independent of the altered growth conditions such as the group-III growth rate and the nitrogen incorporation, suggesting that the flux ratio of antimony-to-arsenic is the deciding factor in determining the composition. This is different than what is seen in GaAsSb growth, where for fixed antimony-to-arsenic flux ratios, different growth rates lead to different incorporation rates.¹⁵ The increase in the nitrogen composition in GaNAsSb, compared to GaNAs in both conditions, was expected and not surprising because this has been reported previously.^{10,13} The mechanism for the increased sticking of nitrogen in the material is not known. However, it is thought that the properties of antimony as a “reactive surfactant” help promote the incorporation of nitrogen into GaAs.^{16,17} As observed from the HRXRD scans, the GaNAsSb was either lattice-matched to GaAs or was very slightly compressively strained for both the compositions. This property is not advantageous when used with the highly compressively strained GaInNAs(Sb) QW materials because the barriers would not provide any strain compensation in the active regions. However, both the compositions of GaNAs showed an appreciable amount of tensile strain. The amount of strain found in $\text{GaN}_{0.019}\text{As}_{0.981}$ and $\text{GaN}_{0.027}\text{As}_{0.973}$ was -0.38% and -0.55% , respectively. From the HRXRD, there does not appear to be much of an improvement or a degradation of material quality upon the addition of antimony to GaNAs for either set of growth conditions. From the Pendellosung fringes, the interfaces for all the materials are of good quality. This could suggest that the GaNAs grown was already of excellent quality or there simply was no effect upon an addition of antimony.

SIMS was performed on the GaNAsSb samples mentioned earlier. Figure 3 shows the SIMS depth profile for the $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample. Although it is relatively straightforward to obtain the depth profiles, attaining the exact compositional values requires a previous calibration due

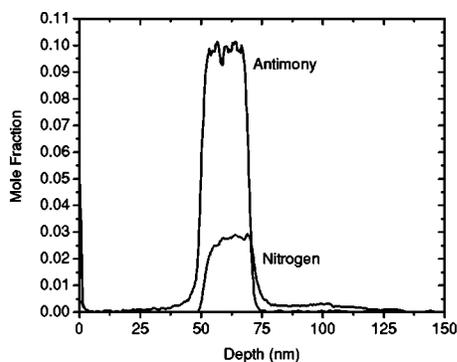


FIG. 3. SIMS depth profile of antimony and nitrogen for a $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample.

to artifacts, such as matrix effects. The SIMS analysis was calibrated by Charles Evans & Associates using the parameters obtained from past growths analyzed with nuclear reaction analysis Rutherford backscattering (RBS) for the nitrogen and particle-induced x-ray emission RBS for antimony.⁵ As mentioned earlier, the compositions were confirmed with the HRXRD data and simulations. An interesting feature to note in the SIMS depth profile is the top interface of the GaNAsSb layer. The nitrogen and antimony profiles do not end at the same location within the sample. This was determined not to be a measurement or a sputtering artifact, as it was repeatable within the same sample and was seen on all the other samples measured with SIMS. Upon examination of the SIMS depth profile, it appears that antimony continues to incorporate $\sim 5\text{--}7$ nm beyond the end of the nitrogen incorporation. Surfactants tend to float on the growth front and do not incorporate into the material. As mentioned in earlier papers relating to GaInNAs(Sb), antimony appears to act as a surfactant and as a group-V component.^{5,13} It is very probable that antimony both incorporates and floats on the growth front for GaNAsSb as well. If this is the case, there will be antimony remaining on the growth front after the shutter is closed to the cell and the residual antimony continues to incorporate or desorb until the supply is exhausted. This growth artifact could be quite detrimental to the devices because there is a thin layer of GaAsSb that could significantly change the originally intended band-structure properties due to the changes of both the composition and strain.

PL measurements were obtained from the $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample. No signal was observed for the $\text{GaN}_{0.034}\text{As}_{0.867}\text{Sb}_{0.099}$ sample, either as-grown or annealed. This suggests that the material was of a very poor optical quality and was not studied further. The PL obtained from the as-grown $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample peaked at 1.316 μm , but was very weak in intensity. This was not surprising because most groups report poor PL intensity for the GaInNAs(Sb) samples which have not been annealed.^{5,8,14} The sample was annealed at a series of temperatures between 720 $^{\circ}\text{C}$ and 820 $^{\circ}\text{C}$ to study the effect upon the optical quality of the material. Similar to GaInNAs(Sb), annealing the PL samples led to a dramatic increase in the PL intensity. As seen in Fig. 4, the PL intensity increased with increasing annealing temperatures until it peaked at 760 $^{\circ}\text{C}$ and decreased beyond this point. The PL peak wavelength also blueshifted with increasing annealing temperatures. Compared to the as-grown PL spectrum, the optimally annealed PL signal was $25\times$ higher in intensity and was blueshifted 70 nm, which is slightly more than the blueshift found in GaInNAs(Sb). Although the blueshifting of the wavelength was expected and is seen in all nitride-arsenide samples, it is interesting to note that the wavelength essentially remains the same past the optimal 760 $^{\circ}\text{C}$ annealing temperature. Unlike GaInNAs(Sb) samples, there is no indium in the samples, and thus, the blueshifting of the PL wavelength upon annealing cannot be explained by the In/Ga/N rearrangement.^{18–20} The sources of blueshifting likely include nitrogen outdiffusion, N/As/Sb rearrangement, and nitrogen declustering. When compared to the typical GaIn-

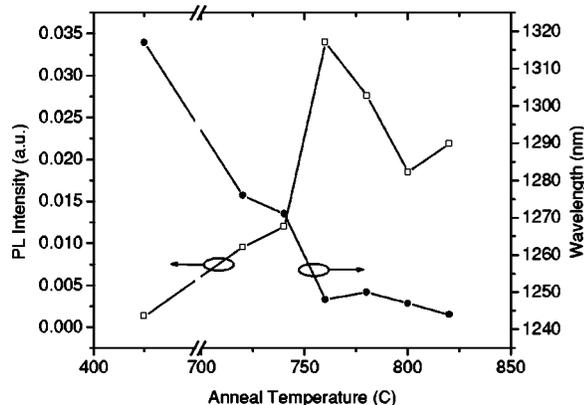


FIG. 4. PL results from the $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample (barrier material for the $1.3\ \mu\text{m}$ QWs). The filled dots show the PL intensity. The empty dots show the peak PL wavelength.

NAs(Sb) PL intensities, the GaNAsSb intensities are at least $25\times$ lower. The low intensity, in comparison to the other nitride-arsenides, could be due to the poor optical material quality or poor band alignment in the active region design. It is unclear which reason is the cause. One final point to note is the actual transition energy of the $\text{GaN}_{0.034}\text{As}_{0.867}\text{Sb}_{0.099}$ sample in comparison with the QW material it surrounds in the devices. If it is assumed that the PL peak wavelength gives a rough estimate of the band gap of the material, then it is seen that the band gap of the $\text{GaN}_{0.034}\text{As}_{0.867}\text{Sb}_{0.099}$ is roughly $0.99\ \text{eV}$ whereas the QW at $1.3\ \mu\text{m}$ is $0.95\ \text{eV}$. With only a $40\ \text{meV}$ difference in the band gap, there is a very poor confinement of electrons and holes within the QW, and it is possible that the alignment between the GaNAsSb and GaInNAsSb at $1.3\ \mu\text{m}$ is not the desired type-I alignment.

B. Arsenic overpressure study

In an attempt to study the general growth properties of GaNAsSb and to potentially improve the PL intensity and material quality, a series of samples with varying arsenic-to-gallium overpressures was grown with the same structure as the previous samples. It is known that in the mixed group-V materials, the relative fluxes of each group-V element play a large role in the composition and growth kinetics. In GaInNAs, there was no significant effect on the nitrogen incorporation by different arsenic fluxes due to the “unity” sticking properties of nitrogen.¹⁴ However, in GaNAsSb, it is suspected that the arsenic and antimony fluxes do indeed affect each other because they do not have the same sticking properties as nitrogen. It is also possible that a variation in the antimony incorporation could affect the nitrogen composition. To test the effects of arsenic overpressure on GaNAsSb, the original $20\times$ arsenic-to-gallium flux overpressure was varied among $15\times$, $25\times$, and $30\times$ when growing $\text{GaN}_{0.034}\text{As}_{0.867}\text{Sb}_{0.099}$, which is the material used as barriers for the $1.3\ \mu\text{m}$ QWs. All the other growth conditions were held constant. In the HRXRD measurements, it was seen that as the arsenic overpressure increased from $15\times$ to $30\times$, the strain in the GaNAsSb layer became less compressive. Since this is a quaternary system, it cannot be determined whether the decrease in the compressive strain is due to a

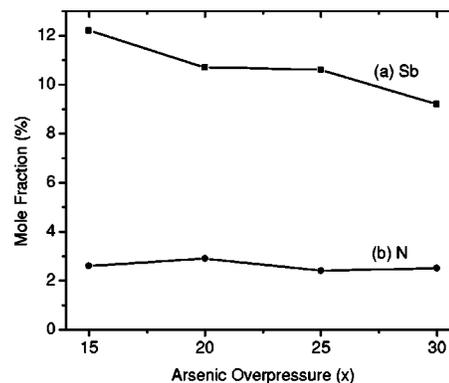


FIG. 5. SIMS results from the arsenic overpressure study. (a) Antimony composition and (b) nitrogen profile.

reduction in the antimony concentration, an increase in the nitrogen concentration, or a combination of both. In all cases, the HRXRD scans did not show any degradation of material compared to the original $20\times$ sample. To determine the origin of the strain reduction, SIMS was performed to measure the composition. Figure 5 plots the results obtained from the SIMS analysis. As the arsenic overpressure is increased from $15\times$ to $30\times$, the data shows that the antimony concentration drops from 12% to 9% , whereas the nitrogen concentration remains roughly constant. This would explain the decrease in the compressive strain with increasing arsenic flux because a reduction in antimony would decrease the lattice constant of GaNAsSb. According to the SIMS data, the increase in arsenic flux has a direct effect on the antimony incorporation rate but had no discernable effect on the nitrogen incorporation (as seen in GaNAs). This decrease in the antimony incorporation with increasing arsenic flux is seen commonly in GaAsSb growth.¹⁵ In addition, the change in the antimony concentration had no effect on the nitrogen incorporation in agreement with the previously obtained results.^{10,13} The data also show an enhanced nitrogen incorporation in the GaNAsSb. GaNAs grown under the same growth conditions yields 1.8% N, much lower than the observed $2.4\text{--}2.9\%$ in GaNAsSb. The PL measurements revealed no significant change in the optical material quality between the different samples. The PL spectra from the varying arsenic overpressure samples are shown in Fig. 6 and are from the optimal annealing temperature. The PL wavelengths

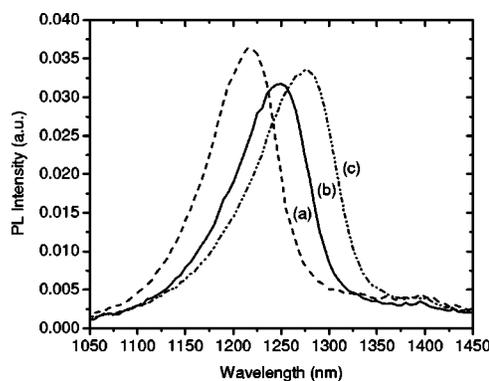


FIG. 6. PL spectra from the $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample grown at different arsenic-to-gallium overpressures. (a) $30\times$, (b) $25\times$, and (c) $15\times$.

were different due to the changing antimony concentration. Going from $15\times$ to $30\times$ arsenic overpressure, we have a 3% Sb decrease, resulting in a peak wavelength shift from 1.275 to 1.220 μm . The PL intensities are within the measurement error and sample repeatability. Changing the arsenic overpressure does not seem to have any major effect on GaNAsSb except for the change in the antimony incorporation. The material quality remains the same, structurally and optically. For the GaInNAs(Sb) QWs with GaNAsSb barriers, it would be beneficial to increase the arsenic overpressure so that the GaNAsSb obtained would have less compressive strain and have a larger band gap for an increased confinement in the wells.

C. Substrate temperature study

The substrate temperature during GaNAsSb growth was also varied to examine the effects on crystal quality and composition. GaInNAs(Sb) was grown at 425 $^{\circ}\text{C}$ to prevent phase segregation and relaxation. One of the driving factors of segregation in GaInNAs(Sb) is the clustering of the indium-rich areas. The GaNAsSb barriers were also grown at the same temperature because it was also thought that the material would segregate. However, indium is not present in this material and thus raised the possibility that the material could be grown at a higher temperature. One problem with nitride-arsenide growth is the low substrate temperature. These low temperatures introduce defects in GaAs materials, such as arsenic antisites and gallium vacancies. It is ideal to grow the material as close to 580 $^{\circ}\text{C}$ as possible to minimize these defects, which may cause nonradiative recombination and a reduced luminescence. A series of samples with the structures and growth conditions identical to those in the initial study were grown with varying substrate temperatures: +50 $^{\circ}\text{C}$ (475 $^{\circ}\text{C}$), +100 $^{\circ}\text{C}$ (525 $^{\circ}\text{C}$), and +150 $^{\circ}\text{C}$ (575 $^{\circ}\text{C}$). Another set of samples with no antimony (GaNAs) was also grown for a comparison. The HRXRD scans of the GaNAs and GaNAsSb showed that temperature did have an effect on the composition (and thus on the strain and structure) of the material. When the substrate temperature was increased, the strain in the GaNAsSb samples shifted from slightly compressive to slightly tensile. As mentioned before, it is unclear whether this is due to a reduction in antimony or an increase in nitrogen incorporation. It is suspected that the shift in the strain is most likely due to a reduction in antimony because it is known that antimony tends to desorb more readily at higher growth temperatures. It is possible that the change in the strain is also due to an increase in the nitrogen incorporation rate, although it is highly unlikely. To determine whether or not temperature has an effect on nitrogen incorporation, GaNAs was grown with varying substrate temperatures as well. From the HRXRD scans, it was seen that the nitrogen composition remained the same, except for the hottest sample in which there was a slight decrease in nitrogen. This would suggest that an increase in the nitrogen incorporation at higher temperatures is not the reason for the shift in the strain in GaNAsSb at higher temperatures. The most surprising result from the HRXRD scans was the fact that the samples did not appear

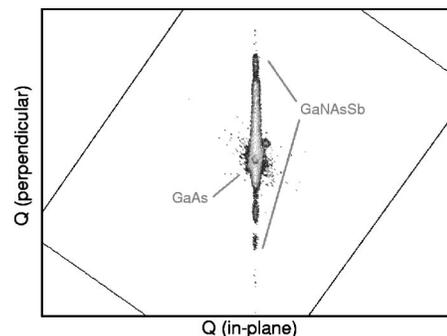


FIG. 7. Reciprocal space map of the $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample grown at a high temperature (575 $^{\circ}\text{C}$). No in-plane components from the QW are seen in the diffraction pattern.

to have any significant segregation, relaxation, or interface degradation when grown at higher temperatures. All the samples had very well-defined QW peaks and Pendellosung fringes. To confirm that the (004) $\omega/2\theta$ scans were not missing any signs of relaxation or segregation, reciprocal space maps (RSMs) were taken of the (224) directions of GaNAs and GaNAsSb. Figure 7 shows the RSM of GaNAsSb grown at 575 $^{\circ}\text{C}$. The GaNAs RSM was not shown because it looks very similar to that of GaNAsSb. The RSMs show the in-plane and perpendicular components of diffraction and the lattice spacing of the material present. If a material is grown coherently, there should be there are no major diffraction peaks in the in-plane direction away from the (224) GaAs peak. However, there is a slight peak to the upper right of the (224) GaAs signal, suggesting that there may be some small amount of segregation and/or relaxation. Since the peak is in the tensile direction compared to the GaAs peak, it would suggest areas of the enhanced nitrogen concentration. The distance of this peak from the main substrate peak is very small and could not be relaxed GaN clusters. It is possible that these are regions of GaNAs(Sb) with higher concentrations of nitrogen than found in the rest of the material. This small peak is found in all nitride-arsenide samples grown at all temperatures, suggesting that it is an inherent property of the growth. Further analysis is required to determine the origin of this peak. SIMS scans were taken of the GaNAsSb samples to measure the composition and depth profiles. As seen in Fig. 8, there is a large decrease in the antimony

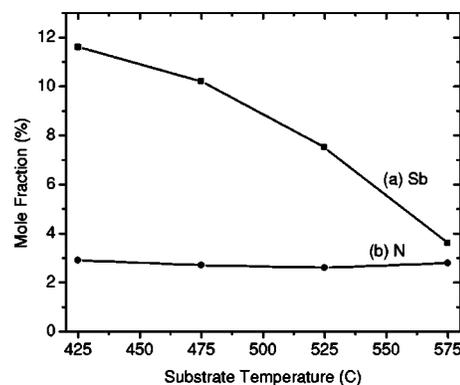


FIG. 8. SIMS results from the substrate temperature study. (a) Antimony composition and (b) nitrogen profile.

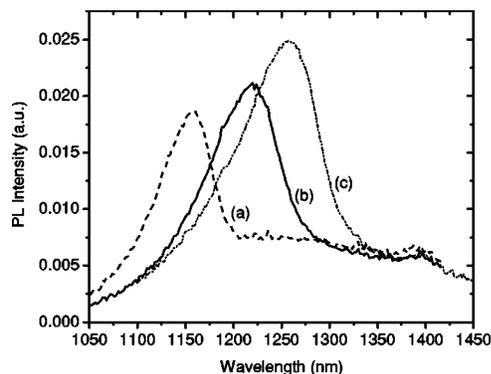


FIG. 9. PL spectra from the $\text{GaN}_{0.029}\text{As}_{0.873}\text{Sb}_{0.098}$ sample grown at different substrate temperatures. (a) +150 °C (575 °C), (b) +100 °C (525 °C), and (c) +50 °C (475 °C). The small peak at 1400 nm is the water present in the testing environment.

concentration with increasing substrate temperature, whereas the nitrogen concentration remained roughly constant. The loss of 8% Sb explains the large shift in the strain observed in the HRXRD peaks. Similar to the SIMS data from the arsenic overpressure study, the nitrogen composition remained constant, even though the antimony concentration changed. With this data, one might be encouraged by the fact that at high temperatures, the material obtained was coherent and had smaller amounts of antimony so that the band gap was larger. However, the PL results displayed in Fig. 9 were unexpected. With increasing substrate temperature, the PL spectra blueshifted, as expected, due to a lower antimony concentration, but also decreased in intensity. There is also a large shoulder to the PL spectra for all three samples, which was not found in the original substrate temperature sample. This longer wavelength shoulder could be a result of microsegregation, which could not be observed easily in HRXRD, but possibly in the RSM. Since there could be areas of clustering of increased nitrogen concentration within the QW, it could lead to areas of luminescence in which the band gap is smaller, leading to a luminescence of a longer wavelength. If these regions occurred only inside the QWs, they would have no effect on the interfaces and thus would not affect or reduce the diffraction thickness oscillations. As evidenced by the poor PL results, GaNAsSb cannot be grown at high temperatures without a large decrease in the optical material quality.

D. Discussion

In choosing a barrier material for a QW, it is desirable to have certain key properties which will enhance the structure without detrimental effects. The barriers must have a sufficient band offset and of the correct type. If the QW is strained, it would be important to strain compensate with the barriers of the opposite strain. This will prevent the relaxation and the formation of defects by effectively increasing the critical thickness of the layers. Finally, the material must be of a good quality. If it is not, the defects present will lead to a higher rate of nonradiative recombination, preventing the carriers from recombining radiatively in the QW. One of the unfortunate limitations of MBE is the limited number of sources, which constrains the choice in alloy compositions

for a barrier material with a defined QW composition. This would not be an issue if the fluxes could be accurately changed during the growth, but most MBE systems do not have this capability. GaNAs was the previous barrier material utilized in the GaInNAs QWs.^{5,14} When antimony was added to GaInNAs, antimony was also added to GaNAs, assuming that the expected surfactant nature of antimony and the improvement in the material to GaInNAs would also apply to GaNAs. However, the GaNAsSb material has several properties that do not make it the desired barrier material of choice. An examination of the RHEED patterns showed that the addition of antimony to GaNAs actually degraded the surface quality and turns the pattern from streaky to spotty. This suggests the surface roughens and there could have been some clustering on the surface that could possibly lead to defects. GaNAsSb does not provide any strain compensation to the highly compressive GaInNAs(Sb) QWs. Without the strain compensation, it is difficult to grow more than 1–2 QWs in a laser structure and thus would not be good for applications such as high-power laser diodes or vertical-cavity surface-emitting lasers (VCSELs). GaNAs, however, is tensile and does provide a strain compensation. The GaNAsSb barriers used for the GaInNAs(Sb) QWs also have a band-gap value which is very close to that of the QW. Having a small band offset in the valence and conduction bands would result in a poor hole and electron confinement and may even reduce the overlap between the two wave functions. This would lead to poor luminescence and also poor thermal characteristics in the device performance. It is also suspected that GaNAsSb has a band-gap region that is shifted toward the conduction-band edge, because antimony tends to increase the valence-band offsets while not affecting the conduction-band offset. This could lead to a type-II alignment with the QW. GaNAs has a larger band gap (~1.1–1.2 eV) and thus provides a better electron and hole confinement. Finally, the optical properties of a GaNAsSb have been found to be of a relatively poor quality. Although PL is not the absolute measure of the optical and electrical quality of a material, it does give a very good idea. From these findings, it is seen that GaNAs is the better barrier material compared to GaNAsSb for the GaInNAs(Sb) QWs.

The device structure for our long-wavelength GaInNAs(Sb) optoelectronic devices was changed to reflect the new knowledge that GaNAs was the preferred barrier. This change led to an improvement in the GaInNAsSb PL quality. For the GaInNAsSb QW samples at 1.55 μm , the full width at half maximum (FWHM) of the PL peak was 57 meV with the GaNAsSb barriers and 44 meV with the GaNAs barriers. The FWHM of the PL peak is a good measure of the quality of the material and structure. Once it was confirmed that the PL quality improved, the edge-emitting lasers and VCSELs were grown with the new device structure. 1.5 μm edge-emitting lasers with GaInNAsSb QWs and GaNAs barriers have been demonstrated.²¹ These devices are the lowest-threshold QW lasers in this wavelength range ever grown on GaAs and show an enormous improvement in the operational parameters. Figure 10 shows the dramatic improvement in our devices compared to other groups and previous studies. It was once thought that a “nitrogen penalty” had existed for

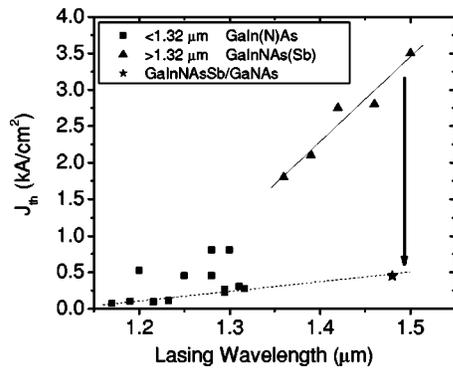


FIG. 10. Compilation of the long-wavelength edge-emitting GaAs lasers threshold current-lasing densities vs wavelength from literature. Utilizing the GaNAs barriers has allowed for the greatly reduced threshold current densities, as shown by the star data point.

devices past 1.31 μm , and the threshold current density inherently increased with more nitrogen. However, we demonstrated that by improving the device structure and material, the nitrogen penalty was not an actual impediment to good devices. VCSELs utilizing the GaNAs barriers have also been demonstrated.²² The previous VCSELs on GaAs have only been able to reach a maximum wavelength of 1.31 μm . The VCSELs grown by this group were found to operate at 1.46 μm and represent the longest wavelength VCSELs ever grown on GaAs.

IV. CONCLUSION

In conclusion, GaNAs(Sb)/GaAs QWs were grown on (100) GaAs substrates by MBE. The properties of GaNAsSb have been studied using RHEED, HRXRD, SIMS, and PL. RHEED showed streaky patterns for GaNAs, but turned spotty when antimony was added to form GaNAsSb. HRXRD showed the GaNAsSb has a lattice constant very similar to that of GaAs for the compositions relevant to the current long-wavelength optoelectronic devices. SIMS, with calibrations, gave the compositions of the GaNAsSb material and confirmed that antimony enhances the nitrogen incorporation. PL showed that the wavelength of GaNAsSb was similar to that of the QWs it was surrounding. Upon annealing the PL intensity increased and the wavelength blue-shifted, similar to that seen in GaInNAs(Sb) QWs. The luminescent properties of the GaNAsSb QWs were significantly lower than those of GaInNAs(Sb) QWs, indicating either a poor optical material quality and/or a poor carrier confinement. Changing the arsenic overpressure had no significant effect on GaNAsSb except for a change in the antimony concentration because arsenic and antimony compete with each other for the group-V sites. Increasing the arsenic flux decreased the antimony concentration. Although GaNAsSb can be grown coherently at higher temperatures (475–575 °C), the optical quality of the material was dras-

tically reduced. Increasing the substrate temperatures blue-shifted the PL peak wavelength (due to the decreased antimony incorporation), but also decreased the intensity. Finally, with these properties in hand, GaNAsSb was shown not to be a good barrier material for GaInNAs(Sb) QWs. The recent laser devices utilizing GaNAs rather than GaNAsSb QW barriers have shown remarkable advances.

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