Research Article

Optimization of $\mu$-c-Si$_{1-x}$Ge$_x$:H Single-Junction Solar Cells with Enhanced Spectral Response and Improved Film Quality

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Effects of RF power on optical, electrical, and structural properties of $\mu$-c-Si$_{1-x}$Ge$_x$:H films was reported. Raman and FTIR spectra from $\mu$-c-Si$_{1-x}$Ge$_x$:H films reflected the variation in microstructure and bonding configuration. Unlike increasing the germane concentration for Ge incorporation, low RF power enhanced Ge incorporation efficiency in $\mu$-c-Si$_{1-x}$Ge$_x$:H alloy. By decreasing RF power from 100 to 50 W at a fixed reactant gas ratio, the optical bandgap of $\mu$-c-Si$_{1-x}$Ge$_x$:H was reduced owing to the increase in Ge content from 11.2 to 23.8 at.%, while Ge-related defects and amorphous phase were increased. Consequently, photo conductivity of $1.62 \times 10^{-5}$ S/cm was obtained for the $\mu$-c-Si$_{1-x}$Ge$_x$:H film deposited at 60 W. By applying 0.9 $\mu$m thick $\mu$-c-Si$_{1-x}$Ge$_x$:H absorber with $X_C$ of 48% and [Ge] of 16.4 at.% in the single-junction cell, efficiency of 6.18% was obtained. The long-wavelength response of $\mu$-c-Si$_{1-x}$Ge$_x$:H cell was significantly enhanced compared with the $\mu$-Si:H cell. In the case of tandem cells, 0.24 $\mu$m a-Si:H/0.9 $\mu$m $\mu$-c-Si$_{1-x}$Ge$_x$:H tandem cell exhibited a comparable spectral response as 0.24 $\mu$m a-Si:H/1.4 $\mu$m $\mu$-Si:H tandem cell and achieved an efficiency of 9.44%.

1. Introduction

Thin-film silicon solar cells have the advantages of low material/energy consumption and the ability of large-area fabrication, which is beneficial for the long term production of photovoltaics. To stay competitive with other technologies, further improvement in conversion efficiency is important. By employing different bandgap absorbers that enable broad-band absorption of solar spectrum, the multijunction solar cells have been demonstrated as a viable approach to achieving high-efficiency devices. Taking advantages of ideal combination of absorber bandgaps [1], the hydrogenated amorphous silicon (a-Si:H)/hydrogenated microcrystalline silicon (µc-Si:H) tandem solar cells with stabilized cell efficiencies of over 10% have been demonstrated by many groups [2, 3]. However, due to the indirect bandgap of µc-Si:H material, absorber with few $\mu$m in thickness is needed for achieving sufficient light absorption in the bottom cell. To further enhance the optical absorption in the long-wavelength region, alloying germanium into µc-Si:H network has been proposed [4, 5].

The bandgap of hydrogenated microcrystalline silicon germanium (µc-Si$_{1-x}$Ge$_x$:H) can be narrowed from 1.12 eV toward 0.67 eV, depending on the Ge content (x) in the alloy [5–8]. Furthermore, Matsui et al. [5] have reported that µc-Si$_{0.2}$Ge$_{0.5}$:H had a high absorption coefficient of $10^4$ cm$^{-1}$ at 1.5 eV, which is approximately one order of magnitude higher than that of µc-Si:H. The thinner Si$_{1-x}$Ge$_x$:H cells have been employed to obtain a comparable photocurrent to µc-Si:H cells [9, 10]. Nevertheless, µc-Si$_{1-x}$Ge$_x$:H alloy is a complicated atomic network consisting of a mixed amorphous-crystalline phase and a Si-Ge-H multielement system. As a result, crystallization and Ge incorporation are bound up with the optical and electric properties for µc-Si$_{1-x}$Ge$_x$:H films. A sufficient crystalline phase is needed for the efficient carrier transport; however, the Ge incorporation in a microcrystalline Si network suppresses the crystallization in the growth of µc-Si$_{1-x}$Ge$_x$:H film [11]. The optimization
of crystalline volume fraction \(X_C\) and Ge content \([\text{Ge}]\) by using the proper process parameters is important in the development of \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H alloy.

Previous studies on \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H alloy have demonstrated the correlation between germane concentration \(R_{\text{GeH}_4}\) and Ge incorporation [5, 9, 10, 12–14]. However, the impact of RF power on \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H film properties has not yet been fully investigated. In this work, the effect of RF power on the optical, electrical, and microstructural properties of \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H films has been presented and discussed in detail. Application of corresponding \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H absorbers in single-junction cells and a-Si:H/\(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H tandem cells has also been performed and presented.

2. Experimental Detail

A 27.12 MHz plasma-enhanced chemical vapor deposition (PECVD) system, having a load-lock and a transfer chamber, was employed for the deposition of doped and undoped silicon based thin films. The process chamber was equipped with a 26 x 26 cm\(^2\) plasma reactor. The interelectrode distance was 8 mm. In order to reduce cross contamination, the NF\(_3\) in situ plasma cleaning was introduced in the single-chamber process. The \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H films were deposited by a highly H\(_2\)-diluted gas mixture of silane (SiH\(_4\)) and germane (GeH\(_4\)). The hydrogen dilution ratio \(R_{\text{H}_2}\), defined as \([\text{H}_2]/([\text{GeH}_4]+[\text{SiH}_4])\) and the germane concentration \(R_{\text{GeH}_4}\), defined as \([\text{GeH}_4]/([\text{GeH}_4]+[\text{SiH}_4])\) were kept at 94.9 and 5.06%, respectively. The RF power was varied in the range of 40–100 W with a pressure of 1000 Pa.

For the characterization of film properties, the 200 nm thick \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H films were prepared on Corning EAGLE XG glass substrates at approximately 200 °C. Besides, p-type Si (100) single side polished wafer was utilized as substrate for FTIR measurement. It is known that the crystalline volume fraction \(X_C\) had variation in the initial growth of microcrystalline materials. In this study, the noncrystallized region occupied only a small part of the films deposited on c-Si and glass substrates. The data obtained from different substrates should be self-consistent in this paper. In addition, FTIR data should be representative for the film and correspond to the cell performance for the 0.9 \(\mu m\) thick \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H solar cells. For the quantitative estimation for \(X_C\), the Raman equipment equipped with a diode-pumped solid-state laser and provided an excited wavelength of 488 nm. By using deconvolution of the Raman spectra, the peaks at 520 cm\(^{-1}\), 494–507 cm\(^{-1}\), and 480 cm\(^{-1}\) correspond to crystalline, intermediate, and amorphous phases, respectively [15–17]. In addition, the Ge-related peaks [18, 19] centered at 400, 370, 300, and 270 cm\(^{-1}\) were attributed to the signals of c-Si-Ge, a-Si-Ge, c-Ge-Ge, and a-Ge-Ge, respectively.

For determining a quantitative composition in the \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H film, an X-ray photoelectron spectroscopy (XPS) was used to measure the intensities of Ge\(_{3d}\) and Si\(_{2p}\) core lines for the estimation of the film Ge content [20]. The hydride bonding configuration was characterized by FTIR spectra. A UV-VIS-NIR spectroscopy was used to measure the transmittance \(T\) and the reflectance \(R\) to obtain the absorption coefficient \(\alpha\). By using \(T\) and \(R\), the intercept of Tauc’s plot of \((\alpha h\nu)^{1/2}\) versus \(h\nu\) (photon energy) is commonly used to evaluate Tauc gap. However, the \(\mu c\)-Si phase segregation in alloys [21] and the mixed phase [22] in microcrystalline materials would produce interference fringe. As an alternative indication for an optical property of \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H film, the optical bandgap \(E_{\text{opt}}\) was used in the study, which was determined by the energy of the photon at the absorption coefficient of \(10^3\) cm\(^{-1}\).

Finally, the 0.9 \(\mu m\) thick \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H single-junction cells and a-Si:H/\(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H tandem solar cells with a thickness of 0.24/0.9 \(\mu m\) were prepared on the commercial textured SnO\(_2\):F-coated glass in a superstrate (p-i-n) configuration. The cell with a device area of 0.25 cm\(^2\) was characterized by an AM1.5G solar simulator and a current-voltage measurement. An external quantum efficiency (EQE) measurement was implemented under both short-circuit and reverse voltage-biased conditions to reveal the behaviors of carrier transport and spectral response in the solar cells.

3. Results and Discussion

3.1. Effect of RF Power on the Properties of \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H Films. Figure 1 shows the Raman spectra of \(\mu c\)-Si\(_{1-x}\)Ge\(_x\):H films deposited with the different RF powers. Of a-Si and c-Si peaks indicated a fraction of Ge was incorporated into amorphous phase and thus suppressed the formation of crystalline phase.
Moreover, the weak peak at approximately 247 cm\(^{-1}\) was observed in \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H alloys, which was originated from the resonant mode \([23, 24]\) and overlapped with amorphous background of Ge–Ge mode. Another weak peak at 430 cm\(^{-1}\) was not found, which may be due to the low Ge content in \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H films \((\text{[Ge]} < 25 \text{ at.\%})\). The studies of Raman spectra in Si\(_{1-x}\)Ge\(_x\) alloys have suggested that the c-Ge-Ge peak was broadened and the intensity rapidly decreased as Ge content was lower than 50 at.\% \([19, 23]\).

Although most peaks can be deconvoluted, the precise estimation for crystalline volume fraction of \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H films is not easy to be determined. With less Ge content \((\text{[Ge]} < 25 \text{ at.\%})\) in \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H alloys, the Ge-related modes having broadened shoulder were difficult to be separated. In this study, weak integrated intensity of c-Ge-Ge mode was ignored in the contribution of crystalline column fraction. To obtain a quantified value to compare the degree of crystallization, the \(X_C\) was calculated by the ratio of \((L_c + I_m)/(L_c + I_m + I_a)\), where the integrated intensities of crystalline \((L_c)\), intermediate \((I_m)\), and amorphous \((I_a)\) Si phases in Raman spectra were used \([16, 25]\).

Figure 2 demonstrates the effects of RF power on \([\text{Ge}], X_C, E_04,\) and the conductivity of \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H films. As the RF power decreased from 100 to 40 W, the \([\text{Ge}]\) significantly increased from 11.2 to 23.8 at.\%. In the plasma of PECVD process, the generation of growth precursor is proportional to the density of energetic electrons which are responsible for the reaction and the dissociation cross section \([26]\). The dissociation energies are 83.4 and 91.7 kcal/mole for GeH\(_4\) and SiH\(_4\), respectively \([27]\). Lower power reduces the energy of electron in the plasma and thus shifts the dissociation thresholds for SiH\(_4\) and GeH\(_4\). As a result, relatively more Ge-related precursors were in the gas phase which leads to more Ge incorporation in the solid phase \([28]\). Moreover, the Ge incorporation efficiency \((\text{[Ge]}/R_{\text{GeH}_4})\) indicates the capability of Ge atom transfer from gas phase into solid state. In the case of all samples, the \([\text{Ge]}/R_{\text{GeH}_4}\) was larger than one, suggesting that Ge was preferentially incorporated in \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H films compared to Si. As the RF power decreased from 100 to 40 W, the \([\text{Ge]}/R_{\text{GeH}_4}\) was enhanced from 2.2 to 4.7. This indicated that the lower RF power significantly promoted Ge incorporation for \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H growth.

In comparison to the effect of RF power, the film Ge content in \(\mu\)-Si\(_{1-x}\)Ge\(_x\):H alloy can also be increased by directly increasing GeH\(_4\) concentration. According to the research results of Matsui et al. and our previous work, the nonlinear behavior of Ge incorporation was observed and reflected a decrease in Ge incorporation efficiency by adding GeH\(_4\) in SiH\(_4\)-GeH\(_4\)-H\(_2\) plasma \([10, 11]\). More sticky GeH\(_3\) growth precursors would be produced and increased the
weak Ge-related bonds on the growth surface [29, 30]. In the hydrogen-containing plasma atmosphere, the probability of SiH₄ precursors replacing the weak Ge-bonded site may be enhanced. Thus the drop in Ge incorporation efficiency was found when the $R_{\text{GeH}}$ was increased.

Matsui et al. [10, 31] have also reported that the Ge content of $\mu$-Si$_{1-x}$Ge$_x$:H films prepared in 100 MHz VHF-PECVD with 130 cm$^2$ reactor was slightly changed by only 2% in the power density range from 0.12 to 0.23 W/cm$^2$. At a higher RF frequency and a relatively higher power density, the GeH$_4$ and SiH$_4$ may be completely dissociated in the plasma. In contrast, a lower RF frequency of 27.12 MHz and a lower power density ranging from 0.06 to 0.15 W/cm$^2$ plasma. In addition, a properly high power density for a-Si$_x$Ge$_{1-x}$:H growth improved film quality [28]. Increasing RF power and NHSMs at 2083, 2102, and 2137 cm$^{-1}$ film prepared at 100 W, relatively high IR absorption from 2080 to 2150 cm$^{-1}$ was observed. In this region, three narrow high stretching modes (NHSMs) at 2083, 2102, and 2137 cm$^{-1}$ which correspond to SiH$_3$, SiH$_2$, and SiH$_x$ at crystalline grain boundaries, respectively [34], were reported as a signature of porous and less-dense structure in high-$X_C \mu$-Si:H network [35]. The presence of NHSMs in our case increased the carrier recombination loss and thus reduced the electrical property (Figure 2). Another two high SMs (HSMs) at 2120 and 2150 cm$^{-1}$ were assigned to SiH$_3$ and SiH$_x$ which resulted from the macroscopic amorphous surfaces in $\mu$-Si:H films. As the power was reduced from 100 to 50 W, both NHSMs and HSMs were reduced due to the reduction in $X_C$. The reduced NHSMs indicated that the micro voids or vacancies at the grain boundary were reduced, leading to a more compact structure [35]. This coincided with the decrease in dark conductivity as the power decreased from 100 to 50 W. In addition, the SMs ranging in 1980–2010 cm$^{-1}$ were assigned to SiH$_3$ and SiH$_x$ bonding, which reflected silicon hydrides in the bulk amorphous phase [36].

In the case of $\mu$-Si$_{1-x}$Ge$_x$:H films prepared at 80 and 100 W, the presence of component at 1880 cm$^{-1}$ reflected the mode of GeH [28, 37]. As RF power was less than 80 W, the GeH bonding was absent. This result suggested that sufficient power is beneficial in providing energy for structure relaxation of Ge-related precursors [38]. A lower RF power for $\mu$-Si$_{1-x}$Ge$_x$:H growth provided less energy to the precursors and shortened the diffusion lengths for the precursors, especially in sticky Ge-related precursors. The Ge-related precursors easily stuck on the growth surface without seeking a minimum energy bond site and formed the Ge-related
weak bonds. According to the XPS result, the increase in Ge incorporation contributed to SiGe network, which narrowed the bandgap. The enhanced light absorption promoted more photo-generated carriers. Nevertheless, the carriers transport was probably recombined by the increase in Ge-induced defects. Ge incorporation could easily induce interconnected microvoids and dangling bonds. This increased the heterogeneity in the SiGe matrix and provided recombination centers for charged carriers, which was different from the midgap defects [28]. Furthermore, previous works showed that Ge incorporation could drive out and recombine carrier traps in the bulk absorber due to the presence of more amorphous phase. At the first few tens of nanometers of the film, more amorphous phase led to a barrier that reduced carrier mobility. This increased the recombination loss and resulted in carrier extraction problems at the p/i interface [39].

To clarify the carrier collection in the $\mu$-Si$_{1-x}$Ge$_x$:H cell, the reverse-bias EQE was measured. Figure 5(b) compares the EQEs of $\mu$-Si$_{1-x}$Ge$_x$:H cell with bias voltage of 0 and −1 V. The cell with $\mu$-Si$_{1-x}$Ge$_x$:H absorber prepared at 50 W was used. When the reverse bias was applied, EQE was enhanced at the wavelength from 350 to 750 nm with an increased current density by 0.89 mA/cm$^2$. This suggested that carriers trapped in the bulk absorber due to the presence of Ge-induced defects were driven out and were collected. In contrast to the cell with $\mu$-Si$_{1-x}$Ge$_x$:H absorber prepared at 50 W, cells with $\mu$-Si$_{1-x}$Ge$_x$:H absorber prepared from 60 to 100 W exhibited additional increase in current density by less than 0.4 mA/cm$^2$ under reverse-bias condition. This suggested that the carrier collection across the p/i region can be improved by reducing Ge-induced defects and amorphous phase.

As compared with the single-junction cell having 0.9 µm thick $\mu$-Si:H absorber, the $\mu$-Si$_{1-x}$Ge$_x$:H cells exhibited a substantial enhancement in EQE at the wavelength from 500 to 1100 nm (Figure 5(b)). The $J_{SC}$ was increased by 2.8 mA/cm$^2$. This shows that the $\mu$-Si$_{1-x}$Ge$_x$:H single-junction solar cells had more superior spectral response than $\mu$-Si:H single-junction solar cells, especially in the red-to-infrared region, which was beneficial in the multijunction configuration.

The performance of $\mu$-Si$_{1-x}$Ge$_x$:H single-junction cells with $\mu$-Si$_{1-x}$Ge$_x$:H absorbers deposited at different RF power is illustrated in Figure 6. The performance of the $\mu$-Si:H cell
with 0.9 μm thick absorber (Xc ~50%) was also shown for comparison. As can be seen from Figure 6, the short-circuit current density \(J_{SC}\) was kept approximately 20 mA/cm\(^2\) as the power decreased from 100 to 60 W. This was due to the enhancement in the long-wavelength response accompanied with the reduction in short-wavelength response. When the RF power was further decreased to 50 W, the \(J_{SC}\) was reduced to 17.3 mA/cm\(^2\) with the corresponding [Ge] increasing to 18.6 at.%. The reduction in \(J_{SC}\) can be attributed to the increase in Ge-related defects which worsened the carrier
transport in the μc-Si1−xGex:H absorber. This can also be supported by the reduced photoconductivity as power reduced to 50 W. On the other hand, the enhancement of the fill factor (FF) from 56.1% to 64.3% was found when the RF power decreased from 100 to 60 W. This was likely due to the reduction in voids and vacancies at the grain boundary in μc-Si1−xGex:H absorber as suggested by the decreased NHSMs (Figure 3). The voids and vacancies induced structural defects where the carrier would be recombed. Similar effect of NHSMs on the performance of μc-SC:H cells has also been reported [35, 40]. In addition, as the RF power decreased from 100 to 60 W, the open-circuit voltage ($V_{OC}$) of μc-Si1−xGex:H cells had a monotonic increase from 0.40 to 0.47 V (average value). This could be due to the suppression of defects at grain boundary, leading to the reduction in reverse saturation current. The reverse saturation current density of μc-Si1−xGex:H cells was decreased from 2.76 × 10−6 to 7.02 × 10−7 A/cm², indicating less leakage path and recombination loss of carriers in the cells. When the power was further reduced to 50 W, the $V_{OC}$ and FF leveled off. This could be due to too much Ge-induced defects in the film, which hindered the carrier transport and degraded electrical property. With the μc-Si1−xGex:H absorber deposited at 60 W and 1000 Pa, the single-junction cell efficiency of 6.18% was obtained with $V_{OC} = 0.475$ V, FF = 64.3%, and $J_{SC} = 20.22$ mA/cm². The corresponding [Ge] and Xc of this μc-Si1−xGex:H film were 16.4 at.% and 48%, respectively.

Compared to the state-of-the-art μc-Si1−xGex:H single-junction solar cell [41] having an efficiency of 8.2% with $J_{SC} = 25.5$ mA/cm², $V_{OC} = 0.494$ V, and FF = 0.65, the comparable $V_{OC}$, FF, and lower $J_{SC}$ were obtained in this research. The difference in front of TCO layer and antireflection coating [42] is likely to be the main reason for the lower $J_{SC}$. For the fabrication of microcrystalline silicon based single-junction solar cells, the front TCO plays an important role in the cell performance. Microcrystalline Si-based materials usually require the highly diluted H2-containing plasma for an adequate $X_C$. Unfortunately, the commercial SnO2:F-coated substrate is much chemically unstable than the ZnO:Ga in the hydrogen-rich plasma. Therefore, the a-Si:H(p)/μc-Si:H(p) bi-layer was used as the p-type window layer for protection of SnO2:F surface for resisting Sn reduction, as reported in our previous work [43].

To demonstrate the improvement of using μc-Si1−xGex:H as the absorber of bottom cell, the a-Si:H/μc-Si1−xGex:H tandem cell was fabricated. The n-type μc-SiO2:F with oxygen content of 8.5 at%, optical gap ($E_{g0}$) of 2.13 eV, and conductivity of $8 \times 10^{-2}$ S/cm was employed as the intermediate reflective layer (IRL) between the component cells. Detail on the study of IRL was reported in our previous work [44]. Table 1 summarizes the cell performance of a-Si:H (0.24 μm)/μc-Si1−xGex:H (0.9 μm) and a-Si:H (0.24 μm)/μc-Si:H (1.4 μm) tandem cells. In comparison with the cell having 1.4 μm thick μc-Si:H bottom absorber, the cell with the 0.9 μm thick μc-Si1−xGex:H bottom absorber exhibited a comparable $V_{OC}$ of 1.33 V with a slightly lower FF of 69.7%. The latter can be ascribed to the Ge-induced defects which adversely influenced the carrier transport in μc-Si1−xGex:H cell. Notably, the tandem cell using 0.9 μm thick μc-Si1−xGex:H as bottom absorber exhibited comparable $J_{SC}$ of 10.18 mA/cm² compared to the cell with 1.4 μm thick μc-Si:H absorber, which was confirmed by the quantum efficiency result shown in Figure 7. There was no significant difference of EQE in short-wavelength region, whereas a slight increase in spectral response was found at the wavelength from 600 to 1100 nm in the case of a-Si:H/μc-Si1−xGex:H tandem cell. Using a 0.9 μm thick μc-Si1−xGex:H bottom absorber in tandem cell exhibited the bottom cell $J_{SC}$ of 11.39 mA/cm², which was 0.39 mA/cm² higher than that of the cell with 1.4 μm thick μc-Si:H bottom absorber. Compared with the μc-Si:H absorber, employment of μc-Si1−xGex:H reduced the absorber thickness by over 30%. The corresponding total current density can reach 22.67 mA/cm². The results indicated that a relative thin bottom absorber can be used for a sufficient IR absorption, fulfilled by applying the μc-Si1−xGex:H bottom absorber in Si-based tandem cells. The conversion efficiency for the 0.24 μm thick a-Si:H/0.9 μm

<table>
<thead>
<tr>
<th>Bottom absorber</th>
<th>$V_{OC}$ (V)</th>
<th>$J_{SC}$ (mA/cm²)</th>
<th>FF (%)</th>
<th>Eff. (%)</th>
<th>$J_{QE}$ (mA/cm²) $J_{QE}$ (total) (mA/cm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.9 μm μc-Si1−xGex:H</td>
<td>1.33</td>
<td>10.18</td>
<td>69.7</td>
<td>9.44</td>
<td>11.28 11.39 22.67</td>
</tr>
<tr>
<td>1.4 μm μc-Si:H</td>
<td>1.34</td>
<td>10.39</td>
<td>72.0</td>
<td>10.07</td>
<td>11.27 11.00 22.27</td>
</tr>
</tbody>
</table>

Table 1: Comparison of cell performance for tandem solar cells, having the 0.24 μm thick a-Si:H top cell stacked with different bottom absorbers.
thick $\mu$-Si$_{1-x}$Ge$_x$:H tandem cell was obtained as 9.44%, with $J_{SC} = 10.18$ mA/cm$^2$, $V_{OC} = 1.33$ V, and $FF = 69.7\%$. The $\mu$-Si$_{1-x}$Ge$_x$:H cell can be an important building block as the bottom cell in high-efficiency triple- or quadruple-junction cells that had the potential to obtain efficiency of 20% [45].

4. Conclusions

In this study, the effects of RF power on optical, electrical, and structural properties of $\mu$-Si$_{1-x}$Ge$_x$:H films were investigated. Decreasing RF power density significantly increased Ge incorporation in $\mu$-Si$_{1-x}$Ge$_x$:H films. The increased Ge content led to the bandgap narrowing. However, low-energy plasma weakened structural relaxation and the crystalline volume fraction decreased. FTIR data showed that the defect- and structural properties of $\mu$-Si$_{1-x}$Ge$_x$:H deposited at high RF power. However, H passivation was less effective at a low RF power. Consequently, photoconductivity of $1.62 \times 10^{-5}$ S/cm and a better film quality of $\mu$-Si$_{1-x}$Ge$_x$:H was obtained at 60 W. The corresponding $X_c$ and [Ge] were 48% and 16.4%, respectively. The cell efficiency for 0.9 $\mu$m thick $\mu$-Si$_{1-x}$Ge$_x$:H single-junction cell achieved 6.18% with $V_{OC} = 0.475$ V, FF = 64.3%, and $J_{SC} = 20.22$ mA/cm$^2$. Compared to the $\mu$-Si:H cell, the QE measurement showed that the long-wavelength response of $\mu$-Si$_{1-x}$Ge$_x$:H cell was significantly enhanced. With a much thinner bottom absorber thickness, 0.24 $\mu$m a-Si:H/0.9 $\mu$m $\mu$-Si$_{1-x}$Ge$_x$:H tandem cell exhibited a comparable spectral response as 0.24 $\mu$m a-Si:H/1.4 $\mu$m $\mu$-Si:H tandem cell and achieved a cell efficiency of 9.44%.

Conflict of Interests

The authors do not have any conflict of interests with the content of the paper.

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