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Electrical properties of boron-doped p -SiGeC grown on n^- -Si substrate

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Electrical properties of fully strained boron-doped $\text{Si}_{0.90-y}\text{Ge}_{0.10}\text{C}_y/n^-$ -Si grown by low pressure chemical vapor deposition have been investigated as a function of carbon content (0.2%–1.5%), using the variable temperature (25–650 K) Hall-effect technique. The results of Hall-effect measurements show that the Si substrate and the SiGeC/Si interfacial layer affect significantly the electrical properties of the SiGeC epitaxial layer. Thus, a three-layer conducting model has been used to extract the carrier concentration and mobility of the SiGeC layer alone. At room temperature, the hole carrier concentration decreases from 6.8×10^{17} to $2.4 \times 10^{17} \text{ cm}^{-3}$ and the mobility decreases from 488 to $348 \text{ cm}^2/\text{Vs}$ as the carbon concentration increases from 0.2% to 1.5%. The boron activation energy increases from 20 to 50 meV as C increases from 0.2% to 1.5% with an increment of 23 meV per atomic % of C. © 2000 American Institute of Physics. [S0003-6951(00)01835-0]

Recently, considerable progress has been made in the growth and characterization of $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y/\text{Si}$ systems because of their potential applications in heterojunction bipolar transistors,¹ resonant tunneling diodes,² and modulation doped field effect transistor (MODFET)³ compatible with Si technology. It has been found that incorporation of carbon in $\text{Si}_{1-x}\text{Ge}_x$ affects the lattice constant, strain compensation, carrier transport, band gap, and band offsets.⁴ Although the structural and optical properties of $\text{Si}_{1-x-y}\text{Ge}_x\text{C}_y$ alloys have been reported, very little study has been carried out on their electrical properties. Therefore, temperature dependent Hall (TDH) effect measurements were made on boron doped $\text{Si}_{0.9-y}\text{Ge}_{0.10}\text{C}_y/\text{Si}$. In particular, the effects of carbon concentration on carrier concentration, mobility, and boron activation energy are reported here.

The $\text{Si}_{0.90-y}\text{Ge}_{0.10}\text{C}_y$ epitaxial layers have been grown using low pressure chemical vapor deposition on an unintentionally doped n^- -Si substrate. The Ge content is fixed at 10%, while the C content was varied between 0.2% and 1.5%. The epitaxial films are 200 nm thick, and the thickness was limited by the critical thickness of the SiGeC single crystal on Si. The growth temperature was fixed at a temperature slightly below 700 °C. Cross section transmission electron microscopy of identical samples show atomically smooth interfaces and pseudomorphic layer growth, for a thickness of 200 nm and the composition range used. The boron partial pressure was kept at a constant value during crystal growth, and thus B-doping levels are expected to be the same for all samples at a nominal value of around $1 \times 10^{18} \text{ cm}^{-3}$. For ohmic contact, Al metal was thermally deposited and was annealed under an argon atmosphere at 550 °C for 5 min.

The apparent measured sheet carrier concentration as a function of inverse temperature is plotted in Fig. 1 for C = 1.0%; the break in the figure separates the high and low temperature regimes. At low temperatures, the p -type carrier concentration remains constant up to about 50 K and then increases exponentially with temperature, denoting the activation of shallow boron acceptors. As the temperature in-

FIG. 1. Apparent measured carrier concentration as a function of inverse temperature for C = 1%. Open symbols indicate p -type conductivity, while solid symbols indicate n type. The solid line is a theoretical fit of a three-conducting layer model to the experimental data.

creases further, the p -type carrier concentration (open symbols) starts to increase rapidly reaching a peak near 375 K, where the apparent conductivity changes from p to n type. Then, the apparent n -type carrier concentration (solid symbols) decreases initially as the temperature increases further, and finally it starts to increase with temperature in the high temperature regime. The measured Hall mobilities, which are not included in the figure, also show a singularity behavior at about 375 K. Similar behavior has been seen in the other $\text{Si}_{0.9-y}\text{Ge}_{0.10}\text{C}_y$ samples, but the singularity peak positions shift to higher temperatures with decreasing C. The apparent measured Hall data of the other samples were reported elsewhere,⁵ and are not included in Fig. 1.

These observations of the Hall coefficient singularities (thus, carrier density and mobility singularities) in all SiGeC samples, which are moderately p -type doped (10^{18} cm^{-3}), are unexpected results. Although this type of Hall coefficient singularity has been observed in very lightly p -doped silicon (10^{12} cm^{-3}),⁶ it is mostly observed in narrow band gap semiconductor materials such as InSb.⁷ For our samples, the epilayer of SiGeC is so thin ($\sim 2000\text{ \AA}$) that the ohmic contact metal apparently penetrates into the n^- -Si substrate, and thus the n^- -Si layer affects the Hall measurements of the p -SiGeC layer. In addition, the constant carrier concentration at temperatures below 50 K shows evidence of a parallel conducting path near the interfacial region between the SiGeC epitaxial layer and the Si substrate.

In order to extract the electrical properties of SiGeC layer alone from the apparent measured Hall data, a three conducting layer model was applied to the temperature dependent measured carrier concentration and mobility data, which takes into account compensation of the Hall constant, R_H , due to multiple conducting layers. The relevant relationships for a three-layer analysis are⁸

$$p = \frac{(\mu_1 p_1 + \mu_2 n_2 + \mu_3 p_3)^2}{\mu_1^2 p_1 - \mu_2^2 n_2 + \mu_3^2 p_3} \quad (1)$$

and

$$\mu = \frac{\mu_1^2 p_1 - \mu_2^2 n_2 + \mu_3^2 p_3}{\mu_1 p_1 + \mu_2 n_2 + \mu_3 p_3}, \quad (2)$$

where p is the apparent measured sheet Hall concentration, μ the apparent measured Hall mobility, μ_i the mobility of layer i , and p_i or n_i the sheet carrier concentration of layer i . In the present case, layers 1, 2, and 3 correspond to the p -type SiGeC layer, the n -type Si substrate, and a p -type interfacial layer between the SiGeC layer and the Si substrate, respectively. Using the charge balance equation, the theoretical temperature dependence of the free hole carrier concentration p_1 for the SiGeC layer alone with a single acceptor density N_{A1} , partially compensated with a donor density of N_{D1} , is given by⁹

$$p_1 = \frac{1}{2} [\phi(T, E_{A1}) + N_{D1}] \cdot \left(\sqrt{1 + 4\phi(T, E_{A1}) \frac{N_{A1} - N_{D1}}{[\phi(T, E_{A1}) + N_{D1}]^2}} - 1 \right), \quad (3)$$

where $\phi(T, E_{A1}) = (g_1/g_0)N_V' T^{3/2} \exp(-E_{A1}/kT)$. Here g_0 is the degeneracy of the unoccupied acceptor state (assume $g_0 = 4$), g_1 the degeneracy of the occupied state (assume g_1

$= 1$), N_V' the effective density of states in the valence band at 1 K, E_{A1} the acceptor activation energy, and k the Boltzmann constant. A similar single-donor model is used for the Si substrate, with E_{D2} chosen as the donor activation energy, N_{D2} the donor concentration, N_{A2} the acceptor concentration, $g_0 = 1$, and $g_1 = 2$. The measured hole concentration in the low temperature regime below 50 K remains constant and independent of temperature, and it is believed to be due to an interfacial layer. Thus, the value of p_3 in Eqs. (1) and (2) is constant for each carbon content y . For the Hall mobility, we used the Mattheissen's rule

$$\frac{1}{\mu_i} = \frac{1}{\mu_I} + \frac{1}{\mu_L}, \quad (4)$$

where μ_i is the mobility of each conducting layer of SiGeC or Si substrate, μ_I the mobility due to ionized impurity scattering with a $T^{3/2}$ dependence at low temperatures, and μ_L the mobility due to phonon scattering with a $T^{-3/2}$ dependence at high temperatures.¹⁰ The mobility in the interfacial layer, which is nearly independent of temperature, is set equal to the mobility at 25 K for each carbon content.

The apparent measured sheet hole concentration p at a given temperature is given by $p = r_H / eR_H$. Here, the Hall factor, r_H , is defined as the ratio of the measured Hall mobility and the drift mobility, and depends on scattering mechanisms, band structure, and temperature. The values of the Hall factor in p -type $\text{Si}_{1-x}\text{Ge}_x$ are variously reported to be between 0.1 and 3. For a Ge content $x = 12\%$ and doping concentration of about $1-2 \times 10^{18}\text{ cm}^{-3}$, Im *et al.*¹¹ have calculated a Hall factor of about 0.48 at room temperature. Also, Joelsson *et al.*¹² have experimentally determined r_H to be equal to 0.50 and 0.21 for $x = 3\%$ and 17% at B doping concentrations of 2.46×10^{18} and $2.07 \times 10^{18}\text{ cm}^{-3}$, respectively. Thus, an extrapolated value of 0.40 for the Hall factor has been used for the current samples with a doping concentration of $1 \times 10^{18}\text{ cm}^{-3}$ and a Ge content of 10%.

A parametric fit of Eqs. (1)–(4) to the measured carrier concentration and mobility has been made, and the solid line in Fig. 1 is the theoretical fit to the experimental data. The resulting fitting parameters of the B acceptor activation energy, E_{A1} , acceptor concentration, N_{A1} , and donor concentration, N_{D1} , for all the SiGeC layers having different carbon contents are listed in Table I. Also, the free hole concentration and Hall mobility of the SiGeC layer alone are plotted in Figs. 2(a) and 2(b) as a function of inverse temperature and temperature, respectively, for the samples of different C contents. Clearly, the boron activation energy increases with C content, and this increase may be mainly due to the increase of band gap energy with C content. A linear fit to E_{A1} versus C data gives an increase in activation energy of $23 \pm 4\text{ meV}$ per atomic % C. At room temperature, the free carrier concentration of the SiGeC layer alone decreases from 6.8×10^{17} to $2.4 \times 10^{17}\text{ cm}^{-3}$ while the measured apparent value varies from 1×10^{18} to $2 \times 10^{17}\text{ cm}^{-3}$ as the C concentration increases from 0.2% to 1.5%. Also, the acceptor concentration of the SiGeC layer alone varies from 2×10^{18} to $1 \times 10^{18}\text{ cm}^{-3}$ as the C concentration increases from 0.2% to 1.5%, although the boron partial pressure was kept constant during crystal growth for all samples. Apparently, the B incorporation in SiGeC decreases with increasing carbon con-

TABLE I. The boron activation energy, acceptor concentration N_{A1} , donor concentration N_{D1} , room temperature hole concentration, and mobility of $\text{Si}_{0.9-y}\text{Ge}_{0.1}\text{C}_y$ layer alone for different carbon contents.

$\text{Si}_{0.9-y}\text{Ge}_{0.1}\text{C}_y$ (% C)	Activation Energy (meV)	N_{A1} (10^{18} cm^{-3})	N_{D1} (10^{17} cm^{-3})	$\mu(300 \text{ K})$ ($\text{cm}^2/\text{V s}$)	$p(300 \text{ K})$ (10^{17} cm^{-3})
0.2	20 ± 5	2.2 ± 1.1	1	488	6.80 ± 0.04
0.5	30 ± 5	1.6 ± 0.5	1	438	5.27 ± 0.08
1.0	45 ± 5	1.4 ± 0.3	1	403	3.42 ± 0.02
1.5	50 ± 5	0.9 ± 0.1	1	348	2.43 ± 0.05

tent, which is consistent with the observation made by Serpentine *et al.*¹³ The room temperature mobility of the SiGeC layer alone decreases from 488 to 348 $\text{cm}^2/\text{V s}$, while the measured apparent hole mobility varies from 183 to 125 $\text{cm}^2/\text{V s}$ as the C composition is varied from 0.2% to 1.5%. The decrease in mobility may be due to an enhanced alloy scattering mechanism as C is added.¹⁴

The Si substrate is *n* type and highly resistive. From the fit, $N_{D2} = 5 \times 10^{11} \text{ cm}^{-3}$ and $N_{A2} = 1 \times 10^{11} \text{ cm}^{-3}$ were obtained for the Si substrate. A donor activation energy of 45 meV reveals the identity of the residual impurity to be phos-

phorus. At room temperature, a mobility of 1400 $\text{cm}^2/\text{V s}$ was extracted, and this value agrees well with the mobility of Si for similar residual donor concentrations.¹⁵ At the interfacial layer, the hole concentration decreases with increasing C content from 4×10^{12} to $7 \times 10^{11} \text{ cm}^{-2}$, while the mobility increases from about 150 to 550 $\text{cm}^2/\text{V s}$ at 50 K. We believe that the *p*-type interfacial layer is generated because of the lattice mismatch between the SiGeC and Si semiconductor materials. The incorporation of C in SiGe materials reduces strain in the SiGeC films, which then reduces the dislocation density, in agreement with our observation of decreasing interfacial hole concentration as C increases.

In summary, Hall-effect measurements show that the Si substrate and a *p*-type interfacial layer near the SiGeC/Si interface affect significantly the electrical properties of the SiGeC layer. The carrier concentration and mobility of $\text{Si}_{0.9-y}\text{Ge}_{0.1}\text{C}_y$ layer alone were extracted by using a three conducting-layer model. At room temperature, the carrier concentration decreases from 6.8×10^{17} to $2.4 \times 10^{17} \text{ cm}^{-3}$ and the mobility decreases from 488 to 348 $\text{cm}^2/\text{V s}$ as the C concentration increases from 0.2% to 1.5%. The boron activation energy increases from 20 to 50 meV as C increases from 0.2% to 1.5% with an increment of 23 meV per atomic % of C.

FIG. 2. The hole concentration (a) and Hall mobility (b) of the B doped $\text{Si}_{0.9-y}\text{Ge}_{0.1}\text{C}_y$ layer alone as a function of inverse temperature and temperature, respectively, for different carbon contents.

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