RESEARCH ARTICLE | JUNE 01 1984

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Appl. Phys. Lett. 44, 1064–1066 (1984)
https://doi.org/10.1063/1.94644

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High mobility electron gas in selectively doped *n*:AlGaAs/GaAs heterojunctions

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(Received 6 February 1984; accepted for publication 9 March 1984)

Selectively doped *n*:AlGaAs/GaAs heterojunctions have been grown by molecular beam epitaxy, employing a slow growth technique, at a substrate temperature of 600 °C. The effect of the undoped AlGaAs spacer thickness on carrier density and Hall mobility was investigated. Mobilities as high as 9200, 200 000, and 1 060 000 cm^2/Vs at 300, 77, and 4.2 K, respectively, were measured in the dark for a spacer thickness of ~ 180 Å and an areal carrier density of $\sim 2.2\times 10^{11}$ cm^{-2}. Surprisingly, samples with spacer thicknesses of 80 Å had 4-K mobilities of $\sim 800\,000$ cm^2/Vs, higher than expected theoretically from the structural parameters.

PACS numbers: 68.55. + b, 81.15.Ef, 73.40.Lq

The proposal and practical realization of modulation doping in semiconductor heterojunctions[1] has created new fields of research in both basic and applied solid state physics. Those led to the discovery of the fractional quantum Hall effect,[2] and new device technology like the high electron mobility transistor.[3]

The most common version of modulation doping is a selectively doped, single heterojunction. The material with the lower electron affinity (in our case, AlGaAs) is doped *n* type, while the neighboring material (GaAs) is grown as pure as possible. Free electrons which are liberated from their parent donors in the AlGaAs will reside in the GaAs, in the vicinity of the interface, forming a quasi-two-dimensional electron gas. In the direction parallel to the interface, these electrons may have very high mobility at low temperatures due to the spatial separation from the ionized donors.

Witkowski *et al.*[4] first introduced an undoped AlGaAs spacer between the doped AlGaAs and the undoped GaAs, achieving in effect a larger separation between the ionized donors and the two-dimensional electron gas, which resulted in a higher mobility but a smaller carrier concentration in the channel. While the room-temperature mobility is limited by phonon scattering (~ 9000 cm^2/Vs), the low-temperature (~ 4 K) mobility is influenced mostly by impurities in the unintentionally doped GaAs and in the doped AlGaAs, and to a lesser extent by residual impurities in the spacer.

We have grown selectively doped structures utilizing a slow growth procedure for the AlGaAs,[5] paying special attention to structure reproducibility. GaAs was grown at 600 °C at a rate of 1 μm/h. Unintentionally doped material is *p* type and has $\sim 1\times 10^{14}$ cm^{-3} free holes, resulting mostly from carbon acceptors. Lightly Si-doped material (with $\sim 2\times 10^{14}$ cm^{-3} free electrons) has a Hall mobility in excess of 140 000 cm^2/Vs at 77 K. The photoluminescence spectrum is dominated by a free exciton with a small carbon-related peak and can be free of defect excitons.[5]

As already known, we had verified[5] that AlGaAs grown at 600 °C at a rate of ~ 1 μm/h is not homogeneous, has a broad exciton line (~ 40 meV wide), and poor surface morphology, while AlGaAs grown at 700 °C is much superior. If AlGaAs is grown at 600 °C, but at a growth rate of ~ 0.1 μm/h, the exciton linewidth and intensity are comparable to those for material grown at 700 °C at a rate of ~ 1 μm/h, and the surface morphology is very good.[5] Moreover, we have found that the carbon-related peak in the "slowly grown AlGaAs" is very small relative to the exciton peak, ensuring a low background of scattering centers in the undoped AlGaAs spacer.

The structures were grown in a RIBER 1000-1 molecular beam epitaxy (MBE) apparatus at a base pressure of $\sim 3\times 10^{-11}$ Torr. Cells were kept continuously idling at ~ 800 °C (As at ~ 150 °C), and the liquid nitrogen shroud was always kept cold. Samples were introduced at least 12 h before growth and were kept at 400 °C until growth was initiated. A buffer layer of ~ 1-μm undoped GaAs was grown at 600 °C with an arsenic-to-gallium flux ratio $F(As_4)/$

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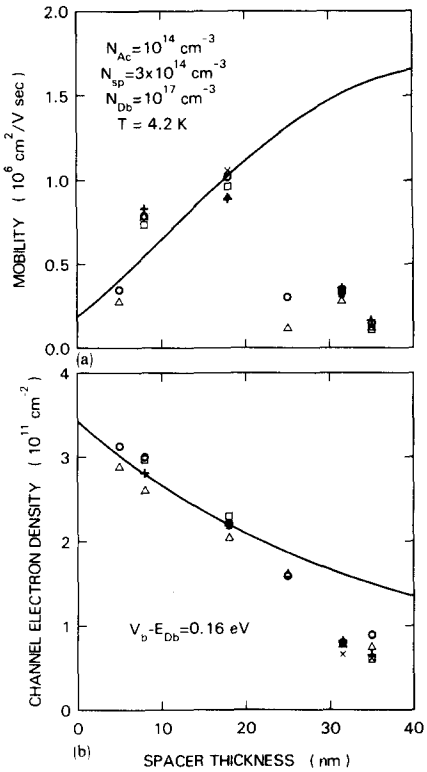

FIG. 1. 4-K Hall mobility and areal carrier density, measured in the dark, as functions of the spacer width. The solid lines are mobilities and carrier densities calculated from the specified parameters using the model of Ref. 8. N_{ac}, N_{sp}: acceptor concentrations in the GaAs and the AlGaAs spacer, respectively, N_{db}: donor concentration in the AlGaAs, V_b: barrier height, and E_{db}: donor binding energy in the AlGaAs.

F(Ga)$\simeq$1, and a growth rate of 1 μm/h. Thereafter, the Ga and As fluxes were gradually reduced over 10–15 min to produce a growth rate of 17 Å/min, and the Al shutter was opened, resulting in the growth of $Al_{0.3}Ga_{0.7}As$. Structures were grown with undoped spacers in the range of 40–350 Å, followed by 400 Å of $\sim 7\times10^{17}$ cm^{-3} silicon-doped $Al_{0.3}Ga_{0.7}As$, and a 100-Å top GaAs layer doped to $\sim 10^{18}$ cm^{-3}. Carrier concentrations and Hall mobilities were measured at 300, 77, and 4 K in the dark. The results obtained at 4 K are plotted in Fig. 1. While the carrier concentration decreases monotonically with increased spacer width, the mobility goes through a peak. The highest mobility sample had $\mu_H = 9200$, 200 000, and 1.06×10^6 cm^2/Vs at 300, 77, and 4 K, respectively, for a channel electron density $N_s = 2.2\times10^{11}$ cm^{-2}. When this sample was exposed to light, the low temperature mobility increased to 1.7×10^6 cm^2/Vs for $N_s = 3.9\times10^{11}$ cm^{-2}.

Average doping was verified by Hall measurements on a test piece. When the silicon profile was investigated by secondary ion mass spectrometry,[6] we found a strong accumulation of Si near the surface of the AlGaAs. By etching and capacitance-voltage profiling the AlGaAs layers, we have found a similar gradient in the doping profile. While the doping near the spacer can be as low as 10^{17} cm^{-3} or even lower, the doping near the outer interface exceeds 10^{18} cm^{-3} (measurements are difficult to perform due to the large leakage currents in the Hg-probe capacitor). This effect has never been reported before, but could be due to Si segregation in AlGaAs, similar to Sn segregation in GaAs.[7] This effect could account for an effectively larger spacer and a correspondingly higher mobility.

We cannot account for the measured channel electron densities in this series of runs with the nominal doping levels and spacer thicknesses. Spacer dimensions were verified by selectively etching the AlGaAs and measuring the total AlGaAs thickness by a step profiler. If the Si is depleted near the channel and segregates near the top of the sample, as suggested by the capacitance profiling results, then the measured channel electron densities N_s can be approximately accounted for, assuming that the difference between the barrier height and the donor binding energy in the AlGaAs is 0.16 eV and that there are 10^{17} donors per cm^3 in the AlGaAs space-charge layer (although the values of N_s in the samples with the largest spacers are lower than expected from these parameters). The curves in Fig. 1 were calculated using the simple model of Ref. 8, in which the envelope wave functions of channel electrons were approximated by a Fang-Howard function $z\exp(-bz/2)$, which has no barrier penetration. Ando[9] showed that such an approximation overestimates the mobility, but the errors are large only for small spacer thicknesses.

With the parameters given above, and neglecting additional scattering from compensation or from a space-charge layer at the top surface of the sample, we can account for the trend of mobilities of the samples with the smaller spacer thicknesses in Fig. 1, but the samples with 80-Å spacers have higher mobilities than can be accounted for, even with these optimistic assumptions. The samples with the larger spacer thicknesses have mobilities well below those predicted from the parameters we used. This may be related to the near exhaustion of the As source when these samples were grown. In an earlier set of samples, heterojunctions with spacers as large as 500 Å, with channel electron densities of 5×10^{10} cm^{-2}, had mobilities measured in the dark as high as 360 000 cm^2/Vs, well above those in this set of runs.

We do not have compelling evidence that the slow growth rate used here is necessary to achieve high mobilities at a 600 °C growth temperature. For structures grown at 600 °C and at a uniform growth rate of 1 μm/h, 4-K mobilities (measured in the dark) higher than 0.5×10^6 cm^2/Vs were routinely achieved. Very recently, Hiyamizu *et al.*[10] reported on mobilities higher than 10^6 cm^2/Vs in samples grown at 680 °C, a range where the quality of the AlGaAs is much better.

In conclusion, we have reported on the fabrication of selectively doped heterojunctions with 4-K mobilities, measured in the dark, as high as 10^6 cm^2/Vs. At 300 and 77 K, the mobilities were 9200 and 200 000 cm^2/Vs, limited by phonon scattering. Samples were grown at 600 °C utilizing a slow growth technique, which minimized diffusion and might be important in multilayer structures. We have found that GaAs and AlGaAs grown at a rate of 0.1 μm/h do not include more unintentional impurities if the MBE system is clean enough. Also, we have evidence that Si segregates toward the surface of AlGaAs. In our structures, this reduces the Coulombic scattering from the area nearby the spacer. Some of the mobilities exceed those calculated from the

theoretical model of Ref. 8, even when measured structural parameters are used.

We thank L. Osterling and L. Alexander for processing the samples and P. J. Price for useful discussions.

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Chemical reactions at the noble and near-noble metal/InP interfaces: Comparison to Si and GaAs

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(Received 22 November 1983; accepted for publication 11 March 1984)

The room-temperature chemical reactions of noble (Cu, Ag, Au) and near-noble transition (Ni, Pd) metals with the vacuum cleaved InP (110) surface have been studied with surface sensitive photoemission spectroscopies. It has been shown that the chemical reactions at these interfaces are closely similar to the reactions taking place on silicon substrates. In particular all metals that react strongly with Si to produce silicides (Cu, Ni, Pd) also form stable phosphides. This reaction is accompanied by phase segregation of metallic In. In addition, for Au (intermixing without a stable compound with P) and Ag (very weak reaction with substrate; growth of metallic Ag islands) the reactions with both InP and Si are qualitatively identical. It has been found that for GaAs the reactions with noble and near-noble metals, though weaker than for InP, follow the same pattern.

PACS numbers: 82.65. − i, 73.30. + y, 79.60.Gs, 68.48. + f

A necessary step in improving the technological control of practical devices and integrated circuits is to understand in detail the chemical reactions that may affect macroscopic properties (such as the Schottky barrier height) and degradation over time of the interfaces. It is the purpose of this letter to show, for the first time, that chemical reactions at the metal/InP interfaces are in some respects strikingly similar to the reactions observed for similar Si junctions, and to make comparisons with the same reactions on GaAs.

In many practical cases the behavior of the metal/semiconductor interface is strongly affected by chemical reactions between metals and the substrate. This fact was recognized many years ago for metal/silicon interfaces. It is now well known that most of the transition metals react strongly with silicon, forming stable silicides[1] following thermal processing. The technical advantage of metal/silicide junctions is that an equilibrium stoichiometric silicide is formed at the interface. This assures the room-temperature stability of the diode, and this limits degradation. It is now recognized that even at room temperature, on a clean silicon surface, a spontaneous reaction takes place between silicon and transition metals such as Pd[2] or Ni.[3] The fact that these reactions occur spontaneously on the surface at such low temperatures cannot be explained by bulk thermodynamics and bulk equilibrium phase diagrams. Despite advances in our phenomenological understanding of metal/silicon interfacial reactions, much still remains to be learned about chemical reactions at metal/III-V semiconductor interfaces. It is, however, well recognized that intermixing of metal and substrate constituents or compound formation may take place.[4] In fact, sharp interfaces may be the exception rather than the rule. Chemical reactions were studied here by monitoring changes in the shape and energy of the core level photoemission spectra of the substrate and metal atoms upon deposition of noble (Cu, Ag, Au) or some near-noble transition (Ni, Pd) metals in ultrahigh vacuum on the clean cleaved InP(110) face. Core level soft x-ray photoemission spectroscopy (SXPS) is particularly suitable for this kind of study since it provides the possibility of studying independently the environment of a particular atom at the interface through new chemically shifted peaks. The photon-induced ($h\nu = 160$ eV) P $L_{2,3}VV$ Auger transition (~ 120 eV) and valence-band photoemission spectra were also monitored. Optimal surface sensitivity of a few angstroms is achieved by tuning the source of monochromatic synchrotron radiation so that the photoelectrons from each particular core line leave the crystal with kinetic energies (E_k) close to the escape depth minimum (electrons E_k between 40 and 60 eV).[5] By separating shifts

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 0003-6951/84/111066-03$01.00