



Graphene nanoribbons on vicinal SiC surfaces by molecular beam epitaxy

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We present a method of producing a densely ordered array of epitaxial graphene nanoribbons (GNRs) using vicinal SiC surfaces as a template, which consists of ordered pairs of (0001) terraces and nanofacets. Controlled selective growth of graphene on approximately 10 nm wide (0001) terraces with 10 nm spatial intervals allows GNR formation. By selecting the vicinal direction of SiC substrate, [1 $\bar{1}$ 00], well-ordered GNRs with predominantly armchair edges are obtained. These structures, the high-density GNRs, enable us to observe the electronic structure at *K* points by angle-resolved photoemission spectroscopy, showing a clear band-gap opening of at least 0.14 eV.

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Graphene nanoribbons (GNRs) are attracting increasing attention in nanoelectronic applications and solid-state physics, where the band-gap opening or modification of the electronic structure at *K* points is a central interest.¹ The electronic structure at *K* points in GNRs has theoretically¹⁻³ and experimentally⁴⁻¹¹ been demonstrated to depend on the type of edge geometry: armchair or zigzag. Semiconducting characteristics are expected in the case of armchair edges owing to the band-gap opening at *K* points.² As the width of GNRs is reduced, the gap is increased by both electron confinement and edge effects. However, realization of GNRs with atomically well defined edges and providing experimental evidences of the gap opening at *K* points remain challenging. Here we demonstrate an approach for producing a densely ordered array of aligned GNRs, which is advantageous for investigating physical properties macroscopically, on unique SiC surfaces as templates via molecular beam epitaxy (MBE) and show band-gap openings at *K* points visualized by angle-resolved photoemission spectroscopy (ARPES). Fabrication of GNRs by a conventional lithographic technique was first reported by Chen⁴ and Han.⁵ GNRs as narrow as 15~20 nm were obtained and the band-gap openings were shown; the band-gap value was dependent on the ribbon width. Since then, new and more advanced approaches have been proposed for GNR fabrication. These include unzipping of carbon nanotubes (CNTs),⁷ chemical,⁸ sonochemical,⁹ and bottom-up molecular precursor methods,¹⁰ and selective thermal decomposition at facets on SiC substrates.¹¹ Each method is unique and all have advantages and disadvantages. The unzipping of CNTs and chemical methods are appropriate for mass production but in their present state require highly complicated treatment and lack sufficient quality control. The primary advantage of the molecular precursor method is that it enables very precise control of the width and edge structure; however, so far, it only works on metal surfaces. The thermal decomposition of SiC substrates, utilized in Ref. 11, is now recognized as a powerful method to obtain high-quality graphene with few structural defects¹² owing to the epitaxial characteristic of the initial carbon layer (buffer layer) with (6 $\sqrt{3} \times 6\sqrt{3}$)R30° (6R3) registry on SiC(1 \times 1).^{13,14} GNRs, approximately 40 nm

wide, were achieved by the thermal decomposition method with selective growth on facets created by reactive ion-etching on a SiC substrate. However, in this method, C atoms are supplied internally from SiC surface steps, and accordingly, unstable growth¹⁵⁻¹⁷ and morphological fluctuation¹¹ are induced. Although all these methods have succeeded in forming GNRs of different quality, the physical characteristics of the resulting GNRs, in particular evident and direct observations of band-gap opening, have not been demonstrated.

Here we report a new method for the fabrication of GNRs on a SiC substrate by molecular beam epitaxy (MBE). C atoms are supplied externally, which effectively addresses the problems described in Ref. 11. Growth selectivity is achieved through the preferential nucleation of GNRs on terraces of vicinal SiC surfaces. The off-axis SiC substrate, often used in the fabrication of SiC electronic devices, is especially noted here because it shows a unique periodic surface structure. We have studied off-axis (vicinal) SiC surfaces that are intentionally off-cut from a (0001) plane and found self-ordered periodic structures consisting of pairs of a (0001) basal plane terrace and a (1 $\bar{1}$ 0*n*) nanofacet (*n* = 35 ~ 37) with a characteristic periodicity of ~20 nm, as shown in Fig. 1. The structural model, illustrated in Fig. 1(b), was derived from high-resolution transmission electron microscopy images. The surface structure is formed by phase separation, quantized step-bunching, and ordering.¹⁸ Such a unique periodic SiC surface, hereafter SiC nanosurface, should be a good template for growing a massive array of GNRs.

In the present study, vicinal 6H-SiC substrates (Si face, 4° off toward [1 $\bar{1}$ 00]) first underwent H₂ gas etching to produce a SiC nanosurface. The resultant SiC samples were then loaded into an ultrahigh vacuum chamber, heated by direct current and cleaned under Si flux at 1050 °C. *In situ* reflection high-energy electron diffraction (RHEED) indicated that the initial broad ($\sqrt{3} \times \sqrt{3}$)R30° (*R3*) structure of native oxide was changed to clear (3 \times 3) of clean SiC(1 \times 1) with Si adlayers,¹⁹ which are free from carbon and oxide contaminations. After terminating the Si beam at the same temperature, the Si adatoms' *R3* structure appeared owing to the evaporation of Si atoms in the Si adlayers.²⁰ The Si adatom *R3* surface is important for

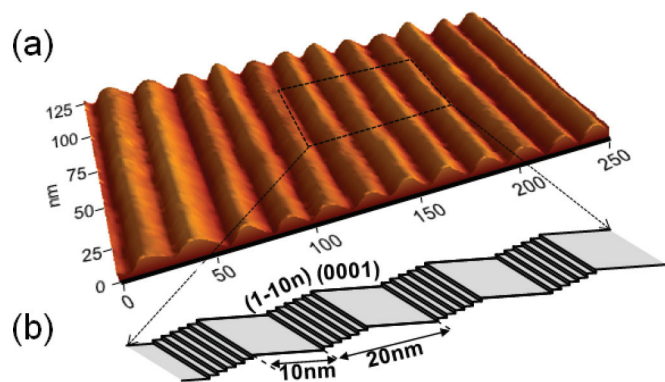


FIG. 1. (Color online) An AFM image and a structural model of SiC nanosurface. (a) An AFM 3D view of the vicinal SiC surface after hydrogen etching, showing periodic array of terraces and facet structures. (b) A schematic drawing of the SiC surface in (a). Each pair of array consists of the (0001) terrace and ($1\bar{1}0n$) facet and is ordered with the periodic distance of ~ 20 nm. The width of each (0001) terrace is ~ 10 nm.

achieving high-quality GNRs on SiC by MBE, because residual Si atoms of previous highly Si rich (3×3) structure interact with C atoms that are externally supplied, and induce undesired SiC growth, resulting in formation of wave-shaped terraces and fluctuant width of GNRs. After that, C atoms were supplied at a constant deposition rate by heating resistive carbon plates at 2200°C . The RHEED pattern was monitored *in situ* at 15 keV and used to terminate the growth. Three samples were prepared by changing the growth time: 5, 20, and 33 min. Each sample was evaluated by atomic force microscopy (AFM), RHEED, low-energy electron diffraction (LEED), and Raman spectroscopy. Finally, the samples were exposed to hydrogen at 600°C for 1 h to transform the $6R3$ structure into quasi-free-standing graphene by hydrogen intercalation.²¹ In addition, scanning tunneling microscopy (STM), polarized Raman spectroscopy, and ARPES were performed.²²

Figure 2(a) shows the RHEED image after 20 min growth. The electron beam was irradiated parallel to the vicinal direction, i.e., perpendicular to step edges [$1\bar{1}00$]. Three sets of diffraction patterns are visible in this image: $\times 1$ SiC surface, $\times 3$ due to $R3$, and satellites due to $6R3$. Figure 2(c) shows the evolution of RHEED intensity with growth time at selected positions, namely, a specular spot and $R3$ and $6R3$ streaks, as indicated in Fig. 2(a). The diffraction intensity of the $R3$ structure ($2/3$ streak) increased initially and decreased after ~ 5 min. The initial increase in the intensity is probably related to the structural change to the C-rich $R3$ surface. The detailed mechanism of structural transformation at this stage is not yet clear. We can speculate that the C atoms may incorporate into the Si-rich phase²³ while maintaining the same surface geometry of $R3$ but resulting in higher intensity of the $R3$ streaks. This structure is now under investigation with *in situ* LEED I-V analysis. As the $R3$ intensity started to decrease, $6R3$ streaks appeared and increased in intensity. This complementary relationship between the intensity profile of $R3$ and $6R3$ structures implies structural transformation from $R3$ to $6R3$ by the incorporation of C atoms. After 33 min growth, the transformation was complete; no $R3$ signal was visible. At this stage, the $6R3$ layer covered the (0001)

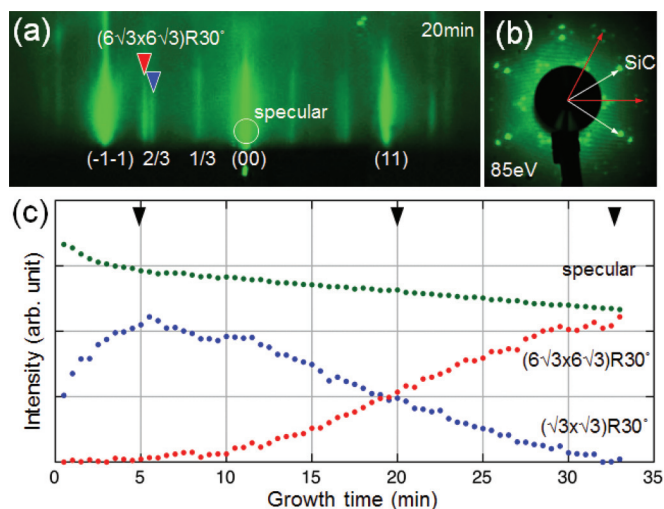


FIG. 2. (Color online) *In situ* RHEED analysis during MBE growth and a LEED image after growth. (a) A RHEED image after 20 min growth, where both $R3$ and $6R3$ structures are visible. (b) A LEED image after 33 min growth, indicating clear $6R3$ satellite spots. (c) RHEED intensity profiles as a function of growth time. The intensities of a specular spot, an $R3$ and a $6R3$ streak, as indicated in (a), are monitored. Note that the initial $R3$ intensity is increased after 5 min growth and the $R3$ and $6R3$ intensities show a complementary relationship.

terraces, as shown in the LEED image in Fig. 2(b). Note that no additional (overlap) spots except for SiC and the $6R3$ structure are seen in this image, indicating that growth takes place selectively on (0001) terraces. The Raman spectrum of the sample grown for 33 min typically indicates no G' (2D) signal [see Fig. 4(a)], which also supports the formation of the $6R3$ structure.

Next, each sample was hydrogen intercalated to obtain quasi-free-standing graphene on the SiC surface.²¹ LEED (not shown) clearly indicated the absence of $6R3$ satellites due to structural transformation and shows only superposition of clear (1×1) SiC and (1×1) graphene patterns. In the AFM height mode, a slight contrast was visible on the (0001) terraces of the sample after 20 min growth, as shown in Fig. 3(a). No contrast was observed before the hydrogen treatment. This contrast results from the difference in height between graphene (previously $6R3$) and the $R3$ regions. As the $6R3$ area transformed to graphene, its height increased owing to hydrogen intercalation at the interface, whereas the $R3$ area was probably etched in the hydrogenation process. The height difference is ~ 0.25 nm, which is in good agreement with the cross-sectional height of graphene on the H-terminated SiC surface.²⁴ This is seen more clearly in Fig. 3(c), which shows the AFM phase image contrast for the corresponding area²⁵ and is also evident from a comparison of cross-sectional height and phase profiles at the line A-B in Figs. 3(a) and 3(c). The sample grown for 33 min, i.e., covered with GNRs, shows a surface morphology that is very similar to that of the SiC nanosurface, as shown in Fig. 3(b), but areas of dark contrast at terraces are evident in the phase image, as shown in Fig. 3(d), owing to the formation of graphene. In XPS measurements (not shown here), the C 1s peak deconvolution clearly shows the presence of graphene on the surface and absence of any

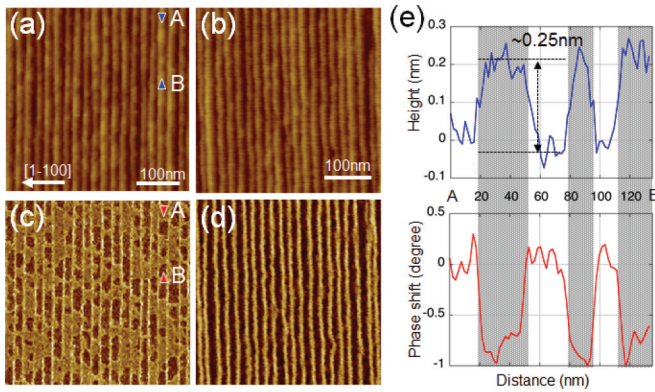


FIG. 3. (Color online) AFM analyses of the hydrogen-intercalated samples. AFM height (upper) and phase images (lower) of the samples after 20 min growth [(a) and (c)] and 33 min growth [(b) and (d)]. Panel (e) represents the cross-section at the line A-B indicated in (a) and (c). ~ 0.25 nm higher regions shaded in (e) show darker contrast in the phase image, i.e., graphene on the hydrogen-terminated SiC surface.

components related to the buffer layer. The qualitative peak ratio analysis of graphene to SiC component shows that estimated graphene coverage in the GNR sample (33 min growth) is 0.5 of the graphene monolayer, which is in good agreement with our GNR model (terrace area is 0.5 of total sample area).

The Raman spectra of the GNR sample (33 min growth) before and after hydrogen intercalation are shown in Fig. 4(a). The spectrum before hydrogen intercalation (as-grown) indicates typical features of the 6R3 structure, including lack of D and G' bands because of the absence of Dirac cones. In contrast, the GNR sample (after hydrogen intercalation) reveals evident D , D' , G , and G' bands. The D and G' bands show a fairly narrow peak fitted with a single Lorentzian function. The full widths at half maximum of the D and G' peaks are 31 and 49 cm^{-1} , respectively. The G -band peak at 1583 cm^{-1} is due to LO phonons at armchair edges, which are softened owing to the Kohn anomaly effect.²⁶ The deconvolution analysis of the G band indicates that there is a small additional peak at a wavelength 15 cm^{-1} higher, which is probably due to

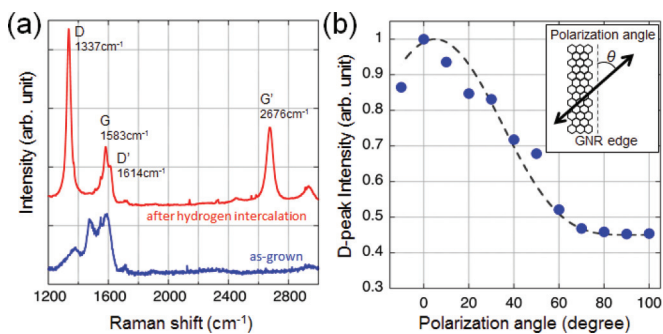


FIG. 4. (Color online) A Raman analyses of GNRs. (a) Raman spectra of the sample after 33 min growth before (blue) and after (red) hydrogen intercalation. (b) Polarization angle dependence of the D -band intensity. D -band intensity at each polarization angle θ , defined in the insert, is fitted as a function of $\cos^4\theta$ shown by a dotted curve.

either bulk LO phonons or TO phonons at armchair edges in the GNRs. The relatively strong D -band intensity due to the presence of high-density armchair edges and possible point defects should be noted. The edge character can be confirmed by the polarization angle dependence of the incident laser on the D -band intensity.²⁷ The integrated intensity of the D band at polarization angle θ is plotted in Fig. 4(b). These are fitted as a function of $\cos^4\theta$, suggesting predominantly armchair-type edges. This is also supported by the STM observation; line nodes²⁸ in the vicinity of the edges due to the interference effects of wave functions were recognized.

The band structures of the GNR samples were studied using ARPES. In Fig. 5(a), the ARPES result along the Γ - K - M lines of the graphene surface Brillouin zone is shown as a photoemission intensity image. The measured Γ - K - M direction in k space corresponds to $[1\bar{1}00]$ direction in real space, which is perpendicular to the step edges and thus not along the nanoribbon. A linear valence band dispersion verifies the presence of single-layer graphene. Note that a conduction band is invisible and the band is folded at the K -point area, as shown in Fig. 5(b). The red dots in Fig. 5(a) indicate the peaks of the energy distribution curves (EDC) shown in Fig. 5(b) and represent band dispersion around the K point. No states are detected between the Fermi energy (E_F) and valence band maximum, and a band-gap is opened at the K point. The minimum band gap of our GNRs can be 0.14 eV, assuming the conduction band minimum at E_F . However, based on the results obtained in the hydrogen intercalation of the 6R3 sheet, the band-gap value is expected to be more than twice 0.14 eV. Quasi-free-standing monolayer graphene shows a slight p -type doping nature;^{21,29} i.e., the Dirac point should be above E_F .

In summary, by templating vicinal SiC surfaces consisting of ordered pairs of (0001) terraces and nanofacets, GNRs are formed by MBE. The carbon atoms supplied to such a surface selectively organize a graphene network on (0001) terraces and form an epitaxial 6R3 layer (buffer layer). Hydrogen intercalation results in transformation to quasi-free-standing graphene and thus GNRs. The edge characteristic of the

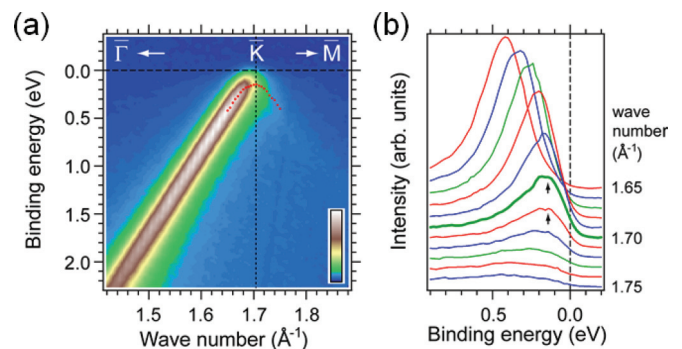


FIG. 5. (Color online) Intensity map and EDCs of the ARPES spectra around the K point. Intensity map (a) and EDCs (b) of the ARPES spectra around the K point of the samples after 33 min growth. The spectra are taken along the Γ - K - M line. The red dots in (a) indicate the positions of the EDC peaks. The intensity map clearly indicates a linear dispersion of the valence band of the single-layer graphene. The EDCs show the folding of the valence band at the K point below E_F .

GNRs grown under these experimental conditions indicates armchair-type edges. The massive arrays of GNRs, ~ 10 nm in width, indicate apparent band-gap openings of at least 0.14 eV at the K point observed by ARPES. A larger band-gap energy should be expected in GNRs ~ 5 nm in width. Such GNRs can be grown on vicinal 4H-SiC surfaces whose terrace width is ~ 5 nm.³⁰

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²²See Supplemental Material at <http://link.aps.org/supplemental/10.1103/PhysRevB.87.121407> for further details on our methods.

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