Highly transparent Nb-doped indium oxide electrodes for organic solar cells

Jun Ho Kim and Tae-Yeon Seong
Department of Materials Science and Engineering, Korea University, Seoul 136-713, South Korea

Seok-In Na
Professional Graduate School of Flexible and Printable Electronics, Chonbuk National University, 664-14, Deokjin-dong, Jeongju-si, Jellabuk-do 561-756, South Korea

Kwun-Bum Chung
Department of Physics, Dankook University, Mt. 29, Anseo-Dong, Chenan 330-714, South Korea

Hye-Min Lee and Han-Ki Kim
Department of Advanced Materials Engineering for Information and Electronics, Kyung Hee University, 1 Seocheon-dong, Yongin, Gyeonggi-do 446-701, South Korea

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The authors investigated the characteristics of Nb-doped In$_2$O$_3$ (INbO) films prepared by co-sputtering of Nb$_2$O$_5$ and In$_2$O$_3$ for use in transparent anodes for organic solar cells (OSCs). To optimize the Nb dopant composition in the In$_2$O$_3$ matrix, the effect of the Nb doping power on the resistivity and transparency of the INbO films were examined. The electronic structure and microstructure of the INbO films were also investigated using synchrotron x-ray absorption spectroscopy and x-ray diffraction examinations in detail. At the optimized Nb co-sputtering power of 30 W, the INbO film exhibited a sheet resistance of 15 $\Omega$/sq, and an optical transmittance of 86.04% at 550 nm, which are highly acceptable for the use as transparent electrodes in the fabrication of OSCs. More importantly, the comparable power conversion efficiency (3.34%) of the OSC with an INbO anode with that (3.31%) of an OSC with a commercial ITO anode indicates that INbO films are promising as a transparent electrode for high performance OSCs. © 2014 American Vacuum Society. [http://dx.doi.org/10.1116/1.4832238]

I. INTRODUCTION

Organic solar cells (OSCs) are creating great interest as a promising alternative to conventional Si-based thin film solar cells due to their cost effectiveness and solution-based simple and fast fabrication process. In particular, bulk-heterojunction OSCs with a blend of poly(3-hexylthiophene) (P3HT) and [6,6]-phenyl-C$_{61}$ butyric acid methyl ester (PCBM) have been mainly investigated and their power conversion efficiency (PCE) has been increased up to 4%–6%. 1–3 To increase the PCE of OSCs, to be acceptable for mass production, most research groups have developed new organic active materials or a new buffer layer, which enhance the exciton formation efficiency or hole/electron extraction efficiency. 4–6 Another effective strategy to increase the PCE of OSCs is to design high-quality transparent anodes with a low resistivity and high transparency for the absorption wavelength of the organic active layer. Because the PCE drop of large area OSCs is critically affected by the series resistance of anode materials, the electrical and interfacial properties of anode materials need to be understood exactly to result in high-performance OSCs. 7–10 Ohmic contact between the anode and organic layer, a low Schottky barrier for hole extraction, and the stability of anode materials against the acidic buffer layer are important criteria for a high quality anode for OSCs. 11 In addition, the transparency of the anode can influence the exciton formation efficiency because the light arrives on the active layer through the transparent anode. Therefore, the development of a new transparent conducting oxide (TCO) films is important as is the development of new organic active layer to improve the performances of OSCs. Although the importance of a TCO anode with a low resistivity and high transparency is well recognized, most OSC researchers in academia or industry still use a sputtered Sn-doped In$_2$O$_3$ (ITO) film as a transparent anode due to its low resistivity (~10$^{-4}$ $\Omega$ cm), high optical transparency (~83% at 550 nm), relatively high work function (~4.8 eV), and easy and fast sputtering process. 12 Considering the variety of oxide materials, the development of high-quality TCO anodes as a substitute for ITO films is important because conventional ITO film has a low transmittance in the near infrared (NIR) wavelength region. 13,14 Among several TCO materials, high mobility TCO (HMTCO) materials, such as Mo, Ti, Ge, and W-doped In$_2$O$_3$ films have been extensively investigated as potential anode materials for OSCs due to their low resistivity and high optical transmittance. 15–18 In particular, HMTCOs have been considered as promising TCOs for tandem-structured photovoltaics because they have a high NIR transmittance as well as in the visible wavelength region. Like other group 5 elements (Ta and V), Nb can act as dopant for In$_2$O$_3$ because the substitution of Nb$^{5+}$ ions into In$^{3+}$ sites effectively produces more excess electrons in Nb-doped In$_2$O$_3$ (INbO) films. Due to the higher amount of excess electrons with the Nb dopant than the Sn dopant in ITO film, TCO anodes with a low resistivity can be realized.
by doping with a small amount of Nb. Although the ionic radius of Nb\(^{5+}\) (0.64 Å) is smaller than that of In\(^{3+}\) (0.79 Å), Nb doping into In\(_2\)O\(_3\) leads to a small lattice deformation due to the low Nb doping weight percentage, unlike Sn (\(~10\) wt. %) dopant in ITO films.\(^{19}\) Gupta et al. reported that a Nb-doped In\(_2\)O\(_3\) film prepared by pulssed laser deposition had a resistivity of 9.61 \(\times\) \(10^{-2}\) Ω cm and a carrier mobility of 65 cm\(^2\)/Vs.\(^{20}\) Recently, epitaxial INbO films prepared by co-sputtering showed a low resistivity of 4.1 \(\times\) \(10^{-4}\) Ω cm and a high optical transmittance of 97% in the visible region.\(^{21}\) However, there are no reports of the application of INbO films as a transparent anode for OSCs or detailed investigation of the Nb doping power effect on the electrical, optical, structural properties, and work function of INbO films.

In this work, we investigated the characteristics of transparent INbO films prepared by a co-sputtering method for application in OSCs as a replacement for conventional ITO films. To optimize the Nb dopant composition, we examined the dependence of the electrical, optical, and structural properties of INbO films on the Nb doping power. In addition, we examined the conduction mechanism of the INbO anode based on low temperature resistivity measurements and the electronic structure of INbO films. Furthermore, we compared the performance of OSCs with the optimized INbO and reference ITO to demonstrate the potential of INbO films as promising anode materials for high performance OSCs.

**II. EXPERIMENT**

To deposit 200-nm-thick INbO films on a glass substrate, we employed a specially designed radio frequency (RF)/direct current (DC) magnetron sputtering system equipped with two tilted cathode guns. Each cathode gun was connected to RF and DC power to sputter the Nb\(_2\)O\(_5\) and the In\(_2\)O\(_3\) targets, respectively. As we previously reported, the co-sputtering process, both Nb\(_2\)O\(_5\) and In\(_2\)O\(_3\) targets attached on tilted cathode guns were presputtered for 30 min to remove the contaminants on the targets. During presputtering, the shutter on the glass substrate can prevent the coating of presputtered Nb\(_2\)O\(_5\) and In\(_2\)O\(_3\) atoms. After presputtering, the INbO films were prepared at a constant DC power of the In\(_2\)O\(_3\) target of 100 W, an Ar\(_2\)O\(_2\) flow rate ratio of 10:0.1 sccm, a working pressure of 3 mTorr, and a Nb\(_2\)O\(_5\) RF power ranging from 0 to 50 W. To investigate the Nb doping power effect on the properties of the INbO films, the DC power applied to the In\(_2\)O\(_3\) target and the thickness of the INbO film were kept constant at 100 W and 200 nm, respectively. During the co-sputtering of the INbO films, the glass substrate was constantly rotated at a speed of 20 rpm to obtain uniformly Nb-doped INbO films. After co-sputtering of the INbO films, all samples were rapidly thermal-annealed at optimized temperature of 600 °C under N\(_2\) (\(~10^{-1}\) Torr) for 10 min. The rapid thermal annealing (RTA) process of the INbO films was conducted using a commercial RTA system (Sintek RTA system) to activate the Nb dopant and crystallize the INbO films. The electrical properties of the INbO films were analyzed by means of Hall measurements (HL5500PC) as a function of the RF power of the Nb\(_2\)O\(_5\) target. The transmittance of the INbO films was measured by a UV/visible spectrometer (UV540). In addition, the resistivity change of the optimized INbO film was measured using a physical property measurement system (Quantum Design) in the temperature range of 2–300 K to investigate the conduction mechanisms of the INbO film. The work function of the INbO film was measured by a Kelvin probe (KP Technology). The structure of the rapid thermal-annealed INbO film was examined by x-ray diffraction (XRD) of the GI-WAXS beam line of the Pohang Light Source (PLS). In addition, the microstructure of the INbO film and interface between INbO and the organic layer were analyzed by high resolution transmission microscopy (HRTEM: JEM-2100F). To investigate the electronic structure near the conduction band of the INbO film, x-ray absorption spectroscopy (XAS) experiments were performed using the total electron yield (TEY) mode at soft x-ray beamline, BL-7A of Photon Factory, Institute of Material Science, High Energy Accelerator Organization (KEK-PF) in Japan.

The feasibility of using INbO films as a transparent anode for OSCs was examined by fabricating OSCs containing the optimized INbO and reference ITO anodes. Based on the conventional P3HT:PCBM-based bulk heterojunction OSCs structure, we compared their short circuit current densities (J\(_{SC}\)), open circuit voltages (V\(_{OC}\)), fill factors (FF), and PCE. Prior to the OSC fabrication, INbO- and ITO-coated glass substrates were cleaned by acetone, methanol, isopropyl, and deionized water, and the surface of the INbO and ITO anode were subsequently UV/ozone-treated for 10 min in order to improve the wetting property of poly(3,4-ethylenedioxythiophene):poly(styrenesulfonate) (PEDOT:PSS) layer. After surface treatment of the INbO- and ITO-coated glass substrates, PEDOT:PSS (Clevios PH510) was spin-coated on the INbO and ITO anodes and subsequently annealed at 120 °C for 10 min in air. Then, a blend of a 50 mg P3HT (Rieke metals) and a 50 mg PCBM (Nano-C) in 1,2-dichlorobenzene (2 ml) was spin-coated onto the PEDOT:PSS layers in a nitrogen-filled glove box. A solvent-annealing treatment, consisting of keeping the spin-coated photoactive films inside a covered glass jar for 120 min, was then performed, and each sample was subsequently annealed at 110 °C for 10 min. The organic layer-coated samples were then transferred to a metal chamber and a Ca/Al (20/100 nm) cathode patterned by a shadow metal mask with an area of 4.65 mm\(^2\) was deposited on the photoactive layer using thermal evaporation. All organic and cathode layers on the INbO and ITO anodes were simultaneously coated by identical coating processes to allow a fair comparison. The photocurrent density–voltage (J–V) characteristics of the OSCs fabricated on the INbO and reference ITO anodes were measured.
using a Keithley 1200 source measurement unit under 100 mW/cm$^2$ illumination with AM 1.5 G irradiation from a solar simulator.

III. RESULTS AND DISCUSSION

Figure 1 shows the Hall measurement results of the INbO films annealed at 600 °C as a function of the RF power of the Nb$_2$O$_5$ target. Figure 1(a) shows the resistivity and sheet resistance of 200-nm-thick INbO films with increasing Nb doping RF power from 0 W (undoped In$_2$O$_3$) to 50 W. All INbO films showed characteristics of degenerate n-type TCO regardless of the Nb doping concentration. Compared to the undoped In$_2$O$_3$ film, the Nb-doped In$_2$O$_3$ films exhibited increased electrical conductivities. An increase of the RF power of the Nb$_2$O$_5$ target from 10 to 30 W led to an increase of the carrier concentration from 2.17 $\times$ 10$^{20}$ to 5.20 $\times$ 10$^{20}$ cm$^{-3}$. In addition, the mobility of the INbO films increased slightly with increasing Nb$_2$O$_5$ RF power from 43.50 (10 W) to 45.40 cm$^2$/V·s (30 W). However, the INbO film showed abruptly decreased carrier mobility at an RF power of 50 W due to impurity scattering or grain boundary scattering. Thus, we calculated the mean free path (MFP) to determine which scattering mechanism is dominant in the INbO films. The MFP of the conducting electrons in the INbO films was calculated by the following equation:

$$l = (3\pi^2)^{1/3} \left(\frac{\hbar}{e^2}\right) \rho^{-1} n^{-2/3},$$

where $\hbar$ is Plank’s constant, $e$ is the electron charge, and $n$ is the carrier concentration. Table I summarizes the MFP values, carrier concentrations ($n$), and resistivities ($\rho$) of the INbO films with increasing RF power. The 30 W Nb-doped In$_2$O$_3$ film showed a MFP value of 6.73 nm, considering the crystallite sizes of the optimized INbO film calculated from XRD measurements of the full-width at half-maximum (FWHM) of 17.15 nm. Thus, electron scattering at the grain boundary is not thought to be an important factor in affecting the mobility of INbO films. As discussed by Lee et al., the carrier mobility in ITO films with a high carrier concentration on the order of 10$^{20}$–10$^{21}$ cm$^{-3}$ is dominated by ionized impurity scattering. Below a Nb doping power of 30 W, the effective activation of the Nb dopant and crystallization of the INbO film simultaneously resulted in increases of carrier mobility and concentration. However, the mobility of the 50 W Nb-doped INbO film showed abruptly decreased carrier mobility due to impurity scattering caused by excessive Nb dopants and segregation. Consequently, we determined that the RF power of 30 W was the optimum doping power to obtain high-quality INbO films for OSCs.

To investigate the degenerated semiconducting properties in the optimized INbO film, we measured the temperature dependency of the 30 W Nb-doped In$_2$O$_3$ film, as shown in Fig. 2. The effective room temperature resistivity of the INbO film was 3.12 $\times$ 10$^{-4}$ Ω cm, which is comparable to a lower resistivity due to the Nb dopant in the In$_2$O$_3$ matrix uniformly doped by the tilted cathode guns. The decreased resistivity of the INbO films with increasing Nb$_2$O$_5$ RF power can be attributed to the increasing carrier concentration, as depicted in Fig. 1(b). The increase of the Nb$_2$O$_5$ RF power from 0 to 50 W led to an increase of the carrier concentration from 2.17 $\times$ 10$^{20}$ to 5.20 $\times$ 10$^{20}$ cm$^{-3}$. In addition, the mobility of the INbO films increased slightly with increasing Nb$_2$O$_5$ RF power from 43.50 (10 W) to 45.40 cm$^2$/V·s (30 W). However, the INbO film showed abruptly decreased carrier mobility at an RF power of 50 W due to impurity scattering or grain boundary scattering. Thus, we calculated the mean free path (MFP) to determine which scattering mechanism is dominant in the INbO films. The MFP of the conducting electrons in the INbO films was calculated by the following equation:

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<table>
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<th>RF power (W)</th>
<th>Resistivity ($\rho$) (x10$^{-4}$ Ω cm)</th>
<th>Carrier concentration ($n$) (x10$^{20}$ cm$^{-3}$)</th>
<th>Mean free path (l) (nm)</th>
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<td>6.73</td>
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<tr>
<td>50</td>
<td>4.08</td>
<td>5.20</td>
<td>4.82</td>
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Table I. Calculated electrical properties including the mean free path value ($l$), resistivity ($\rho$), and carrier concentration ($n$) of the INbO films at various Nb doping powers.
commercial ITO films.29 Like the n-type degenerated ITO film, the INbO film showed a positive dependence on the temperature regardless of sample cooling and heating as a function of temperature from 2 to 300 K, indicating that the INbO film possessed a typical metallic conduction mechanism due to degenerated semiconductor properties. The positive temperature dependence of the resistivity of the INbO film could be explained by carrier scattering in electronic systems.30 As the temperature increased, the conduction electron–phonon scattering increased, which resulted in an increase of the resistivity of the INbO film similar to ITO films. The metallic conductivity observed in the INbO film can be explained by the formation of a degenerate band appearing in heavily doped semiconductors as suggested by Mott.31 Similar observations of high conductivities and a positive temperature coefficient of resistivity have also been reported in other heavily doped TCO systems such as Al-doped ZnO, W-In2O3, Mo-doped In2O3, Ti-doped In2O3, Ge-doped In2O3, and Nb-doped TiO.15–18,32,33

Figure 3(a) shows the transmittances of the Nb-doped In2O3 films annealed at a RTA temperature of 600 °C as a function of the Nb2O5 target RF power in the wavelength range from 200 to 2500 nm. Table II shows the optical transmittances of INbO films at 550 nm and the average transmittance between wavelengths of 500 and 800 nm and the calculated bandgap of the INbO films. In the visible wavelength range, all INbO films showed a high optical transmittance. In addition, at a wavelength of 550 nm, all INbO films showed a high optical transmittance. In addition, at a wavelength of 550 nm, all INbO films showed high optical transmittances above 80%. At a wavelength of 550 nm, the 10 and 20 W doped INbO films show similar optical transmittances of 81.43% and 81.91%, respectively. A further increase of the RF power up to 50 W led to an increase of the optical transmittance at a wavelength of 550 nm. However, the optical transmittance of the INbO film in the NIR wavelength region decreased with increasing RF power. Due to the high carrier concentration, which is comparable to ITO films, the INbO film showed lower NIR transmittances, similar to ITO films. In general, the NIR optical transmittance of TCO films is directly affected by the carrier concentration, because the plasma frequency is directly related to the carrier concentration.34,35 Therefore, as expected from the Hall measurement results shown in Fig. 1(b), the increased carrier concentration in the INbO films with increasing RF power led to the decrease of the NIR transmittance shown in Fig. 3(a). However, considering the absorption wavelength of the P3HT:PCBM active layer, the high optical transmittance of the INbO film could be adapted as transparent anodes for OSCs. To investigate the Nb doping effect on the bandgap of the INbO film, we plotted the optical bandgap of the INbO film determined from the optical transmittance, as shown in Fig. 3(b). The optical bandgaps of the INbO films can be obtained by plotting $x^2$ (absorption coefficient) versus $h\nu$ (photon energy) based on a direct band gap semiconductor.36 Then, we
extended the plot to the photon energy axis by extrapolating the linear region of the plot to $h\nu = 0$ to yield the band gap value.\textsuperscript{19} The optical bandgaps of the 10, 20, 30, and 50 W Nb-doped In$_2$O$_3$ films were 3.72, 3.78, 3.85, and 3.87 eV, respectively. The obtained optical band gaps of the INbO films were larger than that of undoped In$_2$O$_3$ (3.70 eV) because of the Burstein–Moss effect.\textsuperscript{35}

The electronic structure of unoccupied states near the conduction band of the INbO films was analyzed by XAS measurements, as shown in Fig. 4. Renormalizations of the XAS spectra were carefully performed by subtracting an x-ray beam background and scaling of the edge level from the raw data.\textsuperscript{37} The qualitative changes and comparison of the conduction band features, such as the number of unoccupied states, could be analyzed by renormalizations of the XAS spectra.\textsuperscript{38} The normalized oxygen K edge spectra of In$_2$O$_3$ are directly related to the oxygen p-projected states of the conduction band, which consists of unoccupied hybridization orbitals for In 5$p$ + O 2$p$.\textsuperscript{37} The increase of the Nb target plasma power causes modification of the conduction band by incorporating NbO$_x$ molecular structure, attributed to the oxygen 2$p$ states hybridized with Nb 4$d$ and 5$p$ states based on octahedral symmetry.\textsuperscript{39} As the incorporation of NbO$_x$ increased, the conduction band area increased up to a plasma power of 30 W. This demonstrates that the increase of unoccupied states can enhance the charge transport. However, when the plasma power was increased to 50 W, the band edge states and the conduction band states above $\sim$535 eV changed significantly. The increase of the band edge states and decrease of the conduction band can induce an increase of the charge transport scattering and a decrease of the probability for charge transport into unoccupied states, respectively.\textsuperscript{40} These changes of the conduction band and band edge states as a function of Nb doping provide evidence of changes of the resistivity which are related to the modification of electronic structure in the conduction band.

To investigate the effect of RTA on the structure of the optimized INbO film, synchrotron XRD analysis was performed using the GI-WAXS beam line of the PLS. The wavelength of the incident x rays was set to 1.243 Å by a double-bounce Si (111) monochromator. The diffraction peaks could be indexed to cubic bixbyite In$_2$O$_3$ structure (JCPDS card no. 06-0416).\textsuperscript{41} Figure 5 exhibits the synchrotron XRD plot of the undoped and 30 W Nb-doped In$_2$O$_3$ film. The XRD plots of the undoped In$_2$O$_3$ film exhibited several peaks at 2$\theta$ = 21.49° (211), 30.02° (222), 35.50° (400), 50.10° (440), and 59.76° (622), indicative of a typical bixbyite structure. The 30 W Nb-doped In$_2$O$_3$ films showed similar crystalline peaks as the undoped In$_2$O$_3$ film due to their similar lattice constant and polycrystalline structures.\textsuperscript{42} There are no peaks corresponding to the new phase, indicating that the Nb dopant was completely incorporated into the In$_2$O$_3$ lattice due to lower ion radius of Nb$^{5+}$. The 30 W Nb-doped In$_2$O$_3$ film also showed a strong (222) intensity, which indicates that the 30 W Nb-doped In$_2$O$_3$ film maintained a similar bixbyite structure as the undoped In$_2$O$_3$ because of the low doping concentration of the Nb dopant.

Figure 6 exhibits a cross-sectional TEM image of the optimized INbO film in an OSC. The INbO anode showed a uniform thickness because it was fabricated by a co-sputtering method. The annealed INbO film showed a columnar structure with a smooth surface morphology. In addition, it showed a stable interface between the PEDOT:PSS and INbO anode. The enlarged high resolution TEM images in Figs. 6(a)–6(c) show the glass/INbO film interface, a crystalline INbO film, and the INbO/PEDOT:PSS interface, respectively. As shown in Figs. 6(a) and 6(c), the INbO films have a crystalline structure and the fast Fourier transformation (FFT) pattern demonstrates the polycrystalline structure of the INbO films. The diffraction spots in the FFT pattern are scattered because of their orientations and larger grain sizes.\textsuperscript{43–45} The enlarged image in Fig. 6(c) shows the stability of INbO anode against the acidic PEDOT:PSS buffer layer. The INbO anode forms a well-contacted interface with the PEDOT:PSS buffer layer. Due to the stable interface between the PEDOT:PSS buffer layer and INbO anode, it was expected that the hole extraction from the active layer into the INbO anode was improved. The inset shows the FFT pattern obtained from the

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**Fig. 4.** (Color online) XAS measurements of the annealed INbO films as a function of the Nb$_2$O$_5$ RF power.

**Fig. 5.** (Color online) Synchrotron XRD plots of the undoped and optimized INbO films.
PEDOT:PSS and INbO anode. The strong spots in the inset of the HRTEM image confirm the well-developed crystallinity of the INbO films.

Figure 7(a) shows the energy band diagram of the OSC fabricated on the optimized INbO anode as well as a cross-sectional TEM image of the device. The work functions of the optimized INbO and reference ITO anodes were measured by a Kelvin probe. The INbO film showed a higher work function of 5.22 eV than the reference ITO film (4.81 eV). When considering that the energy level between the work function of anode electrodes and the HOMO of donor polymers must be matched for Ohmic contact to improve the built-in potential and hole extraction, it is expected that the INbO film has a desirable work function for use as transparent electrodes in organic devices. To study the feasibility of the use of INbO films as a transparent electrode, we fabricated typical P3HT:PCBM-based heterojunction OSCs, and a reference ITO-based OSC was also fabricated for comparison. Figure 7(b) shows the current density–voltage (J–V) curves of OSCs with the INbO and ITO anodes, and the picture in the inset shows the device structure. As depicted in Fig. 7(b), the OSC fabricated on the typical ITO anode showed a $V_{OC}$ of 0.61 V, a $J_{SC}$ of 9.07 mA/cm$^2$, a FF of 59.80, and a PCE of 3.31%. More importantly, the OSC fabricated on the INbO anode also showed excellent cell-characteristics including a $V_{OC}$ of 0.60 V, a $J_{SC}$ of 8.59 mA/cm$^2$, a FF of 65.13%, and a PCE of 3.34%, which are comparable to the OSC with the reference ITO anode. Although the work function of the INbO anode is higher than that of the ITO anode, both OSCs showed similar $V_{OC}$ values due to the formation of Ohmic contact between the anode and active layer by the use of the conducting PEDOT:PSS buffer layer. Therefore, the comparable performance of the INbO-based OSC to the ITO-based OSC could be attributed mostly to the low sheet resistance of 15 $\Omega$/sq and high transmittance of 86.04% at 550 nm. Based on these results, we can conclude that the INbO anode is a promising replacement for conventional ITO anodes.

IV. SUMMARY AND CONCLUSIONS

In summary, we proposed a low sheet resistance and high transmittance INbO anode prepared by co-sputtered Nb$_2$O$_5$ and In$_2$O$_3$ targets as an alternative to ITO anodes for OSCs. At the optimized Nb$_2$O$_5$ doping RF power of 30 W and RTA at 600 °C, we obtained a high-quality INbO anode with a low resistance of 15 $\Omega$/sq, a high mobility of 45.40 cm$^2$/Vs, a high optical transmittance of 86.04%, and a high work function of 5.22 eV. In addition, we investigated the effect of the Nb doping power on the electrical, optical, and structural properties of the INbO anode. Effective activation of the Nb$^{5+}$ dopant in the In$_2$O$_3$ matrix resulted in an n-type degenerate InbO TCO film with a metallic conduction mechanism which is similar to conventional ITO films. With the aid of the optimized INbO anode, a high-performance OSC with a PCE of 3.34% was achieved, which is comparable to that of the OSC fabricated on a commercial ITO anode (3.31%). This result indicates that the INbO film is a high-quality In$_2$O$_3$-based TCO anode material and may be a substitute for conventional ITO films.
ACKNOWLEDGMENTS

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