Scanning tunneling microscopy and spectroscopy of ion-bombarded Si(111) and Si(100) surfaces

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Surfaces of Si(111)-(7×7) and Si(100)-(2×1) were bombarded by 3-keV Ar⁺ ions at doses of $\leq 10^{12}$ ions cm⁻² to study the effect of individual ion impacts on the atomic structure of surfaces. Atom-resolved images show damaged regions of missing and displaced atoms. Current-imaging tunneling spectroscopy shows rest-atom states in the bombarded areas on Si(111), in agreement with line-scan measurements indicating a monatomic step-height difference between bombarded and unbombarded areas. Upon annealing to 750°C for 2 min, complete (7×7) was restored for the bombarded Si(111) surface whereas ordering of vacancies into line defects was observed for Si(100).

I. INTRODUCTION

The study of the removal of atoms, topographic development, and annealing behavior when energetic ions interact with a solid surface is of fundamental interest as well as practical importance. Recently, the microscopic understanding of the effects of ion bombardment of metal and semiconductor surfaces has been advanced by investigations using a variety of techniques. High-dose experiments to study the transition from the formation of vacancy islands to layer-by-layer sputtering have been conducted by workers using averaging techniques such as thermal energy atom scattering^{1,2} (TEAS) and reflection high-energy electron diffraction (RHEED).³ The development of the scanning tunneling microscope has brought about a unique opportunity to examine ion-bombardment-induced surface morphology on a nanometer scale. High-dose bombardment studies by scanning tunneling microscopy (STM) have been carried out on Au(111) (Ref. 4), Pt(111) (Ref. 5), and Si(100) (Ref. 6) surfaces, but the images are not atomically resolved. Low-dose experiments, i.e., $\leq 10^{12}$ ions cm⁻², have been conducted with the scanning tunneling microscope on surfaces of Si(100) (Ref. 7), PbS(100) (Ref. 8), and Au(111) (Ref. 9) to study the effects of single-collision cascades and the initial stages of sputtering. However, only one atomically resolved image, that of PbS(100),⁸ has been reported. Moreover, the bombardment of the PbS was carried out *ex situ*, and atomic imaging was possible only because of the resistance to oxidation of the PbS(100) surface. In this report, we present atomically resolved scanning tunneling microscopy (STM) and spectroscopy (STS) studies of the Si(111)-(7×7) and Si(100)-(2×1) surfaces bombarded *in situ* by energetic Ar⁺ ions at doses of $\leq 10^{12}$ ions cm⁻² (corresponding to ≤ 0.001 monolayer coverage).

II. EXPERIMENT

The base pressure in the ultrahigh-vacuum (UHV) chamber during the STM experiments was $\leq 1 \times 10^{-10}$

torr. Both Si(111) and Si(100) surfaces were cleaned by annealing to 1200°C for several seconds to eliminate the oxide layer. During annealing, the chamber pressure did not rise above 2×10^{-9} torr. After the annealing treatment, the sample was transferred to the scanning tunneling microscope where the usual 7×7 and 2×1 reconstructions were observed. The sample was then transferred to another part of the chamber where *in situ* bombardment took place. 3-keV Ar⁺ ions focused to a 5-mm-diam spot were produced by a differentially pumped ion gun, and they were incident normally onto the sample surface. The ion current was measured by a Faraday cup and it was adjusted to give various doses ranging from 7×10^{11} to 6×10^{12} ions cm⁻². Typical ion currents used were 5 and 10 nA. The base pressure in the differentially pumped beam line was 8×10^{-10} torr. During Ar⁺ ion bombardment, the pressure in the chamber rose to 3×10^{-8} torr. After bombardment, the sample was transferred back to the scanning tunneling microscope for imaging and spectroscopic studies of the ion impacts on the surface.

The Si(111)-(7×7) and Si(100)-(2×1) reconstructions are easily contaminated by oxygen and the contamination produces surface features which could be mistaken as ion-bombardment damage.¹⁰⁻¹⁴ To ensure that the STM images obtained were not the result of oxygen contamination, we performed two tests. The first test was to leak the 99.9999% pure argon gas via the ion gun into the UHV chamber to a pressure of 1×10^{-7} torr for several minutes, and then image the Si surface with the scanning tunneling microscopy to confirm that the reconstruction was unaffected. The second test was to place the sample in the proximity of, but not actually in, the path of the Ar⁺ ion beam for periods longer than the typical dose period. The unbombarded surface was then examined by the scanning tunneling microscope to verify that the reconstruction did not show signs of being contaminated. These tests were carried out separately for the 7×7 and the 2×1 surfaces. In particular, the second test was performed many times during the course of the experiments.

FIG. 1. STM image of the empty states of a clean Si(111)-(7 $\times$ 7) surface prior to ion bombardment. Scan area was 40 $\times$ 40 nm².

FIG. 2. (a) Empty-state image of a Si(111)-(7 $\times$ 7) surface bombarded by 3-keV Ar⁺ ions at a dose of 3 $\times$ 10¹² ions cm⁻². Sample voltage was +1.8 V; tunnel current was 0.5 nA; scan area was 20 $\times$ 20 nm². (b) Filled-state image of the same bombarded area (including some drift) as in (a). Sample voltage was -1.8 V; tunnel current was 0.5 nA; scan area was 20 $\times$ 20 nm².

III. RESULTS

A. Si(111)

The typical clean Si(111)-(7 $\times$ 7) surface prior to bombardment is shown in Fig. 1. It has a very low density of defects which consist mainly of single missing adatoms. Figures 2(a) and 2(b) show, respectively, the empty- and filled-state images of essentially the same 20 $\times$ 20 nm² area (the difference being due to drift) on the 7 $\times$ 7 surface after bombardment. The ion dose on this surface was 3 $\times$ 10¹² ions cm⁻², which translates into approximately 12 ion impacts on the imaged area. An enlarged image of an impacted region, with a scan area of 11 $\times$ 11 nm², is shown in Fig. 3. Figure 3(a) shows an empty-state image of some large "craters" taken at sample bias +1.8 V and tunnel current 0.5 nA. Figure 3(b) is a constant current topograph taken with the same bias and current as in 3(a)

FIG. 3. (a) Empty-state image of a Si(111)-(7 $\times$ 7) surface bombarded by 3-keV Ar⁺ ions at a dose of 3 $\times$ 10¹² ions cm⁻². Sample voltage was +1.8 V; tunnel current was 0.5 nA; scan area was 11 $\times$ 11 nm². The 7 $\times$ 7 unit cells are outlined to define the position of the craters. (b) Constant-current topograph of the same area as in (a) acquired during the CITS operation with the feedback loop stabilized at +1.8 V and 0.5 nA.

FIG. 4. Differential CITS images between biases (a) -1.5 and -1.0 V, (b) -1.0 and -0.5 V, (c) $+1.5$ and $+2.0$ V, and (d) $+2.0$ and $+2.5$ V. These images were taken simultaneously with that of Fig. 3(b), and the 7×7 unit cell outline from Fig. 3(b) is superimposed.

FIG. 5. (a) Average I - V curves taken over bombarded areas, i.e., craters, and unbombarded areas in the STM image shown in Fig. 3(b). (b) Tunneling spectra, $(dI/dV)/(I/V)$ as a function of energy, of bombarded and unbombarded areas in Fig. 3(b).

FIG. 6. (a) Filled-state image of a clean Si(100)-(2 $\times$ 1) surface prior to bombardment. Scan area was 24×24 nm². (b) The Si(100)-(2 $\times$ 1) surface after bombardment by 3-keV Ar⁺ ions at a dose of 1.5×10^{12} ions cm⁻². Scan area was 20×20 nm². (c) Bombarded surface as in (b) after annealing at 750°C for 2 min. Scan area was 20×20 nm². (d) Bombarded surface as in (c) after annealing at 850°C for 2 min. Sample voltage was -2.0 V and tunnel current was 0.5 nA for all four images. Images (c) and (d) were taken at room temperature.

on the same area a few minutes later. Figure 3(b) was acquired simultaneously as current-imaging tunneling spectroscopy¹⁵ (CITS) was carried out at sample voltages between -2.5 and $+2.5$ V, i.e., the feedback loop was stabilized at $+1.8$ V and 0.5 nA. The 7×7 unit cells are outlined by white lines in both 3(a) and 3(b).

Figure 4 shows differential CITS images taken at voltage intervals of (a) -1.5 to -1.0 V, (b) -1.0 to -0.5 V, (c) $+1.5$ to $+2.0$ V, and (d) $+2.0$ to $+2.5$ V, on the same area as in Fig. 3(b) with identical 7×7 unit cells outlined. Figure 5(a) shows the averaged I - V curves on the cratered and intact 7×7 regions in Fig. 3(b). The $(dI/dV)/(I/V)$ tunneling spectrum reflecting the surface density of states¹⁶ (DOS) is given in Fig. 5(b).

B. Si(100)

Figure 6 is a sequence of images showing (a) the clean (2 $\times$ 1) dimer rows with low defect density prior to ion bombardment; (b) the (2 $\times$ 1) surface after bombardment with 3-keV Ar⁺ ions at a dose of 1.5×10^{12} ions cm⁻²; (c) a different scanned area of the surface as in (b), after annealing to 750°C for 2 min; and (d) a different scanned area of the surface as in (c), after annealing to 850°C for 2 min.

IV. DISCUSSION

A. Si(111)

Figure 2(a) represents a typical empty-state image of the bombarded 7×7 surface. We have acquired such images on the bombarded surface at four different ion doses,

namely, 7.5×10^{11} , 1.5×10^{12} , 3.0×10^{12} , and 6.0×10^{12} ions cm^{-2} . The scan area varies from $10 \times 10 \text{ nm}^2$ to $40 \times 40 \text{ nm}^2$. The nature of the damage and its features appear unchanged from one ion dose to another although the density of defects changes. The 7×7 unit cell has 12 Si adatoms. We label each darkened adatom position as a "hole." A dark region of two or more "holes" constitutes a "crater." From the analysis of over 20 empty-state images of various scan sizes, there appears to be an approximate 1:1 correlation between the density of craters and the ion dose. Since the mean implanted depth of 3-keV Ar^+ ions in Si is 6 nm,¹⁷ it is unlikely that the craters are electronic in origin due to the presence of Ar on or near the surface. Diffusion of implanted Ar to the surface at room temperature is also highly unlikely since previous ellipsometry studies by Holtslag and Silfhout¹⁸ show that desorption of implanted Ar in Si(111) begins at an annealing temperature of about 400°C. We conclude that the observed craters are associated with ion impacts on the surface.

There are three categories of darkened Si adatoms, or "holes," in the craters in Fig. 2(a). The first category consists of dark holes, representing missing adatoms; we count 52 of these. The second category of holes are not totally dark, but grayish. A third category of holes have atom-size bright spots in the vicinity. We call the last two categories "displaced atoms," i.e., Si adatoms that are displaced but not completely removed; there are 25 of them in Fig. 2(a). The expected number of ion impacts on the $20 \times 20 \text{ nm}^2$ scanned area is 12, while we count 15 craters in the image. If we define sputtering yield as the ratio of missing adatoms (excluding displaced atoms) to craters, i.e., 52:15, representing the number of sputtered atoms per incident ion, then we arrive at a yield of 3.5 for this image.

The filled-state image shown in Fig. 2(b) of the same bombarded area has slightly inferior resolution, which we find to be the case in virtually all of our filled-state images. Some of the dark craters in the empty-state image [2(a)] now appear only grayish in the filled-state image [2(b)], and some of the craters in image 2(a) are even replaced by bright spots. Using the same definitions as before, we find that there are 18 missing atoms, 18 displaced atoms, and 12 craters in image 2(b). This gives a sputtering yield of 1.5.

Similar counting exercises performed on 12 additional images gave a sputtering yield of 3.0 ± 0.5 for the empty-state images and 1.8 ± 0.4 for the filled-state images. The sputtering yield of Si bombarded by 3-keV Ar^+ ions at normal incidence is ~ 1 determined by conventional methods,¹⁹ which seems to favor the yield derived from the filled-state images. The poor resolution in most of the filled-state images, however, introduces greater uncertainty in the counting.

In estimating the sputtering yield, we have assumed that each crater is caused by an ion impact. Such an assumption may, of course, not be valid. It is conceivable that an ion of the size of Ar^+ can penetrate an open structure such as the 7×7 surface without causing a crater. On the other hand, the propagation of collision cascades could give rise to craters (due to sputtering) on

parts of the surface other than the points of impact. If we use the formula given by Morehead and Crowder²⁰ to estimate the area of damage A caused by an energetic ion, then

$$A = (S_n/E_d)N_{\text{Si}},$$

where $S_n = (dE/dx)_n$ which is the nuclear stopping power for the 3-keV Ar^+ ion, E_d is the effective energy to displace a target atom and $E_d \approx 15 \text{ eV}$,²¹ and N_{Si} is the density of the Si target which equals $0.05 \text{ \AA}^{-3}$. For 3-keV Ar^+ ion bombardment of Si, $S_n \approx 40 \text{ eV/\AA}$,¹⁷ giving $A = 53 \text{ \AA}^2$. This produces an average diameter of $8 \text{ \AA}$ for a circular damaged region or "crater." The diameters of craters observed in the STM images of bombarded Si(111) surfaces are considerably larger than this value.

In an earlier report on ion-bombardment PbS(100) surfaces,⁸ we proposed that some of the missing or displaced atoms that give rise to the appearances of craters could be caused by surface relaxation to relieve the strain due to subsurface damage created by the incoming ion. If the same "sink hole" mechanism operates on the Si(111) surface, then the craters will appear larger than expected and the sputtering yield estimated by taking the ratio of "missing atoms" to "craters" in STM images will always be higher than the real value. However, it is also conceivable that the sputtering yields determined by conventional weight loss or eroded volume methods are steady-state values as a result of high-dose ion bombardment, whereas the higher value determined from the STM images represents the initial conditions before steady state or complete disordering of the surface is reached.

Figures 3(a) and 3(b) show closeup views of craters on the same bombarded surface where Fig. 2 was imaged. Both 3(a) and 3(b) are empty-state images, the only difference being that 3(b) was acquired simultaneously with CITS a few minutes after image 3(a) was taken. There are some striking differences between the two images. Most noticeable is that some of the 7×7 adatoms have moved, disappeared, or appeared, especially in the vicinity of the craters. This is a very interesting phenomenon, particularly in view of the fact that these are not foreign atoms, and they move at room temperature. This effect is similar to that reported recently by Lyo and Avouris,²² where clusters of atoms or individual atoms on the Si(111)-(7 $\times$ 7) surface were moved by applying voltage pulses at different tip-surface separations. In our case, the CITS operation involving rapid ramping of voltages from -2.5 to $+2.5 \text{ V}$ caused the atoms to move in an analogous manner. We noted that the movement of atoms only occurred during the CITS operation. Normal scanning of the surface at constant bias and constant current did not cause noticeable atom movements.

Figures 4(a)–4(d) are differential CITS images taken at fixed voltage intervals. The resolution is poor because CITS was performed only on 50×50 points compared with the usual 200×200 (or 400×400) pixels in a normal STM image. Each pixel on these images represents the difference in current at the two voltages chosen. So in image 4(c), for example, the difference between the current at $+1.5$ and $+2.0 \text{ V}$ would be almost constant

since the feedback loop was stabilized at +1.8 V and 0.5 nA. Indeed, in Fig. 4(c), the differential current image shows very little contrast. Since these images effectively represent the average dI/dV values over the prescribed voltage intervals at the surface, they represent, to some extent, the DOS features at the surface. Thus the large crater in Fig. 3(b) and its immediate surroundings to the left appear bright in images 4(a) and 4(b), indicating the existence of states below the Fermi level between -0.5 and -1.5 eV. In image 4(d), the existence of states above the Fermi level between 2.0 and 2.5 eV is also evident in and around the large crater. The tunneling spectra $(dI/dV)/(I/V)$ shown in Fig. 5(b), averaged over cratered and intact areas in Fig. 3(b), confirm the proposition derived from the CITS images. The bombarded curve shows a characteristic rest-atom band²³ between -0.5 and -1.5 eV below the Fermi level. The intensity of empty states above +1.5 eV also shows an increase for the bombarded curve compared to the unbombarded curve.

The appearance of the band due to rest-atom states in the cratered areas suggest the removal of the 7×7 adatoms, exposing the first layer of rest atoms underneath. So while the scanning tunneling microscope could not image any Si atoms below the adatom layer, the CITS images of Figs. 4(a) and 4(b) and the rest-atom band in the tunneling spectrum over the bombarded region in Fig. 5(b) provide the clue that only the Si adatoms are removed in the cratered areas. The average crater depth measured in the STM empty-state images is ~ 2.1 Å, which also suggests that only the adatoms are removed. This fits in well with the conventional models of sputtering where sputtering involves layer-by-layer removal.^{1-3,24} Note that, however, our STM observation does not preclude the possibility that sputtering can occur from the second layer as demonstrated recently by molecular-dynamics simulations²⁵ and experiment.²⁶ Because of the open structure of the Si(111)-(7×7) surface, ejection of second-layer atoms can conceivably take place leading to the formation of a depression, or crater, as part of the self-annealing process described earlier as the "sink hole" effect. STM images will not be able to "tell" the difference between this process and sputtering of atoms from the top layer. The ion-bombardment-induced defects observed in the STM images disappeared and the 7×7 reconstruction was completely restored after annealing the surface to 750°C for 2 min.

B. Si(100)

The most notable difference between ion-bombarded Si(111) and Si(100) surfaces is the absence of large craters on the (100) surface. The starting Si(100)-(2×1) surface is relatively defect-free as shown in Fig. 6(a). Most of the defects present appear to be type-C defects as defined by Hamers and Köhler²⁷ with some type-A single dimer vacancy defects. Very few type-B double dimer vacancy defects were found on this surface and there were no Ni-contaminated line defects. After bombardment by 3-keV Ar⁺ ions at a dose of 1.5×10^{12} ions cm⁻², which translates into six ions per 20×20 nm² imaged area, the density of defects increased dramatically. There are 21

defects in filled-state image 6(b) with more than one missing dimer, and out of these, eight of them have more than two missing dimers. A survey of the images of the bombarded surfaces shows that regardless of the ion dose used, there appears to be a greater number of defects than the expected ion impacts. This contrasts sharply with the approximate 1:1 craters to the ion-impact ratio found in ion-bombarded Si(111)-(7×7) surface. It is conceivable that some of the defects in image 6(b) are electronic rather than topographic in nature. Sampling the empty states and performing CITS on the bombarded surface would provide some insight. Unfortunately, we found it difficult to obtain good images of the empty states of the bombarded (2×1) surface and our CITS attempts on this surface also did not produce good resolution. This makes estimation of a sputtering yield for the (2×1) surface difficult and unreliable.

Annealing the bombarded surface as in image 6(b) at 750°C for 2 min produced line defects shown in Fig. 6(c). The width of the line defects appears to become broader after further annealing at 850°C for 2 min as shown in Fig. 6(d). After annealing at 950°C for 2 min, however, the line defects completely vanished (thus ruling out Ni contamination²⁸) and the (2×1) reconstruction was restored, although the surface was rough at places, perhaps indicating carbide segregation at the surface at 950°C. Annealing at 1200°C for a few seconds would restore large flat terraces of (2×1) as in image 6(a).

The formation of line defects in images 6(c) and 6(d) suggests an ordering of the random ion-bombardment-induced defects as shown in image 6(b). Depth measurements of the defects in image 6(b) by simple line scans yield an average depth of ~ 1.4 Å, which is close to the single-step height on the 2×1 surface. If indeed the ion-bombardment defects consist mainly of missing dimers, then we must examine the mechanism which gives rise to the ordering of random defects into line defects at elevated temperatures. This ordering effect is similar to the formation of hexagonal vacancy islands on Au(111) (Ref. 4) and Pt(111) (Ref. 5) surfaces subjected to high-dose ion bombardment at high temperatures.

Recently the step energy for the two types of monatomic steps on Si(100), S_A and S_B , were determined experimentally.^{29,30} Due to the symmetry of the diamond lattice, dimer rows are normal to each other on alternate monatomic steps. The step energy per unit length a ($a=3.84$ Å) for the step edge with the dimer rows of the upper terrace perpendicular to the step edge (S_B type) is 0.09 eV (Ref. 30) [0.09 eV (Ref. 29)]; while for the step edge with the dimer rows of the upper terrace parallel to the step edge (S_A type) a value of 0.06 eV (Ref. 30) [0.028 eV (Ref. 29)] is obtained.

If the step-edge energies play an important role, then ordering of missing dimers ought to be elongated along the direction of the dimer rows of the top layer. After annealing at 750°C, however, we find that the missing dimer defects order into elongated vacancy islands perpendicular to the dimer rows.

Martin *et al.*³¹ have conducted LEED studies on the Si(100) surface with excess missing dimer defects created by sputtering or oxidation. They found $2\times n$ structures,

where $6 < n < 10$, indicating ordering of the defects perpendicular to the dimer-row direction. However, they have no explanation for the ordering mechanism. Reflection electron microscopy (REM) studies by Kahata and Yagi³² showed a coalescence of vacancy defects, created by oxidation of the Si(100)-(2×1) surface at 700°C, into hollow lines perpendicular to the dimer rows. The proposed explanation was easy diffusion of the vacancies along the dimer rows and the attractive interaction of the vacancies led to the line defects perpendicular to the dimer-row direction.

Alerhand *et al.*³³ have calculated the anisotropic intrinsic stress tensor on the two degenerate phases, i.e., (2×1) and (1×2), of the Si(100) surface. Their calculation predict that the surface is under tensile stress in the direction perpendicular to the dimer rows and under compressive stress along the dimer rows. From the consideration of relieving the stress on the surface, one sees intuitively that the ordering of vacancies into lines perpendicular to the dimer rows has the effect of relieving the compressive stress. If the vacancies were to line up along the dimer rows, they would do nothing to relieve the surface stresses. From this argument, it appears that surface stresses play a more significant role than step energies in the ordering of vacancy defects at elevated temperatures.

The coarsening of the vacancy islands requires higher temperatures, e.g., the 850°C anneal in Fig. 6(d), since the activation energy for diffusion is involved in making the vacancies migrate along the dimer rows. In this respect, it is similar to silicon homoepitaxy,³⁴ except we now have the inverse phenomenon, i.e., vacancy islands. At 950°C, the line defects (or vacancy islands) disappeared; presumably sublimation began to take effect at this temperature.

V. CONCLUSIONS

3-keV Ar⁺ ion bombardment of Si(111)-(7×7) and Si(100)-(2×1) surfaces leads to the formation of craters. The sputtering yield estimated by taking the ratio of missing atoms to craters is consistently higher than the steady-state sputtering yield of 1 determined by conventional methods. The craters have an average depth equivalent to a monatomic step height. Tunneling spectra of the craters on the Si(111)-(7×7) surface show a characteristic rest-atom band, indicating the disappearance of only the top-layer atoms in the craters. Annealing of the bombarded surface at 750°C for 2 min results in the complete restoration of the (7×7) reconstruction on Si(111), but ordered vacancy line defects are formed perpendicular to the dimer rows on Si(100) for annealing at the same temperature. The mechanisms responsible for the ordering are thought to be preferential diffusion along the dimer rows, attractive interaction of the vacancies, and relief of the compressive stress in the direction of the dimer rows.

Note added. After the submission of this work, two papers by P. Bedrossian and T. Klitsner³⁵ on the sputtering of Si(111) and Si(100) surfaces by 225 Xe⁺ ions have come to our attention. However, their papers mainly concern the layer-by-layer removal and step retraction phenomena under high-dose bombardment at high temperatures, in contrast to our low-dose individual ion-impact approach.

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FIG. 1. STM image of the empty states of a clean Si(111)- (7×7) surface prior to ion bombardment. Scan area was 40×40 nm².

FIG. 2. (a) Empty-state image of a Si(111)-(7 $\times$ 7) surface bombarded by 3-keV Ar⁺ ions at a dose of 3×10^{12} ions cm⁻². Sample voltage was +1.8 V; tunnel current was 0.5 nA; scan area was 20 $\times$ 20 nm². (b) Filled-state image of the same bombarded area (including some drift) as in (a). Sample voltage was -1.8 V; tunnel current was 0.5 nA; scan area was 20 $\times$ 20 nm².

FIG. 3. (a) Empty-state image of a Si(111)-(7 $\times$ 7) surface bombarded by 3-keV Ar⁺ ions at a dose of 3×10^{12} ions cm⁻². Sample voltage was +1.8 V; tunnel current was 0.5 nA; scan area was 11 $\times$ 11 nm². The 7 $\times$ 7 unit cells are outlined to define the position of the craters. (b) Constant-current topograph of the same area as in (a) acquired during the CITS operation with the feedback loop stabilized at +1.8 V and 0.5 nA.

FIG. 4. Differential CITS images between biases (a) -1.5 and -1.0 V, (b) -1.0 and -0.5 V, (c) $+1.5$ and $+2.0$ V, and (d) $+2.0$ and 2.5 V. These images were taken simultaneously with that of Fig. 3(b), and the 7×7 unit cell outline from Fig. 3(b) is superimposed.

FIG. 6. (a) Filled-state image of a clean Si(100)-(2 $\times$ 1) surface prior to bombardment. Scan area was 24 $\times$ 24 nm². (b) The Si(100)-(2 $\times$ 1) surface after bombardment by 3-keV Ar⁺ ions at a dose of 1.5 $\times$ 10¹² ions cm⁻². Scan area was 20 $\times$ 20 nm². (c) Bombarded surface as in (b) after annealing at 750°C for 2 min. Scan area was 20 $\times$ 20 nm². (d) Bombarded surface as in (c) after annealing at 850°C for 2 min. Scan area was 20 $\times$ 20 nm². Sample voltage was -2.0 V and tunnel current was 0.5 nA for all four images. Images (c) and (d) were taken at room temperature.