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### Influence of Pressure and Plasma Potential on High Growth Rate Microcrystalline Silicon Grown by Very High Frequency Plasma Enhanced Chemical Vapour Deposition

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Microcrystalline silicon ( $\mu$ c-Si) based single junction solar cells have been deposited by very high-frequency plasmaenhanced chemical vapor deposition (VHF-PECVD) using a showerhead cathode at high pressures in depletion conditions. The i-layers are made near the transition from amorphous to crystalline. An energy conversion efficiency of 9.9% is obtained with a single junction solar cell that is deposited on a texture-etched ZnO:Al front contact. The  $\mu$ c-Si i-layer is 1.5  $\mu$ m thick, deposited at a rate of 0.5 nm/s. In order to control the material properties in the growth direction, the hydrogen dilution of silane in the gas phase is graded following different profiles with a parabolic shape. Materials with higher deposition rates were developed by increasing the RF power and the total gas flow such that the depletion condition is constant. At a deposition rate of 4.5 nm/s, a stabilized conversion efficiency of 6.7% is obtained for a single junction solar cell with a  $\mu$ c-Si i-layer of 1  $\mu$ m. It is found that the defect density increases one order of magnitude upon the increase in deposition rate from 0.45 to 4.5 nm/s. This increase in defect density is partially attributed to the increased energy of the ion bombardment during the plasma deposition. We have introduced an additional method to limit the ion energy by controlling the DC self bias voltage using an external power source. In this way, the defect density in the  $\mu$ c-Si layers is decreased and the performance of the solar cells is further improved. It is observed that the performance of solar cells deposited at high rate improves under light soaking conditions at 50 °C, which we attribute to post deposition equilibration of a fast deposited transition material. [DOI: 10.1143/JJAP.45.6166]

KEYWORDS: Solar cells, microcrystalline, silicon, chemical vapour deposition, plasma, thin film

#### 1. Introduction

Thin-film solar cells based on microcrystalline silicon ( $\mu$ c-Si) [single junction hydrogenated  $\mu$ c-Si ( $\mu$ c-Si:H) and "micromorph" tandem hydrogenated amorphous silicon (a-Si:H)/ $\mu$ c-Si:H] are advantageous compared to cells solely based on a-Si:H, because  $\mu$ c-Si:H has a lower band gap, due to which the optical absorption in the infrared region of the solar spectrum is higher. Transition-type  $\mu$ c-Si:H (deposited close to the regime of a-Si:H growth) has promising optoelectronic properties for absorber layers applications. However, inhomogeneous growth and small but significant light induced degradation for cells with these types of materials limit their ultimate efficiency. This paper addresses these issues, especially for  $\mu$ c-Si:H made at high deposition rate.

Traditionally,  $\mu$ c-Si:H is made by plasma-enhanced chemical vapor deposition (PECVD) under high hydrogen dilutions of the silane source gas. When high deposition rates are involved, care has to be taken to reduce the energy of the ion bombardment and to allow for a homogeneous deposition. Single junction  $\mu$ c-Si:H and tandem  $\mu$ c-Si:H/ a-Si:H solar cells are being developed by several groups<sup>1–5)</sup> using different types of deposition configurations and plasma regimes. Here, we combine the approaches of very highfrequency (VHF) PECVD<sup>4–8)</sup> at 60 MHz, high pressure depletion (HPD) conditions,<sup>7–10)</sup> and a shower head gas inlet,<sup>2,10)</sup> that have separately been studied by other groups.

#### 2. Experimental

Transition-type  $\mu$ c-Si i-layers have been made at an RF frequency of 60 MHz, at pressures of 5–10 mbar, using an electrode distance of 6 mm. The substrate temperature is

estimated to be 180 °C. The hydrogen dilution  $d_{\rm H}$  ( $d_{\rm H} = f_{\rm H_2}/f_{\rm SiH_4}$ : the mass flow ratio of the hydrogen and silane source gases), is chosen such that the material is deposited in a regime close to the microcrystalline-to-amorphous transition.

The optical emission of the plasma (due to transitions from excited states of molecules that are formed due to electron impact reactions) is detected through a quartz window view port. An Avantes MC2000 calibrated chargedcoupled device (CCD) spectrometer is used to record the spectrum. A lens system and fiber optics are used to focus on the emission from the pre-sheath region near the substrate. The spectral resolution is in the order of 1 nm; integration times of around 0.5 s are required. By means of a calibration routine, the spectra are corrected for the decreasing transmission of the quartz window due to the deposition on the window. With optical emission spectroscopy (OES), the intensity of the Si\* line (289 nm) and the Balmer alpha (H $\alpha$ ) line (656 nm) are recorded and analyzed.

Additionally, a one-dimensional (1D) plasma model is used to calculate concentrations and reaction rates spatially resolved. In the model, the electric field, densities, and fluxes of the particles are calculated self-consistently.<sup>11</sup>

The crystallinity of the deposited material is estimated using Raman spectroscopy. The crystalline ratio  $R_c$  was obtained as

$$R_{\rm c} = (I_{520} + I_{510}) / (I_{480} + I_{520} + I_{510})$$

where  $I_{520}$  and  $I_{510}$  are the intensities of crystalline contribution whereas  $I_{480}$  is that of amorphous contribution to the transverse optic Si–Si vibrations in the Raman spectra. The value of  $R_c$  is underestimated compared to actual crystalline volume fraction  $X_c$ .

Single junction p-i-n solar cells have been made with transition-type  $\mu$ c-Si:H i-layers deposited on two types of

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Fig. 1. Schematic cross-sectional view of the device configuration of a pi-n solar cell. The texture of the TCO (that extends through the whole pi-n stack) is not drawn.

substrates: Asahi U-type, cut from "large area" rough  $SnO_2$ :F substrates (protected with 10 nm RF magnetron sputtered ZnO:Al) and texture-etched ZnO:Al<sup>12,13</sup> (made at FZ Jülich, Germany) coated glass. The  $SnO_2$ :F was protected with a 10 nm thick sputtered ZnO:Al. A µc-Si:H p-layer and an a-Si:H n-layer were used as doped layers. The n-layer is chosen amorphous to prevent post-deposition oxidation of the i-layer and to reduce peripheral current collection. A 100 nm thick sputtered ZnO:Al back reflector was used to improve the reflection at the back of the cell. The back contacts were made as a Ag/Al double layer, evaporated through a mask defining the cell size (4 × 4 mm<sup>2</sup>). A schematic cross section of a p–i–n solar cell is given in Fig. 1.

The solar cells have been tested by current–voltage (I-V) measurements at 25 °C under simulated AM1.5 light (Wacom dual beam solar simulator) and in dark. The parameters efficiency ( $\eta$ ), open circuit voltage ( $V_{oc}$ ), short circuit current ( $J_{sc}$ ), fill factor (FF), parallel resistance ( $R_p$ ), and series resistance ( $R_s$ ) are derived from the I-V measurements. In the following,  $R_s$  and  $R_p$  are defined as the inverse slope of the I-V curve at the  $V_{oc}$  and  $J_{sc}$ , respectively. Spectral response measurements have been performed under white bias light (spectrum and intensity close to AM1.5). The light soaking experiments of the  $\mu$ c-Si:H cells have been performed under white light (spectrum and intensity close to AM1.5) that is filtered with a filter that approaches the

absorption in a typical top cell of an a-Si/ $\mu$ c-Si tandem cell. The degradation took place at a controlled temperature of 50 °C, whereas the periodic measurements took place at 25 °C.

To determine the defect density  $N_d$  in the i-layers of the solar cells, Fourier-transform photocurrent spectroscopy (FTPS) was performed in Prague.<sup>14)</sup> The absorption coefficient  $\alpha$  in a broad wavelength range is measured in solar cells by analyzing the generated current. The  $\alpha$  spectra are calibrated to the c-Si absorption spectrum at 1.4 eV. This overcomes problems with the signal amplification due to light scattering. Light scattering is dominated by a surface scattering in our case, due to rough TCO used as the substrate. The absorption coefficient  $\alpha$  at 0.8 eV can be considered to be proportional to the midgap defect density of states. The proportionality constant of  $\alpha(0.8 \text{ eV}) = 0.12 \text{ cm}^{-1}$  corresponding to  $N_d = 2 \times 10^{16} \text{ cm}^{-315}$  is used to estimate  $N_d$ . Free carrier absorption of TCO was measured in near IR and it was taken into account when evaluating  $\alpha$ .

#### 3. Results and Discussion

#### 3.1 Optimum hydrogen dilution

Due to the development of crystallites during the deposition, the crystalline fraction  $X_c$  of  $\mu$ c-Si:H increases with thickness. In the case of a p–i–n solar cell this results in a higher crystalline fraction at the i/n interface compared to the p/i interface. It was found that there is an optimum in cell performance at a certain crystalline fraction (close to the transition-type  $\mu$ c-Si).

The I-V parameters of cells, for Asahi U-type transparent conductive oxide (TCO), with i-layers grown at 0.45 nm/s, made at the optimized constant hydrogen dilution  $(d_{\rm H})$  and a higher dilution are listed in Table I. The i-layer thickness is 1.5  $\mu$ m. It is observed that at the lower  $d_{\rm H}$  the cell parameters, especially the FF, are not strongly dependent on the type of TCO. At the higher  $d_{\rm H}$ , however, the solar cell performance is much more sensitive to the substrate. In both cases the trend in  $V_{oc}$  is to decrease and that in  $J_{sc}$  to increase due to the increasing  $X_c$  with higher  $d_{\rm H}$ . At the higher  $d_{\rm H}$ , the FF of cells on Asahi U-type TCO decreases while for the cells on texture-etched ZnO:Al it increases. Moreover,  $J_{sc}$ and  $V_{\rm oc}$  are higher in case of texture-etched ZnO:Al compared to Asahi U-type SnO<sub>2</sub>:F. For an i-layer thickness of 1.5 µm, this leads to an excellent cell performance of  $\eta = 9.9\%$  with a high fill factor of 0.73 and a low series resistance of  $1.43 \,\Omega \,\mathrm{cm}^2$ .

#### 3.2 Plasma conditions for high deposition rates

The depletion of silane is defined as the fraction of the  $SiH_4$  molecules that is converted to other species (growth precursors) by means of dissociation reactions. The number

Table I. I-V parameters of cells on Asahi U-type SnO<sub>2</sub>:F and texture-etched (t.-e.) ZnO:Al.

TCO	$d_{ m H}$	η (%)	$J_{\rm sc}$ (mA/cm <sup>2</sup> )	V <sub>oc</sub> (V)	FF	$R_{\rm s}$ ( $\Omega  {\rm cm}^2$ )	$R_{\rm p}$ ( $\Omega  {\rm cm}^2$ )	п	$J_0$ (A/cm <sup>2</sup> )
Asahi SnO <sub>2</sub> :F	24.5	8.1	20.9	0.55	0.70	2.6	670	1.52	$9.4 \times 10^{-9}$
te. ZnO:Al	24.5	8.9	21.9	0.56	0.71	2.2	1163	1.24	$5.2  imes 10^{-10}$
Asahi SnO <sub>2</sub> :F	27.6	7.5	23.9	0.48	0.64	3.3	432	1.57	$1.1 \times 10^{-7}$
te. ZnO:Al	27.6	9.9	25.9	0.52	0.73	1.4	559	1.36	$8.0  imes 10^{-9}$



Fig. 2. Gas utilization parameter for the i-layer deposition and Raman crystalline ratio measured on the cell (n-layer side) versus the i-layer deposition rate  $r_d$ .

of SiH<sub>4</sub> dissociation reaction can be estimated by measuring the optical emission intensity  $I_{OES}$  of the Si<sup>\*</sup> line. The growth rate  $r_d$  per SiH<sub>4</sub> flow unit [expressed in the gas utilization parameter  $c_d = r_d/f_{SiH_4}$  (nm s<sup>-1</sup> sccm<sup>-1</sup>)] gives an indication of the available density of growth precursors. Both methods give indirect information on the degree of depletion.

A series of  $\mu$ c-Si:H based p–i–n solar cells has been made in which the plasma power for the deposition of the  $\mu$ c-Si:H i-layer is increased for each cell in order to increase the deposition rate. In first instance, the total gas flow  $f_t = f_{SiH_4} + f_{H_2}$  is increased with the same ratio as the power with the intention to keep the degree of depletion constant. The dilution  $d_H$  is kept constant.

Figure 2 shows the gas utilization parameter  $c_d$  and the Raman crystalline fraction  $R_c$  measured at the n-layer side (resulting in an underestimated value for  $R_c$ , due to the a-Si:H n-layer) as a function of the deposition rate for this series. A solar cell with a deposition rate of 1.5 nm/s was deposited at an RF power 50 W; a deposition rate of 2.3 nm/s was obtained at 100 W (0.60 W/cm<sup>2</sup>). An increase of RF power and  $f_t$  with another factor of two, resulted in the data plotted with the round symbol in the figure at  $r_d = 3.9$ nm/s. Obviously, the  $r_d = 3.9$  nm/s cell is amorphous. On the left axis of Fig. 2, the gas utilization parameter  $c_d$  is given. It is calculated that the maximum  $c_d$  (where all Si atoms are deposited) based on geometrical considerations is around  $c_d = 0.29 \text{ nm s}^{-1} \text{ sccm}^{-1}$ . It is clear that in the 3.9 nm/s case,  $c_d$  is lower compared to the other cases, attributed to a lower degree of depletion. The loss in crystallinity is attributed to the loss of atomic hydrogen through abstraction reactions with SiH<sub>4</sub>:

#### $H + SiH_4 \rightarrow SiH_3 + H_2$ ,

because the availability of SiH<sub>4</sub> is higher as the degree of depletion is lower. This principle is confirmed by plasma modeling (1D): Figure 4 shows the creation rate of atomic hydrogen as well as the loss to abstraction reactions for different total gas flows  $f_t$  as a function of position between the electrodes. It is observed that the loss of atomic hydrogen decreases with decreasing  $f_t$ , from which it can be concluded that source gas depletion contributes to a high atomic hydrogen concentration.



Fig. 3. (a) Variation of Si<sup>\*</sup> intensity as a function of silane flow using electrode configuration A. The hydrogen dilution  $d_{\rm H}$  is fixed for each series. The RF power is kept constant at P = 15 W, the pressure at p = 1.1 mbar,  $d_{\rm H}$  is indicated. (b) Variation of Si<sup>\*</sup> intensity and H $\alpha$ /Si<sup>\*</sup> ratio as a function of silane flow ( $d_{\rm H}$  is constant) for different powers using electrode configuration B.



Fig. 4. Spatially resolved hydrogen balance for different total gas flows.

This phenomenon is also reflected in the OES data: Figure 3 shows the  $H\alpha/Si^* I_{OES}$  ratio as a function of total flow using different RF powers. At each power, an increase of  $H\alpha/Si^*$  with decreasing total gas flow is observed. The increase becomes more pronounced when the total flow is below a certain threshold, reflecting the decrease of the loss of atomic hydrogen to abstraction reactions with increasing silane depletion.

The cells with the deposition rates of 3.5 and 4.5 nm/s are made at 200 and 400 W respectively. The total flows are adjusted to the deposition rate in order to maintain a sufficient degree of depletion (i.e., a sufficiently high  $c_d$ ) in order to maintain a sufficient atomic hydrogen concentration. The cells made under these conditions were microcrystalline (as plotted on the right axis of Fig. 2).

When the total gas flow and the RF power are increased with equal factors, in first order approximation  $c_d$  is expected to remain constant, as the degree of depletion is not expected to change. However, a decrease in  $c_d$  with increasing power and total flow in steps with equal factors was observed (the data at  $r_d = 1.5$ , 2.3, and 3.9 nm/s in Fig. 2). This can be attributed to the lower sticking probability of the growth precursors due to the shorter residence time  $t_r$  (the step from 100 W, 310 sccm to 200 W, 621 sccm implies a change of  $t_r$  from 0.1 to 0.05 s). It is remarkable that when the indirect depletion parameter  $c_d$ —that is based on the sticking and deposition of growth precursors rather than the gas phase reactions—is used to tune the total flow, the deposited material has more or less similar crystalline fractions. While when  $I_{OES}(Si^*)/f_{SiH_4}$  is kept constant, the material obviously becomes amorphous at higher powers. In other words, higher degrees of depletion are necessary for crystalline growth at high rates. However, because  $R_c$  increases with P, we speculate that when the approach using  $c_d$  is followed, the increase of depletion with power is higher than necessary for maintaining similar  $R_c$  at higher  $r_d$ .

# 3.3 Cell performance and defect density at high deposition rates

A series of  $\mu$ c-Si:H based p–i–n solar cells has been made in which the plasma power for the deposition of the  $\mu$ c-Si:H i-layer is increased for each cell in order to increase the deposition rate. The total gas flow is increased with the same ratio as the deposition rate, to keep the depletion condition constant, following the procedure as described in §3.2. The dilution  $d_{\rm H}$  is kept constant. The cell parameters  $V_{\rm oc}$  and FF decrease with increasing deposition rate.

Figure 5 shows typical FTPS sub band gap absorption spectra of a cell deposited on Asahi U-type SnO<sub>2</sub>:F as a function of deposition rate  $r_d = 0.45$  nm/s. The absorption coefficient  $\alpha$  at 0.8 eV is considered to be proportional to the midgap defect density of states. However, there are many uncertainties in the interpretation of the absolute value of the FTPS signal of solar cells deposited on rough substrates, such as the effect of light scattering, the free carrier absorption in the TCO and the presence of the interference fringes. A very rough estimation of the defect density could be made using the relation that at 0.8 eV, an  $\alpha$  of 0.12 cm<sup>-1</sup> corresponds to  $N_d = 2 \times 10^{16}$  cm<sup>-3</sup>.<sup>15</sup>) The TCO used is not optimized for microcrystalline silicon and residual signal interferences are seen; they should be averaged on the log scale for evaluation of  $\alpha(0.8 \text{ eV})$ . The optical absorption of



Fig. 5. Sub band gap absorption measured by FTPS on  $\mu$ c-Si:H solar cells on Asahi U-type SnO<sub>2</sub>:F deposited at different deposition rates from  $r_{\rm d} = 0.45$  to 4.5 nm/s. Low photon energy optical absorption coefficient  $\alpha$  is set to the absolute (cm<sup>-1</sup>) scale at 1.4 eV.

different TCOs influence the  $\alpha$  values, and has to be taken into account; but in our series always the same TCO was used. A small systematic error can be due to a different crystalline content and the calibration constant given above is known just within a factor of 2. Therefore, the absolute values of defect density are not very precise. On the other hand relative changes are enough precise, due to the same scattering and absorption effect of the same TCO in the series and just very slowly varying crystalline content (see Fig. 2).

From our data, we can unambiguously see that an increase in deposition rate of one order of magnitude leads to an increase in  $N_d$  of roughly one order of magnitude. The decrease of the solar cell performance can—at least partially—be explained by this decrease of the layer quality.

Figure 6 shows the correlation between the solar cell parameters and  $N_{\rm d}$  (as obtained using the above mentioned proportionality constant and precautions described above) for solar cells with i-layers deposited at different rates. For all the cells from this series,  $d_{\rm H}$  was chosen such that the crystallinity of the i-layer is in the transition regime, so that the performance is optimal. The cells deposited at low rate show very low defect densities. The deposition rates  $r_{\rm d}$  are indicated in the figure. It is observed that with an increase of  $r_{\rm d}$  from 0.45 to 4.5 nm/s,  $N_{\rm d}$  increases roughly one order of magnitude. The overall trend shows a strong correlation between the solar cell parameters and  $N_{\rm d}$ . This suggests that the cell performance is limited by the number of mid-gap states in the i-layer as it should be expected.<sup>17)</sup> A higher  $N_{\rm d}$ leads to an increased dark saturation current density  $J_0$  and lower  $V_{oc}$ . The diode quality factor n is higher and FF is lower due to the increased recombination rate through the mid-gap defect states.

In the depositions of the cells with the i-layers deposited at P = 200 and 400 W with constant  $c_d$ , also texture-etched ZnO:Al on glass was added as a substrate, which leads to considerably better cell performance. The best cell deposited on texture-etched ZnO:Al at the highest rate of  $r_d = 4.5$ nm/s using an RF power of 400 W (2.4 W/cm<sup>2</sup>) has the following *I*–V parameters:  $\eta = 6.4\%$ ,  $J_{sc} = 22.3$  mA/cm<sup>2</sup>,  $V_{oc} = 0.45$  V, FF = 0.63. Most probably  $\eta$  can be improved by fine-tuning  $R_c$  by varying  $d_{\rm H}$ .

The initial cell efficiencies obtained for the cells with high i-layer deposition rates, are compared to similar results obtained by other groups in Fig. 7. It shows that, in a deposition rate regime that is not yet well-explored, we obtained excellent results with i-layers deposited in our reactor with VHF PECVD under HPD condition with showerhead electrode.

#### 3.4 Stability of cell performance under light soaking

The solar cells with the high-rate  $\mu$ c-Si:H i-layers have been exposed to a-Si:H cell filtered light (simulating top cell of a micromorph cell) at 50 °C after the initial measurements. The solar cell parameters have been checked under AM1.5 light at room temperature on a logarithmic time scale. The results are given in Fig. 8. It can be seen that the *I–V* parameters of these solar cells deposited at high-rate actually improve upon light soaking, contrary to the 5–10% degradation of the transition materials with light soaking reported in the literature. It should be noted here, that in our



Fig. 7. Cell performance vs deposition rate for μc-Si:H based single junction solar ells. Solar cells from the present studies are compared to results published by other groups (based on ref. 16 and recent reports).

deposition rate  $r_{d}$  (nm/s)



experiment, after the deposition and the metal evaporation, the cells are not thermally treated ("annealed"). The changes in cell performance with light soaking is mainly due to the improving FF, which can be partially explained by the decrease of the series resistance and also to post-deposition defect equilibration of the material deposited at high rates, induced by the temperature and/or the light exposure. The decrease of diode quality factor (*n*) and reverse saturation current ( $J_0$ ) with light soaking points towards a decrease in defect density that may have been caused by this equilibration. A detailed experiment on thermal treatments in the dark is in progress to distinguish between the effect of temperature and light exposure on the improvement of the cell characteristics and the results will be published in a separate article.



Fig. 8. Cell parameters vs light soaking time for cells made at different deposition rates (indicated) on Asahi U-type substrate.



Fig. 9. Solar cell parameters (best cell) for cells with i-layers deposited at a growth rate of 3.4 nm/s at 200 W on Asahi U-type SnO2:F and texture-etched ZnO:Al with different values of  $V_{dc}$  (The self bias is at  $V_{dc} = 7 \text{ V}$ ).

#### 3.5 Influence of the ion energy on cell performance

The use of high RF powers to obtain high deposition rates, leads to an increased ion energy of the positive ions bombarding the substrate. If the ion energies are too high, the material can be adversely influenced (increasing defect density, decreasing crystallinity).<sup>18,19)</sup> In  $\mu$ c-Si:H growth, the contributions of ions to the deposition can be much higher than in the a-Si:H case.<sup>20)</sup> The increase of the ion bombardment might play a role in the increase in  $N_d$  and the decrease of performance with increasing  $r_d$ , although the use of VHF PECVD and high pressures are already contributing to a reduction of the ion energy bombardment. However, because the electrode configuration is effectively very much symmetric due to the small electrode distance, the DC self bias voltage  $V_{dc}$  is low, due to which the plasma potential  $V_{pl}$  is relatively high, leading to a high ion energy.

A series of solar cells was made with the i-layers deposited at 200 W, varying the DC bias voltage  $V_{dc}$  from the DC self bias (-7 V in these conditions) to more negative values. By fixing  $V_{dc}$  to a lower negative value, using an external voltage source, the ion energy can be reduced. This would improve the quality of the µc-Si:H layers and the performance of the solar cells. The three cells with varied external bias had very similar deposition rates of  $r_d = 3.5$ nm/s. Figure 9 shows the solar cell parameters from the AM1.5 *I–V* measurements plotted versus the  $-V_{dc}$  for cells deposited on Asahi U-type SnO2:F and texture-etched ZnO:Al in the same deposition runs. It is observed that the  $V_{\rm oc}$  decreases and the  $J_{\rm sc}$  increases with increasing  $-V_{\rm dc}$ . This is an indication for increasing crystallinity. Raman spectroscopy measurements confirm this hypothesis. In the self bias case the maximum FF is 0.67 on texture-etched ZnO:Al. With increasing  $-V_{dc}$ , an improvement is observed: for the  $V_{\rm dc} = -20$  V cell, the FF is 0.69.

FTPS measurements show that  $N_d$  is almost a factor of 2 lower for the sample with  $V_{dc} = -20$  V compared to the self-bias condition (shown in Fig. 6). The increase in crystallinity might be explained by the decreased impact by ions during crystal evolution. Optical emission spectroscopy experiments under such bias change showed that

the gas phase reactions reflected in the  $H\alpha/Si^*$  intensity ratio are hardly influenced by the applied DC bias. It can be concluded that the ion bombardment does play a role in the increase of  $N_d$  and the decrease in cell performance with ilayers deposited at high rate, even if high RF frequencies and high pressures are used.

#### 4. Conclusions

Using VHF PECVD in the high-pressure depletion (HPD) regime with a small electrode distance and a shower-head gas inlet, we developed p-i-n solar cells, on textured TCOs, based on transition-type µc-Si:H i-layers. The cell performance is strongly dependent on the crystalline fraction. It was found that by grading the hydrogen dilution during the i-layer deposition, a more homogeneous crystalline fraction in the growth direction could be obtained, with which the cell performance could be improved. The optimum crystalline fraction appeared to be strongly substrate dependent: on texture-etched ZnO:Al it is possible to fabricate high performance cells at a higher crystalline fraction than on ZnO:Al protected Asahi U-type SnO<sub>2</sub>:F. On texture-etched ZnO:Al, a p-i-n cell has been made at 180 °C with an i-layer deposition rate of  $r_{\rm d} = 0.45 \text{ nm/s}$ :  $\eta = 9.9\%$ ,  $V_{\rm oc} = 0.52 \text{ V}$ , FF = 0.73. This cell is stable under light soaking conditions.

By increasing the power and adapting the total flow such that the depletion condition (gas utilization) is kept constant, it was possible to make good µc-Si based solar cells at deposition rates up to  $r_{\rm d} = 4.5 \,\rm nm/s$  with  $\eta = 6.4\%$ ,  $V_{\rm oc} =$ 0.45 V, FF = 0.63, which is better than other groups reported in literature. The influence of  $r_{\rm d}$ , on the defect density in the i-layer is studied with FTPS. An increase in deposition rate by one order of magnitude, led to an increase in defect density of one order of magnitude. It is concluded that the cell performance is determined mainly by the midgap defect density in the i-layer. The decrease in FF follows the trend of the increasing  $N_d$  with  $r_d$ . Reducing the ion bombardment by applying an extra DC bias helps to decrease the defect density, indicating that ions are at least partly responsible for the decrease of the cell performance with increasing  $r_{\rm d}$ . It is observed that the cells made at high deposition rates improve under light-induced degradation, which we attribute to post deposition equilibration of a fast deposited transition material.

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