

Article

The Influence of a-SiC_x:H and a-SiO_x:H Barrier Layers Embedded in the p/i and i/n Interfaces on the Performance of a-Si:H p-i-n Solar Cells

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Abstract

In this work, hydrogenated amorphous silicon carbide (a-SiC_x:H) and hydrogenated amorphous silicon oxide (a-SiO_x:H) films with similar optical bandgaps (E_g), refractive indices (n), and extinction coefficients (k) were fabricated using pulse-wave modulation (PWM) plasma technology by controlling the plasma turn-on to turn-off time ratio (t_{on}/t_{off}). These films were placed at the 1/5 position of the p/i and i/n interfaces of hydrogenated amorphous silicon (a-Si:H) p-i-n solar cells to investigate their influence on solar cell performance. The experimental results confirmed that the deviations in E_g , n , and k were controlled to within 0.2%, 1.4%, and 4.1%, respectively. Under these conditions, placing a-SiC_x:H and a-SiO_x:H films at the p/i and i/n interfaces successfully increased the open-circuit voltage (V_{oc}). However, this also led to a decrease in the short-circuit current due to valence band (ΔE_v) or conduction band (ΔE_c) offsets. The reduction in cell fill factor (FF) and efficiency (η) caused by placing a-SiC_x:H and a-SiO_x:H films at the p/i interface was greater than that caused by placing them at the i/n interface. Placing the a-SiC_x:H film at the p/i interface significantly improved the V_{oc} to 0.8998 V. Due to the n-type doping effect of oxygen atoms, the a-SiO_x:H film exhibited the lowest FF of 43.99% and η of 4.850% at the p/i interface; however, when placed at the i/n interface, it yielded an FF of 67.38% and an η of 7.43%, which are comparable to the standard cell. Appropriately placing the a-SiC_x:H film at the p/i interface and the slightly n-type a-SiO_x:H film at the i/n interface can effectively improve the V_{oc} , FF , and η of p-i-n solar cells.

Keywords: a-Si:H; a-SiC_x:H; a-SiO_x:H; p/i and i/n interfaces; p-i-n solar cells



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1. Introduction

The material properties at the p/i and i/n interfaces of hydrogenated amorphous silicon a-Si:H p-i-n solar cells influence the electric field distribution in the i-layer [1–6]. The transport of photogenerated carriers (electrons and holes) in the i-layer is primarily driven by the drift effect of the i-layer's built-in electric field; electrons drift from the i-layer to the n-layer, while holes drift from the i-layer to the p-layer. The i-layer material properties at the p/i and i/n interfaces strongly influence the built-in electric field distribution and the collection efficiency of carriers transported to the p and n layers. Inserting high-bandgap

hydrogenated amorphous silicon carbide (a-SiC_x:H) or hydrogenated amorphous silicon oxide (a-SiO_x:H) films at the p/i interface blocks back diffusion of photogenerated electrons toward the p-layer via the conduction band offset (ΔE_c), while the valence band offset (ΔE_v) reduces the number of holes injected from the p-layer into the i-layer, which can effectively improve the open-circuit voltage (V_{oc}) of the cell [7,8]. However, high-bandgap materials have higher defect densities, which make carriers more prone to recombination. A high ΔE_v also reduces the collection efficiency of photogenerated holes crossing from the i-layer to the p-layer [9].

High-bandgap materials at the p/i interface can enhance the electric field at the p/i interface but weaken it in the subsequent bulk region of the i-layer, making photogenerated carriers produced by short-wavelength photons at the p/i interface easier to collect. In contrast, the collection efficiency of photogenerated carriers produced by medium- and long-wavelength photons inside the i-layer decreases [1–6,10–15]. The effect of adding high-bandgap materials at the i/n interface on the built-in electric field of the i-layer is smaller than at the p/i interface. Generally, the impact on cell characteristics is lower than at the p/i interface.

Common a-SiC_x:H and a-SiO_x:H high-bandgap silicon-based alloy materials may exhibit different effects on the conduction and recombination of photogenerated carriers, due to distinct defect structures arising from the addition of carbon and oxygen atoms. The a-SiO_x:H film exhibits n-type behavior due to micro n-type doping by oxygen atoms, meaning that in addition to the effects of ΔE_c and ΔE_v and defect density, the a-SiO_x:H film also experiences a positive O₃⁺ space charge that alters the built-in electric field distribution [1–6]. Adding an a-SiO_x:H high-bandgap film at the p/i interface significantly enhances the electric field at this interface due to its n-type property, while reducing the built-in electric field in the bulk region of the i-layer, causing a significant decrease in the fill factor of the cell [4–8,10–15]. In contrast, the a-SiC_x:H film, unlike the a-SiO_x:H film, does not possess n-type properties and does not interfere with the electric field via space charge formation.

The novelty of this study lies in the use of pulse-wave modulation (PWM) plasma technology to fabricate a-SiC_x:H and a-SiO_x:H films with nearly identical optical bandgaps (E_g), refractive indices, and extinction coefficients. By eliminating optical variations, we can explicitly isolate and compare the effects of the chemical bonding states—specifically, oxygen-induced n-type doping versus carbon incorporation—on the internal electric field distribution and solar cell performance.

2. Materials and Methods

a-Si:H, a-SiC_x:H, and a-SiO_x:H films, as well as p-i-n solar cells, were deposited on glass, single-crystalline silicon, and fluorine-doped tin oxide (FTO: SnO₂:F) glass substrates using PWM plasma-enhanced chemical vapor deposition (PECVD) technology [14–17]. A 13.56 MHz radio-frequency (RF) power source controlled the on and off states of the RF power using a square-wave pulse. Plasma excitation occurred during the plasma t_{on} , and plasma termination occurred during the plasma t_{off} time. Gases within the vacuum chamber were excited or ionized into active radicals during t_{on} , and these active radicals were neutralized back to their original neutral gaseous state during t_{off} . By adjusting the RF power and the t_{on}/t_{off} time ratio, the generation and neutralization of active radicals can be precisely regulated to control the bonding composition and the optical and electrical properties of the deposited films [14–17]. The PWM method was specifically chosen because it allows decoupling of the electron temperature and gas dissociation efficiency, enabling precise optical matching of E_g , refractive index (n), and extinction coefficient (k) required for this comparative study.

The process conditions included a fixed chamber pressure of 0.9 Torr and a substrate temperature of 210 °C. For the a-Si:H films, the SiH₄/H₂ flow ratio, $t_{\text{on}}/t_{\text{off}}$ time, and RF power were 20/80 sccm, 20/5 ms, and 10 W, respectively. For the a-SiC_x:H films, the SiH₄/CH₄/H₂ flow rates, $t_{\text{on}}/t_{\text{off}}$ time, and RF power were 20/2.5/80 sccm, 20/1 ms, and 30 W, respectively. For the a-SiO_x:H films, the SiH₄/CO₂/H₂ flow rates, $t_{\text{on}}/t_{\text{off}}$ time, and RF power were 20/2.5/80 sccm, 20/5 ms, and 15 W, respectively.

Figure 1 shows schematic diagrams of the p-i-n solar cells: (a) the a-Si:H reference cell (b) with a-SiC_x:H or a-SiO_x:H placed at the p/i interface, and (c) a-SiC_x:H or a-SiO_x:H placed at the i/n interface. The i-layer of the a-Si:H p-i-n reference cell was a 500 nm thick a-Si:H film, serving as a baseline for comparing cell characteristics. To verify the influence of the wide-bandgap a-SiC_x:H and a-SiO_x:H films, the film thickness was set to 100 nm, accounting for 1/5 of the 500 nm i-layer thickness. Specifically, 1/5 of the a-Si:H film at the p/i interface (where light enters the front of the i-layer from the p-layer) and at the i/n interface (where light penetrates to the rear of the i-layer) was replaced with a-SiC_x:H or a-SiO_x:H films. This resulted in two types of cells with embedded buffer layers (i_B): a p/ i_B /i/n structure of p/ i_B -a-SiC_x:H or a-SiO_x:H (100 nm)/a-Si:H (400 nm)/n, and a p/i/ i_B /n structure of p/a-Si:H (400 nm)/ i_B -a-SiC_x:H or a-SiO_x:H (100 nm)/n. The five cells were designated as follows: the reference cell (C_i-a-Si:H), cells with a-SiC_x:H added at the p/i interface (C_p/ i_B -a-SiC_x:H) and a-SiO_x:H (C_p/ i_B -a-SiO_x:H), and cells with a-SiC_x:H added at the i/n interface (C_ i_B -a-SiC_x:H/n) and a-SiO_x:H (C_ i_B -a-SiO_x:H/n).

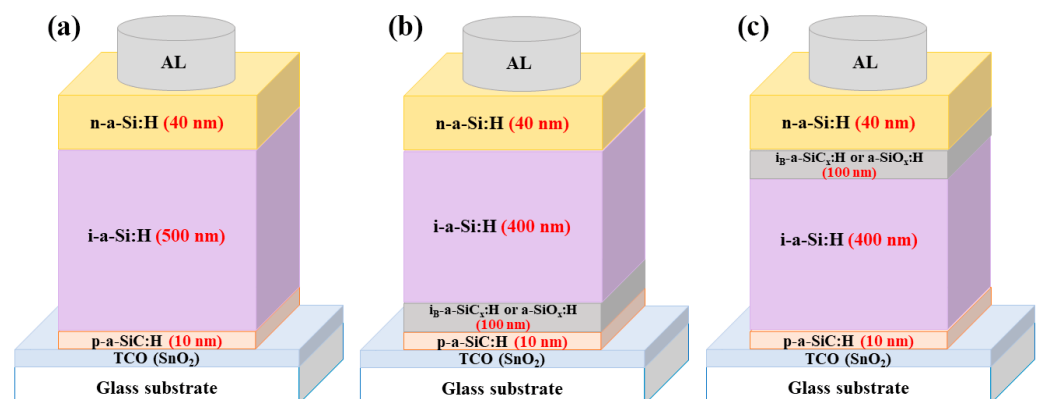


Figure 1. Schematic diagram of the p-i-n solar cells: (a) the i-a-Si:H reference cell (b) with i_B -a-SiC_x:H or a-SiO_x:H placed at the p/ i_B interface, and (c) with a-SiC_x:H or a-SiO_x:H placed at the i_B /n interface.

The n , k , and E_g of the a-Si:H, a-SiC_x:H, and a-SiO_x:H films were measured using a J. A. Woollam M-2000 ellipsometer (J. A. Woollam, Lincoln, NE, USA). Measurements were conducted over the wavelength range of 190–1700 nm at incident angles of 55°, 60°, 65°, and 70°. The polarization state (p- and s-planes) of light reflected from the film surface was measured to obtain amplitude and phase differences, and the thickness (d), E_g , n , and k were then simulated using the Tauc–Lorentz model [17,18]. The Si-H, Si-C, and Si-O bonding in a-SiC_x:H and a-SiO_x:H was measured using a Thermo Scientific Nicolet iS5 Fourier-transform infrared spectrometer (FTIR) (Thermo Fisher Scientific, Waltham, MA, USA). The current density–voltage (J - V) characteristics of the solar cells were measured using an Agilent B2912A Source Measure Unit (SMU) (Agilent Technologies, Santa Clara, CA, USA) under AM 1.5 solar illumination provided by a SAN-EI ELECTRIC XES-40S1 solar simulator (SAN-EI Electric, Osaka, Japan).

3. Results and Discussion

The optical and structural properties of the a-SiC_x:H and a-SiO_x:H films were primarily controlled by RF power and $t_{\text{on}}/t_{\text{off}}$. Figure 2 shows the E_g values of the a-SiO_x:H and

a-SiC_x:H films. For the a-SiO_x:H films, E_g gradually increased from 1.759 to 1.802 to 1.826 eV as the RF power increased from 8 to 15 to 20 W, with $t_{on}/t_{off} = 20/5$ (ms). For the a-SiC_x:H films, due to the lower dissociation efficiency of CH₄, E_g increased only from 1.777 to 1.794 eV when the RF power was raised from 20 to 30 W under the condition of $t_{on}/t_{off} = 20/5$ (ms). To increase E_g , the t_{on}/t_{off} ratio was increased to 20/1 (ms), yielding an E_g of 1.806 eV. The difference in E_g between the 1.802 eV a-SiO_x:H film and the 1.806 eV a-SiC_x:H film was less than 0.2%; thus, a-SiO_x:H and a-SiC_x:H films under these conditions were used as buffer layer materials.

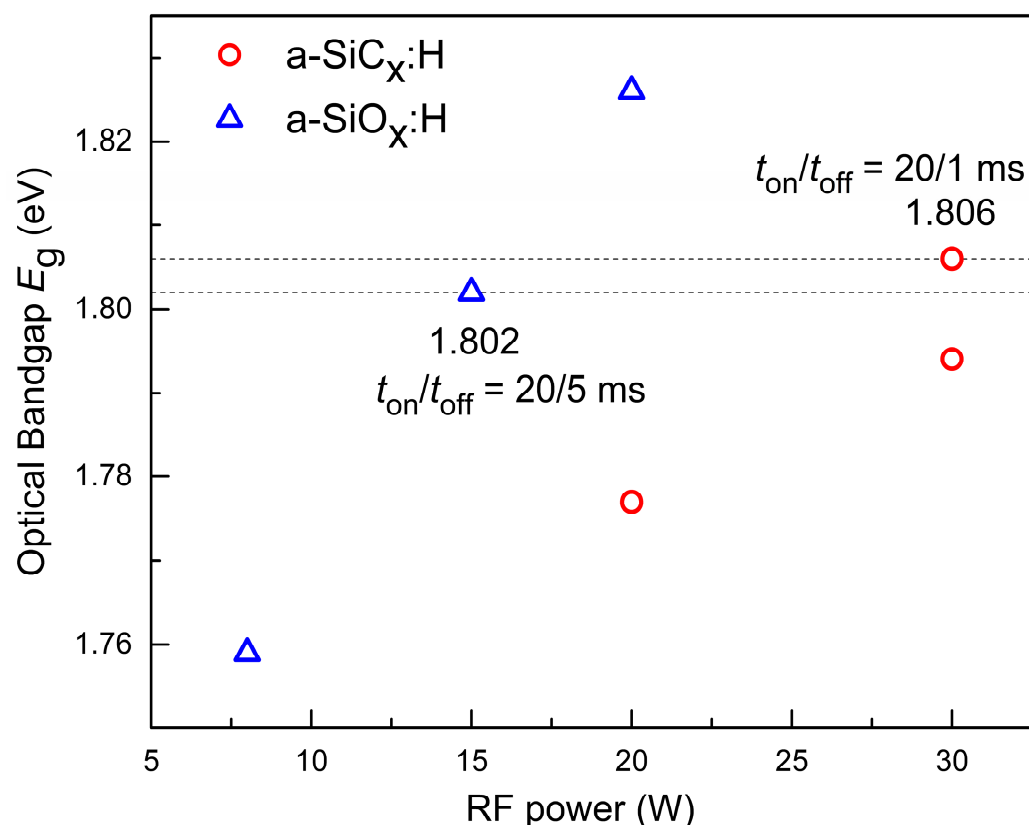


Figure 2. E_g values of a-SiO_x:H and a-SiC_x:H films fabricated using RF power and t_{on}/t_{off} ratios.

To ensure the relative relationship between the n and k of the a-SiO_x:H and a-SiC_x:H films, the Tauc–Lorentz model [18,19] was used for analysis. Figure 3 presents the full spectrum of (a) n and (b) k for a-Si:H, a-SiC_x:H, and a-SiO_x:H films from 190 to 1680 nm, along with (c) the difference in n between a-SiC_x:H and a-SiO_x:H films (Δn) divided by the n value of the reference a-Si:H film ($n_{a-Si:H}$), i.e., $\Delta n/n_{a-Si:H}$, and (d) the difference in k (Δk) divided by the k value of the reference a-Si:H film ($k_{a-Si:H}$), i.e., $\Delta k/k_{a-Si:H}$. These variations are defined in Equations (1) and (2):

$$\Delta n/n_{a-Si:H} = (n_{a-SiC_x:H} - n_{a-SiO_x:H})/n_{a-Si:H}, \quad (1)$$

$$\Delta k/k_{a-Si:H} = (k_{a-SiC_x:H} - k_{a-SiO_x:H})/k_{a-Si:H}. \quad (2)$$

Figure 3a,b show that the n and k values of the wide-bandgap a-SiC_x:H and a-SiO_x:H films were close to each other but lower than those of the a-Si:H film used in the reference cell. Figure 3c presents the $\Delta n/n_{a-Si:H}$ of the a-SiC_x:H and a-SiO_x:H films, with a very small variation error ranging from +1.4% to −0.98%. The refractive index (n) values of the a-SiC_x:H and a-SiO_x:H films were nearly identical. Figure 3d shows the $\Delta k/k_{a-Si:H}$ of the a-SiC_x:H and a-SiO_x:H films, with a variation error ranging from +1.2% to −4.1%.

Although slightly larger than $\Delta n/n_{\text{a-Si:H}}$, the values remained very small, indicating that the k values of the a-SiC_x:H and a-SiO_x:H films were also close to each other. The selected a-SiC_x:H and a-SiO_x:H films possessed similar properties in terms of E_g , n , and k . Under these conditions, the effects on the p/i and i/n interfaces could be comparatively analyzed; in particular, the n-type doping effect of the a-SiO_x:H film compared to the a-SiC_x:H film could be examined under conditions of similar optical characteristics.

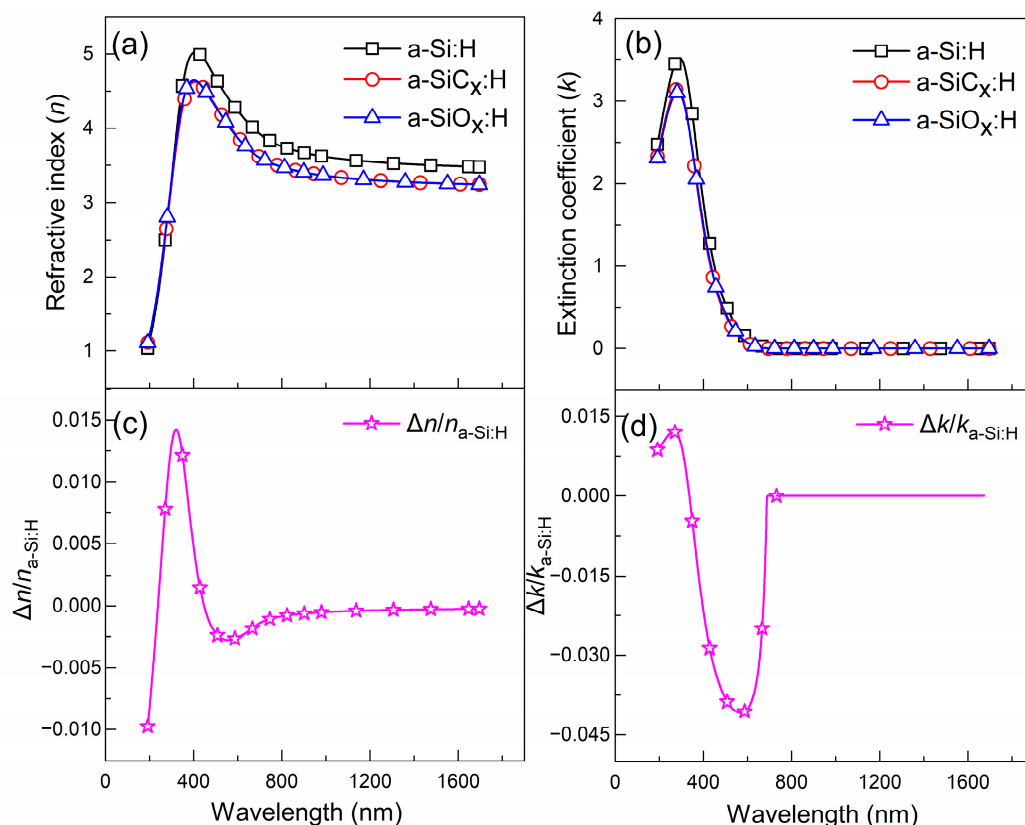


Figure 3. (a) n and (b) k of a-Si:H, a-SiC_x:H, and a-SiO_x:H films; (c) $\Delta n/n_{\text{a-Si:H}}$ and (d) $\Delta k/k_{\text{a-Si:H}}$ of a-SiC_x:H and a-SiO_x:H films.

To confirm the differences in Si-H, Si-C, and Si-O bonding between the a-SiC_x:H and a-SiO_x:H films, we examined the FTIR spectra, shown in Figure 4, of 200 nm thick a-SiC_x:H and a-SiO_x:H films for (a) the wagging and bending bands and (b) the stretching band. Table 1 summarizes the hydrogen contents C_H , the integrated area of Si-C bonds $I_{\text{Si-C}}$ or Si-O bonds $I_{\text{Si-O}}$, the integrated area of C-H₂ bonds $I_{\text{C-H}_2}$ or Si-O-Si bonds $I_{\text{Si-O-Si}}$, and the microstructural ratio (R_S) of the a-SiC_x:H and a-SiO_x:H films.

Table 1. The C_H , $I_{\text{Si-C}}$ or $I_{\text{Si-O}}$, $I_{\text{C-H}_2}$ or $I_{\text{Si-O-Si}}$ bonds, and microstructural ratio R_S of the a-SiC_x:H and a-SiO_x:H films.

	C_H (%)	$I_{\text{Si-C}}$ or $I_{\text{Si-O}}$ (cm ⁻¹)	$I_{\text{C-H}_2}$ or $I_{\text{Si-O-Si}}$ (cm ⁻¹)	R_S
a-SiC _x :H	21.18	19.7	9.33	0.57
a-SiO _x :H	21.06	49.7	330	0.60

Figure 4a shows that the a-SiC_x:H film had a larger absorbance amplitude than the a-SiO_x:H film in the Si-H wagging mode (550 to 740 nm). The integrated absorption strength, I , is calculated as follows:

$$I = \int \frac{\alpha}{\omega} d\omega, \quad (3)$$

where α is the absorption coefficient. The hydrogen concentration, N_H , is then determined using [20–22]

$$N_H = A_H \cdot I = A_H \cdot \int \frac{\alpha}{\omega} d\omega, \quad (4)$$

where A_H is the proportionality constant ($1.6 \times 10^{19} \text{ cm}^{-2}$) [21,22]. Finally, the hydrogen content of the film is defined as

$$C_H (\%) = \frac{N_H}{N_H + N_{Si}} \times 100\%, \quad (5)$$

where N_{Si} is the atomic density of silicon ($5 \times 10^{22} \text{ cm}^{-3}$) [21,22].

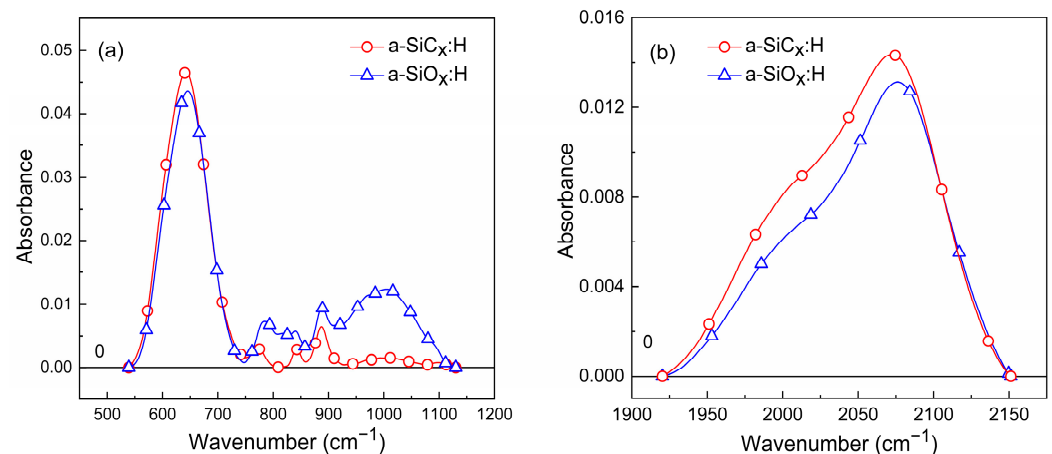


Figure 4. FTIR spectra of a-SiC_x:H and a-SiO_x:H films with a thickness of 200 nm: (a) wagging and bending modes, and (b) stretching modes.

The C_H 21.18% of the a-SiC_x:H film is slightly larger than 21.06% of the a-SiO_x:H film. The difference between the two films in the main Si-H bond intensity was not significant.

The Si-C stretching mode and C-H₂ wagging mode of the a-SiC_x:H film are primarily located at the peak positions of 770 and 1000 cm^{−1} [23]. The integrated areas of I_{Si-C} and I_{C-H_2} are 19.7 and 9.33 cm^{−1}, respectively. The Si-O bending band and Si-O-Si stretching band are located at the peak positions of 780 and 1000 cm^{−1} [20–22,24]. The integrated areas of I_{Si-O} and $I_{Si-O-Si}$ are 49.7 and 330 cm^{−1}, respectively. These results indicate that the primary difference between the two films stems from the distinction in oxygen content.

Figure 4b shows the Si-H stretching mode of the a-SiC_x:H and a-SiO_x:H films. The SiH and SiH₂ peaks of the a-SiC_x:H film are at 2004 and 2077 cm^{−1}, and the integrated areas are I_{SiH} 44.11 cm^{−1} and I_{SiH_2} 57.73 cm^{−1}. The SiH and SiH₂ peaks of the a-SiO_x:H film are at 2007 and 2080 cm^{−1}, and the integrated areas are I_{SiH} 34.69 cm^{−1} and I_{SiH_2} 52.06 cm^{−1}. R_S was estimated from Equation (6) and is 0.57 and 0.60 for the a-SiC_x:H and a-SiO_x:H films, respectively [22]. The microstructures of the a-SiC_x:H and a-SiO_x:H films are similar.

$$R_S = \frac{I_{SiH_2}}{I_{SiH} + I_{SiH_2}} \quad (6)$$

The FTIR data show that the C_H and R_S of the a-SiC_x:H and a-SiO_x:H films are similar, but the large difference stems from the oxygen content.

Figure 5 presents the J - V characteristic curves for the reference cell and the four test cells containing a-SiC_x:H and a-SiO_x:H buffer layers in the 1/5 position of the p/i and i/n interfaces. The curves cover (a) 0 to V_{oc} , (b) -0.5 to 0 V reverse bias, and (c) V_{oc} to 1.0 V forward bias ranges. The data for the short-circuit current density (J_{sc}), V_{oc} , maximum

output current density (J_m), maximum output voltage (V_m), maximum output power (P_{max}), fill factor (FF), and efficiency (η) of the five cells are listed in Table 2.

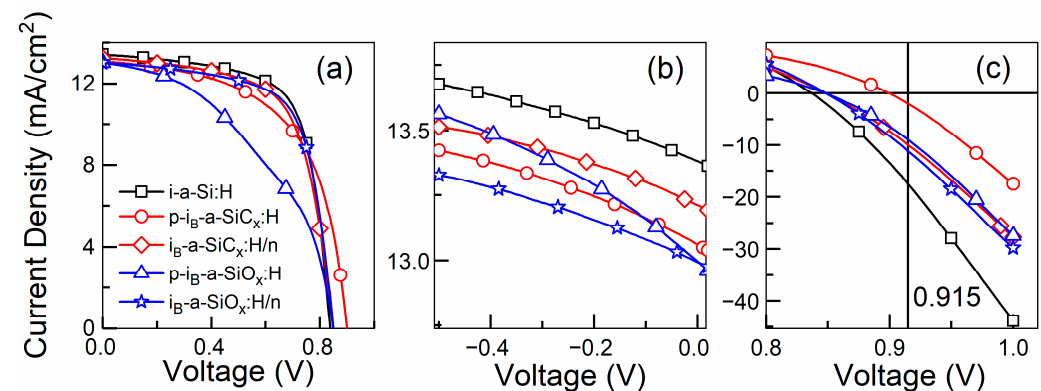


Figure 5. J - V characteristics of the five cells, including the reference cell and cells with a-SiC_x:H and a-SiO_x:H buffer layers at the 1/5 position of the p/i and i/n interfaces: (a) range from 0 to V_{oc} ; (b) reverse bias range from -0.5 to 0 V; and (c) forward bias range from V_{oc} to 1.0 V.

Table 2. The J_{sc} , V_{oc} , J_m , V_m , FF , P_{max} , and η of the five cells.

	J_{sc} (mA/cm ²)	V_{oc} (V)	J_m (mA/cm ²)	V_m (V)	FF (%)	P_{max} (mW)	η (%)
C_a-Si:H	13.38	0.8369	11.31	0.6800	68.69	0.5437	7.692
C_p/i _B -a-SiC _x :H	13.06	0.8998	9.994	0.6800	57.82	0.4804	6.796
C_p/i _B -a-SiO _x :H	13.00	0.8482	8.738	0.5550	43.99	0.3428	4.850
C_i _B -a-SiC _x :H/n	13.21	0.8486	11.02	0.6600	64.84	0.5139	7.270
C_i _B -a-SiO _x :H/n	12.99	0.8487	10.93	0.6800	67.38	0.5252	7.430

Figure 5a and Table 2 show that the reference cell C_i-a-Si:H had the highest J_{sc} of 13.38 mA/cm²; the lowest V_{oc} of 0.8369 V; the highest J_m , V_m , and FF of 11.31 mA/cm², 0.6800 V, and 68.69%, respectively; and the highest η of 7.692%. In the four test cells, J_{sc} was consistently lower than that of the reference cell C_i-a-Si:H, and V_{oc} was consistently higher. The FF values for cells with buffer layers at the p/i interface (57.82% and 43.99%) were significantly lower than those for cells with buffer layers at the i/n interface (64.84% and 67.38%, respectively). Figure 5b shows the variation in current density (J) from -0.5 to 0 V. In this range, the J of the four test cells remained lower than that of the reference cell. Notably, the C_p/i_B-a-SiO_x:H sample with a-SiO_x:H at the p/i interface showed a current that gradually increased from the second lowest (-13.00 mA/cm²) to the second highest (-13.56 mA/cm²) as the voltage changed from 0 to -0.5 V. At the same time, the magnitude relationship of the J values for the other three cells remained basically unchanged. This result indicates that when a-SiO_x:H is placed at the p/i interface, J increases rapidly with increasing reverse bias; that is, photogenerated carriers in the i-layer can continuously increase the current because the reverse bias enhances the electric field intensity in the i-layer. Figure 5c shows the variation in J from V_{oc} to 1.0 V forward bias. Within this range, the increase in the four test cells was smaller than that in the reference cell. For example, at 0.915 V, the buffer layers at the p/i interface exhibited stronger suppression of forward current than those at the i/n interface. Specifically, the C_p/i_B-a-SiC_x:H cell showed the lowest current density, which correlates with its highest V_{oc} of 0.8998 V. This demonstrates that the impact of the a-SiC_x:H and a-SiO_x:H films at the p/i interface on blocking hole injection from the p-layer into the i-layer was higher than their impact on blocking electron injection from the n-layer into the i-layer.

Placing a-SiC_x:H and a-SiO_x:H buffer layers at the p/i interface improved the V_{oc} ; the a-SiC_x:H film increased it to 0.8998 V, which is higher than the 0.8482 V of the a-SiO_x:H film. This implies that the ΔE_v in the a-SiC_x:H film was higher than that in the a-SiO_x:H film. Because this energy barrier is more difficult to cross, there was greater blocking of holes injected from the p-layer into the i-layer or of photogenerated holes flowing from the i-layer to the p-layer, resulting in a larger V_{oc} with the addition of the a-SiC_x:H film. The cell with the a-SiO_x:H p/i_B buffer layer had the lowest FF of 43.99%, while the cell with the i_B/n buffer layer had the second-highest FF (67.38%). This is distinct from the cell with the a-SiC_x:H p/i buffer layer, which had a slightly higher FF (57.82%), while the cell with the i/n buffer layer had a lower FF (64.84%). It is worth noting that while oxygen incorporation in a-SiO_x:H induces an n-type doping effect, creating a positive space charge (O_3^+), carbon incorporation in a-SiC_x:H primarily widens the bandgap and increases structural disorder without acting as a donor. This structural distinction allows a-SiC_x:H to provide a high ΔE_v at the p/i interface without collapsing the bulk electric field, unlike the n-type a-SiO_x:H.

The significant decrease in FF caused by the a-SiO_x:H film had a low correlation with ΔE_v . Instead, this decrease was due to the micro n-type doping of oxygen atoms in the a-SiO_x:H film, which caused misalignment and a reduction in the built-in electric field in the subsequent i-layer, leading to a significant reduction in overall FF and η when the a-SiO_x:H film was added to the p/i buffer layer [5,6,10,12]. This is also reflected in the fact that when the a-SiO_x:H film was moved to the i/n interface, the n-type doping enhanced the built-in electric field of the entire i-layer, resulting in the a-SiO_x:H film at the i/n interface achieving the second-highest FF (67.38%), only slightly lower than that of the a-Si:H reference cell [14,15].

The effects of high-bandgap a-SiC_x:H and a-SiO_x:H thin films at the p/i and i/n interfaces were further validated using a second set of films with higher E_g : a-SiC_x:H (E_g : 1.820 eV) and a-SiO_x:H (E_g : 1.826 eV). These films were deposited with a CH₄ flow rate of 5.6 sccm, RF power of 20 W, and t_{on}/t_{off} of 20/5, and a CO₂ flow rate of 2.5 sccm, RF power of 20 W, and t_{on}/t_{off} of 20/5, respectively. Figure 6 compares the J - V parameters (J_{sc} , V_{oc} , FF , and η) of the original set (a-SiC_x:H, E_g : 1.806 eV; a-SiO_x:H, E_g : 1.802 eV) with those of the second set. Table 3 lists the J_{sc} , V_{oc} , FF , and η results for the second set.

Table 3. The J_{sc} , V_{oc} , FF , and η for the second set of samples, including a-SiC_x:H (E_g : 1.820 eV) and a-SiO_x:H (E_g : 1.826 eV).

	J_{sc} (mA/cm ²)	V_{oc} (V)	FF (%)	η (%)
C_a-Si:H	13.37	0.8366	68.25	7.635
C_p/i _B -a-SiC _x :H	12.55	0.8874	45.10	5.021
C_p/i _B -a-SiO _x :H	12.52	0.8468	39.73	4.210
C_i _B -a-SiC _x :H/n	12.70	0.8333	56.13	5.939
C_i _B -a-SiO _x :H/n	12.27	0.8377	65.50	6.732

Figure 6a,c show that across the four test cells, J_{sc} was consistently lower than that of the reference cell (C_i-a-Si:H), while V_{oc} was, in general, higher. Placing a-SiC_x:H and a-SiO_x:H buffer layers at the p/i interface improved V_{oc} . The a-SiC_x:H film increased this to 0.8998 V (original) and 0.8874 V (second set), which is higher than the 0.8482 V (original) and 0.8468 V (second set) obtained with the a-SiO_x:H film. As shown in Figure 6b,d, both sets of samples showed that placing a-SiO_x:H at the p/i interface yields the lowest FF (original: 43.99% and second set: 39.73%) and η (original: 4.850% and second set: 4.210%) compared to a-SiC_x:H's FF (original: 57.82% and second set: 45.10%) and η (original: 6.796% and second set: 5.021%). Conversely, when a-SiO_x:H is placed at the i/n interface, the FF (original: 67.38% and second set: 65.50%) and η (original: 7.430% and second set: 6.732%)

significantly improved, surpassing those of a-SiC_x:H's *FF* (original: 64.84% and second set: 56.13%) and η (original: 7.270% and second set: 5.939%). These results confirm that the oxygen-doping effect of a-SiO_x:H thin films decreases the internal electric field in the rear i-region when placed at the p/i interface, leading to lower *FF* and η compared to the a-SiC_x:H film. In contrast, at the i/n interface, this effect enhances the electric field in the front i-region, resulting in higher *FF* and η compared to the a-SiC_x:H film.

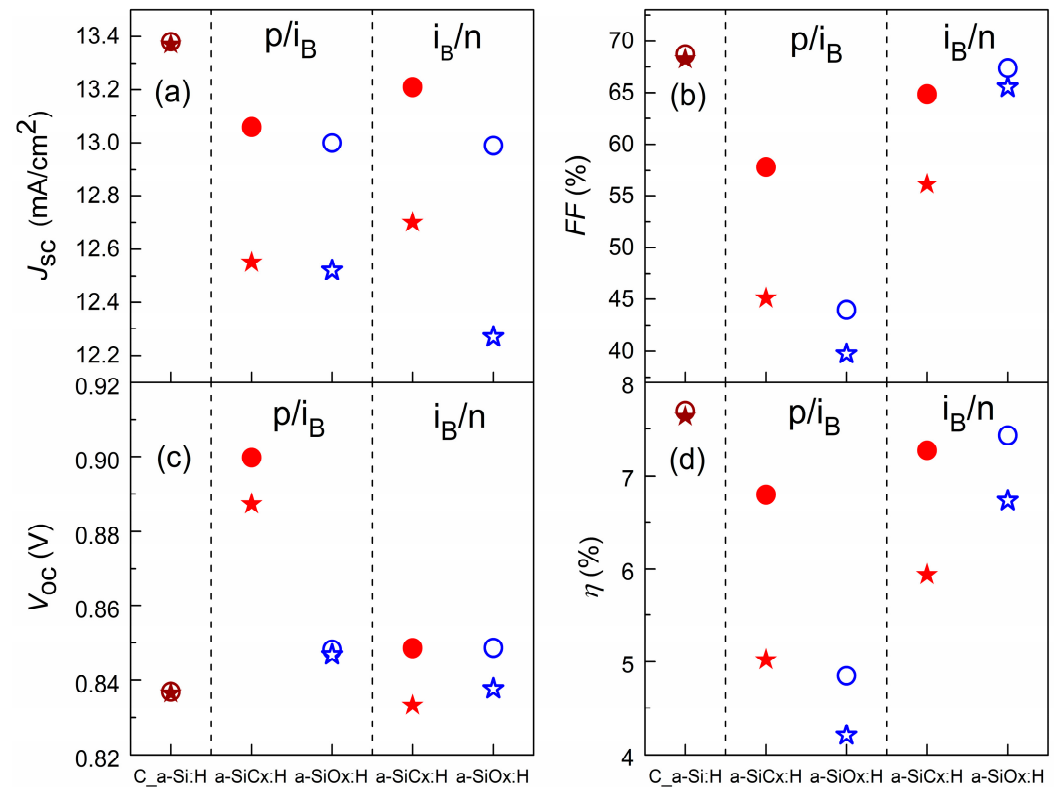


Figure 6. Comparison of (a) J_{sc} ; (b) *FF*; (c) V_{oc} ; and (d) η between the original set and the second set. The original set includes a-SiC_x:H (E_g : 1.806 eV, red solid circles) and a-SiO_x:H (E_g : 1.802 eV, blue hollow circles), while the second set includes a-SiC_x:H (E_g : 1.820 eV, red solid stars) and a-SiO_x:H (E_g : 1.826 eV, blue hollow stars).

The higher E_g of the a-SiC_x:H (E_g : 1.820 eV) and a-SiO_x:H (E_g : 1.826 eV) films resulted in reduced film quality. Consequently, all parameters (J_{sc} , V_{oc} , *FF*, and η) decreased due to high recombination rates as carriers passed through the high-density defects within the a-SiC_x:H or a-SiO_x:H films.

The experimental results above indicate that a-SiC_x:H at the p/i interface significantly increases the V_{oc} value due to a higher ΔE_v effect. a-SiC_x:H and a-SiO_x:H films at the p/i interface have a greater attenuation effect on the fill factor. The electric field is too concentrated between the p-layer and these two buffer layers, weakening it in the subsequent i-layer and causing a significant drop in *FF*; this effect is more pronounced for the a-SiO_x:H film due to its slight n-type doping. This is a substantial distinction in the application of a-SiO_x:H and a-SiC_x:H films. Regardless of placement at the p/i or i/n interface, the open-circuit voltage is higher, and the short-circuit current is lower than that of the reference cell, with band mismatches ΔE_v and ΔE_c contributing. Part of the J_{sc} decrease at the i/n interface is also due to reduced generation of photogenerated carriers from medium- or long-wavelength photons.

4. Conclusions

In this study, a-SiC_x:H and a-SiO_x:H films with approximately identical E_g were fabricated using PWM plasma and placed at the p/i and i/n interfaces. Because the E_g values were similar, the influence of n-type doping in the a-SiO_x:H film could be inferred. The experimental results showed that n-type doping in the a-SiO_x:H film significantly reduced the electric field intensity in the i-layer at the p/i interface while increasing it at the i/n interface. This manifested in a significant reduction in FF at the p/i interface and a substantial increase in FF at the i/n interface. The ΔE_c and ΔE_v of the two films contributed to the increase in the cells' V_{oc} . Appropriately placing the a-SiC_x:H film at the p/i interface and the a-SiO_x:H film at the i/n interface is the optimal strategy. Specifically, the a-SiC_x:H p/i buffer can maximize the V_{oc} (up to ~0.90 V), while the a-SiO_x:H i/n buffer can maintain a high η (7.43%) and FF (67.38%), with values comparable to those of the reference cell.

Future work will focus on optimizing the thickness and doping concentration of the a-SiO_x:H rear buffer layer and exploring its application in tandem solar cell structures.

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