

See discussions, stats, and author profiles for this publication at: <https://www.researchgate.net/publication/253725042>

Formation of buried layers of β -SiC using ion beam synthesis and incoherent lamp annealing

Article in *Applied Physics Letters* · December 1987

DOI: 10.1063/1.98953

CITATIONS

36

READS

13

5 authors, including:



[Karen Joy Kirkby](#)

The University of Manchester

306 PUBLICATIONS 3,371 CITATIONS

[SEE PROFILE](#)



[George Celler](#)

Rutgers, The State University of New Jersey

245 PUBLICATIONS 4,336 CITATIONS

[SEE PROFILE](#)

Some of the authors of this publication are also working on these related projects:



The Malthus Project [View project](#)



Computational Oncology [View project](#)

Formation of buried layers of β -SiC using ion beam synthesis and incoherent lamp annealing

K. J. Reeson and P. L. F. Hemment

Department of Electronic and Electrical Engineering, University of Surrey, Guildford, Surrey, United Kingdom

J. Stoemenos

Department of Physics, University of Thessaloniki, Thessaloniki, Greece

J. Davis

British Telecom Research Laboratories, Martlesham Heath, United Kingdom

G. E. Celler

AT&T Bell Laboratories, Murray Hill, New Jersey 07974

(Received 11 August 1987; accepted for publication 26 October 1987)

It is demonstrated that well-defined buried layers of β -SiC can be grown epitaxially within a silicon substrate. This structure is formed by implanting high doses of carbon ions ($> 3 \times 10^{17} \text{ C}^+ \text{ cm}^{-2}$) at 200 keV into a (100) single-crystal silicon which is maintained at a temperature of approximately 550 °C. During the subsequent anneal at 1405 °C for 90 min redistribution of the implanted species occurs, enabling the formation of a buried layer of β -SiC overlain by high-quality single-crystal silicon ($\chi_{\text{min}} = 4.1\%$).

Several groups have successfully formed discrete buried layers of SiO_2 and Si_3N_4 in silicon utilizing high doses of energetic oxygen or nitrogen ions followed by a high-temperature anneal.^{1,2} Attempts to form distinct buried layers of SiC using the same method have until recently singularly failed. This is thought to be due to the stability of the implanted carbon in substitutional sites³ which leads to higher activation energies being required for active chemical segregation to occur. Previous work⁴⁻⁸ on carbon-implanted silicon has indicated the presence of β -SiC and upon annealing at temperatures of up to 1200 °C has shown some lattice restructuring in the radiation-damaged silicon overlay. The redistribution kinetics typified in the O/Si and N/Si systems has, however, until now been unobserved.

Unlike buried layers of SiO_2 and Si_3N_4 , silicon carbide layers produced by ion beam synthesis do not have a direct application to current silicon on insulator (SOI) technology. Silicon carbide is, however, attracting interest as a wide-band-gap semiconductor which exhibits properties as a radiation hard substrate for use in high-temperature transistors. The ready synthesis of silicon carbide in the clean environment of an ion implanter and the fact that the implant parameters can be tailored to produce the optimum structure makes this method a desirable route for the fabrication of silicon carbide.

Device grade 3 in. (100) single-crystal silicon wafers were implanted with carbon ions to doses of 0.35×10^{18} , 0.72×10^{18} , and $0.95 \times 10^{18} \text{ C}^+ \text{ cm}^{-2}$ at an energy of 200 keV. The silicon substrates were maintained at a temperature of approximately 550 °C during implantation using ion beam heating. After implantation the wafers were capped with 3000 Å SiO_2 and annealed at 1405 °C for 90 min using an incoherent halogen lamp annealer.⁹

Figure 1 shows the nonchanneled Rutherford back-

scattering, (RBS) spectra, prior to annealing, for specimens implanted with 200 keV carbon ions to doses of 0.35×10^{18} , 0.72×10^{18} , and $0.95 \times 10^{18} \text{ C}^+ \text{ cm}^{-2}$, hereafter referred to as the low, medium, and high dose specimens and denoted by a, b, and c in Fig. 1. The dip in the silicon signal between channels 160 and 220 gives a qualitative approximation of the carbon depth distribution which in all three cases is skew Gaussian.

Figure 2 is an x-ray transmission electron microscopy (XTEM) micrograph of the high dose specimen prior to annealing; this shows seven well-defined regions from the surface. These are (1) 2550 Å of highly defective single-crystal (100) silicon; (2) 850 Å of defective (100) single-crystal silicon which shows contrast suggesting a laminar structure. Selected area diffraction from this region shows

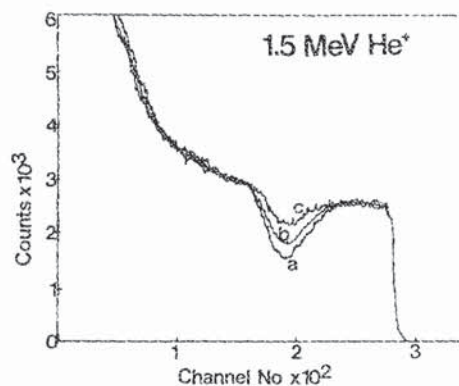


FIG. 1. Nonchanneled RBS spectra, prior to annealing, for the low, medium, and high dose specimens.

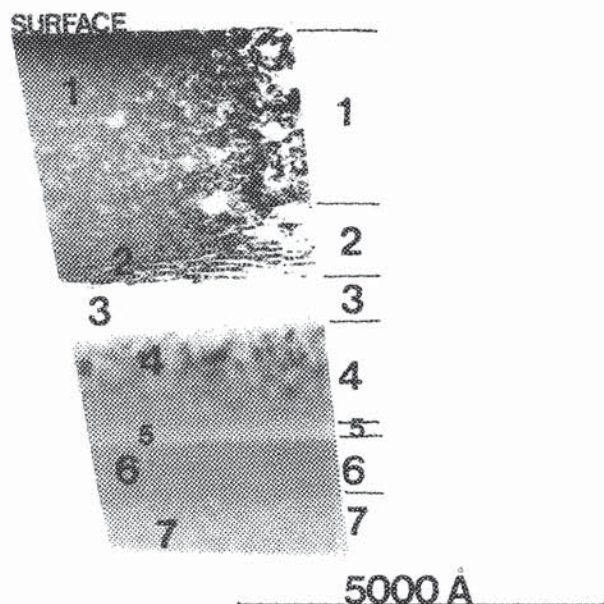


FIG. 2. Cross-section TEM micrograph for the high dose specimen, prior to annealing.

the manifestation of the laminar structure as an elongation in the silicon diffraction spots, (3) 600 Å of amorphous material, (4) 1300 Å of polycrystalline β -SiC, (5) 150 Å of amorphous material, (6) 780 Å of highly defective (100) single-crystal silicon, lying above, (7) the single-crystal silicon substrate. The asymmetry in the thickness of the two amorphous regions, (3) and (5), is consistent with the RBS results and suggests that the distribution of the implanted impurity ion is skew Gaussian.

Figure 3 shows the nonchanneled and channeled RBS results for the low and high dose specimens after annealing. The nonchanneled spectra, curves a and b of Fig. 3, show that for both doses the distributions are now essentially rectangular, saturating at a level commensurate with stoichiometric SiC. Examination of the channeled spectra, curves c and d of Fig. 3, shows them to have χ_{\min} values equal to 4.1% and 4.9%, respectively. These values are similar to

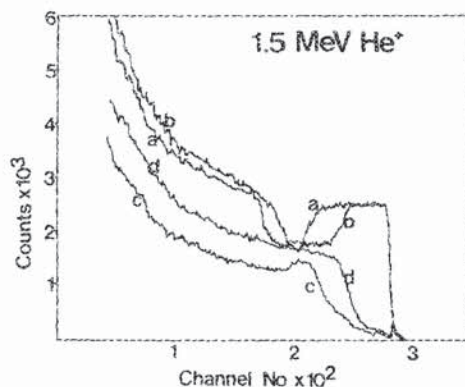


FIG. 3. Nonchanneled and channeled RBS spectra, after annealing at 1405 °C for 90 min, for the low and high dose specimens.

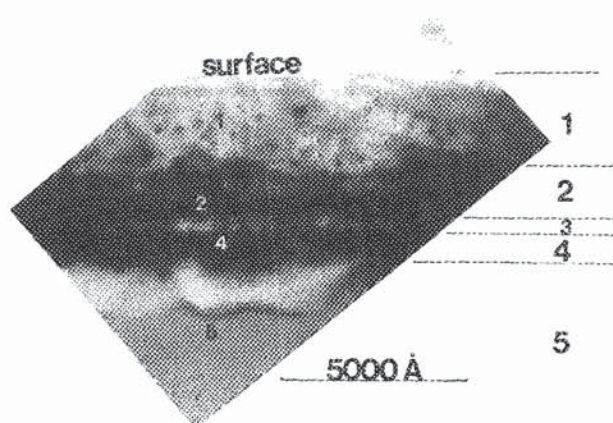


FIG. 4. Cross-sectional TEM micrograph for the high dose specimen, after annealing at 1405 °C for 90 min.

those for perfect single-crystal silicon and indicate that defect annihilation and lattice restructuring have occurred during the annealing process. At greater depths, however, the dechanneled fraction increases, indicating that the silicon overlayer is becoming progressively more defective.

The reason for the increased dechanneled yield in Fig. 3 curve d becomes apparent upon examination of an XTEM micrograph for this specimen in Fig. 4. This shows that in region (1) the silicon overlay contains small precipitates of β -SiC with the same general orientation as the silicon matrix which increase in both size and density with increasing depth. Associated with these precipitates are threading dislocations which likewise show an increased density with depth. Beneath the single-crystal silicon overlayer (1) the buried layer can be divided into three regions: (2) 1700 Å of polycrystalline β -SiC, (3) 350 Å of mixed polycrystalline β -SiC and silicon, (4) 1000 Å of large crystallites (> 1500 Å) of β -SiC growing epitaxially on the silicon substrate region (5). Selected area diffraction of the buried layer shows diffraction spots due to single-crystal β -SiC and diffraction rings due to polycrystalline β -SiC. These rings are composed of small spots due to the large crystallite size of the β -SiC.

To achieve the type of structures observed in Fig. 4 it is necessary to involve two different nucleation sites in the formation of the buried layer.¹⁰ It is postulated that these sites lie, respectively, at the peaks of the damage and range profiles of the as-implanted impurity depth distribution. Such phenomena have been previously reported by the authors^{11,12} for high dose oxygen and nitrogen implantation. The polycrystalline nature of region (2) in Fig. 4 can thus be explained as the nucleation of β -SiC in a highly damaged environment. At a depth commensurate with the peak of the range profile, however, region (4) of Fig. 4, the β -SiC is able to grow epitaxially using the silicon substrate as a seed.

Other analogies can be drawn between the processes observed in the O/Si system and those observed in the C/Si system, although the latter occurs at higher temperature. In the O/Si system,¹³ oxygen is removed from the silicon over-

layer into the buried layer by a process of thermally activated precipitate dissolution. For the C/Si System the topotactic relationship between β -SiC and Si means that the activation energy required to initiate precipitate dissolution is significantly greater than that in the O/Si system. In addition the lower diffusion coefficient of carbon in both SiC and Si with respect to oxygen in SiO₂ and Si also means that higher temperatures are required for active chemical segregation to be achieved.

Carbon, like implanted oxygen and nitrogen, can, upon annealing at sufficiently high temperatures, be made to undergo chemical segregation to achieve a distinct buried layer of β -SiC overlain by single-crystal silicon. The extreme temperature/time environment required for this process to occur can be explained by the stability of substitutional carbon within the silicon matrix and consequently the high activation energies required for precipitate dissolution. The production of buried layers of β -SiC growing epitaxially in a matrix of single-crystal silicon is one of the novel applications of ion beam synthesis and highlights the versatility of the technique in the field of semiconductor and device physics.

The authors would like to thank the staff of the Accelerator Laboratory, University of Surrey for their technical

assistance. They would also like to thank Mrs. B. Doré for typing this manuscript. The work is supported in part by the U.K. Science and Engineering Research Council (SERC).

- ¹J. Margail, I. Stoemenos, C. Jassaud, M. Dupuy, P. Martin, B. Blanchard, and M. Bruci, *les Éditions de Physiq. IV*, 519 (1985).
- ²P. L. F. Hemment, *les Éditions de Physiq. IV*, 475 (1985).
- ³I. H. Wilson, in *Ion Beam Modification of Insulators*, edited by P. Mazoldi and G. Arnold (Elsevier, The Netherlands, 1987) Chap. 7.
- ⁴J. A. Borders, S. T. Picraux, and W. Beezhold, *Appl. Phys. Lett.* **18**, 509 (1971).
- ⁵T. Kimura, S. Kagiya, and S. Yugo, *Thin Solid Films* **81**, 319 (1981).
- ⁶T. Kimura, S. Kagiya, and S. Yugo, *Thin Solid Films* **94**, 191 (1982).
- ⁷L. Kroko, I. Golecki, and H. L. Glass, *Proc. Mater. Res. Soc.* **45**, 323 (1985).
- ⁸K. J. Reeson, P. L. F. Hemment, R. F. Peart, C. D. Meekison, C. Marsh, G. R. Booker, R. J. Chater, J. A. Kilner, and J. Davis, *Radiat. Eff. (Special Edition)* **2**, 555 (1986).
- ⁹G. K. Celler, P. L. F. Hemment, K. W. West, and J. M. Gibson, *Appl. Phys. Lett.* **48**, 532 (1986).
- ¹⁰K. J. Reeson, P. L. F. Hemment, J. Stoemenos, J. R. Davis, and G. K. Celler, 5th Oxford Conference on the Microscopy of Semiconducting Materials, Oxford, April 1987, to be published in *Inst. Phys. Conf. Ser.*
- ¹¹K. J. Reeson, *Nucl. Instrum. Methods B* **19/20**, 269 (1986).
- ¹²P. L. F. Hemment, K. J. Reeson, J. A. Kilner, R. J. Chater, C. D. Marsh, G. R. Booker, G. K. Celler, and J. Stoemenos, *Vacuum* **36**, 877 (1986).
- ¹³J. Stoemenos and J. Margail, *Thin Solid Films* **135**, 115 (1986).