Hydrogen plasma treatment of very thin p-type nanocrystalline Si films grown by RF-PECVD in the presence of B(CH₃)₃

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Abstract
We have characterized the structure and electrical properties of p-type nanocrystalline silicon films prepared by radio-frequency plasma-enhanced chemical vapor deposition and explored optimization methods of such layers for potential applications in thin-film solar cells. Particular attention was paid to the characterization of very thin (∼20 nm) films. The cross-sectional morphology of the layers was studied by fitting the ellipsometry spectra using a multilayer model. The results suggest that the crystallization process in a high-pressure growth regime is mostly realized through a subsurface mechanism in the absence of the incubation layer at the substrate-film interface. Hydrogen plasma treatment of a 22-nm-thick film improved its electrical properties (conductivity increased more than ten times) owing to hydrogen insertion and Si structure rearrangements throughout the entire thickness of the film.

Keywords: nanocrystalline silicon, trimethylboron, incubation layer, subsurface growth, PECVD

1. Introduction

Hydrogenated nanocrystalline silicon (nc-Si : H) has attracted considerable attention during the last two decades owing to its superior properties, as compared with hydrogenated amorphous silicon (a-Si : H), such as higher stability, carrier mobility, and ability to absorb low-energy photons. Intrinsic nc-Si : H was successfully employed as an absorber layer in single-junction fully nanocrystalline solar cells [1], with the record-high initial efficiency of 10.1% [1], or in tandem (micromorph) solar cells, with an initial efficiency of up to 14.7% [3] and the best stable confirmed value of 11.7% [1]. P-type nc-Si : H can also be used as a window layer in a-Si : H solar cells [12].

Nc-Si : H benefits from full compatibility with amorphous silicon thin-film technology. The standard a-Si : H deposition method is based on the dissociation of silane (SiH₄) in radio-frequency plasma enhanced chemical vapor deposition (RF-PECVD) reactors and production of SiHₓ radicals. The growth of nc-Si : H by PECVD requires a very high hydrogen dilution of silane, as the hydrogen radicals assist in the formation of nc-Si : H via (a) hydrogen surface coverage [7, 8], (b) creation of surface dangling bonds, (c) chemical etching (creation of dangling bonds and breakage of weak Si-Si bonds) [9] and (d) relaxation of Si network by hydrogen propagation into the bulk [10].

The growth of nc-Si : H strongly depends on the substrate [11] and typically starts with the formation of an
a-Si : H incubation layer ranging in thickness from a few nanometers to half micron [12]. Deposition of nc-Si : H films typically involves four stages, namely, incubation, nucleation, growth and steady state. The presence of the incubation layer complicates the deposition of very thin nc-Si : H films (<20 nm). This is an important issue since very thin p-type nc-Si : H films are necessary for a-Si and nc-Si solar cells. Several techniques can be used to eliminate the incubation layer during the first stages of nc-Si : H growth: variable hydrogen dilution [13] and cyclical variation of hydrogen dilution [14] during hot-wire CVD process were employed to create a seed layer of Si nanocrystals. Layadi et al [15] used layer-by-layer deposition with an appropriate ratio of a-Si : H deposition time to hydrogen plasma treatment time to produce a fully crystallized interface. Van den Donker et al [16] tailored the initial SiH4 density to prevent the formation of the incubation layer and deposit microcrystalline silicon films from pure SiH4. Roca i Cabarrocas et al [17] proposed to synthesize Si nanocrystals in the gas phase during deposition, thereby avoiding incubation and nucleation phases on the substrate.

During the growth of doped nc-Si : H films, the presence of dopant elements adds disorder to the Si structure and, consequently, complicates the formation of nc-Si : H without an a-Si : H incubation layer. Typically, diborane (B2H6) or trimethylboron (B(CH3)3 or TMB) are used for growing p-type Si films. In the case of TMB, the presence of the supplementary carbon atoms suppresses crystalline growth [18]. On the other hand, the use of TMB as dopant gas, when compared with diborane, yields p-type Si thin films with increased optical gap [19], favoring the use of such films as a window layer in thin-film Si solar cells. Although the properties of silicon films with a thickness above 100 nm are well documented, they differ significantly from those of very thin (10–30 nm) films, especially at the interface with the substrate. In this paper, we present results of the deposition of very thin (~20 nm) highly conductive p-type nc-Si : H films, produced under high pressure, high power density and high H2 dilution regime with the use of TMB as dopant gas by standard RF-PECVD technique. The impact of post-deposition hydrogen plasma treatment on the nc-Si : H film morphology and electrical conductivity is examined.

2. Experimental details

The films were deposited on 1-mm-thick glass substrates in an RF-PECVD chamber, previously described in [20], at a substrate temperature of 150 °C. The base pressure in the vacuum chamber before deposition was lower than 10⁻⁶ Torr. The deposition parameters, optimized for deposition of thick (>150 nm) p-type layers, were presented previously [21, 22], and are listed in table 1. The hydrogen dilution, D_H, is calculated as D_H = (F_H2/(F_H2 + F_SiH4 + F_TMB)) × 100%, where F_s are the corresponding flows, and F_TMB is the ratio of TMB flow to silane flow: R_TMB = (F_TMB/F_SiH4) × 100%.

Since all the deposition parameters were optimized previously, in this work, we studied the effect of deposition time and post deposition hydrogen plasma treatment (HPT) on the film thickness and crystallinity. The flow of hydrogen and pressure were kept constant during deposition and HPT. To prevent surface oxidation, hydrogen plasma treatment was applied to as-grown Si films without exposing them to the atmosphere.

Structural characterization of the nc-Si : H films was performed using Raman spectroscopy and spectroscopic ellipsometry (SE). Raman scattering experiments were performed at room temperature, in the backscattering configuration, using a LabRam 300 Horiba Jobin Yvon spectrometer and a 532 nm laser. The power of the incident beam was small enough (1.7 mW) to avoid recrystallization of the films. The crystalline volume fraction, X_{CRS}, was determined using the procedure previously reported by Yue et al [23]:

\[
X_{CRS} = \frac{A_C + A_{GB}}{A_C + A_{GB} + \eta A_A} \times 100%,
\]

where A_C, A_A and A_{GB} are the integrated intensities of the 520, 480 and 495–510 cm⁻¹ peaks, respectively; \( \eta \) is the ratio of the cross sections for the amorphous and crystalline phases, which was taken as \( \eta = 0.8 \). Estimation of this coefficient is rather complicated and no consensus exists on its value in the literature. Although the uncertainty in \( \eta \) may lead to errors in the absolute crystalline volume fractions, it does not affect the comparison of crystal volume fractions in samples grown under similar deposition conditions.

Ellipsometry spectra were measured using a Horiba Jobin Yvon UVISEL ellipsometer, with a fixed 70° incidence angle, in the range of 1.5–6.5 eV to determine the crystalline fraction, X_{CSE}, of the films and its variation with thickness. The results were fitted as described in\(^4\) (figure 1). The modeling consisted of dividing the film into the bulk, subsurface and surface regions. In these three layers, two material references (a-Si and nc-Si), with optical properties presented in figure 1, and voids were mixed at different ratios to minimize the difference (\( \chi^2 \leq 0.5 \)) between the measured and simulated data [21].

We also modeled the spectra using the Tauc–Lorentz (TL) model [25] to estimate the values of the optical gap E_{opt}. We

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4 Reference spectra are supplied by Jobin Yvon software.
followed this approach to determine $E_{opt}$ as the data obtained from transmission spectroscopy did not follow the Tauc plot due to the nanocrystalline structure of the thin films. The film surface roughness was modeled using the Bruggeman effective medium approximation (BEMA) with 50% TL and 50% voids.

Electrical properties of the films were studied via measurements of conductivity versus temperature, from which the room-temperature conductivity, $\sigma_d$, and activation energy, $E_a$, were calculated. The linearity of $I(V)$ dependence was confirmed before each conductivity measurement. Low voltages (0.1–1 V) were used to reduce high-field effects, such as field-enhanced hopping transport. Aluminum contacts (180 nm thick, 5 mm long and 1 mm apart) were formed on Si films by physical vapor deposition. Chromium contacts with the same geometry were evaporated on a glass substrate prior to Si film deposition.

Cross sections of the nc-Si : H films were examined using a ZEISS Auriga CrossBeam workstation operated at 5 kV with a working distance of 3.8 mm. Asylum MFP-3D atomic force microscopy (AFM) was used in tapping mode with Si tips vibrated at 300 kHz to analyze the effect of the HPT on the surface roughness.

### 3. Results and discussion

Previous results of fitting SE data for thick (>140 nm) films with a multilayer model suggested a presence of a subsurface incubation layer (approximately 15-nm-thick layer located under the surface of the nc-Si : H layer), instead of the typical incubation layer at the substrate-film interface [21]. To support this growth model, a set of electrical measurements was carried out on (a) samples with different configurations of electrical contacts, and (b) on films with different thicknesses. In the former case, the parallel chromium contacts were formed on a glass substrate prior the deposition of p-type nc-Si : H film. Aluminum contacts were then evaporated on the nc-Si : H film ~7 mm away from the Cr contacts (see the inset in figure 2). If an amorphous incubation layer with a thickness of several nanometers existed at the substrate-film interface, it would create a potential barrier for the carriers injected from the Cr contacts. This would lead to lower conductivity values measured between the Cr (bottom) contacts than between the Al (top) contacts. An opposite behavior should be observed in the absence of an amorphous incubation layer at the substrate-film interface. The surface/substrate of the film is typically terminated with a native oxide layer, but this layer is expected to be much thinner than the amorphous incubation layer, and so it should not significantly affect the measurement. Moreover, before measurements, the samples were annealed (150 °C, 1 h, 10^{-3} Torr) to improve the contacts and mitigate the effect of native oxides, which were present on the Cr surface prior to the silicon film deposition.

Figure 2 shows the values of $\sigma_d$ as a function of inverse temperature (Arrhenius plot) for a 191-nm-thick p-type nc-Si : H film. To reduce statistical errors, conductivity measurements were performed on three different pairs of Cr contacts (with the same geometry) and on three pairs of Al contacts. The conductivity measured between the Cr contacts placed underneath the p-type nc-Si : H film was always higher than between top Al contacts: $\sigma_d = 7.10 \pm 0.13$ (Ω cm)^{-1} versus $\sigma_d = 6.20 \pm 0.08$ (Ω cm)^{-1}, respectively. This observation indicates the presence of some ‘defect-rich’ layer on or near the film surface; it does not exclude the presence of an amorphous incubation layer, and only shows that there are more defects at the surface of the film than at its interface with the glass substrate.

The second set of electrical measurements involved conductivity measurements between Al contacts on the surface of p-type nc-Si : H films with different thicknesses. Figure 3(a) shows $\sigma_d$ and $E_a$ data measured for the films with thicknesses varying from 22 to 212 nm. The conductivity values remain high, above 1 (Ω cm)^{-1} in a wide range of film thicknesses. However, for very thin layers, a minor change in the film thickness results in an abrupt decrease of $\sigma_d$, namely, as the thickness decreases from 28 to...
and dense structure. Region near the surface, whereas in the bulk and close to the above. Each fitting indicates the presence of a void-rich and the spectra were modeled using three layers as described the in-depth morphology, SE measurements were conducted averaged morphology of the nc-Si : H films. To characterize excitation, the Raman signal represents the thickness-(figure fitting results with equation (a)). Structural and morphological changes in very thin films may be responsible for such a sudden drop of σ22 nm, the σd drops by more than two orders of magnitude (figure 3(a)). Structural and morphological changes in very thin films may be responsible for such a sudden drop of σd. Similar dependences of conductivity with film thickness were reported previously [26, 27]; however, the decrease in σd (to 10−4 (Ω cm)−1 or less) was less abrupt and occurred at a film thickness of 50 nm. This effect has been mainly ascribed to a significant decrease in the crystalline fraction when the film thickness approaches the thickness of the amorphous incubation layer and, consequently, the percolation path becomes discontinuous.

In our case, however, the crystalline fraction of the thin films (determined by Raman spectroscopy, equation (1)) shows an insignificant decrease with film thickness (figure 3(b)) as compared with the σd values. For 532 nm excitation, the Raman signal represents the thickness-averaged morphology of the nc-Si : H films. To characterize the in-depth morphology, SE measurements were conducted and the spectra were modeled using three layers as described above. Each fitting indicates the presence of a void-rich region near the surface, whereas in the bulk and close to the substrate–film interface, the material has a highly crystalline and dense structure.

The morphology of the studied films and their growth mechanism can be explained in the framework of surface and subsurface reactions of the growing film with hydrogen ions [28]. It is also expected that Si nanoclusters, formed in the gas phase during the deposition in the high-pressure regime, contribute to the formation of a seed layer directly on the substrate, since Si clusters in the gas phase are formed at a much higher rate than the Si layer on the substrate. Consequently, during the initial stages of growth, a porous layer is formed on the substrate, composed of voids and Si nanoparticles deposited directly from the gas phase and Si amorphous phase as a result of the interaction of SiH₃ radicals. This layer is exposed to several competing processes, such as chemical etching, bombardment of heavy (silicon) and light (H⁺, H₂⁺) ions, among others. The amount of crystalline phase in the porous material increases with time. At this stage, heavy and light ions start playing a crucial role in the formation of surface and bulk layers, respectively. Hydrogen ion implantation (with higher implantation depth) increases the H content at the bottom of the subsurface layer, favoring chemical annealing [28] and the formation of a dense nanocrystalline bulk layer. The surface and subsurface layers are affected by amorphization induced by heavy ions (such as Si₂⁺2H₂⁺ [29]) [28], and, simultaneously, new Si clusters constantly arrive to the surface from the gas phase. The competition between these processes is responsible for the formation of a subsurface layer with a constant thickness, which is moving upwards leaving behind a dense nanocrystalline material.

Thus far, the roles of the C and B atoms on the subsurface nucleation process remained unclear. According to the previous results [21, 22], the dopant elements do contribute to the suppression of crystallinity under the mentioned growth conditions. Thus, intrinsic films grown under the same deposition conditions in absence of TMB have a crystalline volume fraction above 80%, as compared to 69% for the p-type nc-Si : H [22]. The decrease of the doping effect (observed via conductivity) with decreasing power density and deposition pressure [21] suggests that formation of the Si clusters in the plasma (at higher pressure) and enhanced implantation of light H ions into the subsurface region (at increased power) play a crucial role in the formation of highly conductive p-type nc-Si : H. Furthermore, the optical gap, E,opt, defined from the modeling of ellipsometry spectra, is 1.75–1.77 eV. Its higher value than the typical gap of the intrinsic nc-Si : H indicates the formation of Si-C bonds.

Figure 4(a) shows the evolution of the film’s morphology with the growth time. This figure is a compilation of the results obtained from the SE modeling of the film microstructure: the ‘highly dense nc-Si’ area represents the bulk layer (including the interface region between the substrate and the film) with the crystalline fraction varying from 70 to 89%. The ‘void and a-Si rich’ interlayer mainly consist of nanostructured material with high fractions of amorphous inclusions (20–60%) and voids (10–34%). The ‘mixed phase surface layer’ has a thickness of 4.2–5.4 nm and a high fraction of voids (44–48%), the rest being a mixture of Si nanocrystals and amorphous phase.
Figure 4. (a) Evolution of the film morphology with deposition time. The bars indicate microstructure of the thin films: black represents highly dense nc-Si : H material; blue—void and a-Si : H rich subsurface layer, and red—a mixed-phase surface layer. The areas between the bars are filled for presentation purposes. The inset shows a cross-sectional SEM image of the 140-nm-thick film. (b) Cross-sectional SEM image of another p-type nc-Si : H thin film.

Since the deposition of our thin films occurs in a ‘dusty-plasma’ regime, we expect that Si clusters 2–4 nm in size formed in the gas phase are constantly incorporated into the growing film surface. Interactions in the subsurface region with light H ions would result in the growth and coalescence of the smallest clusters into bigger ones, with sizes up to 20–40 nm (distinguishable in figure 4(b)). The space between these crystals is filled with a-Si : H tissue and voids.

We shall, however, distinguish the voids in the surface layer and in the bulk and subsurface regions. The thickness of the surface layer, estimated from SE modeling, represents irregular distributions of the smallest clusters into bigger ones, with sizes up to 20–40 nm (distinguishable in figure 4(b)). The space between these crystals is filled with a-Si : H tissue and voids.

During the growth of nc-Si : H, its surface is quickly covered with new nanocrystals arriving from the plasma, and light ions contribute to the recovery of the crystalline structure in the subsurface area. The thickness of the subsurface layer (estimated from the SE modeling) lies in the range of 15–25 nm for all the studied samples. It is plausible to assume that the distributions of the sizes and densities of voids and Si clusters in the subsurface region are not uniform along the thickness, and that their properties are strongly dependent on the damage/reconstruction by the heavy/light ions. Correspondingly, the size of the Si clusters may vary from 2–4 nm (under the surface) up to 20–40 nm at the bottom of the subsurface region. From the percentage of voids in the amorphous and nanocrystalline Si phases, the size of voids present in the subsurface layer is estimated as 0.5–2.0 nm, and their shape is likely irregular. The amount of voids (porosity) in the subsurface and bulk regions represents the hydrogen content distributed in the thin films. The bulk layer is protected from bombardment by the heavy ions and consists mostly of compact Si crystals (several tens of nanometers in size) separated by amorphous grain-boundaries and a very small amount of voids.

According to the SE modeling results, the thinnest film in this study (22.0 nm) does not have any ‘mixed-phase surface layer’. In other words, during the fitting, the thickness of the surface layer was set to zero. The dense crystalline phase (81% of nc-Si : H, 18% of a-Si : H, 1% of voids) of this film is rather thin, only 2.4 nm, and most of the film thickness (19.6 nm) corresponds to a mixture of 23% nc-Si : H, 43% a-Si : H and 34% voids (see figure 6(c)). The thickness of the dense crystalline phase of this layer is not sufficient for the formation of a percolation path for the transport of electrons between the silicon nanocrystals; consequently, the transport is mostly realized through the thicker mixed-phase layer. The thicker films in this study are similar in terms of morphology and thickness of the surface and subsurface layers; however, the dense bulk layer grows faster after its ‘delayed’ formation. This explains the rather abrupt increase in $\sigma_d$ (two orders of magnitude) with a small change in the film thickness: as the total film thickness increases from 22.0 to 28.2 nm, the bulk dense layer grows from 2.4 to 7.3 nm, allowing the formation of a percolation path for efficient charge transport. A further increase in the film thickness does not significantly affect the conductivity (figure 3(a)), suggesting that the dense bulk layer plays a major role in the conductivity.

To confirm this assumption, we performed a simple modeling of the conductivity versus thickness dependence shown in figure 3(a). During the conductivity measurements with coplanar Al contacts, electrons cross both the bulk and the top (surface plus subsurface) layers. Consequently, the total current is the sum of the individual currents flowing in...
each layer (figure 5). The total resistance of the film is then represented as

$$R_{\text{total}} = \frac{R_{\text{Top}} \times R_{\text{B}}}{R_{\text{Top}} + R_{\text{B}}}.$$  (2)

with

$$R_{\text{top}} = \frac{l}{\sigma_{a-Si} d_{\text{Top}} m}$$ and $$R_{\text{B}} = \frac{l}{\sigma_{nc-Si} (d_{\text{total}} - d_{\text{Top}}) m}.$$  (3)

Here, $R_{\text{Top}}$ and $R_{\text{B}}$ are resistances of the top and bulk layers, respectively; $l$ is the distance between the contacts; $m$ is the length of the contacts; $\sigma_{a-Si}$ and $\sigma_{nc-Si}$ are, respectively, the conductivities of the top layer, which mainly consists of amorphous phase, and nanocrystalline (bulk) silicon; $d_{\text{Top}}$ and $d_{\text{B}}$ are the thicknesses of the top and bulk layers, respectively. The conductivity of the entire film can be calculated as

$$\sigma_{d} = \frac{l}{R_{\text{total}} d_{\text{total}} m}.$$  (4)

where $d_{\text{total}} = d_{\text{Top}} + d_{\text{B}}$. The fitting results are shown by the solid line in figure 3(a), and a good agreement between the experimental and calculated data points is observed. The fitting parameters were the thickness of the top layer, $d_{\text{Top}}$, and the conductivities of both layers, $\sigma_{a-Si}$ and $\sigma_{nc-Si}$. The optimized values are $d_{\text{Top}} = 22.0$ nm, $\sigma_{a-Si} = 1.0 \times 10^{-4}$ ($\Omega \text{cm})^{-1}$ and $\sigma_{nc-Si} = 0.8$ ($\Omega \text{cm})^{-1}$. The fitted thickness is close to the value for the top layer estimated from the SE fitting (19.6 nm). The conductivity values are typical for doped amorphous and nanocrystalline silicon thin films.

According to the above results, the electrical properties of the p-type nc-Si : H films with a total thickness of $\sim 20$ nm are strongly affected by minor changes in the films’ morphology or thickness. To increase the electrical conductivity of a 22-nm-thick film without increasing its thickness, we have treated it with a hydrogen plasma. It is expected that in the absence of the heavy ions from silane plasma, which cause surface amorphization of the growing film, light hydrogen ions would assist the recrystallization process, at least during the initial period of the treatment.

Figures 6(a) and 6(b) show Raman and ellipsometry spectra of the as-deposited and hydrogen-plasma-treated (3 min) thin films. Modeling of the Raman spectra suggests a slight increase of crystallinity from 60 to 63%. Multilayer fitting of the ellipsometry spectra of these two thin films (figure 6(b)) reveals changes in the film morphology along the entire thickness. According to the SE fitting results, the HPT film contains two layers with similar thicknesses: 10.0 nm for the bulk and 9.9 nm for the top layer, whereas the corresponding values for the as-deposited film were 2.4 and 19.6 nm. As can be seen in figure 6(c), the crystalline fraction in the plasma-treated film did not increase at the bottom of the film; however, it became more homogeneously distributed along the entire thickness. During the interaction of hydrogen plasma with a thin film, a hydrogen-rich layer is formed [31] and rearrangements of the Si structure occur. Diffusion of
hydrogen atoms into a Si film is realized mainly through a combination of bond breaking and insertion of H in weak and dangling bonds [32]. Our as-deposited nc-Si : H film consists mostly of a porous mixture of a-Si : H and Si nanoclusters; consequently, structural rearrangements due to H insertion may easily occur. It can be noticed in figure 6(c) that after 3 min of HPT, the film contains fewer voids, suggesting the formation of a denser layer at its top. The AFM images of figure 6(d) partially confirm the SE fitting results, since the thickness of the top layer is related to the film roughness. Root mean square (RMS) roughness, measured by AFM, decreased from 1.50 nm for as-deposited to 1.06 nm for HPT thin film.

All the modifications of the film morphology due to plasma treatment should lead to improvements of the electrical properties. A thicker and denser bulk layer with hydrogen-passivated defects should facilitate electron transfer. Figure 7(a) shows the evolution of conductivity with H₂ plasma treatment time. The maximum conductivity is achieved after 3 min of plasma treatment and the resultant values are more than ten times higher than those of the as-deposited thin film. Figures 7(b)–7(d) show changes in the film morphology and thicknesses of the bulk and top layers with HPT time. Most changes occur during the first minute of the plasma treatment, which indicates successful H incorporation into the thin film through its entire thickness. The behavior of conductivity closely follows the amount of crystalline phase in the top layer (figure 7(c)), while the crystalline phase in the bulk layer is not affected after an initial decrease. Prolonged exposure of the layers to the hydrogen plasma causes degradation of the film’s structure due to excessive hydrogen incorporation into the film, as well as film etching (at > 8 min). The etching effect is weak in this experiment because the selected HPT conditions, namely, the power and pressure, are the same as those used during the film deposition. Improved plasma treatment conditions may lead to a more optimized layer structure, and, consequently, to higher values of conductivity.

Additional studies are required for a better understanding of the interactions between the hydrogen plasma and the film’s structure. Optimized p-type nc-Si : H thin films with a thickness below 20 nm can be used as window layers in the amorphous and nanocrystalline solar cells and photodetectors.

4. Conclusions

We presented a growth mechanism of p-type nc-Si : H thin films with its structural properties based on Raman spectroscopy results and multilayer fitting of ellipsometry spectra. Usually, nc-Si : H thin films have an amorphous incubation layer at the substrate-film interface, which is absent in our samples. To explain this difference, we proposed a growth mechanism based on direct inclusion of the Si nanoparticles formed in plasma into the film and interaction of the film surface with heavy and light ions. These growth conditions lead to the formation of a subsurface region with increased amount of voids, which significantly deteriorates the electrical properties of very thin (~20 nm) layers. Moderate (up to 3 min) exposure of these films to hydrogen plasma modifies the Si structure and increases the conductivity from $5 \times 10^{-3}$ to $6 \times 10^{-2}$ (Ω cm)$^{-1}$. These films can be used as window layers both in amorphous and nanocrystalline silicon thin-film solar cells.

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