Impact of repetitive and random surface morphologies on the ripple formation on ion bombarded SiGe-surfaces

Subhendu Sarkar1, Benny Van Daele and Wilfried Vandervorst
IMEC, Kapeldreef 75, B-3001 Leuven, Belgium
E-mail: subhendu.sarkar@gmail.com

New Journal of Physics 10 (2008) 083012 (16pp)
Received 27 March 2008
Published 7 August 2008
Online at http://www.njp.org/
doi:10.1088/1367-2630/10/8/083012

Abstract. Single (strained and relaxed) and polycrystalline Si1−xGe x samples have been bombarded using low-energy oxygen ions to study the impact of strain and relaxation-induced dislocations on their surface morphologies and subsequent formation of ripples. The as-grown relaxed sample surface exhibits cross-hatch patterns due to relaxation along its surface. Compared to Si, all the surfaces of these materials exhibit ripple formation at lower fluences. Moreover, the nature of the induced surface topography depends largely on the original surface morphology. Single crystalline layers show clear ripple formation whereas the polycrystalline surface leads to more isolated features. On a relaxed surface that exhibits the cross-hatch patterns due to relaxation along its surface, very long ripples tend to form along the dislocation ridges which eventually become shorter (and finally break up) with increasing distance from a particular ridge. A detailed temporal study has been done to understand the nature of ripple formation on the relaxed sample. A theoretical estimation of the crossover time indicates that the morphology of the relaxed sample is primarily governed by the nonlinear regime of ripple formation. It is found that the surface follows the scaling laws as dictated by the isotropic Kardar–Parisi–Zhang equation. The growth exponent (n) is found to be ~0.23, which is in agreement with the theoretical value. Ripple coarsening is observed which fits well with the recently predicted model indicating the contribution of nonlinear effects arising out of redeposition of the sputtered atoms and presence of mobile surface adatoms. The subsequent saturation of wavelength takes place as a result of geometrical shadowing.

1 Author to whom any correspondence should be addressed.
1. Introduction

The future of nanotechnology demands controlled and cost-effective fabrication, integration and mass production of nanostructures. The various available methods like photolithography, single-particle manipulation, self-organization, etc generally suffer from one or more shortcomings like limitations to further size reduction, throughput considerations, limitations in control, etc. On the other hand, ion beam sputtering (IBS) offers a controlled and cost-effective method for the production of nanostructures. Principally, this is a self-organization mechanism where the growth kinetics is determined by etching instead of growth. Ever since their first observation [1], the production of nanostructures in the form of ripples, dots, cones, etc have been reported by innumerable researchers [2]–[7] with well-defined vertical roughness, lateral periodicity, controlled step size and orientation. The most common form of nanostructures observed are in the form of ripples exhibited due to IBS in a wide class of substrates from amorphous or amorphizable [2], [8]–[10] to metallic targets [11, 12]. The production of semiconductor nanostructures has attracted the interest of many research groups because of the important applications in optoelectronic and quantum devices. While most of the work on semiconductors has however been done on elemental substrates there are only a few studies on compound semiconductors [10], [13]–[16]. In view of the increasing importance of compound semiconductors and in particular of SiGe in the omnipresent Si-technology, a more detailed study of surface topography development under IBS is of importance.

The first successful theory of ripple formation was developed by Bradley and Harper (BH) [17] based on Sigmund’s linear cascade approximation of sputtering processes [18, 19]. Their calculations showed that etching of a surface by ion sputtering is fundamentally unstable because the erosion rate depends on the surface curvature implying a protrusion eroding more slowly than a depression. The linear equation derived by BH describes satisfactorily some properties of IBS ripples, such as their alignment and wavelength. Other features, such as ripple stabilization, surface roughening, etc which could not be explained by the BH equation needed further nonlinear extensions leading to an anisotropic generalization of the well-known Kuramoto–Shivashinsky (KS) equation [20, 21]. This predicts the formation of a ripple pattern as a consequence of a competition between a roughening process caused by surface sputtering and a smoothening process caused by surface diffusion of thermal and/or ion beam origin. While all predictions of the linear theory can be calculated analytically, the discussion of the nonlinear
effects requires a combination of analytical and numerical tools [21], even with which the understanding of nonlinear effects is far less complete than that of the linear theory. Moreover, many experimental findings remain unexplained at low temperatures both in the linear and nonlinear regimes, as well as regarding the nonlinear regime at high temperatures.

Although effects like ion-induced effective surface diffusion (ESD) [22], ballistic diffusion [23, 24], viscous relaxation ([8, 25] and references therein) and Ehrlich–Schwoebel (ES) barrier diffusion [11] have already been incorporated in the growth equation, the theory essentially does not take into account any contributions from the underlying substrate or the morphology of the substrate surface (having a large surface curvature), both of which decide the fate of the ripples to a large extent. A recent work on allotropes of carbon surfaces [26] indicates that rippling depends on the nature of the underlying carbon materials. Moreover, the work of Aziz and co-workers [27] has indicated that the presence of a sharp boundary affects the rippling phenomenon drastically.

In the present study, we have investigated ripple formation due to low-energy bombardment at oblique incidence on various Si$_{1-x}$Ge$_x$ surfaces which differ by crystallinity and also the topography of the substrate surface. Comparisons with these samples were done with Si (001).

2. Experiment

For our experiments, Si$_{80}$Ge$_{20}$ samples were epitaxially grown using an ASM Epsilon 2000 CVD reactor. Silane and germane were the precursors and hydrogen was used as a carrier gas. The growth was done at 600 °C under 40 Torr. The nominal thickness of the strained wafer was 150 nm. The relaxed wafer was annealed at 900 °C for 5 min following the deposition. The nominal thickness for this layer was 250 nm. Polycrystalline (poly) Si$_{70}$Ge$_{30}$ samples were grown at 610 °C and 275 Torr pressure using pure SiH$_4$ and 1% GeH$_4$ in H$_2$ as reactants. We compared our results for the Si$_{1-x}$Ge$_x$ samples using a standard Si (001) wafer.

Secondary ion mass spectrometry (SIMS) experiments were carried out on the samples using an Atomika 4500 instrument with an oxygen beam of 1 keV O$_2^-$ and 18 nA current. The ion beam was aligned at 48° to the surface normal in all cases. Secondary positive ions were detected for all the samples during the SIMS analyses. The raster size was 250 × 250 µm. The choice of an oxygen beam (rather than the classical inert gas approach) is dictated by the interest in mimicking the normal SIMS conditions where oxygen beams are extensively used in view of their strong impact on the ionization probability [28]. Our experiments were divided into two parts. In the first part, we tried to focus on the difference in ripple characteristics for different Si$_{1-x}$Ge$_x$ samples linking the observed evolution with the surface morphologies of the different samples. For this set, bombardment was carried out for ion fluences of $4.3 \times 10^{17}$ and $1 \times 10^{18}$ ions cm$^{-2}$ (corresponding to 40 and 100 min, respectively). In the second set of experiments, we tried to understand the ripple evolution on a relaxed SiGe wafer. For these experiments, the ion fluence varied in the range $1 \times 10^{17} - 1 \times 10^{18}$ ions cm$^{-2}$ (from 10 to 100 min).

In order to study the resulting surface morphology, tapping mode atomic force microscopy (AFM) measurements were done $ex~situ$ inside the SIMS craters using a Dimension 3000 Nanoscope instrument. The root mean-square (rms) roughness of the surface is obtained from the AFM data using the relation $w(r) = \sqrt{\langle (h_i - h_j)^2 \rangle}$, where $h_i$ and $h_j$ are the heights of the surface at the positions $i$ and $j$ which are separated from each other by a distance $r$. The ripple
wavelength for a particular image is obtained from the distance between the central maxima and first-order peak of the auto-correlation function of the same image. All the AFM analysis has been done using WSxM software [29].

3. Results and discussion

3.1. Initial sample surface and shadowing

Figure 1 shows AFM images of the unbombarded surfaces of strained, relaxed and poly samples. The rms roughnesses of these surfaces have been found to be 0.22, 9.32 and 9.42 nm, respectively. It is clear from the image that the relaxed wafer exhibits cross-hatch patterns on its surface. These are the dislocation lines (DLs) that arise due to relaxation of the overlayer on top of the Si substrate [30]. The average periodicity of the DLs in the two perpendicular directions are 792 and 743 nm. The height is around 30 nm. Line profiles drawn on AFM images of the relaxed substrates reveal that the periodicity is almost sinusoidal. Comparing the strained and relaxed sample roughnesses over shorter length scales, it is seen that the roughness of the strained sample remains almost the same whereas that of the relaxed sample decreases and almost reaches the roughness of the strained case under very short length scales. This is quite evident since if we ignore the DLs, the surface should be otherwise similar to the strained case in terms of roughness. It is to be noted that the surface of the strained wafer is almost as smooth as that of pure Si (having an rms value of ∼0.12 nm). On the other hand, the case is just the opposite for the relaxed and the poly samples. Although they have a similar overall roughness, under shorter length scales the relaxed sample has a much smaller roughness but the poly sample continues to have large roughness values due to its granular nature. It is to be noted here that comparable overall roughness values for these samples arise from two different aspects—random morphology for the poly sample and patterned one for the relaxed sample.

It is important to investigate the effect of shadowing due to the dislocation ridges for the relaxed sample. For a sinusoidal morphology, the limiting condition for shadowing not to occur is given by [31] \( \tan\left(\frac{\pi}{2} - \theta\right) \geq \frac{2n h_0}{l} \) which places an upper limit on \( h_0/l \) for any ion incidence angle \( \theta \). Here \( h_0 \) is the amplitude of the sinusoid and \( l \) is the periodicity. If this ratio tends to be exceeded, the valleys of the sinusoid will not be eroded but the peaks will be planarized thus changing the sinusoid into a sawtooth-like wave form. The periodicity (for DLs aligned perpendicular to ion beam) and the height of the DLs for the relaxed SiGe sample are 792...
and 30 nm, respectively at the beginning. Therefore, for our case \( \tan(\frac{\pi}{2} - \theta) = 0.9004 \) \( \geq \frac{2h_0}{l} \) (= 0.237). Hence, we do not have any shadowing effect due to the DLs for the relaxed case. This condition remains valid even up to the maximum sputter time considered.

### 3.2. Ripple behaviour on different Si\(_{1-x}\)Ge\(_x\) surfaces

Figure 2 shows AFM images of the sputtered craters of all the samples including Si after a bombardment time of 40 and 100 min shown in the left and right columns, respectively. The strained and poly SiGe layers are eroded after a bombardment time of about 50 and 80 min, respectively. For all the samples the ripples tend to align themselves perpendicularly to the ion beam direction, i.e. the ripple wave vector is parallel to the ion beam direction. It is evident from the figures that ripples evolve much earlier for SiGe samples than that for pure Si even when accounting for the slightly higher sputter yield of SiGe compared to Si [32]. This might be an outcome of the influence of the difference in sample compositions. Comparisons between the SiGe samples also reveal certain important features. In particular, for the relaxed case the ripples seem to be aligned with the direction of the DLs. This is however governed by the direction of the ion beam and has been discussed in detail in the next section. It can be seen that the DLs act as a guiding boundary for the ripples leading to unbroken ripples, which are much longer when compared to those for the strained case. This is exemplified in figure 3 which shows a histogram of the lengths of the ripples formed for the different films. It is clear that in the relaxed case extremely long ripples (up to 8 \( \mu \text{m} \)) can be found whereas in the strained case they hardly exceed 2 \( \mu \text{m} \). Moreover, in the strained case the length distribution is fairly uniform across the crater surface whereas in the relaxed case a shortening of the (long) ripples occurs when moving away from the DL towards the valley in between two DLs. Even further away from the edge the well-formed ripples eventually break up into small islands. The scenario is altogether very different for the poly sample. Here the ripples are not as well formed as those for the single crystalline samples. We have tried to quantify the important morphological features in table 1. It is evident that the relaxed sample finally ends up with the surface having the maximum roughness, despite the fact that the relaxed and the poly samples start from the same initial roughness. Crystalline Si exhibits a rapid change in surface roughness from 40 to 100 min whereas the strained one shows a more gradual increase although both of them start with almost the same roughness values. So far as the ripple wavelengths are concerned, they are greater for the strained sample than those for pure Si and the relaxed sample. However, the ripple wavelength for the poly sample is by far the largest compared to all the other cases. These differences suggest more than one factor which could be responsible for the ongoing processes. The surface curvature (due to the DLs) surely plays a major role in deciding the graded ripple pattern for the relaxed sample. Otherwise, the ripple pattern should have looked similar to the strained case. For the fate of the poly sample, the overall granular nature plays a major role. It is probably for this that we get very different values for the poly sample when compared to the other two. The other factor existing at such low energies could be the crystallinity. Earlier studies on different allotropes of carbon have already hinted at such a possibility [26]. For semiconductors bombarded at high energies (which is not our case), this effect might be very low since the ion beam completely amorphizes the topmost layer of the sample [33]. Surface patterning determined by effects of surface crystallinity is of course quite common for metallic surfaces and has been investigated quite extensively [11]. Differences also exist in the ripple heights of the samples. For the strained and relaxed samples they are almost the same indicating
Figure 2. AFM images of ion-bombarded surfaces of Si (a) 40 min, (b) 100 min; strained SiGe, (c) 40 min, (d) 100 min; relaxed SiGe (e) 40 min, (f) 100 min and poly SiGe, (g) 40 min, (h) 100 min. Arrows in each of the images indicate the ion beam direction. Insets of (e) and (f) show magnified views of the ripples more prominently.
that small length scale morphologies could be the deciding factor. However, the ripple heights for the poly sample are much greater when compared to any of the other cases. These differences again seem to be a result of the surface curvature and overall morphology of the pristine samples. The ripple characteristics evaluated from the AFM images (table 1) for the 100 min case of strained and poly samples indicate that the underneath Si surface of these samples tends to adhere to the values obtained with the corresponding layers. Therefore, it is clear that the history of the sample surface is remembered for quite some time even after the original layer is eroded (e.g. for almost double the time for the strained case). This suggests a possible route to get ordered nanostructures of a certain material by first creating them on an overlayer for which we know the exact conditions.

3.3. Estimations from theoretical calculations

The generalized sputter-induced surface evolution equation for a semiconductor surface is given by [21]

\[ \partial_t h = v \nabla^2 h - D \nabla^4 h - K \nabla^4 h + \frac{\lambda}{2} (\nabla h)^2 + \eta(x, y, t). \] (1)

In the above equation, \( v \) is the roughening prefactor dependent on the angle of incidence of the ions. They are also known as the effective surface tensions generated by the erosion process and usually have a negative value leading to a primary (linear) surface instability. \( D \) is the relaxation constant for ion-induced ESD. \( K \) is the relaxation rate due to thermal...
Table 1. Roughness ($w$), ripple wavelength ($l$) and height ($h$) of all samples for 40 and 100 min. (All values are in nm.)

| Parameter | Time | Strained Si | SiGe | Relaxed SiGe | Poly SiGe |
|-----------|------|-------------|------|--------------|----------|
| $w$       | 0    | 0.12        | 0.22 | 9.42         | 9.32     |
|           | 40   | 0.18        | 2.83 | 16.18        | 9.78     |
|           | 100  | 4.38        | 3.99$^a$ | 21.88        | 13.5$^a$ |
| $l$       |      | 40          | 42   | 71           | 53       |
|           | 100  | 70          | 88$^a$ | 60           | 250$^a$ |
| $h$       | 40   | –           | 6    | 6            | 36       |
|           | 100  | 11          | 8$^a$ | 9            | 52$^a$  |

$^a$ Layer eroded.

surface diffusion. The nonlinear term $\lambda$ gives the slope-dependent erosion rate, $\nabla h$ denoting the actual slope. This term incorporates the lateral growth or erosion and is proportional to the ratio between flux and the penetration depth. It brings forth the saturation of the surface roughness in time. $\eta$ is a white noise term reflecting the random fluctuations in the growth process and is a set of uncorrelated random numbers with zero configurational average. It is to be noted that all the above terms have $x$ and $y$ components which can be evaluated independently [21]. Calculations made from equation (1) clearly separates two regimes in the sputter erosion process. For short erosion times (before a characteristic time $t < t_c$), only the linear terms determine the outcome of the growth equation. This is commonly known as the linear regime where a competition between the unstable negative sputter erosion term $(-|v|\partial^2 h)$ trying to roughen the surface and the positive surface diffusion terms ($D$ and $K$) trying to smoothen it eventually gives rise to ripples with wavelength $l_i = \frac{2\pi \sqrt{D_{tot}/|v_i|}}{\lambda}$, where $i$ indicates the direction ($x$ or $y$) along which the corresponding $|v_i|$ is the largest. $D_{tot}$ is the sum of the surface diffusion terms $K$ and $D$ and its value depends on the independent strength of these two terms. While successful in predicting the ripple wavelength and orientation [34], this linear theory cannot explain a number of experimental features, such as the saturation of the ripple amplitude [35]–[37], the observation of rotated ripples [38], and the appearance of kinetic roughening [39, 40]. Consequently, nonlinear terms had to be brought into consideration which come into effect after the characteristic time $t_c \sim (D_{tot}/v^2)ln(v/\lambda)$ [41] and the surface morphology is determined by either destroying the ripples or generating a new rotated ripple structure.

The numerical values of $v_i$, $D_{ij}$ and $\lambda_i$ can be found out from the equations given by Makeev et al [21]. For our case, we took the incident energy as 0.5 keV since a 1 keV O$^+$ ion sputters the surface as two O atoms. In order to do the calculations one needs to have the values of the energy deposition depth $a$, longitudinal straggling $\alpha$ and lateral straggling $\beta$. For this purpose, we follow Bolse’s approach [42, 43] for normal ion incidence. The depth profile of the damage distribution function (target displacements and replacements) is obtained from SRIM code [44] with full cascade option and the resulting plot is fitted with the deposited energy.
density function \( F_D \) given by [18, 19]

\[
F_D = \frac{E}{(2\pi)^{3/2} \alpha \beta^2} \exp \left( -\frac{(z - h_0 - a)^2}{2\alpha^2} - \frac{x^2 + y^2}{2\beta^2} \right),
\]

Putting our experimental conditions into the equation, we get \( a = 1.52 \text{ nm} \) and \( \alpha = 1.39 \text{ nm} \) for SiGe and \( a = 1.88 \text{ nm} \) and \( \alpha = 1.35 \text{ nm} \) for Si. The \( \beta \) values were taken from the Winterbon tables [45] and are 0.86 and 0.68 nm for SiGe and Si, respectively. With these values, we see that for SiGe the ripples are theoretically predicted to be parallel to the ion beam, thus contradicting our experiments. The nearest value of \( \beta \) that is in agreement with our observations is 0.73 nm (\( \sim \alpha/2 \) for SiGe). This discrepancy could be due to the fact that the value obtained from [45] is for an elemental target but we are actually working with a compound one. For Si, we get the correct directionality of the ripples using the \( \beta \) value from the Winterbon tables. Results obtained following the above procedure are tabulated in table 2. It is worthwhile to mention here that the early onset of ripples that we observe for the SiGe samples compared to Si could be due to the corresponding material parameters that governs the competition between the roughening and relaxation terms and the initial surface roughness, which is nearly impossible to quantify. Moreover, we do not know at present anything about the transient evolution of the surface when an ion beam first hits it. Recent molecular dynamics (MD) simulations by Johnson and co-workers [46] have shown that surface corrugations can evolve out of a flat surface just after about 50 ion impacts. This is seen for both Si and SiGe \( |v_x| > |v_y| \). Thus the ripples will be aligned perpendicular to the ion beam direction which is what we actually observe from the AFM images. The characteristic or crossover time \( (t_c) \) obtained for SiGe is \( \sim 20 \text{ sec} \). It is therefore evident that the nonlinear terms will dominate the surface morphology just after a few moments of sputtering. The morphology will now depend on the relative signs of the nonlinear terms \( \lambda_x \) and \( \lambda_y \). For both SiGe and Si, \( \lambda_x \cdot \lambda_y > 0 \), thus signifying an eventual kinetic roughening of the surface. The scaling properties of the surface under such situations are then governed by the isotropic Kardar–Parisi–Zhang (KPZ) equation given by [47]

\[
\partial_t h = \nu \nabla^2 h + \frac{\lambda}{2} (\nabla h)^2 + \eta(x, y, t).
\]

According to the scaling theory [48] the roughness exponent \( (m) \) and the growth exponent \( (n) \) vary as \( w(l) \sim l^m \) and \( w(L, t) \sim t^n \), respectively, where \( w \) is the surface width. It is argued that with time, the surface develops coarser and coarser features characterized by the length \( l \) parallel to the substrate and \( l \sim t^{1/2} \), where \( z \) is defined as the dynamical exponent. For the present case, \( n \simeq 0.25 \) theoretically. The experimental value of \( n \) has been discussed later in the text. The thermal relaxation factor \( K \) is usually very low at room temperatures. Using the same input values as used by Erlebacher et al [49] this is estimated to be \( 3 \times 10^{-11} \text{ nm}^4 \text{ s}^{-1} \) which in turn gives an estimated ripple wavelength of about \( 4 \times 10^{-4} \text{ nm} \). It is evident that this value is nowhere close to the experimentally observed values. Therefore, it is argued that ion-induced ESD plays a key role in the ripple formation. The calculated value (table 2) is smaller by almost a factor of 7–10 than the experimental values. The maximum deviation is however found for the poly SiGe sample whose ripple wavelength differs by almost a factor of 30 from the theoretical value considering a bombardment time of 100 min. The deviations are principally caused by the choice of the bombardment parameters which in practice could be slightly different from the chosen theoretical values. Moreover, major differences in the starting surface morphology and inherent properties (e.g. crystallinity) of the samples can also account for such deviations and differences between the samples. It is to be borne in mind that such differences were not
Table 2. Theoretically estimated values of ripple characteristics for SiGe and Si for 1 keV O$_2^+$ at 48° incidence.

| Sample | $v_x$ (nm$^2$ s$^{-1}$) | $v_y$ (nm$^2$ s$^{-1}$) | $\lambda_x$ (nm$^2$ s$^{-1}$) | $\lambda_y$ (nm$^2$ s$^{-1}$) | $D_x$ (nm$^3$/s$^{-1}$) | $D_y$ (nm$^3$/s$^{-1}$) | $l_x$ (nm) | Growth rate (s$^{-1}$) |
|--------|----------------|----------------|----------------|----------------|----------------|----------------|----------|----------------|
| SiGe   | -1.34 ($\times 10^{-2}$) | -1.29 ($\times 10^{-2}$) | -0.98 ($\times 10^{-2}$) | -3.92 ($\times 10^{-2}$) | 1.23 ($\times 10^{-2}$) | 8.5 ($\times 10^{-2}$) | 3.67 |
| Si     | -1.52 ($\times 10^{-2}$) | -1.06 ($\times 10^{-2}$) | -2.25 ($\times 10^{-2}$) | -0.12 ($\times 10^{-2}$) | 1.16 ($\times 10^{-2}$) | 7.8 ($\times 10^{-2}$) | 4.99 |

taken care of in the theoretical estimation of the ripple parameters. All the theoretical models expand the morphology in powers of the surface slope, which is treated as a small parameter. Recent theoretical calculations [50] suggest that for a particular material the resulting surface morphology can vary depending on the surface roughness owing to the difference in sputter yields.

3.4. Time evolution study of ripple formation on cross-hatched SiGe surfaces

It has already been stated earlier that the ripples formed on relaxed SiGe samples exhibit certain characteristics that are absent in any of the other samples. To study the evolution of these ripples we bombarded the sample surfaces for different times and the resulting craters were analysed using AFM. Figure 4 shows the AFM images obtained. Since the sample surface already had DLs on it, it would be interesting to see the fate of these DLs with time. Figure 5 plots the height, periodicity and slopes of the DLs with the sputtering time. The heights of the DLs that are parallel to the ripples gradually increase with time and finally tend to saturate. This further proves that there is no shadowing of the incident beam due to the DLs. The average periodicity of the DLs lying perpendicular to the ripple direction gradually increases suggesting a gradual increase of the correlation length in this direction. In other words, the DLs lying perpendicular to the ripples gradually tend to die out with sputtering time. Another interesting feature of the DLs is that their slopes undergo a sharp change after a sputter time of about 40 min.

As discussed in an earlier section, the lengths of the individual ripples vary to a large extent in the direction of the ion beam for the relaxed SiGe sample. This is a bit similar to what has been observed previously [27] where a template with a sharp vertical boundary served as the driving force for the orientation of the ripples. In comparison to that study, we have in the present case a sinusoidal boundary. Moreover, in contrast to the earlier case, the ion beam direction in the present case is not aligned with the parallel DLs. It is important to note here that a defect free ripple can actually evolve even in the absence of a sharp boundary although the maximum range of the lateral templating effect might be lower. Currently, there is no theory that takes care of such a situation in ripple formation. In order to account for such an effect one needs to consider the periodic features of the starting surface (DLs for our case) in the growth equation and carry out calculations beyond the small slope approximation. The fact that shadowing by the DLs is not a cause for ripple shortening is strengthened by the fact that calculations done for the DLs at the end of sputtering (100 min) rule out any such possibility ($\frac{2\pi h_0}{l} = 0.721$).

The gradual formation and propagation of the ripples over the DLs are depicted in the line profile diagrams of figure 6. The large humps are the DLs that are parallel to the ripples. The small fluctuations on the top of the large humps denote the actual ripples. The ripples initially
Figure 4. AFM images of ion-bombarded surfaces of relaxed SiGe sputtered for different times. Arrows in each of the images indicate the ion beam direction.
The first ripple that forms at the start of the leading slope is always found to be the longest. Subsequent ripples follow the first one. The entire group of ripples for a particular dislocation ridge is then seen to migrate to the top of a dislocation ridge. Finally, the top of the ridge flattens giving way to the ripple pattern. Ripples are observed as early as 10 min but are quite distinct from about 22 min. Estimates from the AFM images reveal that defects in the ripples start appearing after about 2 \(\mu\)m distance from the edge. Eventually the ripples become shorter finally ending up into small islands or mounds over a distance of about 0.4 \(\mu\)m. The driving force behind the long ripple formation is so strong that the ripples are almost unperturbed by the DLs that run perpendicular to them. It is clear that the strength of this factor falls off with the distance from the edge of the dislocation ridge. Figure 7 shows the variation of the roughness, periodicity and height of the ripples formed on the relaxed SiGe surface due to sputtering. The growth rate obtained by an exponential fit to the roughness values gives 1.8 \(\times\) 10\(^{-4}\) s\(^{-1}\). The calculated value (table 2) is greater than this by an order of magnitude. The growth exponent \(n\) is found to be 0.23 which is quite close to what is predicted theoretically. But on the other hand, the dynamical exponent \(z = 3.03\). Although this value is almost double what has been predicted theoretically, other studies on semiconductor surfaces have also predicted values close to this [51]. It is therefore evident that we are in the nonlinear regime of ripple formation. The variation of the average height of the ripples also follows a similar nature as the above two. All these parameters tend to saturate after a bombarding time of around 60 min. This is also found to be true for the fate of the DLs in general.

To account for the ripple coarsening as observed in figure 7, we can consider a recent work by Cuerno and co-workers [52]. Their ‘hydrodynamic’ model is a suitable generalization of continuum descriptions of IBS ripples formed under oblique incidence ions. It is proposed that the near surface atoms receiving enough energy and momentum to break their bonds are
Figure 6. Time evolution of the ripple profiles with the sputtering time.

Figure 7. Variation of roughness (upper panel), periodicity (middle panel) and height (bottom panel) of the ripples formed with the sputtering time.
in principle sputtered away, although they may join the current of surface adatoms that are available to other relaxation mechanisms, such as surface diffusion, before incorporating back into the bulk solid. In addition to the time evolution of the surface height, this model also considers the time evolution of thickness of the mobile surface adatoms layer, \( R(\vec{r}, t) \). The latter is a function of the fraction \( (1 - \phi) \) of eroded material that redeposits onto the growing surface, \( \phi \) being the fraction of atoms that are sputtered away from the surface. The surface evolution is then described by the equation

\[
\frac{\partial}{\partial t} h = \sum_{i=x,y} \left[ -v_i \partial_i^2 h + \lambda_i^{(1)} (\partial_i h)^2 \right] + \sum_{i,j=x,y} \left[ -K_{ij} \partial_i^2 \partial_j^2 h + \lambda_{ij}^{(2)} \partial_i^2 (\partial_j h)^2 \right]. \tag{4}
\]

For simplicity, the zeroth- and first-order terms as well as the terms that are not responsible for pattern formation itself are omitted from the above equation. \( v_i \) and \( K_{ij} \) have the same meaning of sputter erosion and surface diffusion as in equation (3). However, the relaxation term \( K_{ij} \) and the nonlinear terms \( \lambda_i^{(1)} \) and \( \lambda_{ij}^{(2)} \) depend now on the redeposition parameter \( \phi \), angle of ion incidence \( \theta \), thermal diffusion and a nonzero average fraction of mobile atoms that also contribute to smoothing. Under the nonlinear regime, the ripple characteristics depend on the relative strength of the terms \( \lambda_i^{(1)} \) and \( \lambda_{ij}^{(2)} \). When the values of the coefficients \( \lambda_{ij}^{(2)} \) increase relative to \( \lambda_i^{(1)} \), coarsening stops later, and the amplitude and wavelength of the pattern also increase. The coarsening exponent \( 1/2 \) will take an effective value that will be larger the later coarsening stops and may depend on simulation parameter values. It is owing to these nonlinear \( \lambda_{ij}^{(2)} \) terms that we observe a coarsening in our case. The exact form is of course unknown at present due to lack of quantified information of the redeposited and mobile atoms. Subsequent to coarsening, a saturation is observed. This is actually due to geometrical shadowing estimated from the shadowing condition of [31] which sets in at around 60 min for our case. But this shadowing is due to the ripples formed on top of the DL ridges. The amplitude to height ratio of the ripples attains such a value that it blocks the incoming beam and hinders the further growth of amplitude resulting in the eventual saturation that we observe here. Probably this also explains the flattening of the dislocation ridges (figure 6) at later stages.

4. Conclusion

Our study deals with ripple formation in morphologically different Si\(_{1-x}\)Ge\(_x\) surfaces under 1 keV O\(_2^+\) ion bombardment at an ion incidence angle of 48°. Strained, relaxed and polycrystalline SiGe samples are chosen for the experiments. While the as-grown strained and polycrystalline samples are smooth and rough, respectively, the relaxed one exhibits the presence of DLs on its surface. AFM results indicate that the morphology of these three types of surfaces subsequent to bombardment is vastly different with respect to roughness, ripple wavelength and height. Attempts have been made to compare these results with the theoretical estimates. Differences primarily arise due to sample surface curvature and the vast morphological difference especially with respect to the polycrystalline sample. Comparison of the SIMS yields with the roughnesses suggest that small scale roughness primarily decides the yield enhancement observed for the samples. A time evolution study of the ripples has been done on the relaxed SiGe sample. Very long ripples are formed along the boundary of the dislocation ridges thereby suggesting a strong boundary effect (for the formation of the ripples) which is essentially unperturbed by the perpendicular DLs. These ripples shorten in length as one goes away from the boundary. The morphology is primarily determined by the nonlinear regime of
ripple formation. The surface features follow scaling laws that are governed by the isotropic KPZ equation. Ripple coarsening is also observed as predicted by the theoretical hydrodynamic model. Finally, the ripples exhibit a shadowing effect thereby explaining the observed saturation.

Acknowledgments

We thank Professor Jonah Erlebacher of the John Hopkins University, Maryland, USA and Professor T K Chini of the Saha Institute of Nuclear Physics, Kolkata, India for helpful discussions.

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New Journal of Physics 10 (2008) 083012 (http://www.njp.org/)