

# Fabrication of SiC Films on Si(100) using a C<sub>60</sub> molecular source

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*Indexing terms: Silicon carbide, Thick films, Vacuum deposition*

70nm thick SiC films have been fabricated on Si(100) wafers under ultrahigh vacuum conditions, by depositing C<sub>60</sub> molecules on substrates held at 900°C. The composition and morphology of the films have been measured using infra-red spectroscopy and atomic force microscopy.

**Introduction:** The search for applications for fullerene materials has been an ongoing effort, giving rise to several novel concepts [1-3]. In this Letter we report a new application of C<sub>60</sub> molecules as the source material for the growth of continuous SiC films on a silicon substrate held at 900°C under ultrahigh vacuum conditions. The importance of silicon carbide as a semiconductor material makes it the material of choice for fabricating semiconductor devices operating under adverse conditions [4]. A possible advantage of the fabrication method presented here is that the process involves only pure silicon and pure carbon in the form of C<sub>60</sub> molecules, with no need to introduce any impurities or toxic gas into the chamber.

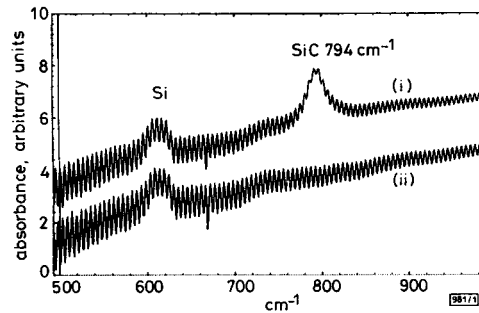


Fig. 1 Infra-red spectrum of a bare silicon sample, and the grown film, showing an absorption peak characteristic of a continuous α-SiC film

- (i) Si(100) + SiC film
- (ii) Si(100)

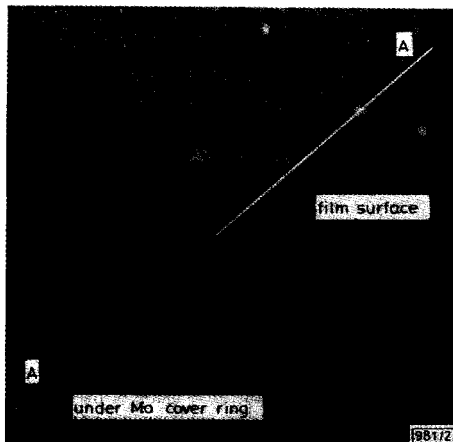


Fig. 2 AFM image of SiC films and Si substrate

Upper right-hand side: SiC film  
Lower left-hand side: Si substrate

**Experiment:** The SiC films were fabricated in an ultrahigh-vacuum chamber with a base pressure of  $5 \times 10^{-11}$  torr. Commercial *n*-type Si(100) (phosphorus doped,  $1.0 \Omega\text{cm}$ ) wafers were used as sample substrates. Clean Si(100) surfaces were prepared by annealing to 1200°C using electron beam bombardment on the back of the

sample. Pure C<sub>60</sub> molecules (99.8%) were sublimed at 350°C from a Knudson cell equipped with a BN crucible. The deposition rate was  $\sim 0.5$  monolayer C<sub>60</sub> molecules per minute. During the 1h C<sub>60</sub> molecule deposition, the substrate was held at 900°C, and the chamber pressure was kept below  $3 \times 10^{-9}$  torr. The sample was then removed from the UHV chamber and characterised by infra-red (IR) spectroscopy and atomic force microscopy (AFM) under ambient conditions. Fig. 1 displays two IR spectra of (i) a bare Si(100) sample, and (ii) the sample with the grown film. The absorption peak at  $793\text{cm}^{-1}$  corresponds to a continuous α-SiC film with crystallites having either disc or needle shapes [6]. Fig. 2 is a  $7 \times 7 \mu\text{m}^2$  AFM image of the sample, showing the area that was exposed to the C<sub>60</sub> source on the upper right-hand side, and the area under an Mo cover ring on the lower left-hand side. Fig. 3 is a cross-section along the line A-A shown in Fig. 2. A pronounced boundary between these two regions is observed at the centre of the image. Note that while the film is continuous, only homogeneously distributed clusters formed under the Mo cover ring. Note also the presence of deep pits between the clusters in the vicinity of the film boundary.

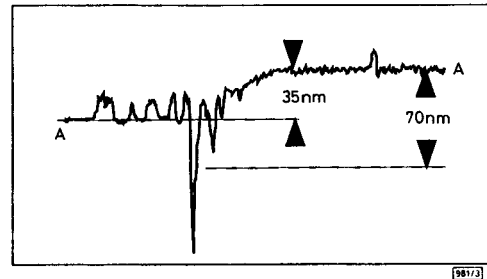


Fig. 3 Cross-section along line A-A shown in Fig. 2

**Discussion:** In previous scanning tunnelling microscopy, temperature programmed desorption, and Auger electron spectroscopy studies, it has been shown that C<sub>60</sub> adsorbates on a silicon surface decompose when annealing the sample to above 750°C [7-9]. It was found that at this temperature, the carbon atoms released from the decomposed C<sub>60</sub> molecules react with the Si atoms belonging to the substrate, and form SiC. An attempt to grow thick SiC films from room temperature grown multilayers of C<sub>60</sub> molecules and annealed to elevated temperatures failed because the molecules will desorb at elevated temperatures, leaving only a single monolayer behind. We have therefore deposited the C<sub>60</sub> molecules on a silicon substrate held at 900°C and obtained a film whose thickness can be estimated from the cross-section profile of Fig. 2. The height from the substrate surface to the top of the film surface is 35nm. However, the formation of the SiC film consumes substrate silicon atoms. Therefore, approximately half of the SiC film forms below the original substrate surface, yielding a thickness of  $\sim 70\text{nm}$ . Note that the SiC clusters formed under the Mo cover ring are probably due to C<sub>60</sub> molecules that diffused under it, and interacted with the silicon substrate to form SiC clusters [9].

**Acknowledgment:** We would like to thank D. R. Huffman and L. D. Lamb for helpful discussions and for providing the C<sub>60</sub> samples. This research is supported by the National Science Foundation, Air Force Office of Scientific Research, and Ballistic Missile Defense Initiative.

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3 May 1994

Electronics Letters Online No: 19940683

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## High-quality InGaAs/InP multiquantum-well structures on Si fabricated by direct bonding

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*Indexing terms: Gallium indium arsenide, Indium phosphide, Semiconductor quantum wells, Wafer bonding*

High-quality InGaAs/InP multiquantum-well (MQW) structures have been successfully fabricated on Si substrates by direct bonding. These structures were first grown on InP substrates, then bonded at 700°C on to Si substrates with buffer layers. The etch-pit densities of the InP surfaces are significantly low,  $\sim 10^6 \text{cm}^{-2}$ , the lowest values ever reported. Furthermore, strong relative photoluminescence intensity from the MQW structures, over 70% of that before bonding, is obtained.

**Introduction:** The successful integration of Si electronics and III-V optoelectronic functions would facilitate the fabrication of large-scale optoelectronic integrated circuits (OEICs). For applications to optical communications and interconnections, integration with long-wavelength devices such as lasers, detectors, and waveguides is favourable.

The main approach to fabricating III-V optical devices on Si has so far been heteroepitaxial growth. However, this approach suffers from several significant disadvantages, such as the large lattice mismatch which causes high density of threading dislocations, and the large difference in thermal expansion coefficients which leads to large residual thermal stress. Moreover, it has been reported that the residual threading dislocations are re-introduced from the III-V/Si interface due to thermal stress in the cooling stage [1].

Although much research has been carried out and a long-wavelength laser with excellent reliability has been demonstrated [2], the dislocation density in InP/Si systems remains in a much higher range (around  $10^7 \text{cm}^{-2}$ ) than required for high-performance laser operation.

On the other hand, a direct bonding technique has been developed as another approach to fabricating III-V optical devices on Si. This technique has already been applied in cases involving III-V/III-V systems, such as long-wavelength lasers on GaAs [3, 4]. Moreover, Lo *et al.* have demonstrated a high-performance InGaAs/GaAs MQW laser on Si [5]. However, no long-wavelength device has yet been fabricated on Si.

In this work, the bonding technique has been applied to fabricate long-wavelength, InGaAs/InP, multiquantum-well (MQW) structures on Si substrates. In our process, heteroepitaxially grown buffer layers on Si substrates are used to relax the thermal stress which may cause the propagation of threading dislocations through the bonded interface in the cooling stage.

**Experiment:** The samples used here were grown with a Varian Modular Gen II MBE system. Gaseous arsine and phosphine cracked at 950°C, as well as In, Ga, and Al metals, were used as sources. An InGaAs layer (0.3 μm) which acted as an etch-stop

layer during the InP substrate removal after bonding was first grown on an Fe-doped InP(100) substrate. An InGaAs/InP (5nm/20nm) multiquantum-well (MQW) structure, sandwiched between two InP layers (1.5 μm × 2), was then grown. In addition, a multiple-layer stacked buffer layer was heteroepitaxially grown on an  $Si(100)2^\circ \text{off} \langle 011 \rangle$  substrate. This buffer layer consisted of a GaP layer (0.3 μm), a GaAs layer (1 μm), an InAlGaAs step-graded buffer layer (1.3 μm), and finally, an InP layer (1.5 μm).

The InP surfaces of these two kinds of grown sample were mirror-polished to reduce surface roughness formed during epitaxial growth, cleaned with  $H_2SO_4:H_2O_2:H_2O$  solutions (3:1:1), and rinsed in de-ionised water. The samples were then placed with their mirror-polished sides in contact under a weight of  $\sim 300 \text{g/cm}^2$  and loaded immediately into an annealing furnace for bonding. They were annealed at 700°C for 1 h in  $H_2$  ambient. After the samples were bonded, the InP substrate as well as the InGaAs etch-stop layer were removed by conventional wet etching.



Fig. 1 Nomarski contrast photograph of an InP surface etched with  $HBr:H_3PO_4$  solution (1:2) for 8s, revealing a significantly low EPD of  $\sim 10^6 \text{cm}^{-2}$

**Results and discussion:** The etch-pit density (EPD) was measured by etching the InP surface with  $HBr:H_3PO_4$  solution (1:2) for 8s. A Nomarski contrast photograph of the etched surface is shown in Fig. 1, revealing a significantly low EPD of  $\sim 10^6 \text{cm}^{-2}$ ; this value is the lowest ever reported and is comparable to that of Fe-doped InP substrates used for MQW growth.

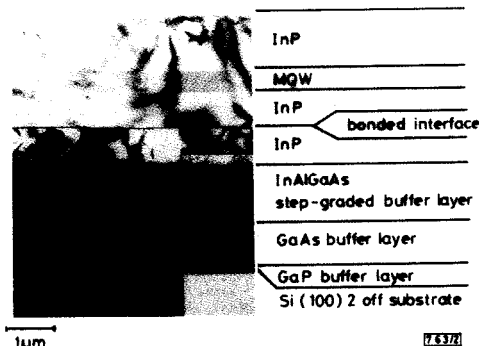


Fig. 2 Cross-sectional TEM image of same sample shown in Fig. 1

The dislocation propagation characteristics of the bonded structure were studied by transmission electron microscopy (TEM). Fig. 2 shows the cross-sectional TEM image of the same sample shown in Fig. 1. This observation confirms that an abrupt interface without any voids is formed in the bonding process. Furthermore, no defects were found in the bonded MQW layer over the whole region being inspected. These results indicate that the buffer layer, which is heteroepitaxially grown on the Si substrate, plays an important role in relaxing the thermal stress and avoiding the propagation of threading dislocations in the buffer layer (as high as  $\sim 10^8 \text{cm}^{-2}$ ) during the cooling stage.

To investigate the optical quality of the bonded MQW structures, room temperature photoluminescence (PL) measurements were carried out. The samples were excited using a YAG laser with a power density of  $\sim 10^3 \text{W/cm}^2$  on the sample surfaces. A typical