Stress Stability of Poly-SiGe and Various Oxide Films in Humid Environments

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ABSTRACT

This paper presents the stress stability of thin films for MEMS structural and sacrificial layers. The average residual stresses of the thin films were monitored via wafer curvature measurement over a long period of time. Poly-Si, poly-SiGe, poly-Ge and thermally growth SiO₂ films are found to be stable in humid environments. Moisture makes LPCVD and TEOS-based PECVD SiO₂ films more compressive over time. Multi-layer thin film stress is modeled with the same methodology used to derive the Stoney Equation [1].

INTRODUCTION

Reliability specifications typically require MEMS structural layers to have long-term material stability. Poly-SiGe is a promising structural material for modular integration of MEMS with electronics, because of its relatively low deposition temperature (<450°C) [2]. SiO₂ films can also be deposited at high rates at low temperatures, and are attractive for use as sacrificial layers because they can be easily etched selectively with respect to poly-SiGe.

In past research, results from the analysis of wafer curvature over time appeared to indicate that poly-SiGe films experience a stress drift in humid environments [3]. This created a major challenge for the plausibility of poly-SiGe MEMS technology. The stress drift phenomenon has been investigated further in this research. Results show that the residual stress of poly-SiGe films is, in fact, stable in ambient conditions. The apparent residual stress drift of the poly-SiGe films in Ref. [3] was caused by the unstable low temperature LPCVD oxide on the backside of the wafers.

EXPERIMENTAL DETAILS

The average residual stresses of various thin films were determined with wafer curvature measurements before and after thin film deposition using a Tencor FLX-2320. Long term average residual stress monitoring was done with various layer stacks as shown in Fig. 1. Poly-Si, poly-SiGe, and poly-Ge, as well as various oxides, were deposited and removed from single crystal silicon (SCS) wafers under different conditions as summarized in Table I. Initial wafer curvature measurements were taken from the bare Si wafer for the single layer stacks (Fig. 1a & b), and from the oxidized wafer before poly-Si, poly-SiGe or poly-Ge deposition for the bi-layer stacks (Fig. 1c).

![Fig. 1 Layer stacks for stress monitoring](image-url)
Table I. Deposition and removal conditions of the various thin films.

<table>
<thead>
<tr>
<th>Film</th>
<th>Deposition Method</th>
<th>Removal Method</th>
</tr>
</thead>
<tbody>
<tr>
<td>Poly-Si (0.6 µm)</td>
<td>LPCVD @ 620 °C</td>
<td>RIE @ 60 °C</td>
</tr>
<tr>
<td>Poly-SiGe (0.2-1 µm)</td>
<td>LPCVD @ 400-450 °C</td>
<td>RIE @ 60 °C</td>
</tr>
<tr>
<td>Poly-Ge (0.4 µm)</td>
<td>LPCVD @ 350 °C</td>
<td>RIE @ 60 °C</td>
</tr>
<tr>
<td>Dry Thermal Oxide (1200 Å)</td>
<td>Thermally growth @ 1050 °C</td>
<td>HF solution @ 21 °C</td>
</tr>
<tr>
<td>Wet Thermal Oxide (1600 Å)</td>
<td>Thermally growth @ 1050 °C</td>
<td>HF solution @ 21 °C</td>
</tr>
<tr>
<td>LPCVD Oxide (2 µm)</td>
<td>LPCVD @ 450 °C</td>
<td>HF solution @ 21 °C</td>
</tr>
<tr>
<td>PECVD Oxide (0.5 µm)</td>
<td>PECVD @ 390 °C</td>
<td>Single side deposition</td>
</tr>
</tbody>
</table>

RESULTS and DISCUSSIONS

Wafers used in Ref. [3] to monitor the stress stability had a poly-SiGe film deposited on top of a 2 µm LPCVD oxide, as shown in Fig. 1c. This layer stack is commonly used in MEMS: the thick oxide serves as a sacrificial layer and the poly-SiGe serves as the structural layer. Under further investigation, the results reported in Ref. [3] have been reproduced in this work, as plotted in Fig. 2. The measured stresses of poly-Si, poly-SiGe and poly-Ge on LPCVD oxide become more tensile over time, but all poly-Si, poly-SiGe and poly-Ge films on thermal oxide or SCS are stable. These results indicate a problem with the LPCVD oxide. It should also be noted that poly-Si films on LPCVD oxide are more stable than poly-SiGe and poly-Ge films on the same oxide. This is because, during the poly-Si deposition, the LPCVD oxide is annealed at 620 °C.

![Fig. 2 Stress stability of poly-Si, SiGe and Ge on various substrates.](image1)

![Fig. 3 Stress stability of poly-SiGe on LPCVD oxide](image2)

Further experimentation was done with the poly-SiGe on LPCVD oxide wafers (Fig. 3). If the backside poly-SiGe films of two similar wafers are removed at different times, the drift profiles and absolute stresses of the wafers are nearly identical, with an offset in the x-axis. When the backside poly-SiGe and LPCVD oxide films are both removed, the stresses of the topside poly-SiGe and LPCVD oxide films become stable. This data suggests that the stress drift reported in Ref. [3] is due solely to the instability of the LPCVD oxide film exposed to the ambient on the backside of the wafer.
Low temperature (450 °C) LPCVD oxide films are known to be porous and of poor quality [4]. In this investigation, different experiments were done with LPCVD oxide wafers to characterize its stability (Fig. 4). The residual stress of the as-deposited LPCVD oxide wafer becomes more compressive over time. The “drift” rate slows over time. The backside oxide removal date does not affect the measurement. Putting the LPCVD oxide films in a desiccated environment slows down the stress drift. Annealing the films at 615 °C for five hours results in a tensile film that becomes more compressive more slowly than unannealed films. The above facts suggest that the absorption of ambient water into the LPCVD oxide films is the major cause of the observed stress drift. As water is absorbed, the films become more compressive. The 615 °C N2 annealing appears to densify the oxide and decreases the diffusion constant. It is also shown in Fig. 4 that if a wafer is annealed 136 days after deposition in N2 at 425 °C for an hour, the resultant stress was approximately that of a fresh oxide film. The stress drift of this 425 °C annealed wafer is faster than a monitor wafer with the backside also etched on day 136. Since the 425 °C anneal is lower than the deposition temperature of 450 °C, it is unlikely that the anneal rearranges or densifies the oxide molecules; rather, the anneal most likely drives out the absorbed water. Ultimately, the stress drift of the 425 °C annealed wafer is faster than the control wafer on day 136 due to a larger moisture gradient.

Returning to the data for poly-SiGe on LPCVD oxide wafers in Fig. 3, as the backside oxide films became more compressive, the poly-SiGe film appeared to become more tensile. Data in Fig. 3 also indicates that poly-SiGe is an effective barrier to moisture. The backside LPCVD oxide did not drift until it was exposed to the ambient after the poly-SiGe layer was removed. Finally, the stress of the frontside LPCVD oxide did not drift under the poly-SiGe cap.

Fig. 5 shows the stress stability of different oxide films. The stress of the 450 °C LPCVD oxide becomes more compressive while the stresses of the 1050 °C dry and wet thermal oxides are nearly stable. The stresses of TEOS-based PECVD oxides deposited at 390 °C also become more compressive, but at a much higher rate than the LPCVD oxide films. Also, the TEOS-based PECVD oxide films begin to saturate within 30 days of being exposed to the ambient. Their remarkably high stress drift rate is expected since TEOS-based oxides are generally more porous [5] and the diffusion constant of water is much higher. In contrast, silane-based PECVD oxide films were found to have better stress stability in humid environments (data not presented here).

Fig. 4 Stress stability of LPCVD oxide

Fig. 5 Stress stability of various oxides
MODELING THIN FILM STRESS

Let us now turn to the stress analysis of \( n \) thin films on the frontside and \( m \) thin films on the backside of a single-crystal substrate wafer. In this case, the thin films experience nearly ideal biaxial stress, or plane stress. Moreover, the residual stresses in the films are uniform over the wafer, not varying with direction or position. We shall assume that the thin films are linear, isotropic materials. This is reasonable for amorphous and poly-crystalline materials. The single-crystal substrate, on the other hand, is anisotropic. However, proper choice of the Young’s modulus will minimize the error introduced by this assumption [6].

To proceed, we assume that there exists a neutral axis in the substrate whose position is unaffected by the existence of the films. Further, it is assumed that the deflections are small and the shear forces are negligible. Finally, we shall assume the residual stress in the thin films is small compared to the substrate stiffness, allowing us to neglect the contraction/extension of the wafer caused by the thin films. The material and geometric constants of the substrate and thin films are listed in Table II. Fig. 6 illustrates the setup for two frontside and two backside films.

**Table II. Material and geometric constants**

<table>
<thead>
<tr>
<th></th>
<th>Substrate</th>
<th>Film ( f1 )</th>
<th>\ldots</th>
<th>Film ( fn )</th>
<th>Film ( b1 )</th>
<th>\ldots</th>
<th>Film ( bm )</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Young’s Modulus</strong></td>
<td>( E_s )</td>
<td>( E_{f1} )</td>
<td>\ldots</td>
<td>( E_{fn} )</td>
<td>( E_{b1} )</td>
<td>\ldots</td>
<td>( E_{bm} )</td>
</tr>
<tr>
<td><strong>Thickness</strong></td>
<td>( H )</td>
<td>( h_{f1} )</td>
<td>\ldots</td>
<td>( h_{fn} )</td>
<td>( h_{b1} )</td>
<td>\ldots</td>
<td>( h_{bm} )</td>
</tr>
<tr>
<td><strong>Residual Stress</strong></td>
<td>( \sigma_s = 0 )</td>
<td>( \sigma_{f1} )</td>
<td>\ldots</td>
<td>( \sigma_{fn} )</td>
<td>( \sigma_{b1} )</td>
<td>\ldots</td>
<td>( \sigma_{bm} )</td>
</tr>
</tbody>
</table>

![Diagram](a) Before Release  
(b) After Release

Fig. 6 Cross-sectional views of a substrate with thin films on both sides for \( n = 2 \) and \( m = 2 \).

For uniform, constant biaxial stress, the constitutive relation relating stress, \( \sigma \), to strain, \( \varepsilon \), is:

\[
\sigma = \frac{E'}{1 - \nu} \varepsilon = E' \varepsilon
\]

where \( E' \) and \( \nu \) are the biaxial elastic modulus and Poisson’s ratio of the material, respectively. Conceptually, we imagine the wafer clamped so that, despite the thin film residual stresses, there is no deflection of the wafer or stress-relaxation of the thin-films. This is the “Before Release” state. Next, we imagine the wafer released from its clamps and deforming to an equilibrium state: the “After Release” state.

The system is in static equilibrium after release. Therefore, the net moment per unit length of the system after release, \( M_{ar} \), must equal the net moment before release, \( M_{br} \). Before release, the moments are due only to the thin films:
\[
M_{br} = \sum_{i=1}^{n} \int_{-H/2}^{H/2} \sigma_{f i} z dz - \sum_{j=1}^{m} \int_{-H/2}^{H/2} \sigma_{b j} z dz
\]

\[
F(i) = \sum_{i=1}^{n} h_{f i}, \quad B(j) = \sum_{j=1}^{m} h_{bj}
\]

\[
M_{br} \approx \frac{H}{2} \left[ \sum_{i=1}^{n} \sigma_{fi} h_{fi} - \sum_{j=1}^{m} \sigma_{bj} h_{bj} \right]
\]  \hspace{1cm} (2)

After release, the substrate develops a balancing moment via pure bending, \( M_{ar} \):

\[
M_{ar} = \int_{-H/2}^{H/2} \sigma_{s} z dz = \frac{H^2}{12} \frac{E_s}{R} \int_{-H/2}^{H/2} z^2 dz = \frac{H^2}{12} \frac{E_s}{R} \frac{H^3}{3}
\]  \hspace{1cm} (3)

where \( R \) is the radius of curvature of the wafer. Recall that, for static equilibrium after release, we require \( M_{ar} = M_{br} \). Hence, equating (2) and (3):

\[
\frac{1}{6} \left( \frac{E_s}{R} \right) H^2 = \sum_{i=1}^{n} \sigma_{fi} h_{fi} - \sum_{j=1}^{m} \sigma_{bj} h_{bj}
\]  \hspace{1cm} (4)

The Tencor FLX-2320 measures the change in the radius of curvature of a substrate caused by the stress of a thin film. The stress of the thin film is calculated with a simplified version of Eq. (4):

\[
\sigma_{measured} = \frac{1}{6} \left( \frac{E_s}{R} \right) H^2 = \sigma_f
\]  \hspace{1cm} (5)

which is appropriate for single layer thin films as shown in Fig. 1a & b.

For the bi-layer stacks shown in Fig. 1c, Eq. (4) can be simplified to:

\[
\sigma_{measured} = \frac{1}{6} \left( \frac{E_s}{R} \right) h_{SiGe}^2 = \sigma_{SiGe} + \sigma_{ox(f)} \frac{h_{ox(f)}}{h_{SiGe}} - \sigma_{ox(b)} \frac{h_{ox(b)}}{h_{SiGe}}
\]  \hspace{1cm} (6)

If the stresses of the frontside oxide and the backside oxides are cancelled out, Eq. (6) would be equivalent to Eq. (5) and the measured stress would be the true SiGe film stress. However, the backside oxide absorbs moisture and becomes more compressive; whereas the stress of the frontside oxide is constant under the SiGe cap. Therefore, the measured stress is not the true stress of SiGe film once the backside oxide starts to absorb moisture. An apparent stress drift of the SiGe film is observed and qualitative agreement with the oxide stress change is found.

CONCLUSIONS

The average residual stress of poly-Si, poly-SiGe, poly-Ge, and thermal oxide are stable in ambient conditions. Poly-SiGe remains a promising material for modular MEMS integration. LPCVD and TEOS-based PECVD oxides absorb moisture and become more compressive in a humid environment. Due to their high deposition rates, LPCVD and PECVD oxides are often used as MEMS sacrificial layer. However, the stress drift in sacrificial materials is not anticipated to affect the mechanical properties of the MEMS structure layers since they are eventually removed.

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REFERENCES