The connection between observed, physical properties of adsorbed surfaces and their microscopic origins requires structural determination at the atomic level. Yet, investigation of such systems at that scale reveals local inhomogeneities in both reconstruction and individual atomic species at various stages of adsorbate incorporation. We have studied the submonolayer growth of Ga and B on Si(111) with tunneling microscopy and spectroscopy, and we show for each known phase assumed by these systems that images of mixed-phase surfaces provide structural information for new reconstructions which would not be accessible from images of a homogeneous, single-phase surface. Where mixing of adsorbate and substrate atoms occurs within a surface phase, individual atoms of two species may be identified. We demonstrate that controlled surface doping of Si(111) by boron induces identifiable changes in the local character of specific adatom sites leading to device characteristics associated with the presence of the tunneling tip over particular atoms.

1. Introduction

The sensitivity of the scanning tunneling microscope (STM) to surface topography and electronic character at the atomic scale [1] offers the potential for revealing the microscopic origins for macroscopic physical properties of adsorbed systems. Additionally, atomic-level examination of semiconductor surfaces uncovers inhomogeneities that may have no, or at best an ambiguous, macroscopic manifestation, but which might still be exploited practically. These capabilities of the STM, based on the technique's unique connection with evanescent waves and localized surface states, therefore average spatial information over macroscopic dimensions.

We demonstrate the use of the STM to these ends with the contrasting cases of gallium and boron adsorption on Si(111). Both Ga and B are trivalent elements occupying Group III of the periodic table, and both induce the $\sqrt{3} \times \sqrt{3}$ R30° periodicity at $\frac{1}{2}$ monolayer coverage (1 ML = $7.8 \times 10^{14}$ atoms/cm²) in low-energy electron diffraction (LEED) [2]. Nevertheless, there are a number of intriguing differences in the adsorption behaviors of these two elements. Above $\frac{1}{2}$ ML but below 1 ML, Ga induces an incommensurate, ordered overlayer on Si(111) [3], but boron does not. Above 1 ML, Ga does not wet a clean Si(111) surface at all [4]. Even within the $\sqrt{3} \times \sqrt{3}$ R30° structure, the B-terminated surface is stable to prolonged annealing above 1000°C, whereas Ga evaporates readily below 650°C. It has also been reported [5] that epitaxial silicon growth on Si(111)$\sqrt{3} \times \sqrt{3}$ R30°-B leaves the ordered boron layer undisturbed, but attempting similar Si growth on Si(111)$\sqrt{3} \times \sqrt{3}$ R30°-Ga immediately disrupts the Ga superstructure.

We shall show that atomic-scale characteristics of these systems, revealed by STM, account at least qualitatively for the microscopic origins of the fundamental physical properties above. Specially prepared, inhomogeneous surfaces are particularly useful in this regard; in addition, such surfaces display unique properties that might not be anticipated from macroscopic measurements. We first present STM data for each of the known phases of submonolayer Ga adsorption on Si(111) below, illustrating the methods with the relatively simple $\sqrt{3} \times \sqrt{3}$ R30° structure and then proceeding to the incommensurate phases. The contrasting case of boron is treated next. Finally, the implications of the topographic and spectroscopic data for both macroscopic materials considerations and atomic-scale device characteristics are discussed.

2. Gallium on Si(111)

Each of the experiments described in this work took place in a UHV chamber, of base pressure $10^{-10}$ Torr, that included facilities for cleaning and deposition. LEED optics, and an STM. Prior to Ga deposition, 0.04 Ω cm As-doped Si(111) crystals were sputter-cleaned in vacuum and annealed at 1000°C to exhibit the $7 \times 7$ reconstruction characteristic of the atomically clean surface [6]. Lateral distance measurements in the STM derive from the known periodicity of this reconstruction. Ga was then evaporated from an effusion cell at ~ 1 ML/min, while the sample was held at 550°C. Coverages were evaluated by in situ Auger spectroscopy or by Rutherford backscattering after the experiment.
If deposition was terminated when the LEED pattern transformed to $\sqrt{3} \times \sqrt{3} \text{R}30^\circ$, subsequent examination of the surface by STM resulted in images such as fig. 1, in which the measured adatom spacing of 6.7 Å corresponds to the periodicity of the diffraction spots. This and subsequent images were acquired by rastering the tip across the surface and recording the tip height, which is continually adjusted to maintain a constant tunneling current arriving on the sample. A greater tip displacement away from the surface is displayed by brighter shading, and the protrusions in the figure represent individual adatoms.

The image in fig. 1 might have been anticipated from the LEED periodicity and the trivalency of Ga, for a structure such as appears in fig. 1, if composed of Ga adatoms, would completely terminate the dangling bonds of an unreconstructed Si substrate. Nevertheless, the STM does not indicate the chemical nature of the observed adatoms, so the vertical position of the Ga must be determined independently. In this case, adatom occupation by the adsorbate was verified by X-ray standing-wave interferometry [7].

In order to determine the structure of the Si(111)$\sqrt{3} \times \sqrt{3} \text{R}30^\circ$–Ga surface completely, it remains to distinguish which of the two alternate three-fold binding sites on the substrate is occupied by the adatom. In the $T_4$ or "filled" position, the adatom would sit directly above a fivefold coordinated atom in the lower plane of the outer substrate double layer, while an adatom in the $H_3$ or "hollow" site would sit directly above no atom in the first substrate double layer. The resulting surface structures would be inequivalent, but both would give rise to the identical STM image, as in fig. 1. $T_4$ adatom occupation in this case was predicted from LDA calcu-
Fig. 3. Corrugation trace representing tip height vs lateral distance, along the line in fig. 2 spanning the $7 \times 7 - \sqrt{3} \times \sqrt{3} R30^\circ$ interface, demonstrating T$_4$ adatom occupation.

lations [7] and has been established for the case of Si adatoms in the $7 \times 7$ reconstruction. Experimental site determination for the adsorbed system here would therefore test both the predictive power of LDA structure calculations and the generality of the T$_4$ phenomenon.

Such a determination can be made with STM by preparing a surface that exhibits both $7 \times 7$ and $\sqrt{3} \times \sqrt{3} R30^\circ$ diffraction spots simultaneously. The binding site in the latter, undetermined structure, can then be found by examining interfaces with regions of the former, known structure. When such interfaces are located, as in fig. 2, the analysis consists of examining corrugation traces spanning the boundary. Thus, the cut shown in fig. 3, acquired along the line indicated in fig.

Fig. 4. 470×400 Å STM image of the Ga-induced overlayer characteristic of ≈1 ML Ga coverage. Tip bias = −3.3 V, 800 pA.

Fig. 5. 100×100 Å STM image showing the atomic-level structure of the supercells in a locally formed Ga overlayer, bordering reconstructed regions of lower local Ga concentration.
demonstrates that the Si(111)$\sqrt{3} \times \sqrt{3}$ R30° surface does assume the T4 configuration. The trace that would have resulted from H4 filling is indicated to illustrate that the STM resolves the distinction comfortably. It was therefore possible to employ an inhomogeneous surface to determine the lateral binding position of Ga in the new structure. We are unaware of ion scattering attempts to distinguish T4 and H4 sites on this surface, and we suggest that such measurements may be instructive.

Above 1/2 ML Ga coverage, the $\sqrt{3} \times \sqrt{3}$ R30° LEED pattern disappears and is replaced with hexagonal arrays of spots centered on the first-order diffraction spots of the substrate. We have seen that it was difficult to determine even the simple $\sqrt{3} \times \sqrt{3}$ R30° surface structure completely from diffraction data alone; for the more complicated incommensurate overlayer at higher coverage, there was no atomic-scale model prior to the STM study that accounted for any LEED spots at all. Aside from its practical implications, incommensuration is interesting in its own right and worthy of examination, and tunneling images would illustrate just how a surface accommodates competing adatom–adatom and adatom–substrate interactions.

An STM image of a sample with an average 1 ML Ga coverage is shown in fig. 4. The hexagonally packed protrusions in that image are not individual atoms, but rather supercells spaced 24 Å apart and forming a superlattice. Immediately, the origin of one of the sets of LEED spots, previously identified as 6.3 x 6.3 [3], is seen to be an incommensurate superlattice. Two terraces are also apparent in the image, and in fact the surface is decorated with remarkably straight steps that align themselves with the three symmetry directions of the surface. The step height of 3.2 ± 0.2 Å corresponds to the spacing of bulk Si planes in the (111) direction, suggesting that single-atomic silicon steps lie under the observed material.

Once again, an inhomogeneous surface can illuminate the initial growth of the overlayer and its binding to the substrate. For example, the image in fig. 5 was obtained from a surface that gave rise to 7 x 7, $\sqrt{3} \times \sqrt{3}$ R30°, and overlayer LEED patterns simultaneously. The presence of the two former structures allows precise calibration of the lateral distances for a particular image. Thus, the 24 Å observed spacing of the supercells within a prominent island matches the periodicity in fig. 4 and confirms that the island, which appears to nucleate along the surface’s symmetry directions, actually represents initial growth of the superlattice material. Also visible now are individual atoms within the supercells, in turn packed hexagonally with a 4.1 Å lattice constant.

The overlayer material, then, consists of an incommensurate superlattice, whose elements are in turn made up of an incommensurate, hexagonal lattice of atoms. We note that we have observed the previously described 11 x 11 and 6.3/3 x 6.3/3 LEED patterns for this system in addition to the 6.3 x 6.3 spots. The first might arise from the internal 4.1 Å spacing of the supercells, taking multiple scattering into account. While we do not detect obvious manifestations of the second pattern in the STM images, we emphasize that new periodicities might arise from subsurface rearrangements that could not be observed by STM [8].

Previous X-ray standing-wave measurements on the Ga overlayer placed the absorbed atom only 0.25 Å above the extrapolated lower half of a double layer,
compared with the bulk planar spacing of 0.77 Å within a Si double layer [9]. Based on these data alone, a natural first attempt at a structural model might have placed Ga atoms substitutionally in the first layer of an unreconstructed surface; however, the lattice spacing would then be 3.8 Å rather than the larger 4.1 Å measured by STM.

Evidently, the Ga-induced compression of the outer double layer, measured by the interferometric technique, leads to a lateral, outward relaxation in that layer to relieve the resulting bond angle stress, at least partially. The observed boundaries between the supercells then arise as a periodic array of misfit dislocations from the incommensurate relationship between substrate (3.8 Å) and surface (4.1 Å) atomic lattice constants. We note that a compressed double layer, as described here, resembles a graphite plane; in this case, the plane would be composed of interpenetrating, hexagonal Si and Ga sublattices. The inertness of such a planar structure might account for the inability of Ga to wet the Si(111) surface beyond 1 ML, to which we alluded above.

The formation of a planar outer layer follows a trend toward increasing, electronically driven compression in 1 × 1 substitutional geometries from right to left along a row of the periodic table, predicted theoretically and supported by X-ray standing-wave measurements [10,11]. Gallium, being trivalent, is expected to favor rehybridization in an sp² configuration, which would lead to a flattening of its nearest-neighbor bonds. At the other extreme, arsenic, a pentavalent element, would favor hybridization with smaller bond angles, inducing a vertical expansion of the outer layer.

Having determined the surface structures induced by various stages of Ga adsorption on Si(111), we now illustrate the identification of atomic defects on an adsorbed surface. It should be emphasized that the STM's sensitivity to the mid-gap surface bands involved in the tunneling process does not offer the chemical specificity offered by core-level probes. Yet if the impurities on the surface are known and controlled, they might be distinguished under STM by their effects on localized states.

Fig. 6 presents two tunneling images, acquired at different tip bias voltages, of one area of √3 × √3 R30° periodicity near a 7 × 7 domain, as in fig. 2, and which might therefore be expected to incorporate silicon.
adatoms as defects. Two distinct types of atoms in the figure are distinguished by their relative brightness, which reverses between the images. The reversal in shading with bias suggests that differing electronic character associated with the two types of sites accounts for the appearance of the images.

Inverse photoemission measurements on Si(111)√3 × √3 R30°-Ga have previously indicated an unoccupied, dangling-bond orbital atop Ga adatoms, ≈ 1.1 eV above the surface fermi level [12]. The corresponding dangling orbital anticipated for T4 Si atoms in the √3 × √3 R30° structure is expected to lie near the Fermi level because of its partial occupation by the silicon's additional electron [13]. On a mixed surface incorporating Ga and Si adatoms, we might therefore expect the species dependence of the dangling orbital's energy to be reflected in the respective local contributions to the tunneling current.

At negative tip bias, the tunneling current includes contributions from surface states lying between the Fermi level and the bias energy, so a greater tip retraction in a constant-current topographic image is induced over sites with a greater integrated, local density of states (DOS) up to the tip bias. Thus, the greater tip retraction over Si, the minority species, at low tip bias results in its brighter appearance. The Ga atoms dominate at the higher bias, where their unoccupied dangling orbitals contribute to the tunneling current. The identification of the randomly distributed defect sites is clearly a unique capability of STM.

3. Boron on Si(111)

Each of the Group-III elements through indium is known to induce a √3 × √3 R30° LEED pattern on Si(111) at low coverage. T4 adatom occupation has now been established for both Si and Ga atoms, as indicated above, and is strongly favored theoretically for the Al- and In-terminated √3 × √3 R30° surfaces [12]. In light of these results, an examination of the B-terminated surface might appear repetitious; nevertheless, the considerably smaller covalent radius of B (0.8 Å) compared with Ga (1.25) or Si (1.17) [14] calls attention to boron

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as a candidate to test the \( T_4 \) pattern. As an important dopant, boron's incorporation in a silicon surface merits examination.

We have employed three methods of preparing Si(111)\( \sqrt{3} \times \sqrt{3} \) R30° - B, each of which invariably results in an apparently random mixing of two types of adatom sites, again distinguished by relative brightness in STM images. Unlike the Ga case, however, the relative brightnesses of the two sites on the B-terminated surface persists regardless of the value or polarity of the tip bias used. Moreover, their relative proportion in STM images can be adjusted according to the parameters of preparation. By ion beam sputtering of boron on clean, As-doped Si(111) initially displaying the \( 7 \times 7 \) reconstruction, the concentration of “bright” adatoms could be lowered to 5%, as in fig. 7a. In the second preparation, Si was reevaporated onto a boron-rich surface, which was then annealed briefly at 800°C to recover the \( \sqrt{3} \times \sqrt{3} \) R30° periodicity in LEED and STM. The resulting concentration of “bright” adatoms could then be raised to 80%, as in fig. 7b. We note that similar results are obtained by annealing a heavily B-doped \( (10^{20} \text{ cm}^{-3}) \) Si(111) crystal at 1000°C [15], and the resulting STM images then show a concentration of 15–20% “bright” sites.

Longer boron deposition therefore raises the proportion of “dark” sites, while silicon deposition diminishes their presence. Apparently, the dark sites are connected with the presence of the dopant, and the bright sites with its absence. Once these sites are identified, it still remains to account for the appearance of the images, particularly the bias-independent persistence of the relative brightness of the two sites. The position of the boron itself must first be determined.

The lateral position of the \( \sqrt{3} \times \sqrt{3} \) R30° adatoms is found to be \( T_4 \), after analyzing interfaces between \( 7 \times 7 \) and \( \sqrt{3} \times \sqrt{3} \) R30° domains on an inhomogeneous surface, as demonstrated for the case of Ga above. The two alternative, vertical positions for the boron atom that are consistent with the full symmetry of both the LEED and STM images would place the adsorbate either (a) in the adatom position, following the pattern established for the Al-, Ga, and In-terminated structures, or (b) in the fivefold coordinate site directly beneath the adatom (\( B_5 \)).

The surfaces of the former case have been found theoretically to experience tensile stress, resulting from the unfavorable bond angles of a \( T_4 \) adatom in an unrelaxed structure [10]. Substantial atomic displacement, involving a vertical separation of adatom and \( B_5 \) atom, with a compression of the second double layer, is required to alleviate the stress partially. However, the smaller size of the boron atom would permit it to fit in the \( B_5 \) site (case b) maintaining favorable bond angles in the first double layer, and transferring minimal stress to the second. Ab initio total energy calculations comparing cases (a) and (b) for boron adsorption, which we have described elsewhere [16], have placed the latter structure, with boron underneath the adatom, lower in energy by 0.93 eV/adatom cell, or 0.31 eV/(1 \times 1) cell. To underscore the magnitude of this difference, we note that it exceeds the previously estimated energy difference between the \( 7 \times 7 \) reconstruction and an unreconstructed Si(111) surface [17]. According to this picture, all of the observed atoms in figs. 7a and b would be silicon; the “dark” sites would sit directly above a dopant atom, while the “bright” positions would represent silicon atoms sitting on silicon. We shall therefore refer to the former sites as “boron-occupied,” and the latter as “boron-free”.

Subsurface boron substitution accounts qualitatively for the contrasting physical properties of the B and Ga-adsorbed \( \sqrt{3} \times \sqrt{3} \) R30° surfaces mentioned in the introduction. The relative stability of the substitutional site itself explains the corresponding stability of the B-terminated surface to prolonged annealing, compared with the Ga-terminated surface that incorporates unfavourable tensile stress. Additional, ordered boron incorporation beyond \( \frac{1}{2} \) ML would disrupt the subsurface substitutional structure; consequently, no new structure comparable to the Ga-induced overlayer is observed with additional boron deposition. However, epitaxial Si growth, which disrupts the Ga \( \sqrt{3} \times \sqrt{3} \) R30° structure, can bury a B layer intact as arriving Si atoms bind to Si adatoms already at the surface.

Fig. 8. Typical tunneling spectra recorded over (a) bright, B-free, and (b) dark, B-occupied sites on Si(111)\( \sqrt{3} \times \sqrt{3} \) R30° - B/Si. The arrow indicates −1.4 V tip bias, where a peak in the LDOS appears on dark sites, and a minimum occurs on bright sites. The open-circle spectrum, which falls below zero at that bias, is characteristic of those boron-free sites that display negative differential conductivity.
The expected height difference between boron-occupied and boron-free sites accounts only partially for the observed tip height difference in topographic images, which can exceed 1 Å. We therefore consider the electronic contributions to tunneling that would affect the observed corrugation. Local tunneling spectra associated with each site can be acquired by freezing the lateral and vertical tip positions directly above an atom momentarily, sweeping the tip bias, and recording the tunnel current and differential conductivity. Fig. 8 presents representative spectra of the differential conductivity, which has been normalized to the total conductivity. The sample from which these spectra were taken was prepared by annealing degenerately B-doped Si(111), and negative tip bias represents tunneling into unoccupied surface states. Peaks in such spectra have previously been connected with the presence of surface states at the energy of the tip bias [18].

The occupied-state peak at +1.5 V tip bias in both spectra may be associated with the “back-bonding” state identified by photoemission [12,13]. On the boron-free sites, features in the local DOS appear in the vicinity of the Fermi level, and on either side. These might be associated with the predicted, partially occupied, dangling orbital of $\sqrt{3} \times \sqrt{3} \text{R} 30^\circ$ Si adatoms on a Si substrate [13]; their presence on the boron-free sites and absence on the boron-occupied sites undoubtedly contributes to the brighter appearance of the former positions at both polarities down to low tip biases. We note that this observation underscores the contrasting behavior of this surface to Si(111)$\sqrt{3} \times \sqrt{3} \text{R} 30^\circ$-Al, for which it was reported that no corresponding, low-lying unoccupied state peak was observed at Si defect sites [19]. Finally, the presence of an unoccupied-state peak 1.4 eV above the Fermi level on the boron-occupied sites is in qualitative agreement with a predicted upward movement of the dangling orbital on those sites, resulting from charge transfer from the Si adatom to the dopant underneath [16].

On most boron-free sites, the differential conductivity at −1.4 V tip bias is low, but non-negative. In fact, a continuous drop in the DOS for ≈ 0.8 eV above the 0.6 eV state has been reported for Si adatoms on clean Si(111) [20]. The unfilled curve in fig. 8a presents a spectrum, typical of B-free sites that are surrounded predominantly by B-occupied sites, which assumes

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Fig. 9. STM topographic image (a) and $dI/dV$ image (b) of one region, both acquired at −1.4 V tip bias. Regions of negative differential conductivity are shaded solid white in (b).

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negative values at $-1.4$ V tip bias. The surface B-doping evidently induces negative differential conductivity which is not present on the clean surface.

The spatial distribution of both the unoccupied state, associated with "dark" sites, and the centers of negative differential conductivity, associated with certain "bright" sites, may be determined by acquiring from the same region both a topographical image and an image of the differential conductivity at $-1.4$ V tip bias. The latter is obtained by placing a small (30 mV rms, 6.25 kHz) modulation on the tip bias and recording as a function of lateral position the output of a lock-in amplifier that senses the tunneling current. A pair of such images is displayed in fig. 9, from which it is apparent that spatial maxima in the differential conductivity at $-1.4$ V tip bias occur directly above boron-occupied sites, confirming the strong spatial localization of the state density on these sites. The centers of negative differential conductivity indicated in fig. 9b are also localized to atomic dimensions but occur directly on certain boron-free positions.

Having identified the local electronic character of each type of atomic site, we may associate specific $I-V$ characteristics of the tunnel junction according to the presence of the tip over specific sites, as in fig. 10. The curve shown for a boron-free site with the conductance anomaly at $-1.4$ V bias is reminiscent of the $I-V$ characteristic of a tunnel diode. In the present case, the lateral isolation of the centers of negative differential conductivity and their confinement to atomic dimensions enable those sites to act as independent, isolated, atomic-scale tunnel diodes in concert with the STM's tip.

The tunnel diode, originally described by Esaki in 1958 [21], exhibits the conductance anomaly when one electrode is biased in a range of energies where tunneling is suppressed, below which lies an energy where tunneling is favored. We propose that the same general mechanism is responsible for the conductance anomaly observed in the STM on Si(111)$(\sqrt{3} \times \sqrt{3})$R30°-B. At bright, boron-free sites, tunneling is strongly favored into the low-lying unoccupied dangling bond state, but is suppressed at slightly higher energies as the local surface DOS falls off. While $dI/dV$ is low, but non-negative, at $-1.4$ V tip bias even in the absence of boron, the presence of the dopant surrounding an isolated boron-free site would be expected to narrow the dispersion of the low-lying state on the latter site, further reducing the differential conductivity at $-1.4$ V bias. In this way, the dopant provides the small perturbation to drive the differential conductivity to a negative value at that bias.

The identification and spatial characterization of the surface states which contribute to the tunneling characteristics of each type of surface atomic site evidently constitute applications of STM that would fall outside the realm of traditional, averaging surface probes. It should be emphasized, however, that the very appearance of the conductance anomaly underscores the role of the entire tip–vacuum–sample system in STM data.

Practical exploitation of the conductance anomaly or any other local electronic feature would undoubtedly rely on an ability to position dopant atoms purposefully at the surface. While the average surface boron concentration can be adjusted in the preparation techniques we have employed, the local dopant distribution remains a product of random diffusion processes. Still, the determination of this distribution at the atomic scale is an important contribution of STM which would not be accessible by other means.

4. Conclusion

The study of adsorbed systems on semiconductor surfaces at the atomic scale is presented with spatial inhomogeneities in both surface lattice structure as well as the distribution of atomic species within a given structure, neither of which is amenable to analysis by many traditional techniques that must rely on long-range order. Mixing of phase domains is easily detected by STM, and we have illustrated the role that images of interfaces between phase domains have played in revealing the previously incompletely determined structures induced by submonolayer Ga and B adsorption on Si(111). Structural determination is the first step toward understanding physical or electronic properties of surfaces, and we have indicated how the observed, contrasting structures that these two adsorbates induce might influence the contrasting macroscopic properties assumed under Ga and B coverage.
Tunneling microscopy of Si(111)√3 × √3 R30° – B revealed an apparently random mixing of B and Si atoms on that surface in each of three methods of preparation employed. Under appropriate conditions, similar mixing could be induced on the Si(111)√3 × √3 R30° – Ga surface, and tunneling-spectroscopic investigations led to the assignment of particular local surface states to sites occupied by silicon or by adsorbate atoms. Such local spectroscopic information can complement the excellent k-space resolution of photoemission measurements. We have also indicated that the measured topographic corrugation and tunneling spectra of the inhomogeneous, B-doped √3 × √3 R30° surface diverged significantly from patterns established by other Group III adsorbates. Further examination by theoretical methods strongly indicated that boron atoms occupy a subsurface substitutional site, unlike other Group III atoms on Si(111). We would also like to emphasize that observations of inhomogeneous surfaces with STM methods can be quite helpful in assessing the validity of simple models in quantitative analysis of those surfaces by finite-size probes, for example ion beams, that may average over various kinds of surface structures.

Finally, it was shown that surface states, surface stress and atomic mixing could be brought to bear on device parameters on the atomic scale, using the STM. A small perturbation in the surface bands induced by surface boron doping led to the appearance of centers of negative differential conductivity, with atomic dimensions, at specific binding sites on the Si(111)√3 × √3 R30° – B surface. Identification of the surface states contributing to measured, local I–V characteristics led to an explanation for the appearance of the conductance anomaly according to the mechanism of a tunnel diode. A dopant was introduced, surface states were modified, and a new I–V curve could be engineered, this time with individual atoms.

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