are bias dependent as shown in Figs. 5 and 6, respectively. $R_P = 0{\cdot}5\,\Omega$, $L_p = 0{\cdot}3$ nH and $C_P = 0{\cdot}2$ pF do not depend on applied voltage and are attributed to parasitic elements associated with the mounting of the device in the package.

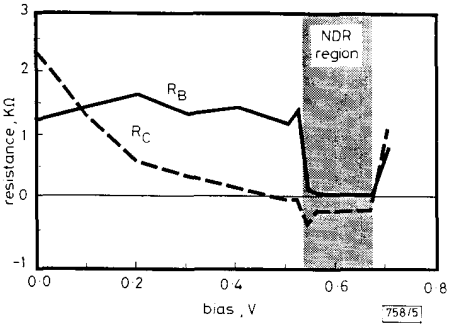


**Fig. 5** *Variation of resistors $R_B$ and $R_C$ with applied bias*

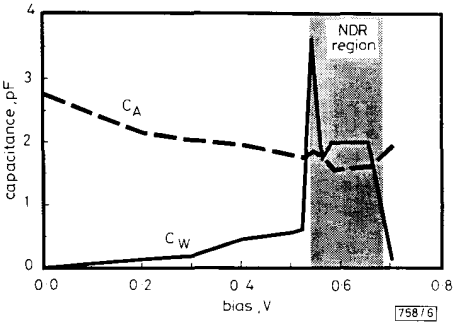


**Fig. 6** *Variation of capacitors $C_A$ and $C_W$ with applied bias*

are bias dependent as shown in Figs. 5 and 6, respectively. $R_P \ominus 0{\cdot}5\,\Omega$, $L_p = 0{\cdot}3$ nH and $C_P = 0{\cdot}2$ pF do not depend on applied voltage are attributed to parasitic elements associated with the mounting of the device in the package.

An analysis of the equivalent circuit shows that the cutoff frequency $f_T$ is about 3 GHz and the devices when operated as oscillators were found to produce about 1 $\mu$W of power up to 2·8 GHz.

*Equivalent circuit component behaviour:* $R_C$ behaves as expected taking on negative values in the region just above resonant tunnelling. Similarly $C_A$ shows a trend typical of a depletion layer capacitance. The magnitude of $C_W$ rises rapidly as the DC current approaches its peak value. This is attributed to charge transfer into and out of the quantum well near the resonant peak and the overall potential distribution remains largely unchanged, being determined by the charge in the depletion and accumulation layers set by DC conditions. $R_B$ is nearly constant up to the NDR region and then falls to a value below 100 Ω. In this region, under small-signal excitation, an increase in the applied voltage moves the device away from resonance and results in electrons moving out of the well into the accumulation and depletion regions. The energy levels in the emitter are thus raised relative to those in the quantum well enhancing the current across the barrier. The net effect is a lowering in the differential resistance of the barrier in the NDR region.

*Conclusion:* Accurate electrical characterisation of a resonant tunnelling diode has yielded a small signal equivalent circuit which is valid over a wide frequency band and whose elements can be identified with the structure and processes occurring in the device. The model exhibits good agreement with experimental results, both in a stabilised reflection amplifier circuit and as an oscillator.

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# INVESTIGATION OF INDIUM DOPING AND INCORPORATION IN AlGaAs/GaAs DOUBLE-BARRIER RESONANT TUNNELLING STRUCTURES

*Indexing terms: Semiconductor devices, Tunnelling structures*

In doping and incorporation in the barrier layers of AlGaAs/GaAs double-barrier resonant tunnelling structures (DBRTSs) have been studied. It was found that the peak-to-valley current ratio (PVCR) can be improved by the proper amount of In doping. This is attributed to the improvement in the quality of the AlGaAs barrier layers due to the high surface migration rate of In atoms that reduces group III vacancies. Also pseudomorphic $In_x(Al_{0.5}Ga_{0.5})_{1-x}As/GaAs$ ($x = 0{\cdot}12$) strained-layer DBRTSs have been fabricated by incorporating a sufficient amount of In into the AlGaAs barrier layers. PVCRs as high as 27·5 at 77 K have been obtained. This is the first realisation of such DBRTSs with lattice-mismatched quaternary barrier layers.

Double-barrier resonant tunnelling structures (DBRTSs) have attracted considerable attention in recent years because of their potential applications and interesting transport properties. It is well known that the performance of the negative differential resistance (NDR) of a DBRTS strongly depends on the quality of the active region, consisting of the quantum well and the barrier layers. For the DBRTSs using AlGaAs as the barrier layers grown by molecular beam epitaxy (MBE), because of the low surface migration rate of Al atoms during growth (especially at the substrate temperature suitable for the GaAs growth), vacancies are easily formed in the AlGaAs barrier layers. Recently, we have shown that the proper amount of In doping can improve the quality of MBE-grown AlGaAs, and this improvement was attributed to the reduction in the vacancy concentration of group III atoms due to the high surface migration rate of In atoms.[1] We report the effect of In doping in the AlGaAs barrier layers of AlGaAs/GaAs DBRTSs on the NDR performance. InAlGaAs/GaAs strained-layer DBRTSs, obtained by incorporating a sufficient amount of In into the AlGaAs barrier layers, are also reported for the first time.

The devices in this work were grown by MBE in a Varian Gen II system on (100)-oriented $n^+$-GaAs substrates at 580°C. First, a 0·3 $\mu$m $n^+$-GaAs ($n = 2 \times 10^{18}\,cm^{-3}$) buffer layer was

grown on the substrate, followed by a 0·2 $\mu$m undoped GaAs layer for the electrode. The tunnelling structure was then grown, which was composed of a 50 Å undoped GaAs quantum well sandwiched between two 90 Å undoped $Al_{0.4}Ga_{0.6}As$ barrier layers. The barrier layers were doped with In from 0 to $1{\cdot}2 \times 10^{20}\,cm^{-3}$. The In concentration was determined from the logarithmic plot of the In growth rate against the reciprocal temperature of the In furnace. A strained-layer DBRTS with 90 Å quaternary $In_{0.12}Al_{0.44}Ga_{0.44}As$ barrier layers was also grown. On top of the tunnelling structure, a 0·2 $\mu$m undoped GaAs layer was grown for the other electrode, followed by a 0·3 $\mu$m $n^+$-GaAs ($n = 2 \times 10^{18}\,cm^{-3}$) layer. The growth ended with a 500 Å $n^+$-GaAs ($n = 5 \times 10^{18}\,cm^{-3}$) layer for ohmic contact.

After growth, AuGe was applied to the top side of the sample by the liftoff technique and then alloyed for $n$-type ohmic contact. Circular devices with 120 $\mu$m diameter were defined with conventional photolithography and mesa etching. The bottom ohmic contact was provided by In which had been used for holding the substrate during growth. The samples were then diced and bonded for testing.

Fig. 1 shows the current-voltage (I/V) characteristics of the AlGaAs/GaAs DBRTS with an In doping concentration of $4 \times 10^{19}\,cm^{-3}$. The peak-to-valley current ratio (PVCR) at

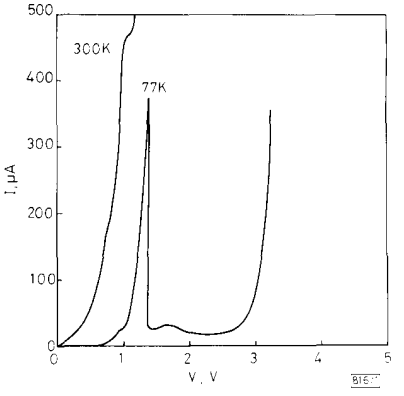


**Fig. 1** *I/V characteristics of AlGaAs/GaAs DBRTS with In doping concentration of* $4 \times 10^{19}\,cm^{-3}$

77 K is about 22·2, a very high value among AlGaAs/GaAs DBRTSs. At 77 K, the bump in the valley region of the NDR region has been attributed by Goldman *et al.*[2] to the LO-phonon-emission-assisted tunnelling of electrons from the emitter electrode to the quantum well. The kink at about 1 V is due to the resonant tunnelling of electrons from the first excited quantum level in the accumulation layer that is caused by the large band bending across the undoped electrodes.[3] Because of the thick barriers used in our structures, the tunnelling current at 300 K is small as compared to the thermionic emission current over the barriers. Therefore the NDR (and even the tunnelling-related features) at 300 K is not very clear in the I/V curve. The I/V characteristics of other DBRTSs with or without different In doping concentrations are similar to that shown in Fig. 1.

Fig. 2 shows the PVCR at 77 K as a function of the In concentration. We can see that a proper amount of In doping increases the PVCR. The PVCR reaches the maximum as the In concentration is $4 \times 10^{19}\,cm^{-3}$. The improvement in the NDR performance with In doping is due to the In atoms being able to occupy the Al or Ga vacancies owing to their high surface migration rate[1] and thus reduce the defect- or dislocation-related scatterings. This results in a better resonant tunnelling process. However, when the In concentration is too high, the NDR performance degrades. A probable cause is the serious incorporation of additional impurities associated with the In source,[1] which cancels the positive effect of the In doping on the NDR performance.

When a sufficient amount of In is incorporated into the AlGaAs barrier layers, lattice-mismatched quaternary InAlGaAs epitaxial layers are formed. Fig. 3 shows the I/V characteristics of the strained-layer DBRTS with 90 Å $In_{0.12}Al_{0.44}Ga_{0.44}As$ barrier layers. The PVCR at 77 K reaches 27·5, higher than those of the above DBRTSs with

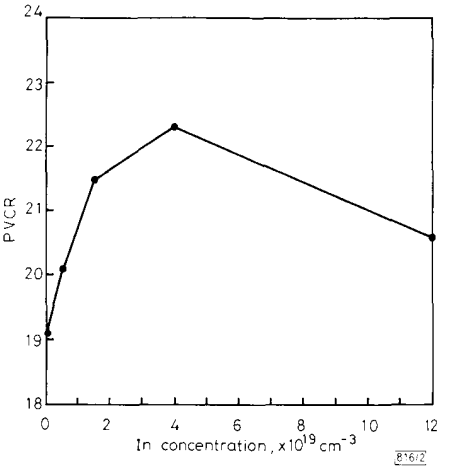


**Fig. 2** *PVCR at 77 K as function of In doping concentration*

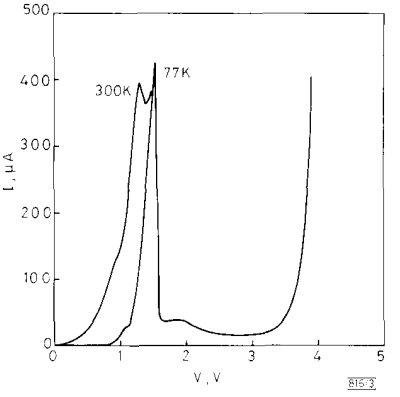


**Fig. 3** *I/V characteristics of strained-layer DBRTS with 90 Å quaternary* $In_{0.12}Al_{0.44}Ga_{0.44}As$ *barrier layers*

In-doped AlGaAs barrier layers. This indicates the good quality of the pseudomorphic $In_{0.12}Al_{0.44}Ga_{0.44}As$ epitaxial layers. Because the lattice constant of AlGaAs is very close to that of GaAs, the lattice constant of $In_{0.12}Al_{0.44}Ga_{0.44}As$ should be approximately equal to that of $In_{0.12}Ga_{0.88}As$. For $In_{0.12}Ga_{0.88}As$, its critical layer thickness[4] is much larger than the thickness of the present $In_{0.12}Al_{0.44}Ga_{0.44}As$ barrier layer. Therefore, the lattice mis-match is expected to be taken up by the elastic strain in $In_{0.12}Al_{0.44}Ga_{0.44}As$ epitaxial layers and does not degrade the NDR performance. On the other hand, the energy gap of $In_{0.12}Al_{0.44}Ga_{0.44}As$ is is about 1·93 V (obtained from Reference 5), very close to that of $Al_{0.4}Ga_{0.6}As$ (1·92 V). If the conduction-band offset fraction of the $In_{0.12}Al_{0.44}Ga_{0.44}As/GaAs$ heterointerface is assumed to be equal to that of the $Al_{0.4}Ga_{0.6}As/GaAs$ heterointerface, the barrier height in the $In_{0.12}Al_{0.44}Ga_{0.44}As/GaAs$ DBRTS is very close to that in the $Al_{0.4}Ga_{0.6}As/GaAs$ DBRTS. Therefore, the resonant tunnelling current levels of the DBRTSs with $Al_{0.4}Ga_{0.6}As$ or $In_{0.12}Al_{0.44}Ga_{0.44}As$ barrier layers should not differ too much, as shown in Figs. 1 and 3.

In summary, the effect of In doping in the barrier layers on the NDR performance of AlGaAs/GaAs DBRTSs has been investigated. It is found that a proper amount of In doping improves the PVCR. This is attributed to the reduction of group III vacancies due to the high surface migration rate of In atoms. Also we have fabricated pseudomorphic InAlGaAs/GaAs strained-layer DBRTSs for the first time. For the In content of 0·12, the PVCR of 27·5 at 77 K has been obtained.

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## PREPARATION AND GROWTH OF THIN FILM $Pb_2Sr_2Ca_{0.5}Y_{0.5}Cu_3O_{8+\delta}$ SUPERCONDUCTORS BY LASER ABLATION

*Indexing terms: Superconducting devices, Superconductors*

For the first time, the preparation of thin films of superconducting $Pb_2Sr_2Y_{0.5}Ca_{0.5}Cu_3O_{8+\delta}$ material using standard bulk target preparation conditions and the laser ablation technique is reported. In common with much of the bulk characteristics, it is found that the width of the superconducting transition is large, extending from $T_{c,onset}$ between 70–83 K and $T_{c,zero}$ between 16–20 K.

Since the discovery of high-transition temperature superconductivity in La–Ba–Cu–O,[1] progress has continued at a rapid rate in designing new materials with and without copper oxide layers. At present, four families of Cu-oxide based high $T_c$ superconductors have been identified. These include the $(La, M)_2CuO_4$, the $LnBa_2Cu_3O_7$, the $(Tl, Bi)_m(Ba, Sr)_2CaL_{n+1}Cu_nO_{m+2n+2}$ and the $Pb_2Sr_2ACu_3O_{8+\delta}$ series, where M can be a substitutional metal cation, Ln a Lanthanide, and A a Lanthanide Ln, Y, or a mixture of (Ln or Y) and either (Sr or Ca). Cava *et al.*[2] have shown that one particular composition of the latter, $Pb_2Sr_2Y_{0.5}Ca_{0.5}Cu_3O_8$, exhibits superconductivity around 68 K, and suggests that many possible avenues may be explored to optimise the superconducting properties. Up to now, however, very little work has been reported in this system. We describe the preparation of bulk ceramic targets of this material, and proceed to demonstrate for the first time the growth of thin films using the laser ablation technique.

The crystal structure of $Pb_2Sr_2Ca_{0.5}Y_{0.5}Cu_3O_8$ is characterised by double $CuO_2$ pyramidal layers and a $c$-axis layer sequence of $-A-CuO_2-SrO-PbO-CuO_\delta - PbO-SrO-CuO_2-$. It can be related to the $YBa_2Cu_3O_{7-\delta}$ system by considering the $PbO-CuO_\delta-PbO$ layers in place of the $CuO_{1-\delta}$ layer. It is the high superconducting transition temperatures of the double $CuO_2$ pyramidal structures that have attracted very much attention in recent years. However, for $Pb_2Sr_2Ca_{0.5}Y_{0.5}Cu_3O_8$, the electrical qualities reported have generally compared rather poorly with those of the other double pyramid families, and superconducting widths around 40 K have been typical.[2] Work is in progress to optimise the preparation of ceramic samples of this system, and currently the best reported values of $T_{c,zero}$ are around 75 K.[3] There have been no reported studies of the preparation of thin films of $Pb_2Sr_2Ca_{0.5}Y_{0.5}Cu_3O_8$. It is important, however, not only for possible future applications to be able to prepare such films, but also for fundamental studies of the structural and electrical properties of the system.

Bulk samples of $Pb_2Sr_2Y_{0.5}Ca_{0.5}Cu_3O_8$, to be subsequently employed as targets for laser ablation, were fabricated using a two-step solid-state reaction related to that reported by Cava *et al.*[2] First, the precursor was prepared by mixing and grinding an appropriate mixture of the oxides and carbonates ($CaCO_3$, $SrCO_3$, $Y_2O_3$, and CuO), which was then pelletised and heated at 960°C for 16 hours in air with intermediate grinding. The second step involved the introduction of the PbO, which was accomplished in this case by grinding the precursor yet again and adding PbO power to the mixture. This was then mixed and ground thoroughly for 30 minutes, pelletised into a 13 mm die by using 10 ton pressure, and heated at 864°C for 7 hours in a mixture of 99% $N_2$ and 1% $O_2$ flowing at the rate of 12 cc/minute. After 7 hours the pellet was cooled in the gas stream in 10 minutes. It was observed at this stage that the pellet had expanded by approximately 1·2 mm across its diameter and had become very fragile. To ensure a homogeneous reaction between the PbO and the precursor, the sample was further reground and after a final pelletising step, was sintered at 864°C for 18 hours, using the same gas flow conditions as for the heating stage described above. Fig. 1 shows the DC resistivity of a typical bulk pellet prepared in this way. In common with other published results, the value of $T_{c,onset}$ is around 83 K, while the transition remains relatively wide, with $T_{c,zero}$ at 43 K.

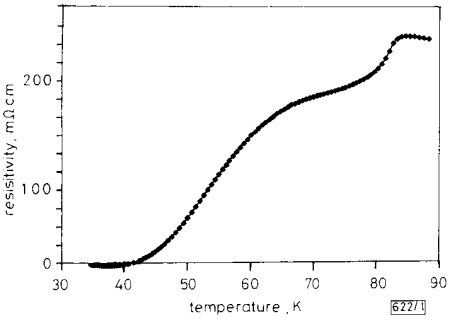


**Fig. 1** *Resistivity property of target*

We have found that the ultimate characteristics of the material are extremely sensitive to some of the preparation conditions. For example, when the sintering temperature was 855°C (instead of 864°C) under otherwise identical gas flow conditions, the transition temperature, $T_{c,onset}$ was no longer found to be at 83 K but at 70 K, whereas $T_{c,zero}$ did not even reach 20 K. When this pellet was further sintered in Ar for 6 hours at 855°C, the $T_{c,onset}$ transition shifted from 70 K to 40 K and the material became highly resistive. Further annealing in air for one hour at 855°C made the pellet completely semiconducting.

This transition, however, could be reversed. For example, when the semiconducting pellet was resintered again in the same $N_2/O_2$ gas mixture (flowing at 6 cc/min) for 36 hours at 855°C, the $T_{c,onset}$ was regained at 70 K, with a lower value of resistivity, but $T_{c,zero}$ was below 15 K. Further sintering for 12 hours at 855°C under identical conditions produced the same $T_{c,onset}$ at 70 K and $T_{c,zero}$ was increased to 20 K. This shows that the re-establishment of superconductivity is possible by heat treatment at the same temperature by choosing the correct gaseous mixture. We have found that the time required for this transformation is inversely proportional to the flow rates used. When the temperature was increased by 5°C to 860°C, and the sample resintered for 35 hours under