Faculty and Staff Publication:

**Shailesh Madiseti, Vadim Tokranov, Andrew Greene,
Michael Yakimov, Makoto Hirayama,
& Serge Oktyabrsky**

Citation for Final Published Work:

Madiseti, S., Tokranov, V., Greene, A., Yakimov, M., Hirayama, M., Oktyabrsky S., . . . Jacob, A. P. (2014). Growth of strained InGaSb quantum wells for p-FET on Si: Defects, interfaces, and electrical properties. *Journal of Vacuum Science & Technology B, Nanotechnology and Microelectronics: Materials, Processing, Measurement, and Phenomena*, 32(5).
doi:10.1116/1.4892797

<http://avs.scitation.org/doi/full/10.1116/1.4892797>

Growth of strained InGaSb quantum wells for p-FET on Si: Defects, interfaces, and electrical properties

Shailesh Madiseti, Vadim Tokranov, Andrew Greene, Michael Yakimov, Makoto Hirayama, and Serge Oktyabrsky^{a)}

SUNY College of Nanoscale Science and Engineering, Albany, New York 12203

Steven Bentley and Ajey P. Jacob

GLOBALFOUNDRIES at Albany NanoTech, Albany, New York 12203

(Received 6 May 2014; accepted 29 July 2014; published 13 August 2014)

A study of heteroepitaxial molecular beam epitaxy growth of strained p-channel InGaSb quantum well (QW) on lattice mismatched Si (100) using Al(Ga)Sb metamorphic buffers is presented in this paper. The migration enhanced epitaxy (MEE) technique was employed for AlSb nucleation layer (NL) on Si and analyzed using atomic force microscopy and *in-situ* Auger electron spectroscopy techniques to optimize growth conditions for continuous 2D buffer layers and improve surface quality of subsequent layers. Growth-related defects (threading dislocations, microtwins, and anti-phase boundaries) and their effect on surface morphology and electrical properties of the QWs are analyzed with scanning electron microscope and transmission electron microscopy and correlated to the NL properties. The baseline data for defect density in the layers and resultant surface morphology are presented. Room temperature p-channel Hall mobility of $660 \text{ cm}^2/\text{Vs}$ at $3 \times 10^{11} \text{ cm}^{-2}$ sheet hole concentration is achieved in InGaSb QWs using an optimized 15 monolayer AlSb MEE NL at 300°C growth temperature. © 2014 American Vacuum Society.

[<http://dx.doi.org/10.1116/1.4892797>]

I. INTRODUCTION

Integration of III–V on a Si platform is of considerable interest to the electronics industry to make both high electron and high hole mobility metal–oxide–semiconductor field effect transistors (MOSFET) commercially viable as Si offers low cost and high quality substrate. However, III–V growth on Si poses a number of challenges due to large lattice mismatch and thermal mismatch, which result in growth-related defects such as threading dislocations (TDs) and planar defects [microtwins (MTs) and stacking faults]. Additionally, growth transition from nonpolar to polar lattice results in the formation of antiphase domain boundaries (APB). Most of these defects hinder the progress of these materials since they can be electrically active, degrade transport properties, and lead to degraded surface morphology. Substantial work has been done on growth of GaAs and InGaAs on Si over the last two decades,^{1–4} and comparable progress of III–Sb on Si is also required for p-type MOSFET for complementary metal oxide semiconductor (CMOS) applications. Earlier research on the growth of GaAs on Si reported significant progress on substrate preparation,⁵ optimization of AlAs and AlGaAs buffers,⁶ thermal annealing methods to reduce defect density,⁷ use of tilted/misoriented substrates,⁸ or the use of SrTiO₃ buffer layers.² Other groups employed metal–organic chemical vapor deposition (MOCVD) using composite buffers such as SiGe (Ref. 3) or superlattices as intermediate layers.⁹ MOCVD epitaxy of GaAs requires gallium and hydride (AsH₃) precursors, whereas III–Sb MOCVD is difficult due to the instability of

SbH₃ molecule at as low as 100°C ; hence, the development of alternative Sb precursors will be required.^{10,11}

MBE growth of In_xGa_{1–x}Sb on Si is a relatively new field when compared to III–As on Si. Recent progress in III–Sb on Si MBE has shown the benefits of an AlSb nucleation layer (NL) that limits surface migration in assisting high quality 2D growth of GaSb,^{12–14} defect reduction using InSb quantum dot multilayers,¹⁵ GaSb/AlSb,¹⁶ or InGaSb (Ref. 17) quantum wells (QWs).

However, these papers have not reported on growth-related defects, defect-related surface morphology, and subsequent electrical properties of these films. The work described in this paper provides baseline data for growth of AlSb nucleation layers and defect density, resultant surface morphology and transport properties of strained buried p-type In_{0.36}Ga_{0.64}Sb QW structures grown on Si for CMOS applications.

II. EXPERIMENT

Structures were grown using EPI GEN II MBE system with group-III SUMO[®] cells and Sb and As valved cracker sources equipped with *in-situ* Auger electron spectroscopy (AES). Along with Si, silicon-on-insulator (SOI) substrates (fully depleted, 20 nm thick Si as measured) were used for growth to avoid leakages through Si substrate. Both Si and SOI wafers had exact (001) $\pm 1^\circ$ orientation standard for electronic industry.

After dipping into hydrofluoric acid solution, a substrate was immediately loaded into the vacuum chamber and subjected to consecutive UHV thermal treatment steps at $280^\circ\text{C}/500^\circ\text{C}$ for 10 h in the buffer chamber before transferring it into the growth chamber for final anneal at 850°C for

^{a)}Electronic mail: soktyabrsky@albany.edu

1 h to remove traces of carbon and oxides on the surface, which were monitored by *in-situ* AES. The removal of hydrogen and oxygen from the surface was also confirmed by reflection high-energy electron diffraction (RHEED) 2×2 patterns of surface reconstruction after which, with no other exposures, the substrate was further cooled down to NL growth temperature (250–535 °C) and soaked under Sb flux for 5 min to reduce antiphase domain boundaries.¹⁵ The substrate temperature was calibrated directly in the range of 400–450 °C using (1×3) to (4×4) RHEED pattern transition of AlSb surface along $[110]$ azimuth¹⁸ (fully relaxed 1 μm -thick AlSb on Si-substrate) and further interpolated using a substrate thermocouple temperature measurements.

The procedure for III/V flux ratio calibration was similar to that described in Ref. 18. The group III-limited growth rate was measured using RHEED oscillations. Then, with the fixed Al flux, the Sb flux was reduced to obtain Sb-limited growth as indicated by suppressed RHEED oscillations and metal-rich RHEED reconstruction.

The AlSb NLs were grown by migration-enhanced epitaxy (MEE) that involved alternating deposition fluxes of Al and Sb at different growth temperatures, while the coverage of AlSb on Si was studied using *in-situ* AES and *ex-situ* atomic force microscopy (AFM) imaging. The MEE growth sequence was fixed for all NL substrate temperatures; a cycle consisted of 2 s Al supply, 1 s interruption, and 2 s Sb supply corresponding to growth of 1 monolayer (ML) of AlSb. It is possible (especially at very low substrate temperatures 250–300 °C) for Sb to accumulate on NL surface but no noticeable increase of Sb/Al signal ratio was observed by *in-situ* AES measurements. For comparison, a NL was grown at 535 °C by conventional MBE (no MEE) with simultaneous Al and Sb components delivery on a similarly prepared Si substrate.

Heteroepitaxial $\text{Al}_{0.93}\text{Ga}_{0.07}\text{Sb}$ metamorphic buffers were then grown at 560 °C on the AlSb NLs [Fig. 1(a)]. The buried $\text{In}_{0.36}\text{Ga}_{0.64}\text{Sb}$ p-QW were grown at 450 °C and covered at the same growth temperature with 2 nm p-type $\text{Al}_{0.24}\text{Ga}_{0.76}\text{Sb}$ top barrier, Be-doped at $1\times 10^{18}\text{cm}^{-3}$, and 1 nm InAs surface layer to prevent oxidation of this Al-containing barrier.^{19–21} The biaxial strain in the QWs was previously optimized at 1.8% on GaAs substrates to maximize the hole mobility,²¹ and the similar composition of QWs was maintained on the Si substrates.

The AlGaSb layers were grown at approximately 1 ML/s, AlSb NLs, and InGaSb QW were grown at 0.45 ML/s, and top InAs at 0.2 ML/s. After completing the III-V epitaxy on SOI, the structures were transferred to another chamber for reactive evaporation of Al at 10^{-6} Torr in oxygen ambient to deposit *in-situ* Al_2O_3 gate oxide.²² The oxide was annealed in the forming gas at 350 °C that reduces the interface and oxide trap densities.²⁰ AFM (DI Nanoscope III) was used to examine surface morphology, while cross-sectional transmission electron microscopy (TEM, FEI Titan G2 300 kV) was used to study growth defects in the layers. Figure 1(b) shows a cross-sectional TEM image of a complete structure on SOI and indicates growth-related defects which are discussed below. Hall mobility and hole density in the QWs

10nm in-situ Al_2O_3
1nm InAs
2nm AlGaSb top barrier
10nm i-$\text{In}_{0.36}\text{Ga}_{0.64}\text{Sb}$, 450°C
0.1 μm i-$\text{Al}_{0.8}\text{Ga}_{0.2}\text{Sb}$, 560°C
1.3 μm i-Al(Ga)Sb, 560°C
AlSb NL, varying temperature
5min Sb_2 -soak
SOI Substrate (Si-20nm)

(a)

Fig. 1. (Color online) (a) Schematic cross-section of a heterostructure with $\text{In}_{0.36}\text{Ga}_{0.64}\text{Sb}$ QW grown on SOI substrates illustrating the sequence of the layers and the growth temperature. (b) Large area dark-field TEM image showing growth-related defects in the heterostructure: TDs, MT, and APB within the film.

were measured in Van der Pauw configuration in magnetic field of 0.5 T.

III. RESULTS AND DISCUSSION

A. AlSb NL growth

To improve the quality of the heterostructure, we studied the effect of growth temperature on the AlSb NL using MEE growth technique by alternative supply of Al and Sb_2 fluxes. Due to the large lattice mismatch between Si and AlSb, high strain initiates the growth in the Volmer–Weber island-like growth mode. MEE facilitates low temperature growth by the effective increase of group-III adatom diffusion lengths, and results in films with reduced defect densities.²³ With the goal of increasing nucleation density while keeping high crystal quality in the NL and the following III-Sb heterostructure, the AlSb NLs were grown at different temperatures. During the growth, the coverage of AlSb on Si was monitored by a surface sensitive *in-situ* AES. In this experiment, the substrate was moved between the growth and AES chambers several times until the Si signal vanished and Al signal saturated.

FIG. 2. (Color online) (a) Decay of the substrate Si AES signal with the average thickness (in ML) of AlSb nucleation layer grown at different temperatures. Solid lines are fitting curves with equation $I(t) = I(0)\exp(-Bt^a)$, with t —NL thickness, B , a —fitting constants. (b) AlSb thickness needed to reduce Si AES signal to 50% and 10% at different growth temperatures. Dashed lines are drawn to guide the eye.

As expected, low temperature growth of AlSb promoted faster coverage of AlSb on Si (Fig. 2). The Si AES signal decays quite slowly, indicating the formation of 3D AlSb islands, with the Si signal originating mostly from uncovered areas. The solid curves in Fig. 2(b) are fitted using a simplified version of an analytical model of condensation with migration.²⁴ Figure 2(b) plots the average AlSb thickness needed for 50% and 90% coverage of the Si substrate.

After completing the AES experiments (when Si Auger signal almost vanished), each structure with AlSb thicknesses from 8 to 38 MLs (as indicated in Table I), was

unloaded from the MBE system and imaged with AFM. Figure 3 shows images of the AlSb nucleation layer grown at 250°–535 °C using MEE and grown at 535 °C with conventional MBE for comparison [Fig. 3(f)] and the growth of NL is observed to be uniform over large areas as seen in Fig. 3(g). It can be observed that low temperature promotes faster coverage with higher nucleation density [Fig. 3(h)], which results in earlier transition to the continuous layer growth mode resulting in flatter surfaces described as the “self-flattening effect.” The results of AFM studies of the AlSb NL are shown in Fig. 4. The nucleation density was obtained from the AFM granular contrast and is presented as Arrhenius plot in Fig. 4(a). The nucleation density increases with the reduction of growth temperature with activation energy of 0.38 eV. This value corresponds to the diffusion activation values (0.1–0.7 eV) reported for different materials.^{25,26} MEE almost doubles the nucleation as compared to conventional MBE growth with simultaneous Al and Sb components delivery.

The average roughness of the AlSb NLs [Fig. 4(b)] reduces with decrease of the growth temperature, similarly to the minimum continuous layer thickness [Fig. 2(b)]. It can be concluded that reduction of the NL growth temperature below 350–400 °C is necessary to promote dense nucleation and faster transition to the continuous layer growth. On the other hand, low growth temperatures may result in poor layer crystallinity with high density of defects and even misoriented grains. Therefore, it is further important to study the properties of the consequent layers in the device structure.

B. InGaSb quantum well on AlGaSb metamorphic buffer

Heteroepitaxial 1.2 μm of $\text{Al}_{0.93}\text{Ga}_{0.07}\text{Sb}$ metamorphic buffers with strained buried $\text{In}_{0.36}\text{Ga}_{0.64}\text{Sb}$ QW channels were grown on the AlSb NLs (Fig. 1). The NL thicknesses were chosen to provide complete coverage of the SOI substrates as indicated in Table I. Also summarized in Table I are the average roughness values (from $1 \times 1 \mu\text{m}^2$ images) of AlSb NLs, of the buried p-InGaSb QW-channel heterostructures with different NLs, and corresponding thicknesses of AlSb NLs used for AFM analysis and growth of QW structures. Contrary to the AlSb NL, the roughness of QW structures [Fig. 4(b)] increases at 250 °C NL growth temperature likely due to degradation of crystallinity of the NL.

TABLE I. Specifications of samples studied: growth temperature of AlSb NL and corresponding AFM ($1 \times 1 \mu\text{m}^2$ and $10 \times 10 \mu\text{m}^2$) average roughness of AlSb NLs and thick heterostructures with QWs, thickness of NL used for AFM measurements and thick heterostructure growth.

Growth temperature (°C)	AFM NL $1 \times 1 \mu\text{m}^2$ roughness (R_a), nm	AFM NL $10 \times 10 \mu\text{m}^2$ roughness (R_a), nm	R_a of the completed heterostructure, nm	NL thickness for AFM (for thick heterostructures), in ML
250	0.3	0.22	0.99	8 (10)
300	0.6	0.37	0.87	15 (17)
330	0.9	0.47	0.9	18 (20)
390	1.6	0.59	0.91	30 (32)
535	3.4	2.57	1.33	38 (40)
535 (no MEE)	5.86	2.93	2.12	38 (40)

FIG. 3. (Color online) (a)–(f) AFM images ($1 \times 1 \mu\text{m}^2$) of AlSb nucleation layer grown at (a)–(e) 250–535 °C using MEE, (f) at 535 °C with conventional MBE (no MEE), (g) AFM image $20 \times 20 \mu\text{m}^2$ scan area at 300 °C to demonstrate the uniformity of growth across larger area, and (h) schematics of NL layer formation at different growth temperatures.

The Si/AlSb interface of a completed QW structure with NL grown at 300 °C is shown in Fig. 5(a). A fast Fourier transform (FFT) power spectrum of the interface region [Fig. 5(b)] shows perfect epitaxy, distinct lateral periodicities from Si and AlSb crystal lattices, and allows for evaluation of lattice spacings. The two lattices have the mismatch of $\delta = (a_{\text{AlSb}} - a_{\text{Si}})/a_{\text{AlSb}} = 11.49\%$, if lattice parameters $a_{\text{Si}} = 5.431 \text{ \AA}$ and $a_{\text{AlSb}} = 6.136 \text{ \AA}$ are used. The FFT measured splitting $\delta_{002} = 11.44\%$ and $\delta_{220} = 11.39\%$ for normal and in-plane periodicities prove that the lattice relaxation is completed within a few monolayers of AlSb and the residual strain and tetragonal distortion can be estimated to be less than 0.1% within the accuracy of the instrument and the measurements. The lattice relaxation takes place mostly by the array of 90° misfit dislocations²⁷ as can be seen from Fig. 5(c) with enhanced lattice contrast from the boxed region of the interface in Fig. 5(a) by FFT filtering out of zero and higher-order periodicities. From the image, the misfit dislocations have average periodicity of 8.42 (220) planes that corresponds to relaxation of $\delta_{220_BV} = 11.9\%$ with dislocation Burgers vectors. This value proves the full relaxation of AlSb with 90° misfit dislocation array although the accuracy might be affected by

nucleation of planar defects (microtwins) at the interface [Figs. 5(a) and 5(c)].

Figure 6 shows images of the completed QW heterostructures with AlGaSb buffers. The overall uniformity of the structure is illustrated in a low-magnification SEM image [Fig. 6(a)], showing that the morphology is dominated by small features. These features observed by AFM are 3D bumps elongated along $\langle 110 \rangle$ direction [Fig. 6(b)]. Cross-sectional TEM correlates this to undulations on AlGaSb surface and resulting features in the InGaSb layer as shown in Fig. 6(c). The morphological growth instabilities that occur between or within the closely spaced or wide MTs are likely the origin of this bump formation. The thin InGaSb layer essentially decorates the MTs and may further result in excessive scattering of carriers. This problem needs to be further addressed. These surface features are different from those observed in thick GaSb films grown on similar Si substrates (though using different buffer layer) and showing formation of similar bulk defects. The most prominent features on GaSb-on-Si surface were winding loops from antiphase domain boundaries, and MTs resulted in surface steps with the height in monolayers equal to the number of faulted MT planes.²⁸

FIG. 4. (Color online) (a) Arrhenius plot of nucleation density of AlSb on Si showing 0.32 eV activation energy for Al adatom surface migration. Nucleation density for conventional (no MEE) MBE growth is $2 \times$ lower. (b) Surface roughness of AlSb NL (rhombus) and 1.2 μm thick QW heterostructure (squares) obtained from $10 \times 10 \mu\text{m}^2$ AFM images as a function of the NL growth temperature.

The other surface features due to TDs and APBs are less pronounced, as shown in Fig. 7. TDs show a monolayer-high step on the QW surface as observed from AFM (Ref. 22) and from dark-field STEM image [Fig. 7(a)]. APBs create

FIG. 5. (Color online) (a) High-resolution TEM micrograph of Si/AlSb interface of the completed QW structure with NL grown at 300 °C. (b) FFT power spectrum of the micrograph confirming the epitaxial growth and complete relaxation of the AlSb layer. (c) FFT filtered image of the area boxed in (a) showing formation of 90° misfit dislocation array at the interface.

FIG. 6. (Color online) Surface images of the completed QW structure with NL grown at 300 °C: (a) A typical low magnification SEM image; (b) AFM topographic image of $10 \times 10 \mu\text{m}^2$ scan area ($1 \times 1 \mu\text{m}^2$ in the inset) of the InGaSb QW capped with 2 nm-AlGaSb/1 nm-InAs/10 nm- Al_2O_3 grown on SOI; (c) TEM cross-sectional micrograph of the surface as in the AFM image. The surface 3D features are the result of surface growth instability of InGaSb at double or wide microtwins.

FIG. 7. (Color online) (a) Cross-sectional scanning high-angle annular dark field TEM image of a threading dislocation, causing a monolayer surface step. The inset shows a low-angle DF STEM image of the same spot with the higher contrast from dislocation. (b) Cross-sectional HAADF STEM micrograph showing an antiphase domain boundary creating a surface dip on AlGaSb surface which after InGaSb growth is recovered. Polarity of the adjacent domains is clearly visible from the dumbbell contrast.

1–3 nm high surface dips on the AlGaSb surface but they are mostly “repaired” after the growth of the strained InGaSb layer likely due to the high adatom mobility of the constituents of the QW layer [Fig. 7(b)]. Interestingly, we have not seen this type of surface planarization in GaSb structures where the APBs produced 1–3 nm high surface steps.²⁸

The defect density was quantified from cross-sectional TEM images [Fig. 8(a)], which show fast reduction of both MTs and TDs with the buffer thickness. The baseline defect density at the surface for 1.8 μm thick buffer includes threading dislocation density of $\sim 3 \times 10^8 \text{ cm}^{-2}$, microtwin density of $\sim 2 \times 10^3 \text{ cm}^{-1}$, and APB density of $\sim 10^4 \text{ cm}^{-1}$. These values are about 2–3 times higher than those when grown on GaAs or InP substrates.²²

High nucleation density results in early coverage of AlSb NL on the Si surface at low growth temperatures and also reduces the density of surface defects on the completed structures with QWs. Figure 8(b) plots the density of the MT-related defects quantified from AFM images as in Fig. 6(a). The ratio of surface defect density to the MT density gives the width of the MT, which is about 1 μm in all the samples. The surface defect density curve is expectedly similar to the surface roughness curve in Fig. 4(b), indicating that this roughness is mostly associated with the MT-related surface features.

The QW sample grown with 300 °C NL has exhibited the lowest roughness and lowest MT density and also the highest room temperature Hall p-mobility of the buried QW channel of 660 $\text{cm}^2/\text{V s}$ and almost the lowest sheet density of

Fig. 8. (Color online) (a) Reduction of TDs density and density of MTs with the metamorphic AlGaSb layer thickness quantified from cross-sectional TEM images. (b) Density of surface defects related to microtwins quantified from AFM images as in Fig. 5(a).

Fig. 9. (Color online) Room temperature Hall mobility and sheet hole density in the strained InGaSb QW as a function of nucleation layer growth temperature.

$3.2 \times 10^{11} \text{ cm}^{-2}$ (Fig. 9). It should be noted that since all the structures were similarly doped, the difference in resulting hole density is likely due to electrically active defects, which are normally p-type in III-Sbs. The reduction of defect densities on the optimized AlSb NL grown at 300 °C results in the highest mobility; however, this mobility is significantly less than the value of 900 $\text{cm}^2/\text{V s}$ obtained on a similar buried channel structure grown on GaAs.²¹

IV. SUMMARY

In summary, we studied AlSb nucleation layer growth by migration-enhanced MBE on Si at various temperatures to enhance nucleation density, accelerate coverage, and reduce roughness for the subsequent heterostructure growth with metamorphic AlGaSb buffer layers. The baseline data for defect density in the layers and resulting surface morphology are presented. The dominant surface features are elongated bump-like defects due to microtwins, while the presence of other defects such as APBs and TDs do not have a prominent effect on the surface morphology. The optimum growth temperature of ~ 300 °C is found for AlSb NL, resulting in room temperature Hall mobility of 660 $\text{cm}^2/\text{V s}$ at $3 \times 10^{11} \text{ cm}^{-2}$ sheet hole density in the strained InGaSb QW p-channel. These data will provide the baseline for further research in III-Sb MOSFETs on Si.

ACKNOWLEDGMENTS

The authors would like to thank Shun Sasaki for his valuable technical help and discussions. The work was carried out with the support of Semiconductor Research Corporation under the contract 2013-VJ-2395, GLOBALFOUNDRIES, and SUMCO Corporation.

¹S. F. Fang, K. Adomi, S. Iyer, H. Morkoc, H. Zabel, C. Choi, and N. Otsuka, *J. Appl. Phys.* **68**, R31 (1990).

²K. Eisenbeiser, R. Emrick, R. Droopad, Z. Yu, J. Finder, S. Rockwell, J. Holmes, C. Overgaard, and W. Ooms, *Electron. Dev. Lett.* **23**, 300 (2002).

³M. E. Groenert, C. W. Leitz, A. J. Pitera, V. Yang, H. Lee, R. J. Ram, and E. A. Fitzgerald, *J. Appl. Phys.* **93**, 362 (2003).

⁴M. K. Hudait, *ECS Trans.* **45**, 581 (2012).

⁵V. A. Joshkin, S. R. Oktyabrsky, I. A. Bogonin, and A. A. Orlikovsky, *J. Cryst. Growth* **132**, 209 (1993).

- ⁶H. Kobayashi and M. Kawabe, *Jpn. J. Appl. Phys.* **29**, L1342 (1990).
- ⁷J. W. Lee, H. Shichijo, H. L. Tsai, and R. J. Matyi, *Appl. Phys. Lett.* **50**, 31 (1987).
- ⁸R. Fischer, D. Neuman, H. Zabel, H. Morkoc, C. Choi, and N. Otsuka, *Appl. Phys. Lett.* **48**, 1223 (1986).
- ⁹T. Soga, S. Sakai, M. Umeno, and S. Hattori, *Jpn. J. Appl. Phys.* **25**, 1510 (1986).
- ¹⁰M. A. Todd, G. Bandari, and T. H. Baum, *Chem. Mater.* **11**, 547 (1999).
- ¹¹C. A. Wang, *J. Cryst. Growth* **272**, 664 (2004).
- ¹²K. Akahane, N. Yamamoto, S. I. Gozu, and N. Ohtani, *J. Cryst. Growth* **264**, 21 (2004).
- ¹³K. Akahane, N. Yamamoto, S. Gozu, A. Ueta, and N. Ohtani, *J. Cryst. Growth* **283**, 297 (2005).
- ¹⁴J. Tatebayashi, A. Jallipalli, M. N. Kutty, H. Shenghong, K. Nunna, G. Balakrishnan, L. R. Dawson, and D. L. Huffaker, *J. Sel. Top. Quantum Electron.* **15**, 716 (2009).
- ¹⁵K.-M. Ko, J.-H. Seo, D.-E. Kim, S.-T. Lee, Y.-K. Noh, M.-D. Kim, and J.-E. Oh, *Nanotechnology* **20**, 225201 (2009).
- ¹⁶H. Toyota, S. Fujie, M. Haneta, A. Mikami, T. Endoh, Y. Jinbo, and N. Uchitomi, *Phys. Proc.* **3**, 1345 (2010).
- ¹⁷J. B. Rodriguez, L. Cerutti, P. Grech, and E. Tournie, *Appl. Phys. Lett.* **94**, 061124 (2009).
- ¹⁸A. S. Bracker, M. J. Yang, B. R. Bennett, J. C. Culbertson, and W. J. Moore, *J. Cryst. Growth* **220**, 384 (2000).
- ¹⁹B. R. Bennett, M. G. Ancona, J. B. Boos, and B. V. Shanabrook, *Appl. Phys. Lett.* **91**, 042104 (2007).
- ²⁰A. Greene, S. Madiseti, P. Nagaiah, M. Yakimov, V. Tokranov, R. Moore, and S. Oktyabrsky, *Solid State Electron.* **78**, 56 (2012).
- ²¹S. Madiseti, T. Chidambaram, P. Nagaiah, V. Tokranov, M. Yakimov, and S. Oktyabrsky, *Phys. Status Solidi (RRL)* **7**, 550 (2013).
- ²²V. Tokranov, S. Madiseti, M. Yakimov, P. Nagaiah, N. Faleev, and S. Oktyabrsky, *J. Cryst. Growth* **378**, 631 (2013).
- ²³W. Stolz, Y. Horikoshi, and M. Naganuma, *Jpn. J. Appl. Phys.* **27**, L1140 (1988).
- ²⁴V. I. Trofimov, K. Jongil, and B. Seongi, *Mater. Sci. Eng. C* **27**, 1397 (2007).
- ²⁵J. A. Venables, *J. Vac. Sci. Technol. B* **4**, 870 (1986).
- ²⁶D. Leonhardt and S. M. Han, *Surf. Sci.* **603**, 2624 (2009).
- ²⁷J. Narayan and S. Oktyabrsky, *J. Appl. Phys.* **92**, 7122 (2002).
- ²⁸S. Madiseti, V. Tokranov, A. Greene, S. Novak, M. Yakimov, S. Oktyabrsky, S. Bentley, and A. P. Jacob, *MRS Proc.* **1635**, T02.03 (2014).

Copyright of Journal of Vacuum Science & Technology: Part B-Nanotechnology & Microelectronics is the property of AVS, The Science & Technology Society and its content may not be copied or emailed to multiple sites or posted to a listserv without the copyright holder's express written permission. However, users may print, download, or email articles for individual use.