The role of the silyl radical in plasma deposition of microcrystalline silicon

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(Received 10 May 2004; accepted 18 July 2004)

Expanding thermal plasma chemical-vapor deposition has been used to deposit microcrystalline silicon films. We studied the behavior of the refractive index, crystalline fraction, and growth rate as a function of the silane (SiH₄) flow close to the transition from amorphous to microcrystalline silicon. It was found that the refractive index, a measure for film density, increases when the average sticking probability of the depositing radicals decreases. Furthermore, we studied the influence of the position at which SiH₄ is injected in the expanding plasma on the film density. It was found that the film density becomes higher when the SiH₄ is injected closer to the substrate. Both findings strongly suggest that the film density benefits from a high contribution of the SiH₃ radical to the growth of microcrystalline silicon. © 2004 American Institute of Physics.

[DOI: 10.1063/1.1790577]

I. INTRODUCTION

Thin silicon films are widely applied in large-area electronic devices like flat displays and solar cells. Hydrogenated amorphous silicon (a-Si:H) has already been studied and applied since 1965. Quite some knowledge has been gained about the deposition mechanisms and the optimum plasma chemistry. It is now widely agreed that the SiH₃ radical plays an important role in the deposition process of high-quality a-Si:H. A higher surface mobility is often suggested as one of the reasons. Hydrogenated microcrystalline silicon (μc-Si:H) has become popular for application in large-area electronics about ten years ago. There is still a lot of debate about the growth mechanism, and several growth mechanisms have been proposed, many of them inspired on the successful growth model of a-Si:H. It is not yet clear what the optimum deposition plasma chemistry should look like and which radicals contribute favorably to the growth of high-quality μc-Si:H.

The industry standard for the production of thin silicon films is radio-frequency plasma-enhanced chemical-vapor deposition (RF PECVD), but other techniques have been developed, mainly with the aim to increase the growth rate. Very-high frequency (VHF) PECVD, hot wire (HW) CVD, and electron cyclotron resonance (ECR) PECVD are some of the alternative techniques investigated. In this paper, μc-Si:H films have been prepared using expanding thermal plasma (ETP) CVD. ETP CVD is a remote plasma deposition technique, in which the plasma generation takes place in a relatively high pressure chamber (the cascaded arc plasma source), separate from the plasma chemistry that takes place when the precursor is injected in the low-pressure process chamber. This simplifies the chemistry to such an extent that ETP CVD is a suitable plasma deposition technique for the investigation of the influence of the plasma chemistry on the film deposition. This has been demonstrated, in particular, for the ETP CVD of a-Si:H, and a-C:H and a-SiOₓCₓ:H. The deposition of thin silicon films with RF/VHF PECVD, HW CVD, and ETP CVD all rely on the decomposition of the precursor gas silane (SiH₄) into radicals and ions that eventually stick to the substrate surface, where film growth occurs. The decomposition mechanism of SiH₄ is totally different for the three deposition techniques mentioned. In RF/VHF PECVD, electrons gain kinetic energy driven by the alternating electric field that is applied between two parallel electrodes, and the SiH₄ molecules are dissociated by the electron impact. In HW CVD, the SiH₄ molecules are dissociated by a catalytic reaction at hot tungsten or tantalum wires. In ETP CVD, a beam of atomic hydrogen is created, which interacts with the SiH₄ molecules injected downstream. In this way, hydrogen abstraction reactions dominate the dissociation of SiH₄, and it depends on the surplus of hydrogen and the path length to the substrate to what extent subsequent abstraction reactions are capable of creating the various silane radicals. These abstraction reactions are:

\[
\text{SiH}_4 + H \rightarrow \text{SiH}_3 + \text{H}_2,
\]

\[
\text{SiH}_3 + H \rightarrow \text{SiH}_2 + \text{H}_2,
\]

\[
\text{SiH}_2 + H \rightarrow \text{Si} + \text{H}_2.
\]
SiH + H → Si + H₂.  

(4)

The reaction rates of these reactions can be found in Refs. 22 and 23. Note that we neglect radical-radical as well as radical-silane reactions because we only consider here the high dilution conditions to deposit μc-Si:H. Furthermore, radical-H₂ reactions are neglected because the reaction rate is very small. Since the H₂-dilution is high, reactions involving ions can be neglected too (<10¹⁶ m⁻³). ²⁴

It has been demonstrated that for the deposition of high-quality a-Si:H, it is required that the flux of SiH₃ radicals toward the substrate dominates the deposition. In the deposition of high-quality a-Si:H with ETP CVD, this is realized as follows. The SiH₄ flow is set at a higher value than the flow of atomic hydrogen from the plasma source. The atomic hydrogen created in the cascaded arc plasma source is totally consumed by the first hydrogen abstraction reaction [Eq. (1)] with the relatively high density of SiH₃. ²⁵ For these conditions, subsequent reactions of SiH₃ with atomic hydrogen are suppressed and a dominant SiH₃ beam is generated. The latter is confirmed by means of cavity ring-down spectroscopy (CRDS) and threshold ionization-mass-spectrometry measurements. ²⁵

It is generally accepted that for the deposition of μc-Si:H, atomic hydrogen is essential at the surface of the growing film to enhance the crystallization. ²⁵, ²⁶, ²⁷ Therefore, commonly, the gas flows are adjusted to a high (1%-5%) dilution of the SiH₄ in H₂ irrespective of the deposition technology. In RF PECVD, the high-energy electrons also dissociate H₂. In HW CVD, H₂ is catalytically dissociated at the tungsten or tantalum wire. In ETP CVD, increasing the H₂ flow in the cascaded arc plasma source results in a higher atomic hydrogen flow into the reaction chamber, and a simultaneous decrease in SiH₄ flow leads to a lower consumption of H by SiH₃. Both effects result in a higher H flux at the surface of the growing film. As a consequence of the high atomic hydrogen density in the reaction chamber, the depletion of SiH₄ is very high and can even reach 100%. ¹⁸ Furthermore, subsequent hydrogen abstraction reactions will take place, creating SiH₃ with x < 3. This is clearly illustrated by the CRDS measurements by Hamers et al., in which the SiH₄ flow is varied for a constant Ar and H₂ flow. ²⁸

In general, μc-Si:H, especially when deposited at higher deposition rates and at lower temperatures, is porous to some extent, decreasing the applicability in solar cells. This characteristic we have discussed in another article. ²⁹ In principle, apart from the crystallite size, there will be no difference in the quality of the crystalline silicon phase in microcrystalline silicon. It is the amorphous phase between the crystallites that contains the pores and has the varying defect density. This suggests that the quality of μc-Si:H can be improved when the SiH₃ radicals dominate the deposition of microcrystalline silicon because the amorphous phase between the crystalline will be of better quality, similar to the case of entirely hydrogenated amorphous silicon films. Figure 1, in which we plotted the μc-Si:H refractive index at a photon energy of 2 eV versus the SiH₄ depletion for various arc and reactor chamber settings, corroborates more or less this hypothesis. If the SiH₄ depletion is close to 100%, the reaction chain from Eqs. (1) and (4) goes to a full completion leading to a distribution of high sticking radicals and this is accompanied by a low refractive index.

In this paper, we will discuss two additional observations to Fig. 1, which were obtained during the systematic variation of the process conditions to achieve better quality μc-Si:H (see Ref. 29). This discussion might give a leading principle in obtaining an even better quality μc-Si:H in the near future. The first observation is related to the behavior of the refractive index, crystalline fraction, and growth rate as a function of the SiH₄ flow, in particular, close to the a-Si:H/μc-Si:H transition. The other observation discussed is related to the effect of the varying SiH₄ injection ring position on the quality of the μc-Si:H.

We first describe the ETP CVD setup and the film diagnostics employed. Subsequently, the results are presented and discussed. As we will argue, the implications of the observations strongly suggest the beneficial role of the SiH₃ radical in the deposition of μc-Si:H. Finally, the conclusions are presented.

II. EXPERIMENTAL SETUP AND PROCEDURES

The cascaded arc solar cell apparatus Delft Eindhoven (CASCADE) setup is designed to prepare thin-film silicon solar cells, in which the absorber layer is deposited using ETP CVD. ²⁰ It consists of a load lock, a RF PECVD chamber for the deposition of the doped layers, and a reaction chamber for the deposition of silicon films with the ETP CVD. In Fig. 2, a schematic representation is shown of ETP part of the CASCADE setup. This deposition technique has been extensively described elsewhere; ¹⁵ in the following discussions, we will give a brief summary and give the relevant process parameters, which are varied.

On top of the ETP reaction chamber is the cascaded arc plasma source. This source generates a plasma between three cathodes and an anode plate at a dc discharge current of 50 A at a pressure of about 0.1 bar. Two different geometries for the cascade arc plasma source have been employed. The
The substrate is positioned at 410 mm below the plasma source, which can be fed with different gases. In this case, H₂ or an Ar/H₂ mixture is used (100–300 mbar). The plasma expands into the reaction chamber (~0.2 mbar), and downstream, the precursor gas SiH₄ is injected. The silicon film is deposited on the substrate, which is on a temperature-controlled holder. A shutter is used in order to start the deposition at stable plasma conditions. On the side of the reaction chamber, below the substrate, the pump line is situated. A valve on the other side of the chamber opens the way to the load lock and the RF PECVD chamber.

The advantage of this source is that it can be operated at relatively low gas flows (typically 600 sccm (denotes cubic centimeter per minute at stp)) Ar and 200 sccm H₂ and, therefore, a low pump capacity is sufficient to obtain the required process pressure of about 0.2 mbar. This source, referred to as the 2.5-mm arc, cannot be operated on hydrogen gas only, but a minimum argon flow of twice the hydrogen flow must be added. In this research, this source is always operated at an argon-hydrogen flow ratio of 2:1, because for the deposition of microcrystalline silicon, the hydrogen dilution is important. In order to be able to operate a pure hydrogen plasma, a source with a different geometry is installed. The discharge channel is now created by four copper cascade plates with holes having a diameter of 4 mm, forming a central channel of about 20-mm long. This source, referred to as the 4-mm arc, is operated at 2000 sccm H₂. The substrate is positioned at 410 mm below the plasma source exit. The substrate temperature is 300°C throughout this work, unless stated otherwise.

After leaving the nozzle of the cascaded arc, the plasma expands supersonically into the reaction chamber, which is at a pressure of about 0.25 mbar. The expanding plasma shocks when it collides with the background gas, after which it flows toward the substrate with a velocity that gradually decreases to about 100–200 m/s right above the substrate. The precursor gas, SiH₄, is injected into the expanding plasma through a ring-shaped (80-mm diameter) gas line, which is concentric with the reaction chamber, with holes (1 mm diameter) pointing to the axis of the reaction chamber. The atomic hydrogen in the plasma dissociates the SiH₄ molecules into radicals, which are deposited at the substrate at 410 mm below the plasma source exit. The consumption of the SiH₄ by the hydrogen plasma is measured with a mass spectrometer, which is mounted on the side of the reactor at the substrate level.

Using the 2.5-mm arc fed with 1200 sccm Ar and 600 sccm H₂, a series of samples has been prepared with a varying SiH₄ flow in order to investigate the behavior of the refractive index, crystalline fraction, and growth rate close to the α-Si:H/μc-Si:H transition.

A second series of samples has been prepared using the 4-mm arc fed with 2000 sccm of H₂ to investigate the influence of the position of the injection ring. Three injection ring positions have been employed, 365, 110, 55 mm above the substrate (for brevity referred to as high, middle, and low injection ring position), and for any of these positions, the SiH₄ flow is varied from 5 to 25 sccm. The same deposition series is repeated using a hydrogen gas flow of 1500 sccm and the injection ring at the high and the low position.

The layers are simultaneously deposited on Coming 1737 glass substrates and n-type crystalline Si wafers. Care was taken to have the thickness of all the films in the range from 650 to 750 nm. The films are analyzed using reflection-transmission spectroscopy to determine the thickness and the refractive index at a photon energy of 2 eV. Raman spectroscopy was used to determine the crystalline fraction in the microcrystalline film, as described previously.

III. RESULTS AND DISCUSSION

A. SiH₄ variation

In most PECVD technique, there is a threshold dilution $R = \frac{R}{(H₂+(H₂+SiH₄))}$ for the deposition of μc-Si:H (e.g., Ref. 33). In Fig. 3, the results of the film series deposited using the 2.5-mm arc are shown. From the crystalline fraction in Fig. 3(a), the transition region between the amorphous and microcrystalline material can be determined at a SiH₄ flow somewhere in between 5 and 10 sccm, as is indicated by the dashed vertical line. In Fig. 3(b), the depletion is shown as a function of the SiH₄ flow. For the small SiH₄ flows (<3 sccm), the depletion is almost 100% because there is a surplus of H available in the plasma to dissociate all the injected SiH₄, (H₂ dissociation degrees of 0.3% (Ref. 34) and 0.4% (Ref. 35) have been reported for an Ar/H₂ ratio of 6/1 and for a pure H₂ plasma, respectively. For H₂ flow of 600 sccm, this corresponds to H flow of about 1.8 and 24 sccm, respectively. The H flow in this case with an Ar/H₂ ratio of 2 is expected to be in between these two flows.) When the SiH₄ flow increases, the depletion decreases because, gradually, there is not enough H to decompose all SiH₄. In Fig. 3(c), the deposition rate is shown as a function of the SiH₄ flow. As the SiH₄ flow increases, the growth rate increases because more depositing radicals are created. Note that there seem to be two distinct slopes in the graph: a steep slope for the data points with a low SiH₄ flow (≤3 sccm) and a less-steep slope for the higher flows.

The observation of the different slopes might be interpreted as follows. First, it could be argued that this is due to the varying mass density of the films, which is suggested by the varying refractive indices in Fig. 3(d). To exclude this, the mass density of the films is calculated using the infrared
refractive index, the hydrogen content, and the crystalline fractions. From the mass density, the thickness, and the deposition time, the growth flux can be estimated. Furthermore, the SiH₄ flow multiplied by the depletion will give the rate at which the radicals (SiHₓ with 0 ≤ x ≤ 4, we neglect the production of higher radicals SiₓHₘ with n > 1) are produced and are eventually deposited. These two quantities are plotted in Fig. 4. The slope of the line through the data points in this graph indicates the number of Si atoms incorporated in the film per created SiHₓ radical (0 ≤ x ≤ 4). This means that the slope is proportional to an effective, or averaged, sticking probability. In Fig. 5, we calculated the ratio of the growth flux and the radical production rate as a function of the SiH₄ flow. The values plotted in Fig. 5 are proportional to the sticking probability. It is clear that the sticking probability is much higher for the low SiH₄ flows than for the high SiH₄ flows. This is due to the fact that for low SiH₄ flows, follow-up hydrogen abstraction reactions creating SiHₓ with 0 ≤ x ≤ 3 are more likely, and these radicals have sticking probabilities which are close to 1.10,37 The values given in Fig. 5 do not directly correspond with the averaged sticking probability on the substrate. The material is also deposited on the walls of the reactor, although probably at lower rates than at the substrate. Therefore, the effective deposition area is larger than the substrate size of 100 cm², which is used in the calculation of the growth flux. However, this geometrical factor will be constant as long as only the SiH₄ flow is varied since the plasma expansion properties, and with it the spatial distribution of the radicals in the reaction chamber, remain unaltered.

We can go a step further by making an additional assumption for the low SiH₄ flow. Under these conditions, we assume that SiH₄ is entirely stripped of all hydrogen, i.e., the deposition is dominated by the silicon radical, which has a measured sticking coefficient of 1.10,37 Since we find values of about 0.4/100 cm² in Fig. 5, a scaling factor of 2.5 should be introduced that corrects for the effective, growth-rate-weighted deposition area. This scaling factor is introduced on the right-hand axis of Fig. 5. Consequently, for the highest SiH₄ flows, a sticking coefficient of 2.5 × 0.136 = 0.34 is found. For comparison, the surface reaction probability of SiH₃ is reported to be 0.28 (Ref. 38) and 0.30.37 Probably,
for these high SiH₄ flows, the SiH₃ radicals contribute significantly to the deposition, although the contribution of other radicals is not yet fully minimized.

What are the implications of this observation? The refractive index in Fig. 3(d) shows a typical value of about 4 for good-quality a-Si:H at a SiH₄ flow of 40 sccm at a growth rate of 4 nm/s. But when the SiH₃ flow decreases, the refractive index also decreases, indicating that the material becomes porous. This also holds for the microcrystalline films. At a SiH₄ flow of 5 sccm (close to the transition to a-Si:H), the refractive index is 2.8 [Fig. 3(d)]. At this SiH₄ flow, the average sticking coefficient is 0.6 (Fig. 5). At the lowest SiH₄ flows, the refractive index is 2.7 and the average sticking coefficient is almost 1. This correlation between the average sticking coefficient and the refractive index strongly suggests that SiH₃ plays a role in the improvement of the properties of the a-Si:H tissue in between the crystalline material. However, keeping both the dilution high (and thus, a high H flux to the substrate) to obtain the μc-Si:H films and, simultaneously, the SiH₃ flux, relative to the other silane radicals (and thus, improve the quality of the a-Si:H fraction in μc-Si:H), is a requirement that is hard to fulfill because SiH₃ and H react with each other.

B. Ring position variation

How can we obtain a substantial contribution of the SiH₃ radicals to the deposition of the film while simultaneously a high flux of atomic hydrogen to the substrate? This can be achieved by decreasing the interaction time of the injected SiH₄ with the expanding hydrogen plasma. In this case, only one hydrogen abstraction reaction can take place before SiH₃ is deposited at the substrate surface. With ETP CVD, this can be realized by injecting SiH₄ into the expanding plasma close to the substrate (here we take 55 mm above the substrate). On injection, the SiH₃ molecules will be dragged along with the plasma at a velocity of 100–200 m/s. The atomic hydrogen density is in the range of 10^{19}–10^{20} m^{-3}. It can be estimated, based on the rates for reactions (1)–(4) reported in literature, that only a small fraction of the SiH₃ radicals created in reaction (1) will react with the atomic hydrogen [reactions (2)–(4)] to form SiH₄ with 0≤x≤3 before they reach the substrate surface.

In Fig. 6, the crystalline fraction, the SiH₄ depletion in the plasma, the growth rate, and the refractive index are shown for the varying SiH₃ flow and for the different injection ring positions. As can be seen from Fig. 6(a), the SiH₃ flow at which the transition from μc-Si:H to a-Si:H occurs shifts to higher SiH₃ flows for the lower injection position. This is a first indication that the plasma chemistry is influenced as the injection position is changed, resulting in a different film deposition mechanism. On the other hand, the refractive index does not seem to depend much on the SiH₃ flow. However, if we look at the films containing an equal crystalline fraction, for example 50%, then the refractive index increases from 3.1 to 3.4 if the injection position is decreased from 365 to 55 mm above the substrate [Fig. 6(d)]. This implies that the lowering of the SiH₃ injection position improves the material properties of the μc-Si:H deposited under these conditions. Similar trends are observed for a hydrogen flow of 1500 sccm and a varying injection ring position. The improvement of the μc-Si:H properties is also reflected in the optoelectronic properties; for the lowest injection ring position, a photoresponse of 50 at a light conductivity of 10^{-6} S/cm is obtained, whereas for the high injection ring position, the photo-and dark conductivity are almost equal at about 10^{-7} S/cm.

These observations imply that the composition of the particle flux that arrives at the substrate surface depends on the injection position. To obtain more insight in the nature of the influence of the injection position on the mix of reactive species arriving at the substrate, we describe the deposition plasma with a simple plug-down model.39,40 In this model, we apply a number of simplifications for the sake of clarity.
The SiH₄ is injected (white arrow) homogeneously in a plane (4) and reacts with the atomic hydrogen on the way toward the substrate. The deviations that form the simple model are also indicated. The hydrogen plasma expands, therefore, atomic hydrogen will also flow to the side of the reactor (3), where it reacts with the SiH₄ that did not deposit on the substrate and was not pumped (6) but recirculates into the reactor (7). The reaction products can enter the plasma expansion (8) and find their way to the substrate again.

Therefore, the absolute values that result from the subsequent calculations might not be entirely correct, but the trends will provide a good illustration of the effect of the SiH₄ injection position. We consider a parallel beam of atomic hydrogen having a uniform density of 10²⁰ m⁻³ (valid for a hydrogen flow of about 2000 sccm) flowing toward the substrate with a velocity of v=200 m/s. The heavy particle temperature in the plasma beam at substrate level is set at 600 K. These values are obtained from two-photon absorption-laser-induced fluorescence and electron-beam-induced fluorescence measurements and mass spectrometry carried out on comparable plasmas. The gray column between the plasma source and the substrate in Fig. 7 represents this beam. In reality, the beam expands, resulting in a decreasing atomic hydrogen density going from the plasma source to the substrate. We approximate the SiH₄ injection (with flow Q=5 sccm) through the eight holes in the injection ring by a homogeneous injection in a plane with an area A=64 cm² parallel to the substrate surface, as indicated by the white arrow. Since there is no signature of the eight injection holes in the thickness variation of the deposited films, the injection can indeed be assumed to be uniform. This is explained by the fact that the thermal velocity (about 500 m/s) is higher than the drift velocity of the plasma expansion. The vertical position of the injection plane is allowed to vary. We assume that the SiH₄ molecules instantaneously pick up the drift velocity of the plasma toward the substrate. Furthermore, we assume that the atomic hydrogen density is higher than the SiH₄ density, so we can neglect the atomic hydrogen consumption and keep the density constant. The SiH₄ depletion depends only slightly on the SiH₄ flow [see Fig. 6(b)], which justifies this assumption. Furthermore, if the atomic hydrogen density decreases due to the reactions with SiH₃, the reaction rates of reactions (1)–(4) will not change relative to each other, and the qualitative results of the model calculations are still valid.

In order to calculate the radical densities SiHₓ, the following set of differential equations has to be solved:

\[
\nu \frac{\partial n_{\text{SiH}_x}}{\partial z} = k_n \rho H \rho_{\text{SiH}_x} - k_n \rho H n_{\text{SiH}_{x-1}},
\]

in which \(n\) is the density in m⁻³ (\(n_{\text{SiH}_0} = 0\) for \(x < 0\) and \(x > 4\)), \(\nu\) is the drift velocity, and \(z\) is the distance from the injection ring toward the substrate. The following boundary condition should be satisfied:

\[
n_{\text{SiH}_4}(z = 0) = \frac{QN_z}{V_m \nu A},
\]

in which \(Q\) is the SiH₄ mass flow (5 sccm), \(N_a\) is Avogadro’s constant (6.022 × 10²³ mol⁻¹), and \(V_m\) is the volume per mol (24.5 × 10⁻³ m³).

The set of differential equations with all the relevant experimental parameters was solved using the MAPLE V software. In Fig. 8, the results of the calculation of the radical and the molecule densities at substrate level as a function of the distance of the injection ring to the substrate is shown. Clearly, the subsequent hydrogen abstraction reactions become more important as SiH₄ is injected more remote from the substrate. The SiH₄ density at substrate level decreases as the distance from the injection ring increases; the SiH₃ density first increases, until the SiH₄ density is so low that the consumption rate in reaction (3) becomes more important as SiH₄ is injected more remote from the substrate. The SiH₃ density at substrate level decreases as the distance from the injection ring increases; the SiH₂ density first increases, until the SiH₃ density is so low that the consumption rate in reaction (2) becomes more important as SiH₄ is injected more remote from the substrate. The SiH₂ density at substrate level decreases as the distance from the injection ring increases; the SiH density first increases, until the SiH₂ density is so low that the consumption rate in reaction (1) becomes more important as SiH₄ is injected more remote from the substrate.
the injection ring is to the substrate, the higher the SiH₃ density, relative to the density of the other silane radicals, becomes.

The model described previously is a rather simple and linear model, and here, we want to investigate the limitations of the model using the experimental results from Fig. 6. The proposed model predicts a decrease in the SiH₃ depletion as the injection position becomes lower (cf. SiH₃ density in Fig. 8) because the interaction time with the atomic hydrogen is shorter. The experimental results, however, show hardly any influence: the depletion is about 80% for all applied injection positions. This immediately reveals the biggest weakness of the model: the recirculation of the particles in the reaction chamber is not accounted for. Recirculating SiH₃ (and even SiH₄, which is reflected at the substrate) will have a relatively long interaction path with H in the background gas and with the expanding plasma. This recirculating gas will generate a SiH₃ flux (σ < 3) at the substrate surface, which partly masks the increase in the SiH₃ flux due the lowering of the injection position. Therefore, the growth rate and the depletion are not significantly influenced by the injection position. Probably, a process chamber in which recirculation plays a smaller role (e.g., larger diameter of the reaction chamber or metal plates that block recirculating gas flows) can emphasize the influence of the injection position on the material properties and might even be the most promising way to further improve the quality of ETP CVD-deposited µc-Si:H.

IV. CONCLUSIONS

Two observations have been described in this paper. Firstly, the investigation of the growth rate as a function of SiH₃ flow around the transition region between amorphous and microcrystalline material reveals that the average sticking coefficient of the depositing radicals decreases and the refractive index increases with increasing the SiH₃ flow. Secondly, the material properties of µc-Si:H improve when SiH₃ is injected closer to the substrate, which can only be explained by a change in the radical composition at the substrate. A model that we applied to describe the deposition plasma indicates that there is a relative increase in the SiH₃ radical flux when the injection position is closer to the substrate. These two observations strongly suggest that an increase in the contribution of SiH₃ to the growth of µc-Si:H improves the material quality.

ACKNOWLEDGMENTS

Ries van de Sande and Jo Jansen are acknowledged for the design and the construction of the deposition system and for their technical assistance. Martijn Tijssen and Arjan Driessen are acknowledged for their skillful technical assistance. This research was financially supported by NOVEM.

39S. Mazoufere, M. G. H. Boogaards, I. S. J. Bakker, P. Pandan, R. Engeln,