Impact of Substrate Thickness on the Degradation in Multicrystalline Silicon

Utkarshaa Varshney, Moonyong Kim, Muhammad Umair Khan, Phillip Hamer, Catherine Chan, Malcolm Abbott, and Bram Hoex

Abstract—Light and elevated-temperature-induced degradation (LeTID) is a well-known phenomenon that reduces the bulk lifetime in silicon wafers. The cause of this degradation mechanism is still under investigation. However, a wide range of empirical trends that correlate LeTID with multiple physical and processing parameters have been reported, including the observation that wafers thinner than 120 µm do not show significant LeTID. In this work, we extend that study by varying the thickness of the wafers, the temperature of the firing step, and testing LeTID at the accelerated stability testing conditions. We demonstrate that the extent of degradation reduces with the thickness of the wafer, in agreement with the earlier work. However, silicon wafers with a thickness below 120 µm still suffer from LeTID when fired at sufficiently high temperatures, demonstrating that thinner wafers are not inherently immune to LeTID. By performing accelerated testing using a high-intensity laser and fitting the degradation and regeneration data, we observe that thinner wafers do not necessarily exhibit a faster recovery, as suggested earlier. However, their reduced degradation extent could be a consequence of relatively higher out-diffusion of hydrogen per unit volume in thinner wafers during firing. We further report that the method used for thinning the wafers results in a variation in the surface morphology of the samples, and that may partly be responsible for the observed correlation between the thickness of the wafers and LeTID extent. Finally, we discuss how these new findings can be explained by the involvement of hydrogen and other impurities in LeTID.

Index Terms—Firing, hydrogen, light and elevated-temperature-induced degradation (LeTID), silicon, surface morphology, wafer thickness.

I. INTRODUCTION

LIGHT and elevated-temperature-induced degradation (LeTID) is known to deteriorate the performance of silicon wafers and cells. It has been shown to exist in nearly all silicon wafer types, namely multicrystalline silicon (mc-Si) [1], Cz-Si [2], float zone (FZ-Si) [3], and n-type Si [4]. It thus became a pressing concern in the past years for both the academic and industrial photovoltaic communities. Its mitigation is crucial as the current industrially dominant [5] passivated emitter and rear contact solar cell was reported to exhibit more than 10% rel. loss in the power output [6].

Despite numerous studies that focused on examining the cause of this degradation, the structure and composition of the underlying defect that causes LeTID is unknown. However, important clues to determine the primary source of degradation have come through numerous studies that tested the impact of sample processing on the behavior of LeTID. Earlier, an increase in the extent of degradation with an increase in the firing temperature has been reported, while minimal or no degradation was observed in the samples fired at low temperatures or the samples without firing [6]–[8]. It has also been shown that using different dielectric layers [9]–[11], the application of gettering steps [12], variations in the firing speed [13], and the postprocess annealing [8], [14] impact both the degradation and recovery.

Particularly relevant to the work presented in this article are the results by Bredemeier et al. [15], where it was shown that the extent of degradation strongly depends on the thickness of the silicon wafer. It was also implied that the silicon substrates thinner than 120 µm would not suffer from a significant degradation. In light of the existing LeTID theories, this observation can be explained by two potential sources of degradation. Numerous recent publications that focused on identifying the root cause of LeTID have suggested metallic impurities as a part of the defect complex causing LeTID [6], [7], [15], [16]. Titanium, molybdenum, tungsten [6], nickel, and cobalt [15] were identified as potential candidates contributing to LeTID based on their activation energies and capture cross-sectional ratios. This is a reasonable assumption as the mc-Si material has been long known to suffer from metallic impurities pertaining to its crystallization process [17]. Furthermore, the involvement of metallic impurities in causing LeTID is also supported by the observations of reduced degradation by gettering [12]. In parallel, there is a growing consensus on the involvement of hydrogen in causing this degradation [3], [4], [11], [15], [18]–[21], which can explain the existence of LeTID in all-silicon types [1]–[4]. It also agrees with the reported dependence of degradation extent on the properties of the hydrogen-rich SiNₓ:H [10], [11], [22], and a lower degradation when using a hydrogen-lean AlOₓ surface passivation layer [9], [23]. Additionally, degradation was also observed when a silicon wafer is exposed to a

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microwave-induced remote hydrogen plasma [19], implying that it is not the firing process itself but rather the introduction of hydrogen that activates LeTID. Furthermore, another model suggesting the plausibility of a metal-hydrogen complex causing LeTID has also been presented [20], [21].

As a strong decrease in the extent of LeTID was reported by thinning the silicon substrate, it is interesting to understand the cause of this reduced degradation and examine whether thin wafers are intrinsically immune to LeTID. In this article, we extend the earlier work from Bredemeier et al. [15] by studying the impact of silicon wafer thickness (115–190 μm thick mc-Si lifetime test structures) on LeTID, examining the stability of the samples when fired at three different set peak firing temperatures to determine the potential role of thermal variations in thinner substrates, and investigating the differences in surface morphology between the samples. We confirm a strong correlation between the reduction in wafer thickness and the extent of LeTID. In addition, we report that at sufficiently high firing temperatures, it is possible to observe LeTID also in thinner wafers. Furthermore, it was demonstrated that the process of thinning the wafers used in this work led to some unintended yet significant variations in the surface morphology, which should be considered when trying to determine the impact of the chemical thinning processes on the reduction in the LeTID extent.

II. Sample Preparation and Experiment

For this study, p-type mc-Si wafers (∼190 μm and 1.6 Ω·cm) from neighboring ingot positions (sister wafers) were chosen as substrates. All the samples were first cleaned and underwent a POCl₃ gettering diffusion to form an n⁺ emitter on both the wafer sides with a sheet resistance $R_{sh} \sim 60$ Ω/sq. Subsequently, these n⁺ layers were chemically removed using a solution of nitric acid and hydrofluoric acid (HNO₃/HF) for multiple durations to obtain wafers in the range of 115–190 μm thickness. The wafer thicknesses were measured using a gauge meter (Mitutoyo, 7001-10) at multiple positions on each wafer and the average values are reported. These wafers were then symmetrically passivated with aluminum oxide (AlOₓ:H) ~20 nm and silicon nitride (SiNx:H) ~100 nm using a remote plasma-enhanced chemical vapor deposition (r-PECVD, Roth & Rau MAiA) at a substrate temperature of 350 °C and 400 °C, respectively [23] to create symmetrical lifetime test structures with AlOₓ:H/SiNx:H passivation stacks. The samples were further laser cleaved into 3.9 cm × 3.9 cm tokens and fired at a set peak firing temperature of 775 °C (actual sample temperature: 610 ± 16 °C), 855 °C (actual sample temperature: 714 ± 9 °C), and 950 °C (actual sample temperature: 839 ± 6 °C) in a fast firing belt furnace (Schmid) at a speed of 450 cm/min. The actual sample temperatures were measured for each sample with a type-K thermocouple and a Datapaq (DP 5, Fluke process instruments) thermal profiler on identical samples. Average values over three measurements are reported. These samples were kept at room temperature in the dark prior to the stability testing experiments. Fig. 1 shows the process sequence of the sample preparation and the testing.

A. LeTID Testing

To test for LeTID, a set of samples with multiple thicknesses was tested under an illumination intensity of 1 kW/m² using halogen lamps in ambient conditions on a hotplate at 130 °C [24]. Another identical set of tokens underwent accelerated testing using 938 nm high-intensity laser at an illumination intensity of 30.5 kW/m² (~54 suns) at 168 ± 6 °C [25]. The effective minority carrier lifetimes $\tau_{eff}$ were characterized ex situ using quasi-steady-state photoconductance measurements using a photoconductance lifetime tester (WCT120-TS, Sinton Instruments). $\tau_{eff}$ was extracted at an injection level $\Delta n$ equal to 0.1 x the base doping level. For the comparison of the defect concentration, the normalized defect density (NDD) was the chosen metric and was calculated [26] as

$$NDD = \frac{1}{\tau(t)} - \frac{1}{\tau_{fired}}$$

where $\tau(t)$ and $\tau_{fired}$, respectively, are the effective minority carrier lifetimes at time $t$ during the stability testing and after firing.

The NDD results observed under high-intensity illumination were fitted using a single exponential decay function to calculate the degradation and recovery rate constants

$$NDD(t) = N_p \cdot \left(1 - e^{-k_{deg} \cdot t}\right) \times \left(e^{-k_{rec} \cdot t}\right)$$

where $k_{deg}$ and $k_{rec}$ signify the degradation and recovery rate constants, respectively, and $N_p$ implies the defect precursor concentration. The quality of the fit was determined using the regular residual ($r^2$) analysis. The closeness of $r^2$ to one implies the goodness of the fit with the measured data.

B. Optical and Surface Morphology Characterization

To understand the influence of sample processing on the optical parameters, the reflectance of the samples was measured with a UV-Vis spectrophotometer (LAMBDA 650, Perkin Elmer). The surface morphology of the samples was characterized by scanning electron microscopy (SEM) images (FEI Nova NanoSEM 450). This was accompanied by the atomic force microscopy (AFM, Bruker Dimension Icon SPM) measurement to examine the variation in texturing within the samples. The
AFM data were analyzed using Gwyddion [27]. A mean plane subtraction leveled the data and the minimum value was shifted to zero to extract the surface areas.

III. RESULTS

A. Impact of Substrate Thickness on LeTID

The progression of NDD as a function of time for mc-Si samples with a thickness ranging from 115–190 μm, fired at a sample temperature of 839 ± 6 °C, is shown in Fig. 2. Two identical sets of samples were tested for LeTID at 130 °C under 1 kW/m² illumination using halogen lamps [see Fig. 2(a)] and at a sample temperature of 168 ± 6 °C under high-intensity laser illumination of 30.5 kW/m² [see Fig. 2(b)].

In both cases, an initial increase in NDD was observed, the extent of which was significantly influenced by the wafer thickness. The graphs clearly show a strong positive correlation between increasing wafer thicknesses and higher defect concentrations, except for the 125 μm sample; however, the difference is within the measurement error. This trend is consistent with the results reported by Bredemeier et al. [15]. However, an interesting observation was the clear difference in the shape of the graphs, pointing toward a significant difference in the defect formation and recovery dynamics. Earlier, Kwapil et al. [28] and Bredemeier et al. [29] have also reported an (almost) linear dependence of degradation rate on Δn during dark annealing and illuminated annealing up to 1.5 suns. For the recovery rates, while the exact dependence with the Δn is still unknown, one study by Kersten et al. [30] showed a clear correlation of the recovery rates with Δn. With our data in Fig. 2(a), the time required to reach the NDD max on the thinnest wafer (∼115 μm) was about ∼30× (from ∼10³ to ∼2.9 × 10⁴ s) faster than that of the thickest wafer (∼190 μm). However, when a high illumination intensity was used, the difference in the time to reach NDD max was almost negligible. Therefore, it suggests that the testing conditions affect the reaction dynamics. Based on the Δn dependence of reacton rates, this difference can be explained. Tables I and II state Δn at the implied one sun (at 1 kW/m²) or ∼54 suns (at 30.5 kW/m²) at the different stages of LeTID (initial, maximum NDD, and fully recovered).

The one sun data were extracted from Sinton lifetime raw data, whereas the data at ∼54 suns were extrapolated by fitting the lifetime curves using the injection-dependent lifetime spectroscopy using the methods proposed by Nampalli et al. [31]. For both the samples, the variation in Δn at ∼54 suns is less than a factor of 4. However, the Δn value at implied one sun for the halogen lamp testing condition showed variation from 1.7 × 10¹⁵/cm³ to 1.1 × 10¹⁵/cm³ and 1.3 × 10¹⁵/cm³ to 0.034 × 10¹⁵/cm³ for the 115 μm and 190 μm samples, respectively. This suggests a dramatic reduction in Δn at one sun for the thicker wafer during the degradation and recovery cycle. Considering the correlation of both the degradation and recovery rates on Δn, the rates were

![Fig. 2](image_url)

**Fig. 2.** Progression of NDD as a function of time for mc-Si samples fired at a sample temperature of 839 ± 6 °C and tested at (a) 130 °C using halogen lamps under 1 kW/m² illumination. The dotted lines are guide to the eye and (b) at 168 ± 6 °C under high-intensity laser illumination of 30.5 kW/m². The connecting lines show the actual fitting of the data. The inset shows the result in different scales.

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expected to be significantly smaller for the thicker samples as $\Delta n$ was significantly lower because of the lower minority charge carrier lifetime. In contrast, the small variation of $\Delta n$ at $\sim$54 suns (because of the Auger limit) indicates the small variation in reaction rates. Although it is clear that the degradation and recovery rates have a correlation with $\Delta n$, quantifying their correlation will require further investigation.

It is worth mentioning that the NDD, as shown in Fig. 2, is attributed to a bulk Shockley-read-hall (SRH) defect with a capture cross-sectional ratio of 28.4±6, which is similar to the values reported corresponding LeTID [6], [16], [32], [33]. However, minor instabilities in the surface passivation quality cannot be ruled out, despite the surface-related degradation typically occurring on significantly longer timescales [34]. In addition, the corresponding surface saturation current density $J_{0s}$ values are not shown as the extraction of $J_{0s}$ to examine the deterioration of AlO$_x$H/SiN$_x$H passivated surfaces seems inaccurate. It is likely because of the loss of fixed negative charges, which leads to the $J_{0s}$ analysis being inconclusive, as also reported earlier by Sperber et al. [35].

### B. Variation in Reaction Rates

The extracted reaction rate constants corresponding to the degradation and recovery data at high intensity [see Fig. 2(b)] are shown in Fig. 3. It was found that the rate constants of degradation and regeneration ($k_{deg}$ and $k_{rec}$, respectively) were similar for all the samples, and only minor variations were observed. The $r^2$ from each fitting is also shown, where all the values were in the range of 0.94–0.98, which suggests the close agreement between the fitting and the experimental results. The fact that both $k_{deg}$ and $k_{rec}$ were found not to be thickness dependent seems to disagree with the findings of a faster regeneration in thinner wafers, as reported in [15]. However, it is also important to note that Bredemeier et al. [15] fitted their data using a diffusion-based model where the recovery was assumed to be because of the diffusion of interstitial metal atoms to the wafer surfaces, whereas in this work, the degradation and regeneration were modeled using an exponential decay function to extract the rate constants. In addition, these results were obtained under very high illumination intensity as compared to one sun illumination used in the study by Bredemeier et al. [15].

### C. Variation in Firing and Impact of Peak Firing Temperature on the Degradation Extent

The firing process has shown to be a significant factor in controlling LeTID, with an increase in the firing temperature leading to the increased degradation intensity [8], [36]. This is because (at least in part) of an increase in the amount of hydrogen released into the silicon bulk with a higher firing temperature. Therefore, it is important to consider the variations in the actual sample temperature during the firing. Fig. 4(a) showed the sample temperatures when they were fired at a set peak firing temperature of 775 °C (actual: 610±16 °C), 855 °C (actual: 714±9 °C), and 950 °C (actual: 839±6 °C). We ensured that the temperature of each sample with different thickness is similar for each set peak firing temperature by firing them in the same run on the same day when the furnace was in thermal equilibrium. Interestingly, the thinnest sample experienced the lowest temperature regardless of the setpoint, which was significantly lower compared with the thickest wafer (as much as 30 °C for a setpoint temperature of 775 °C). It is worth noting that the surface morphology of the samples might also have an influence on the lower temperatures observed in thinner wafers. As thinner wafers are optically more reflective (discussed in Section III-D), they will likely absorb less heat during the firing. As the thermal budget is a critical factor in LeTID studies, the low degradation observed in the 115 μm sample might have been influenced by the lower sample temperature. In addition to the actual peak temperature, the ramp-down rates have also been shown to have an impact on the degradation extent [13], [14]. From 32 to 37 s, the temperature is roughly stable at 300 °C. This occurred in the nonactive cooling zone, which has also been typically observed in our previous studies [4] and is attributed to our firing furnace. However, as shown in Fig. 4(b), the overall thermal profiles with the set peak temperature of 950 °C did not show a significant variation for the samples with different thicknesses.

To understand the influence of the firing temperature on the extent of degradation, a parallel set of samples fired at three different peak firing temperatures of 610±16, 714±9, and 839±6 °C was treated under accelerated LeTID testing conditions. Fig. 5 shows the evolution of NDD as a function of time for the thinnest [115 μm, see Fig. 5(a)] and the thickest [190 μm, see Fig. 5(b)] sample. The extent of maximum degradation is nearly one order of magnitude higher for the 190 μm sample; however, a positive dependence of the defect concentration with the firing temperature was observed in both cases. It is important to note that a clear degradation and regeneration cycle was observed even for the 115 μm sample when fired at a sufficiently high
firing temperature of 839 °C, which implies that thinner wafers are not immune to LeTID. In addition, a negligible degradation was observed in the 115 \( \mu \text{m} \) sample when fired at lower temperatures. Therefore, using both thinner wafers and lower firing temperatures can provide a potential LeTID mitigation strategy.

D. Influence of Surface Morphology

Fig. 6 shows the SEM images of selected mc-Si samples used in this study. It was observed that the chemical etching process that was used for thinning the wafers resulted in a significant variation in the surface morphology of the samples. The cross-sectional image of the thickest sample [190 \( \mu \text{m} \), see Fig. 6(a)] revealed a surface with inverted spherical pits that are similar to commonly used iso-texture (∼2.6 \( \mu \text{m} \)), the depth of which was reduced to ∼1.5 \( \mu \text{m} \) for the 155 \( \mu \text{m} \) sample [see Fig. 6(b)] and changed to an almost planar surface in the case of the thinnest sample [115 \( \mu \text{m} \), see Fig. 6(c)].

To quantify the variation in the surface morphology of the samples, Fig. 7 shows the correlation between the surface area to the projected area ratio measured by AFM and maximum NDD \( \text{NDD}_{\text{max}} \) extracted from Fig. 2(b) as a function of wafer thickness. It shows a clear increase in \( \text{NDD}_{\text{max}} \) with the increase in the thickness of the wafers, along with an enhancement in the surface area ratio. This implies that the process of thinning the wafers led to a partial loss of their texture, as also demonstrated in Fig. 6. It is worth noting that the difference in surface areas with wafer thickness was relatively small compared with the drastic variation observed in their degradation extents. Therefore, it was unlikely to be the primary factor. For, e.g., the 155 \( \mu \text{m} \) sample showed a similar surface area as the 135 \( \mu \text{m} \) sample but a significantly larger degradation intensity. It is important to mention that we could not separate the effect of substrate thickness and surface morphology (in our samples) on the observed degradation. The impact of varying surface morphology remains an area for future work.
In addition, the difference in the front surface texture of the samples was also characterized by reflectance measurements (not shown here). The reflectance of the 190 μm sample at 938 nm (the wavelength of the laser used) was 0.7% compared with 4.6% for the 115 μm sample, resulting in a small difference in the electron–hole pair generation rate per cm² for the samples. This partly compensates for the increase in the volume electron–hole pair generation rate because of the difference in the thickness of the samples [37]. Depending on the effective minority carrier lifetime, this may affect the generation and recovery rates because of a difference in Δn; however, this would not explain the significant difference observed in NDD between the samples.

IV. DISCUSSION

In earlier sections, we showed a direct correlation between the reduction in the extent of LeTID with the decrease in silicon wafer thickness, which supports the earlier work presented in [15]. In addition, thinner wafers also showed a strong firing dependence exhibiting a degradation and recovery cycle when they were fired at higher firing temperatures. Furthermore, a minor variation in the surface areas was observed in the thinner samples that would have affected the light trapping, surface passivation, and in- and out-diffusion of hydrogen. Although this article does not directly examine the defect species involved in LeTID, the trends we report provide a valuable insight into understanding the defect mechanism.

A. Interpretation of the Degradation Extents and Rates

The thinner wafers experienced the lowest sample temperatures during firing [see Fig. 4(a)]. However, the trend of increasing degradation extent with wafer thickness could not be completely explained with this difference in the temperature. For example, a negligible variation in the sample temperature for 125, 135, and 155 μm samples when fired at a set peak temperature of 950 °C was observed despite a drastic difference in their maximum NDDs, as shown in Fig. 2. In addition, a possible reason behind thinner wafers experiencing lower sample temperatures could be the relatively faster cooling, as observed by Kulakov et al. [38], where a 85 μm silicon sample had double the cooling rate in comparison with a 200 μm sample. In terms of reaction rates, similar reaction rate constants observed in Fig. 3 render faster recovery in thinner samples unlikely to explain the lower degradation extents in thinner wafers, as shown in Fig. 2.

B. Impact of Effective Surface Area

In addition to the variation in the thickness of the substrates, there appears to be a second factor that may have contributed toward the observation of a lower level of degradation in thinner silicon wafers. As described earlier, the processing of the samples used in this work led to some changes in their surface morphology. The thicker samples tend to have more textured surfaces with a higher surface area, which decreased simultaneously with the thickness of the wafer. The surface area of the texture has been previously reported as an important factor in controlling the LeTID extent [39]. It has also been shown by Sopori et al. [40] that the diffusion and solubility of deuterium in the wafer increases with the surface roughness, and it is reasonable to assume a similar trend for hydrogen. Therefore, during high-temperature firing, hydrogen in-diffusion from the
dielectric stack into the bulk could likely be more effective in the more textured (and thicker) samples. However, the large variation (over an order of magnitude) in the NDD$_{\text{max}}$ could not be solely explained by a relatively minor variation in the surface morphology (~10%).

C. Outlook on the Defect Model

Over the past few years, there has been a growing consensus that the concentration of hydrogen in the silicon bulk is linked to the magnitude of LeTID [18], [19], [21], [36], [41]. One suggested hypothesis was hydrogen being the only defect precursor and its movement within the wafers explains the degradation and recovery [18], [41]. In addition, the involvement of hydrogen in conjunction with another impurity [7], [21], [42] was also suggested. This hydrogen is typically resulting from hydrogenated surface passivation layers into the bulk during the high-temperature firing [8], [43], the amount of which increases with the increase in the firing temperature [44]. While the wafer is ramped up to higher temperatures during firing, some hydrogen diffuses into the wafer. Subsequently, during the rapid cool down [see Fig. 4(b)], hydrogen out diffuses from the wafer [44]–[46]. The amount of bulk hydrogen is, therefore, determined by both these processes involving the in-diffusion and effusion during firing. As similar peak firing temperature and cooling rates were observed for all the samples (see Fig. 4), we can rule out both these factors in explaining the results shown in Fig. 2. In addition, Fig. 7 clearly shows that the observed variation in NDD$_{\text{max}}$ was significantly higher than the difference in the surface area, indicating that it might play a role but cannot fully explain the higher degradation extents in thicker wafers.

Using Fig. 2, we showed that the maximum NDD and the concentration of defect precursors, $N_p$, increases strongly with wafer thickness up to 190 $\mu$m. There is plenty of evidence in the literature pointing toward the possibility of hydrogen plus another impurity (say, an H–Y species) as precursors for LeTID. Under this assumption, thinner wafers likely result in a lower concentration of either H or Y or both. Considering hydrogen, the absolute amount that in diffuses into the silicon bulk during firing would likely be similar for all the wafers, owing to the identical SiN$_x$:H and AlO$_x$:H layers used for all samples [9]–[11], [22]. However, while cooling down, the redistribution of hydrogen would likely be influenced by the proximity of the surfaces. Therefore, it is plausible that during firing, a higher amount of hydrogen per unit volume out diffuses from the wafer, leaving the silicon bulk of thinner wafers comparatively hydrogen-lean. This can potentially explain the lower LeTID extent in thinner wafers. On the other hand, considering the involvement of a second precursor species (Y), a lower degradation extent can also be a consequence of less Y in the bulk of thinner wafers. Possible reasoning for this effect is better surface gettering in thinner wafers during firing, as also suggested in [15], where adjusting the defect concentration to decrease with wafer thickness resulted in a good fit of lifetime data.

An interesting implication of this work is that the influence of wafer thickness was most dominant during firing as opposed to during the light soaking process. As discussed earlier, most of the hydrogen (and/or other LeTID constituents) out diffuses in this step, and this appears to be the key in determining the maximum degradation extent, as shown in Fig. 2. In contrast to this, Fig. 3 reveals that the formation and recovery rates are independent of wafer thicknesses under high-injection conditions. This implies that the diffusion of impurities to the surface during light (or current) soaking is not a rate-limiting step and may not be the mechanism for regeneration of the bulk lifetime.

V. Conclusion

In this study, we tested the stability of mc-Si lifetime test structures based on 115–190 $\mu$m thick substrates using two different stability testing conditions. A significant dependence of the reaction kinetics on the injection level was found. It was observed that the extent of LeTID decreases with the reduction in silicon substrate thickness. However, thinner samples also showed a complete degradation and regeneration cycle when fired at higher peak firing temperatures, demonstrating that thinner wafers are not intrinsically immune to LeTID. The temperature profiles for the samples with different thicknesses were similar (albeit thinner wafers experiencing lower sample temperatures), suggesting the difference in the degradation extent is less likely because of the thermal impact. The reduced degradation in thinner wafers was explained by the plausible increase in the hydrogen effusion during firing. Furthermore, we found that the process of thinning the wafers resulted in a variation in their optical and physical properties. This could be a secondary factor responsible for lower degradation in thinner wafers. However, the separation of the influence of the thickness and surface morphology needs further work.

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