5

Low Thermal Budget Techniques for Enhancing Crystallization

5.1 Introduction

It was demonstrated in the previous chapters that the processing temperature of polycrystalline semiconductor materials is always higher than the CMOS backend thermal budget, and this is mainly due to the relatively high transition temperature from amorphous to polycrystalline. In Chapter 4, the possibility of reducing the transition temperature by alloying silicon and germanium was investigated. The main drawback of this approach is that the Ge content should be higher than 65%, which is not fully compatible with standard VLSI processes [1], and which might affect device reliability since Ge is more affected by humidity compared to silicon [2, 3], and surface roughness is also increased [4].

The crystallization temperature of silicon can be significantly reduced to 500°C or lower by using metal induced crystallization (MIC), and the realized devices have outstanding performance compared to those employing conventional solid-phase crystallization [5, 6]. The crystallization thermal budget can be further reduced by using laser induced chemical vapor deposition (LICVD) and/or laser induced crystallization (LIC), which was proposed for low thermal budget applications such as TFTs [7], solar cells fabricated on glass substrates [8], and for monolithic integration of MEMS devices on top of standard driving electronics using SiGe as an structural material [9].

The following sections give an overview of the effect of metal induced crystallization and pulsed laser annealing on the morphology of silicon germanium.
5.2 Metal Induced Crystallization of SiGe

The main concept of metal induced crystallization is based on reducing the solid phase crystallization temperature of silicon due to the interaction between the silicon covalent bond and the free electrons of certain metals such as aluminum [6], copper [10], gold [11], silver [12], or nickel [13]. The crystallization temperature of silicon depends on the type of metal and varies between 350°C for Ag [12] and 485°C for Cu [10]. The main disadvantage of this technique is the metal contamination into the crystallized film, which might imply a significant reduction in the electrical conductivity especially for semiconductor materials. Thus, this technique might not be suitable for applications that require high electrical resistivity or a wide range of conductivity by controlling the doping level. Furthermore, the presence of the metal atoms inside the crystallized film might affect the internal dissipation, thus reducing the quality factor, which means that this technique might not be appropriate for the realization of high quality resonators.

Zhang et al. [14] investigated the effect of Ni seeding on the structural properties of Si$_{1-x}$Ge$_x$ deposited at 450°C for TFTs. Compared to excimer laser annealing and solid phase crystallization, this approach is expected to give better uniformity over large areas and to be more cost effective. Figure 5.1 shows that increasing the thickness of the Ni layer to 0.2 nm results in a sharp increase in the grain size. Further increase of the thickness of the Ni film results in a decrease in the grain size till it reaches a minimum at 0.5 nm. Increasing the thickness of the Ni layer from 0.5 to 1 nm is accompanied by a gradual increase in the grain size. For Ni layers thicker than 1 nm, the grain size is almost independent of the thickness of the Ni layer. This behavior is mainly due to the increase in density of Ni nuclei with thickness. For very thin layers (~0.05 nm), the Ni nuclei density is $7.5 \times 10^8$/cm$^2$ [14]. The deposition of a thin Ni layer

![Figure 5.1](image-url) Poly Si$_{1-x}$Ge$_x$ grain size as a function of the Ni seed layer thickness. (From: [14].)
directly on clean c-Si results in NiSi$_2$ [15], which has a cubic structure and a lattice constant closely matched to that of silicon, thus enhancing the preferential growth rate of silicon as confirmed by the Raman spectrum displayed in Figure 5.2. This explains the steep increase in the grain size observed in Figure 5.1. The increase in the Si peak observed in the Raman spectrum displayed in Figure 5.2 (see peak located at 500 m$^{-1}$) shows that this process proceeds until the Ni thickness reaches 0.2 nm. The decreased grain size accompanying the increase in the thickness of the Ni film more than 0.2 nm is due to the preferential crystallite growth of Si$_{1-x}$Ge$_x$ induced by the seeding effect of the Ni nuclei rather than the spontaneous formation of Si$_{1-x}$Ge$_x$ nuclei on the surface [14]. As the Ni thickness exceeds 1 nm, the Ni nuclei coalesce to form a continuous film, and accordingly, the nuclei formation of Si$_{1-x}$Ge$_x$ on Ni starts to be the decisive factor for grain growth. Hence, the grain size is independent of the thickness of the Ni seed layer, as clear from Figure 5.1. It is also interesting to note that the changes in the intensity of the Si and Ge peaks observed in the Raman spectrum displayed in Figure 5.2 shows that the thickness of the Ni layer not only affects the grain size but also affects the film composition.

Chen et al. [16] investigated the possibility of using Au to enhance the crystallization of amorphous Si$_{1-x}$Ge$_x$ deposited on glass substrates for pin infrared detectors. This technique allows maximizing optical gain and response speed. Figure 5.3 shows the effect of the annealing temperature and period on the induced lateral crystallization (ILC). This figure shows that the lower the hydrogen content in the film, the larger the induced crystallization. This is mainly due to the higher density of dangling bonds which captures the gold atoms, encouraging the formation of meta stable Au-SiGe compound, which enhances Si diffusion and causes dendritic growth of SiGe crystals.

![Figure 5.2 Raman spectra of poly Si$_{1-x}$Ge$_x$ films deposited on Ni-seeding layers with different thickness. (From: [14]. © 2003 American Institute of Physics.)](image-url)
5.2.1 MIC for MEMS Application

The previous section gave an overview of the concept of metal induced crystallization. At this point it is instructive to investigate the impact of the metal seed layer on grain microstructure, mean stress, and stress gradient of the MEMS structural layer. It is recommended to select a seed layer that is compatible with standard CMOS fabrication tools, as this allows using advanced facilities for depositing and patterning the MEMS structural layer. Hence, a reasonable choice for the seed layer would be either aluminum or titanium. Due to the attractive physical properties of silicon germanium, it might be interesting to consider this material as the MEMS structural material. To enhance crystallization, the structural layer can be deposited directly on top of the metal seed layer or the seed layer can be inserted in the middle of the structural layer as shown schematically in Figure 5.4.

For the same seed layer type and the same Si$_{1-x}$Ge$_x$ deposition temperature, crystallization enhancement depends strongly on the deposition technique being either LPCVD or PECVD. By investigating the TEM cross sections in Figure 5.5 [17], it is clear that LPCVD Si$_{0.37}$Ge$_{0.63}$ deposited at 370°C on top of Al is fully polycrystalline with a columnar grain microstructure. This might be due to the low growth rate, which enables better mixing and diffusion of Al with Si$_{1-x}$Ge$_x$ to form an alloy having lower crystallization temperature. On the other hand, PECVD Si$_{0.29}$Ge$_{0.71}$ deposited at the same temperature has a few columnar grains embedded in an amorphous matrix as inferred from the TEM cross section displayed in Figure 5.5(b). Depositing PECVD Si$_{0.29}$Ge$_{0.71}$ on top of Ti reduces significantly the grain size of the columnar crystals, and the amorphous phase dominates, as demonstrated in Figure 5.5(c). Finally, the TEM cross section displayed in Figure 5.5(d) illustrates the effect of a long anneal at a
temperature close to the deposition temperature for PECVD Si$_{0.29}$Ge$_{0.71}$ deposited on top of a thin Al layer. Annealing results in the generation of small crystals (~10 nm) embedded inside the amorphous matrix. Thus, it can be concluded that for low thermal budget applications, it is recommended to use Al as this enhances crystallization at lower temperatures.

It is important now to have an idea about the mean stress and the sheet resistance in the different layers under consideration. The data displayed in Table 5.1 shows that in general metal induced crystallization results in a compressive stress. The minimum stress is obtained by LPCVD depositions; while depositing PECVD Si$_{1-x}$Ge$_x$ on top of Ti results in the highest mean compressive stress. On the other hand, the measured sheet resistances of the different Si$_{1-x}$Ge$_x$ layers deposited on top of Al are close to each other in spite of the significant differences in the grain microstructure. This might indicate that the conductivity is dominated by the bottom Al layer, or it might be due to the diffusion of Al throughout the Si$_{1-x}$Ge$_x$ layer. To discriminate between the two effects, it might be instructive to check the Al distribution throughout the film using secondary ion mass spectroscopy (SIMS). Figure 5.6 shows that for the LPCVD Si$_{0.37}$Ge$_{0.63}$
Figure 5.5  TEM cross section of $\text{Si}_{1-x}\text{Ge}_x$ deposited at 370°C: (a) LPCVD $\text{Si}_{0.37}\text{Ge}_{0.63}$ deposited on Al; (b) PECVD $\text{Si}_{0.29}\text{Ge}_{0.71}$ deposited on Al; (c) PECVD $\text{Si}_{0.29}\text{Ge}_{0.71}$ deposited on Ti; and (d) PECVD $\text{Si}_{0.29}\text{Ge}_{0.71}$ deposited on Al and annealed for 21 hours at 370°C.  
(From: [17].)
Figure 5.6 Al profile in Si$_{1-x}$Ge$_x$ deposited at 370°C as determined by SIMS: (a) LPCVD Si$_{0.38}$Ge$_{0.62}$; (b) PECVD Si$_{0.29}$Ge$_{0.71}$; and (c) PECVD Si$_{0.29}$Ge$_{0.71}$ annealed for 21 hours at 370°C. (From: [17].)
layer, Al diffused significantly throughout the total film thickness. This large amount of diffusion can be explained by the long deposition period (4 hours) [17]. On the other hand, for the as-grown PECVD films, there is a significant amount of Al in the top 50 nm of the film [Figure 5.6(b)]. This might explain the relatively low sheet resistance, in spite of the significant amorphous phase observed in the TEM cross sections [Figure 5.5(b)]. Annealing a stack of 50-nm Al/0.29 μm Si0.29Ge0.71 for 21 hours at 370°C resulted in a significant increase in the Al concentration in the Si1-xGe layer up to a depth of 1 μm from the Al/SiGe interface, as clear from the SIMS profile displayed in Figure 5.6(c). Thus, the low sheet resistance observed for the different stacks is mainly due to the significant diffusion of Al or Ti into the Si1-xGe film.

For MEMS applications we are interested in having layers with low mean tensile stress. Hence, it is instructive to investigate the possibility of reducing mean stress by sandwiching the metal layer in between two structural layers [Figure 5.4(c, d)]. Such structure is characterized by a stronger texture even for the plasma enhanced deposited films, as clear from the TEM cross section displayed in Figure 5.7. Accordingly, the stress tends to be more tensile. The mean

<table>
<thead>
<tr>
<th>Stack (defined from bottom to top)</th>
<th>Stress</th>
<th>Sheet Resistance</th>
</tr>
</thead>
<tbody>
<tr>
<td>SiO2/Al/LPCVD Si0.37Ge0.63</td>
<td>−97 MPa</td>
<td>0.45 ± 0.0123 Ω/Sq</td>
</tr>
<tr>
<td>SiO2/Al/PECVD Si0.29Ge0.71</td>
<td>−196 MPa</td>
<td>0.524 ± 0.0156 Ω/Sq</td>
</tr>
<tr>
<td>SiO2/Ti/PECVD Si0.29Ge0.71</td>
<td>−307 MPa</td>
<td>8.98 ± 0.331 Ω/Sq</td>
</tr>
<tr>
<td>SiO2/Al/PECVD Si0.29Ge0.71 (annealed for 21 hours)</td>
<td>−197 MPa</td>
<td>0.55 ± 0.035 Ω/Sq</td>
</tr>
</tbody>
</table>

Table 5.1 Stress and Sheet Resistance in Different Stacks of Si1-xGe Deposited at 370°C (From: [17].)

Figure 5.7 TEM cross section for a stack of Ti/PECVD Si29Ge71/Al/PECVD Si0.29Ge0.71. (TEM cross section has been done by Hugo Bender and Oliver Richard IMEC, Belgium.)
stress reported for a stack of either Ti/Si$_{0.29}$Ge$_{0.71}$/Al/Si$_{0.29}$Ge$_{0.71}$ or Al/ Si$_{0.29}$Ge$_{0.71}$/Al/Si$_{0.29}$Ge$_{0.71}$ is 66 MPa or 90 MPa tensile, respectively [17].

In addition to mean stress, stress gradient is also an important issue, as it might dramatically affect the functionality of surface micromachined structures. By inspecting the deflection profile of the surface micromachined cantilevers displayed in Figure 5.8, it is clear that a stack of Ti/PECVD Si$_{1-x}$Ge$_x$ is always bending upwards, indicating that the bottom layers are much more compressive than the upper ones. On the other hand, the profiles of cantilevers realized by stacks of Ti/ Si$_{0.29}$Ge$_{0.71}$/Al/Si$_{0.29}$Ge$_{0.71}$ or Al/ Si$_{0.29}$Ge$_{0.71}$/Al/Si$_{0.29}$Ge$_{0.71}$ indicate that the upper layers are more compressive than the lower ones. This is most probably due to the fact that the middle Si$_{0.29}$Ge$_{0.71}$ layer is more crystalline than the upper one (as inferred from TEM cross sections). Further reduction of the stress gradient might be possible by fine-tuning the thickness of the individual layers of the stack. It is also interesting to note that the minimum bending profile is obtained for a stack of Al/PECVD Si$_{0.29}$Ge$_{0.71}$, in spite of the fact that the mean stress is highly compressive.

King et al. [9] investigated the effect of using Ni as a seed layer for enhancing the crystallization of Si$_{1-x}$Ge$_x$ having a Ge content between 25% and 30%. To study the effect of Ni, a thin Ni layer has been deposited on top of α-Si$_{0.7}$Ge$_{0.3}$ film. After annealing at 425°C for a long time, the structure is completely crystalline, as clear from the TEM cross section displayed in Figure 5.9. The mean stress of the crystallized film has been found to be compressive (~135 MPa). Whereas, the strain gradient is $3 \times 10^{-4}$ μm$^{-1}$ [9], which is rather high for applications requiring large suspended structures such as inertial sensors.

![Figure 5.8](image_url) Bending profile of surface micromachined cantilevers realized by different stacks of thin metal films and PECVD Si$_{1-x}$Ge$_x$. (From: [17].)
5.3 Laser Induced Crystallization

Over the last two decades excimer laser annealing was considered an efficient low thermal budget technique for locally modifying the physical properties of thin films without introducing any damage or modifications to the underlying layers. Early motivation for using pulsed laser annealing was to control the grain size and crystallinity of amorphous silicon [18, 19], which was attractive for the fabrication of thin film transistors having high field effect mobility on glass substrates [20, 21]. Also, it was commonly used to tune the electrical properties of implanted semiconductors [22–24] especially for devices that require shallow doped regions. Furthermore, it was demonstrated that pulsed laser annealing can noticeably improve the efficiency of solar cells as it enhances the minority carrier diffusion length [25, 26]. The fact that pulsed laser annealing reduces the defect density due to the melting and recrystallization mechanism widened the application of this technique to improve the electrical properties of metal induced crystallized amorphous silicon thin films [27]. Microsecond and nanosecond lasers are now used for low temperature deposition of a wide variety of materials like bismuth telluride [28], aluminum nitride [29], vanadium oxide [30], and ferromagnetic materials [31].

Many studies have been performed to understand the effect of laser annealing on the average grain size and stress of silicon films deposited by LPCVD [18] or RF sputtering [32]. The effects of pulsed laser annealing in locally tuning the electrical and structural properties of silicon [33–35] and Si_{1-x}Ge_x [36–39] are widely investigated. A novel application for pulsed laser annealing is to control stress and stress gradient in surface micromachined MEMS that are integrated on top of standard prefabricated driving electronics [40]. This application is much more challenging as MEMS implies using rather
thick layers and requires the optimization of mechanical and electrical properties, and accordingly, the laser annealing conditions are completely different from those conventionally implemented for TFTs or pulsed laser deposition.

5.3.1 Laser Setup

The most important criteria for selecting the laser source are the wavelength, pulse fluence, repetition rate, and the intensity profile along the beam cross section. The type of the MEMS structural material determines the wavelength, as it should be efficiently absorbed by the material. The combination of the pulse fluence and rate determines the grain microstructure, electrical conductivity, surface roughness, mean stress, and stress gradient. To guarantee a uniform treatment all over the wafer, the beam should have a uniform intensity across its wave front, which can be realized by using a set of lenses for beam reshaping and homogenization as explained in [41].

5.3.2 Effect of Laser Annealing on Grain Microstructure, Electrical Conductivity, and Surface Roughness

One of the attractive features of pulsed laser annealing is that it enables a broader range of Ge contents, and hence, the required physical properties of the MEMS structural layer can be tailored according to the application without being limited to high Ge contents as in the case of conventional thermal treatments discussed in Chapter 4.

To study the interaction between laser pulse and the MEMS structural layer, we consider Si$_{1-x}$Ge$_x$ films, having low Ge content (∼25%), and deposited at 400°C on top of thermal oxide. The as-grown films are amorphous, as clear from the TEM cross section displayed in Figure 5.10 [42]. Exposing this film to a single, 440 mJ/cm$^2$ laser pulse, having a wavelength of 248 nm and duration of 24 ns, crystallizes the film, resulting in a grain microstructure characterized by two distinct regions: an upper low defect density region having blocky grains and a bottom region having high defect density fine grains. Thus, this fluence lies in the near complete melting regime [43]. Increasing the pulse fluence is accompanied by an increase in the maximum temperature, and accordingly, the melt depth is increased. This results in the generation of blocky coarse grains close to the surface and fine bottom grains. The depth of the blocky grain region significantly increases by increasing the pulse fluence, where as the fine grain zone diminishes, as clear from the TEM cross sections in Figure 5.11(b, c). Increasing the number of pulses to 100, at an average pulse energy density varying between 400 and 600 mJ/cm$^2$, the thickness of the columnar layer is slightly increased [Figure 5.11(d, e)] and the defect density is somewhat lower than for single pulse. On the other hand, applying 100 pulses at a fluence of 790 mJ/cm$^2$
converts the grain microstructure from equiaxed to columnar [Figure 5.11(f)]. The XRD patterns displayed in Figure 5.12 show that after laser crystallization, the film always has a dominant \{111\} texture regardless of the average laser energy density.

Figure 5.10  TEM cross section of as-grown Si\textsubscript{0.75}Ge\textsubscript{0.25} deposited at 400°C and 800 mTorr.  
(From: [42].)

Figure 5.11  TEM cross sections demonstrating the effect of average laser energy and number of pulses on grain microstructure of Si\textsubscript{0.75}Ge\textsubscript{0.25} films deposited at 400°C: (a) single pulse at 440 mJ/cm\textsuperscript{2}; (b) single pulse at 616 mJ/cm\textsuperscript{2}; (c) single pulse at 790 mJ/cm\textsuperscript{2}; (d) 100 pulses at 440 mJ/cm\textsuperscript{2}; (e) 100 pulses at 616 mJ/cm\textsuperscript{2}; and (f) 100 pulses at 790 mJ/cm\textsuperscript{2}.  
(From: [42].)
It is also interesting to correlate structural changes associated with laser annealing to changes in electrical conductivity. Figure 5.13 shows that for a single laser pulse, the resistivity tends to increase with the laser fluence. The noticeable increase at 790 mJ/cm² is likely due to the formation of an equiaxed grain microstructure [see Figure 5.11(c)]. On the other hand, the squares in

**Figure 5.12**  XRD diffraction analyses of laser-annealed poly-Si₀.₇₅Ge₀.₂₅ deposited at 400°C: (a) single laser pulse at 760 mJ/cm²; (b) single laser pulse at 440 mJ/cm²; (c) 100 laser pulses at 10 Hz and 760 mJ/cm²; and (d) 100 laser pulses at 10 Hz and 440 mJ/cm². (From: [42].)

![XRD diffraction analyses](image)

**Figure 5.13**  Dependence of resistivity of Si₀.₇₅Ge₀.₂₅ deposited at 400°C on pulse fluence for single pulse (diamonds) and 100 laser pulses at 10 Hz (squares). (From: [42].)
Figure 5.13 shows that the resistivity drops significantly with multiple pulses. This is more pronounced at the lowest and highest energy densities. The low resistivity at high energies may be due to the combined effects of the transformation from equiaxed to columnar grain structure and a noticeable increase in average grain size [see Figure 5.11(d)]. The low resistivity at low energies may be due to the reduced defect density, as inferred from the TEM cross-sections displayed in Figure 5.11. It is interesting to note that similar results are reported for laser-crystallized silicon [44].

For MEMS, surface roughness is a critical issue that might affect the functionality of the realized devices. Hence, it is instructive to investigate how it is affected by laser fluence and pulse rate. Figure 5.14 provides quantitative data on the effect of laser energy density on surface roughness of a Si$_{0.12}$Ge$_{0.88}$ film deposited at 400°C. Exposing this film to a single laser pulse at 792 mJ/cm$^2$ decreases the root-mean-square (rms) surface roughness from 160 to 80 nm. The TEM cross sections displayed in Figure 5.15 show the effect of the number of pulses on surface roughness. In this case the Ge content is much higher than the film considered in Figure 5.10. This high Ge content enhances the crystallization of the as-grown material deposited at 400°C, as clear from the TEM cross section displayed in Figure 5.15(a), where the surface roughness is also evident. Comparing Figure 5.15(b) and Figure 5.15(c), it can be concluded that surface roughness is reduced by increasing the number of pulses from 1 to 10. Similar behavior is reported for laser-annealed PECVD amorphous-Si deposited at temperatures between 310°C and 360°C [6]. It should be noted that a large number of laser pulses (100) results in a dramatic increase in surface roughness, as is shown in Figure 5.15(d).
In addition to the changes introduced by the laser pulse to the grain microstructure and surface roughness, it is essential to check if the laser treatment has any influence on doping or germanium segregation. By investigating the SIMS profile displayed in Figure 5.16 we notice that exposing a 5 µm thick Si_{0.08}Ge_{0.92} film to a single laser pulse having a fluence of 728 mJ/cm² increased the boron concentration close to the surface by almost a factor of five. Furthermore, there is a noticeable increase in the oxygen content, which extends for 1 µm below the film surface. On the other hand, after laser treatment, the Ge profile seems to be more uniform over the first 100 nm of the film.

5.3.3 Film Melt Depth

In order for the pulsed laser annealing technique to be suitable for MEMS monolithic integration on top of standard prefabricated electronics, the absorbed energy must be retained within the Si_{1-x}Ge_x film. The penetration depth of the laser-generated heat into the absorbing layer is characterized by the
larger of the optical absorption length and the thermal diffusion length determined at the end of the laser pulse. In general, for the 248-nm wavelength, the thermal absorption length dominates over the optical absorption for Si$_{1-x}$Ge$_x$ and it will depend mainly on the initial microstructure of the SiGe film (being either amorphous or polycrystalline) as well as any phase transformations that occur during the laser pulse. The TEM cross sections displayed in Figure 5.17 provide some evidence for the depth over which the laser pulse altered the film microstructure. By investigating this figure, we notice that a 1-µm-thick Si$_{0.4}$Ge$_{0.6}$ film deposited at 400°C and 650 mTorr exhibits polycrystalline regions that nucleated during growth, presumably at the growth surface. Exposing this film to a single laser pulse [Figure 5.17(b)] or 10 laser pulses at 1 Hz

Figure 5.16 SIMS profile of germanium, boron, and oxygen in 5 µm thick Si$_8$Ge$_92$ film: (a) as-grown; and (b) after a single laser pulse having a fluence of 782 mJ/cm$^2$. (From: [42].)
[Figure 5.17(c)] at an average energy density of 728 mJ/cm² results in crystallization of the amorphous material and recrystallization of the regions that were originally crystalline. The crystallization depth is limited to approximately 0.8 µm, as inferred from the thickness of the bottom amorphous layer in Figure 5.17(b, c). It should be noted that the crystallization temperature for a 60% Ge film is ~450°C [45]. Therefore, if the Si₁₋ₓGex film thickness is greater than 1 µm, there should be no significant thermal impact on the underlying layers (which can withstand exposure to temperatures at or below 450°C). If the film thickness is less than 1 µm, then the thermal penetration depth can be reduced by lowering the pulse fluence. Figure 5.18 demonstrates this for a 160-nm-thick Si₀.₄Ge₀.₆ film deposited at 400°C: a 616-mJ/cm² pulse completely penetrates the film [Figure 5.18(a)]; by reducing the laser fluence to 120 mJ/cm², the melt depth is noticeably reduced [Figure 5.18(b)].

At this point it is interesting to note that the effective penetration depth of the laser pulse (i.e., the depth through which microstructural changes are evident) depends strongly on the grain microstructure of the as-grown film. If the film is initially polycrystalline, the activation energy required for grain growth is much higher than that required for crystallization of an amorphous film, as previously reported for silicon [46], and hence, for the same pulse fluence, the molten depth is significantly reduced as compared to the amorphous film. Figure 5.19(a) clarifies this issue where it can be noticed that the changes in the grain microstructure of poly Si₄₄Ge₅₆ after being exposed to a single laser pulse at 320 mJ/cm² are limited to a depth of 100 nm, most likely indicative of a shallow molten depth and a limited diffusion of the self-propagating silicon germanium liquid. For an amorphous film, this same effective penetration depth is achieved.
by a pulse having energy 2.5 times lower than this value [see Figure 5.18(b)] due to the lower melting temperature of the amorphous phase and the longer diffusion length of the molten.

Finally, it is interesting to compare pulsed laser annealing to rapid thermal annealing (RTA), which is also a low-thermal-budget crystallization technique. For this comparison, we consider amorphous Si$_{0.75}$Ge$_{0.25}$ films deposited at 400°C and 650 mTorr (Figure 5.10). After annealing at 500°C for 3 minutes, the film is composed of randomly oriented tiny crystals [Figure 5.20(a)]; in contrast, if the film is exposed to a single laser pulse at 760 mJ/cm$^2$, the grains are columnar and much larger [Figure 5.20(b)]. The differences in grain size and microstructure are more evident in the higher sheet resistance of the RTA film, as inferred from the data displayed in Figure 5.21. It is clear that even if the

Figure 5.18  TEM cross section of 180-nm-thick Si$_{0.82}$Ge$_{0.18}$ deposited at 400°C and 1 Torr and exposed to a single laser pulse: (a) 616 mJ/cm$^2$; and (b) 120 mJ/cm$^2$. (From: [42].)
RTA temperature is increased to 900°C, the sheet resistance is still more than a factor of 2.5 higher than that of the laser-annealed film.

### 5.3.4 Effect of Laser Annealing on Mean Stress and Stress Gradient

For an understanding about the effect of pulsed laser annealing on mean stress and stress gradient, we start by considering as-grown, 5-µm-thick, poly Si$_{0.08}$Ge$_{0.92}$ film deposited at 400°C and 650 mTorr. TEM cross sections show that the as-grown film is polycrystalline with a V-shaped, columnar texture [Figure 5.15(a)]. The stress gradient of the as-grown film is relatively low, as clear from the bending profile of the surface micromachined cantilevers displayed in Figure 5.22(a). Exposing this film to a single laser pulse having an energy density varying from 244 to 568 mJ/cm$^2$, at atmospheric pressure, results in an increase in both the stress gradient and in the mean tensile stress, as clear from
the deflection of the surface micromachined cantilevers and the rotating pointer [47] displayed in Figure 5.22(b–d). The increased tensile stress, especially for the top layers of the film, is mainly due to the recrystallization of the film by the laser pulse, which results in contractions against the grain boundaries. This effect decreases gradually below the film surface due to the limited penetration of the laser pulse, and accordingly, the stress gradient is increased.
Thus, to control stress gradient using pulsed laser annealing, we should start with an as-grown material having upper layers more compressed than the lower ones. This can be achieved either by tuning the deposition conditions (which is not always feasible) or by depositing Si$_{1-x}$Ge$_x$ bilayers having different mean stresses and tuning the energy of the laser pulse impinging the surface of the top or bottom film (or perhaps both) until the stress gradient is completely eliminated. Figure 5.23 gives more insight about how this concept can be

Figure 5.22 Effect of single laser pulse on stress gradient and mean stress of 5-$\mu$m-thick poly Si$_8$Ge$_{92}$ film deposited at 400°C and 650 mTorr: (a) as-grown; (b) laser annealed with a single pulse at 244 mJ/cm$^2$; (c) laser annealed with a single pulse at 408 mJ/cm$^2$; and (d) laser annealed with a single pulse at 568 mJ/cm$^2$.

(From: [40]. © 2004 IEEE. Reprinted with permission.)
implemented. In this case, we consider the effect of single laser pulse on two different Si$_{1-x}$Ge$_x$ stacks deposited at 400°C [Figure 5.23(a–c)] or at 425°C [Figure 5.23(d–f)]. For the first stack, the as-grown bottom film is amorphous, whereas the top film is polycrystalline and the bilayer bends upwards [Figure 5.23(a)].

To study the effect of pulsed laser annealing on eliminating the stress gradient, first the bottom Si$_{0.6}$Ge$_{0.4}$ layer is deposited, and then the sample is removed from the LPCVD furnace and exposed to a single laser pulse at 320 mJ/cm$^2$ under normal atmospheric conditions. Then the sample is cleaned in hydrofluoric acid for 1 minute to eliminate the native oxide. Immediately after

Figure 5.23 Surface micromachined cantilevers, 5 µm wide, and having a length varying from 25 µm to 1 mm, realized by Si$_{1-x}$Ge$_x$ bilayer: (a–c) films deposited at 400°C, bottom film: 1.4-µm Si$_{0.6}$Ge$_{0.4}$, top film: 1.4-µm Si$_{0.44}$Ge$_{0.56}$; (d–f) films deposited at 425°C, bottom film: 0.9-µm Si$_{0.44}$Ge$_{0.56}$, top film: 0.6-µm Si$_{0.57}$Ge$_{0.43}$. (a) As-grown films; (b) bottom film exposed to a single laser pulse at 320 mJ/cm$^2$; (c) bottom film exposed to single laser pulse at 380 mJ/cm$^2$; (d) as-grown; (e) bottom film exposed to a single laser pulse at 320 mJ/cm$^2$; and (f) bottom film exposed to a single laser pulse at 400 mJ/cm$^2$. (From: [40]. © 2004 IEEE. Reprinted with permission.)
that, the sample is placed in the LPCVD furnace to deposit the top $\text{Si}_{0.44}\text{Ge}_{0.56}$ film. By investigating the cantilever profile displayed in Figure 5.23(b), it is clear that the stress gradient is significantly reduced. This is mainly due to the increased tensile stress in the bottom $\text{Si}_{60}\text{Ge}_{40}$ film after laser annealing. If the laser energy is tuned in such a way that makes the tensile stress balance the bending moments around the interface of the two films, the stress gradient is almost eliminated. A slight increase in the laser energy results in an undesired out-of-plane deflection as demonstrated in Figure 5.23(c). Now, if we consider the as-grown $\text{Si}_{1-x}\text{Ge}_x$ stack deposited at 425°C [Figure 5.23(d)], it would be logical to expose the top film, only, to a single laser pulse to eliminate the out-of-plane downward bending. But Figure 5.23(e) shows that stress gradient is completely eliminated if the bottom film is exposed to a single laser pulse at 320 mJ/cm$^2$ and the top film is not laser treated. Further increase in the energy of the laser pulse, impinging the surface of the bottom film, results in an upward bending rather than an expected downward bend.

For a better understanding of this behavior, it is instructive to refer to the TEM cross sections of the as-grown and laser annealed films displayed in Figure 5.24. These cross sections show that the single laser pulse affected not only the microstructure of the bottom film, but also the microstructure of the top film. Figure 5.24(b) shows that the texture of the top film is more columnar when deposited on a laser-annealed lower film. Thus, we can classify the effect of laser annealing according to the type of the as-grown film being either amorphous or polycrystalline. If the bottom film is amorphous, which is the case at 400°C deposition temperature, the single laser pulse results in crystallization and a huge tensile stress. This stress has a dominant effect in determining the bending moments of the composite film, rather than the effect of the changes in the microstructure of the top film. If the bottom film is initially polycrystalline, the changes in the grain microstructure of the top film play the major role in determining the stress gradient. Finally, it should be noted that in addition to eliminating stress gradient, the single laser pulse significantly reduces average stress. The average stress of the bottom layer of the bilayer deposited at 425°C is reduced from 120 MPa compressive to 20 MPa compressive. An SEM image of surface micromachined interdigitated structure realized by a $\text{Si}_{1-x}\text{Ge}_x$ bilayer is displayed in Figure 5.25. It is clear from the figure that the low mean stress results in perfect alignment of the stationery and movable comb fingers.

5.4 Summary and Conclusion

This chapter explores the possibility of enhancing crystallization of the MEMS structural layer by either metal induced crystallization or pulsed laser annealing. The grain microstructure and texture of PECVD $\text{Si}_{1-x}\text{Ge}_x$ deposited at 370°C on
top of thin Al or Ti metal films is investigated. In general, Al enhances the crystallization of as-grown Si$_{1-x}$Ge$_x$, but the mean stress is highly compressive. The electrical conductivity is close to that of Al and this is mainly due to the
significant diffusion of Al into the Si$_{1-x}$Ge$_x$ films. Stacking Ti/Si$_{1-x}$Ge$_x$/Al/Si$_{1-x}$Ge$_x$ noticeably strengthens texture and results in a relatively low mean tensile stress. The stress gradient needs further reduction.

The effect of pulsed excimer laser annealing (ELA) on the microstructure, texture, electrical conductivity, and surface roughness of Si$_{1-x}$Ge$_x$ films is investigated. ELA is a useful technique for locally modifying the physical properties of Si$_{1-x}$Ge$_x$ films, and it is suitable for annealing films deposited on top of completed CMOS electronics. Laser-annealed in situ–doped p-type Si$_{1-x}$Ge$_x$ films are characterized by a dominant {111} texture, an average grain size of 300 nm, and an average resistivity of 0.7 m$\Omega$.cm. The maximum depth to which the microstructure can be significantly modified is 0.8 $\mu$m in Si$_{0.8}$Ge$_{0.2}$ that is amorphous as deposited. This effective penetration depth can be reduced by either decreasing the laser pulse fluence or by using polycrystalline films. Finally, the surface roughness can be reduced with a moderate number of laser pulses ($\sim$10 pulses). Stress gradient can be completely eliminated by pulsed laser annealing of Si$_{1-x}$Ge$_x$ bilayers.

Finally, it is worthwhile to mention that hydrogenated microcrystalline silicon germanium deposited at temperatures varying between 300°C and 400°C is currently investigated as a material suitable for low thermal budget MEMS applications [48, 49]. Preliminary results on boron in situ doped as grown material shows that electrical resistivity, mean stress and strain gradient can be tuned respectively to an optimal value of 7 m$\Omega$.cm, 34 MPa tensile and $1.5 \times 10^{-4}$ $\mu$m$^{-1}$ [49]. This approach might be attractive as you can realize the required MEMS physical properties in situ without any additional steps. The main drawback for such approach is after depositing the microcrystalline film, the wafer temperature cannot exceed the deposition temperature otherwise there are voids due to hydrogen evolution.

References


