High-precision deformation mapping in finFET transistors with two nanometre spatial resolution by precession electron diffraction

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Precession electron diffraction has been used to systematically measure the deformation in Si/SiGe blanket films and patterned finFET test structures grown on silicon-on-insulator type wafers. Deformation maps have been obtained with a spatial resolution of 2.0 nm and a precision of $\pm 0.025\%$. The measured deformation by precession diffraction for the blanket films has been validated by comparison to energy dispersive x-ray spectrometry, X-Ray diffraction, and finite element simulations. We show that although the blanket films remain biaxially strained, the patterned fin structures are fully relaxed in the crystallographic planes that have been investigated. We demonstrate that precession diffraction is a viable deformation mapping technique that can be used to provide useful studies of state-of-the-art electronic devices. *Published by AIP Publishing.*
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Strain is routinely used to improve the performance of transistor devices.¹ For traditional planar devices, the application of strain is straightforward. Recessed SiGe sources and drains are used to compress the channel for p-type metal oxide semiconductors (pMOS) and SiC, or nitride films are used to provide tensile strain for n-type metal oxide semiconductors (nMOS). However, the miniaturisation of semiconductor devices has led to new problems such as leakage current between the source and drain when the device is nominally switched off. Solutions that have been proposed include: (1) the finFET architecture where the gate is wrapped across three sides of the channel to reduce the leakage current or (2) silicon-on-insulator (SOI) technology where the devices are grown onto a buried oxide film to protect them from leakage through the substrate. Of course, if fin structures can be grown on SOI wafers, then the leakage current is controlled around the whole channel. For both the finFET and SOI type devices, new methods of applying strain need to be developed. As the fabrication of modern semiconductor devices can involve hundreds of processing steps, it is difficult to assess from only electrical measurements whether the improvements and differences between the differently processed devices are only due to the applied strain. In addition, the manufacture of fully processed wafers for testing is both time consuming and expensive. Thus, accurate and precise metrology with an appropriate spatial resolution is required to assess the effectiveness of the different strategies that are used to strain the devices at an early stage of wafer processing. There are many different techniques that can be used to measure strain in these devices;

however, only transmission electron microscopy (TEM) based techniques have the nm-scale spatial resolution that is required in order to observe individual structures. Driven by the semiconductor industry, the last 10 years has seen a great deal of activity in the field of strain analysis using the TEM.^{2–6} Precession electron diffraction (N-PED) is one of the recently developed techniques, which can be used to provide deformation maps with an excellent spatial resolution (< 2 nm) and precision ($\pm 0.02\%$).^{7,8}

In this study, four different specimens prepared from two different wafers were examined. The first wafer is comprised of a 20-nm-thick Si_{0.60}Ge_{0.40} layer grown onto an 8-nm-thick silicon layer on SOI, now referred to as *as-deposited*. The second wafer, now referred to as *annealed*, was nominally identical to the first except that it had been annealed such that the Si and SiGe films would mix. From each wafer, one specimen was prepared from a region containing a blanket film and another from a region with patterned 12-nm-wide finFET structures each separated by 45 nm. As the regions of interest were on the surface of the wafer, it was important to be able to prepare the specimens without exposing them to the ion beam. Here, a marker pen with a wide nib was used to flow ink onto the regions of interest. The specimens were then prepared as usual by focused ion beam milling (FIB). A metal top layer was deposited using ion-beam assisted deposition to protect the region of interest from the tails of the beam. *In-situ* lift out was then performed, and the specimen was thinned for examination using an operating voltage of 30 kV for the rough milling and then 8 kV for the final steps until specimens with a crystalline thickness of 150-nm were obtained. Finally, the specimens were plasma cleaned in order to remove the ink such that it did not interfere with the measured strain.

The STEM energy dispersive x-ray spectrometry (EDX) experiments were performed using an FEI Osiris equipped

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FIG. 1. HAADF STEM images of (a) as-deposited blanket film and (b) patterned as-deposited film showing finFET structures, (c) EDX map of the as-deposited blanket film, and (d) composition profiles extracted across the indicated region. (e) Annealed film and (f) patterned annealed film showing finFET structures. (g) EDX map for the annealed film and composition profiles taken from the indicated region. To better observe the Ge distributions, the Si contribution has been removed from the EDX maps. Deformation maps have been acquired for the regions indicated by the black dotted rectangles.

with four windowless silicon drift detectors (SDDs) integrated into the pole piece, which allows X-ray detection over a 0.8 sr solid angle. The EDX maps were acquired with a probe current of 300 pA. The spatial resolution for the EDX analyses (measured from the width of intensity profiles at interfaces) is evaluated at 2 nm, which is very close to the full width at half maximum (FWHM) of the beam size used for strain mapping. The sample was slightly tilted off the zone axis condition to reduce effects of electron channeling. The quantification was performed using the intensity of Si K and Ge L peaks. As these two peaks are very close in energy (Ge L peak at 1.19 keV and Si K peak at 1.74 keV), they are thus subject to similar absorption effects. The quantification was performed using the Cliff Lorimer method. Figure 1 shows high-angle annular dark field (HAADF) STEM images and EDX maps acquired from the different specimens. Figures 1(a) and 1(b) show the presence of the bottom Si layer and the top SiGe layer for the *as-deposited* blanket film and patterned finFET structures, respectively. Figure 1(c) and (d) shows an EDX map of the blanket film and quantitative composition profiles, respectively, revealing that the Ge concentration increases from 32.5 to $40.5 \pm 2\%$ from bottom to top in the film. Figures 1(e) and 1(f) show that the anneal has led to complete mixing of the two layers for both the blanket film and the finFET structures. Figures 1(g) and (h) show an EDX map and composition profiles, respectively, revealing that the Ge content is uniformly $27 \pm 2\%$ in the film. The EDX results can be compared with XRD measurements on the blanket films, which provided values of 37% Ge content for the specimen before anneal and 27% Ge content after anneal. The reduction in the Ge content for the annealed specimen is consistent with the mixing of the Si and SiGe regions. For the etched finFET structures, the STEM results reveal a 2-nm-thick oxide layer present on the surfaces, which was confirmed using Energy-dispersive X-Ray spectroscopy (EDX) maps.

N-PED experiments were applied to the different devices using a double aberration corrected FEI Titan Ultimate TEM equipped with a high-brightness X-FEG electron gun and a monochromator. The probe corrector was used to provide an electron probe with a full width at half maximum (FWHM) of 0.8 nm at a convergence angle of 2.4 mrad

using a $10 \mu\text{m}$ C2 aperture. The electron beam is preprocessed using the STEM scan coils, and software developed at CEA is then used to the scan beams across the specimens while acquiring and processing the diffraction patterns. The key is to acquire the data while limiting specimen drifts, which is not necessarily linear or always in the same direction; for this reason, 512 pixel² images were recorded using a 2048 pixel camera as it is quicker to acquire and save the data. A strain precision of $\pm 0.025\%$ was determined by acquiring a map of the unstrained Si substrate and calculating the standard deviation of the ε_x map. The precision of these measurements can be improved by using more highly sampled images and longer recording times at the expense of the specimen drift in the microscope, which is variable from day to day.⁶ By using these settings, it is possible to record the

FIG. 2. (a) A typical diffraction pattern acquired from the annealed film and (b) size of the electron probe acquired from a SiGe region of the finFET specimen orientated in the $[110]$ direction acquired using a convergence angle of 2.4 mrad and a precession angle of 0.25° . (c) A STEM image of an atomically abrupt Si/Si_{0.65}Ge_{0.35} interface with the electron probe superimposed on it. (d) A ε_z deformation profile of the region indicated in (c).

FIG. 3. (a) ϵ_z and (b) ϵ_x maps for the as-deposited film. (c) Deformation profiles extracted across a single pixel from bottom to top for the as-deposited film. (d) ϵ_z and (e) ϵ_x maps for the annealed film, and (f) profiles extracted across a single pixel from bottom to top for the annealed film. (g) ϵ_z and (h) ϵ_x maps for the as-deposited patterned finFET sample. (i) Profiles extracted across a single pixel from bottom to top for the as-deposited patterned finFET samples. (j) ϵ_z and (k) ϵ_x maps for the annealed patterned finFET sample. (l) Deformation profiles for the annealed patterned finFET samples.

deformation maps of the finFET devices during a 3 min total acquisition time. Figure 2(a) shows a typical diffraction pattern of the silicon reference substrate acquired in our experiments orientated on the [011] zone axis using a precession angle of 0.25° such that the first two families of diffraction spots are evenly illuminated. Figure 2(b) shows an image of the precessed probe and its intensity profile with the probe resting on the 150-nm-thick specimen. The 0.8 nm probe is scattered through the specimen, and a FWHM of 2.0 nm is now measured whose value we use as the spatial resolution in the deformation maps.⁸ Figure 2(c) shows a STEM image of an atomically sharp Si/Si_{0.65}Ge_{0.35} interface from a calibration structure. The ϵ_z profile shown in Figure 2(d) has been acquired across the dashed line in (c) with a measurement pitch of 1.4 nm. As there is only one data point in an intermediate position between the strained SiGe and the Si, this suggests that the spatial resolution is higher than 2 nm. An additional advantage is that as the intensity in the diffracted beams is averaged and their positions are measured using template matching, the accuracy of the measurements are robust against experimental problems such as specimen thickness variations and bending.

Figures 3(a) and 3(b) show ϵ_z and ϵ_x maps for the as-deposited film acquired from the indicated region in Figure 1(a). The maps have been calculated using an array of 60×30 diffraction patterns with a sampling of 1 nm. From the ϵ_x map, no deformation is observed relative to the silicon substrate, suggesting that the as-deposited SiGe film is perfectly in epitaxy with the silicon layer. For the ϵ_z map, it can be seen that the lattice parameter is expanded in the Ge

regions. Deformation profiles have been extracted and are shown in Figure 3(e), showing quantitatively that ϵ_z increases from 1.75 just above the silicon layer to 2.70 % at the specimen surface. The EDX results show that this increase corresponds to a change in the Ge content in the specimens rather than effects from relaxation of the thin TEM specimen.¹⁰ For ϵ_x , the results suggest a light compression in the film. The compressive ϵ_x is unexpected but can be attributed to the presence of the buried oxide.⁹ For the annealed specimen, the ϵ_z and ϵ_x maps are shown in Figures 3(c) and 3(d), respectively. Figure 3(f) reveals that ϵ_z is uniformly 1.80%. For the ϵ_x component, the values show slight compression, which is unexpected and could be due to the presence of the surface Si₃N₄ film.

Finite element simulations were performed to assess the effects of thin foil relaxation on the measured deformation. This is especially relevant as surface SiGe films with high Ge contents are known to relax strongly.¹¹ In this case, the situation is more complicated due to the presence of the buried oxide and the Si₃N₄ film on the top of the SiGe wafers. Figure 4(a) shows the simulated deformation profiles for the as-deposited sample in the bulk state and for a 150-nm-thick specimen with and without the surface oxide and Si₃N₄ layers. The simulations suggest that the Si₃N₄ film reduces the relaxation of the SiGe layer by a factor of two. The experimental results are also superimposed onto the simulations, showing that the simulations are only accurate in the central part of the SiGe layers and there is some discrepancy at the top and bottom of the as-deposited film. The experimental results suggest that the top Si₃N₄ layer almost

FIG. 4. (a) FE simulations for the as-deposited bulk and a 150nm-thick TEM specimen with and without the surface Si_3N_4 film compared to experimentally obtained results. (b) Experimentally measured values of deformation for the different specimens plotted against the Ge content measured by EDX. The results are compared with the expected Lagrange ϵ_z as a function of the fully biaxially strained and fully relaxed specimens. For the pre-annealed specimens, the values measured at both the top and the bottom of the Ge layer are shown. For the annealed specimens, the deformation measurement has been taken from the mid-point.

completely prevents thin foil relaxation and the buried oxide introduces a compression.

The expected ϵ_z for biaxially strained SiGe films as a function of the Ge content can be calculated using

$$\epsilon_z = -\frac{2C_{12}}{C_{11}}\epsilon_x + f\left(1 + \frac{2C_{12}}{C_{11}}\right) \quad (1)$$

where

$$f = \frac{a_{\text{SiGe}} - a_{\text{Si}}}{a_{\text{Si}}}. \quad (2)$$

Here, C_{11} and C_{12} are the elastic constants of SiGe and ϵ_x is calculated using Vegard's law using the lattice parameters for Si, a_{Si} , and SiGe, a_{SiGe} . Therefore, for a fully relaxed device, we have $\epsilon_x = \epsilon_z = f$, and if the device is biaxially strained, then $\epsilon_x = 0$ giving

$$\epsilon_z = f\left(1 + \frac{2C_{12}}{C_{11}}\right). \quad (3)$$

In Figure 4(b), the experimentally measured values of ϵ_z are plotted against the Ge concentration measured by EDX. These results are also compared with the values of biaxially strained and fully relaxed specimens, which have been calculated using Equation (1). For the as-deposited specimens where the Ge concentration varies, the deformation values measured at the top and the bottom of the as-deposited films are shown. The results show that for the biaxially strained blanket films, the measurements that have been obtained by EDX, XRD, and N-PED are consistent with the calculated values to within the experimental error.

Figures 3(g) and 3(h) show deformation maps for the as-deposited patterned fins for the ϵ_z and ϵ_x directions, respectively. The experimentally measured values of deformation from bottom to top of the fin are 1.10–1.40% for ϵ_z and 0.95–1.20% for ϵ_x . It has already been shown that for the blanket films, we measure $\epsilon_x = 0$, demonstrating that the SiGe is fully strained. For fully relaxed films, Vegard's law predicts a lattice expansion of ϵ_x in the range of 1.03–1.43% for the Ge concentrations of 32.5–40.5 that are present in the fin. Thus,

the experimental measurements indicate that the patterned structures have relaxed and are no longer strained in this direction. Figures 3(j) and 3(k) show the ϵ_x and ϵ_z maps for the annealed fins. For ϵ_x , the deformation increases from 1.0% at the base of the fin to 1.2% at the top. If the fins are fully relaxed, then a value of 1.04% would be expected, and therefore, again the films are partially relaxed. For ϵ_z , the deformation varies from 1.4% at the base of the finFET, which reduces to 1.1% in the midsection and increases again to 1.5%. The experimentally measured values at the base and top of the finFET are higher than that would be expected for a fully relaxed structure, which suggests that the surface oxide layers are contributing to the deformation.

In conclusion, the deformation in both SiGe blanket films and patterned SiGe fin structures has been examined using N-PED. The advantages of state-of-the-art TEM are that a range of different techniques such as STEM, EDX, and deformation mapping can be quickly be applied to the same specimen to provide quantitative results. Of course, deformation and EDX mapping could in principle be applied simultaneously, especially because it has been shown that precession also reduces EDX quantification problems by reducing electron channeling.¹² We have shown that N-PED can be used to accurately measure the deformation in full sheet SiGe films grown on SOI wafers. The measured deformation for the blanket films is consistent with that measured using XRD, EDX, and theory. For the patterned structures, there are many factors which contribute to the deformation, including the Ge concentration, the finFET shape, the oxidation of the fins and the crystal quality of the film. For this reason, it is important to be able to experimentally measure the deformation. Here, we have shown that the patterned finFET structures are relaxed both before and after anneal. In conclusion, we have shown that today N-PED has the capability to systematically measure the deformation in technologically relevant semiconductor devices with a spatial resolution higher than 2 nm and a precision of around 0.025%.

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