

# Highly-ordered graphene for two dimensional electronics

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With expanding interest in graphene-based electronics, it is crucial that high quality graphene films be grown. Sublimation of Si from the 4H-SiC(0001) (Si-terminated) surface in ultrahigh vacuum is a demonstrated method to produce epitaxial graphene sheets on a semiconductor. In this paper we show that graphene grown from the SiC(000 $\bar{1}$ ) (C-terminated) surface are of higher quality than those previously grown on SiC(0001). Graphene grown on the C-face can have structural domain sizes more than three times larger than those grown on the Si-face while at the same time reducing SiC substrate disorder from sublimation by an order of magnitude.

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An increasingly large effort is underway to create materials suitable for nanometer-scale electronic devices. One proposed avenue is to take advantage of the unique electronic properties of carbon nanotubes to make gates and ballistic conducting wires.[1] Challenges for such an approach are: Control of the properties of individual nanotubes (e.g. diameter, helicity), inherent heterojunction impedances associated with interconnection of nanotubes, and the assembly of vast networks from individual nanotube devices. Similar challenges are inherent to any approach that relies on preformed nanoscale objects. A more conventional means to achieve large-scale integration of nanoelectronic devices would be to rely on the continued scaling of lithographic techniques, which have been the semiconductor industry's greatest feat. Assuming such advances in lithography, a key issue is then the choice of material for nano-patterning. A suitable material should have excellent transport properties (e.g. mobility) and allow for control of electronic properties (band gap, doping) down to nanometer sizes.

It has been proposed that the unique electronic properties of carbon nanotubes could be obtained if graphene sheets were limited to nanometer-scale dimensions. [2] Recent experiments have demonstrated the unique electronic properties of graphene [3, 4, 5] thus charting a potential route to nanoelectronics based on epitaxial graphene (EG). [6] A requirement for further progress will be the preparation of a controlled number (thickness) of very large epitaxial graphene sheets on semiconductor substrates. These can be lithographically-patterned into narrow ribbons or other shapes, providing the necessary confinement for devices. Thus a scalable assembly of nano-patterned epitaxial graphene (NPEG) devices should be possible, with ballistic graphene conductors as interconnects.

Prior investigations of 6H- and 4H- SiC(0001) and (000 $\bar{1}$ ) surfaces showed that graphite films can be grown on these surfaces by sublimating Si from SiC during heating above  $\sim 1200^\circ\text{C}$  in ultrahigh vacuum (UHV).[7, 8, 9] These studies showed that graphite grows epitaxially on the (0001) Si-terminated (Si-face) surface of SiC, while graphite grown on the C-terminated (000 $\bar{1}$ ) (C-face) surface was rotationally disordered and under some conditions formed nanocaps instead of a smooth film.[10] Consequently, the C-face was initially overlooked as a potential substrate for graphene-based elec-

tronics.

In this paper we report that the classification of the C-face graphite as poorly ordered compared to the Si-face is incorrect. We instead show that the mean structural domain size on C-face graphite is much larger than on the Si-face, and that the inherent substrate roughness from sublimation is dramatically suppressed compared to the best previously-reported Si-face films. The improved structural order correlates with recent magnetotransport measurements, which show electron mobilities of over  $1\text{ m}^2/\text{V} \cdot \text{s}$  and coherence lengths exceeding  $1\text{ }\mu\text{m}$  in graphite films prepared on the C-face of SiC.[11] These observations have important implications for the science and technology of graphene.

All substrates were 4H-SiC purchased from Cree, Inc. [12] Prior to graphitization the samples were ultrasonically cleaned in acetone and ethanol, then hydrogen-etched at  $1600^\circ\text{C}$  for 30min. The  $\text{H}_2$  etching was done in a vacuum RF-induction furnace with a 200 ccm flow of 5%  $\text{H}_2/95\%$  Ar at  $\sim 1$  atm. This process removed all surface scratches and left a regularly stepped surface, as characterized by AFM. A typical step terrace width is  $\geq 1\text{ }\mu\text{m}$  with a step height of  $5\text{ }\text{\AA}$  (this corresponds to a  $\sim 0.03^\circ$  miscut).

Si-face 4H-SiC(0001) samples were prepared in UHV ( $P < 1 \times 10^{-10}$  Torr) by electron-bombardment heating. Substrates were first heated to  $1100^\circ\text{C}$  for 6 min, then to  $1320^\circ\text{C}$  for 8 min to remove surface contamination and form a well ordered  $\sqrt{3} \times \sqrt{3}$  reconstruction. They were subsequently heated to  $1400\text{--}1440^\circ\text{C}$  for 6–12 min to create graphene films 1–2 layers thick, as determined by x-ray reflectivity. C-face 4H-SiC(000 $\bar{1}$ ) samples were heated to  $1430^\circ\text{C}$  for 5–8 min in a vacuum RF-induction furnace ( $P = 3 \times 10^{-5}$  Torr). These parameters produced graphitic films of 7–13 graphene layers.

The x-ray scattering experiments were performed at the Advanced Photon Source, Argonne National Laboratory, on the 6ID-B- $\mu\text{CAT}$  beam line at 16.2 keV photon energy. The graphite film thickness for all samples was determined by measuring the x-ray intensity as function of  $\ell$  along the graphite  $(1, \bar{1}, \ell)_G$  rod.[8] The notation  $(h, k, \ell)_G$  identifies a reciprocal-space point in units of the graphite hexagonal reciprocal lattice basis vectors:  $a_G^* = 2.9508\text{ \AA}^{-1}$  and  $c_G^* = 1.8829\text{ \AA}^{-1}$ . Unsubscripted reciprocal-space coordinates  $(h, k, \ell)$  refer to the substrate 4H-SiC hexagonal recip-

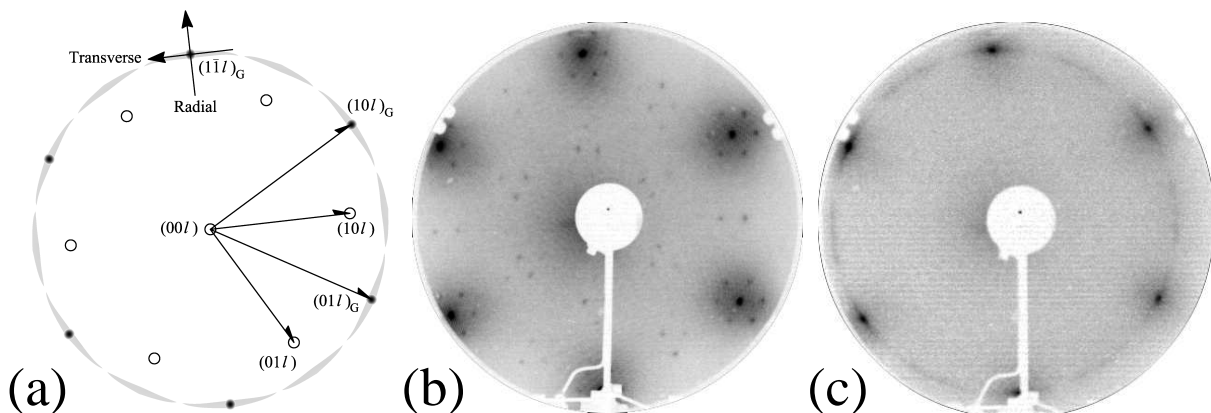


FIG. 1: (a) A schematic of reciprocal space for graphite on SiC. Graphite rods ( $\bullet$ ) are rotated  $30^\circ$  from the SiC rods ( $\circ$ ). Shaded regions represent the effect of orientational disorder in the graphite. Radial and transverse x-ray scan directions for Fig. 2 are depicted in (a). (b) and (c) show LEED images acquired at 75-eV electron energy: (b) a SiC(0001) Si-face surface with 2 graphene layers (UHV synthesis); (c) SiC(000 $\bar{1}$ ) C-face surface with 7 graphene layers (induction furnace synthesis).

rocal lattice units:  $a^* = 2.3552 \text{ \AA}^{-1}$  and  $c^* = 0.6230 \text{ \AA}^{-1}$ .

A reciprocal space schematic for epitaxial graphene on SiC is shown in Fig. 1(a). Open circles depict the  $(1 \times 1)$  low-energy electron diffraction (LEED) pattern from an unreconstructed SiC surface, while the filled circles are the  $(1 \times 1)_G$  pattern for a graphene or graphite film with hexagonal unit cell rotated azimuthally by  $30^\circ$  ( $R30^\circ$ ) relative to the SiC  $(1 \times 1)$  cell. The diffuse ring through the  $(1 \times 1)_G$  graphite spots in Fig. 1(a) represents the  $(1 \times 1)$  LEED pattern from a graphene film with rotational disorder relative to the SiC surface.

Figure 1(b) shows the LEED pattern obtained from a 2-graphene-layer film grown on the Si-face of SiC. In addition to the graphite and SiC  $(1 \times 1)$  patterns, it shows a complex  $6\sqrt{3} \times 6\sqrt{3}R30^\circ$  reconstruction, which is well known and has been studied extensively in the literature (see e.g. Refs. 9, 13 and references therein). The LEED pattern from a C-face sample with 7 graphene layers is shown in Fig. 1(c). The LEED only shows the 6-fold graphite pattern because the thicker C-face film attenuates the electron beam. The azimuthal streaks in Fig. 1(c) may indicate some rotational disorder in the graphene sheets (as previously observed[9]).

More detailed information on the structural order of the samples was acquired by surface x-ray scattering. We have measured the width of the graphite  $(00\ell)_G$  and  $(1\bar{1}\ell)_G$  rods on C- and Si-face samples. Figure 2 shows radial scans [see Fig. 1(a)] across the graphite  $(1, \bar{1}, 1.5)_G$  crystal truncation rod for both the Si- and C-face samples. The Si-face samples have a radial width of  $\Delta q_r \sim 0.022 \text{ \AA}^{-1}$  corresponding to a graphite mean coherent domain size[14]  $L = 2\pi/\Delta q_r \sim 290 \text{ \AA}$ . This is very similar to the graphite domain size observed by Charrier *et al.*[8] and is typical of the quality of graphite grown on the SiC(0001) surface reported to date in the literature. Although their surface treatment was different from ours, the fact that the domain sizes are similar suggests a limit on the graphite quality other than surface preparation.

In contrast, graphite grown on the C-face has much larger

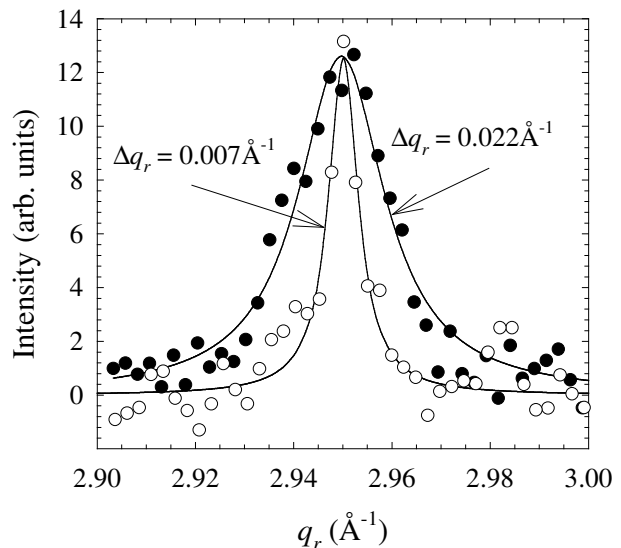


FIG. 2: Radial scans through the graphite crystal truncation rod  $(1, \bar{1}, 1.5)_G$  for both 2-layers of graphene grown on a Si-face sample ( $\bullet$ ) and 7-layers of graphene grown on a C-face sample ( $\circ$ ). Solid lines are Lorentzian fits.

domain sizes demonstrated by the smaller radial widths in Fig. 2:  $\Delta q_r \sim 0.007 \text{ \AA}^{-1}$ , corresponding to  $L \sim 900 \text{ \AA}$ . So while the LEED patterns show azimuthal disorder in the C-face graphite, the coherent graphite domains are more than 3 times larger than for the Si-face graphite. This improved structural coherence correlates with the high carrier mobility measured recently for C-face graphene[11] versus Si-face.[6] We note that the difference in film thickness may play a role in the long range order of the films. However, the C-face  $\Delta q_t$  do not change for films between 7–13 layers. There is also little difference in the long range order of 1–2 layer graphene films on the Si-face. In addition growth of 4–5 graphene layers on Si-face requires temperature above  $\sim 1500^\circ\text{C}$  where substrate disorder becomes problematic (see below).

Why are the C-face graphite films so much better than the

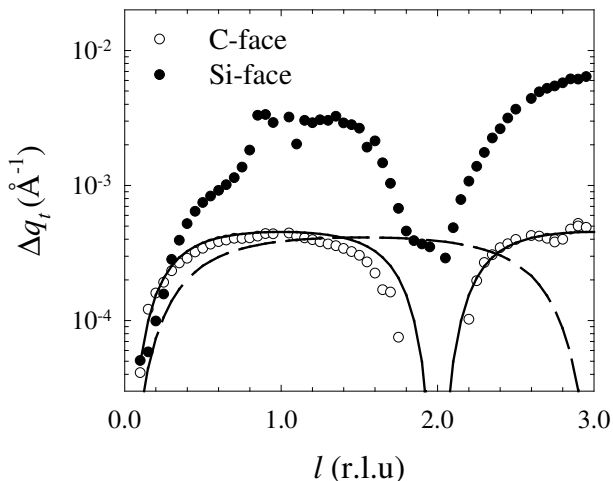


FIG. 3: Comparison of the FWHM ( $\Delta q_t$ ) of the  $(00\ell)_G$  rod vs.  $q_z = \ell c_{SiC}^*$  from (●) a 2-layer graphene film grown on the Si-face and (○) an 8-layer film grown on a C-faces 4H-SiC substrate. Instrument broadening has been removed for clarification. The lines are fits to a geometric distribution of steps and step heights with either the (solid line) 1/2 4H step height or the (dashed line) graphite step height.

Si-face films? While we are not able to explain the details of the growth mechanism leading to this difference, the x-ray data does point out a dramatic difference in the surface morphology of the SiC substrate after film growth. Transverse scans along the specular  $(00\ell)$  rod were taken on both C- and Si-face graphitized surfaces. The transverse peak widths  $\Delta q_t$  are plotted in Fig. 3 versus  $\ell$  (SiC units). These scans reveal a modulation of the width with  $\ell$  that is very different for the C- and Si-face graphitized surfaces. The peak-width modulation is due to atomic steps.[14] In this case it is due to steps on the SiC substrate and not steps in the graphite. We know this for two reasons. First, graphite steps would cause a width modulation of the graphite  $(1, \bar{1}, \ell)_G$  rod that is not observed.

Second, the modulation period  $\Delta\ell$  is inversely proportional to the step height:  $d_{step} = c_{SiC}/\Delta\ell$ . For both C- and Si-face samples,  $\Delta\ell$  corresponds to half of the 4H unit cell height (5.043Å) and not to the graphite step height (3.337Å) or any multiple. This is clearly demonstrated by the fits in Fig. 3. The fits are to a model of a geometric distributions of steps and step heights based on either the half 4H unit cell height or the graphite step height. It is clear that the graphite steps produce the wrong period.

The maximum width in Fig. 3 is inversely proportional to the mean distance between steps on the SiC substrate,  $D$ ,  $D = 2\pi/\Delta q_t$ . [14] Because the modulation amplitude is much larger for the Si-face sample, we conclude that the SiC step density is more than an order of magnitude greater than on the graphitized C-face (we note that the starting SiC step density before graphitization was nearly the same for both samples).

Quantitatively, the C-face samples show that the mean terrace width of the SiC substrate is  $D_C \sim 1.4 \mu\text{m}$ , while the Si-face samples have a terrace size of  $D_{Si} \sim 0.2 \mu\text{m}$ . The higher step densities observed after graphitization on the Si-face substrate correlate with the poorer long range order of

the graphite grown on this face. Whether this is a cause or effect relation, remains to be determined.

In conclusion we have shown that ultrathin films of well-ordered graphite (graphene) can be grown on the  $SiC(000\bar{1})$  surface via Si sublimation. In spite of a small orientational disorder, the long range order of the graphite on this surface is more than three times larger than previously reported for graphite growth on the  $SiC(0001)$  surface.

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