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Growth studies on $Si_{0.8}Ge_{0.2}$ channel two-dimensional hole gases

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We report a study of the influences of MBE conditions on the low-temperature mobilities of $Si/Si_{0.8}Ge_{0.2}$ 2DHG structures. A significant dependence of 2DHG mobility on growth temperature is observed with the maximum mobility of 3640 $cm^2\,V^{-1}\,s^{-1}$ at 5.4 K being achieved at the relatively high-growth temperature of 640 °C. This dependence is associated with a reduction in interface charge density. Studies on lower mobility samples show that Cu contamination can be reduced both by growth interruptions and by modifications to the Ge source; this reduction produces improvements in the low-temperature mobilities. We suggest that interface charge deriving from residual metal contamination is currently limiting the 4-K mobility.

Since work began on remote doping in the Si/SiGe system in 1982,[1] the peak two-dimensional electron gas (2DEG) 4-K mobility, μ_e, has increased by two orders of magnitude from $\approx 1\times10^3$ $cm^2\,V^{-1}\,s^{-1}$ to 1.7×10^5 $cm^2\,V^{-1}\,s^{-1}$ in 1992.[2] By contrast, in SiGe channel two-dimensional hole gas (2DHG) structures, irrespective of growth technique, the maximum reported μ_h (4 K) (Ref. 3) is 4×10^3 $cm^2\,V^{-1}\,s^{-1}$, as compared to 3.3×10^3 $cm^2\,V^{-1}\,s^{-1}$ seen in 1984.[4] This work describes a study of such structures, specifically the single quantum-well normal structure in which the B dopant is incorporated above the well. The structures were grown by molecular beam epitaxy (MBE) in a Vacuum Generators V90*S* system at a growth rate of 0.1 nm s^{-1} on ⟨100⟩ Si substrates. Ge was co-evaporated by electron beam from a 2-in. Ge charge in a water-cooled Cu hearth.

To achieve high-quality, fully strained SiGe, the growth temperature, T_s, is usually restricted to the range 450–650 °C. These limits are determined by the requirement for sufficient adatom mobility (at normal growth rates) and the retention of 2D growth processes. SiGe channel 2DHG structures were grown at various values of T_s in this temperature range. The mobility data are shown in Fig. 1 for normal structures with a 25-nm spacer and a sheet-carrier concentration of 2×10^{11} cm^{-2}. The most striking feature is the increase in low-temperature mobility with increasing growth temperature, a maximum of 3640 $cm^2\,V^{-1}\,s^{-1}$ at 5.4 K being observed for T_s=640 °C. This result is unexpected and may have ramifications beyond a study of the remote doped SiGe channel class of structures considered here.

It is conceivable that the mobility may be dominated by ionized impurity scattering, but this is considered unlikely for the following reasons. Ionized impurity scattering could arise from the background *n*-type impurities sited in the SiGe channel or by remote doping either in the spacer layer or in the depleted doped region. The background impurity (phosphorus) concentration in the V90*S* system is $<1\times10^{15}$ cm^{-3} and shows no discernible dependence on T_s in Si,[5] although background doping has not been studied for SiGe. In addition, B diffusion into the spacer layer, which would degrade the mobility, will increase with increasing T_s; B segregation effects, which do reduce with increasing T_s over this temperature range,[6] are not pertinent for normal structures other than to increase the effective spacer-layer thickness. An important phenomenon with the correct temperature dependence is Ge segregation. Nakagawa and Miyao[7] assert that Ge on Si segregation reduces with increasing T_s over the range of this study. However, there is no clear reported evidence for significant segregation in the present, relatively dilute $Si_{0.8}Ge_{0.2}$/Si system; such a segregation dependence was not, moreover, identified by x-ray analysis in similar samples. Increasing T_s, however, may act to smooth the SiGe/Si interface thereby relieving interface roughness scattering.[8] On close examination, the mobility-temperature plots of Fig. 1 are found to exhibit a minimum at about 10 K which has been attributed in a GaAs/AlAs system to the scattering of a degenerate two-dimensional gas of carriers by interface roughness.[9] Cross-sectional transmission electron microscopy (TEM) of this sample set reveals abrupt interfaces to a resolution of 1 nm. Alternatively, the accumulation of a sheet of charge at the SiGe/Si interface is another potential source of scattering[10] and this accumulation could be partially relieved at increased T_s.

A theoretical analysis of mobility as a function of sheet-carrier density[11] has considered the following scattering mechanisms; alloy, remote impurity, interface roughness, and interface charge scattering. This work has concluded that scattering by a sheet of charge at the interface is the mechanism currently limiting 4-K mobilities in samples grown at 550 °C. Further work relating to the low-temperature (0.3–20 K) conductivity and Hall coefficient[12] has confirmed this conclusion and demonstrated that the increase of mobility with increasing T_s is due to a reduction in interface charge density. Capacitance voltage (*C-V*) profiles of similar structures grown by the identical technology have clearly shown such charge at the interface;[13] the temperature dependence of which is indicative of deep acceptor states. The remainder of this letter may provide a clue to the source of this charge.

We encountered two problems in the early part of these studies before the samples of Fig. 1 were grown; 2DHG 4-K mobilities were $<1\times10^3$ $cm^2\,V^{-1}\,s^{-1}$, and

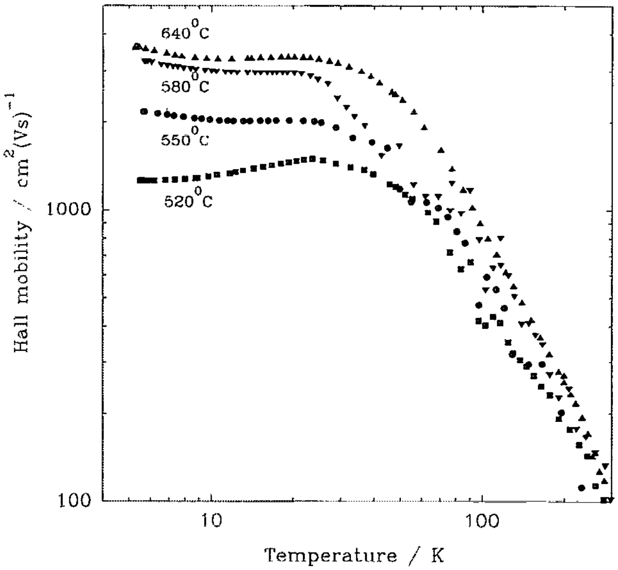

FIG. 1. Mobility as a function of growth temperature for 2D hole gases of carrier concentration 2×10^{11} cm^{-2}.

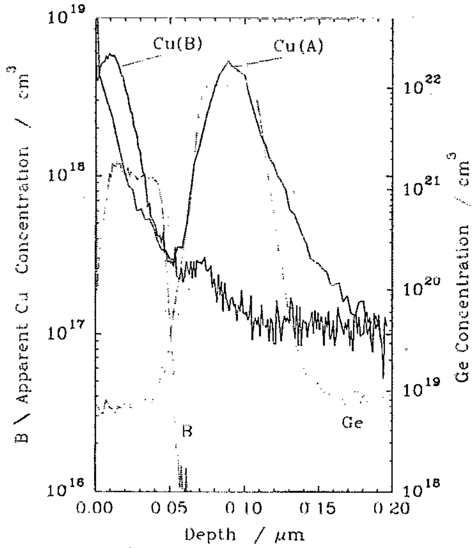

FIG. 2. SIMS profiles of samples A (no growth interruption) and B (growth interruption). The Ge and B concentrations are shown for one sample only to aid clarity.

importantly, these low values were not reproducible. These problems were solved by, first, the incorporation of a growth interruption during growth, and second, by the introduction of a Si liner around the Ge source. We believe both of these modifications produce a reduction in metal contamination in the SiGe channel. The growth interruption had the effect of increasing the 4-K 2DHG mobility from bulklike values (<100 cm^2 V^{-1} s^{-1}) to 500 cm^2 V^{-1} s^{-1} and, in otherwise identical samples, the additional use of a Si liner increased this value to 2200 cm^2 V^{-1} s^{-1}. We shall initially discuss the more subtle influence of the growth interruption.

A 3 min growth interruption at, or 0.5-nm before, the initiation of B doping was found to be an essential element of the growth schedule to obtain highly reproducible 4-K 2DHG mobilities. A growth interruption on the cessation of SiGe growth was not found to have a similarly beneficial effect. The growth interruption produced no significant increase in the time for which a layer is maintained at the growth temperature and occurred after the growth of, effectively, a Si-capped SiGe layer. These observations therefore suggested that the interruption alters the behavior of unintentional impurities which may vary within a growth series, rather than a fundamental growth process such as segregation.

Two samples, A and B, were grown prior to the use of a Si liner for the Ge source. In sample A, there was no growth interruption and in sample B, a growth interruption was employed immediately prior to B doping. As can be seen from the secondary ion mass spectrometry (SIMS) profiles shown in Fig. 2, the Cu concentration is very high, which reflects the unscreened nature of the Ge source. SIMS analysis also demonstrates that the Cu remains in the SiGe channel for sample A, but for sample B, is gettered from the channel and resides mainly in the B-doped Si region. The Cu level shown in Fig. 2 is greater than the solid-solubility concentration in Si,[14] as such the absolute level as determined by SIMS cannot be considered accurate. However, the repositioning of Cu into the B-doped region is a reliable observation.

For sample A, the 4-K mobility is bulklike and the resistivity versus temperature dependence suggests that states at the band edge are strongly localized due to distortion of the lattice along the interface. The resistivity dependence of sample B, on the other hand, is indicative of conduction in a degenerate 2D gas with no strong localization (4-K mobility is 500 cm^2 V^{-1} s^{-1}). The relatively low mobility reflects the partial retention of Cu within the conduction channel. We consider, therefore, that the beneficial effect of the growth interruption is produced by diffusion of residual metallics to a region of disorder at the interruption, created probably by residual gas uptake and subsequent retention of the metals in the B-doped material, which acts as a preferential gettering domain.

The inclusion of a Si liner to the Ge source, in conjunction with the growth interruption, increased the 2DHG mobilities fourfold from typical values of 500 cm^2 V^{-1} s^{-1} to the values shown in Fig. 1, which are highly reproducible. The effect of the Si liner is readily explained. At typical Ge deposition rates (0.01–0.06 nm s^{-1}), the entire surface of the Ge charge melts, leading to inevitable contact between molten Ge and the Cu hearth of the electron-beam evaporator. Interposing the Si liner, which remains solid at Ge evaporation temperatures, contains the Ge and minimizes Cu dissolution into the Ge melt. SIMS of the samples of Fig. 1 indicated Cu levels below the detection limit (1×10^{17} cm^{-3}) throughout the structure, hence, further studies relating to these phenomena are not trivial. While it is conceivable that the gettering phenomenon is concentration dependent or specific to fast diffusers in Si, such as Cu, the improved reproducibility suggests that it is effective with the metal concentrations expected in MBE-grown Si, i.e., $\sim10^{14}$ cm^{-3}.

In light of the identification of scattering by interface charge, we suggest that this charge is a result of unintentional contamination, almost certainly metals, that is further reduced by the use of high-growth temperatures. Although we have not demonstrated a clear correlation between the interface charge and metal concentration for the low levels encountered with the use of a Si liner, this correlation is made for higher Cu concentrations. Moreover, the nature of the dominant scattering mechanism is common to all samples with 4-K mobilities >500 $cm^2 V^{-1} s^{-1}$, i.e., including those in which metal contamination has been directly observed. The observation of deep acceptor states at the interface by *C-V* profiling has been made in material grown under identical conditions to those of the high-mobility samples. We note that Cu is a deep acceptor in Si.

To conclude, 2DHG mobilities of $\approx 3640\ cm^2 V^{-1} s^{-1}$ at 5.4 K have been obtained in the SiGe channel of the SiGe/Si system. Increasing growth temperature was found to produce significant increases in 4-K mobility, with the highest value being observed at the highest T_s (640 °C) used in this study. This is attributed to a reduction in interface charge. Studies in low-mobility samples suggest that metal-induced interface charge dominates the 4-K mobility. Analysis of the high-mobility samples in which metals cannot be directly observed also indicates that interface-charge scattering dominates. We speculate that the improvement in mobility with increasing T_s derives from a reduction in residual metal contamination at the Si/SiGe interface.

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[1] H. M. Manasevit, I. S. Gergis, and A. B. Jones, Appl. Phys. Lett. **41**, 464 (1982).
[2] F. Schäffler, D. Többen, H-J. Herzog, G. Abstreiter, and B. Holländer, Semicond. Sci. Technol. 7, 260 (1992).
[3] T. Mishima, C. W. Fredriksz, G. F. A. van der Walle, D. J. Gravesteijn, R. A. van den Heuvel, and A. A. van Gorkum, Appl. Phys. Lett. **57**, 2567 (1990).
[4] R. People, J. C. Bean, D. V. Lang, A. M. Sergent, H. L. Stormer, K. W. Wecht, R. T. Lynch, and K. Baldwin, Appl. Phys. Lett. **45**, 1231 (1984).
[5] R. F. Houghton, Ph.D. thesis, University of Warwick, 1991.
[6] C. P. Parry, R. A. Kubiak, S. M. Newstead, T. E. Whall, and E. H. C. Parker, MRS Proc. **220**, 121 (1991).
[7] K. Nakagawa and M. Miyao, J. Appl. Phys. **69**, 3058 (1991).
[8] S. Mori and T. Ando, J. Phys. Soc. Jpn. **48**, 865 (1980).
[9] H. Sakaki, T. Noda, K. Hirakawa, M. Tanaka, and T. Matsusue, Appl. Phys. Lett. **51**, 1934 (1987).
[10] A. Gold and V. T. Dolgopolov, Phys. Rev. B **33**, 1076 (1986).
[11] C. J. Emeleus, D. W. Smith, T. E. Whall, E. H. C. Parker, and M. J. Kearney (unpublished).
[12] C. J. Emeleus, T. E. Whall, D. W. Smith, E. H. C. Parker, and M. J. Kearney, Thin Solid Films (accepted for publication).
[13] J. Brighten, R. A. Kubiak, P. J. Phillips, T. E. Whall, E. H. C. Parker, I. D. Hawkins, and A. R. Peaker, Thin Solid Films (accepted for publication).
[14] R. N. Hall and J. H. Racette, J. Appl. Phys. **35**, 379 (1964).