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Strain relaxation and induced defects in InAsSb self-assembled quantum dots

J. F. Chen,[a] R. S. Hsiao, and W. D. Huang
Department of Electrophysics, National Chiao Tung University, Hsinchu 30050, Taiwan, Republic of China

Y. H. Wu and L. Chang
Department of Materials Science and Engineering, National Chiao Tung University, Hsinchu 30050, Taiwan, Republic of China

J. S. Wang
Department of Physics, Chung Yuan Christian University, Chung-Li 32023, Taiwan, Republic of China

J. Y. Chi
Industrial Technology Research Institute (OES/ITRI), Hsinchu 310 Taiwan, Republic of China

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The onset of strain relaxation and induced defects in $InAs_{0.94}Sb_{0.06}$ quantum dots are investigated. We show that the relaxation causes partial carrier depletion in the dots and drastic carrier depletion in the top GaAs layer due to the introduction of two defect traps at 0.35 and 0.64 eV. This result is consistent with transmission electron microscopy data which show misfit dislocations on the edges of the dot upper boundary and threading dislocations in the top GaAs layer. The bottom GaAs layer is dislocation-free, and thus the strain relaxation may initially occur on the edges of the dots. © *2006 American Institute of Physics*. [DOI: 10.1063/1.2212064]

Self-assembled InAs quantum dots[1–5] (QDs) are of great interest for practical applications and scientific studies. Such dots are usually formed by Stranski-Krastanow growth. When depositing the InAs layer to about 1.7 ML (monolayer),[2–4] strain is partially relieved by forming the dots coherently. To extend the emission wavelength, it is necessary to increase the dot size by increasing the InAs thickness.[6] However, when the InAs thickness reaches a critical thickness[7] (about 3 ML), strain is further relaxed by introduction of misfit dislocations. This strain relaxation is more complicated than in a planar system and may be affected by the incorporation of isoelectronic dopants. The introduction of antimony (Sb) as a surfactant[8,9] into a strained InGaAs quantum well (QW) has been shown to delay three-dimensional (3D) growth and thus extend the emission wavelength. Recently, dilute Sb has been incorporated into InAs QDs (Ref. 10) and an emission at 1.3 μm at room temperature has been demonstrated.[11] Despite these efforts, however, the properties of strain relaxation in InAsSb QDs have never been characterized. Thus, in this letter, we report our studies of the strain relaxation in InAsSb QDs by performing capacitance-voltage (*C-V*), transmission electron microscopy (TEM), and deep-level transient spectroscopy (DLTS) measurements.

InAsSb QDs were grown on n^+-GaAs (100) substrates by molecular beam epitaxy in a Riber Epineat machine. The QDs were formed in Stranski-Krastanow growth mode by depositing the InAsSb layer at 485 °C and a growth rate of 0.256 Å/s. A conventional Sb *K* cell was set at 305 °C to provide a beam flux of 3.5×10^{-7} torr, which corresponds to 6% Sb composition estimated from the x-ray data of GaAsSb layers. Reflection high-energy electron diffraction (RHEED) patterns showed a transition to spotty after ~25 s growth of the InAsSb layer, in comparison to ~20 s for the QDs without Sb, indicating a retarded QD formation by the Sb incorporation.[10] For electrical characterizations, the QDs are sandwiched between two 0.3-μm-thick Si-doped GaAs (6×10^{16} cm^{-3}) layers. Detailed growth of the QDs can be found elsewhere.[12] Schottky diodes were realized by evaporating Al on the sample with a dot diameter of 1500 μm.

Figure 1 shows the 20 K carrier distributions converted from the *C-V* spectra of the $InAs_{0.94}Sb_{0.06}$ QDs Schottky diodes with 2-, 2.2-, and 2.8-ML-thick InAsSb layers. The 2 and 2.2 ML samples show a strong carrier-accumulation peak at the dots. Both peaks show no frequency-dependent attenuation, reflecting a very fast electron emission for quantum states. The peak linewidth of the 2.2 ML sample is much narrower than that of the 2 ML sample due to improved size uniformity for larger dots, as indicated by photolumines-

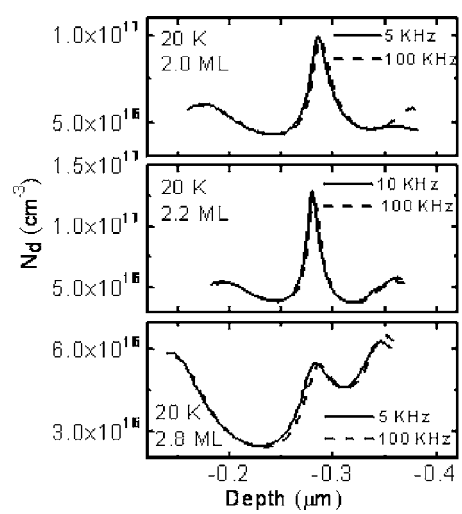

FIG. 1. Carrier distributions at 20 K of the InAsSb QDs samples with 2-, 2.2-, and 2.8-ML-thick InAsSb layers. The 2 and 2.2 ML samples show strong carrier accumulation at the dots but the 2.8 ML sample exhibits weak accumulation at the dots and drastic carrier depletion in the top GaAs layer.

[a]Electronic mail: jfchen@cc.nctu.edu.tw

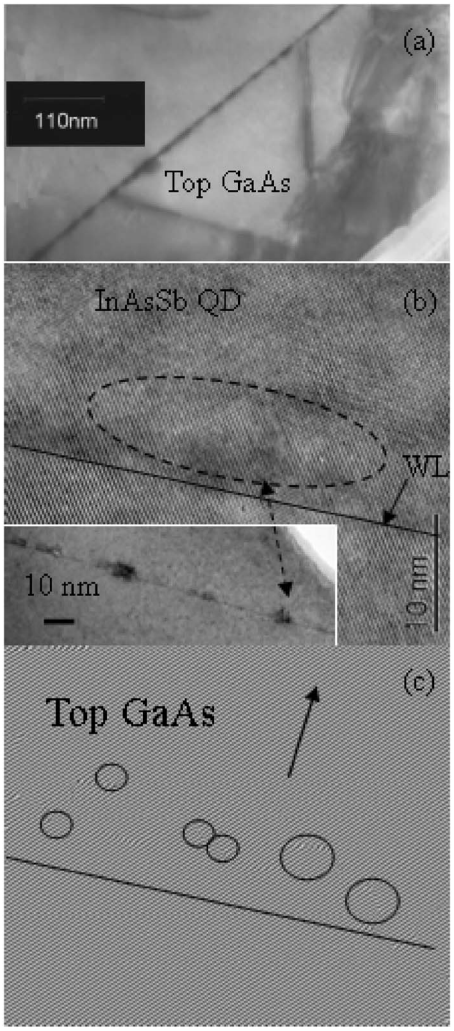

FIG. 2. Cross-sectional TEM picture of the 2.8 ML sample, illustrating a line of dots and threading dislocations in the top GaAs layer. (b) The HRTEM picture of a dot (dashed ellipse), corresponding to the one at the right-hand side in the inset. (c) The corresponding Fourier transformed image, showing two misfits in each of the two large circles at the right-hand side and one misfit in each of the rest.

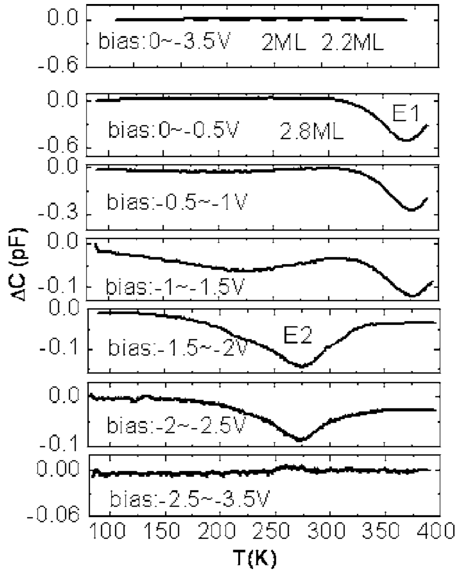

FIG. 3. DLTS spectra of the studied samples. The 2 and 2.2 ML samples exhibit no traps due to a coherent formation of the QDs but the 2.8 ML sample exhibits a trap E_1 in the GaAs top layer and a trap E_2 near the dots. No traps are visible in the GaAs bottom layer.

cence (PL) spectra which show a redshift of 300 K QD ground-state emission from 1235 to 1255 nm and decreased linewidth as increasing the InAsSb thickness from 2 to 2.2 ML. In contrast, the 2.8 ML sample exhibits weak accumulation at the dots (at 0.28 μm) and drastic carrier depletion (at around 0.23 μm) in the top GaAs layer. This sudden degradation is accompanied with a broadening of a PL emission at 1285 nm and can be attributed to strain relaxation by introduction of carrier-depletion traps. As consistent with the normal carrier distribution in the bottom GaAs layer, the weak accumulation peak at the dots shows no frequency-dependent attenuation, suggesting that no traps are induced in the neighboring GaAs bottom layer, since any traps there would considerably increase the emission time.[13] Accordingly, the critical thickness for defect introduction should be between 2.2 and 2.8 ML, a value slightly smaller than those (between 2.7 and 3.06 ML) previously obtained in InAs QDs,[6,13] probably due to an increased lattice mismatch due to the Sb incorporation.

The normal carrier distributions in the 2 and 2.2 ML samples are consistent with their TEM data, where no apparent dislocations were observed. On the other hand, the cross-sectional TEM picture of the 2.8 ML sample, as shown in Fig. 2(a), shows threading dislocations in the top GaAs layer. Traps associated with these threading dislocations can be responsible for the drastic carrier depletion in the top GaAs layer. Figure 2(b) shows the high resolution TEM (HRTEM) picture of a typical dot (dashed ellipse) which is the one at the right-hand side in the inset of the figure. This dot has a height of ~5 nm and a width of ~20 nm. To see lattice misfits, Fig. 2(c) shows the corresponding Fourier transformed image. A total of eight misfits with two in each of the two large circles at the right-hand side and one in each of the rest circles can be seen. Among them, six are on the edges of the QD upper boundary and two are inside the dot. For a dot density of $\sim 3\times 10^{10}$ cm^{-2} observed by atomic force microscopy (AFM), the average density of the misfits is $\sim 2.4\times 10^{11}$ cm^{-2}. Defects associated with these misfits can trap electrons and explain the partially carrier depletion in the dots. Note that the bottom GaAs layer is dislocation-free and thus the dot lower interface likely remains coherently strained. From the spatial distribution of these misfits, we deduce that the strain relaxation initially occurs at the edges of the dot upper boundary where considerable stress may be accumulated or inhomogeneities or threading dislocations are probably produced for providing nucleation sources for misfit dislocations.

The relaxation-induced traps were revealed by the DLTS spectra shown in Fig. 3. As shown in the top figure, the 2 and 2.2 ML samples, measured by sweeping from 0 to -3.5 V, exhibit no traps, indicating a coherent formation of the QDs. In contrast, the 2.8 ML sample exhibits a trap E_1 in the top GaAs layer and a trap E_2 near the dots. No traps are visible in the GaAs bottom layer. Arrhenius plots yield an activation energy (capture cross section) of 0.64 eV $(2.11\times 10^{-15}$ cm$^{-2})$ for the E_1 trap and 0.35 eV $(1.18\times 10^{-17}$ cm$^{-2})$ for the E_2 trap. As filling pulse width increases, the intensity of the E_1 trap continuously increases without saturation. Figure 4(a) shows that its 300 K transient response, measured after sweeping from 0 to -0.5 V, can be fitted by a logarithmic function (open squares). This signa-

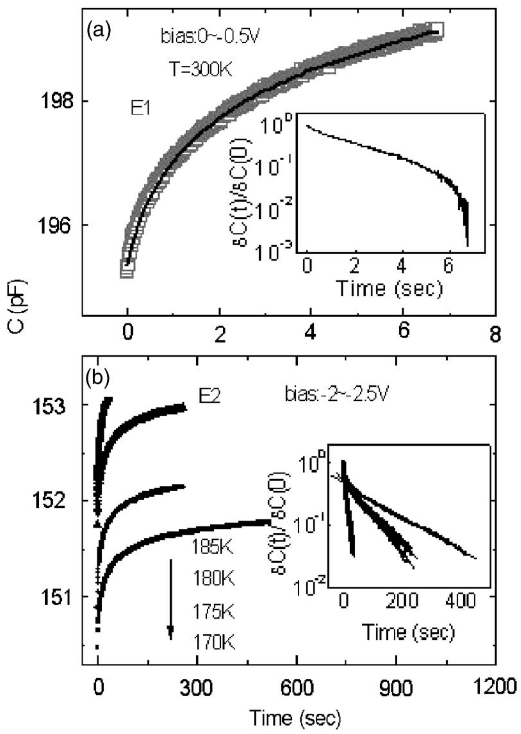

FIG. 4. (a) Transient response of the E_1 trap, measured after sweeping from 0 to -0.5 V. This response can be fitted by a logarithmic function (open squares). (b) Transient response of the E_2 trap, measured after sweeping from -2 to -2.5 V, which, except at the very initial time, can be fitted by an exponential function.

ture is characteristic of Coulombic repulsion of the carriers captured at the traps along the linearly arrayed dislocation lines,[14] and thus confirms that the E_1 trap is associated with the threading dislocations in the top GaAs layer. On the other hand, the intensity of the E_2 trap is found to saturate at ΔC=0.125 pF with filling pulse width. From this saturated intensity, a sheet density is obtained to be 10^9 cm^{-2} from $N_T=N_D(\Delta C/C_0{}^2)\varepsilon A$ where $N_D=6\times10^{16}$ cm^{-3}, C_0=220 pF, area $A=5\times10^{-3}$ cm^2, and permittivity $\varepsilon=1.14\times10^{-10}$ F/m. This concentration is about two orders of magnitude less than that of the misfits estimated from the TEM data, and thus the E_2 trap is unlikely associated with the localized state due to the destruction of lattice translational symmetry by misfit dislocations. Figure 4(b) shows that its transient response, by sweeping from -2 to -2.5 V, except at the very initial time, can be fitted by an exponential function. This signature suggests that the E_2 trap may be isolated point defects associated with misfit dislocations. By comparison, these two traps are believed to be the E_1 and E_5 traps previously observed in relaxed InAs QDs,[7] indicating that they are common relaxation-induced traps and are independent of Sb incorporation. The other three traps observed in Ref. 7 are not detected in this study. Since isoelectronic dopants have been shown to reduce the dislocation density[15] or inhibit the motion of dislocations,[16] we suspect that the Sb incorporation may play a role of preventing the occurrence of these traps. Detailed studies are underway in this direction.

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