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Characterization of structure and defects in dot-in-well laser structures

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Abstract

Recent progress in the development of 1.3 μm InAs/InGaAs dots-in-a-well (DWELL) laser structures has led to efficient CW room temperature laser operation with low current thresholds. However, present devices suffer from non-ideal temperature characteristics due to gain saturation, a consequence of the finite dot density, and to carrier escape due to the small energy separation between the quantum dot (QD) ground and first-excited states. In order to improve device performance, we have examined methods to increase the QD quality and density. In these studies, we have examined the effect of using thin InAlAs capping layers and high temperature buffer layers. Both effects are observed to strongly modify the structure of the QDs producing significant improvements in the InAs QDs optical properties at room temperature.

Initial attempts at multilayer QD structures showed substantial degradation in optical and electrical properties compared to single layer structures. Analysis by Transmission Electron Microscopy (TEM) has identified the presence of defects arising from the complex interaction of QDs, which propagate through the QD layers into the upper regions of the structure as being the primary cause of the poor electronic device characteristics. The use of high growth temperature spacing layers (HGTSLs) have recently allowed us the fabrication of a defect free five layer-stacked structure with record low threshold current density.

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1. Introduction

InAs-based Quantum Dot (QD) lasers, using GaAs substrates, have been subject of considerable recent research effort due to their potential for application within the 1300–1550 nm fiber optical communication waveband. The InAs QD laser offers potential improved performance in terms of low threshold current density and improved thermal stability, whilst exploiting lower cost GaAs technology.

At present, 1.3 μm GaAs/InAs QD lasers with ultra-low threshold current densities of ~20 A/cm^2 have now been routinely demonstrated [1,2]. Values of characteristic temperature (TO) for 1.3 μm QD lasers as high as 80 K, but more typically 30–50 K, have been reported [3]. Of the two main techniques used to fabricate 1.3 μm emitting QDs, atomic layer Epitaxy (ALE) and InAs dots in a InGaAs quantum well (dot-in-a-well—DWELL), where the InAs QD is positioned within an $In_xGa_{1-x}As$ quantum well (QW) of low indium composition (x) of typically 0.1–0.2, have been successful. The latter technique has the advantage that it results in a higher dot density, potentially allowing the fabrication of laser devices with uncoated facets. However, the presence of Stacking Faults (SF) and threading dislocations (TD) are often associated with the large lattice mismatch in most III–V semiconductor films. Several investigators have suggested that the defects in III–V semiconductors originate during the three-dimensional (3D) island coalescence stage of Stranski-Krastanow (SK) growth [4–8]. In particular, Ernst and Pirouz [4] have suggested, based on transmission electron microscope (TEM) investigations of GaAs, that {111} twins and stacking faults form during growth on {111} faceted 3D islands.

This paper presents a structural characterization study of single and multilayer 1.3 μm multilayer DWELL structures. The study shows that defect formation in these devices is a significant problem, but that the effects can be minimized by two techniques: the use of a high growth

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temperature GaAs spacer layer (HGTSL) and the use of thin AlInAs capping layers.

2. Experimental

Single and multilayer InAs DWELL devices were grown on an Oxford Instruments VG Semicon V90+ solid-source molecular beam epitaxy system on 3-in. n+ GaAs (100) substrates. The DWELL structure consists of a 2 nm $In_xGa_{1-x}As$ layer, a 3.0 ml InAs quantum dot and a 6 nm $In_xGa_{1-x}As$ cap layer all grown at a temperature of 510 °C and with $x=0.15$. The dot is therefore placed asymmetrically within the well, which has both thickness and composition previously optimized as a function of room temperature photoluminescence intensity. Both the details of the DWELL structure and the full laser designs have been previously reported [14].

The first set of samples consists of a single buried DWELL layer, with the 6 nm cap $In_xGa_{1-x}As$, but without the 2 nm $In_xGa_{1-x}As$ below the dot which is usual in the full DWELL structure. On the surface of this sample, at a height of 200 nm above the buried DWELL layer, a second QD layer was grown. In sample A1 the DWELL cap layer was composed of $In_xGa_{1-x}As$ in the usual way, whilst sample A2 had a 0.9 nm InAlAs layer inserted between the QD and the $In_xGa_{1-x}As$ cap.

The second set of samples consists of multilayer DWELL laser structures. Samples B1 and B2 are structurally similar except that for B1 all the active region was grown at 510 °C whereas, in the case of the sample B2, the GaAs barrier layer was divided in two parts, the first 15 nm at 500 °C and the second 35 nm at an elevated temperature of 585 °C.

The morphology of the superficial InAs QDs was investigated using a Digital Instruments Dimension Atomic Force Microscopy (AFM) system in ambient conditions in tapping mode. Cross-sectional transmission electron microscopy (XTEM) specimens were prepared by mechanical polishing followed by ion milling in a Gatan PIPS instrument at low incidence angle and examined in a Philips 420 microscope. PL measurements were performed at room temperature with an Ar^+ laser emitting at 532 nm and a cooled Ge detector.

3. Results

Room temperature PL spectra of samples A1 and A2, with and without the InAlAs capping layer respectively, Fig. 1, showed that there were no significant differences in the emission wavelengths. However a large difference in both the emission intensity and line width was observed. Sample B2 is ~10 times brighter and exhibits a reduction from 55 to 38 nm of the PL full width at half maximum (FWHM).

Dark field (DF) Cross-sectional Transmission Electron Microscopy (XTEM) images with the (200) diffraction vector showed that the density of QDs is 6.5×10^9 cm^{-2} in the case of sample A1 and 3.5×10^9 cm^{-2} in the case of sample A2. The surface dot density observed by AFM was the same within experimental errors. In both samples it was observed that approximately 1.1×10^9 cm^{-2} (~30%) of the QD islands are relaxed, from the extended strain contrast they exhibit in (004) BF images [9,10]. The presence of these relaxed QDs leads to the formation of single or double stacking faults, the latter having a characteristic 'v-shape' gliding on the {111} planes towards the surface, Fig. 2. The presence of these SFs is observed to create surface QDs with a much higher volume than the rest of the dots grown on top of these structures. We suggest that the surface steps created at the intersection of the stacking fault to the surface are responsible for this phenomenon.

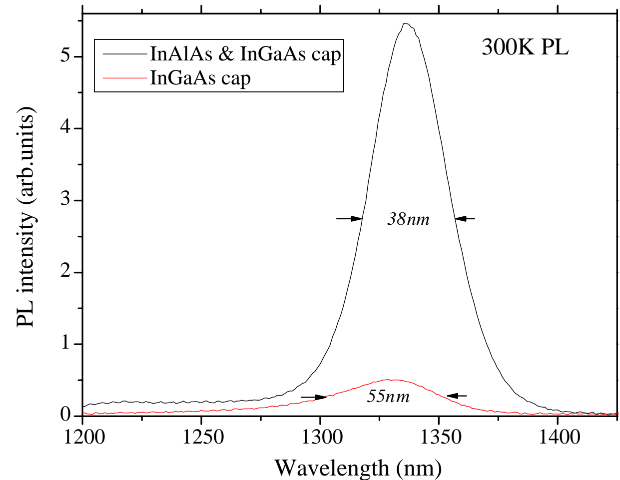

Fig. 1. Room temperature PL spectra of samples A1 and A2.

In the analysis of the surface QDs by AFM large number of QDs with round regular shape without any preferential direction were observed. However, it was clear that two groups of QDs with identical shape but widely different size (defining size as height/diameter ratio [11]) were observed in both samples: small-size QDs (~0.1) and big-size QDs (~0.2).

Fig. 3 shows TEM cross-sectional images of a 5-layer DWELL non-HGTSL laser structure (B1) where large defected areas are observed. These defected areas are formed from the complex interaction of a number of QDs within one of the lowest (typically either the 1st or 2nd) lying dot planes. Dislocation loops and threading dislocations are generated in these planes extending to the upper layers generating a V-shape defected area. These defects exist at a density of $\sim1\times10^4$ cm^{-1} and their presence locally reduces the QD density e.g., the density of dots is reduced from 2.2×10^5 cm^{-1} to 8.3×10^4 cm^{-1} from the lowest (1st) layer to the upper (5th) layer in Fig. 3b. Distortions in the GaAs barrier region are also observed around these defects. These are seen more clearly in Fig. 4, where thin AlAs marker layers have been inserted in the GaAs barrier to show the profile. Modulations are observed

Fig. 2. Dark field (200) TEM cross-sectional images of stacking faults originated from dislocated QDs in samples A1 and A2.

to a depth of up to 50 nm above the defective region. In low magnification TEM micrographs, as Fig. 3a, it was observed that extended defects (threading dislocations) are propagated into the upper cladding regions, which destroy the optical and electronic properties of the devices.

Detailed TEM analysis of these defects has shown that they are formed from a complex entanglement of dislocation loops and, unusually, different Burgers vectors are present, both of the 60° and screw orientation at the interface. Since the dislocations glide on the (111) planes, these must have formed after the cap was grown, but since these are only mobile at high temperatures they must have been formed in the very latter stages of the growth process.

In contrast to the above results, the introduction of the HGTSLs, as in sample B2, suppresses the formation of dislocated QDs and the presence of threading dislocation is removed. No defects were observed from a five-layer device grown with HGTSLs and the QD density was estimated as 1.9×10^5 cm^{-1} for all the QD layers, see Fig. 5.

4. Discussion

Typical InAs QDs grown on GaAs have densities $\sim 1 \times 10^9$ cm^{-2}. However, even at this relatively low density, there exist a significant number of dislocated QDs within the distribution. In the first set of samples (samples A) we observed contrast effects of InAlAs and InGaAs overgrowth layer upon the optical and the structural properties of 1.3 μm InAs/GaAs QDs.

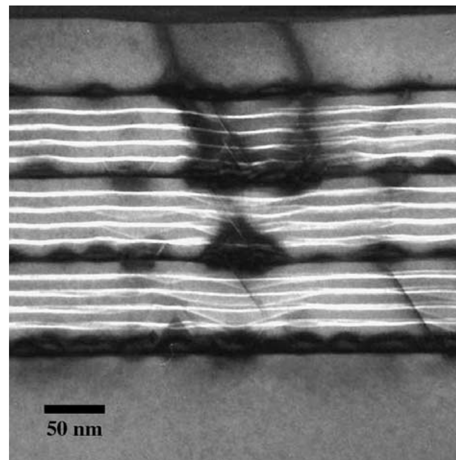

Fig. 4. Dark field (200) TEM cross-sectional image of a structure grown without HGTSLs where AlGaAs layers were used as makers during the GaAs barrier layers growth.

Significant improvements of the InAs QD optical properties have been observed at room temperature in sample A2, which has the InAlAs overgrowth, compared to that of A1. Whereas in sample A1, 30% of the total amount of QDs showed associated defects, the use of 0.9 nm InAlAs layers in sample A2 gives a significant reduction in the number of dislocated QDs, resulting in an enhancement of the RT PL intensity. At present, the precise mechanism is unclear. However we would suggest this could be related to dislocation pinning or another barrier to dislocations occurring at the microscopic level.

Samples B1 and B2 were grown with a 2 nm InGaAs layer under the islands with the purpose of reaching higher QD density suitable for DWELL laser studies. Moreover, the structure included five layers of QDs at 50 nm distances to improve device performance.

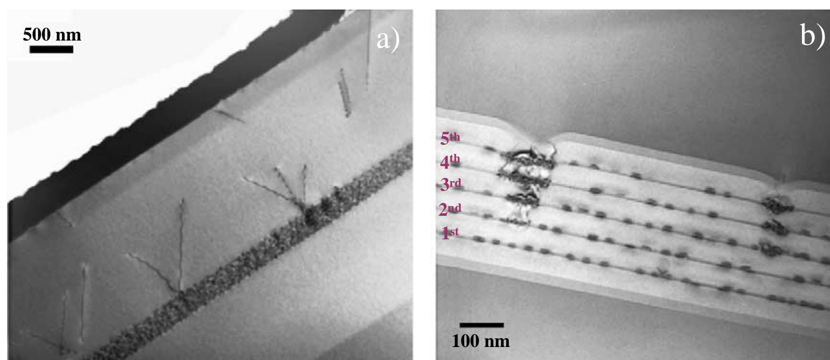

Fig. 3. Cross-sectional TEM micrographs of sample B1. (a) Low magnification micrograph using the 220 reflection and (b) 200 dark field image at high magnification.

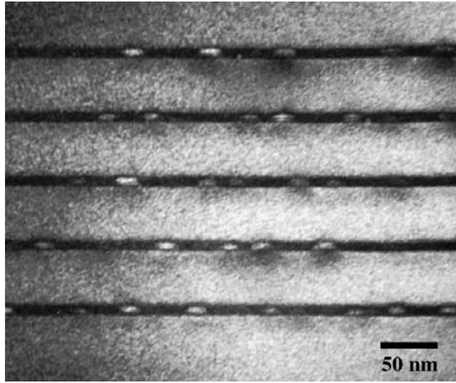

Fig. 5. Dark field (200) TEM cross-sectional image of B2 sample.

Effectively, an increment of one order of magnitude of these islands was obtained using the 2 nm underlying layer, which is highly desirable for reducing gain saturation effects. However, it is evident that this is not without cost, since the generation of a different kind of defect is observed as a consequence of the high density. These are dislocation loops and threading dislocations generated from the coalescence of QDs, an example of which is observed in sample B1. The use of AlAs thin layers sandwiched in the GaAs barrier layers as markers has allowed us to follow the profile of the growth front in a similar sample. The presence of the dislocated QDs considerably distorts the barrier region. This can be understood since relaxed, or dislocated, QDs are not pseudomorphic, so GaAs layers grown on top of these islands suffer a tensile deformation. When the following QD layer is deposited these depressions act as an effective sink for incoming indium adatoms. The preferential formation of InAs islands in this region creates locally very high dot concentrations, which inevitably makes close contact between dots, introducing more dislocations that propagate along the {111} planes to the next QD layer, where the process takes place again. This chain-effect originates a macroscopic V-shape defective region, which causes the degradation of the optical properties of the device. The tensile deformation acts to encourage incoming Ga adatoms to migrate away from this region, producing the resulting depression. This depression can be seen in Fig 6, which shows AFM images (a) immediately after the growth of the QD layer and (b) after the completion of the DWELL and a further 10 nm of GaAs cap.

The use of the HGTSL in B2 suppresses the formation of these V-shape defects; however, the mechanism by which this suppression takes place is as yet unclear. Ledentsov et al. [12,13] have reported a technique in which a very thin GaAs cap layer is deposited after the InAs/InGaAs DWELL, followed by a high temperature (>600 °C) selective evaporation of uncovered larger dislocated dots. However, this mechanism is unlikely to operate in the present structures that use a much thicker cap (21 nm) and a lower temperature (580 °C). In addition there is no evidence for large unfilled voids in the TEM images. At this stage, we do not have a comprehensive understanding of the effects of growth temperature on the surface morphology, however we suggest that Ga adatom migration driven by the strain field associated with the defective InAs islands and the depression in the barrier is negligible beyond a certain thickness [14]. However, the non-planar surface will remain. The higher surface mobility of the Ga atoms will then play an important role in smoothing this strain-free non-planar GaAs surface [14] and the rate of planarization will be increased by the increased Ga mobility that occurs at elevated temperatures [11].

5. Conclusions

We have observed that relaxation of InAs QDs based on the DWELL structure is a significant phenomenon and that its effect on subsequent layers must be considered in multilayer samples with finite spacer thicknesses. Such dislocations nucleate around incoherent islands and may extend through subsequent dot layers. The use of thin InAlAs cap layers would appear an effective way of

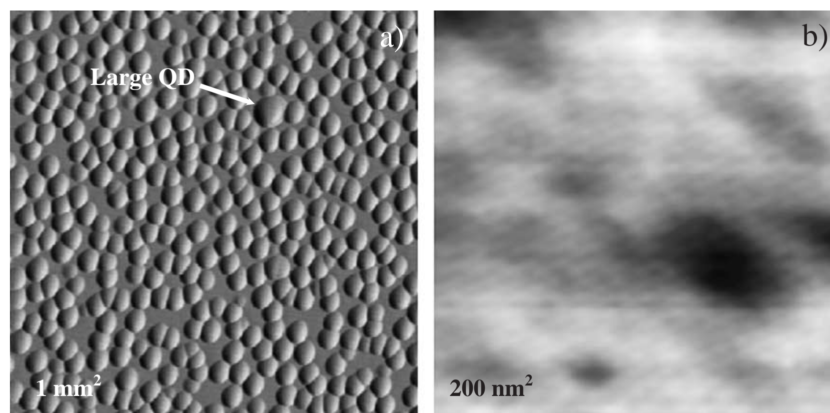

Fig. 6. AFM micrographs of (a) InAs QD layer where the presence of a large QD is arrowed and (b) of a depression in the 10 nm GaAs capping layer grown on top of InAs DWELL QDs.

reducing the relaxation of the dot layers by acting on the dislocation mechanism at a microscopic level. The result is that these structures exhibit a stronger PL intensity and a narrower line width than samples with only InGaAs cladding layers.

For DWELL samples with 2 nm InGaAs below the QD, a significantly higher dot density is observed which is highly desirable for efficient optical emission. However, due to the high island density, coalescence of QDs occurs, generating dislocations that can propagate upwards. In multilayer structures the interaction of these dislocations with subsequent layers is observed to locally distort the lattice, encouraging dot formation above the relaxed region. The overall effect is to create a macroscopic defected region composed of clusters of dislocated dots that severely degrade the optical and electronic quality of laser structures. The use of the HGTSLs inhibits the propagation of the disturbed GaAs spacer layer profile avoiding in this way the coalescence of QDs and consequently the formation of threading dislocations.

The use of both techniques offer promise to improve the quality of DWELL laser structures designed for 1.3 μm emission.

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