Absolute in situ measurement of surface dangling bonds during a-Si:H growth

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Hydrogenated amorphous silicon (a-Si:H) is regarded as a model system in mechanistic studies of amorphous semiconductor film growth, where the key reactive site is generally considered to be a surface “dangling bond.” Employing an ultrahigh-\(Q\) (~10\(^{10}\)) monolithic optical resonator, the authors probe the creation of dangling-bond defects during growth of a-Si:H from a predominantly SiH\(_x\) (\(x=0–3\)) radical flux by detecting the associated near-IR subgap absorption with evanescent wave cavity ringdown spectroscopy. They find the apparent dangling-bond creation rate \((5\pm3) \times 10^{12} \text{ cm}^{-2} \text{ s}^{-1}\) and steady-state surface density \((5\pm2) \times 10^{11} \text{ cm}^{-2}\) to be considerably lower than expected from dangling-bond-based growth mechanisms.

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Widely used in electronic and optoelectronic devices, hydrogenated amorphous silicon (a-Si:H) has unique properties derived from structural disorder and may enable the development of large-area low-cost flexible solar cells.\(^1\) In mechanistic studies of amorphous semiconductor film growth, a-Si:H also serves as a model system, where surface defects are generally thought to play a key role by providing reactive sites for binding of vapor-phase radical precursors.\(^2\)\(^-\)\(^4\) Based on epitaxial growth studies of c-Si (Ref. 5) and vibrational spectroscopy of a-Si:H,\(^6\) the primary defect site has long been considered to be an undercoordinated Si atom possessing a singly occupied sp\(^3\)-hybridized orbital or “dangling bond” (DB), which binds radical precursors through a facile radical-radical coupling.\(^2\)\(^-\)\(^4\)\(^7\) However, the bulk DB density in device-grade a-Si:H is typically \(\approx 10^{16}/\text{cm}^2\), while a surface defect density as low as \(\approx 10^{11}/\text{cm}^2\) is conceivably compatible with DB-based growth mechanisms. Therefore, highly sensitive in situ techniques are required to probe DB defects and yield decisive mechanistic insight.

Photothermal deflection,\(^8\) constant photocurrent,\(^9\) electron spin resonance (ESR),\(^10\) and cavity ringdown\(^11\) spectroscopic techniques have been applied to DB detection. Although constant photocurrent and photothermal deflection methods provide the requisite sensitivity, these techniques are not easily adapted to in situ studies under growth conditions. ESR has been applied during growth and H-atom exposure of a-Si:H films,\(^10\)\(^-\)\(^11\) although surface, bulk, and interface defects are not readily distinguished: By probing the subgap DB defect absorption,\(^8\)\(^,\)\(^9\)\(^,\)\(^12\) cavity ringdown spectroscopy (CRDS) enables determination of very low absolute defect densities.\(^11\) However, the conventional implementation of CRDS using a linear resonator is not readily adapted to in situ studies of a-Si:H growth.

Alternatively, by employing an ultrahigh-\(Q\) miniature monolithic optical resonator, evanescent wave CRDS (Refs. 13 and 14) enables in situ studies with high sensitivity, as the concomitant evanescent wave at a total-internal reflection (TIR) surface serves as an ideally suited thin-film probe. The application of high-\(Q\) monolithic resonators to probe surfaces, films, and nanoscale materials shows considerable promise. For example, we demonstrated detection of a three-quantum vibration-combination transition on atomically smooth SiO\(_2\), revealing a unique quasi-ice monolayer at ambient temperature.\(^14\) Moreover, Borselli et al.\(^15\) used a monolithic microresonator to detect surface defect states on crystalline Si, revealing the role of surface passivation in the performance of microphotonic components. Here, we employ a monolithic folded resonator\(^13,\)\(^14\) to probe the DB defect states during growth of a-Si:H from a predominantly SiH\(_x\) (\(x=0–3\)) radical flux on atomically smooth SiO\(_2\). Assuming an absorption cross section of \(10^{-16} \text{ cm}^2/\text{defect}\), a minimum detectable surface DB defect density of \(\approx 10^8/\text{cm}^2\) is predicted, enabling mechanistic details to be elucidated.

Fabricated from ultralow-OH amorphous silica (Suprasil 300; Heraeus Tenevo), the folded resonator (Fig. 1, inset) has two orthogonal planar highly-reflective mirror surface \((R=99.99\% \text{ at } 1200 \text{ nm})\) separated by a 2 cm unfolded path and a convex TIR surface. Having 0.05 nm rms roughness, the TIR surface forms the working surface for deposition, where an approximately 0.04 mm\(^2\) region is sampled at the apex by the resonator modes. The idler beam of a seeded-tripled-neodymium-doped-yttrium-aluminum-garnet-pumped optical parametric oscillator (30 Hz, \(\approx 0.5 \text{ mJ/pulse, } 6 \text{ ns,}\) linewidth < 10 cm\(^{-1}\)) was employed to generate orthogonally polarized ringdown transients, which were simultaneously detected by free-space-coupled, high-speed, 125 MHz, InGaAs detectors. The resonator was mounted in a high-vacuum chamber (\(<10^{-9} \text{ mbar}\)) and maintained at 150 \(^\circ\)C.
by radiative heating. The a-Si:H film was deposited by thermal decomposition of SiH₄ at 4 × 10⁻⁵ mbar on a hot tungsten filament (~2050 °C) located at a distance of 6.8 cm from the TIR surface, producing primarily SiH₃ as the growth precursor. Separate real-time spectroscopic ellipsometry measurements under identical conditions determined the growth rate to be 3.5 nm/min. The growth process was studied by performing many growth cycles (flux on/growth/flux off), resulting in a final thickness of 800 nm.

Figure 1(a) shows the absolute intrinsic loss per pass of the resonator prior to deposition for s and p polarizations between 1170 and 1245 nm using a 0.05 nm step size. Each data point is an average of 75 measurements, which yielded a typical 1σ noise level of 2 × 10⁻⁶ optical loss. The loss per pass reaches a minimum of 18 × 10⁻⁶ at 1205 nm, which yields a base ringdown time of 5.3 μs and a quality factor of Q = 8.3 × 10⁸. At 150 °C in vacuum, the spectral dependence of the intrinsic loss arises mostly from mirror reflectance, although very weak features can be seen (arrows) at 1229, 1232, and 1236 nm, which we tentatively attribute to residual (100)- and (111)-cristobalite-like surface hydroxyl species. In Fig. 1(b), the solid and dashed plots show the intrinsic-loss-subtracted, polarized absorption spectra for 25 and 60 nm a-Si:H films, respectively, obtained after termination of growth. As the DB defect state produces a very large discontinuity in the normal component of the bulk fields, including excitation of quasiwaveguide modes having broad radiative linewidths associated with coupling into the resonator half-plane. Specifically, the polarization anisotropy observed in Fig. 1(b) arises from the influence of the quasi-TE₀ mode, which leads to a larger average electric field intensity inside the film for s compared to p polarization for both h values considered. Similarly, a larger optical loss is observed for the 25 nm compared to the 60 nm film as optimal coupling into the broad quasi-TE₀ mode occurs at a thickness of <20 nm. Furthermore, Eq. (1) accounts for the surface contribution to the optical loss through N_{DB}(h) and I_{s,p}(h,λ). We use the s-polarized field to evaluate the DB defect contribution to the growth mechanism, since for the p-polarized field Fresnel analysis implies a large discontinuity in the normal component of the surface electric field. Using the s-polarized field provides therefore a quantitative description that is insensitive to the precise relationship between the chemical versus optical surface.

Employing the single-laser-shot minimum detectable absorption of 10⁻⁷, we probed the film growth kinetics with a repetition-rate-limited time resolution of 33 ms. Figure 2 shows the s-polarized optical response of a 60 nm a-Si:H film to a 20 s radical flux exposure, where the base loss of the resonator has been subtracted. Figure 3 shows the optical loss for both s and p polarizations up to a film thickness of 800 nm, as obtained by combining sequential growth cycles similar to the growth cycle shown in Fig. 2. The large variation in the optical loss as a function of thickness in Fig. 3 derives from excitation of aforementioned quasiwaveguide modes, while the gradual increase in optical loss with thickness is caused by absorption by DBs in the bulk a-Si:H. The total optical loss of the film as a function of wavelength

\[ \Gamma(\lambda) = \int_0^h I_{s,p}(z,\lambda) N_{DB}(z) \sigma_{DB} dz, \]
noise level in the data is enhanced at high optical loss values due to the reduced $Q$ factor of the resonator+film configuration. Figure 2 shows that the film response to the radical flux attains steady state within several seconds of exposure, followed by comparatively slow recovery after flux termination. Using the measured DB absorption cross section, a radical flux-induced surface DB density can be obtained as shown on the right ordinate, where a local field correction has been applied using the electric field intensity just below the film surface [i.e., $I_0 \langle \delta E(L) \rangle$]. Because the s-polarized electric field is continuous across the surface region, DB defects on the surface cannot be distinguished from those just inside the surface, although clearly defects on the surface are relevant to growth. However, the measured s-polarized loss is at least the sum of the true surface+subsurface contributions, which provides an upper bound for the steady-state surface DB density and the DB creation rate during growth.

Figure 3 shows the radical flux-induced surface DB density, as obtained from data in Figs. 2 and 3 at different film thicknesses. A thickness-independent steady-state surface DB density of $(5 \pm 2) \times 10^{11}$ DB/cm$^2$ is found, which corresponds to a fractional surface coverage of $\sim 5 \times 10^{-4}$. Such a low coverage precludes a mechanism based on the direct precursor-defect binding, as a coverage of $\sim 0.1$ is required to account for typical a-Si:H growth rates, assuming a sticking probability for radical-radical coupling of 0.1. However, a low surface coverage does not eliminate mechanisms involving surface diffusion of weakly bound precursors or a direct insertion mechanism. Yet the role of the DB is significantly different for these latter two mechanisms. For precursor-mediated growth, the creation rate of DB defects is critical and must at least equal the growth rate (DB passivation rate), while for direct insertion the DB defect is not specifically relevant. As also shown in Fig. 4, the initial DB creation rate, which was obtained from many growth cycles at various film thicknesses, is found to be $(5 \pm 3) \times 10^{12}$ DB/cm$^2 \text{s}$. Comparing this value with the actual growth rate of $2.9 \times 10^{14}$ Si/cm$^2 \text{s}$, assuming a Si density of $5 \times 10^{22}$ cm$^{-3}$ in a-Si:H, the DB creation rate is $60 \times$ too small to account for the actual growth rate. Therefore, we conclude that the DB defect is not crucial in growth mechanism of a Si:H from radical precursors.

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