Low-temperature plasma-deposited silicon epitaxial films: Growth and properties

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Low-temperature (≤200 °C) epitaxial growth yields precise thickness, doping, and thermal-budget control, which enables advanced-design semiconductor devices. In this paper, we use plasma-enhanced chemical vapor deposition to grow homo-epitaxial layers and study the different growth modes on crystalline silicon substrates. In particular, we determine the conditions leading to epitaxial growth in light of a model that depends only on the silane concentration in the plasma and the mean free path length of surface adatoms. For such growth, we show that the presence of a persistent defect-free interface layer between the crystalline silicon substrate and the epitaxial layer stems not only from the growth conditions but also from unintentional contamination of the reactor. Based on our findings, we determine the plasma conditions to grow high-quality bulk epitaxial films and propose a two-step growth process to obtain device-grade material. © 2014 AIP Publishing LLC.

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I. INTRODUCTION

Homo-epitaxy enables the growth of high-quality semiconductor films with a well-defined doping-concentration profile and accurate layer-thickness control. For crystalline silicon (c-Si) films, the first successful implementation was found in bipolar transistor fabrication,1 which rapidly extended to advanced microelectronics processing technology, including metal-oxide-semiconductor integrated circuits.2,3 Silicon epitaxial growth has been studied since the early 1950s, using a variety of methods such as vapor-phase, liquid-phase, solid-phase, and molecular beam epitaxy (MBE). Initially, high-temperature processes (>800 °C) were preferred to guarantee epitaxial layers of sufficient quality. Elevated temperatures clean the surface by volatilizing contaminants present prior to the growth and sustain defect-free crystal growth. However, at high temperatures, unintentional dopant and impurity diffusion may also occur, detrimentally affecting the bulk electronic properties of the substrate (e.g., the charge-carrier lifetime) or impairing the properties gained by earlier processing steps. This problem may be circumvented by using plasma-enhanced chemical vapor deposition (PECVD), which dissociates process gasses at significantly lower temperatures compared to thermal CVD.4 For photovoltaic applications, low-temperature (≤200 °C), PECVD is a well-established technique to grow (hydrogenated) amorphous (a-Si:H) and microcrystalline (μc-Si:H) silicon thin films, with important applications in thin-film silicon and high-efficiency silicon heterojunction solar cell fabrication. For these films, the plasma properties dictate the microstructural and electronic quality of the deposited material.5 Even though epitaxial growth by PECVD was first reported in the early 1970s, since then it has been only occasionally further investigated.6–10 Recently, PECVD epitaxial layers regained increased interest, however, as it was recognized that such layers are ideally suited to engineer silicon solar cells, either as (relatively thick) optically active absorber layers that replace the still-costly wafer11,12 or as thin layers for homojunction formation (electron or hole collectors, depending on the doping type).13–16 Such thin epitaxial layers have also found application in certain high-efficiency heterojunction solar cell architectures and heterojunction field-effect transistors, illustrating that the same PECVD tools can be used to deposit various materials to engineer electronic devices.17–19 Despite this, the precise influence of the plasma conditions on the electronic and microstructural quality of the grown layers has been elusive. Characterization of these plasma conditions is important to gain insight into the growth of device-grade epitaxial layers. Moreover, such knowledge is also valuable in the fundamental understanding how to avoid epitaxial growth and how a-Si:H films can be deposited on c-Si wafers with atomically sharp interfaces. Sharp interfaces are critical for high-efficiency crystalline silicon/amorphous silicon heterojunction solar cells.20–22 the current world-record c-Si solar cell technology.23

In this paper, we first discuss the growth regimes arising from different substrate and surface structures. We determine the plasma conditions needed to obtain epitaxial growth on a (100) c-Si wafer and discuss the relevant parameters that determine such growth. Second, we find that an interfacial layer between the c-Si substrate and the epitaxial layer is systematically present and we discuss the origin of this layer. Third, we investigate the microstructural quality of the epitaxial bulk by transmission electron microscopy (TEM) and...
identify the lowest defect density growth conditions. Finally, we study the electronic properties of the epitaxial films and outline the consequences for future device incorporation.

II. EXPERIMENTAL SETUP

Three types of substrates were used: (a) Schott AF 45 glass (b) thin PECVD μc-Si:H layers (~55 nm) deposited on such glass and (c) float-zone 2 Ωcm phosphorus-doped mirror-polished c-Si (100) wafers. Surface cleaning of the wafers involved removal of the native oxide in a diluted hydrofluoric acid solution (HF 5%) for 45 s. The substrates were then loaded directly into a PECVD system for intrinsic silicon film deposition using silane (SiH₄) mixed in hydrogen gas (H₂). A KAI-M PlasmaBox™ reactor (Oerlikon/Tokyo Electron) in parallel-plate configuration powered at very high frequency (VHF, 40.68 MHz) with an inter-electrode distance of 12.5 mm was used for the depositions.²⁴,²⁵ The base pressure of the reactor was typically about 4 × 10⁻⁷ mbar. The three substrates were always co-deposited for given conditions. The deposition pressure (1–7 mbar), power (0.098–0.42 W cm⁻²), and gas flux were varied at a fixed reactor temperature to change the SiH₄ depletion fraction.²⁷ As the used wafers have very high lifetimes and a-Si:H films yield excellent passivation, this method gives a qualitative evaluation of the defect density of the epitaxial films, assuming the field-effect passivation of the undoped epitaxial layers to be similar for each sample.

To characterize the plasma during the depositions, a plasma diagnostic tool built in-house was used to measure in situ the SiH₄ depletion fraction.²⁸ In this setup, a quantum cascade laser probed the rotovibrational absorption lines that are characteristic of the SiH₄ molecule.²⁹ The SiH₄ depletion fraction, D, can then be accessed according to

\[ D = 1 - \frac{A_{\text{SiH₄}}^{\text{on}}}{A_{\text{SiH₄}}^{\text{off}}} \]  

where \( A_{\text{SiH₄}}^{\text{on}} \) and \( A_{\text{SiH₄}}^{\text{off}} \) are the absorbances measured at the relevant wavenumber before and after ignition of the plasma, respectively. The SiH₄ depletion is a direct consequence of the reduction of silane partial pressure due to dissociation by the plasma. Silane dissociation into radicals (Si, SiH, SiH₂, SiH₃) is essential for film growth. Notably, SiH₃ has a low reactivity and therefore long-lived nature, and is often argued to be the dominant film growth precursor for high-quality a-Si:H and μc-Si:H films.³⁰ Next, the silane concentration in the plasma, \( c_p \), is defined as

\[ c_p = c (1 - D), \]  

deleted the overlines in the equation for clarity.

where \( c \) is the input silane concentration before plasma ignition. For the gases used in this study, \( c \) is the silane flow rate divided by the total flux of standard cubic centimeters per minute.³¹ Thus, \( c_p \) is directly linked to the flux ratio of hydrogen atoms and silane radicals arriving at the growing film surface. Low \( c_p \) values imply that a significant fraction of SiH₄ is dissociated into radicals. Conversely, high values indicate low depletion in the plasma.³² Strahm et al. demonstrated that this parameter is fundamentally linked to the microstructural transition from amorphous to microcrystalline silicon.³¹ We will also use \( c_p \) in this study and confirm its relevance for the growth of epitaxial silicon films. In the appendix, another method to access this parameter is discussed.

III. RESULTS AND DISCUSSION

A. Determining parameters for epitaxial growth

In this section, the plasma conditions triggering the a-Si:H, the μc-Si:H, and the epitaxial c-Si growth modes are determined. First, for reference, we plot in Fig. 1(a) the Raman crystallinity fraction of films deposited on glass as a function of \( c_p \). On this type of substrate, μc-Si:H growth occurs for low \( c_p \) (< 1%), whereas for higher \( c_p \) (> 2.7%), the deposited films are fully amorphous. These two growth modes are separated by a transition regime in which both a-Si:H and μc-Si:H growth may happen. This graph clearly demonstrates a direct correlation between the phase of the deposited films and \( c_p \), confirming earlier findings.³¹

\[ \text{SiH₄ conc. in the plasma, } c_p \% \]

![FIG. 1. Raman crystallinity fraction as a function of \( c_p \) on (a) a glass substrate and (b) a μc-Si:H layer (70% Raman crystallinity fraction). The deposition pressure (1–7 mbar), power (0.098–0.42 W cm⁻²), and gas flux were varied at a fixed reactor temperature to change \( c_p \).} \]
Deposition on a substrate precoated with μc-Si:H (70% Raman crystallinity fraction, without native oxide removal) provides a similar $c_p$ dependence as on glass, as seen in Fig. 1(b). As the optical penetration depth of the used Raman laser (∼50 nm in a-Si:H and ∼150 nm in μc-Si:H) is less than the film thickness, the μc-Si:H underlayer should not contribute to the Raman signal. In this figure, we see that the transition regime is not shifted with respect to $c_p$ when changing the substrate surface. However, in the transition regime, we observe a higher Raman crystallinity fraction of these samples compared to a deposition directly on glass. Despite the presence of a surface oxide on the μc-Si:H substrate, this increase likely stems from the promotion of additional nucleation sites due to the local underlying crystalline structure.

Fig. 2 shows film growth on a perfectly crystalline substrate, i.e., a mirror-polished ⟨100⟩ c-Si wafer after native oxide etching in HF. Again we observe three main growth modes, where now low $c_p$ conditions result in epitaxial rather than μc-Si:H growth. The similarity in the $c_p$ dependence confirms that $c_p$ is a fundamental parameter for epitaxial growth as well, underlining how plasma conditions directly dictate the phase of the growing film. We observe that the transition region lies now at higher $c_p$ values (3% ≤ $c_p$ ≤ 9%), compared to the previously discussed cases.

For epitaxial film growth, the transition from epitaxial to amorphous growth may occur as well (stars in Fig. 2), which is usually referred to as epitaxial breakdown.33,34 This phenomenon is not studied here and its dependence on $c_p$ is unknown. The absence in the transition regime of data points with partial crystallinity between 1% and 99% is related to the optical penetration depth of the Raman laser: if epitaxial breakdown occurs at a depth >50 nm measured from the top surface, Raman spectroscopy will probe only the amorphous part of the deposited layer. For $c_p < 3\%$, sustainable epitaxial growth is observed up to the deposited thickness for these films. However, for $c_p < 1\%$, we will show in Sec. III C that these films are very defective, pointing to a fourth growth regime.

Even though the data in Fig. 2 prove that $c_p$ plays an essential role, it cannot be the only parameter dictating the phase of the deposited film in the transition regime. Arguably, the mean free path length of the adatoms on the film growth surface must be equally important. Several factors impact this parameter, such as the surface microstructure, the deposition rate, the deposition temperature, deposition-related ion bombardment, and other deposition-specific parameters.35 In the following paragraphs, we briefly discuss two of these factors, the surface microstructure and deposition rate, as the other parameters cannot be easily accessed.

First, we consider co-deposition of films on ⟨100⟩ and ⟨111⟩ wafers of which cross-sectional TEM micrographs are shown in Figs. 3(a) and 3(b). The growth on the ⟨100⟩ surface is homo-epitaxial, whereas the one on the ⟨111⟩ substrate is purely amorphous. This dependence of the growth mode on the crystal orientation can have significant implications for fabrication of devices such as silicon heterojunction solar cells.36 Such solar cells usually feature...
textured surfaces obtained by alkaline etching of (100) wafers, yielding pyramids with (111) facets. Practically, however, epitaxial needles may be initiated, when the surface orientation locally deviates from the (111) pyramidal facets (Fig. 3(c)), due to locally non-ideal surface texture. This difference in growth mode between (100) and (111) surfaces is in line with earlier findings using PECVD. The origin of this phenomenon has been elusive. Interestingly, this crystalline-orientation-dependent growth mode shows a striking resemblance with observations made for low-temperature epitaxial growth using MBE in ultra-high vacuum conditions. In the latter case, the presence of strings of Si(100)–(2×1) dimers are known to play an essential role for epitaxial growth. An ideal (111) surface contains only a single silicon dangling bond per surface atom; whereas for (100) surfaces, two dangling bonds are present (insets in Figs. 3(a) and 3(b)). Such ideal (100) surfaces can easily reconstruct: between the dangling bonds of adjacent surface atoms, a σ bond—the so-called dimer bond—is formed. Such dimers are unique to the Si (100) surface and strongly depend on the specific surface preparation. On Si(100)–(2×1) reconstructed surfaces, epitaxial growth then occurs as long dimer strings, perpendicular to the present surface strings, at remarkably low temperatures. This phenomenon is unknown for (111) surfaces for which significantly higher temperatures are usually needed and where epitaxial growth follows island formation. For our PECVD experiments, even though the involved surfaces are far from ideal, we speculate that the microscopic differences between (100) and (111) surfaces during deposition critically affect the phase of the deposited material.

Second, Fig. 4(a) shows the Raman crystallinity fraction of different films as a function of their deposition rate. Outside the transition regime, the deposition rate can vary significantly without impacting the growth mode. Conversely, in the transition regime, the growth mode is markedly influenced by the deposition rate, since increasing deposition rates will lower the mean free path length of the adatoms, suppressing the probability of crystalline film growth.

Combining all our findings so far, we draw a simplified view of the amorphous-to-crystalline transition during PECVD using only two main parameters: the SiH₄ concentration in the plasma, ϵ_p, and the mean free path of the adatoms on the film growth surface. This is illustrated in Fig. 4(b). The width of the transition regime in our model stems from additional “hidden” parameters, including ion bombardment and secondary radical gas phase reactions (e.g., Si₂H₆, etc.), which could decrease the number of adatoms without altering ϵ_p.

B. Interface layer of epitaxial films

The surface structure and plasma conditions influence nucleation, as discussed in the previous section. However, impurities in the reactor may also affect growth of the first few nanometers, which we now discuss.

The presence of a thin interface layer between the c-Si substrate and PECVD epitaxial film is indicated in the HRTEM micrographs shown in Fig. 5. Such interface layers have been reported for epitaxial layers grown by low-temperature MBE, hot-wire CVD, and PECVD. We now address the question of why such an interface layer occurs and start by analyzing its composition. SIMS measurements point to an excess of hydrogen and oxygen (Fig. 6) at this interface. Hydrogen is abundantly present in the plasma and may thus indeed easily be incorporated in the deposited films. Conversely, oxygen could originate only from a remaining native oxide film prior to deposition, or from oxygen contamination in the reactor or gas bottles.

Interestingly, in SE measurements, this interface layer yields interference fringes that occur between photon energies from 1 to 3 eV (due to light reflected at this interface and the surface roughness layer). These interferences allow accurate monitoring of the interface and epitaxial layer thicknesses by SE data fitting. To model this interface layer, an effective medium approximation (EMA) of c-Si bulk material and voids is used. If these interferences are absent (because the interface layer is either too thin or simply absent), the thickness of the epitaxial layer can be
determined only by co-deposition on an amorphous test-substrate, assuming similar growth rates for both substrates. Intriguingly, we deduce from Fig. 7(a) that the interface layer thickness does not depend on the amount of time the wafer surfaces are exposed to air after native oxide removal and loading in the reactor. Its thickness rather depends on the number of subsequent depositions in the reactor (Fig. 7(b)). This strongly suggests that the interface layer thickness depends on the presence of impurities such as oxygen in the reactor, which accumulate in the reactor for each loading and unloading of the samples. The decreasing oxygen content throughout the film grown at very low $c_p$ confirms this hypothesis. For these conditions, a high impurity concentration could further be explained by the very low growth rate of these films ($<2 \text{Å s}^{-1}$), which enables more impurities to be incorporated.

To avoid the presence of such undesired oxygen at the interface between the substrate and the epitaxial layer, the reactor was cleaned using a nitrogen trifluoride or a sulfur hexafluoride plasma with subsequent conditioning of the reactor following each deposition. The pumping time was kept constant between the cleaning and the deposition. Under such conditions, for the samples in Fig. 2, we observe a direct dependence of the interface layer thickness on the tested deposition parameters (dilution, power, and pressure, Figs. 7(c)–7(e)). Each of these tested parameters influences $c_p$. In Fig. 7(f), we plot the interface layer thickness as a function of these corresponding $c_p$ values and for the remaining data from Fig. 2, allowing for direct comparison of each deposition parameter. Here, we observe a certain spreading in the data, indicating that the interface layer thickness does not depend solely on the plasma parameters.

### C. Epitaxial bulk

Following the first nanometers of growth, further crystal propagation depends mainly on the conditions in the plasma. To determine which conditions yield high-quality epitaxial films, we discuss now the microstructural quality of the epitaxial bulk films.

For this purpose, we took SE data of the samples discussed in Fig. 2. These data were then fitted using an EMA featuring $c$-Si and $a$-Si:H fractions from which a qualitative estimate of the amorphous fraction of the deposited films can be made, as plotted in Fig. 8. This SE amorphous signature observed in several epitaxial films probably stems from bonds with non-ideal angles and lengths compared to $c$-Si bonds.

We note that Raman spectroscopy did not offer sufficient sensitivity to probe such differences (Fig. 2). We observe here the presence of a possible fourth regime for very low $c_p$ values in which the layers show a high amorphous fraction, which we discuss below.

To reveal the microscopic structure and local details of the films and to relate them to the SE data, we also performed cross-sectional analyses using HRTEM imaging. Our epitaxial layers typically contain dislocations or dislocation loops on {1 1–1} planes, as for instance shown in Fig. 9. This micrograph was taken from the middle (bulk) of a
90-nm-thick layer deposited at $c_p = 1.6\%$. We see a clear correlation between the defect densities extracted by HRTEM and bright-field (BF) TEM and the amorphous fraction in deposited films extracted by SE (data not shown).

STEM BF micrographs of layers deposited at very low $c_p$ values reveal that such films are indeed composed of both amorphous and crystalline regions. EDX analysis of such films, shown in Fig. 10(b), reveals that oxygen is present within the film in the form of irregular hazy layers. Fourier filtering obtained at different positions of a HRTEM micrograph (Fig. 10(c)) demonstrates that the film features epitaxial as well as amorphous regions. Notably, the regions rich in oxygen correspond to the amorphous zones. Likely, the (unintentional) incorporation of oxygen provokes a disruption of crystalline growth. In addition, we observe a rotation of a few degrees in the crystalline orientation of the growing layer in this sample, possibly due to an accumulation of dislocations or cracks. On a glass substrate and in such low $c_p$ regimes, the layer peels off as soon as it is exposed to air, indicative of high stress buildup in the deposited material. Decreasing $c_p$ even further provokes wafer etching.55

To further investigate the bulk quality of the epitaxial layers, we turn back to the SIMS results shown above (Fig. 6). Expectedly, the amorphous film has the highest hydrogen content, while the epitaxial films grown at very low $c_p$ ($c_p = 0.7\%$) and in the transition regime ($c_p = 5.8\%$) show a hydrogen density of about $2 \times 10^{21} \text{ cm}^{-3}$. Finally, the purely crystalline ($a$-Si:H content of 0%) epitaxial layer, deposited at $c_p = 2.1\%$, has the lowest bulk hydrogen content. The epitaxial film grown in the transition regime ($c_p = 5.8\%$) has a higher hydrogen content than the one grown in the nearby epitaxial regime ($c_p = 2.1\%$). This $c_p$ dependence of the hydrogen incorporation is similar to that observed for $\mu$-c-Si:H film growth for which the transition regime contains the highest hydrogen content.56,57 We remark that the hydrogen content throughout the film may either decrease (films with $c_p = 4\%$ and 2.1%) or increase (films with $c_p = 0.7\%$) with thickness.

Quite generally, in epitaxial as well as in microcrystalline films, compressive stress—linked to the hydrogen content of the films—may cause defects or a distortion of the crystalline lattice.53 We now briefly outline how the observed differences in hydrogen content in our films may be traced back to their precise deposition details. According to the different film growth models available in the literature,58 hydrogen can (1) enhance the diffusion process of SiH$_3$ by full surface coverage and local heating (surface diffusion model), (2) break weak Si-Si bonds (etching model), and (3) penetrate the subsurface region and cause crystallization (chemical annealing model). Therefore, we have two fundamentally opposing effects of hydrogen: hydrogen promotes epitaxial growth, and conversely when incorporated in

FIG. 8. Ellipsometry data fitted using the model shown in the inset. The stars indicate epitaxial breakdown after 2–5 nm of growth, observed by TEM. The model uses an EMA of $c$-Si and void to simulate the interface layer between the substrate and the film (see Sec. III B). The second and third layers from the substrate represent the epitaxial and the roughness layer ($c$-Si and void), respectively.

FIG. 9. HRTEM micrograph of the bulk of an epitaxial layer with a dislocation/dislocation loop on a $\{11\overline{1}\}$ plane in the region marked by the yellow box. The top right inset shows the diffractogram of the region marked in the yellow box; Fourier filtering of the region marked in yellow using the $\{11\overline{1}\}$ planes indicated by yellow circles in the diffractogram highlights the plane termination at the end of the dislocation (bottom right inset).

FIG. 10. STEM micrographs of a low $c_p$ sample (0.7%). (a) STEM BF micrograph and (b) STEM EDX map of the BF micrograph. (c) HRTEM of the first 100 nm of a film and diffractograms of the four regions marked in white boxes. Starting from the $c$-Si substrate (region 1), epitaxial growth (regions 2 and 4) is observed to alternate with $a$-Si:H growth (region 3).
the film, it can then distort the crystalline lattice, ultimately yielding epitaxial breakdown. These two effects compete with each other and explain why sustainable epitaxial growth only occurs for intermediate \( c_p \) values.

Summarizing our findings so far, from previous studies linking plasma conditions to deposited material quality, device-grade \( \mu c\)-Si:H on glass should be deposited close to the amorphous-to-crystalline silicon transition, i.e., in the transition regime of Fig. 1(a).\(^{59,60} \) The validity of this argument was extended for thin-film \( a\)-Si:H absorber layers, as well as for layers designed for \( c\)-Si surface passivation.\(^{61} \) For epitaxial films, it also appears that bulk quality deteriorates for films deposited far away from the transition region, namely under very low \( c_p \) conditions. However, in the transition regime, the epitaxial film contains more hydrogen than in the nearby epitaxial regime, and hydrogen strains the crystalline lattice and creates defects. Hence, for high-quality bulk epitaxial films, the ideal conditions appear to be close to but not necessarily within the transition regime. Conversely, in the case of \( \mu c\)-Si:H layers, the higher hydrogen content in the transition regime has rather a beneficial effect and likely passivates the grain boundaries, leading to higher open-circuit voltages in solar cells.\(^{56,62,63} \) Moreover, the stress induced by this excess of hydrogen in the \( \mu c\)-Si:H layers can be accommodated by the flexible amorphous network around the crystalline grains, which is not the case for an epitaxial film. These two reasons could explain why, microscopically, high-quality epitaxial films are obtained at lower \( c_p \) values than are necessary for high-quality \( \mu c\)-Si:H layers.

Figure 11 illustrates the three main regimes outlined in this paper: (1) defective epitaxial layers are grown with very low \( c_p \) values, (2) the least defective layers are grown close to the transition regime, and (3) the thinnest interface layers are obtained close to the amorphous regime.

D. Electronic properties of epitaxial layers and consequences for solar cells

Finally, to verify the electronic quality of the deposited films, photoconductance measurements were performed on symmetrical samples composed of an epitaxial film and a passivating intrinsic \( a\)-Si:H layer (see inset Fig. 12). As stated earlier, we used exclusively high-quality wafers and we know that the amorphous passivating layers are excellent, yielding a carrier lifetime of over 7 ms when no epitaxial layer is present (data not shown). Therefore, we expect the minority-carrier lifetime to be limited by carrier recombination occurring at the interface layers between the \( c\)-Si substrate and epitaxial film or within the epitaxial films themselves. Photoconductance measurements are then ideal probes to assess the electronic quality of these epitaxial films. Figure 12 shows how the carrier lifetime of a wafer featuring such epitaxial/\( a\)-Si:H(i) stack varies as function of the thickness of the interface between the wafer and the epitaxial layer. The data presented here were obtained by growing epitaxial films using increasing hydrogen dilution in the plasma. For this series, TEM micrographs show that the number of defects in and the thickness of the interface layer between the wafer and the epitaxial film do not dictate the bulk epitaxial quality. Indeed, starting from a thick interface layer, the bulk epitaxial layer can exhibit fewer defects compared to the case with a thin interface layer (micrographs not shown here). From this, we conclude that the defects at the interface and within the interface layer limit the effective carrier lifetime in these samples.

As a consequence, if the epitaxial layer is used to form the electron or hole collector of a device (either as the emitter or back-surface field), the defective interface layer should be as thin as possible to avoid increased recombination losses. To accomplish this, the first nanometers should be deposited in a regime as close as possible to the amorphous-to-crystalline transition. Subsequently, to sustain a device-grade epitaxial layer, the deposition conditions should be changed to lower \( c_p \) values, as discussed in Sec. III C.\(^{64} \) Conversely, if the epitaxial layer is grown as an optically active absorber (i.e., as the replacement for a \( c\)-Si wafer), the interface layer can be maximized to act as a weak layer that can later be exploited to separate the thin film from the wafer as suggested in Ref. 12.

IV. CONCLUSIONS

Low-temperature epitaxial growth by PECVD could have a wide range of applications in advanced-design semiconductor devices if the determinant parameters for such growth are understood and mastered. In this paper, we show that the silane concentration in the plasma, \( c_p \), is a crucial parameter for determining the growth mode of epitaxial silicon thin films on \( c\)-Si substrates. However, the mean free path of adatoms, which is influenced by substrate type and growth rate effects, also plays an important role. The bulk of the
epitaxial film is found to show fewer dislocations and lower hydrogen content for films close to the amorphous-to-crystalline transition regime. Between the epitaxial layer and the c-Si substrate, an interface layer limits the minority-carrier lifetime. The thickness of this interface layer—a porous layer with an excess of hydrogen and oxygen—depends not only on the conditions in the plasma but also on impurities in the reactor. The layer thickness is effectively reduced by using \( c_p \) in the transition regime and by cleaning the reactor between each deposition. Therefore, to grow a high-quality epitaxial layer with the thinnest possible interface layer, we suggest a two-step process composed of a high \( c_p \) step to grow the first nanometers and then a lower \( c_p \) step to have stable, high-quality epitaxial growth.

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APPENDIX: SIMPLE ESTIMATION OF THE SILANE CONCENTRATION IN THE PLASMA

We have shown in this paper that \( c_p \) is a relevant parameter to determine the precise film growth mode. In this appendix, we briefly discuss a simple way to experimentally access this parameter.

First, we estimate the maximum possible deposition rate, \( R_{\text{max}} \). For this, we assume that all present Si atoms injected in the reactor volume contribute to the film growth, which yields \(^{31}\)

\[
R_{\text{max}} = \frac{\Phi_{\text{SiH}_4} V m_{\text{Si}}}{A \rho}, \quad \text{(A1)}
\]

where \( \Phi_{\text{SiH}_4} \) is the molecular flux in \( \text{particles m}^{-3}\text{s}^{-1} \), \( V \) the reactor volume, \( A \) the area of the reactor exposed to the plasma and \( \rho \) the silicon density. Using the ideal gas equation, the molecular flux can be converted into the flow rate in standard cubic centimeters per minute (sccm)

\[
R_{\text{max}} = 20.97 \frac{F_{\text{SiH}_4}}{A \rho}, \quad \text{(A2)}
\]

with \( F_{\text{SiH}_4} \) being the silane flow rate. If we consider deposition on glass of \( a\)-Si:H or \( \mu c\)-Si:H, the hydrogen content of the film can be assumed to be around 10% (2180 kg/m\(^3\)). Second, the silane dissociation efficiency, defined as the ratio between the measured deposition rate \( R \) and ideal \( R_{\text{max}} \), is linked to \( c_p \) in a simple plasma model as follows: \(^{65}\)

\[
\frac{R}{R_{\text{max}}} = \frac{c - c_p}{c(1 + c_p)}. \quad \text{(A3)}
\]

With this simple model, measuring the deposition rate gives an estimate of the \( c_p \) value in the plasma. This method is valid only if no powder, dust, or polysilane molecules (\( \text{Si}_2\text{H}_6, \text{Si}_3\text{H}_8, \ldots \)) are formed in the plasma. These species are formed by secondary reactions. To verify this simple model experimentally, in Fig. 13, we show for various plasma conditions the linear dependence between \( c_p \) values obtained by optical measurements using the quantum cascade laser and those obtained by the growth rate method. We see a slight but systematic overvaluation of \( c_p \) by the growth rate method, which may come from the limits of the used plasma chemistry model and the non-uniform deposition in the reactor.

The three data points which are clearly off the trend line stem from dusty plasma regimes and were not taken into account for the fitting. The formation of powder occurs in regimes at high pressure, high RF power, or high silane concentration.

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