Strain engineered InAs/GaAs quantum dots for 1.5 μm emitters

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We show that it is possible to obtain emission up to 1.5 μm from InAs/GaAs quantum dots using a combination of low growth rates, a seed layer, variable substrate temperature, and InGaAs capping. These strain engineered structures exhibit a remarkably small linewidth (14 meV) consistent with islands that are very uniform in both composition and size.

1 Introduction Optoelectronic components for telecommunications applications around 1.3 and 1.55 μm are currently based on InP substrates. Over the last few years it has been shown that InAs/GaAs quantum dots (QDs) are capable of reaching the lower wavelength and although commercial products are beginning to become available, there are still problems that need to be overcome in order to exploit fully the potential of QD emitters. Key issues are the low gain associated with the QD ground state (GS) and the large inhomogeneous broadening which means that only a subset of the ensemble contributes to the gain at the operating wavelength.

We have pioneered a method of achieving emission around 1.3 μm based on reducing the growth rate of the QDs which results in relatively large islands with a composition that is close to pure InAs [1, 2]. There is a concomitant reduction in the inhomogeneous linewidth to ~25 meV consistent with the increased island height but the island density is reduced by a factor of three. This exacerbates the problem of low gain, which we have attempted to overcome by growing multilayer samples optimised so that each layer emits at the same wavelength [3]. Other techniques such as atomic layer epitaxy [4], seed layers [5], strain reducing layers [6] and dots in a well (DWELL) [7] have also demonstrated emission around 1.3–1.35 μm but this appears to be the limit for structures which show good room temperature emission. In this paper we demonstrate that the emission can be extended to 1.5 μm by relieving the strain using a seed layer and by growing and capping at a low temperature with InGaAs. The seed layer also dictates the dot density allowing independent control of density and size (composition). The linewidth is reduced to ~14 meV. These results suggest that QDs may become the active layer of choice for telecommunications applications.

2 Experimental details and growth The structures were grown using solid source molecular beam epitaxy (MBE) in a combined MBE-Scanning Tunnelling Microscope (STM) system. After oxide removal a 0.5 μm GaAs buffer was grown at a substrate temperature, T_S, of 580 °C on the epi-ready n^+ GaAs wafer. T_S was then reduced to 500 °C and the sample annealed for 10 minutes under an As_2 flux RHEED measurements then showed a $c(4 \times 4)$ surface reconstruction prior to InAs deposition at 0.016 monolayers per second as determined from the 2D–3D transition. This layer constitutes the seed layer. A GaAs spacer of thickness d was then deposited and T_S increased to 580 °C to reduce surface undulations

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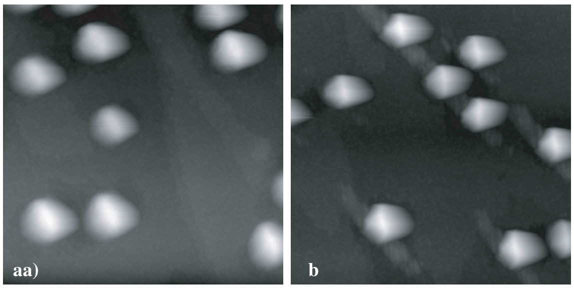

Fig. 1 STM images of the 2nd QD layer for uncappped versions of samples A and B. Despite the difference in T_S the density is the same due to the underlying seed layer.

induced by the seed layer [3, 8]. T_S was then reduced to values in the range 450–500 °C in preparation for deposition of the 2nd QD layer. This 2nd layer can be grown at a rate different to the first but this is not critical. Finally the structure was overgrown with 40 nm of GaAs before increasing T_S to 580 °C for the remaining 100 nm cap.

STM constant current images were obtained with a sample bias of –3.5 V and tunnelling currents in the range 0.05–0.2 nA. Photoluminescence (PL) was excited in capped structures with an Ar^+ laser, the emission dispersed using a SPEX 0.5 m monochromator and detected with a North Coast cooled Ge-diode and lock-in amplifier.

3 Experimental results

3.1 Structural assessment Figure 1 shows STM images of the 2nd layer of two samples, which are uncapped versions of samples A and B in Table 1. Sample A consists of two nominally identical seed layers formed by deposition of 2.5 ML of InAs at T_S = 500 °C, separated by an 11 nm thick GaAs spacer. Sample B was similar except that the 2nd layer was deposited at a lower substrate temperature T_S = 470 °C. The islands have good uniformity with similar diameter ~30 nm and are present at a density of $2.0 \pm 0.2 \times 10^{10}$ cm^{-2} which is the same as that of the seed layer (not shown here). In both cases the 2D–3D transition for the 2nd layer occurred after deposition of 1.4 ML rather than 1.9 ML for the seed layer. There is a slight difference in the average height of the islands; for sample A this is 6.16 nm whereas for sample B it is 6.96 nm.

3.2 Optical properties In order to deconvolve the effects of strain and indium segregation a series of optical samples was grown whose details are listed in Table 1.

Figure 2 compares the low temperature emission obtained from samples A, B and C under the conditions of low excitation to minimise the influence of excited states. Sample A yields a PL spectrum consisting of a single, relatively broad (FWHM ~40 meV), asymmetric peak centred around 1.12 µm. This is consistent with emission from a single layer grown and capped at 500 °C. For sample B the 2nd layer (and the first 40 nm of the cap) is deposited at T_S = 470 °C, the emission is redshifted and the linewidth

Table 1 Details of the optical samples.

sample	T_S (growth)	T_S (cap)	PL (10 K) (µm)	FWHM (meV)
A	500	500	1.123	40
B	470	470	1.255	17
C	470	470	1.278	14
D	500	500	1.178	23
E	470	500	1.201	19
F	500	470	1.210	23
G*	500	470	1.178	27

* denotes a single layer

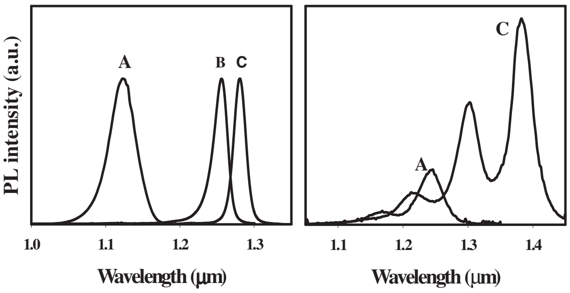

Fig. 2 Left panel shows the low excitation 10 K PL for samples A, B and C. Right panel compares the RT emission from samples A and C.

reduces to ~17meV. Sample C is nominally identical to sample B except that more material (3.2 ML) was deposited in the 2nd layer. This results in a further redshift and narrowing of the peak to only 14 meV. Note that in each case only GaAs is used as the cap material. InGaAs capped samples will be discussed later. The right hand panel of Fig. 2 compares the room temperature emission from samples A and C obtained under the conditions of high excitation. The GS emission of C is close to 1.4 µm, which is the longest wavelength yet reported for InAs/GaAs QDs capped only with GaAs. Interestingly, the first excited state emits around 1.3 µm and this may be advantageous in laser structures since the gain of this transition is twice that of the GS. The emission from the GS and first excited state of sample A indicates where the seed layer of sample C should emit and shows that there is a strong overlap with the second excited state of the second layer. The integrated intensity obtained from sample C is a factor of five higher than that of sample A.

4 Discussion It might be expected that the seed layer would induce a tensile strain in the second layer leading to a redshift in its emission. However, as our previous studies have shown [3, 8], there is strain induced In/Ga intermixing in the 2nd layer, which induces a blueshift. The emission from sample A therefore consists of two closely overlapping peaks yielding a broad asymmetric emission feature. Sample B was grown and capped at a lower substrate temperature of 470 °C in order to reduce the effects of intermixing in the 2nd layer. This is the main reason for the observed redshift. There is also a significant reduction in the linewidth, which we attribute to a greater uniformity in the composition of the dots owing to the reduced intermixing effects. The increased redshift and decreased linewidth observed for sample C is again a manifestation of increased island height due in this case to deposition of more InAs. A comparison of samples A and C at high excitation (Fig. 2) shows that there is strong coupling between the two QD layers and in the case of sample C emission from the seed layer is hardly detectable (this is also true at 10 K). This implies that carriers captured by the seed layer relax to the GS and then rapidly tunnel to an excited state of the 2nd layer. When the excitation level is increased further so that the first excited state of the 2nd layer is close to saturation, the GS of the seed layer becomes detectable.

In order to deconvolve the effects of strain reduction and suppression of intermixing, additional samples (D, E and F in Table 1) nominally identical to sample C were grown but the growth and capping of the 2nd layer were varied independently. Comparing the emission of sample D with that of C the decrease in growth and capping temperature results in a redshift of 82 meV. If the *lower growth temperature* alone was responsible then sample E should emit at the same energy as C, but E exhibits only a 20 meV redshift compared with D. Similarly if a *lower capping temperature* were the main factor, sample F should emit at the same energy as sample C. In fact sample F redshifts relative to sample D but only by 27 meV. We conclude that both parameters contribute to the redshift but in different ways. For instance, the islands in the 2nd layer of samples B and C are taller than for sample A. Therefore we associate a lower growth temperature with taller islands. This explains the redshift of sample E compared with D. We propose that the capping temperature determines the degree of intermixing and this explains why sample F is redshifted compared with D since the islands prior to capping should be the same. Implicit in this analysis is the assumption that the seed layer induces a strain relaxation in the 2nd layer. The purpose of sample G (see Table 1) was to compare the emission from a single layer with those of a seeded 2nd layer keeping the dot size and density the same. Consequently sample G consists of a single layer of dots

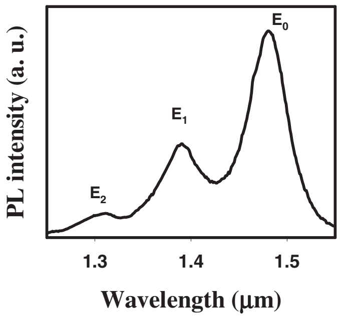

Fig. 3 Room temperature emission form a strain engineered InGaAs capped structure. The GS (E_0) emission occurs around 1.5 µm. Note that the second excited state, E_2, emits at 1.3 µm and has a potential gain three times E_0.

grown under the same conditions as sample F. STM images of an uncapped sample confirmed that the density and island size were similar. Sample F exhibits a redshift of 27 meV compared with sample G and this difference is attributed to strain relaxation of the 2^{nd} layer. All three factors contribute to the long wavelength emission seen for sample C; strain relaxation, increased dot height and suppression of intermixing.

5 InGaAs cap It has been argued that capping with InGaAs reduces the strain in the dots and inhibits indium segregation [6]. We have therefore investigated whether additional benefits can be obtained using InGaAs capping applied to the strain engineered structures described in the previous section. Figure 3 shows the room temperature emission obtained from sample H, nominally identical to sample C but capped with $In_{0.25}Ga_{0.75}As$. At this relatively high excitation level three emission peaks (E_0, E_1 and E_2) corresponding to transitions involving the GS and the first two excited states are detected. At low excitation the linewidth of the GS is again ~14 meV demonstrating the uniformity of size and composition offered by this method of growth. Although emission wavelengths around 1.5 µm will have obvious applications in telecommunications it should be noted that these structures may provide a route to efficient emitters at 1.3 µm since the second excited state can have a gain three times that of the GS and this might obviate the need for more layers to enhance the gain. This would be especially useful for Vertical Cavity Surface Emitting Lasers.

6 Conclusions We have demonstrated that it is possible to extend the emission wavelength of InAs/GaAs QDs to ~1.5 µm through a combination of a seed layer which fixes the dot density regardless of substrate temperature, low growth rates and suppression of intermixing effects during the capping stage. It is anticipated that these techniques will be refined and 1.55 µm QD devices will soon be competing with InP based quantum well emitters.

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