Room temperature emission at 1.6 μm from InGaAs quantum dots capped with GaAsSb

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Room temperature photoluminescence at 1.6 μm is demonstrated from InGaAs quantum dots capped with an 8 nm GaAsSb quantum well. Results obtained from various sample structures are compared, including samples capped with GaAs. The observed redshift in GaAsSb capped samples is attributed to a type II band alignment and to a beneficial modification of growth kinetics during capping due to the presence of Sb. The sample structure is discussed on the basis of transmission electron microscopy results. © *2005 American Institute of Physics*. [DOI: 10.1063/1.2130529]

In the last few years there has been a strong interest in expanding the usable wavelength range of GaAs based optoelectronic devices towards 1.5 μm and beyond.[1] Only a handful of research groups have achieved photoluminescence (PL) in the 1.5–1.7 μm range from GaAs based samples, using different approaches. By growing quantum dots (QDs) on InGaAs metamorphic layers, Ledentsov *et al.* have reached room temperature electroluminescence at 1.49 μm.[2] This group had previously obtained PL at 1.7 μm by growing InAs/GaAs QDs at unusually low temperatures (325 °C).[3] Growing QDs on metamorphic buffer layers also enables long wavelength luminescence (1.63 μm) as shown by Xin *et al.*[4] Le Ru *et al.* have obtained photoluminescence at 1.48 μm by growing double layers of InAs QDs capped with InGaAs.[5] Sun *et al.* obtained 1.55 μm emission from GaInNAs quantum wells.[6] Akahane *et al.* and Liu *et al.* recently reported photoluminescence up to 1.4 μm using GaAsSb capping layers on InAs QDs,[7,8] and Pötschke *et al.* have described a QD ripening effect when exposing InAs QDs to an Sb flux.[9] In this letter we report room temperature photoluminescence at 1.62 μm from InGaAs QDs capped with GaAsSb.

Samples are grown on GaAs(001) substrates using a home built solid source molecular-beam epitaxy (MBE) system of custom design. During sample growth the As_4 and Sb_4 beam equivalent pressures are 1.7×10^{-6} Torr and 2.2×10^{-7} Torr, respectively. The InGaAs QDs layer is grown below, above, or in the middle of an 8 nm GaAsSb quantum well (QW). Another sample with an 8 nm GaAsSb QW is grown to calibrate the fraction of incorporated Sb. X-ray diffraction indicates an average 17% Sb content.

The first step in our research, before using Sb, has been the optimization of the InGaAs QD layer. The emission wavelength from InGaAs quantum dots is limited by the appearance of structural defects at high In coverages. A compromise thus exists between the In concentration (which determines the bulk band gap) and the size of the dots (which determines the confinement energy). Most groups have attempted to obtain pure InAs QDs with sharp interfaces with the GaAs matrix. This results in a hard confinement potential and consequently the wavelength is mostly limited by the size of the nanostructures. Instead, we have grown large InGaAs alloyed QDs, so that the emission wavelength is mostly limited by the QD composition rather than by the quantum dot size. There are various advantages to our approach: Larger critical volume for dislocation formation, smaller compressive stress, and larger electron effective mass, which in turn lowers the ground-state energy relative to the bottom of the potential well.

We have used 2.2 ML (500 °C) of InAs as a QD seed layer, followed by 8 ML of $In_{0.5}Ga_{0.5}As$. Even if no seeding layer is used, the critical thickness for InGaAs QD nucleation in our growth conditions is 4 ML, so the InGaAs layer can not be expected to be in the form of a thin quantum well (as in dot in a well or DWELL structures). Chang *et al.* recently showed that a substantial fraction of the In deposited during growth of the InGaAs layer contributes to enlarging the underlying QDs.[10]

The QD nucleation density of InGaAs is larger than that of InAs. As a consequence, if both the InAs and the InGaAs layers are grown at the same temperature, some InGaAs QDs will nucleate on the wetting layer rather than onto the InAs seeding dots. StGirons *et al.* first observed this effect to result in two families of QDs, with the InGaAs/InAs QDs emitting at a lower energy than the InGaAs QDs.[11] Thus the growth temperature of the InGaAs layer was optimized to 520 °C in order to decrease the InGaAs QD nucleation density so that most of the InGaAs would grow onto the seeding InAs QDs rather than on the wetting layer. A 1.4 μm limit was reached in the room temperature photoluminescence from the InGaAs QD samples, at which point we began to explore the possibilities afforded by Sb incorporation either above or below the InGaAs QD layer.

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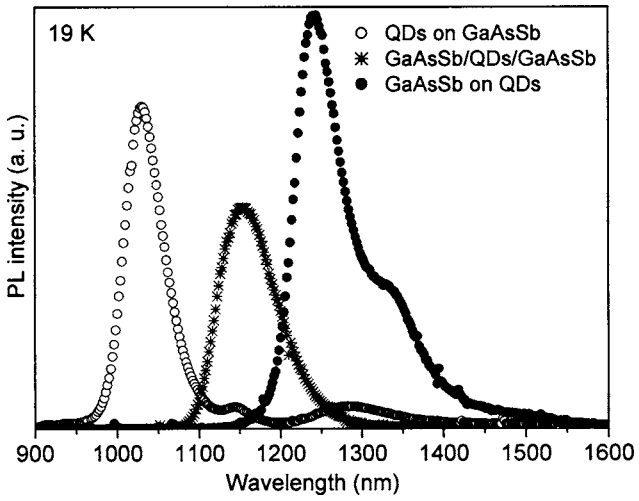

FIG. 1. Photoluminescence at 19 K of an 8 nm GaAsSb QW with InGaAs QDs either above, below, or in the middle of the well.

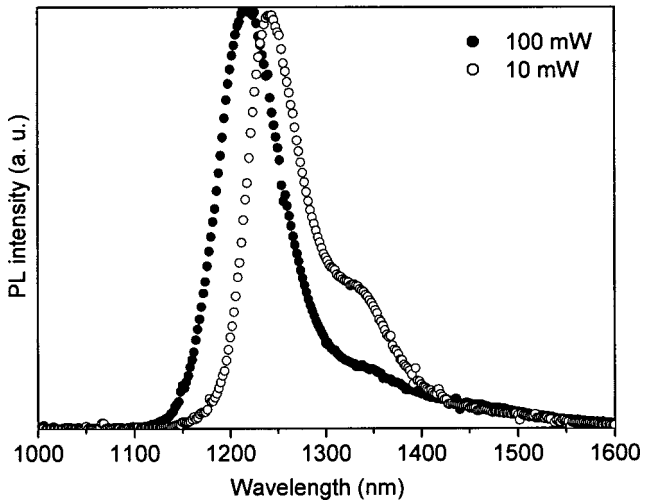

FIG. 3. Excitation dependence of the photoluminescence at 19 K of InGaAs QDs capped with GaAsSb.

In Fig. 1 we compare the low temperature PL spectra of samples with an InGaAs QD layer grown below, above, or in the middle of an 8 nm GaAsSb QW. The sample with the InGaAs QDs on top of the GaAsSb QW has blueshifted PL at 1030 nm. QD nucleation on the GaAsSb QW results in a blueshift relative to samples without Sb. This is most likely due to an increased nucleation density, and therefore, smaller QD volume. Sun *et al.* have reported an increased nucleation density and reduced critical thickness for InAs QDs grown on Sb covered GaAs (001) surfaces.[12] Much lower luminescence energy (wavelength range between 1200 and 1400 nm) is obtained when the GaAsSb QW is grown as a cap layer on top of the GaInAs QDs. As we are mostly interested in reaching long wavelength emission from GaAs based devices, we will now focus our attention on the case of InGaAs QDs capped with a GaAsSb QW.

Figure 2 shows a comparison of the room temperature spectra for InGaAs QD samples capped with either GaAs or a GaAsSb 8 nm quantum well with 17% Sb content. The GaAs capped InGaAs QDs have a narrow inhomogeneous broadening (28 meV at 19 K). The high energy shoulder in the spectrum from the GaAs capped sample is an excited state emission (51 meV above the main peak). The presence of Sb during capping of the InGaAs QDs has two major effects: A 146 meV redshift and a broadening of the main PL peak. The maximum intensity of luminescence is located at a wavelength of 1.62 μm for the GaAsSb capped InGaAs QDs layer. The reduced photoluminescence intensity from GaAsSb capped InGaAs quantum dots might be related to the type II band alignment in these heterostructures. The separate electron and hole confinement is likely to lead to long radiative recombination lifetimes, and this makes the samples more sensitive to non radiative recombination at structural defects, such as those due to the large lattice misfit induced by Sb and In incorporation.

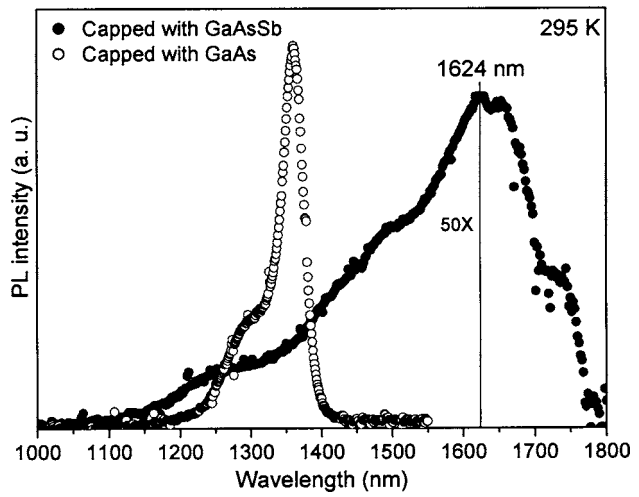

FIG. 2. Comparison of the room temperature photoluminescence of GaAs and GaAsSb capped InGaAs QDs.

The low temperature (19 K) PL from the GaAsSb QW sample without QDs has its maximum at 1108 nm. When coupled to the InGaAs QDs the PL from the GaAsSb QW redshifts to 1244 nm. This peak shows a 23 meV blueshift when the excitation intensity is increased by an order of magnitude (see Fig. 3). This is a typical fingerprint of type II spatially indirect recombination. Liu *et al.* have shown[8] that such type II alignment in GaAsSb capped QDs occurs when the Sb content of the capping layer is above 14%, in good agreement with our own results. No additional peaks due to state filling appear with increasing excitation intensity. The width of the main peak at 1244 nm increases by 8 meV when the excitation power is increased by an order of magnitude. At room temperature, emission from QDs prevails, and emission from the GaAsSb QW is quenched due to thermal escape to the QDs. In stark contrast, the GaAsSb QW sample without QDs shows intense room temperature PL peaking at 1168 nm.

In Fig. 4 we show a transmission electron microscopy (TEM) micrograph of the sample with an 8 nm GaAsSb QW grown on top of the InGaAs QDs. The thickness of the GaAsSb layer ranges from about 5 to 10 nm. The increased width of the PL spectra of the GaAsSb capped sample is most likely due to these fluctuations in QW thickness. The GaAsSb layer follows the morphology of the underlying QD layer, and a planar growth front is only recovered after GaAs growth, in agreement with the *in situ* reflection high energy electron diffraction (RHEED) observations.

In conclusion we demonstrate GaAsSb capping of InGaAs QDs as a viable alternative to develop GaAs based epitaxial materials active in the 1.5–1.7 μm spectral region.

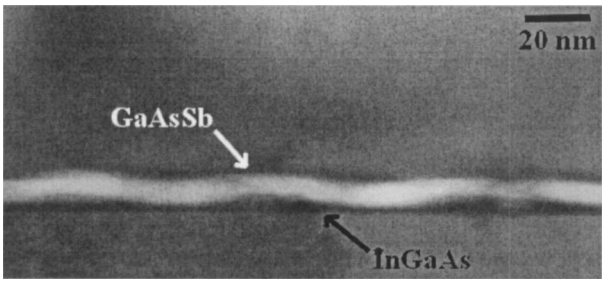

FIG. 4. (002) DF reflection TEM micrograph near the [001] zone axis of GaAsSb capped InGaAs QDs.

Clearly much remains to be done. Enhancements in luminescence intensity and nanostructure size distribution are needed, but these initial experiments are promising given the wide range of possible growth conditions.

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