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Long wavelength InAs self-assembled quantum dots embedded in GaNAs strain-compensating layers

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Abstract

We investigated the effect of GaNAs strain-compensating layers (SCLs) on the properties of InAs self-assembled quantum dots (QDs) grown on GaAs (0 0 1) substrates. The GaNAs material can be used as SCL thereby minimizing the net strain, and thus is advantageous for multi-stacking of InAs QDs structures and achieving long wavelength emission. The emission wavelength of InAs QDs can be tuned by changing the nitrogen (N) composition in GaNAs SCLs due to both effects of strain compensation and lowering of potential barrier height. A 1.3 μm photoluminescence emission at 77 K was clearly observed for sample with $GaN_{0.024}As_{0.976}$ SCL. Further, we observed an improvement of optical properties of InAs QDs by replacing the more popular GaAs embedding layers with GaNAs SCLs, which is a result of decreasing non-radiative defects owing to minimizing the total net strain.

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Keywords: Quantum dots; MBE; Strain compensation; GaNAs

1. Introduction

Self-assembled and self-organized quantum dots (QDs) have been extensively investigated for future applications to high-performance QD lasers [1], single electron transistors [2], and high-efficiency compound solar cells [3]. Long emission wavelengths of 1.3 and 1.55 μm have been the focus of intense research for application to future optical fiber communication systems. In the case of InAs QDs/GaAs systems, the optical emission appears usually around 1.0–1.1 μm due to a strong three-dimensional quantum confinement as well as compressive strain accommodated around the QDs. To realize long wavelength emission from InAs QD structure, InGaAs [4,5] and GaAsSb [6] strain-reducing layers (SRLs) have been recently used to bury InAs QDs. Recently, long wavelength emission from QDs without SRLs, such as InGaSb QDs and In(Ga)AsN QDs, have been also proposed [7,8]. However, the accumulation of strain with multi-stacking is still a limiting factor. On the other hand, if InAs QDs were embedded in GaNAs strain-compensating layers (SCLs), one can expect long wavelength emission while retaining a high optical property as a result of net strain reduction [9]. Consequently, multi-stacking of InAs QD structure becomes possible [10,11]. In this work, we report on the properties of self-assembled InAs QDs embedded in GaNAs SCLs grown on GaAs (0 0 1) substrate.

2. Experiments

We fabricated self-assembled InAs QDs embedded in GaN_xAs_{1-x} SCLs on GaAs (0 0 1) substrates by atomic H-assisted RF molecular beam epitaxy (H-MBE) [12]. It is possible to obtain higher quality Ga(In)NAs materials by atomic H irradiation during growth [13]. First, each GaAs substrate was subjected to cleaning in the growth chamber at 580 °C for 30 min with atomic H irradiation in order to obtain an atomically flat surface. Then a 300 nm-thick GaAs buffer layer was grown at a growth rate of 1μm/h. After lowering the substrate temperature to 480 °C, 2.0 monolayers (MLs) of InAs QD layer were deposited at 0.1 μm/h. Subsequently, 20 nm-thick GaN_xAs_{1-x} SCL and

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20 nm-thick GaAs cap layer were further grown. The growth rate of GaNAs layer was varied from 0.2 to 1.5 μm/h in order to control the N composition. The RF power and N_2 flow rate were set to 200 W and 1.0 sccm, respectively. The RF plasma was generated before the onset of QDs layer growth, and GaNAs SCL was grown immediately after the QDs without growth interruption. In this work, we varied the N composition of GaNAs SCL from 0.15% to 2.4%. No post growth annealing was performed. The As_4 and H_2 back pressures during the growth were 1.2×10^{-6} and 5.0×10^{-6} Torr, respectively. The growth process and surface morphology were studied in situ by reflection high-energy electron diffraction (RHEED), and ex situ by using atomic force microscopy (AFM). The photoluminescence (PL) was measured by using a standard lock-in technique at 77 K with the second harmonic of a YAG laser at a wavelength of 532 nm as the excitation source and InGaAs photodetector.

3. Results and discussion

Fig. 1 shows the AFM image of InAs self-assembled QDs on GaAs (0 0 1) substrate without the top capping layer. The average QD diameter, height and area density were 35.3 nm, 7.3 nm, and $3.7\times10^{10}\,cm^{-2}$, respectively. This image shows that a good size uniformity in the diameter with dispersion of only 15%.

Fig. 2 shows the PL spectra at 77 K for single InAs QD layers embedded in GaN_xAs_{1-x} SCLs with varying N composition, (a) $x = 0\%$ (GaAs cap), (b) 0.4%, (c) 1.6%, and (d) 2.4%, respectively. We also summarized the PL peak wavelength as a function of N composition of GaNAs SCL in the inset. First, PL peak is immediately red-shifted upon replacing the GaAs embedding layer with GaNAs

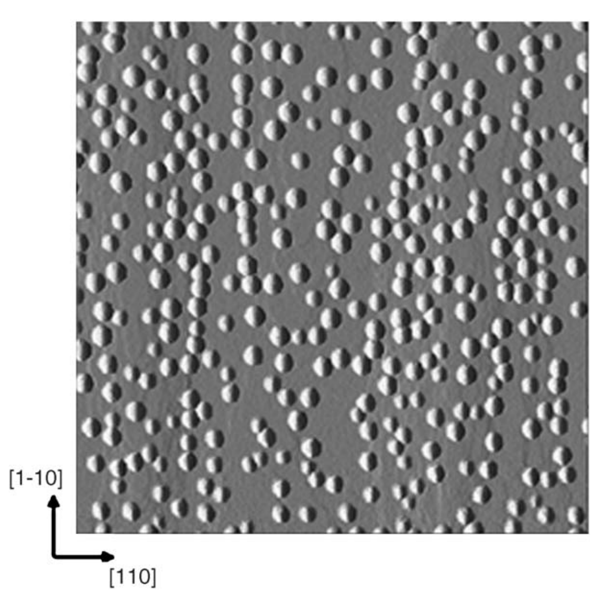

Fig. 1. AFM image of typical InAs QD structure grown on GaAs (0 0 1) substrate. Scan size is 1 μm × 1 μm.

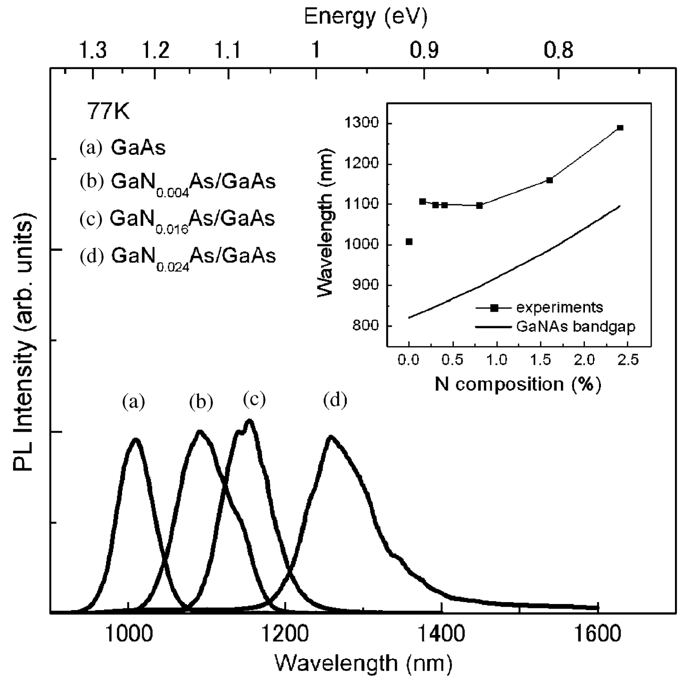

Fig. 2. PL spectra at 77 K from InAs QDs embedded in GaN_xAs_{1-x} SCLs with varying N composition. Inset shows center peak wavelength as a function of N composition.

SCL, which is primarily due to the effect of strain compensation. Second, PL peak shifts to longer wavelength with increasing N composition. This is due to lowering of quantum energy levels as the barrier height of GaNAs SCLs is reduced. The band gap energy of GaNAs as a function of N composition is also plotted as a reference in the inset of Fig. 2. It can be seen that the slope shows a good fit to that of experimental results in the higher N composition (0.8–2.4%) range. Since the addition of nitrogen increases mainly the conduction band offset [14], lowering of quantum energy level is expected mainly to occur at the conduction band. Here, in order to investigate interface diffusion of N atoms into InAs QDs layer, we performed post-growth anneal at 600 °C for 5 min in N_2 ambient. The emission wavelength from InAs QDs showed slight blueshift. This suggests that the diffusion of In atoms to GaNAs layer occurs more easily than that of N atoms to InAs QDs layer. Therefore, we attribute the main mechanism of redshift to strain compensation effect and/or reduction of barrier height. Further, a 1.3 μm PL emission at 77 K was clearly observed for sample with 2.4% nitrogen composition of GaNAs SCL. Therefore, the emission wavelength of InAs QDs is shown to be controllable by changing the N composition of GaNAs SCLs. But further investigation is still needed in order to achieve longer wavelength of 1.55 μm range.

Fig. 3 plots the integral PL intensity and full-width at half-maximum (FWHM) as a function of N composition of GaNAs SCL. We notice that PL intensity becomes stronger up to 0.8% compared to GaAs cap ($x = 0$), although the potential barrier height becomes lower.

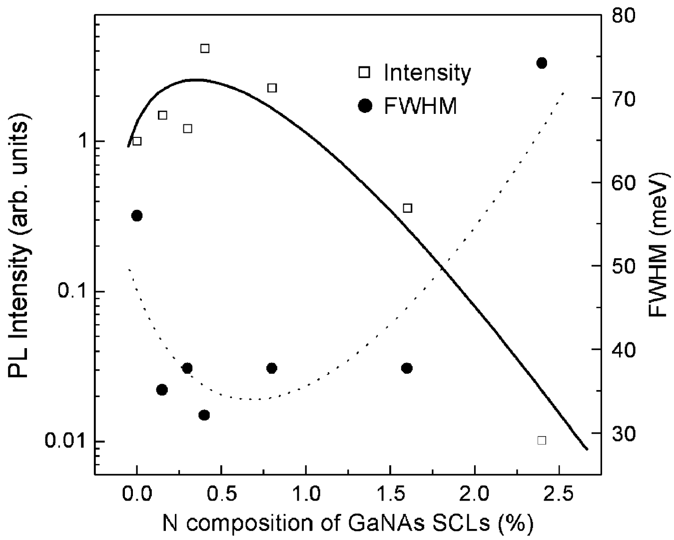

Fig. 3. Integral PL intensity (open square) and full width at half-maximum (FWHM) (circle) as a function of N composition of GaNAs SCL.

This indicates that generation of defects near the QDs is effectively suppressed due to reduction of total lattice strain. However, PL intensity starts to decrease with increasing of N composition up to 2.4% as a result of, (1) lowering of potential barrier height of GaNAs SCLs, and/or (2) degradation of GaNAs material quality caused by phase separation and localization of N atoms, which become strongly apparent at low growth rates in Ga(In)NAs matrix. On the other hand, the FWHM decreases with increasing N composition up to 0.8%. This improvement suggests that QD shape and size homogeneity are modified by the net strain compensation by covering the InAs QDs with GaNAs SCLs. Similar improvement of optical characteristics has been observed by other growth methods [9].

To compare the radiative emission efficiency of InAs QDs covered by GaAs and GaN_xAs_{1-x}, the PL properties were evaluated as a function of temperatures from 20 to 250 K. Fig. 4 shows the temperature dependence on PL intensity from InAs QDs covered by (a) GaAs, (b) $GaN_{0.004}As$, and (c) $GaN_{0.016}As$. The experimental data can be fitted with the equation below,

$$I = I_0/\{1 + C\exp(-E_a/k_B T)\}, \quad (1)$$

where E_a and k_B are the activation energy of confined carriers and Boltzman constant, and I_0 and C are the constants independent of temperature, respectively. C is proportional to the density of non-radiative recombination centers and it is defined as the ratio of the probability of non radiative to radiative recombination process at high temperatures [15]. Although the activation energy of the sample with GaNAs SCLs obtained from the fits are slightly smaller than that of GaAs capped sample (120 meV), which is due to lowering of potential barrier height, we note that C constants related to the non-radiative recombination probability becomes smaller by a factor of 4–30. Therefore, the optical quality of the InAs QDs buried in GaNAs SCLs is much improved by adopting net strain compensation growth technique.

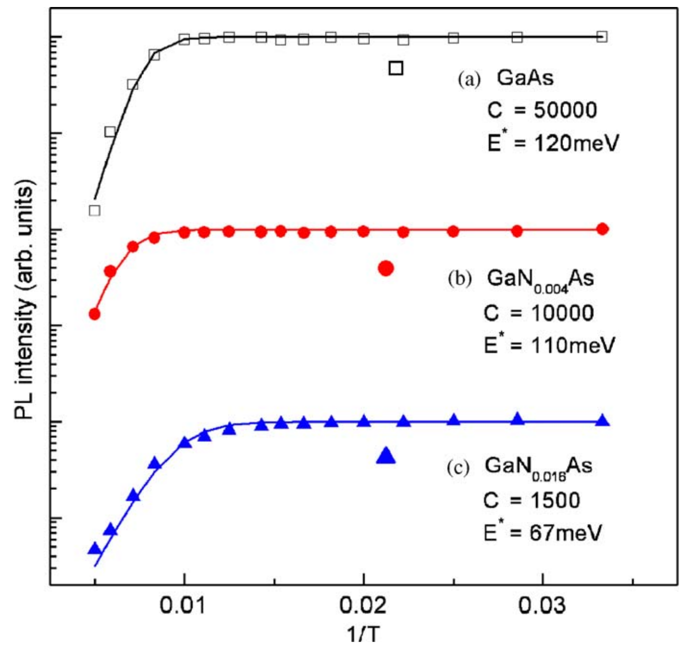

Fig. 4. Temperature dependence of PL intensity from InAs QDs covered by (a) GaAs, (b) $GaN_{0.004}As$, and (c) $GaN_{0.016}As$, respectively.

4. Summary

In summary, we investigated the effect of GaNAs SCLs on the optical properties of InAs self-assembled QDs grown on GaAs (0 0 1) substrates. By replacing the GaAs embedding layer with GaNAs SCL, (1) PL peak shifts toward longer wavelength and 1.3 μm emission was successfully obtained from InAs QDs embedded in $GaN_{0.024}As$ SCL, (2) PL intensity becomes stronger up compared to GaAs cap peak, although actual potential barrier becomes lower, and (3) density of the non-radiative recombination centers is decreased for samples embedded with GaNAs SCLs. These features are as a consequence of strain compensation. Therefore, GaNAs SCLs may be useful in realizing high performance long wavelength devices.

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