

Highly tensile strained silicon–carbon alloys epitaxially grown into recessed source drain areas of NMOS devices

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Abstract

Growth of high quality $\text{Si}_{1-y-z}\text{C}_y\text{P}_z$ (SiCP) with high substitutional carbon levels has been demonstrated. Growth rates up to 82 nm min^{-1} (at 550°C) led to films incorporating up to 3.6% substitutional carbon $[\text{C}]_{\text{VL}}$ according to Vegard's law (VL) or 2.8% $[\text{C}]_{\text{KB}}$ after Kelieres/Berti (KB). The films were fully strained with no sign of relaxation and displayed stress well above 2 GPa. Optionally, these films can be heavily doped with P, which then results in resistivities as low as $0.5 \text{ m}\Omega \text{ cm}$. The α/epi growth rate ratio was optimized to 1. This high throughput SiCP epitaxial growth process is an excellent candidate to be combined with a selective etch process. Used with repeated deposition/etch cycles, the final result will be identical to a true selective epitaxial growth process, but without any loading effect on $[\text{C}]$ or $[\text{P}]$ due to the blanket deposition steps.

(Some figures in this article are in colour only in the electronic version)

1. Introduction

$\text{Si}_{1-y}\text{C}_y$ (SiC) alloys are complementary in nature to the $\text{Si}_{1-x}\text{Ge}_x$ system. Due to the smaller lattice constant of C relative to Si, pseudomorphic $\text{Si}_{1-y}\text{C}_y$ films are under tensile strain. Grown into recessed source/drain areas, tensile stressed $\text{Si}_{1-y}\text{C}_y$ films can induce a tensile strained Si channel which enhances electron mobility [1]. This can be achieved either by true, conventional selective epitaxial growth (SEG) or by a combination of a non-selective deposition process followed by a selective removal of undesired material. A non-selective (blanket/mixed) deposition is a combination of an epitaxial film growth over the exposed crystalline silicon areas of the source and drain, while amorphous or polycrystalline deposition occurs over the dielectric areas. A selective etch removes the amorphous material from the insulator, leaving only the epitaxial film in the source and drain areas. To make such an approach production worthy, two requirements are essential: (1) a high growth rate with a low amorphous to epitaxial growth rate ratio, and (2) a high etch rate with a high amorphous to epitaxial etch rate ratio. In this paper we focus on

the deposition process on recesses formed in Si (100) substrates.

Epitaxial growth of SiC alloys is very challenging due to the large lattice mismatch between diamond and silicon, the low solubility of C in Si and the tendency of $\text{Si}_{1-y}\text{C}_y$ to precipitate. In order to grow metastable pseudomorphic films beyond the critical thickness and to avoid silicon carbide precipitation, non-equilibrium growth conditions (low growth temperature, high growth rate) are necessary [2]. On the other hand, thermal decomposition of the typical silicon precursor gases decreases strongly with decreasing temperature [3, 4]. These two contradictory conditions limit the possible process window of operation and the best choice of the precursor.

In situ doped SiCP films with very low film resistivity allow us to significantly reduce external transistor resistance. Additionally, a high level of electrically active (substitutional) P contributes to the tensile stress [5]. *In situ* doping eliminates the need for dopant implantation and subsequent dopant activation anneal that could initiate C precipitation and thus reduce stress. Due to the sensitivity of highly metastable (highly strained) SiC films, extensive post-epi implantation and thermal treatment are undesirable and should be avoided.

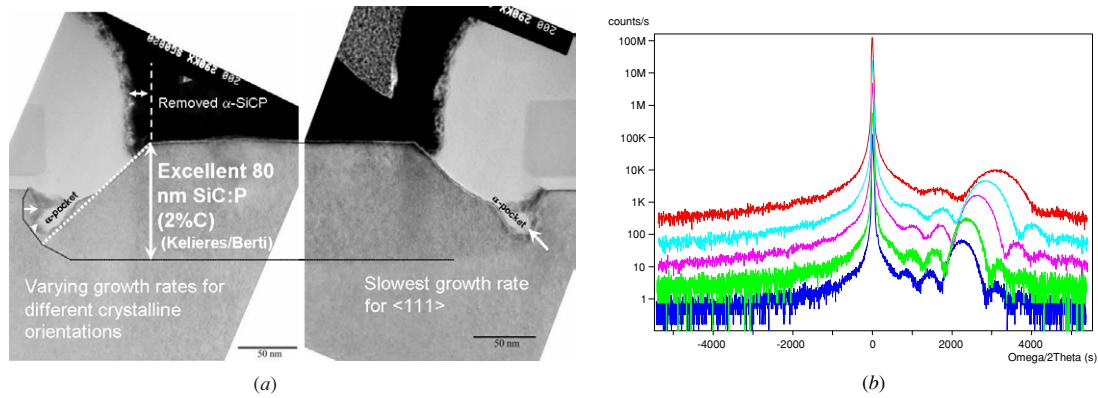


Figure 1. (a) XTEM of $\text{Si}_{0.97}\text{C}_{0.02}\text{P}_{0.01}$ in a RSD area. (b) XRD of SiCP films with $[\text{C}]_{\text{VL}}$ of 2.5–3.6%.

Junction implant and activation treatment should therefore be performed before the epi growth process.

It has been shown that SiC improves the contact resistance by $\sim 40\%$ due to the reduced SiC Schottky barrier height. The energy band gap for SiC material is smaller than that for Si [6]. Under the Fermi pinning assumption, this translates to an equivalent reduction in the Schottky barrier height which benefits the contact resistance and overall transistor access resistance. Moreover, P diffusion is suppressed in the presence of substitutional carbon and very abrupt dopant profiles can be achieved and maintained during further thermal processing [7]. In contrast, when As is used as an n-type dopant [8], dopant diffusion is enhanced in the presence of C. Phosphorous should then be the dopant of choice for SiC strained NMOS transistors.

2. Experimental details

Intrinsic, As and P doped SiC layers were grown at 550°C in an Epsilon[®] reduced pressure (RP) CVD epitaxial deposition tool manufactured by ASM. The Epsilon is a horizontal, single wafer, load-locked reactor, featuring a lamp heated and rotated silicon-carbide coated graphite susceptor in a cold wall quartz tube. The silicon, carbon and phosphorous source gases used in the described experiment were Silcore[®] (proprietary ASM version of specifically purified and packaged Si_3H_8), (mono-) methylsilane (20% MMS in H_2) and AsH_3/PH_3 (1% in H_2), respectively.

3. Results and discussion

Doped and undoped SiC films were epitaxially grown at 550°C with growth rates up to 82 nm min^{-1} . Due to the surface segregation behaviour [9] of C, As and P, the incorporation depends strongly on growth rate and temperature. The carbon is mainly incorporated into substitutional sites, with very little interstitial carbon, resulting in extraordinarily high carbon substitutionality. This was achieved with a sufficiently high growth rate and sufficiently low growth temperature, as predicted by Osten *et al* [10].

One challenge, particularly with P-doped SiC films, is the crystalline orientation dependent growth rate of SiCP. Figure 1 clearly shows that a reduction of the growth rate in the

(1 1 1) direction occurs. There is some limited growth on the vertical sidewall that pinches off a certain volume, forming an amorphous pocket in the corner of the recess. There are some indications in the literature that P can poison a (1 1 1) growth surface [11] which could also contribute to a reduced growth rate along the surface. A proper recess shape (isotropic versus anisotropic etch, recess depth) and pre-epi surface treatment (e.g. HF-last, *in situ* bake, chemical vapour etching (CVE), Si-seed layer) are considered important steps to obtain good quality epitaxial SiC(P).

High-resolution x-ray diffraction measurements of the epi layers grown on bare Si substrates reveal up to 3.6% substitutional carbon $[\text{C}]_{\text{VL}}$ according to the commonly used Vegard's law (linear interpolation between Si and diamond) or 2.8% $[\text{C}]_{\text{KB}}$ according to the relation of Kelieres/Berti [12]. In this paper Vegard's law is used to extract the carbon concentration unless stated differently. This corresponds to a biaxial stress well above 2 GPa. Pronounced interference fringes suggest that no relaxation of the films has occurred (figure 1(b)). The extracted values are shown in figure 2 (open symbols).

Figure 2(a) shows that the growth rate of the SiCP is essentially constant and independent of the MMS flow. Figure 2(b) shows that the level of substitutional carbon increases monotonically with increasing MMS flow. Figures 2(c) and (d) show film resistivities as low as $0.5\text{ m}\Omega\text{ cm}$, depending on the MMS flow and the substitutional carbon concentration in the film, respectively. The open symbols indicate twice the Silcore flow compared to the full symbols. With increasing carbon concentration levels the film resistivity goes up, most likely because the carrier mobility goes down due to various scattering mechanisms. The large size difference of C and Si atoms and the large bond-length difference between Si–C and Si–Si bonds cause a significant displacement of the four nearest neighbours of the C atom. This might cause an increase in resistivity even for fully substitutional carbon atoms in the crystal lattice. With increasing carbon concentration, the likelihood of forming C–C bonds also goes up. The [P] dopant levels ranged from the low to mid 10^{20} cm^{-3} .

Figures 3(a) and (b) show the effect of pressure and H_2 carrier flow on substitutional carbon incorporation and growth rate. The level of substitutional carbon is affected by both the total pressure and the H_2 carrier gas flow. The growth

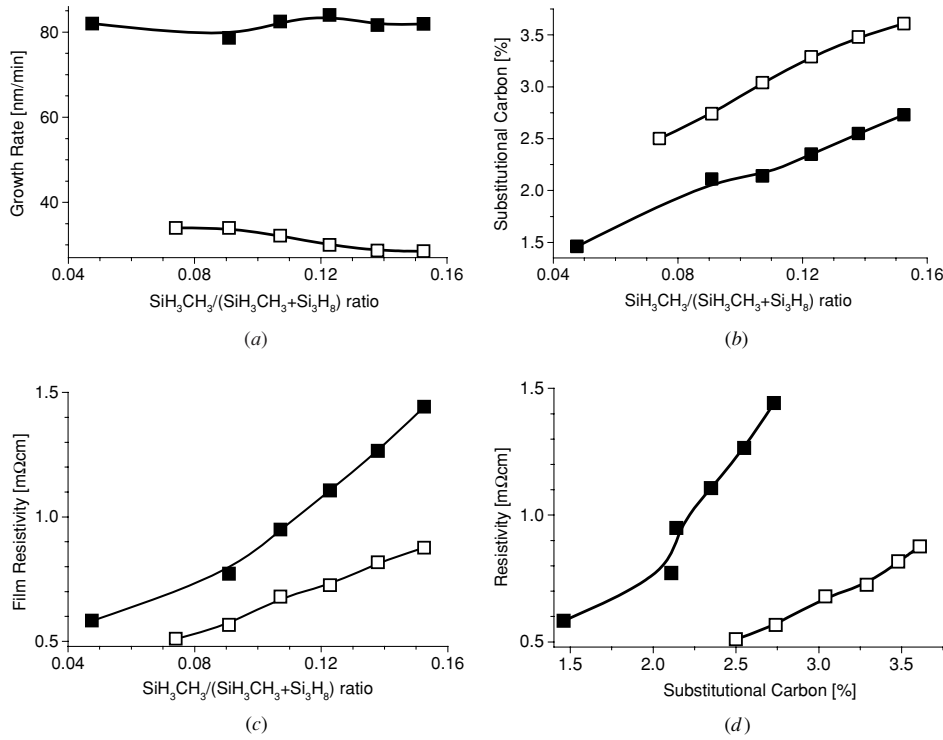


Figure 2. (a) Growth rate as a function of gas ratio. (b) Substitutional $[\text{C}]_{\text{VL}}$ as a function of gas ratio. (c) Film resistivity as a function of gas ratio. (d) Resistivity as a function of substitutional $[\text{C}]_{\text{VL}}$ content.

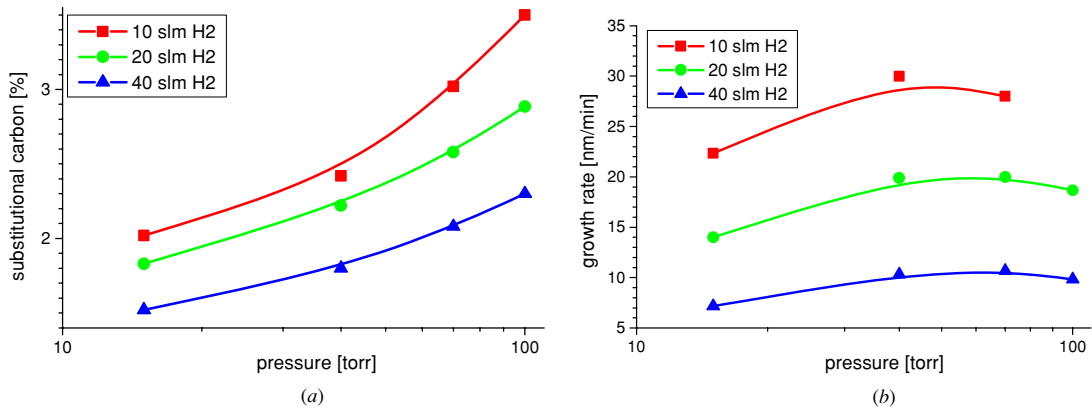


Figure 3. Dependence of substitutional carbon $[\text{C}]_{\text{VL}}$ and GR on pressure and H_2 carrier flow for fixed Silcore flow.

rate is mostly affected by the H_2 dilution, gas velocity and residence time in the reactor, but is not affected by the total system pressure.

Figures 4(a) and (b) show the substitutional carbon as a function of the growth rate. In order to vary the growth rate, only the total amount of MMS, Silcore and (Arsine) was varied, while the gas composition (MMS/Silcore/(Arsine)-ratio) was kept constant throughout a series. Again higher MMS/Silcore ratios result in higher substitutional carbon levels. A higher pressure results in a higher substitutional carbon level. It is worth mentioning that for growth rates less than 5 nm min^{-1} , the substitutional carbon level falls off, but saturates for higher growth rates at a level determined by the gas flow ratio and the total system pressure. In the case of the As doped sample (Figure 4(b)) the slope is somehow different due to a

surfactant effect of As that enhances the carbon incorporation and improves film quality compared to the undoped case. For the lowest growth rate, a sharp increase in the resistivity is observed, associated with a rapid degradation in the film quality. The last statement regarding the film degradation with low growth rates holds for all experiments described above.

Figure 5 shows the result of a variation in growth temperature with otherwise fixed process conditions. An increase of the temperature slightly increases the growth rate and increases the level of substitutional carbon. However, the quality degrades quickly and the critical thickness before silicon carbide precipitation occurs becomes very small. A decrease in temperature lowers the observed substitutional carbon, which is somehow counterintuitive. As shown in the graph, the unexpected drop in substitutional carbon can be

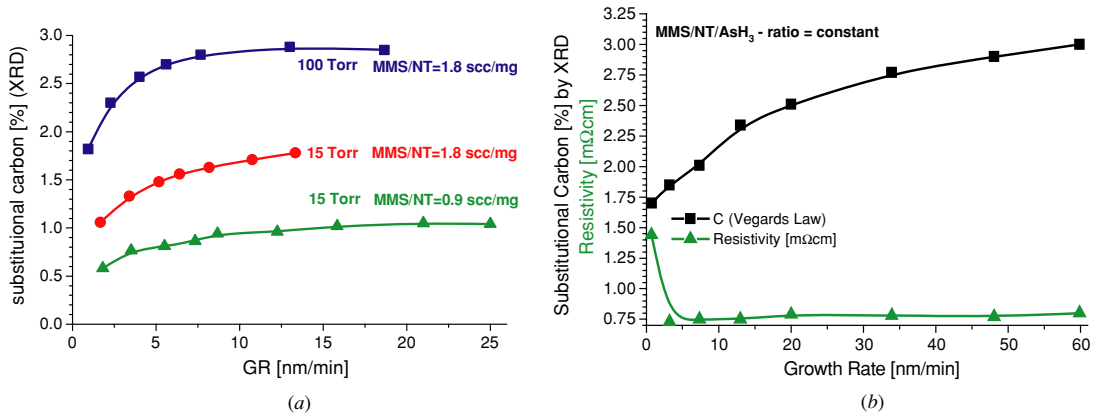


Figure 4. Substitutional carbon incorporation $[C]_{VL}$ as a function of GR, (a) undoped, (b) [As] doped.

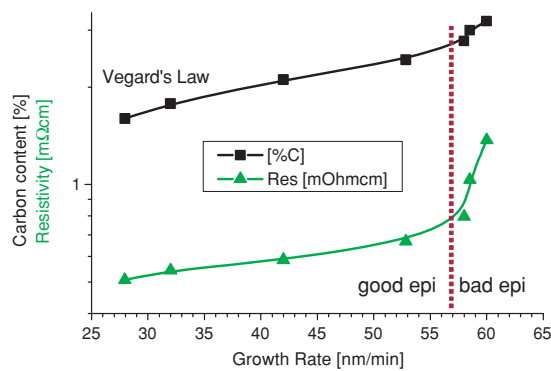


Figure 5. Substitutional carbon $[C]_{VL} = f(GR)$; $GR = f(T)$; from left to right: $T = (520 \text{ to } 580 \text{ } ^\circ\text{C})$ increased in $10 \text{ } ^\circ\text{C}$ increments).

explained by a simple growth rate effect. In the region of good epitaxial growth ($520\text{--}550 \text{ } ^\circ\text{C}$), an increased growth rate results in an increased level of substitutional carbon. However, at lower temperatures even higher levels of substitutional carbon can be achieved by simply increasing the MMS flow (not shown in the graph).

Figure 6 shows FTIR absorption spectroscopy of SiC samples with 1.8% $[C]_{VL}$, and thicknesses of 20 to 208 nm. All samples present an absorption peak at about 607 cm^{-1} , which is due to the localized vibrational mode (LVM) of substitutional carbon in Si indicating 100% substitutionality. No β -SiC precipitation, which would yield an absorption peak near 800 cm^{-1} , was found for these samples. The localized LVM at 607 cm^{-1} can be observed as well by Raman spectroscopy (not shown). Another method that also works for heavily doped SiCP films is to compare total (chemical) carbon concentration $[C]_{tot}$ as measured by SIMS with the substitutional carbon content as determined from the perpendicular lattice constant measured by x-ray diffraction, extracting $[C]_{VL}$ or $[C]_{KB}$ using a known relationship [12].

Perfect amorphous deposition without crystallinity was confirmed by XRR on blanket oxide wafers. The α /epi growth rate ratio as determined by TEM was confirmed on patterned substrates by spectroscopic ellipsometry. In order to achieve local tensile strained SiCP in the recessed source/drain areas, the amorphous SiCP over the insulator has to be selectively removed by one of the following methods.

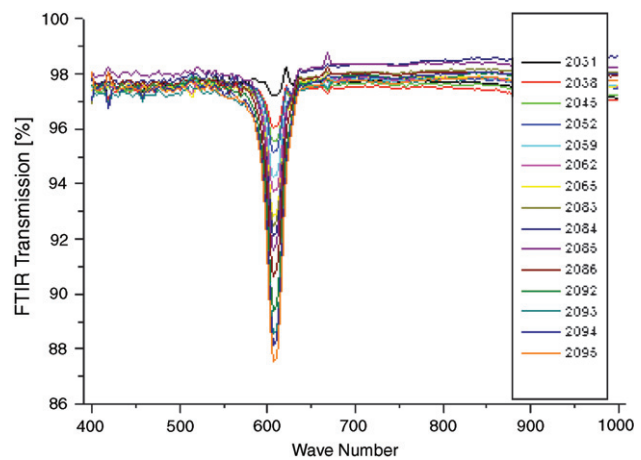


Figure 6. FTIR transmission spectra of undoped 20–208 nm thick SiC films with 1.8% substitutional carbon $[C]_{VL}$.

- (1) *In situ* post-epi (CVE) with appropriate etching chemistry (either one post-epi etch-step as shown in figure 1(a) or with repeated deposition/etch cycles [13]).
- (2) *Ex situ* wet chemical etch removal so far works only for undoped SiC films, but selectivity of the wet etch rate decreases with increasing [P] and [C] concentration.
- (3) Lithographic patterning with subsequent dry etch and/or CMP. This is an expensive, high risk, low yield process, only to prove the concept.

The α to epi interface (figure 1) that is formed might be a concern, since it is not clear yet how transmission of the high stress will be affected by the presence of an amorphous pocket, a micro twin or a stacking fault. Moreover, the silicidation process must be adjusted to these boundary conditions in order to avoid a penetration of the silicide in the junction area.

4. Conclusions

Fast growth of high quality SiC(P) with high substitutional carbon levels has been demonstrated. We have clearly shown how carbon segregation can be overcome and substitutional carbon concentration can be maximized with high growth rates. Growth rates up to 82 nm min^{-1} (at $550 \text{ } ^\circ\text{C}$) led to

films incorporating up to 3.6% substitutional carbon $[C]_{VL}$ according to Vegard's law; (2.8% $[C]_{KB}$ after Kelieres/Berti) and an α /epi growth rate ratio of 1. The films were fully strained with no sign of relaxation and displayed stress well above 2 GPa. Optionally, these films can be heavily doped with P, which then results in resistivities as low as 0.5 m Ω cm. Transmission electron microscopy of the P doped SiC films confirms that a high quality epitaxial growth of SiC(P) films has been achieved. The demonstrated epitaxial growth process is an excellent candidate to be combined with a selective etch process in repeated deposition etch cycles [13]. The final result will be identical to a true selective epitaxial growth process, but without any loading effect on [C] or [P].

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