Size, strain and band offset engineering in GaAs(Sb)(N)-capped InAs quantum dots for 1.3 – 1.55 μm emitters

J. M. Ulloa*[a], M. del Moral[a], M. Montes[a], M. Bozkurt[b], P. M. Koenraad[b], A. Guzmán[a], A. Hierro[a]

[a]Institute for Systems based on Optoelectronics and Microtechnology (ISOM), Universidad Politecnica de Madrid, Ciudad Universitaria s/n, 28040 Madrid, Spain;

[b]Photonics and Semiconductor Nanophysics, Department of Applied Physics, Eindhoven University of Technology, P. O. Box 513, NL-5600 MB Eindhoven, The Netherlands

ABSTRACT

The optical and structural properties of InAs/GaAs quantum dots (QD) are strongly modified through the use of a thin (~ 5 nm) GaAsSb(N) capping layer. In the case of GaAsSb-capped QDs, cross-sectional scanning tunnelling microscopy measurements show that the QD height can be controllably tuned through the Sb content up to ~ 14 % Sb. The increased QD height (together with the reduced strain) gives rise to a strong red shift and a large enhancement of the photoluminescence (PL) characteristics. This is due to improved carrier confinement and reduced sensitivity of the excitonic bandgap to QD size fluctuations within the ensemble. Moreover, the PL degradation with temperature is strongly reduced in the presence of Sb. Despite this, emission in the 1.5 μm region with these structures is only achieved for high Sb contents and a type-II band alignment that degrades the PL. Adding small amounts of N to the GaAsSb capping layer allows to progressively reduce the QD-barrier conduction band offset. This different strategy to red shift the PL allows reaching 1.5 μm with moderate Sb contents, keeping therefore a type-I alignment. Nevertheless, the PL emission is progressively degraded when the N content in the capping layer is increased.

Keywords: Quantum dot, photoluminescence, GaAsSb, GaAsSbN, diluted nitrides

1. INTRODUCTION

Since the use of Sb to cover InAs/GaAs quantum dots (QDs) was first reported a few years ago,[1-3] significant activity has been realized in this field, which is becoming increasingly attractive.[4-20] A thin (~ 5 nm) GaAsSb capping layer allows extending the emission wavelength of InAs QDs grown on GaAs substrates to the 1.55 μm region.[1-5] The strong observed red shift has been typically attributed to a type-II band alignment for high Sb contents, with the hole wavefunction being localized out of the QD in the GaAsSb capping layer.[3,4,10,13-15] Nevertheless, apart from a few studies dedicated to analyze the emission from type-II samples,[10-15] not much attention has been paid to the evolution of the optical properties of GaAsSb-capped InAs/GaAs QDs with the amount of Sb in the capping layer. Moreover, the effect that different Sb contents could have in the structural properties of the QDs is still unknown. The fact that GaAsSb acts as a strain reducing layer for InAs/GaAs QDs, together with the surfactant effect of Sb, could lead to an altered capping process that could modify the final size and/or shape of the QDs. It is well known that strong QD size and shape changes take usually place during the capping process, and that these changes are dependent on the capping material used.[21,22] These structural changes are of crucial relevance because they will strongly affect the optical properties of the QD system. In particular, they should significantly affect the carrier escape mechanisms and the temperature evolution of the PL emission in these samples. Although a different PL degradation between 14 K and room temperature for GaAs and GaSb-capped samples has been reported,[7] there are no similar reports on GaAsSb-capped QDs and the influence of different Sb contents has not been reported.

One drawback of the Sb-based approach is that optical emission at long wavelengths (in the 1.4 – 1.6 μm region) can only be achieved in GaAsSb-capped InAs/GaAs QDs with a type-II band alignment that degrades the PL. Nevertheless, introducing small amounts of N in the capping layer would strongly reduce the QD-capping layer conduction band offset, inducing an extra red shift. The addition of Sb to GaAs affects mainly the valence band of GaAs,[23] while N affects only the conduction band,[24] which means that a capping layer made from the quaternary GaAsSbN would allow to

Quantum Dots and Nanostructures: Synthesis, Characterization, and Modeling VIII, edited by Kurt G. Eyink, Frank Szmulowicz, Diana L. Huffaker, Proc. of SPIE Vol. 7947, 79470T · CCC code: 0277-786X/11/$18 · doi: 10.1117/12.875014

independently tune the QD-barrier valence and conduction band offsets. This could allow reaching 1.55 μm while keeping a type-I band alignment.

In this work, we have first used photoluminescence (PL) and Cross-Sectional Scanning Tunneling Microscopy (X-STM) to correlate the Sb-induced changes in the optical properties of the QDs with structural changes. We show that the QD height can be controllably tuned through the Sb content and that taller QDs show a reduced degradation of the PL emission with temperature. The addition of small amounts of N to the thin GaAsSb capping layer allows to strongly reduce the conduction band offset, allowing reaching 1.55 μm with a type-I band alignment. Nevertheless, the PL spectra are significantly degraded when the amount of N is increased.

*jmulloa@die.upm.es; phone +34 915495700 x4415; fax +34-913367323

2. EXPERIMENTAL DETAILS

The samples were grown by solid source molecular beam epitaxy (MBE) on n+ Si doped (100) GaAs substrates. First, a series of eleven samples containing a single QD layer capped with GaAsSb was grown for PL studies. In all of these samples, 2.7 monolayers (ML) of InAs were deposited at 450 °C and 0.04 ML/s on an intrinsic GaAs buffer layer, and capped with a nominally 4.5 nm-thick $GaAs_{1-x}Sb_x$ layer grown at 470 °C. The Sb content was nominally changed from 0 to 25 %. 250 nm of GaAs were finally grown on top of the GaAsSb capping. A layer of similar uncapped QDs was also grown on the surface of every sample for atomic force microscopy measurements (AFM). Four of the GaAsSb-capped QD layers with different Sb contents were reproduced in a single sample (separated from each other by 50 nm of GaAs) for X-STM measurements. Another series of 10 samples with a GaAsSbN capping layer was grown under the same conditions than the previous. In this case, the Sb content in the capping layer was fixed to ~ 13 % and the N content was progressively increased.

The PL was measured from 15 K to room temperature using a closed cycle Helium cryostat with a temperature controller and a He-Ne laser as the excitation source. The emitted light was dispersed through a 1m spectrometer and detected with a liquid nitrogen cooled Ge detector using standard lock-in techniques. The X-STM measurements were performed on a (1 1 0) surface plane of in-situ cleaved samples under UHV ($p < 4\times10^{-11}$ Torr) conditions, at room temperature. Polycrystalline tungsten tips prepared by electrochemical etching were used. The images were obtained in constant current mode at high negative voltages (~3 V).

3. TUNNING THE QD HEIGHT THROUGH THE Sb CONTENT

We investigate first the effect of a thin GaAsSb capping layer with different Sb contents on both the structural and optical properties of the QDs, trying to correlate the Sb-induced changes in the PL with structural changes at the atomic scale.

3.1 Effect of the Sb content on the structural properties

The sample with four QD layers (from now on LA, LB, LC and LD) was analyzed by X-STM. The Sb content in the capping layer was different in each layer, starting from 0 % in LA (reference GaAs-capped QDs). From high negative voltage (-3 V) X-STM images, the Sb content in the capping layer can be deduced by analyzing the outward relaxation of the cleaved surface.[25,26] By comparing the measured outward relaxation to calculations from continuum elasticity theory we obtain an Sb content of 0, 7, 11 and 22 % for LA, LB, LC and LD, respectively.[18]

High resolution images of a QD in LA, LB and LD can be seen in Fig. 1 (a), (b) and (c), respectively. The measurement conditions (negative voltage) allow imaging group V elements so that the bright spots in images (b) and (c) represent individual Sb atoms in the As matrix (due to the different size and bonding configuration, Sb atoms appear brighter than As atoms). The bright spots in image (a) represent individual In atoms, which are visible through their distortion of the surrounding As atoms. The GaAsSb layer appears clearly as a distinct layer, which is hardly intermixed with the QDs and the WL. The interface between the QD and the capping layer is well defined, especially in the lower Sb content layers LB and LC (see LB in Fig. 1 (b)). This suggests that there is no Sb incorporation into the QDs. Further support for this conclusion can be obtained by applying a local mean equalization filter to the images (which eliminates the background contrast due to strain relaxation and enhances the contrast of individual Sb atoms). The average contrast between adjacent atoms in filtered images is the same in GaAsSb-capped QDs than in the QDs of the reference layer. It can therefore be concluded that there is no Sb inside the QDs, contrary to what has been observed in InAs/GaAs QDs capped directly with 2.2 ML of GaSb,[9] or exposed to a Sb flux immediately before capping with GaAs.[20]

Figure 1. Filled states topography images of a QD in (a) LA, (b) LB and (c) LD. In image (a) the bright spots correspond to In atoms. In images (b) and (c) the bright spots correspond to Sb atoms in the As matrix.

Significant differences between the QDs in the different layers are already clear from these images. The size and shape of the QDs in each layer can be deduced by a statistical analysis of the relationship between the height and the base length in a large number of QDs. Considering the high uniformity observed by AFM (not shown), it is assumed for the analysis that all the dots in each layer are identical. After analyzing ~ 15 QDs in each layer, a linear dispersion, similar to that obtained in ref. 27, is found, indicating that the QDs have an ellipsoidal or lens shape (as also apparent from Fig. 1) with a base diameter of 24 ± 1 nm. This correlates very well with the 26 ± 2 nm diameter measured by AFM in similar surface QDs.

However, the capped QD height increases with the Sb content, as shown in Fig. 2, in which the QD height normalized to the height of surface uncapped QDs is plotted as a function of the Sb content in the capping layer. The height differences must be originated during the capping process: the strong QD decomposition that takes place during capping with GaAs[21,22] is reduced by the presence of Sb. Moreover, the reduction is proportional to the amount of Sb in the capping layer. The result is that the QD height is progressively increased when the amount of Sb in the capping layer increases. The dissolution process is found to be completely suppressed for an Sb content of 22 % (the QD height measured by X-STM is the same that the one measured by AFM in uncapped QDs) but the linear regression in the figure shows that QD dissolution stops completely for an smaller Sb content of ~14%. For higher Sb contents, no further change in the QD height is expected. It is possible, therefore, to controllably tune the height of the InAs QDs by changing the Sb content in the capping layer between 0 and ~14 %.

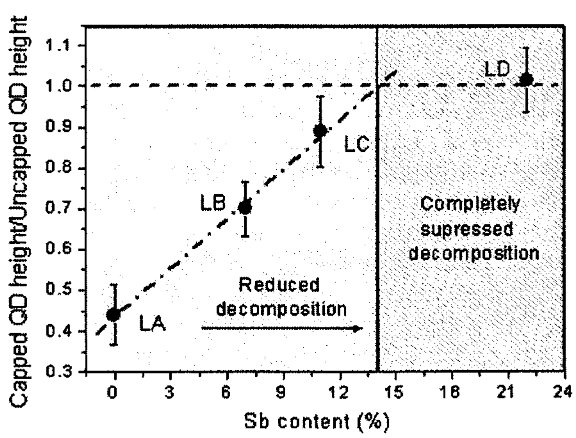

Figure 2. QD height normalized to the height of the equivalent uncapped QDs as a function of the Sb content in the four different layers studied by X-STM. A value of 1.0 indicates a completely suppressed decomposition process. The red dash-dot line is a linear fit to the values of LA, LB and LC. The two different background colors indicate the two different regimes.

3.2 Effect of the Sb content on the room temperature PL

The results in section 3.1 show that the PL peak red shift typically observed when adding Sb to the capping layer is not only due to the reduced strain or the transition to a type-II band alignment, but also to the fact that the QD height increases initially with the Sb content. Fig. 3 (a) shows the normalized room temperature PL spectra of some of the samples in the series with increasing amount of Sb in the capping layer. As already reported,[1-5] by gradually increasing the Sb content, the emission wavelength of the QDs can be red-shifted, reaching almost 1.5 μm. With this QD system, long wavelengths close to 1.55 μm are up to now only achievable in samples with a type-II band alignment. From PL measurements as a function of excitation power (not shown), we know that the transition from a type-I to a type-II band alignment happens somewhere between the samples labeled as I and II in Fig. 3 (a).

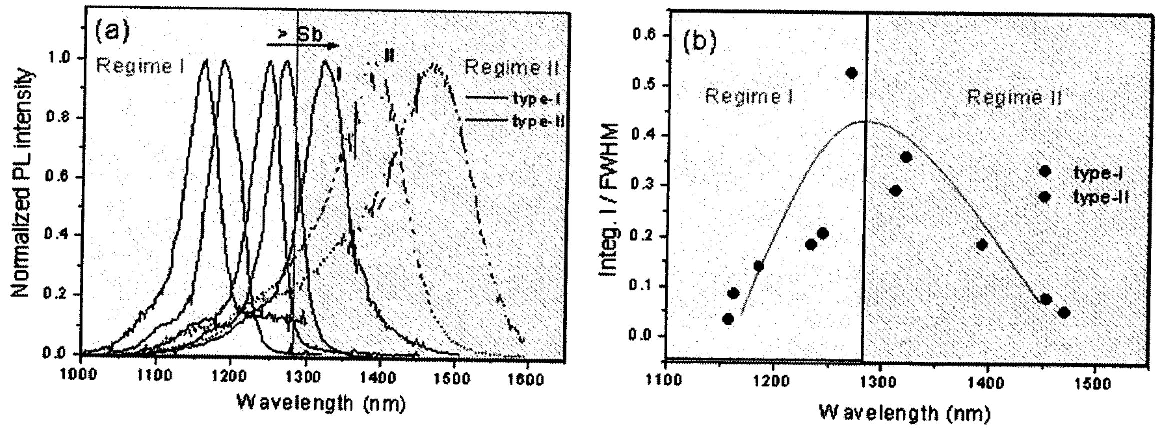

Figure 3. (a) Normalized room temperature PL spectra from a selection of the QD layers with increasing amount of Sb in the capping layer. (b) Dependence of the ratio between the PL integrated intensity and FWHM on the peak wavelength. Lines are guides to the eye. The two different background colors indicate the two different optical regimes.

However, the observed red shift develops within two clearly different optical regimes regarding the PL properties (regimes I and II) which are not determined by this band alignment transition. As a figure of merit of the PL quality we plot in Fig. 3 (b) the ratio between the PL integrated intensity and the full width at half maximum (FWHM) as a function of the peak wavelength. It can be seen that for low Sb contents, the PL emission is progressively improved (reduced FWHM and increased integrated intensity), reaching its optimum at a peak wavelength of ~ 1280 nm. Indeed, the presence of the first regime allows us to obtain PL emission at 1.3 μm with significantly improved optical properties compared to the shorter wavelength PL of the reference GaAs-capped InAs/GaAs QDs. We attribute this improvement to the observed gradual increase in QD height, which improves carrier confinement and reduces the sensitivity of the effective bandgap to QD size fluctuations within the ensemble. In regime II, at higher wavelengths (higher Sb contents), the PL is gradually degraded with increasing Sb, becoming very broad and less intense for the longest wavelengths. This degradation at high Sb contents occurs as a result of composition modulation in the capping layer and strain-induced Sb migration to the top of the QDs, together with the transition to a type-II band alignment.[18] In this case, the hole wavefunction is confined in the capping layer on top of the QDs and, therefore, different Sb contents on top of the QDs will have a very strong impact on the hole energy levels, giving rise to a broad PL spectrum.

3.3 Effect of the Sb content on the temperature evolution of PL

The complex structural evolution, with the initial increase of QD height up to ~ 12-14 % Sb and the following enhancement of composition modulation at higher Sb contents, together with the change in the valence band alignment, should strongly affect the temperature behavior of the PL emission. In order to investigate this effect, the PL spectrum as a function of temperature was measured for several samples in the series between 15 K and room temperature. An Arrhenius plot of the PL integrated intensity as a function of the inverse temperature for the samples with 0 and 12 % Sb content can be observed in the inset of Fig. 4. There is an anomalous increase in the integrated intensity with temperature in the range of 35 – 90 K and 35 – 130 K in the samples with 0 and 12 % Sb, respectively. One possible explanation would be carrier redistribution from smaller QDs to bigger QDs within the ensemble, in which exciton recombination is more efficient due to increased confinement.[28] Nevertheless, this would typically give rise to a reduced emission energy in that temperature range together with a reduction of the FWHM,[29] something that we do not observe (not shown). Moreover, the high QD size homogeneity seen in AFM measurements (see ref. 18) also suggests that redistribution effects should not be strong. Other possible reasons could be carrier escape from traps in the wetting layer (WL),[30] or the presence of strain-induced potential barriers at the QD interface.[31]

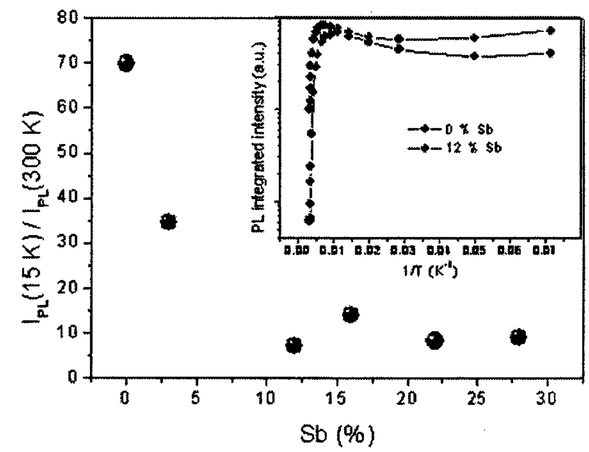

Figure 4. Ratio between the PL integrated intensity at 15 K and 300 K as a function of the Sb content. The inset shows an Arrhenius plot of the PL integrated intensity versus the inverse of the temperature for samples with 0 and 12 % Sb in the capping layer.

The quenching of the PL intensity with temperature is significantly smaller in the Sb-containing sample. Indeed, the integrated intensity decreases a factor of 70 between 15 K and room temperature in the reference sample, while the decrease factor is only 7 in the sample with Sb, so PL degradation with temperature is reduced by one order of magnitude. If the ratio between the integrated intensity at 15 K and room temperature is plotted as a function of the Sb content for all the analyzed samples, it is clear that there is a very strong initial reduction of the PL quenching with temperature when the amount of Sb in the capping layer increases (Fig. 4). For Sb contents above ~ 12 % Sb, the magnitude of the quenching remains approximately constant. The improvement takes place in the first optical regime (Regime I), and could therefore be related to the increased QD height, which increases carrier confinement and reduces carrier escape from the QDs. The analysis of the thermal activation energies together with structural data from X-STM measurements (not shown), allow us to determine that, in the absence of Sb, escape of both electrons and holes to the GaAs barriers is the main PL quenching mechanism.[32] Nevertheless, when adding Sb to the structure, electron escape is suppressed due to increased QD height. At moderate Sb contents holes escape from the QD to the thin GaAsSb capping layer, where redistribution and retrapping processes between QDs can take place. However, at high Sb contents where type-II band alignments are obtained, X-STM analysis shows strain-induced Sb-rich clusters on top of the QDs. In this case, hole escape from the Sb-rich clusters to the extended GaAsSb capping is identified as the dominant quenching mechanism.[32] The reduced PL quenching with temperature in the presence of Sb represents an important advantage regarding the applications since it would improve the temperature performance for long wavelength devices. The possibility of increasing the QD height through the Sb content is therefore very interesting regarding device applications.

4. TUNNING THE CONDUCTION BAND OFFSET THROUGH THE N CONTENT

In the second series of samples, a 5 nm thick GaAsSbN capping layer was grown under the same conditions than the previous GaAsSb layer. In this case, the Sb content was fixed to ~ 13 % trying to keep the type-I band alignment and the N content was progressively increased in order to extend the emission wavelength by progressively reducing the QD-capping layer conduction band offset. The normalized low temperature PL spectra of these samples are shown in Fig. 5 (a). From the point of view of the strain, the presence of N counteracts the strain reducing effect of Sb, and no PL red shift is expected in this sense. Therefore, the observed strong red shift must be mainly due to the progressive reduction of the conduction band offset. Emission at 1.45 μm is obtained at 15 K; unfortunately no room temperature emission was observed from any of the N-containing samples. This can be related to the PL degradation observed when the N content is increased: as shown in Fig. 5 (b) the integrated intensity decreases and the FWHM increases with the amount of N in the capping layer. Indeed, an initial strong decrease of the PL integrated intensity of order of magnitude is obtained when the PL peak is red-shifted by only 70 nm. The reasons for this degradation are being investigated. Possible reasons could be the presence of N-related defects acting as non-radiative recombination centers, the reduced electron confinement due to small conduction band offset, the presence of strong composition modulation in the capping layer etc.

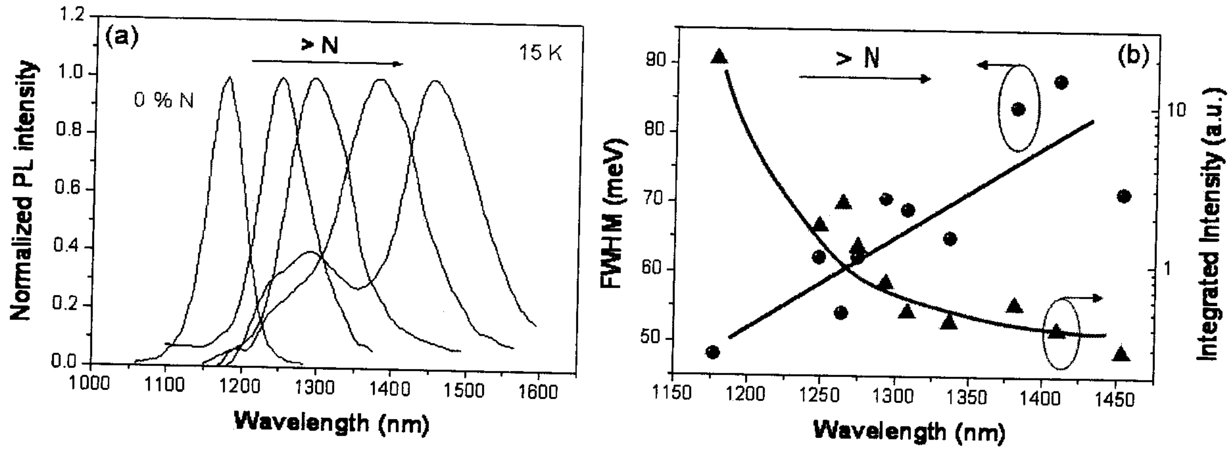

Figure 5. (a) Normalized room temperature PL spectra from some of the GaAsSbN-capped QD layers with increasing amount of N in the capping layer. (b) Dependence of the FWHM and integrated PL intensity on the peak wavelength. The blue signs correspond to the 0% N case. Lines are guides to the eye.

From PL measurements as a function of the excitation power (not shown), we see that all the samples in the series have a type-I band alignment. Therefore, emission in the 1.5 μm region can be obtained with type-I samples by using a GaAsSbN capping layer. Despite this, the room temperature performance of log wavelength type-I GaAsSbN-capped QDs is much poorer than that of the type-II GaAsSb-capped QDs. Further optimization of the growth parameters of the thin GaAsSbN layer could help to solve this problem in the future.

5. CONCLUSIONS

In summary, we have used X-STM and PL measurements to show how the optical and structural properties of InAs/GaAs quantum dots (QD) can be strongly modified through the use of a thin (~ 5 nm) GaAsSb(N) capping layer. The QD height can be controllably tuned through the Sb content in the case of GaAsSb-capped QDs. In addition to a strong red shift, the increased QD height (together with the reduced strain) induces a large enhancement of the PL characteristics. This is due to improved carrier confinement and reduced the sensitivity of the excitonic bandgap to QD size fluctuations within the ensemble. Moreover, the PL degradation with temperature is strongly reduced in the presence of Sb. With GaAsSb-capped QDs, emission in the 1.5 µm region is only achieved for high Sb contents and a type-II band alignment that degrades the PL. Nevertheless, adding small amounts of N to the GaAsSb capping layer allows to progressively reduce the QD-barrier conduction band offset. This approach allows reaching 1.5 µm with moderate Sb contents, keeping therefore a type-I alignment. Nevertheless, the PL emission is progressively degraded when the N content in the capping layer is increased. The reason for this degradation is currently being investigated.

6. ACKNOWLEDGMENTS

This work has been supported by the Comunidad de Madrid through the projects NANOCOMIC (S0505IESP-0200) and "Q&C Light" (P2009/ESP-1503), by the European Union through the SANDiE Network of Excellence (contract No. NMP4-CT-2004-500101), COST Action MP0805 and by STW-VICI Grant No. 6631.

REFERENCES

[1] K. Akahane, N. Yamamoto, and N. Ohtani,"Strong photoluminescence and laser operation of InAs quantum dots covered by a GaAsSb strain-reducing layer," Physica E (Amsterdam) 21, 295 (2004).

[2] H. Y. Liu, M. J. Steer, T. J. Badcock, D. J. Mowbray, M. S. Skolnick, P. Navaretti, K. M. Groom, M. Hopkinson, and R. A. Hogg, "Long-wavelength light emission and lasing from InAs/GaAs quantum dots covered by a GaAsSb strain-reducing layer," Appl. Phys. Lett. 86, 143108 (2005).

[3] J. M. Ripalda, D. Granados, Y. González, A. M. Sánchez, S. I. Molina, and J.M.García, "Room temperature emission at 1.6µm from InGaAs quantum dots capped with GaAsSb," Appl. Phys. Lett. 87, 202108 (2005).

[4] H. Y. Liu, M. J. Steer, T. J. Badcock, D. J. Mowbray, M. S. Skolnick, F. Suarez, J. S. Ng, M. Hopkinson, and J. P. R. David, "Room-temperature 1.6µm light emission from InAs/GaAs quantum dots with a thin GaAsSb cap layer," J. Appl. Phys. 99, 046104 (2006).

[5] T. Matsuura, T. Miyamoto, M. Ohta, and F. Koyama, "Photoluminescence characterization of (Ga)InAs quantum dots with GaInAsSb cover layer grown by MBE," Phys. Stat. Sol. (c) 3, 516 (2006).

[6] D. Guimard, S. Tsukamoto, M. Nishioka, and Y. Arakawa, "1.55µm emission from InAs/GaAs quantum dots grown by metal organic chemical vapor deposition via antimony incorporation," Appl. Phys. Lett. 89, 083116 (2006).

[7] J. M. Ripalda, D. Alonso-Álvarez, B. Alén, A. G. Taboada, J. M. García, Y. González, L. González, "Enhancement of the room temperature luminescence of InAs quantum dots by GaSb capping," Appl. Phys. Lett. 91, 012111 (2007).

[8] J. M. Ulloa, I. W. D. Drouzas, P. M. Koenraad, D.J. Mowbray, M.J. Steer, H. Y. Liu, and M. Hopkinson, "Suppression of InAs/GaAs quantum dot decomposition by the incorporation of a GaAsSb capping layer," Appl. Phys. Lett. 90, 213105 (2007).

[9] S. I. Molina, A M Sánchez, A M Beltrán, D L Sales, T Ben, M F Chisholm, M Varela, S J Pennycook, P L Galindo, A. J. Papworth, P. J. Goodhew, J. M. Ripalda, "Incorporation of Sb in InAs/GaAs quantum dots," Appl. Phys. Lett. 91, 263105 (2007).

[10] T. T. Chen, C. L. Cheng, Y. F. Chen, F. Y. Chang, H. H. Lin, C.-T. Wu, and C.-H. Chen, "Unusual optical properties of type-II InAs/$GaAs_{0.7}Sb_{0.3}$ quantum dots by photoluminescence studies," Phys. Rev. B 75, 033310 (2007).

[11] C. Y. Jin, H. Y. Liu, S. Y. Zhang, Q. Jiang, S. L. Liew, M. Hopkinson, T. J. Badcock, E. Nabavi, and D. J. Mowbray, "Optical transitions in type-II InAs/GaAs quantum dots covered by a GaAsSb strain-reducing layer," Appl. Phys. Lett. 91, 021102 (2007).

[12] J. M. Ulloa, C. Çelebi, P. M. Koenraad, A. Simon, E. Gapihan, A. Letoublon, N. Bertru, J. Drouzas, D. J. Mowbray, M. J. Steer, and M. Hopkinson, "Atomic scale study of the impact of the strain and composition of the capping layer on the formation of InAs quantum dots," J. Appl. Phys. 101, 081707 (2007).

[13] Y. D. Jang, T. J. Badcock, D. J. Mowbray, M. S. Skolnick, J. Park, D. Lee, H. Y. Liu, M. J. Steer, and M. Hopkinson, "Carrier lifetimes in type-II InAs quantum dots capped with a GaAsSb strain reducing layer," Appl. Phys. Lett. 92, 251905 (2008).

[14] Wen-Hao Chang, Yu-An Liao, Wei-Ting Hsu, Ming-Chih Lee, Pei-Chin Chiu, and Jen-Inn Chyi, "Carrier dynamics of type-II InAs/GaAs quantum dots covered by a thin $GaAs_{1-x}Sb_x$ layer," Appl. Phys. Lett. 93, 033107 (2008).

[15] Yu-An Liao, Wei-Ting Hsu, Pei-Chin Chiu, Jen-Inn Chyi, and Wen-Hao Chang, "Effects of thermal annealing on the emission properties of type-II InAs/GaAsSb quantum dots," Appl. Phys. Lett. 94, 053101 (2009).

[16] V. Haxha, I. Drouzas, J. M. Ulloa, M. Bozkurt, P. M. Koenraad, D. J. Mowbray, H. Y. Liu, M. J. Steer, M. Hopkinson and M. A. Migliorato, "Role of segregation in InAs/GaAs quantum dot structures capped with a GaAsSb strain-reduction layer," Phys. Rev. B 80, 165334 (2009).

[17] D. Guimard, M. Ishida, L. Li, M. Nishioka, Y. Tanaka, H. Sudo, T. Yamamoto, H. Kondo, M. Sugawara, and Y. Arakawa, "Interface properties of InAs quantum dots produced by antimony surfactant-mediated growth: etching of segregated antimony and its impact on the photoluminescence and lasing characteristics," Appl. Phys. Lett. 94, 103116 (2009).

[18] J. M. Ulloa, R. Gargallo-Caballero, M. Bozkurt, M. del Moral, A. Guzmán, P. M. Koenraad, and A. Hierro, "GaAsSb-capped InAs quantum dots: From enlarged quantum dot height to alloy fluctuations," Phys. Rev. B 81, 165305 (2010).

[19] J. M. Ulloa, P. M. Koenraad, M. Bonnet-Eymard, A. Létoublon, and N. Bertru, "Effect of a lattice-matched GaAsSb capping layer on the structural properties of InAs/InGaAs/InP quantum dots," J. Appl. Phys. 107, 074309 (2010).

[20] A. G. Taboada, A. M. Sánchez, A. M. Beltrán, M. Bozkurt, D. Alonso-Álvarez, B. Alén, A. Rivera, J. M. Ripalda, J. M. Llorens, J. Martín-Sánchez, Y. González, J. M. Ulloa, J. M. García, S. I. Molina, and P. M. Koenraad, "Structural and optical changes induced by incorporation of antimony into InAs/GaAs(001) quantum dots," Phys. Rev. B 82, 235316 (2010).

[21] J. M. Garcia, G. Medeiros-Ribeiro, K. Schmidt, T. Ngo, J. L. Feng, A. Lorke, J. Kotthaus, P. M. Petroff, "Intermixing and shape changes during the formation of InAs self-assembled quantum dots," Appl. Phys. Lett. 71, 2014 (1997).

[22] P.D. Siverns, S. Malik, G. McPherson, D. Childs, C. Roberts, R. Murray, B.A. Joyce and H. Davock, "Scanning transmission-electron microscopy study of InAs/GaAs quantum dots," Phys. Rev. B. 58, R10127 (1998).

[23] R. Teissier, D. Sicault, J. C. Harmand, G. Ungaro, G. Le Roux, L. Largeau, "Temperature-dependent valence band offset and band-gap energies of pseudomorphic GaAsSb on GaAs," J. Appl. Phys. 89, 5473 (2001).

[24] W. Shan, W. Walukiewicz, J.W. Ager III, E.E. Haller, J.F. Geisz, D.J. Friedman, J.M. Olson, S.R. Kurtz, "Band Anticrossing in GaInNAs Alloys," Phys. Rev. Lett. 82, 1221 (1999).

[25] R. M. Feenstra, "Comparison of electronic and mechanical contrast in scanning tunneling microscopy images of semiconductor heterojunctions," Physica B 273, 796 (1999).

[26] D. M. Bruls, J. W. A. M. Vugs, P. M. Koenraad, H. W. M. Salemink, J. H. Wolter, M. Hopkinson, M. S. Skolnick, F. Long, and S. P. A. Gill, "Determination of the shape and indium distribution of low-growth-rate InAs quantum dots by cross-sectional scanning tunneling microscopy," Appl. Phys. Lett. 81, 1708 (2002).

[27] J. H. Blokland, M. Bozkurt, J. M. Ulloa, D. Reuter, A. D. Wieck, P. M. Koenraad, P. C. M. Christianen and J. C. Maan, "Ellipsoidal InAs quantum dots observed by cross-sectional scanning tunneling microscopy," Appl. Phys Lett. 94, 023107 (2009).

[28] Jeppe Johansen, Søren Stobbe, Ivan S. Nikolaev, Toke Lund-Hansen, Philip T. Kristensen, Jørn M. Hvam, Willem L. Vos, and Peter Lodahl, "Size dependence of the wavefunction of self-assembled InAs quantum dots from time-resolved optical measurements," Phys. Rev. B 77, 073303 (2008).

[29] T. Mano, R. Notzel, Q. Gong, T. van Lippen, G. J. Hamhuis, T. J. Eijkemans, and J. H. Wolter, "Temperature-Dependent Photoluminescence of Self-Assembled (In,Ga)As Quantum Dots on GaAs (100): Carrier Redistribution through Low-Energy Continuous States," Jpn. J. Appl. Phys. 44, 6829 (2005).

[30] G. Saint-Girons, A. Lemaitre, V. Navarro-Paredes, G. Patriarche, E. V. K. Rao, I. Sagnes, and B. Theys, "Photoluminescence probing of non-radiative channels in hydrogenated In(Ga)As/GaAs quantum dots," J. Cryst. Growth 264, 334 (2004).

[31] L. He, G. Bester, and A. Zunger, "Compressive Strain-induced interfacial hole localization in self-assembled quantum dots: InAs/GaAs versus tensile InAs/InSb," Phys. Rev. B 70, 235316 (2004).