$Si/Si_{1-x-y}Ge_xC_y/Si$ Heterojunction Bipolar Transistors

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***Abstract*—In this letter, we report the first $Si/Si_{1-x-y}Ge_xC_y/Si$ n-p-n heterojunction bipolar transistors and the first electrical bandgap measurements of strained $Si_{1-x-y}Ge_xC_y$ on Si (100) substrates. The carbon compositions were measured by the shift between the $Si_{1-x-y}Ge_xC_y$ and $Si_{1-x}Ge_x$ X-ray diffraction peaks. The temperature dependence of the HBT collector current demonstrates that carbon causes a shift in bandgap of +26 meV/%C for germanium fractions of $x = 0.2$ and $x = 0.25$. These results show that carbon reduces the strain in $Si_{1-x}Ge_x$ at a faster rate than it increases the bandgap (compared to reducing x in $Si_{1-x}Ge_x$), so that a $Si_{1-x-y}Ge_xC_y$ film will have less strain than a $Si_{1-x}Ge_x$ film with the same bandgap.**

I. Introduction

Recently, there has been substantial interest in $Si_{1-x-y}Ge_xC_y$ alloy layers for extending the capability of Si-based heterostructures beyond $Si_{1-x}Ge_x$. By now, it has been well established that the small size of the carbon atom enables one to compensate the strain in $Si_{1-x}Ge_x$ alloys grown commensurately on Si (100) substrates [1], [2]. However, no electrical devices or electrically measured bandgap data have been previously reported in the $Si_{1-x-y}Ge_xC_y$ materials system. This paper demonstrates n-p-n heterojunction bipolar transistors with strained $Si_{1-x-y}Ge_xC_y$ bases and Si emitters and collectors. We also present bandgap data for $Si_{1-x-y}Ge_xC_y$, obtained from the HBT electrical characteristics.

II. Growth and Device Structure

To date, $Si_{1-x-y}Ge_xC_y$ layers have been grown by MBE [1] and CVD [2], and also formed by ion implantation and annealing [3]. Our experiments were performed at Princeton University using $Si_{1-x-y}Ge_xC_y$ layers grown by rapid thermal chemical vapor deposition (RTCVD) [4] at both 550°C and 625°C using methylsilane as the carbon precursor. The source gases were methylsilane, germane, and dichlorosilane for the base and collector, and silane for the emitter. The layers were doped *in situ* using B_2H_6 and PH_3. A hydrogen carrier gas was used at a growth pressure of 6 torr. Four wafers were grown at both $Si_{1-x-y}Ge_xC_y$ growth temperatures keeping all gas flows constant between wafers, except for methylsilane which was varied to change the amount of carbon in the layers.

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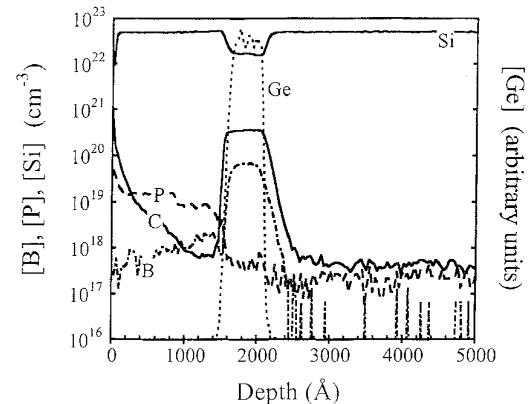

Fig. 1. SIMS of a 1.1% carbon wafer showing B, P, Ge, Si, and C profiles. B and P are calibrated on a logarithmic scale. The C measurement is also plotted on a logarithmic scale with 7 orders of magnitude, but is not absolutely calibrated. Ge is plotted in arbitrary units on a linear scale.

Fig. 1 shows a SIMS profile measured at Motorola of the following layer structure. A 5 μm n^+ buffer layer was grown on p-type, 20 Ω cm, 100 mm wafers at 1000°C, followed by the 6000 Å, n^- collector doped with phosphorus between 10^{16} and 10^{17} cm^{-3}. The 460 Å, 7×10^{19} cm^{-3} doped $Si_{0.739}Ge_{0.25}C_{0.011}$ base was grown at 550°C with ~70 Å $Si_{0.739}Ge_{0.25}C_{0.011}$ undoped spacer layers on each side. The 600 Å, 8×10^{18} cm^{-3} doped emitter was then grown at 700°C followed by a 800 Å, 2×10^{19} emitter contact layer. XTEM performed at Motorola showed no defects, dislocations, or SiC precipitates in any of the 550°C layers. The 625°C structure was similar to the 550°C one, except that the base doping was an order of magnitude lower and the lightly doped collector was 2000 Å.

Fig. 2 shows X-ray diffraction spectra of the [004] reflections from the four wafers with bases grown at 550°C. Two peaks are present from each layer because the X-ray diffractometer did not filter out the $K_{\alpha 2}$ X-ray line. As carbon is added, note that the peaks of the strained $Si_{1-x-y}Ge_xC_y$ layers shift toward the Si substrate peak, indicating a reduction of strain. A shift in the X-ray peak ($\Delta 2\theta$) from the substrate of 1.4° was measured for the control sample, corresponding to $x = 0.25$. In the samples with carbon, it was assumed that the germanium levels did not change as C was added since the germane flow was held constant. This was confirmed by SIMS.

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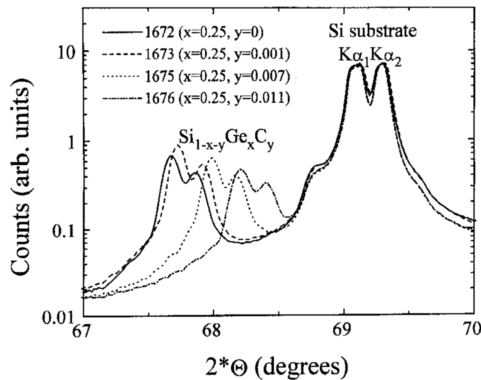

Fig. 2. X-ray diffraction spectra of the 550°C $Si_{0.75}Ge_{0.25}$ sample and the three 550°C $Si_{1-x-y}Ge_xC_y$ samples.

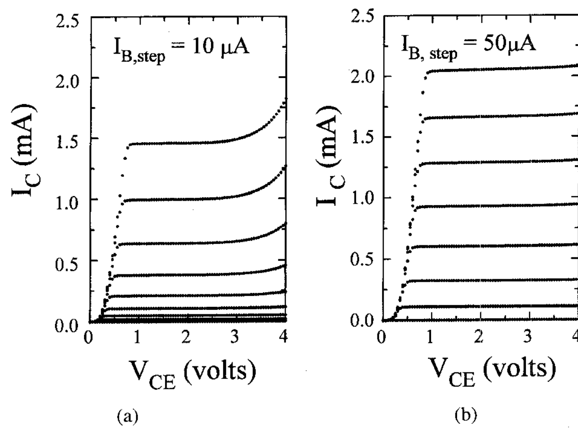

Fig. 3. Room-temperature common-emitter characteristics of the (a) 550°C $Si_{0.75}Ge_{0.25}$ and (b) $Si_{0.743}Ge_{0.25}C_{0.007}$ devices with emitter areas of 900 μm^2.

Carbon fractions of 0.1%, 0.7%, and 1.1% were extracted from the wafers with carbon, assuming that all carbon atoms were substitutional on lattice sites and that one substitutional carbon atom compensates 8.3 germanium atoms [2]. These carbon fractions correspond to strain compensations of 3%, 23%, and 37%, respectively. A Ge fraction of 20% and carbon fractions of 0.45%, 0.55%, and 0.9% were extracted from the X-ray data for the wafers with bases grown at 625°C.

Large area devices were fabricated using a simple double mesa process. The base was revealed by a selective wet etch which defined the emitter-base junction area, while the collector-base junction was defined by dry etching a larger mesa around the emitter-base mesa. While this process leaves the emitter-base surfaces unpassivated, leading to large, non-ideal base currents, the process is adequate for measuring the ideal collector currents needed to measure changes in bandgap. Devices are easily and quickly fabricated by this process, and it has been used to demonstrate high frequency SiGe HBT's [5].

III. Results and Discussion

Fig. 3 shows the room-temperature common-emitter characteristics for both the 550°C $Si_{0.75}Ge_{0.25}$ and the $Si_{0.743}Ge_{0.25}C_{0.007}$ devices with emitter areas of 900 μm^2.

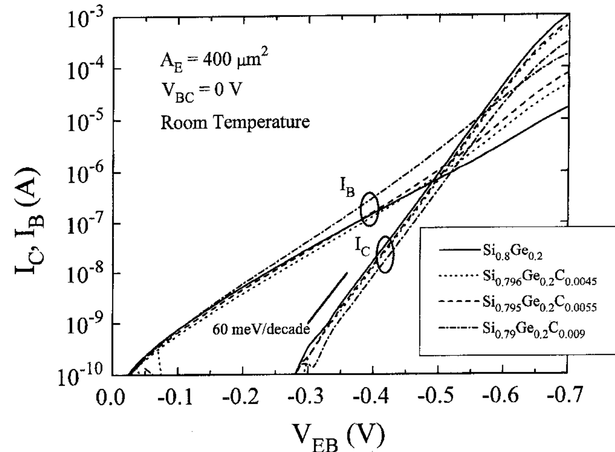

Fig. 4. Room-temperature Gummel plots of the 625°C control device and the 0.45%, 0.55%, and 0.9% carbon devices.

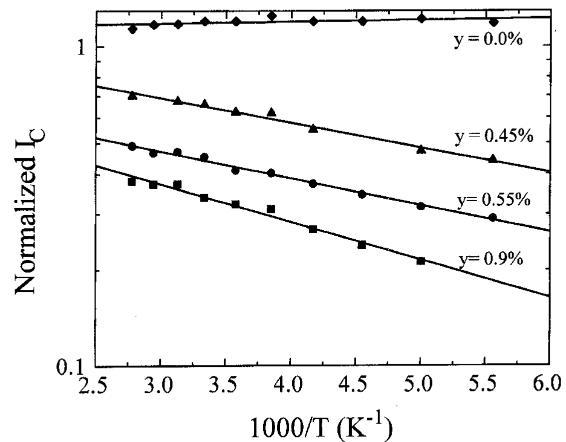

Fig. 5. Normalized collector current, ($I_{C,\ SiGeC}/I_{C,\ SiGe}$) versus inverse temperature for the 625°C devices with 0.0%, 0.45%, 0.55%, and 0.9% C.

The control device shows a maximum h_{fe} of 40 while the 0.7% carbon device had a gain of 7.5. The collector-emitter breakdown voltage was ~15 V with $I_B = 0$ (and $I_C = 0$), but decreased to ~6 V with $I_C \sim 2$ mA, reflecting the fact that the gain was a function of bias. The large Early voltages (>300 V) indicate the absence of parasitic barriers which would be caused by base dopant outdiffusion [6], [7]. From the curve-tracer figure, it is not possible to determine at fixed V_{BE} whether the decrease in gain between the control sample and the sample with carbon is due to a decrease in collector current or an increase in base current.

Fig. 4 shows room-temperature Gummel plots for the 625°C control device as well as for the 0.45%, 0.55%, and 0.9% C devices for $V_{BC} = 0$. Note that the collector current has a slope of 60 meV/decade at room-temperature, as expected for ideal collector currents in a bipolar transistor. The collector currents of devices with different perimeter to area ratios scale with emitter-base area. As the carbon level is increased, note the slight decrease in I_C. Assuming no change in conduction and valence band density of states, base minority-carrier diffusion coefficient, and the base Gummel number, this shift indicates

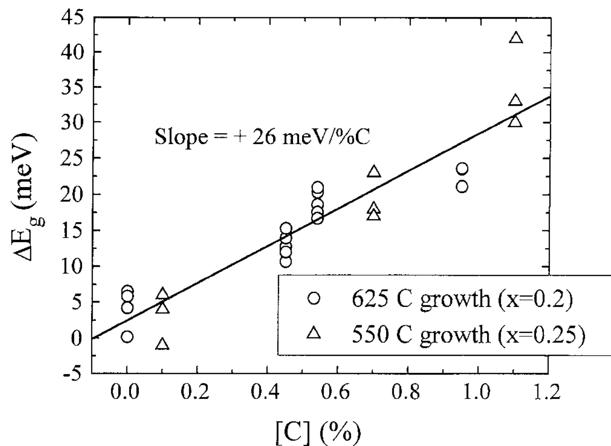

Fig. 6. ΔE_g as a function of carbon fraction for 625°C and 550°C devices and the best linear fit to all devices measured.

an increase in base bandgap as carbon is added. The base currents are nonideal below, and measurements of $Si_{1-x}Ge_x$ devices with different perimeter to area ratios indicated that this base current scaled roughly with emitter-base perimeter. This should be expected due to the unpassivated transistor structure. In addition, above $V_{BE} = 0.5$ V, the devices show a clear increase in base current with increasing carbon fraction, leading to a decrease in transistor gain. The origin of this base current is under further study and may be related to decreasing minority carrier lifetime in the $Si_{1-x-y}Ge_xC_y$, possibly due to an impure methylsilane source. The Gummel plots for the 550°C base samples were similar, although I_B was somewhat higher at low current levels.

It is well known that the difference in bandgap between the base regions of two HBT's may be extracted from the dependence of their collector currents on temperature [8]. The ratio of collector currents in HBT's with different base compositions is

$$\frac{I_{C_1}}{I_{C_2}} = \frac{N_{G_2}}{N_{G_1}} \frac{(N_c N_v)_1}{(N_c N_v)_2} \frac{D_{n_1}}{D_{n_2}} e^{\Delta E_A/kT} \tag{1}$$

where N_{Gi} is the base Gummel number, $(N_c N_v)_i$ is the product of the effective conduction and valence band densities of states, and D_{ni} is the electron diffusion coefficient. If there are no parasitic barriers in the conduction band, such as those resulting from conduction band offsets (ΔE_c) in abrupt junction devices without undoped spacer layers [9], or those resulting from base outdiffusion into the emitter and collector layers [10], [11], then the activation energy is the change in bandgap between the two materials, ΔE_g. This no-barrier assumption is justified in our case since $\Delta E_c \sim 0$ in the Si/strained $Si_{1-x}Ge_x$ system [12] and because the undoped spacers in our devices would prevent any barrier formation from small amounts of boron diffusion or due to any small ΔE_c induced by carbon. The 550°C devices' high Early voltages (>300 V) and identical collector currents in forward and reverse mode support this assumption.

Forward-active Gummel plots were measured from 180 to 360 K. Fig. 5 shows the ratio of $I_{c\,SiGeC}/I_{c\,SiGe}$ at $V_{BE} = 0.55$ V as a function of inverse temperature for the 625°C transistors. Assuming a similar temperature dependence of electron diffusion coefficients and densities of states between the two samples, ΔE_g was then extracted from the slope of the lines. As seen from the graph, carbon incorporation clearly leads to an increased bandgap. Fig. 6 shows the change in bandgap versus carbon percentage for devices with both $x = 0.2$ and $x = 0.25$. The best fit is +26 meV/%C. This data is much more reliable than our preliminary results showing a near zero dependence on C [13]. In this work, a wider temperature measurement range was used, the temperature of the cold finger was measured more accurately, and higher quality epitaxial layers led to better collector current ideality factor (n ranged from 0.995 to 1.005 for all carbon fractions over the entire temperature range). Also note that there is no substantial difference in bandgap between the 625°C and 550°C films.

The effect of C on the bandgap of pseudomorphically strained $S_{1-x-y}Ge_xC_y$ alloys has been measured by photoluminescence (PL) in the range of $x = 0.24$–0.38 and $y = 0.0$–0.011 [14] and in the range of $x = 0.155$ and $y = 0.0$–0.0085 [15]. The measured changes in bandgap were 21–24 meV/%C. That our transport measurements and the PL measurements yield a similar number is significant, since in SiGeC strong relaxation around C sites is known to occur [16]. This relaxation has been theoretically associated with a reduced bandgap [17]. If the bandgap was locally lower near C atoms, PL (which measures the recombination energy of an isolated exciton) might measure a lower bandgap than HBT transport measurements, where carriers must move through materials. That this did not occur suggests that the bandgap is spatially uniform.

As measured in our X-ray results, 1%C leads to a strain reduction of 0.003 in SiGeC films lattice-matched to Si (100). Achieving such a strain reduction by simply reducing the Ge concentration (by ~8%) would cause a bandgap increase of ~60 meV. Adding C (and keeping Ge fixed) to reduce the strain, however, increases the bandgap by less than half this amount. Therefore, to achieve a given bandgap less than that of Si, SiGeC alloys will have less strain than SiGe alloys. Furthermore, an increased thermal stability has been observed by structural and photoluminescence measurements [14]. Therefore, SiGeC films may have technological uses. Finally, note that if the bandgap versus C dependence were linearly extrapolated to 3%C a fully-compensated strain-free film of $Si_{0.72}Ge_{0.25}C_{0.03}$ would be predicted to have a bandgap about 120 meV lower than that of Si. Further work will be needed to justify the validity of such an extrapolation, however.

IV. Conclusions

We have demonstrated the first Si/$Si_{1-x-y}Ge_xC_y$/Si n-p-n HBT's and have used these devices to make the first electrical bandgap measurement of strained $Si_{1-x-y}Ge_xC_y$ on Si (100). For $x = 0.2$ and $x = 0.25$, a bandgap shift of +26 meV%C has been found for carbon fractions up to 1.1%C. SiGeC alloys will have a lower strain than SiGe alloys with a similar bandgap, and hence may find technological applications.

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