Room-temperature optically-pumped InGaSb quantum well lasers monolithically grown on Si(100) substrate

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Room-temperature optically-pumped $In_{0.2}Ga_{0.8}Sb$ quantum well lasers on Si are reported. The defect-free monolithic epistructure growth on a Si(100) substrate is initiated by an AlSb quantum dot nucleation layer followed by an AlSb/GaSb superlattice. The 13% mismatch between the AlSb and Si lattice is accommodated by misfit dislocations and associated crystallographic undulations in the AlSb buffer. The nucleation layer and buffer are characterised by atomic force microscopy and transmission electron microscopy. The lasing spectrum is characterised as a function of pump power and polarisation analysis.

Introduction: The growth of III–V materials on Si has been pursued for over two decades to achieve monolithic integration of light emitters with Si device technology [1–6]. Room-temperature GaAs/AlGaAs lasers [2] and even vertical cavity lasers [3] grown on Si(100) have been demonstrated in the past. While these early results were encouraging, the device characteristics were marginal due to microcracks and high dislocation density in the GaAs buffer [4]. Our approach for mismatched epitaxy on Si is based on an AlSb buffer nucleated on Si via self-assembled AlSb quantum dots (QDs). The growth of high-quality AlSb layers on Si has been observed by others [7]. However, the identification and characterisation of the AlSb QD nucleation to ensure defect-free epilayers has only recently been described [8]. The strain relief is achieved by a combination of surface undulations and misfit dislocations that propagate parallel to the substrate rather than threading dislocations as in GaAs grown metamorphically on Si [8, 9]. In this Letter, we demonstrate room temperature (RT) optically-pumped (OP) lasing of an $In_{0.2}Ga_{0.8}Sb$ quantum well (QW) laser monolithically grown on Si(100) substrate.

Experiment and results: Prior to growth, the Si substrate surface is hydrogen-passivated in an HF etch. The hydrogen is removed by heating the substrate to 500°C in vacuum. A thermal cycle at 800°C ensures the removal of oxide remnants. This is verified by reflection high-energy electron diffraction (RHEED). The substrate temperature is reduced and stabilised at 500°C followed by a five-minute soak in an Sb overpressure. The entire structure is then grown at 500°C. The substrate preparation, nucleation and buffer growth – all crucial to high quality material — are described briefly here, but in detail elsewhere [8]. The AlSb buffer consists of a 100 Å nucleation layer followed by an AlSb(100 Å)/GaSb(100 Å) $\times$ 80 superlattice. The laser structure consists of an $Al_{0.9}Ga_{0.1}Sb$ clad (1.5 µm) and an $Al_{0.3}Ga_{0.7}Sb$ waveguide (0.5 µm), which sandwich the active region. The active region includes three 130 Å $In_{0.2}Ga_{0.8}Sb$ QWs separated by 200 Å $Al_{0.3}Ga_{0.7}Sb$ barriers. Prior to optical pumping, the sample is cleaved into 1 mm $\times$ 0.5 cm bars.

The optical pump source is a modelocked Ti-sapphire laser centred at $\lambda_p = 775$ nm with a laser line width $\Delta\lambda = 7$ nm. The repetition rate is 1 kHz with a pulse width of 200 fs. A cylindrical lens produces a 1 $\times$ 0.2 mm pump spot (in the sample plane) to yield a maximum excitation density of $I_P = 6$ mJ/cm^2. The emission from the pump is linearly polarised and coplanar with both the optical benchtop and the growth direction of the InGaSb/AlSb/Si laser sample. The sample emission is collected from one end facet and analysed for polarisation and spectral content. In these measurements we have accounted for the polarisation response of the spectrometer.

Fig. 1*a–c* show atomic force microscope (AFM) data after 3, 18 and 54 ML of AlSb deposition. At 3 MLs, the QD density is 10^{11} QDs/cm^2 with dot height and diameter of 1–3 and 20 nm, respectively. Fig. 1*b* shows the growth at 18 MLs. The effect of this continued deposition causes the individual islands to coalesce but remain crystallographic, in contrast to InAs/GaAs QD growth where island coalescence leads to large defective islands. Fig. 1*c* shows continued coalescence towards planar growth with 54 ML depositions. The insets show corresponding RHEED patterns at each stage of the nucleation layer growth.

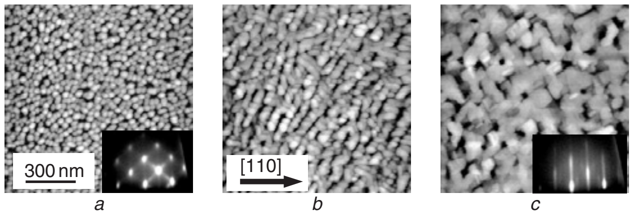

Fig. 1 *AFM images showing surface structure after various MLs of ALSb deposition on Si*
a 3 ML
b 18 MLs
c 54 MLs
Fig. *a* and *c* also show the RHEED image for the corresponding growths

Fig. 2 shows a cross-sectional transmission electron microscope (TEM) image of the (110) crystal plane at the AlSb buffer/Si interface. The image shows the absence of any defects in the AlSb layer (100 nm) under observation. We attribute the defect-free epitaxy to a unique combination of very high strain energy and very strong atomic bonds, which produce the strain-relieving crystal undulations in the AlSb grown on Si. This involves the presence of periodic misfit dislocations at the Si/AlSb interface. The periodic array of misfit dislocations establish alternating regions of strain and relaxation, causing the material to undulate [10]. These undulations increase the surface area of the epimaterial and help to accommodate the 13% mismatch. Owing to the alternate strong and weak bonds, the AlSb/GaSb superlattice helps to smooth the undulations for further heterostructure growth. Both the high-resolution TEM and Fourier transform analyses are discussed elsewhere [8].

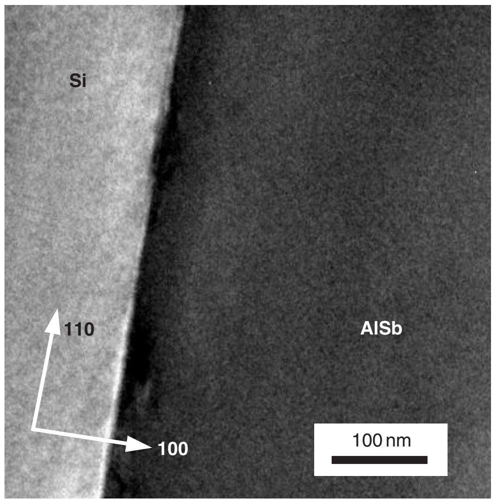

Fig. 2 *Cross-sectional HR-TEM image of (110) plane showing defect-free AlSb on Si*

The RT-OP lasing spectrum is shown in Fig. 3. The plot includes emission spectra at different pump intensities ranging from 0.0001 to 0.63 I_P ($I_P = 6$ mJ/cm^2). The threshold intensity is ~0.002 I_P. The InSb detector used for these measurements is not gated to the pump pulse frequency and measures only an average signal. Below threshold, Fig. 3(i) 0.0001 I_P, the signal is very weak as measured by the DC InSb detector. Above threshold, the emission has a sharp increase in intensity when pumped between 0.005 I_P and 0.01 I_P. The spectrum Fig. 3(ii) 0.01 I_P, peaks at 2.0 µm and appears single-lobed. The full-width at half maximum of the emission spectrum is 45 nm. For pump intensity greater than 0.05 I_P, heating effects result in a slight red shift and a reduced output efficiency. With further increase in pump power, Fig. 3(iii) 0.4 I_P and Fig. 3(iv) 0.63 I_P, the spectral shape changes and the spectrum becomes multi-lobed accompanied by a marginal increase in intensity. Through polarisation analysis of the output beam, we have established that the TE component (perpendicular to growth direction) of the emission increases with pump intensity, while the TM (parallel to the growth direction) component remains clamped. At 0.63 I_P the intensity of the TE component is 30 times that of the TM component.

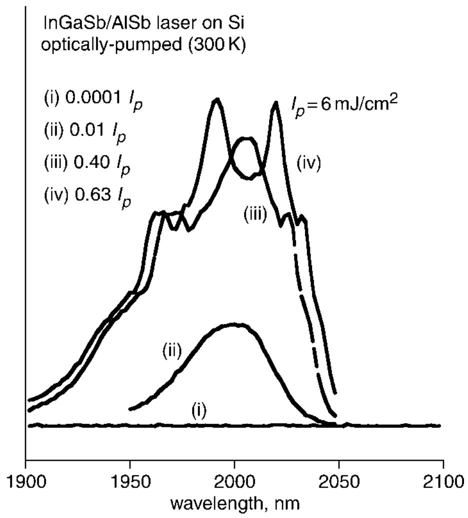

Fig. 3 *RT-OP lasing spectra at pump intensities from 0.0001 to 0.63 I_P*
$I_P = 6$ mJ/cm^2

Conclusion: We have demonstrated a RT-OP InGaSb QW laser on Si(100) with emission at 2 µm. The laser is grown on an AlSb QD nucleation layer that facilitates the 13% mismatch. The laser spectrum is characterised by changes in the spectral shape with increasing pump intensity. Further analyses indicates strong polarisation in direction perpendicular to the growth (TE mode) accompanied by a clamping of intensity in the TM guided modes. Diodes with similar epistructure and growth conditions are under investigation.

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