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Doping effects in AlGaAs

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A summary of the results of a few experiments on a variety of structures based on GaAs–AlGaAs heterojunctions, in which the AlGaAs was Si doped are reported. RHEED and SIMS results are presented for heavily and moderately doped AlGaAs, showing that Si segregated toward the growing front. Photoluminescence spectra of heavily doped AlGaAs has a dominant low energy peak and is absent of the exciton peak. Normal and inverted selectively doped structures were grown at a variety of conditions. Their properties demonstrate that Si segregates toward the growing front, and suggests that this effect could be the main cause of the low mobility in inverted structures.

I. INTRODUCTION

Si is the most widely used donor in GaAs and AlGaAs due to its low amphotericity and presumed lack of segregation, but its incorporation in AlGaAs is not well understood. For example, the persistent photoconductivity effects in Si doped AlGaAs (and for other dopants too) is widely recognized, and generally attributed to the DX centers formed by the Si doping,[1] but a thorough understanding of the phenomenon is still lacking. Another example is the controversy around the explanation of the change in the electron concentration with changing Al mole fraction. It was believed[2] that the donor level gets deeper with increasing the Al mole fraction (for $0.2 < x < 0.45$), but new evidence points out that there are two donor levels: one shallow and one deep (D-X centers). Both are constant in energy with changing the Al mole fraction, but the net effect of free carrier reduction is a result of cross over between the D-X centers energy and the Γ band.[3]

So far, there have been no reports concerning the segregation of Si in GaAs and AlGaAs. Recently, we published on the segregation of Si in very heavily doped GaAs and AlGaAs[4]. Here, we present some more data on the RHEED and PL of heavily doped AlGaAs. C-V and Hall measurements on bulk AlGaAs will be compared, and then the results of transport measurements parallel to GaAs-AlGaAs heterojunction interfaces will be discussed. These results strongly suggest the movement and accumulation of Si with the growing front of AlGaAs, even for moderate doping.

II. HEAVY DOPING IN AlGaAs

We have attempted to heavily dope AlGaAs with Si. To accomplish this without raising the Si temperature above 1200 °C (to prevent CO and N evolution from the cell), the growth rate was slowed down to $\sim 0.14 \mu\text{m h}^{-1}$. In a similar fashion to GaAs,[4] the RHEED pattern converted to metal rich, and a flux ratio of $F(As_4)/F(Ga) \approx 3$ was necessary to maintain a (2×4) As stable pattern, and a smooth surface morphology (note, that in heavy doping of GaAs, a flux ratio of at least 5 was necessary). Since the deterioration of the RHEED pattern was gradual (occurred over a 100–300 Å layer growth, and faster for higher Si temperature or higher Al mole fraction), I believe it is a result of Si accumulation on the surface, as the SIMS results of Fig. 2 clearly show, and may be a formation of some Si–As compound, which leaves the surface As deficient. If extra As was not supplied, the doping level dropped and the morphology deteriorated. The RHEED patterns are very similar to those published in Ref. 4.

The photoluminescence of films with Si incorporation $n^2_{Si} \gtrsim 2\times10^{19}$ cm^{-3}, exhibits in Fig. 1 the collapse of the excitonic emission. The main peak at the energy gap position which is evident for $n^1_{Si} < 10^{19}$ cm^{-3} disappears and a broad peak at a lower energy, with a long low energy tail is dominant. It is believed that this peak is a result of donor Si, complexing with acceptor Si and Ga vacancies, based on the configurational coordinate model.[5]

Two typical SIMS results are shown in Fig. 2 demonstrating that Si piles up considerably near the surface, confirming the RHEED results. The structures are composed of AlGaAs layer on top of GaAs buffer with a 1000–2000 Å of an undoped AlGaAs at the bottom. Au Schottky barriers

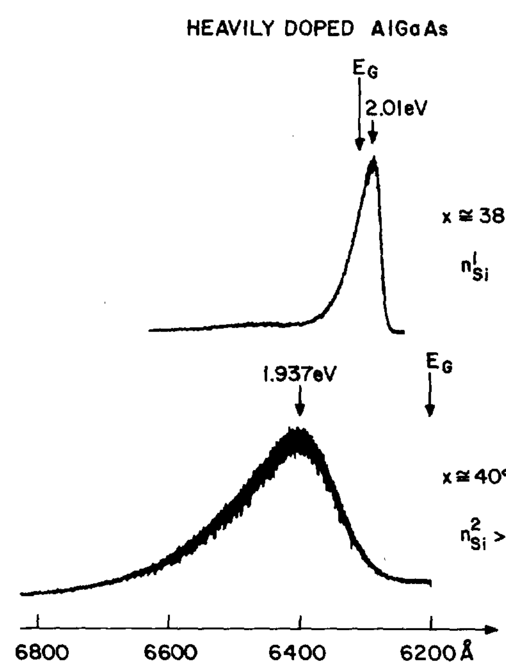

FIG. 1. At the top, a 4 KPL spectrum of a heavily doped AlGaAs, but with $n^1_{Si} < 10^{19}$ cm^{-3}. The bottom is a spectrum of AlGaAs with $n^2_{Si} > 10^{19}$ cm^{-3}. The collapse of the excitonic structure is clearly evident. The Al mole fraction is determined by microprobe.

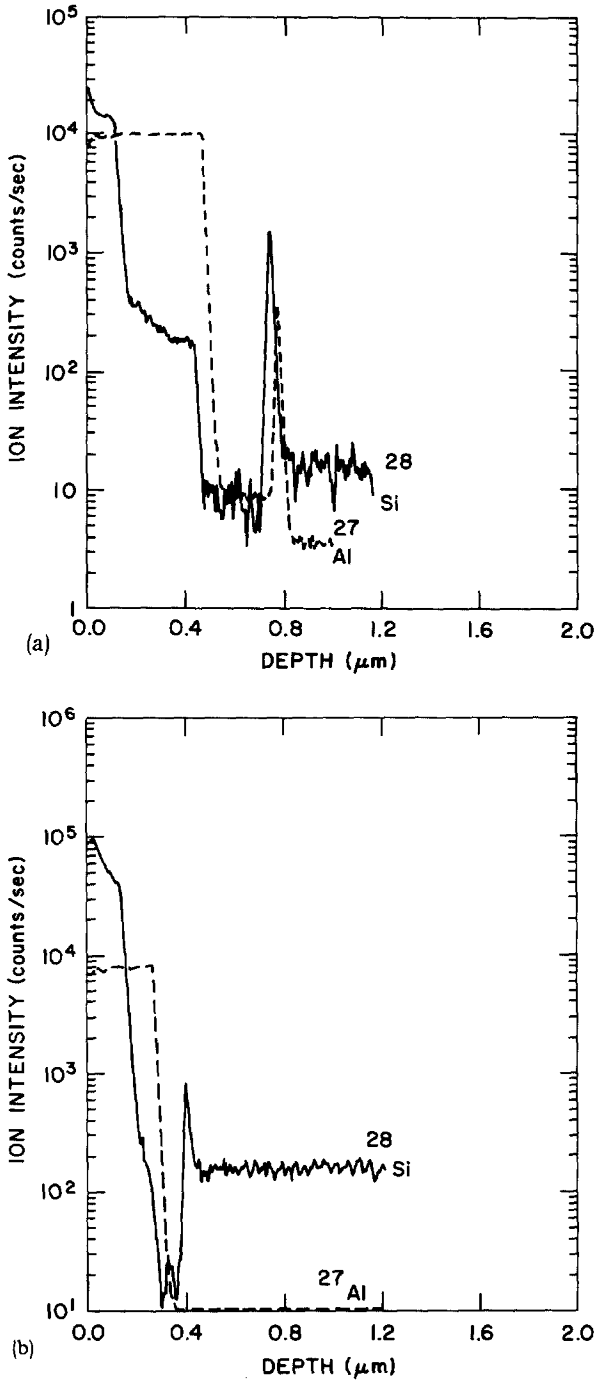

FIG. 2. Two examples of SIMS results of heavily doped AlGaAs. Each structure is undoped GaAs buffer-undoped AlGaAs–Si doped AlGaAs. The initial spikes occurs at the interface of substrate GaAs buffer due to contaminations, and the initial rise of the Si signal in the AlGaAs is due to a higher yield of the secondary ion ejection from AlGaAs (compared with GaAs). (a) AlGaAs grown at $\sim 1.4\,\mu$m h^{-1} with $x \approx 0.25$, and (b) AlGaAs grown at $\sim 0.14\,\mu$m h^{-1} with $x \approx 0.25$, but heavier doping.

formed on surfaces of those heavily doped layers, had very large leakage currents with a near linear I–V, suggesting that tunneling is the dominant transfer mechanism, resulting from the high Si doping level.

III. C-V AND HALL MEASUREMENTS

AlGaAs layers were grown using growth rates of 0.14 and 1.4 μm h^{-1} at 600 °C, on an undoped GaAs buffer. The doped AlGaAs layers were grown on top of an undoped AlGaAs to prevent electron transfer to the GaAs buffer and the creation of a quasi two dimensional electron gas (2DEG) there. Layers were then checked by C-V (Hg probe) and Hall measurements (van der Pauw method).

Consistently the C-V measurements exhibited a maximum doping level which was more than ten times higher than the electron concentration determined by the Hall measurements. Moreover, in all layers, doping near the surface was higher than the doping inside the layer, as Fig. 3 demonstrates.

The difference in the doping levels determined by both methods is easily understood if one invokes deep donor levels (those are filled D-X centers, ~100 meV deep[1]), which get ionized by the application of the negative potential and contribute to the total amount of charges in the depletion layer. On the other hand, the Hall measurements lead only to the number of free electrons in the layer. The reduction of the total number of carriers into the depth of the layer could be explained by Si segregation toward the growing front of the material. Note that at 77 °K, the conductive layers, 2000–4000 Å thick, became semi-insulating due to freeze out of all carriers.

IV. "NORMAL" AND "INVERTED" SELECTIVELY DOPED HETEROJUNCTIONS

"Normal" selectively doped heterojunctions are structures where doped AlGaAs is grown on top of on undoped AlGaAs (spacer), which in turn is grown on top of an undoped GaAs. A quasi 2DEG is formed in the GaAs side of the junction interface.

Measurements of the mobility and carrier concentration in a series of structures with different spacers were performed.[7] For all spacers, the areal carrier density was smaller, and for the smallest three spacers the mobilities were higher than the theoretically predicted. Even though the AlGaAs was doped to a doping level of $\leqslant 10^{18}$ cm^{-3}, an

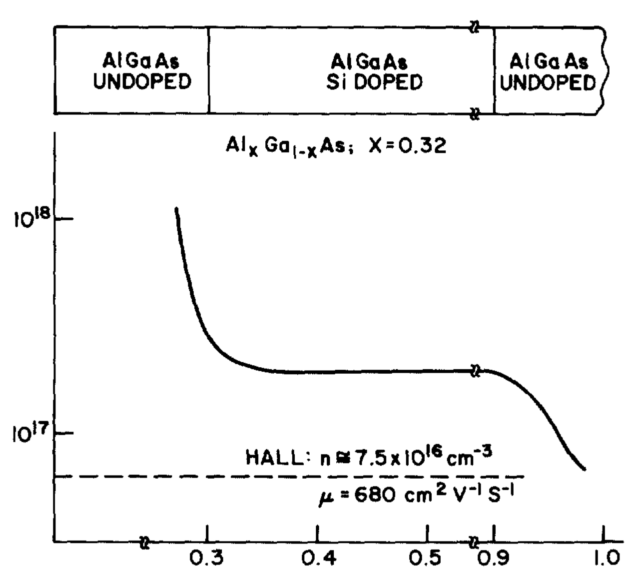

FIG. 3. Typical C-V profile of the structure shown, at the top. The room temperature Hall results are shown at the bottom.

agreement between theory and measurement could be roughly obtained only if the donor concentrations near the spacers were chosen to be $\sim 10^{17}$ cm^{-3}, and the donor level depth 150 meV below the conduction band. Even for those parameters, the high mobilities were difficult to explain.

For comparison, "inverted" heterojunctions were grown at 600 and 700 °C. In those structures the AlGaAs is grown first with a large spacer on the bottom (>600 Å) and a smaller spacer on top (100–200 Å). The doped region is 150–300 Å wide, with doping $\sim 8\times10^{17}$cm^{-3}. This layer is followed by ~ 2000 Å of undoped GaAs and capped by an ~ 150 Å of $\sim 1\times10^{18}$cm^{-3} doped GaAs layer.

When normal and inverted structures were compared, consistently, the inverted ones had higher areal electron densities, for the same spacers at 300 and 77 K ($\sim 1\times10^{12}$ cm^{-2}). By C-V profiling of the inverted structures, it was verified that the cap layer is completely depleted and the doped AlGaAs layer freezes out at 77 K. Aside from the greater carrier concentration of the 2DEG formed in the inverted structures, the mobilities were much lower. The maximum 77 K mobility was $\sim 40\,000$ cm^2/V sec for Al mole fraction of 0.25–0.35. Even though the reason for the low mobility of the 2DEG is still under controversy, I believe that Si segregation plays an important role in: (a) effectively increasing (decreasing) the spacer in normal (inverted) heterojunctions and (b) decreasing the mobility of the 2DEG in the "inverted" structures by moving toward, and maybe into the GaAs top layer. The SIMS data of Fig. 4 give some indication for Si outdiffusion from the AlGaAs layer, but the resolution is not good enough to be conclusive. If the Si segregation is responsible for the poor performance of the inverted structures, the introduction of a series of wells in the top spacer layer, as reported before,[6] could reduce it significantly.

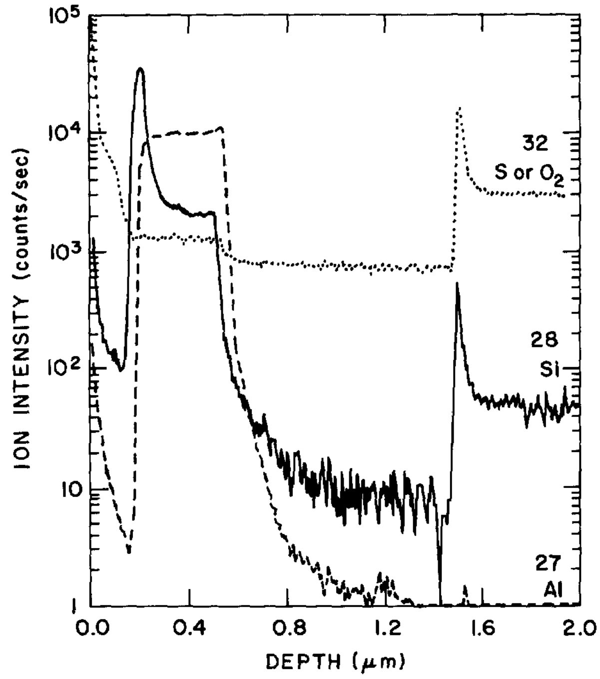

FIG. 4. A SIMS profile of an inverted structure. The Si signal rises when the Al turned on due to the higher yield. After ~ 1000 Å of a back spacer the Si signal rises up due to the opening of the Si shutter, but it fails to decrease until the GaAs layer is grown, even though a top spacer of ~ 200 Å is introduced. The tails toward the surface of the Al and Si signals are due to surface contaminations.

Inverted structures with Al mole fraction of $<10\%$ were grown, in order to reduce the Si segregation. To improve the interface smoothness, a growth temperature of 700 °C was chosen. 77 K low field mobilities as high as 71 000 cm^2/V s were measured with carrier concentration of 3×10^{11}cm^{-2} in the dark, for a spacer of ~ 200 Å wide. The usefulness of those structures is questionable due to the small AlGaAs barriers.

Note the importance of the As flux during growth. When excess As was used during the growth of the spacer region of the inverted structures, the areal carrier densities dropped considerably, but the mobilities didn't improve. This would suggest the creation of traps in the AlGaAs, but not the prevention of Si segregation.

In order to prevent Si movement, another approach was taken and inverted structures were grown at 500 °C with Al mole fraction of $\sim 30\%$ and an undoped spacer of 200 Å. A growth rate of 0.1 μm/h was chosen, to allow reasonable quality AlGaAs.[7] Carrier densities of 0.6–2×10^{11} cm^{-2} with 77 K dark mobilities as high as 50 000 cm^2/V s (and up to 86 000 cm^2/V s in the light) were measured. At 4 K dark mobilities as high as 10^5 cm^2/V s had been measured. Even though those mobilities are still not as high as those measured in the normal structures, the striking reduction of the electron density in the 2DEG channel and the higher mobilities, reconfirm once again the segregation of Si.

V. CONCLUSIONS

I have described results of experiments, where AlGaAs was Si doped. They suggest that the Si segregates in the AlGaAs toward the growing front. In selectively doped structures, Si segregation will enhance mobility in normal structures and degrade it in inverted structures. By growing inverted structures at 500 °C, 77 K mobilities as high as 86 000 cm^2/V s had been achieved.

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