

# Tunable uniaxial vs biaxial in-plane strain using compliant substrates

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In this letter, the relaxation of strained rectangular islands on compliant substrates is used to achieve semiconductor thin films with either uniaxial stress or uniaxial strain in the plane of the film over an area of tens of microns. The work is demonstrated using silicon and silicon–germanium alloy single-crystal thin films, with uniaxial strain values approaching 1%. The biaxially strained SiGe or SiGe/Si films on borophosphosilicate glass (BPSG) were fabricated by a wafer bonding and layer transfer process. When the viscosity of BPSG drops at high temperatures for short times, films patterned in a rectangular shape can move laterally to relieve stress only in one in-plane direction. Thus one can tailor the strain from biaxial to uniaxial in the thin films. © 2005 American Institute of Physics. [DOI: 10.1063/1.2006215]

Strain plays a critical role in electronic properties of many semiconductor materials. Tensile biaxial strain, for example, in silicon has been used to boost electron mobility for *n*-channel metal-oxide-semiconductor field-effect transistors (MOSFETs).<sup>1</sup> Compressive biaxial strain in germanium is found to greatly enhance hole mobility in *p*-channel MOSFETs.<sup>2</sup> Recently, uniaxial strain in silicon thin films has been reported to enhance hole mobility much more than expected<sup>3</sup> and the enhancement was found to hold up at high vertical electric field, at which biaxial-strain induced enhancement usually diminishes.<sup>4</sup>

The conventional methods to engineer strain in semiconductor devices are either heteroepitaxial growth, such as the pseudomorphic growth of strained SiGe alloy on silicon substrates,<sup>5</sup> mechanical approaches, such as substrate bending,<sup>3</sup> or the use of different materials near a device structure to create strain in the device.<sup>4</sup> In this letter, we show how compliant substrates can be used to choose the directionality of strain (biaxial vs uniaxial) to create structures with uniform tunable strain over reasonably large areas (tens of microns). The work is demonstrated using silicon and silicon–germanium alloy single-crystal thin films.

The approach we use is the controlled lateral expansion of compressively strained planar structures on compliant borophosphosilicate glass (BPSG) films.<sup>6–9</sup> A compressively strained pseudomorphic SiGe film is first grown on a sacrificial Si(100) substrate, and then transferred to a second silicon substrate coated with BPSG, followed by the selective removal of the sacrificial substrate and then the patterning of the SiGe layer into islands of edge length 10–300  $\mu\text{m}$ . Typically the BPSG contains 4.4% B and 4.1%

P by weight, and the Si<sub>1-x</sub>Ge<sub>x</sub> layer contains 30% Ge (for an initial biaxial in-plane strain  $\epsilon_0$  of  $-1.2\%$ ) and is 30 nm thick (Fig. 1 inset). When heated above 750 °C, the viscosity of the BPSG drops rapidly, so the SiGe can freely laterally expand to relax the strain. Because the lateral expansion occurs first at the edge of the island and then propagates to the center, large islands relax slower than small islands. The relaxation time is predicted to scale as the square of the island edge length,<sup>9,10</sup> a critical point for this letter. Figure 1 shows data for the biaxial strain in the center of the island, as measured by micro-Raman scattering<sup>9</sup> after a 12 min anneal at 800 °C, as a function of island edge length. Because of the free surface (no stress) in the vertical [001] direction, one can relate the Raman peak shift to the strain  $\epsilon$  in the in-plane [100] and [010] directions by<sup>11</sup>

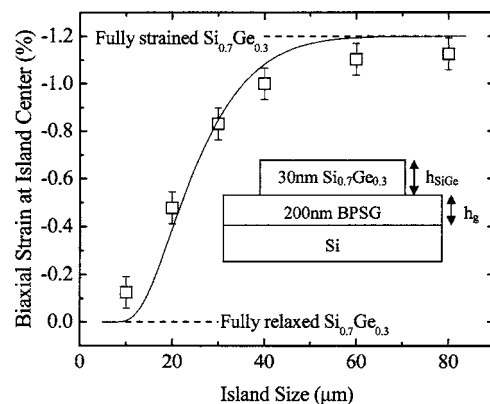


FIG. 1. Island edge size dependence of the biaxial strain in the center of square islands of 30 nm Si<sub>0.7</sub>Ge<sub>0.3</sub> (with initial biaxial strain of  $-1.2\%$ ) after an 800 °C 12 min anneal. The solid line is calculated from two-dimensional lateral expansion theory with parameters described in Table I.

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TABLE I. Deformation potentials ( $p$  and  $q$ ), elastic constants ( $c_{11}$  and  $c_{12}$ ), and Poisson's ratio ( $\nu$ ) of silicon and  $\text{Si}_{0.7}\text{Ge}_{0.3}$  (linearly interpolated between Si and Ge) (see Refs. 12 and 13).

	$p$	$q$	$c_{11}$ ( $10^{10}$ N/m <sup>2</sup> )	$c_{12}$ ( $10^{10}$ N/m <sup>2</sup> )	$\nu$
Si	-1.85	-2.3	16.48	6.35	0.28
$\text{Si}_{0.7}\text{Ge}_{0.3}$	-1.73	-2.20	15.39	5.89	0.28

$$\omega = \omega_0 + \frac{\omega_0}{2} \left( q - \frac{c_{12}}{c_{11}} p \right) (\varepsilon_{[100]} + \varepsilon_{[010]}). \quad (1)$$

Here  $\omega_0$  refers to the peak position for the strain-free case (we measured  $520 \text{ cm}^{-1}$  for silicon and  $502 \text{ cm}^{-1}$  for  $\text{Si}_{0.7}\text{Ge}_{0.3}$ ),  $p$  and  $q$  are the deformation potentials,<sup>12</sup> and  $c_{11}$  and  $c_{12}$  are the elastic constants (Table I<sup>13</sup>). In this biaxial case,  $\varepsilon_{[100]}$  and  $\varepsilon_{[010]}$  are equal by symmetry.

In Fig. 1, one sees that after annealing,  $10 \mu\text{m}$ -square islands have nearly fully relaxed, but the relaxation has not yet reached the center of islands of size  $60 \mu\text{m}$  or larger.<sup>9</sup> Buckling, which can also relieve strain,<sup>9</sup> was negligible in this experiment—the rms surface roughness was  $< 2 \text{ nm}$ . The solid line is a calculation of the two-dimensional relaxation model of Ref. 10, using the parameters of SiGe thickness ( $h_{\text{SiGe}} = 30 \text{ nm}$ ), BPSG thickness ( $h_g = 200 \text{ nm}$ ), and  $\eta = 1.2 \times 10^{11}$  Poise.<sup>9</sup> This viscosity is many orders of magnitude smaller than that of thermally grown  $\text{SiO}_2$ ,<sup>6</sup> which allows lateral displacement across tens of microns within minutes and is why BPSG was used in this work.

In this letter we focus on the SiGe films patterned into *rectangular* islands [Fig. 2(a)] to enable independent control of in-plane strain along different directions. The stripe length was  $150 \mu\text{m}$  in the  $[010]$  direction, and the stripe width ( $[100]$  direction) varied from  $10$  to  $150 \mu\text{m}$ . Consider the extreme case of a long, narrow, compressively strained is-

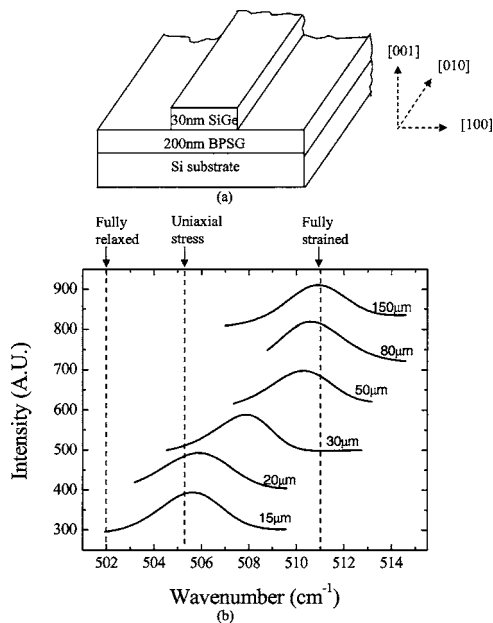


FIG. 2. (a) Diagram of long SiGe stripes on BPSG. (b) Raman spectra for the Si-Si phonon mode measured at the center of  $30 \text{ nm}$   $\text{Si}_{0.7}\text{Ge}_{0.3}$  rectangles of different widths, with initial biaxial strain of  $-1.2\%$ , after a  $15 \text{ min}$  anneal at  $800 \text{ }^\circ\text{C}$  in nitrogen. All rectangles were  $150 \mu\text{m}$  long. Vertical lines indicate conditions of full relaxation, uniaxial stress, and full biaxial strain. The peaks were smoothed according to the Savitzky-Golay technique (see Ref. 18).

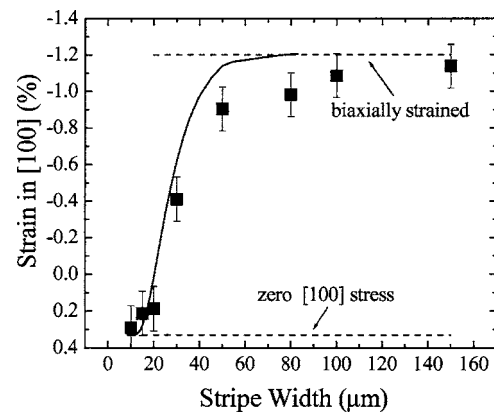


FIG. 3. Strain in the  $[100]$  direction at the center of annealed SiGe rectangular islands as a function of island width, calculated from the positions of the Raman peaks of Fig. 2. The strain condition for zero stress is indicated by a dashed horizontal line. The solid line is calculated to match the data from the one-dimensional lateral expansion theory (see Refs. 14 and 15) with parameters given in Table I.

land. The SiGe stripe is initially under biaxial, compressive in-plane strain  $\varepsilon_0 = -1.2\%$ , with no stress in the  $[001]$  direction because of the free top surface. When the viscosity of the BPSG is reduced at high temperature,  $\sigma_{\text{SiGe}[100]}$ , which denotes the stress in the  $[100]$  direction, should quickly reach zero as the lateral expansion along this short direction is fast, resulting in a condition of uniaxial stress—nonzero stress only in the  $[010]$  direction. The strain  $\varepsilon_{\text{SiGe}[010]}$  in the long direction remains unchanged at  $\varepsilon_0$  because the film has not had time to expand in that direction. From the Poisson effect, one then actually expects a nonzero *tensile* strain of  $-\nu\varepsilon_0 (+0.34\%)$  to be present in the  $[100]$  (zero-stress) direction. Figure 2(b) shows the Raman spectra from the center of islands of SiGe films after a  $15 \text{ min}$   $800 \text{ }^\circ\text{C}$  anneal in nitrogen. On  $150 \times 150 \mu\text{m}$  islands, no shift of the initial Raman peak (no relaxation) is observed, as expected. As the island width is narrowed, the SiGe relaxes perpendicular to the stripe and the Raman peak shifts to lower energies from its initial biaxially strained position of  $511 \text{ cm}^{-1}$ . The shift saturates (with peak at  $505.5 \text{ cm}^{-1}$ ) with island edge length below  $20 \mu\text{m}$ , because for these dimensions all stress has been nearly relaxed in the  $[100]$  direction. From Eq. (1) and the expected final strains described above, one expects a Raman peak at  $505.3 \text{ cm}^{-1}$ , in good agreement with the data. For comparison, a film fully relaxed in both in-plane directions would have a Raman peak near  $502 \text{ cm}^{-1}$ .

Figure 3 summarizes the final strain in the  $[100]$  direction (calculated from the Raman peaks) as a function of stripe width. For stripes wider than  $100 \mu\text{m}$ , little strain relaxation was observed and the original biaxial strain was retained. For stripes narrower than  $20 \mu\text{m}$ , the final strain in the  $[100]$  direction ( $\sim 0.29\%$ ) was indeed near that expected for zero stress in that direction (indicated in Fig. 3). In Fig. 3, the relaxation as a function of stripe width was also modeled with a good agreement using a one-dimensional model.<sup>14,15</sup>

We now demonstrate nonbiaxial tensile strain in a silicon thin film. A  $25 \text{ nm}$  relaxed Si layer was grown on top of the biaxially strained  $30 \text{ nm}$  SiGe layer on the original sacrificial wafer before layer transfer. The bilayer structure of Si/ $\text{Si}_{0.7}\text{Ge}_{0.3}$  was transferred onto  $200 \text{ nm}$  BPSG (with Si now underneath the SiGe) and then patterned into rectangular stripes as before [Fig. 4(a)]. Initially, the  $\text{Si}_{0.7}\text{Ge}_{0.3}$  film

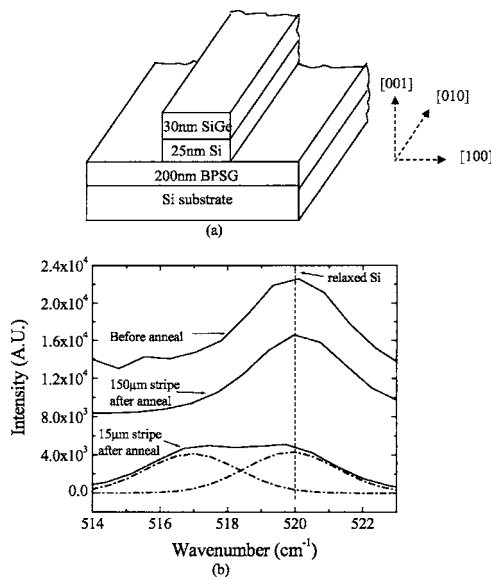


FIG. 4. (a) Diagram of long SiGe/Si stripes on BPSG for creating uniaxial tensile strain in the Si film. (b) Raman spectra of peaks from the silicon layer measured at the center of 150- $\mu\text{m}$ -long 30 nm Si<sub>0.7</sub>Ge<sub>0.3</sub>/25 nm Si rectangular islands (inset) before annealing (identical for 15 and 150  $\mu\text{m}$  widths) and after a 15 min anneal at 800 °C. The solid lines represent the raw data. The dotted lines are the de-convoluted curves after peak fitting for the 15- $\mu\text{m}$ -wide stripe after annealing. The Si film in the 150- $\mu\text{m}$ -wide stripe remains strain free, while the Si film in the 15- $\mu\text{m}$ -wide stripe is under in-plane tensile strain along the short direction ([100]).

was under biaxial compression and the silicon film was strain free. The Raman peak from the Si layer was in the same position as that from the substrate (520  $\text{cm}^{-1}$ ) [Fig. 4(b)]. Upon annealing at 800 °C for 15 min, no relaxation (no shift in Raman peak) in 150  $\mu\text{m}$  squares occurred. Thus we can assume there was no relaxation of rectangular structures in the long ([010]) direction as well, with the Si strain remaining at zero in that direction. In our previous work, the relaxation of smaller *square* islands of this structure was used to generate *biaxial* tensile strain in silicon.<sup>16,17</sup>

Narrow stripes did show a substantial shift in the Raman peak to lower wave numbers (from 520 to 517  $\text{cm}^{-1}$ ) after annealing, splitting away from that of the substrate, indicating tension in the silicon in the [100] direction. In Fig. 4(b), the data after annealing have been deconvoluted into two peaks—one from the strained Si layer and the one from the unstrained Si substrate. As the SiGe expanded in this direction to release its compressive stress, it pulled the Si film into tension, proceeding until a stress balance in the [100] direction was reached:  $\sigma_{\text{SiGe}[100]} \cdot h_{\text{SiGe}} + \sigma_{\text{Si}[100]} \cdot h_{\text{Si}} = 0$ . Thus the final structure has uniaxial strain within the plane of the Si film: zero strain in the [010] direction and tension in the [100] direction. Note after relaxation the SiGe film can be removed by selective etching, to leave an ultrathin uniaxially strained Si film on insulator without any SiGe present.

During relaxation of square islands, it has been shown that the Si/SiGe interface remains coherent (i.e., no misfit dislocations between the two layers)<sup>7</sup> which we also assume here. With such an assumption, the final strain in the Si film in the [100] direction should be

$$\begin{aligned} \varepsilon_{\text{Si}[100]} &= -\varepsilon_0 \frac{(c_{11,b} + c_{12,b} - 2c_{12,b}^2/c_{11,b})h_b}{(c_{11,a} - c_{12,a}^2/c_{11,a})h_a + (c_{11,b} - c_{12,b}^2/c_{11,b})h_b} \\ &= 0.81\% , \end{aligned} \quad (2)$$

where  $a$  and  $b$  refer to Si and SiGe, and the final numerical value of 0.81% assumes the film parameters given earlier. From the observed Raman peak at 517  $\text{cm}^{-1}$  (a shift of 3  $\text{cm}^{-1}$ ) and Eq. (1), one infers a strain of 0.73% in the [100] direction. This is close to the expected 0.81%, giving us confidence that the physical process has been correctly described.

This approach of compliant substrates to achieve tunable strain in different in-plane directions has several practical advantages over other approaches. By using multiple islands sizes and shapes on the same substrate, uniaxial, biaxial, and unstrained films have all been integrated onto a single substrate, without any extra process complexity. The films have a uniform strain over an area of at least tens of microns by tens of microns, which is much larger than the  $\sim 0.1 \mu\text{m}$  strained regions created by process-induced strain approaches such as stressed overlayers or strained epitaxial source/drain regions in MOSFETs. Finally, the process does not rely on lattice mismatched epitaxy and the presence of dislocations, enabling high quality layers. Thus the method might find applications in other material systems beyond silicon or silicon-germanium films.

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<sup>1</sup>J. Welser, J. Hoyt, S. Takagi, and J. Gibbons, Tech. Dig. - Int. Electron Devices Meet. **1994**, 373 (1994).

<sup>2</sup>M. L. Lee, C. W. Leitz, Z. Cheng, A. J. Pitera, T. Langdo, M. T. Currie, G. Taraschi, E. A. Fitzgerald, and D. A. Antoniadis, Appl. Phys. Lett. **79**, 3344 (2001).

<sup>3</sup>B. M. Haugerud, L. A. Bosworth, and R. E. Belford, J. Appl. Phys. **94**, 4102 (2003).

<sup>4</sup>T. Ghani, M. Armstrong, C. Auth, M. Bost, P. Charvat, G. Glass, T. Hoffmann, K. Johnson, C. Kenyon, J. Klaus, B. McIntyre, K. Mistry, A. Murthy, J. Sandford, M. Silberstein, S. Sivakumar, P. Smith, K. Zawadzki, S. Thompson, and M. Bohr, Tech. Dig. - Int. Electron Devices Meet. **2003**, 8 (2003).

<sup>5</sup>E. A. Fitzgerald, Y.-H. Xie, M. L. Green, D. Brasen, A. R. Kortan, J. Michel, Y.-J. Mii, and B. E. Weir, Appl. Phys. Lett. **59**, 811 (1991).

<sup>6</sup>K. D. Hobart, F. J. Kub, M. Fatemi, M. E. Twigg, P. E. Thompson, T. S. Kuan, and C. K. Inoki, J. Electron. Mater. **29**, 897 (2000).

<sup>7</sup>H. Yin, K. D. Hobart, F. J. Kub, S. R. Shieh, T. S. Duffy, and J. C. Sturm, Appl. Phys. Lett. **82**, 3853 (2003).

<sup>8</sup>H. Yin, R. Huang, K. D. Hobart, J. Liang, Z. Suo, S. R. Shieh, T. S. Duffy, F. J. Kub, and J. C. Sturm, J. Appl. Phys. **94**, 6875 (2003).

<sup>9</sup>H. Yin, R. Huang, K. D. Hobart, Z. Suo, T. S. Kuan, C. K. Inoki, S. R. Shieh, T. S. Duffy, F. J. Kub, and J. C. Sturm, J. Appl. Phys. **91**, 9716 (2002).

<sup>10</sup>R. Huang, H. Yin, J. Liang, K. D. Hobart, J. C. Sturm, and Z. Suo, Mater. Res. Soc. Symp. Proc. **695**, 115 (2001).

<sup>11</sup>I. De Wolf, J. Raman Spectrosc. **30**, 877 (1999).

<sup>12</sup>S. C. Jain, B. Dietrich, H. Richter, A. Atkinson, and A. H. Harker, Phys. Rev. B **52**, 6247 (1995).

<sup>13</sup>M. Neuberger, *Group IV Semiconducting Materials* (IFI/Plenum, New York, 1971).

<sup>14</sup>L. B. Freund and S. Suresh, *Thin Film Materials* (Cambridge University Press, Cambridge, 2004).

<sup>15</sup>P. D. Moran and T. F. Kuech, J. Electron. Mater. **30**, 802 (2001).

<sup>16</sup>H. Yin, K. D. Hobart, R. L. Peterson, F. J. Kub, S. R. Shieh, T. S. Duffy, and J. C. Sturm, Tech. Dig. - Int. Electron Devices Meet. **2003**, 53 (2003).

<sup>17</sup>H. Yin, K. D. Hobart, R. L. Peterson, F. J. Kub, S. R. Shieh, T. S. Duffy, and J. C. Sturm, IEEE Trans. Electron. Device (submitted).

<sup>18</sup>A. Savitzky and M. J. E. Golay, Anal. Chem. **36**, 1627 (1964).