1 **Phosphorus- and boron-doped hydrogenated amorphous silicon films prepared**
2 **using vaporized liquid cyclopentasilane**

3
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11
12 **Abstract**

13 A simple, inexpensive method for fabricating a hydrogenated amorphous silicon
14 (a-Si:H) film using thermal chemical vapor deposition from cyclopentasilane (CPS) at
15 atmospheric pressure with a substrate temperature of 370°C is described. The reactant
16 gas was generated from liquid CPS by vaporization in the deposition chamber. The
17 vaporized CPS gas was transformed immediately into a-Si:H film on a heated substrate.
18 The a-Si:H films could be doped either n- or p-type by dissolving appropriate amounts
19 of white phosphorus or decaborane, respectively, in the liquid CPS before vaporization.
20 This process allows deposition of doped a-Si:H films of photovoltaic device-quality
21 without the need for handling, storage, or transportation of large amounts of gaseous
22 reactants.

23
24 **Keywords:** CPS, cyclopentasilane, polysilane, amorphous silicon, solar cell.
25

1. Introduction

Hydrogenated amorphous silicon (a-Si:H) films have received considerable interest because of their immense potential for various applications in large-area electronic devices, such as solar cells and thin-film transistors [1,2]. A number of detailed investigations have been carried out on the fabrication and characterization of a-Si:H films made using a variety of chemical vapor deposition (CVD) techniques [3]. In particular, device quality a-Si:H films have been deposited by plasma discharge of monosilane using plasma-enhanced (PE) CVD [4].

a-Si:H films can also be produced by pyrolytic decomposition of SiH_4 gas using thermal CVD [5]. However, due to the high temperatures (500–600°C) required to achieve an appreciable deposition rate, hydrogen bonding in the resulting films are limited; consequently, the defect density for such films is much larger than for films obtained by PECVD. To use low-cost substrates, the thermal CVD deposition temperature has to be lowered. Process temperatures as low as 300–400°C have been achieved utilizing higher silanes (linear silanes: $\text{Si}_n\text{H}_{2n+2}$, $n > 2$) as reactant gasses, because higher silanes are thermodynamically and kinetically less stable than SiH_4 [6]. For these cases, the temperatures required for thermal CVD are still higher than those for PECVD (~200°C), but thermal CVD production of a-Si:H offers a much larger

throughput at a greatly reduced cost.

In this paper, we report on an inexpensive, practical, simple method for fabricating a-Si:H from thermal CVD of higher silane in nitrogen gas at atmospheric pressure. Cyclopentasilane (CPS) was used as the higher silane because of its easy storage and handling, higher stability than linear silanes, and compatibility with non-vacuum processes. CPS has a boiling point of 195°C and stays in the liquid state at ordinary temperatures and pressures. CPS and its derived polymer was originally synthesized as a solution precursor for semiconductor Si to fabricate solution-processed electronics in our previous reports [7,8]. However, liquid materials can be applied to not only solution-process but also vapor-process. CPS vapor is generated by vaporization of liquid CPS in a heated CVD chamber. CPS as a gas source does not require gas bombs or carrier tubes for its containment and usage because of its liquid state. Furthermore, Si gas can be locally generated anywhere in a deposition chamber through placement of the liquid CPS. This versatility, combined with the potential for non-vacuum processing, gives us large degrees of freedom in equipment design, with the potential for a simple, versatile non-traditional CVD system.

The deposition equipment used in this study was much simpler than in typical CVD systems. In addition, we report a method for phosphorus and boron doping, which

eliminates the hazards of phosphine and diborane gas by replacing them with white phosphorus and decaborane, respectively, dissolved in liquid CPS.

Since little is known about the thermal CVD of cyclic silane and the performance of the method, we analyzed the basic properties of resultant a-Si:H films. In particular, we focused on p-type a-Si:H (p-Si) and n-type a-Si:H (n-Si) films. We also fabricated an a-Si:H solar cell using doped a-Si:H. The properties of i-type a-Si:H (i-Si) films and details of the deposition mechanisms was discussed in another report [9].

2. Experiment

Fig. 1 shows a schematic representation of the apparatus. The deposition chamber consisted of a petri dish with a diameter of 80 mm and a depth of 1 mm. CPS was placed in the corners of the petri dish. Then, the petri dish was put on a substrate which was heated on a hot plate at 370°C. Since the petri dish was readily heated by the hot plate, the CPS vaporized and filled the volume covered by the petri dish. Vaporized CPS decomposed and transformed immediately into a-Si:H on the substrate surface. The film thickness was controlled by the amount of CPS in the petri dish. Deposition time was fixed at 3 min. Si wafer (CZ crystal) substrates with high resistivity ($>1000 \Omega\text{cm}$) were used for Fourier-transform infrared spectroscopy (FTIR) measurements, and non-alkali

glass substrates (OA-10, Nippon Electric Glass) were used for optical and electrical characterization.

CPS was synthesized according to the procedure of Hengge et al. [10,11] and was used for deposition of i-Si film. Decaborane (Taiyo Nippon Sanso) purified by sublimation and white phosphorus obtained by the cracking of red phosphorus (Sigma-Aldrich) were used as p- and n-type dopant material, respectively. Decaborane and white phosphorus were dissolved in CPS at concentrations up to 1 and 3 wt%, respectively, and were used for deposition of doped a-Si:H films. All the procedures were conducted in a nitrogen glove box with oxygen content and dew point less than 0.5 ppm and -75°C , respectively.

The vaporization characteristics of CPS are important, because CPS vapor was used as a gas source in this study. Therefore, the thermal properties of liquid CPS were examined using thermogravimetry (TG)/differential thermal analysis (DTA) (EXTAR 6200, Seiko Instruments). Approximately 10 mg of CPS was added into the sample pan. To prevent oxidation as much as possible during the measurement, the purge gas was purified using a gas purifier (Pureron Japan) to obtain oxygen concentrations <1 ppb, and a flow rate of 200 mL/min was maintained. CPS was heated from 25°C up to 450°C with a heating rate of $20^{\circ}\text{C}/\text{min}$.

The structures of the resultant a-Si:H films were characterized by FTIR (ALPHA, Bruker Optics) in the transmission mode. The binding energy of B 1s was measured by X-ray photoelectron spectroscopy (XPS). The hydrogen, boron, and phosphorus contents were quantified by secondary ion mass spectroscopy (SIMS). The optical gap E_g of the films were estimated using a Tauc plot of data obtained from transmittance and reflectance spectroscopy in the UV–visible range (FilmTek 3000, Scientific Computing International). The film density was measured by X-ray reflection (XRR). The dark conductivity σ at room temperature and its temperature dependence were measured with Al contacts in a coplanar configuration with a 250- μ m gap width. An AM-1.5G solar simulator (WXS-50S, WACOM Electric) with an intensity of 100 mW/cm² was used for illuminated current and voltage measurements.

We also tested our a-Si:H films in a-Si:H solar cells. An Al-doped ZnO (AZO) film with a sheet resistivity of 50 Ω /sq and a thickness of 200 nm was sputtered onto a flat glass substrate (OA-10). Then, a-Si:H films were deposited by Hot-wire (HW) CVD with order n-i-p or p-i-n with respective layer thicknesses of 20, 100, and 20 nm. In processing, particular layers in some of these solar cells were instead prepared by thermal CVD of liquid CPS, as summarized in Table 1. 4 μ l of CPS was used for the thermal CVD source. Next indium tin oxide (ITO) of 100 nm thickness and an area of 1

cm $\times$ 1 cm was sputtered onto the n-i-p and p-i-n structures. Finally, some of the Si film was removed by CF₄ etching to allow for direct electrical contact to the AZO electrode layer. The current–voltage (J – V) characteristics of the cells deposited on glass substrate (OA-10) were measured under 100 mW/cm² illumination.

The HWCVD used was nearly identical to that described in the literature [12]. The apparatus was made of stainless steel with a width of 40 cm, depth of 40 cm, and height of 35 cm. A tungsten wire with a diameter of 0.5 mm and a length of 200 cm was used as the catalyzer. The wire was placed between the substrate holder and gas inlet inside the chamber. The distance between the substrate holder and the wire was kept at 8 cm. After putting in the substrate and evacuating the chamber, the wire was heated to 1850°C by resistive heating. The temperature of the wire was estimated based on the temperature dependence of electric resistivity of the wire. The substrate was also heated to 300°C by a heater mounted at the back of the holder. Reactant gas was introduced into the chamber from the opposite side of the substrate holder across the tungsten wire. The flow rate of the gas was adjusted using flow meters. The flow rate conditions are listed in Table 2. The values of E_g , σ , and photoconductivity σ_p for the resultant films are also listed.

3. Results and discussion

3.1 CPS analysis and deposition process

Thermal analysis of liquid CPS illuminated the nature of the deposition process in the chamber. Fig. 2(a) shows the TG signal (bold solid line) and DTA signal (thin solid line) as a function of the heating temperature. Also shown is the temperature/time derivative DTG (dashed line) of the TG signal.

A broad endothermic peak in the DTA signal extending from room temperature (RT) to higher temperatures was caused by the evaporation of CPS. The TG data indicated a weight loss of approximately 90% with heating up to 360°C; consequently, only 10% of the material remained in the form of a-Si:H after pyrolysis of CPS. A major peak appeared at 180°C in the DTA and DTG signals, corresponding to the boiling point of CPS (194°C [13]). Below 200°C, 70% of the mass readily evaporated. Thermogravimetric-mass spectrometry clarified that this component can be attributed to CPS molecules. Remaining 20% of the mass gradually evaporated between 200°C and 360°C, which we attribute to thermally polymerized silicon compounds. The polymerized silicon has a higher boiling point than CPS. The polymerized silicon converts to a-Si:H as they break down into silane-based gases by heating, which reduces their weight by two-thirds. The observed thermal characteristics are consistent with the

behavior of a polymer derived from CPS [14]. Fig. 2(b) shows the TG and DTG signals for the dopant materials. The DTG peaks correspond to the boiling point of white phosphorus (213°C) and decaborane (280°C) [15].

Fig. 3 shows the time dependence of the film thickness on the substrate and the temperature of the petri dish during the deposition, as shown in Fig. 1. The temperature of the petri dish with the CPS heated up to 300°C within 2 min, and then the temperature stayed constant. The relative thickness, which is the film thickness normalized to the final thickness, increased rapidly within 2 min and then stayed constant. Based on Fig. 2, it is clear that almost all the CPS and dopant materials were vaporized in the petri dish at a temperature below 300°C. Therefore, when the petri dish was set on the **heated substrate**, the CPS evaporated quickly and transformed into an a-Si:H film on the substrate, with incorporation of the dopant material. After 2 min, the Si deposition stopped, because the CPS was gone.

The thermal decomposition of silane compound initiates with formation of a silylene-type radical [16]. The generated radical reacts with CPS by inserting one into Si-H bonds in CPS to form a larger polysilane. This new polysilane then decomposes again. In the course of the silylene generation, hydrogen is eliminated from the CPS and polysilane. Thus the polysilane molecules are continually being broken up and enlarged

on hot surface to form a-Si:H.

3.2 Film properties

Electrical properties

The a-Si:H deposited by thermal CVD from CPS could be doped either n- or p-type by dissolving appropriate amounts of white phosphorus or decaborane, respectively, in the liquid CPS before vaporization. Fig. 4(a) shows a summary of the properties of doped a-Si:H films. On the right-hand side of the graph, σ at room temperature is plotted against phosphorus concentration in Si films. The effect of boron doping is shown on the left-hand side on the plot. The dopant concentration was measured by SIMS (model 6300, PHI). The values of σ for an i-Si film are also given in the center of the graph. The σ for n- and p-Si films showed a significant increase with dopant concentration, indicating that white phosphorus and decaborane were effective dopant materials in the deposition system.

The following expression to describe the dependence of σ on temperature T is generally used:

$$\sigma = \sigma_0 \exp (-E_{act}/kT) \quad (1)$$

where σ_0 is the conductivity prefactor, E_{act} is $(E_C - E_F)_0$ for n-type materials, the energy difference between the conduction band energy E_C and the Fermi level E_F at zero

temperature. In the case of p-Si, E_{act} is $(E_F - E_V)_0$, where E_V is the valence band energy. The values of E_{act} are shown in Fig. 4(b). They achieved 0.15 and 0.28 eV for n-Si and p-Si, respectively, without consideration of the effect of statistical shift [17].

Fig. 4(c) shows σ_0 ; its value mostly decreased with dopant concentration. This behavior is due to a gradual transition from extended-state conduction to hopping conduction. For higher E_{act} , σ_0 showed high mobility, and σ_0 decreased as E_{act} decreased. Our conductivity data followed Meyer–Neldel behavior, as described by Fritzsche [18], in which $\log \sigma_0$ is a linear function of E_{act} . This behavior has been interpreted in an extended-state conduction model by Spear et al. [19] as a shifting of the conduction band mobility edge for $0.6 < E_{act} < 1.0$ eV and $10^2 < \sigma_0 < 10^7$ S/cm. In Spear’s model, the conduction mechanism is hopping for σ_0 below the minimum metallic conductivity ($\sim 10^2$ S/cm), and conduction is thermally activated for σ_0 above the minimum metallic conductivity.

The dashed line in Fig. 4 (c) is the minimum metallic conductivity calculated by Mott and Davis [20]. Our lightly doped films exhibited a high E_{act} and extended-state conductivity, as evidenced by $E_{act} > 0.6$ eV and σ_0 values well above the minimum metallic conductivity. In contrast, highly doped films had σ_0 near or below the minimum metallic conductivity and $E_{act} < 0.6$ eV, as is commonly observed [18]. The conductivity

mechanism in this range probably involves hopping and may be complex, but the changes in σ_0 and E_{act} indicate that in the high-doping regime, decaborane and white phosphorus dope the material in the same sense as does diborane and phosphine, respectively [18,21].

The electrical and optical properties of the doped a-Si:H films with the highest dopant concentrations are summarized in Table 3. As a reference, the values of i-Si prepared by same method are also listed. The optical gap E_g was estimated from the relationship between $(\alpha E)^{1/2}$ and E , i.e., a Tauc plot, in which α is the optical absorption coefficient and E is the photon energy in eV [22]. A film with thickness 80–100 nm deposited on a glass substrate was used for the measurement. The narrowing of E_g in high-boron-doped films is consistent with other results in the literature [23].

The overall effects of doping were to increase σ of a-Si:H by 7–8 orders of magnitude and to move the Fermi level continuously to within 0.15–0.28 eV of the extended states. Although we used decaborane and white phosphorus instead of diborane and phosphine, respectively, the resultant values for σ and E_{act} are comparable to those of the first-ever doped a-Si:H film [24]. Moreover, our high-doped film exhibited small σ_0 near or below the minimum metallic conductivity, indicating that the decaborane and white phosphorus can be used as dopant materials. In addition, the small E_{act} provides a

small barrier at the p-i or n-i interface, making it a possible material for standard p-i-n or n-i-p solar cells. Use of solid decaborane and white phosphorus can eliminate the hazards of diborane and phosphine gas, which are conventional CVD materials.

FTIR and XPS measurements

Fig. 5 shows the infrared absorption spectra obtained by FTIR for p-, n-, and i-Si films. The dopant concentrations of the p- and n-Si films were 3.3×10^{21} and 8.5×10^{20} cm^{-3} , respectively. We assigned the absorption peaks near 2000 and 2070–2090 cm^{-1} to the stretching modes of Si–H bonds embedded in the bulk matrix and Si–H/Si–H₂ bonds at void surfaces, respectively [25]. The microstructure parameter R , defined as $R = I_{2090}/(I_{2000} + I_{2090})$, where I_{2090} and I_{2000} are integrated intensities of each peak at 2070–2090 and 2000 cm^{-1} , respectively, is an indicator of film quality [26]. The domination of the 2070–2090 cm^{-1} peak compared to the 2000 cm^{-1} peak for all of the a-Si:H films means a large value for R of 0.61, 0.58, and 0.53 for the p-, n-, and i-Si films, respectively. These are much larger than the $R < 0.2$ for device-grade a-Si:H films prepared by plasma discharge of monosilane, which show a dominant 2000 cm^{-1} peak [27]. This result suggests the presence of voids in the films. Indeed, the film density of our a-Si:H films, as measured by XRR (X'Pert PRO MRD, PANalytical), was 2.1 g/cm^3 ,

while that of conventionally made a-Si:H film is 2.2 g/cm³ [3].

The densities of bonded hydrogen N_H in the a-Si:H films were evaluated on the basis of the integrated intensity of the Si–H wagging mode absorption at 640 cm⁻¹ using Eq. (2), with proportionality coefficient $A_{640} = 2.1 \times 10^{19}$ cm⁻² [28,29]:

$$N_H = A_i \int \frac{\alpha(\omega)}{\omega} d\omega. \quad (2)$$

The hydrogen contents, as measured by N_H/N_{Si} , in the p-, n-, and i-Si films were 6.1, 4.0, and 6.7 at%, respectively, assuming that N_{Si} , the density of Si, is 5×10^{22} cm⁻³. Since the content of hydrogen compared to silicon in CPS (Si₅H₁₀) is 200 at%, the hydrogen content of 4.0–6.7 at% in the films indicates that dehydrogenation occurred during pyrolysis in the film deposition process.

The hydrogen content of the films, as measured by FTIR and SIMS, are listed in Table 3. Only Si-bonded hydrogen shows up in the FTIR spectra, while all hydrogen in the film is measured with SIMS. The hydrogen content in the n-Si films measured by FTIR and SIMS was lower than that in i-Si films, indicating a low density of Si–H bonds in the n-Si films. The difference in hydrogen content between the n-Si and i-Si films, which is estimated to be $1.4\text{--}1.8 \times 10^{21}$ cm⁻³, is 1.6–2 times greater than the phosphorus concentration (8.5×10^{20} cm⁻³). Therefore, white phosphorus doping has the effect of removing hydrogen atoms in the Si network. It is known that the free

energy of desorption of hydrogen depends on the Fermi level and decreases when the Fermi level shifts towards the extended states [30,31]. This doping dependence of hydrogen desorption energies has been associated with simultaneous electronic transition during rupture of Si–H bonds. We suggest that our n-Si films allow this mechanism, because the Fermi level shifted with white phosphorus doping, as is the case for phosphine doping in PECVD a-Si:H [24].

As for the decaborane-doped sample, the discrepancy between the hydrogen content measured by FTIR (6.1 at%) and by SIMS (10.4 at%) for the p-Si film is due to the existence of B–H bonding, as evidenced by a B–H stretching mode at 2520 cm^{-1} in the p-Si spectrum [32]. From the difference in hydrogen content (4.3 at%), the density of B–H bonds is estimated at $2.2 \times 10^{21}\text{ cm}^{-3}$. It corresponding to the removal of six hydrogen atoms from decaborane. In the pyrolysis of decaborane, it is known that the hydrogen is removed at temperatures greater than 200°C , and non-volatile high-molecular-weight boron hydrides remain [33]. In the case of a mixed decaborane and carbon compound, decaborane readily reacts with carbon at $170\text{--}238^{\circ}\text{C}$, and a large uptake of carbon is observed [34]. Although the reaction mechanism between decaborane and CPS is still unclear, it is reasonable to suppose that decaborane decomposes by heating and reacts with CPS in vapor or on the hot substrate, releasing

its hydrogen during deposition.

Fig. 6 shows the B 1s spectrum measured by XPS (Kratos AXIS ULTRA DLD, Shimadzu) in the binding energy range from 183 to 194 eV for the p-Si film. The B 1s spectrum can be resolved into three peaks at 190.0, 188.0, and 187.2 eV. We attribute the peaks at 190.0 and 188.0 eV to deactivated boron related to B–B and B–H bonds in decaborane. These results can be compared to those of Tsutsui et al., who assigned peaks at 189.6 and 188.3 eV to deactivated boron related to boron clusters and the peak at 187.1 eV to electrically activated boron in a high-doped silicon substrate [35]. The difference in peak position between our results and those of Tsutsui et al. is presumably due to the experimental differences in silicon structure (amorphous vs crystalline), boron concentration, dopant materials, and process conditions.

The percentage of electrically activated boron was 20%, which was estimated using the integrated intensities of each peak. The presence of deactivated boron related to B–H and B–B bonds indicates that not all the hydrogen in decaborane was released, and high-molecular-weight boron hydrides remained in the film at less than 370°C. Unusual thermal stability has been attributed to B–B bridged structures in decaborane. Such species containing more than several boron atoms are highly unlikely to be electrically active substitutional dopants in a-Si:H. Nevertheless, decaborane acts as an

acceptor, and the resultant p-Si shows usable electrical properties, as shown in Fig. 4. Ionization of conventionally used boron compounds, like diborane, trimethyl boron, and boron trifluoride, which result in species containing one or two boron atoms, principally, and higher order reaction required to produce species containing more than several boron atoms have a very low probability of occurring when the boron compound is used as a dopant material. Therefore, the different decomposition/incorporation mechanism of decaborane and other conventionally used boron compound is implied here. This will be investigated in a future study.

3.3 Cell properties

Doped a-Si:H films deposited by thermal CVD from liquid CPS were used as solar cell layers. The deposition methods for the separate film layers are shown in Table 1. All the Si films were deposited by HWCVD in cells 1 and 3. The n-Si layer in the n-i-p structure and the p-Si layer in the p-i-n structure were deposited by thermal CVD in cells 2 and 4, respectively. Properties of the films used are shown in Tables 2 and 3. Figs. 7(a) and (b) show the J - V characteristics of the Glass/AZO/n-i-p/ITO and Glass/AZO/p-i-n/ITO solar cells, respectively. The photovoltaic parameters, such as open-circuit voltage (V_{oc}), short-circuit current (J_{sc}), fill factor (FF), and energy

conversion efficiency (η) of the cells, are also shown in Fig. 7.

Comparing the curves of cells 1 and 2 and the curves of cells 3 and 4, we can see they are very similar, indicating the doped a-Si:H films prepared by thermal CVD were comparable to those prepared by HWCVD, in terms of forming an operational solar cell. E_g in the thermal-CVD-grown p-Si layer was 0.20 eV narrower than that in the p-Si grown by HWCVD. Cells 3' and 4' correspond to cells 3 and 4 with light incident from the p-Si side. A comparison of cells 3 and 4 with cells 3' and 4' indicates that the thermal-CVD-grown p-Si is undesirable as a window material because of its narrow E_g . J_{sc} for cell 4' was 68% of that for cell 4, because more incident light was absorbed in the window layer of the thermal-CVD-grown p-Si. Therefore, we are currently investigating amorphous silicon carbide films as wide-band-gap p-type materials using the same growth process in another report [36].

4. Conclusion

We deposited doped a-Si:H films by thermal CVD from liquid CPS at a temperature of 370°C in nitrogen gas at atmospheric pressure. We also clarified that it is easy to dope the a-Si:H film either n- or p-type by dissolving appropriate amounts of white phosphorus or decaborane, respectively, in the liquid CPS before vaporization.

From a practical and economic point of view, our thermal CVD process has the following advantages: (1) low capital outlay, since a hot plate and a small petri dish can be used as a deposition chamber and no expensive vacuum nor electrical equipment is needed; (2) simplicity; (3) no storage of large amounts of hazardous gases, since CPS, white phosphorus, and decaborane are in liquid or solid states; (4) small volumes of starting materials are needed, since the liquid material expands in volume by several orders of magnitude when it evaporates. These advantages suggest thermal CVD of liquid CPS can be used for much simpler processing of Si devices.

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Fig. 1. Schematic drawing of the deposition system.

Fig. 2. TG (bold solid line), DTG (dashed line), and DTA (thin solid line) curves for (a) CPS, (b) decaborane (red) and white phosphorus (blue).

Fig. 3. Petri dish temperature and Si film thickness during the deposition process.

Fig. 4. The effect of phosphorus and boron doping on the (a) dark conductivity at 25°C, (b) activation energy, and (c) conductivity prefactor in a-Si:H films. The dashed line in (c) is the minimum metallic conductivity from ref. [16]. On the right side of each graph, these quantities are plotted against the phosphorus concentration, and on the left side, they are plotted against the boron concentration. The dopant concentrations in the films were measured by SIMS.

Fig. 5. FTIR spectra of the p-, n-, and i-Si films.

Fig. 6. B 1s spectrum for the p-Si film with a boron concentration of $3.3 \times 10^{21} \text{ cm}^{-3}$. Dotted and thin lines are experimental and fitted data, respectively.

Fig. 7. J - V characteristics of the solar cells with 100 mW/cm^2 illumination. The photovoltaic parameters for the cells are listed on the right-hand side. (a) a-Si:H films fabricated in order n-i-p, and the incident light was from the p-Si side. All the Si was fabricated by HWCVD in cell 1, and only the n-Si layer was fabricated by thermal CVD in cell 2. (b) a-Si:H films fabricated in order p-i-n. All the Si was fabricated by HWCVD in cell 3, and only the p-Si layer was fabricated by thermal CVD in cell 4. The incident light was from the n-Si side for cells 3 and 4, and cells 3' and 4' denote illumination from the p-Si side.

Table 1. Stacking order and deposition method for each layer in a-Si:H solar cells. The films were deposited on a glass substrate.

Table 2. Flow rate of reactant gases in HWCVD and optical gap E_g , dark conductivity σ , and photoconductivity σ_p for the resultant a-Si:H.

Table 3. Film properties of a-Si:H. C_H is hydrogen content measured by FTIR and SIMS. E_g is the optical gap estimated from Tauc plots. E_{act} and σ are activation energy and dark conductivity, respectively. The dopant concentrations for p- and n-Si were 3.3×10^{21}

466 and $8.5 \times 10^{20} \text{ cm}^{-3}$, respectively.

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Table 1. Stacking order and deposition method for each layer in a-Si:H solar cells. The films were deposited on a glass substrate.

Sample	AZO	n-Si	i-Si	p-Si	ITO
Cell 1	Sputter	HWCVD	HWCVD	HWCVD	Sputter
Cell 2		Thermal-CVD	HWCVD	HWCVD	
Sample	AZO	p-Si	i-Si	n-Si	ITO
Cell 3	Sputter	HWCVD	HWCVD	HWCVD	Sputter
Cell 4		Thermal-CVD	HWCVD	HWCVD	

Table 2. Flow rate of reactant gases in HWCVD and optical gap E_g , dark conductivity σ , and photoconductivity σ_p for the resultant a-Si:H.

	Gas flow (sccm)				E_g (eV)	σ (S/cm)	σ_p (S/cm)
	H ₂	SiH ₄	B ₂ H ₆	PH ₃			
p-Si	-	10	5	-	1.76	6.8×10^{-6}	-
n-Si	-	10	-	20	1.61	2.9×10^{-3}	-
i-Si	10	10	-	-	1.79	1.3×10^{-11}	1.3×10^{-5}

Table 3. Film properties of a-Si:H. C_H is hydrogen content measured by FTIR and SIMS. E_g is the optical gap estimated from Tauc plots. E_{act} and σ are activation energy and dark conductivity, respectively. The dopant concentrations for p- and n-Si were 3.3×10^{21} and $8.5 \times 10^{20} \text{ cm}^{-3}$, respectively.

	$C_H(\text{at.}\%)$		$E_{act} \text{ (eV)}$	$E_g \text{ (eV)}$	$\sigma \text{ (S/cm)}$
	by FTIR	by SIMS			
p-Si	6.1	10.4	0.28	1.56	1.4×10^{-4}
n-Si	4.0	3.5	0.15	1.67	9.8×10^{-3}
i-Si	6.7	7.1	0.81	1.65	1.5×10^{-11}

Figures 1

Figures 2

Figures 3

Figures 4

Figures 5

Figures 6

Figures 7

	Cell 1	Cell 2
V_{oc} (V)	0.79	0.82
J_{sc} (mA/cm^2)	5.29	5.20
$F.F.$	0.56	0.59
η (%)	2.34	2.52

	Cell 3	Cell 4
V_{oc} (V)	0.82	0.82
J_{sc} (mA/cm^2)	4.97	4.92
$F.F.$	0.58	0.57
η (%)	2.36	2.30

	Cell 3'	Cell 4'
V_{oc} (V)	0.82	0.81
J_{sc} (mA/cm^2)	5.48	3.34
$F.F.$	0.59	0.63
η (%)	2.64	1.70