Very high two-dimensional hole gas mobilities in strained silicon germanium

E. Basaran, R. A. Kubiak, T. E. Whall, and E. H. C. Parker

Physics Department, University of Warwick, Coventry, CV4 7AL, United Kingdom

We report on the growth by solid source MBE and characterization of remote doped Si/SiGe/Si two-dimensional hole gas structures. It has been found that by reducing the Ge composition to $\leqslant 13\%$ and limiting the thickness of the alloy layer, growth temperatures can be increased up to 950 °C for these structures while maintaining good structural integrity and planar interfaces. Record mobilities of 19 820 $cm^2 V^{-1} s^{-1}$ at 7 K were obtained in normal structures. Our calculations suggest that alloy scattering is not important in these structures and that interface roughness and interface charge scattering limit the low temperature mobilities.

There have been dramatic improvements in 4 K two-dimensional electron gas mobilities in silicon grown on relaxed SiGe buffer layers in recent years with values up to 1.8×10^4 cm^2 V^{-1} s^{-1} now being reported.[1-3] In contrast, two-dimensional hole gas (2DHG) mobilities in coherent low Ge content ($\leqslant 20\%$) SiGe alloys have for some time remained at around 4×10^3 cm^2 V^{-1} s^{-1}.[4] We have, however, very recently reported[5] significantly higher 2D mobilities (9300 cm^2 V^{-1} s^{-1} at 4 K) in strained SiGe and the analysis of mobility behavior indicates that it is limited by scattering associated with charge or roughness at the Si/SiGe interface. High 4 K 2DHG mobilities (55 000 cm^2 V^{-1} s^{-1}) have also recently been obtained in Ge grown on relaxed SiGe buffer layers.[6] In this letter we report on an extension to our studies involving an investigation of how growth temperature schedules influence low temperature 2DHG mobilities in coherent Si/SiGe/Si structures. It was found that Ge concentrations $\leqslant 13\%$ were needed to ensure minimal strain relaxation at the highest growth temperatures, with alloy layer thicknesses comparable to the equilibrium critical thicknesses. The influence of the high growth temperature was primarily to reduce interface charge scattering giving mobilities up to 19 820 cm^2 V^{-1} s^{-1} at 7 K.

The Si/SiGe structures were grown in a VG V90S MBE system, fitted with a 180 cc electron beam evaporator (AP&T) for silicon, a 40cc electron beam evaporator (Edwards Temescal) for Ge, and a special graphite element substrate heater. Total growth rates of between 0.1 and 0.3 nm s^{-1} were controlled using an Inficon Sentinel III system. Elemental B doping was achieved using an in-house designed source. Substrate temperatures between 550 and 950 °C were monitored using an Ircon pyrometer and the reproducibility between growth runs was better than ±10 °C. Low field Hall measurements down to 4 K were carried out on a cross-shaped samples using the van der Pauw methodology. Hall bars were fabricated on selected samples for studies at high magnetic fields (up to 15 T) and down to temperatures of 0.3 K; these are briefly mentioned here but will be the subject of more detailed publications.

Earlier we reported the presence of Cu in Ge-containing structures, whereas Cu was below the secondary ion mass spectroscopy (SIMS) detection limit of (10^{17} cm^{-3}) for Si only structures.[4] The presence of Cu can be readily attributed to the use of Cu hearths in the electron beam evaporators. In the present work both the Si and Ge charges were contained in pyrolytic graphite crucibles which acted as liners in the electron beam hearth. Such a procedure should lead to considerable reductions in residual Cu contamination levels in the 2DHG structures.

The studies were carried on single quantum well Si/SiGe/Si structures grown on 0.3-μm-thick buffer layers on $n^-[100]$ substrates. Elemental B doping of the cap layer at a concentration of $1-5 \times 10^{18}$ cm^{-3} was initiated 20–40 nm after completing the alloy layer growth, thus forming a spacer layer and producing a 2DHG at the upper Si/SiGe interface, (i.e., a "normal" structure). The Ge alloy concentrations in the alloy were in the range 6%–13% and alloy thicknesses were between 30–74 nm. An exception to this specification was sample 33/56, where the spacer layer L_s was ~63 nm (producing a correspondingly lower 2DHG sheet density). The equilibrium critical thickness according to Matthew and Blakeslee for these structures is in the range 20–70 nm so that all of the layers were close to or in the equilibrium regime.[7] XTEM analysis of some of the structures revealed no dislocations and that they were of high crystallographic quality with sharp interfaces. Defect etching also showed very low misfit dislocation densities. X-ray rocking curves obtained using the symmetric [004] reflection were used to determine the Ge concentration in some of the alloy layers assuming the layers were fully strained. Preliminary analysis using both the symmetric [004] and asymmetric [115] reflections indicated the alloy layer in sample 33/56 was >92% strained. Although Pendellosung thickness fringes were evident in nearly all of the X-ray rocking curves, indicating sharp planar interfaces, XTEM on certain samples revealed the upper Si/SiGe interface to be undulating with a period of between 500–1000 nm and the depth of the undulation varied up to a maximum of 20 nm. These observations are consistent with strain-related surface diffusion effects.[8]

The 2DHG 4 K Hall mobilities are shown in Fig. 1 plotted against growth temperature (T_s). The mobility increased from 4000 cm^2 V^{-1} s^{-1} for a 13% alloy grown at 650 °C to 17 650 cm^2 V^{-1} s^{-1} for a 6.5% alloy grown at 900 °C. To achieve the highest mobilities it was found necessary to access growth temperatures $\geqslant 890$ °C and to reduce the growth
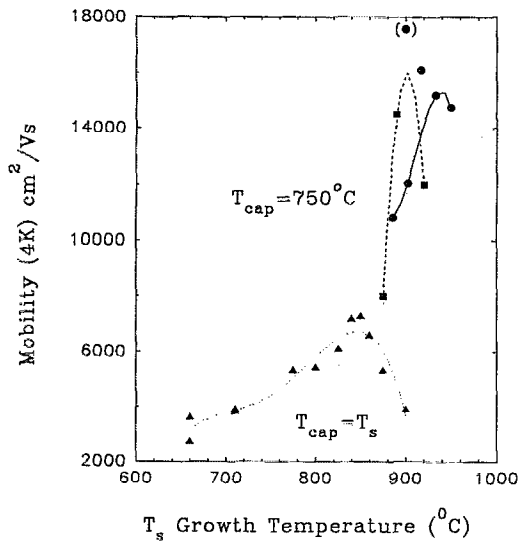
FIG. 1. 4 K 2DHG mobilities in $Si_{1-x}Ge_x$. T_{cap} is the set temperature at which the cap layer was grown. ▲ $x=0.11-0.14$, $L_w=50$ nm, $L_s=20$ nm, $N_s=2-4\times10^{11}$ cm^{-2}, ● $x=0.06-0.09$, $L_w=30$ nm, $L_s=20-40$ nm, $N_s=4-10\times10^{10}$ cm^{-2} (●) $x=0.065$, $L_w=74$ nm, $L_s=63$ nm, $N_s=4\times10^{10}$ cm^{-2}. ■ $x=0.05-0.06$, $L_w=30$ nm, $L_s=30$ nm, $N_s=4-12\times10^{10}$ cm^{-2}.
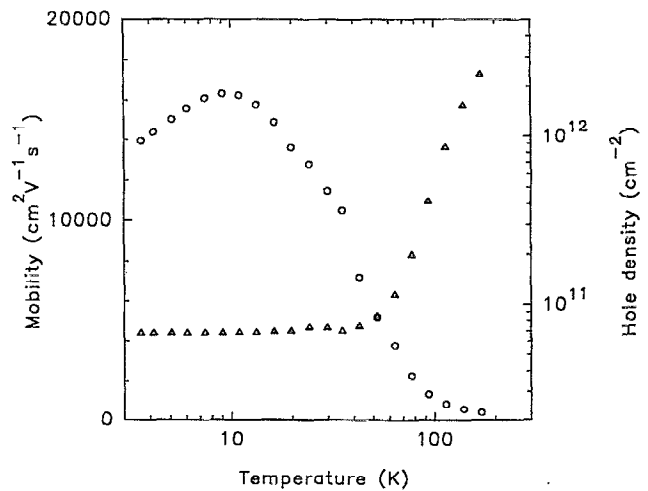

FIG. 2. Hall mobility (O) and carrier sheet density (△) vs temperature for a Si/SiGe$_{0.92}$Ge$_{0.08}$/Si structure. A 2D sheet density $\approx7\times10^{10}$ cm^{-2} is indicated by the Hall measurements.

temperature immediately after termination of the alloy layer growth in order that the doped part of the silicon cap was grown while the wafer was cooling to 750 °C (though it rarely achieved this temperature by the end of growth). The mobility seemed to peak sharply at growth temperatures around 900 °C. The reduction in the mobility for $T_s>900$ °C could be associated with a reduction in the effective spacer width due to the propensity of Si to planarize an undulating surface following growth of SIGe at higher temperature by filling in the depressions, or due to B diffusion during growth of the cap layer. Higher sheet densities observed in samples grown at temperatures above 900 °C provide evidence to support this view.

Table I shows the measured properties of some of the layers. The highest mobility was 19 820 cm^2 V^{-1} s^{-1} at 7 K obtained for sample 33/56, which had a Ge concentration of 6.5% and a sheet density of 4×10^{10} cm^{-2}. Hall data obtained from an 8% alloy layer over the temperature range 4−300 K are shown in Fig. 2, indicating a 2DHG density of 7×10^{10} cm^{-2}, a 4 K mobility of 14 200 cm^2 V^{-1} s^{-1} and a peak mobility of 16 200 cm^2 V^{-1} s^{-1} at 9 K. Shubnikov de Haas oscillations obtained from some of the samples confirmed the 2D confinement of the carriers and gave sheet densities similar to those obtained from the Hall measurements.

(Shubnikov de Haas measurements have not yet been carried out on the highest mobility samples). Measurements down to lower temperatures on some of the samples show that the mobility goes through a minimum at around 3 K and increases with reducing temperature to the lowest temperature (0.3 K). This temperature dependence of the mobility is similar to that which has been seen previously in GaAs/AlAs quantum wells by Sakaki et al.[9] The increase as the temperature is decreased below 3 K is attributed to an increase in screening. A relaxation time which increases with energy can account for the minimum at 3 K, followed by a maximum at around 10 K as phonon scattering becomes dominant. These features are currently the subject of further study.

The interpretation of the mobility data for a range of structures with various sheet densities is shown in Fig. 3. Figure 3 includes data on structures grown by us and discussed previously where maximum 4 K mobilities were of the order 2500 cm^2 V^{-1} s^{-1} (Ref. 10) and 9300 cm^{-1} V^{-1} s^{-1}.[5] We have carried out calculations of screened alloy scattering at 0 K for a 13% alloy. These calculations differ from those carried out previously by us[10] in that we have used a more accurate value of the effective mass, equal to 0.23m_0 as deduced from Shubnikov de Haas measurements.[11] The new calculations also employ a more accurate treatment of screening based on the work of Gold.[12] Both the present and previous calculations indicate that alloy scattering is not an important factor in any of the present

TABLE I. Measured properties of some 2DHG structures.

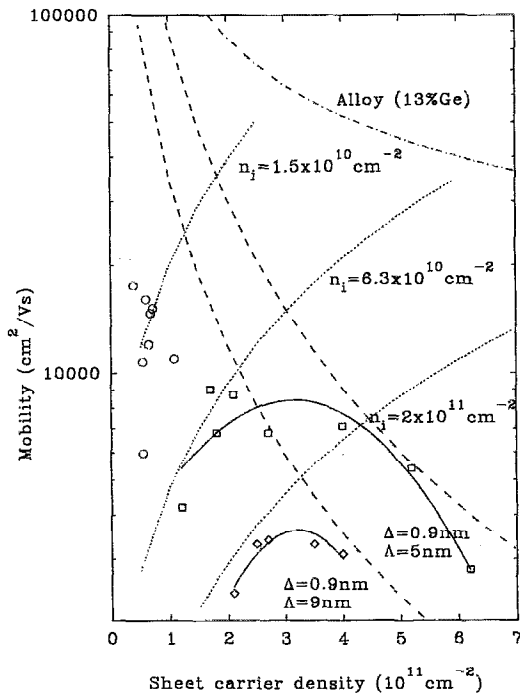| Sample No.	Ge Conc.%	μ(4 K) cm^2 V^{-1} s^{-1}	μ_{peak}(T) cm^2 V^{-1} s^{-1} (K)	μ(0.35 K) cm^2 V^{-1} s^{-1}	N_s (Hall) cm^{-2}	N_s (SdH) cm^{-2}
31/17	13[a]	11,100		15 000	1.8×10^{11}	2.05×10^{11}
33/55	9.1	11,000			6.8×10^{10}	
33/56	6.5	17,580	19 820 (7 K)		3.9×10^{10}	
34/45	8.3	16,100			6×10^{10}	

[a]Indicates uncalibrated concentration.

FIG. 3. 4 K $Si_{1-x}Ge_x$ 2DHG mobility vs sheet density. The dot-dashed curve is screened alloy scattering for $x = 0.13$ (theory), the dashed curves are short range interface roughness scattering for different correlation lengths Λ and roughness depths Δ (theory) and the dotted curves are interface charge scattering for different interface charge densities, n_i (theory). $\Diamond$ experiment $x = 0.20$, $T_s = 550{-}710$ °C. $\square$ experiment $x = 0.13$, $T_s = 875$ °C. $\bigcirc$ experiment $x = 0.06{-}0.13$, $T_s = 890{-}950$ °C.

samples. The two scattering processes which seem to be most dominant in the present material, and which can account for observed dependence of mobility on sheet density, are interface charge and interface roughness scattering—the former dominating at low sheet densities and the latter at high sheet densities. The interface roughness scattering calculations depend on the correlation length Λ and the depth Δ

of the interface roughness and the values chosen are those which give the best fits to the data in Fig. 3. It should be noted that since this is a two parameter fit the choice of these parameters must be regarded as somewhat arbitrary. Insufficient XTEM analysis has been carried out to deduce reliable values of Λ and Δ for the present structures. The mobility analysis indicates that for low sheet densities in the present structures ($\leqslant 1 \times 10^{11}$ cm^{-2}) interface charge scattering is prevalent and that the primary effect of increasing growth temperatures is to reduce the density of this charge to $\approx 2 \times 10^{10}$ cm^{-2}. Additional confirmation that such short range scattering was prevalent in these structures at low temperatures has been obtained from measurements of the quantum lifetime in $Si_{0.87}Ge_{0.13}$ 2DHGs.[11]

In conclusion we have shown that very high mobilities can be obtained in strained SiGe channels through the use of high growth temperatures. It is possible that even higher mobility values can be achieved, offering further prospects for fundamental studies and for improved room temperature FET device operation.

[1] H. Schäffler, D. Többen, H. J. Herzog, G. Abstreiter, and B. Hollander, Semicond. Sci. Technol. 7, 260 (1992).

[2] Y. H. Xie, E. A. Fitzgerald, D. Monroe, P. J. Silverman, and G. P. Watson, J. Appl. Phys. 73, 8364 (1993).

[3] S. Q. Murphy, Z. Schlesinger, S. F. Nelson, J. O. Chu, and B. S. Meyerson, Appl. Phys. Lett. 63, 222 (1993).

[4] D. W. Smith, C. J. Emeleus, R. A. Kubiak, T. E. Whall, and E. H. C. Parker, Appl. Phys. Letts. 61, 1453 (1992).

[5] T. E. Whall, D. W. Smith, A. D. Plews, R. A. Kubiak, P. J. Phillips, and E. H. C. Parker, Semicond. Sci. Technol. 8, 615 (1993).

[6] Y. H. Zie, Don Monroe, E. A. Fitzgerald, P. J. Silverman, F. A. Thiel, and G. P. Watson, Appl. Phys. Lett. 63, (1993).

[7] J. W. Matthews and A. E. Blakeslee, J. Cryst. Growth 27, 118 (1974).

[8] A. J. Pidduck, D. J. Robbins, A. G. Cullis, W. Y. Leong, and A. M. Pitt, Thin Solid Films 222, 78 (1992).

[9] H. Sakaki, T. Noda, K. Hirakawa, M. Tanaka, and T. Matsusue, Appl. Phys. Lett. 23, 1934 (1987).

[10] C. J. Emeleus, T. E. Whall, D. W. Smith, R. A. Kubiak, E. H. C. Parker, and M. J. Kearney, J. Appl. Phys. 73, 3852 (1993).

[11] T. E. Whall, N. L. Mattey, A. D. Plews, P. J. Phillips, O. A. Mironov, R. J. Nicholas, and M. J. Kearney, Appl. Phys. Lett. 64, 357 (1994).

[12] A. Gold, Phys. Rev. B 38, 798 (1988).