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EPITAXIAL GROWTH AND CHARACTERIZATION OF GaAs/Al/GaAs HETEROSTRUCTURES

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We report on transmission electron microscopy, secondary ion mass spectroscopy, X-ray and Raman scattering studies of GaAs/Al/GaAs heterostructures grown by migration enhanced epitaxy.

Epitaxial growth of buried metal films in GaAs has recently received much attention partly because of its relevance to the field of interconnects [1–7]. Compared with semiconductor heterostructures, metal–semiconductor epitaxy usually presents more severe difficulties due to large differences in lattice constants, preferred growth temperatures and growth modes coupled with more complex oxidation and interdiffusion properties. Work by Sands et al. [1–3] and, in particular, by Harbison et al. [4] on GaAs–NiAl–GaAs have shown how to overcome some of these problems. In this work, we report on MEE (migration enhanced epitaxy) growth [5] and characterization of GaAs–Al–GaAs structures deposited on (001) GaAs. Data on two samples for which the top GaAs layer was grown at 200 and 400 °C are presented. Results for the 200 °C structure show that the layer on top consists of both GaAs(001) and GaAs(110) and that Al forms in the (111) orientation. In contrast, the 400 °C sample exhibits better structural quality but at the expense of substantial alloying at the Al/GaAs interface.

In conventional molecular beam epitaxy (MBE), growth involves anion-rich conditions and, therefore, surface kinetics is mostly determined by activated cation migration rates. Growth temperatures need to be large enough so as to overcome the corresponding activation barrier, but sufficiently low to prevent entropy-controlled growth and unwanted diffusion. The main difficulty in GaAs/Al heteroepitaxy is the deposition of the semiconductor on the metal layer [1–6]. Since the energies required to break Ga–As and Al–As bonds are much larger than that of the metal–metal bond, growth temperatures appropriate for Al are too low for GaAs growth (typically ~ 550 °C). As soon as the arsenic flux is turned on, the metal layer will tend to form AlAs. In MEE, one grows under *cation* overpressure. To avoid the incorporation of the excess cations in the lattice, the procedure involves first the deposition of a monolayer of cations which is then followed by anion-impingement. Using this method, the GaAs growth temperature can be reduced considerably. This is because migration becomes limited by the strength of the cation–cation bond which is, on average, a factor of two weaker than that of the cation–anion bond. In regard to the problem of lattice mismatch we note that, while both Al and GaAs have an underlying fcc structure, their lattice constants differ by nearly 14%. The situation is far

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better for Al(110) (on GaAs(001)) where the difference is reduced to ~ 1.4%. However, the X-ray results discussed below reveal that the metal grows in a (111) orientation. This is favored energetically by the close packing of atoms.

The GaAs/Al/GaAs structures were grown on GaAs(001) substrates. They consist of a buffer GaAs–$Al_{0.3}Ga_{0.7}As$ superlattice followed by 0.5 μm of GaAs (growth temperature 600 °C), the Al layer deposited at $T \approx 0$ °C and the MEE-grown top layer. MEE was achieved by alternate impingement of the Ga and As fluxes with growth rate of one monolayer every two seconds.

In the two samples considered in this work, the growth temperatures (thicknesses) of the GaAs layer on top were 200 °C (0.3 μm) and 400 °C (0.5 μm); the corresponding nominal thicknesses of the Al-layer were 300 and 100 Å. For reasons discussed elsewhere, a small amount of In ($\leq 1\%$) was incorporated during growth. The structures were studied using transmission electron microscopy (TEM), secondary ion mass spectroscopy (SIMS), X-ray and Raman scattering. TEM cross-sectional micrographs of the 400 °C sample show substantial outdiffussion and balling of the Al layer. In contrast, the 200 °C structure exhibits weak outdiffusion but considerable twinning of the top GaAs layer (see below). The depth profiles of the different atomic species, as obtained by SIMS using Cs-ions, are shown in fig. 1 for the sample in which the top GaAs layer was deposited at 200 °C. We note the presence of a relatively sharp and nearly symmetric Al peak. For the structure grown at 400 °C (not shown), the Al feature is much broader and asymmetric.

X-ray diffraction data on the GaAs–Al–GaAs structures are shown in fig. 2. We employed a combination of techniques including four-circle diffractometry and double crystal rocking curve analysis. The former method gives overall information on the degree of perfection of the layers, their stacking and orientation, while the double-crystal technique is capable of a very high resolution determination of chemical composition variations and the resulting strains. The data in fig. 2 are for a diffraction vector normal to the (001) substrate surface. Immediately, it is clear that the sample with the top layer grown at 400 °C, is of

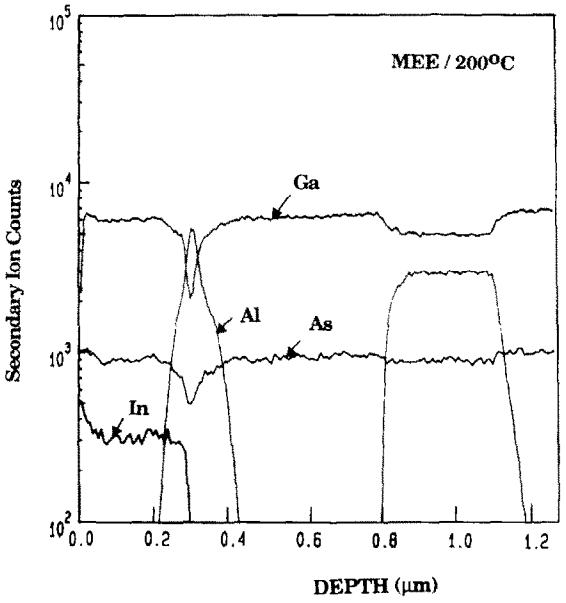

Fig. 1. SIMS profile of the GaAs/Al/GaAs structure (top layer grown at 200 °C).

higher structural quality than the 200 °C structure. In particular, the presence of a significant (220) GaAs peak in the latter sample is evidence

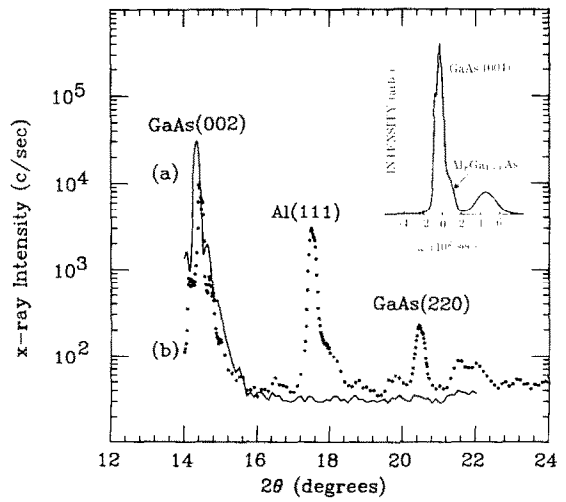

Fig. 2. X-ray diffraction scans of GaAs/Al/GaAs heterostructures with top layer grown at (a) 400 °C (solid line) and (b) 200 °C (dots). Features in the vicinity of (002) are due to the buffer superlattice. Inset: rocking curve of 400 °C sample showing presence of $Al_xGa_{1-x}As$ ($x \approx 0.08$). The peak at $\omega \approx 500$ s is due to In-doping of the top GaAs layer.

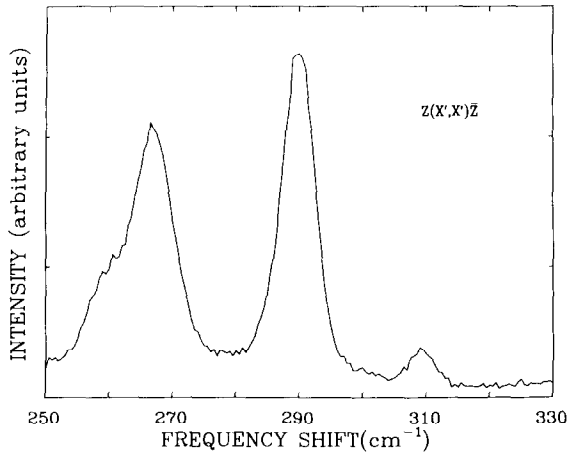

Fig. 3. Raman spectrum of the GaAs/Al/GaAs structure (200 °C top-layer growth temperature) showing phonon scattering.

that twinning occurs in the regrowth of GaAs on top of the Al layer, which we find to be in the (111) close-packed orientation. The 110-twinned regions are found to have very broad rocking curve widths (~ 5° FWHM) indicating poor alignement with the substrate; this is probably a consequence of the lattice mismatch introduced by the (111) Al layer. In contrast, the sample grown at higher temperature shows no evidence of twinning; however there is no X-ray peak from Al in this sample (see fig. 2a) suggesting that alloying has taken place. Our double-crystal results in the inset confirm that this is inded the case.

The Raman spectrum of the 200° C-structure at $T = 300$ K is shown in fig. 3. The data were obtained using the 4880 Å line of an Ar^+-laser in a nearly backscattering configuration. The notation $z(x', x')\bar{z}$ indicates that the incident (scattered) photon propagates along $z \equiv [001]$ ($\bar{z} \equiv [00\bar{1}]$ with polarization $x' \equiv [110]$. Here, the directions refer to the crystal axes of the *substrate*. Our data on this sample show no evidence of alloying effects. The spectral features at 267 and 290 cm^{-1} are due to the bulk transverse (TO) and longitudinal optical (LO) photons of the top GaAs layer (the penetration depth of the light is approximately 0.1 μm). These lines appear in the geometry $z(x', x')\bar{z}$, but not in $z(x', y')\bar{z}$ ($y \equiv [1\bar{1}0]$). In III–V compounds, backscattering from (001) surfaces allows only the LO mode while, for (110), only the TO phonon is allowed. Therefore, the observation of both TO and LO peaks is consistent with the X-ray findings showing both (001) and (110) orientations for the top layer. Moreover, the observed selection rules indicate that the GaAs substrate and the top layer share a common [110] axis. Other than the two optical modes, the spectrum shows weaker features at 260 and 310 cm^{-1}. Unlike the TO and LO phonons, these peaks also appear in the $z(x', y')\bar{z}$ configuration and, furthermore, their intensity and lineshape depend on the angle between the directions of the incident and scattered beams. This and the positions of the peaks suggest that they are due to bulk polaritons; however such excitations are nominally forbidden in backscattering. The structure for which the top GaAs layer was grown at 400 °C also shows the extra lines in the Raman spectrum, but no TO mode. The latter fact indicates that the top layer orientation is (001), in agreement with the X-ray results.

In summary, we have explored the use of MEE for the growth of Al buried-layer heterostructures in GaAs. The structural character of the samples is determined by a complex balance between interdiffusion, alloy formation and the kinetics of metal-semiconductor epitaxy. Further experiments are required to identify optimal growth conditions to suppress twinning.

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References

[1] T. Sands, Appl. Phys. Lett. 52 (1988) 197.
[2] T. Sands, V.G. Keramidas, J. Washburn and R. Gronsky, Appl. Phys. Lett. 48 (1986) 402.
[3] T. Sands, J.P. Harbison, W.K. Chan, S.A. Schwarz, C.C. Chang, C.J. Palmström and V.G. Keramidas, Appl. Phys. Lett. 52 (1988) 1216.
[4] J.P. Harbison, T. Sands, N. Tabatabaie, W.K. Chan, L.T. Florez and V.G. Keramidas, Appl. Phys. Lett. 53 (1988) 1717.

[5] Y. Horikoshi, M. Kawashime and H. Yamaguchi, Jpn. J. Appl. Phys. 25 (1986) L868.
[6] B. Tadayon, S. Tadayon, M.G. Spencer, G.L. Harris, L. Rathbun, J.T. Bradshaw, W.J. Schaff, P.J. Tasker and L.F. Eastman, Appl. Phys. Lett. 53 (1988) 2664.
[7] For early work on GaAs–Al–GaAs, see: R. Ludeke, L.L. Chang and L. Esaki, Appl. Phys. Lett. 23 (1973) 201; K. Okamoto, C.E.C. Wood, L. Rathbun and L.F. Eastman, J. Appl. Phys. 53 (1982) 1532.