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# Effect of growth parameters on the properties of RF-sputtered highly conductive and transparent p-type NiO<sub>x</sub> films

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#### Abstract

Highly conductive and transparent  $NiO_x$  films can be very useful as buffer layers for the optimization of the p-type contacts of optoelectronic devices. Thin  $NiO_x$  films were fabricated by reactive radio frequency (RF) sputtering at room temperature starting from a Ni target. A systematic study of the influence of oxygen partial pressure, RF power and sputtering gas pressure on the films' properties was carried out. The structural, microstructural, optical and electrical properties were affected differently by the sputtering parameters. Resistivity decreased by increasing the oxygen partial pressure and the sputtering total pressure and by decreasing the RF power, while transmittance increased by decreasing the oxygen partial pressure and by increasing the RF power and sputtering pressure. Minimum resistivity of  $1.6 \times 10^{-2} \,\Omega$ cm and a visible transmittance of 40% were achieved for a film grown in a pure oxygen atmosphere, while a higher transmittance of 54% and a resistivity of  $\rho = 1.1 \times 10^{-1} \Omega$ cm were obtained for a film grown at 30% oxygen partial pressure. The trends of transmittance and resistivity as a function of the oxygen pressure during the sputtering process can be explained in terms of the amount of Ni<sup>3+</sup> defects deduced by x-ray photoelectron spectroscopy (XPS) measurements. The full interpretation of the other results is less straightforward and highlights the relevance of the samples' structural properties.

Keywords: nickel oxide, p-type transparent conductive oxide, radio frequency sputtering, optical properties, electrical properties

(Some figures may appear in colour only in the online journal)

#### 1. Introduction

P-type transparent conductive oxides (TCOs) offer tremendous opportunities in the emerging branch of optoelectronics called invisible electronics. Applications are wideranging; they include a hole transport layer for anodes in organic light emitting devices (OLEDs) and solar cells, transparent lighting (window integrated OLEDs), fully transparent displays, optical or chemiresistive gas sensors, optical switching devices, etc. The fabrication of high-performing p-type transparent conductive oxides is very difficult, even using a high-temperature process, and this has limited the use of n-type TCOs to the role of transparent electrodes, preventing their use in transparent p-n junction-based TCOs.

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Several p-type TCOs have been proposed, including ZnO: Mg, ZnO:N, ZnO:In, NiO, NiO:Li, CuAlO<sub>2</sub>, Cu<sub>2</sub>SrO<sub>2</sub>,  $CuYO_2$ :Ca, and  $CuGaO_2$  thin films [1, 2]. Since the discovery of copper aluminum oxide CuAlO<sub>2</sub> [3], progress has been made in developing p-type materials, and some examples of p-n junctions have also been proposed [4-6]. However, only a few examples of high-conductivity p-type films have appeared in the recent literature, such as Ni<sub>x</sub>Co<sub>3-x</sub>O<sub>4</sub> films with a conductivity of  $375 \,\mathrm{S \, cm^{-1}}$  [7], LaCuOSe epitaxial films with a conductivity of  $910 \,\mathrm{S \, cm^{-1}}$  [8], BaCuSF films with a conductivity of  $260 \,\mathrm{S \, cm^{-1}}$  [9],  $\mathrm{Bi_2 Sr_2 CoO_2 O_v}$  films with a conductivity of  $182 \text{ S cm}^{-1}$  [10] and NiO films with a conductivity of 100 S cm<sup>-1</sup> [11]. Although stoichiometric NiO is an insulator at room temperature with a resistivity of  $\sim 10^{13} \Omega$  cm, p-doped NiO may be obtained by introducing acceptor defects such as nickel vacancies and oxygen interstitials and/or by intentional doping with Li, K and Cu ions [12-23]. Resistivity values reported in the literature for transparent NiO<sub>x</sub> films range from  $10^4 \Omega$ cm to  $10^{-2} \Omega$ cm [11, 12, 23], and several methods have been used for growing NiO films, including sputtering [11, 12, 14, 24], e-beam evaporation [25, 26], chemical methods [27, 28], pulsed laser deposition [29], spray pyrolysis [30, 31] and plasmaenhanced chemical vapor deposition [32, 33]. Among the above-mentioned techniques, sputtering is one of the preferred due to its industrial scalability.

In this work, thin non-stoichiometric NiO films  $(NiO_x)$  with a resistivity down to  $\sim 10^{-2} \Omega$ cm were obtained by radio frequency (RF) sputtering at room temperature. Film properties were analyzed not only as a function of a single growth parameter (the effects of oxygen partial pressure or substrate temperature are the most studied in the literature), but as a function of different sputtering parameters such as RF power, total pressure and oxygen partial pressure. A correlation between the growth parameters and the electrical, optical and structural properties of the films was established. The ageing of the film was also discussed.

#### 2. Methods

NiO<sub>x</sub> films were grown on glass or fused silica substrates by RF sputtering in an  $Ar + O_2$  atmosphere starting from a 6.5 inch diameter Ni target of 99.99% purity. A base pressure of about  $2 \times 10^{-4}$  Pa was ensured in the deposition chamber by a turbomolecular pump. The deposition conditions were varied systematically aiming at the maximization of both visible transmittance and film conductivity. Several films were grown at different deposition parameters such as RF power, total pressure and oxygen partial pressure, while the substrate temperature was fixed at 300 K and the target-tosubstrate distance was fixed at 70 mm. The oxygen partial pressure  $p(O_2)/p(Ar + O_2)$  in the sputtering gas was varied in the range 2.5–100%, the RF power in the range 50–400 W and the total sputtering pressure  $p(Ar + O_2)$  in the range 0.4– 1.5 Pa. Thick films (about 200 nm) were grown as a reference to determine the growth rate for each set of deposition parameters. Their thickness was measured by a surface



**Figure 1.** Transmittance curves of 35 nm thick NiO<sub>x</sub> films grown at 0.9 Pa total gas pressure, 66% oxygen partial pressure and different RF powers (set A).

profilometer KLA-Tencor P-10 with an error of about 5%. From these data the sputtering times needed to obtain 35 nm films were deduced. The transmittance and reflectance of the films were measured by a Lambda 950 spectrophotometer and transmittance curves were normalized to the transmittance of the substrate. The sheet resistance of the films was measured by the Van Der Pauw method, and the mobility and carrier concentration were obtained by the Hall method using a HP Profiler Biorad HL 5900. The structural characterization of the films was performed by x-ray diffraction using a Philips X-Pert Pro 500 Diffractometer with Cu K $\alpha$  radiation (1.5418 Å) in  $\theta$ -2 $\theta$  configuration. The film morphology was observed by a field emission scanning electron microscope Leo Supra.

The films were also analyzed by x-ray photoelectron spectroscopy (XPS) with a Vacuum Generator ESCA LAB MKII system using the AlK $\alpha$  line as an x-ray source.

#### 3. Results

#### 3.1. Optical and electrical properties

While stoichiometric NiO is a transparent and insulating material, the transition to defect-rich non-stoichiometric NiO<sub>x</sub> (x > 1) makes the material semitransparent and conductive. The aim is to find the deposition conditions which give high conductivity films without decreasing too much the optical transparency.

Table 1 summarizes the growth parameters and some properties of the investigated NiO<sub>x</sub> samples. Three sets of samples (A, B, C) of approximately 35 nm thickness were fabricated at room temperature. In set A the RF power was varied in the range 50–400 W, while the sputtering pressure was kept constant at 0.9 Pa and the oxygen partial pressure was kept constant at 66% (0.6 Pa). Samples of set B were prepared at a fixed RF power of 250 W, a fixed Ar partial pressure of 0.3 Pa and an oxygen partial pressure variable



**Figure 2.** Sheet resistance  $(\bigcirc)$  and transmittance  $(\blacksquare)$  of NiO<sub>x</sub> films (set A) versus RF power.

from 30 to 66% (0.1–0.6 Pa); one more sample was grown at 250 W in a pure oxygen atmosphere (0.6 Pa). In set C the total sputtering pressure was varied in the range 0.4–1.5 Pa at a fixed RF power of 250 W and a fixed oxygen fraction of 30%.

Figure 1 shows the spectral transmittance of the films of set A grown at a fixed oxygen partial pressure of 66%, a fixed sputtering pressure of 0.9 Pa and a RF power varying in the range of 50–400 W. Film transmittance and sheet resistance increase upon increasing the RF power (figure 2). These results are in accordance with the results from Chen *et al* [14] who report an increase of film resistance upon increasing the RF power from 100 to 200 W.

Samples of set B were prepared at a fixed RF power of 250 W, a fixed Ar partial pressure of 0.3 Pa and an oxygen partial pressure variable from 30 to 100% (pure oxygen atmosphere). Preliminary studies performed on samples grown at a lower oxygen partial pressure of 2.4% (minimum adjustable partial pressure in our system) showed that such a low oxygen pressure is sufficient to grow NiO<sub>x</sub> films (without the Ni phase) starting from the metal target. The sample grown at 2.4% oxygen pressure showed a slight increase of transmittance against a higher resistivity value, if compared to the films grown at higher oxygen pressures; it will therefore not be considered in the following.

Figure 3 shows the transmittance curves of the NiO<sub>x</sub> films of this set. We can observe a decrease of transmittance upon increasing the oxygen content, in accordance with the literature data [17, 18]. The sheet resistance of the films also decreases with an increasing oxygen content, in accordance with the literature data [16, 18, 21, 34], and reaches a minimum value of  $4.8 \text{ k}\Omega/\text{square}$  (resistivity  $\rho = 0.016 \Omega$ cm) for a film grown in a pure oxygen atmosphere, as shown in figure 4. It is worth noting that this resistivity value is among the lowest found in the literature for semitransparent NiO<sub>x</sub> films. The film grown at 30% oxygen content shows the best visible transmittance of about 50%, and a resistivity of about 0.18  $\Omega$ cm.

In set C the total sputtering pressure was varied in the range 0.4–1.5 Pa, at a fixed RF power of 250 W and a fixed

oxygen fraction of 30%. The transmittance of the films increases with increasing total pressure (figure 5), while the sheet resistance decreases between 0.4 and 1.2 Pa. An increase in film resistance is observed at 1.5 Pa (figure 6), followed by a fast degradation in time. Results of the electrical and optical measurements on the films of set C show that, at fixed RF power and oxygen partial pressure, the sputtering pressure giving the highest transmittance and the lowest resistance of the films is 1.2 Pa.

In order to estimate and compare the performances of all films we used the figure of merit *F* defined as  $F = 1/[R_s \ln (T + R)]$ , where  $R_s$  is the sheet resistance of the film, and *R* and *T* are respectively the reflectance and transmittance of the films in the visible region. We have chosen this figure of merit among the others [35, 36] because it takes into account also the visible reflectance of the films, which is up to 25% in our samples. The values of *F* range from 26 (M $\Omega$ )<sup>-1</sup> to 394 (M $\Omega$ )<sup>-1</sup> (see table 1). The film with the highest figure of merit (NiO\_7) was grown at 250 W in a pure oxygen atmosphere at a pressure of 0.6 Pa.

Samples of set B were re-deposited on polished quartz substrates to better evaluate the optical and electrical properties as a function of the oxygen content. The optical band gap  $E_G$  of the NiO<sub>x</sub> films was extrapolated from the Tauc's plot (figure 7(a)) on the basis of the following equation, valid for semiconductors with a direct band gap:  $\alpha h\nu = A$  $(h\nu - E_G)^{1/2}$ , where A is a constant,  $h\nu$  is the photon energy and  $\alpha$  is the absorption coefficient. The absorption coefficient was calculated from the transmittance (T) and reflectance (R) curves of the films with the approximated formula  $\alpha = 1/d \ln [(1 - R)^2/T]$ , where d is the film thickness. As shown in figure 7(b),  $E_G$  decreases slightly from 3.30 to 3.18 eV by increasing the oxygen fraction in the sputtering gas.

Sheet resistance and Hall measurements were performed on samples deposited at different oxygen partial pressures in order to measure the resistivity, carrier concentration and Hall mobility (table 2). Repeating the measurement, the resistivity values were perfectly reproducible while the values of the Hall coefficient were affected by a large casual error. Therefore carrier concentrations and the Hall mobility values must be taken just as an order of magnitude. Even with this limitation, a clear p-type conductivity is found (as confirmed also by hot point probe measurements).

Stability tests were carried out at room temperature on samples stored in an ambient and low-vacuum atmosphere. It is in fact known that the electrical properties of NiO<sub>x</sub> films are affected by ageing [14, 21, 37]; this is due to the release of structural defects and residual stresses in the films and also to the absorption/desorption of oxygen in the film [14]. Ageing is observed mainly in the 35 nm thick films, while the thicker ones are much more stable. Almost all thin films showed a slow increase of the resistance in time, with the exception of the film grown at 1.5 Pa which showed a very fast increase of resistance (almost one order of magnitude after one week of exposure to the ambient atmosphere). The resistance measured on the 35 nm films only doubled after one year of storage in a low vacuum at  $10^{-2}$  mbar, while transmittance increased slightly. For films stored for one year in an ambient

**Table 1.** Deposition parameters and selected properties of NiO<sub>x</sub> films grown at room temperature on glass substrates. Here  $P_{\text{RF}}$  is the RF power, *p* is the pressure of the sputtering gas, *T*(550 nm) is the transmittance at 550 nm,  $R_{\text{s}}$  is the film sheet resistance and *F* is the figure of merit.

Sample	P <sub>RF</sub> (W)	Set	$p(Ar + O_2)$ (Pa)	<i>p</i> (O <sub>2</sub> ) (Pa)	$p(O_2)/p(Ar + O_2)$ (%)	Dep. rate $(nm min^{-1})$	T(550 nm) (%)	$R_{ m s}$ (k $\Omega/ m sq$ )	F $(M\Omega)^{-1}$
NiO_1	50	А	0.9	0.6	66	0.5	39	5.8	311
NiO_2	150	А	0.9	0.6	66	1.6	42	9.2	222
NiO_3	250	A, B	0.9	0.6	66	2.8	44	12.4	168
NiO_4	400	А	0.9	0.6	66	4.3	55	36.4	85
NiO_5	250	B, C	0.4	0.1	30	2.9	50	56.0	47
NiO_6	250	В	0.6	0.3	50	2.6	44	24.4	96
NiO_7	250	В	0.6	0.6	100	2.1	39	4.8	394
NiO_8	250	С	0.8	0.2	30	3.3	52	35.6	75
NiO_9	250	С	1.2	0.4	30	3.4	54	31.2	93
NiO_10	250	С	1.5	0.5	30	3.3	58	140	26



**Figure 3.** Transmittance curves of 35 nm thick  $NiO_x$  films grown at 250 W and different oxygen partial pressures (set B).



**Figure 4.** Sheet resistance  $(\bigcirc)$  and transmittance  $(\blacksquare)$  of NiO<sub>x</sub> films (set B) versus oxygen partial pressure.



**Figure 5.** Transmittance curves of 35 nm thick NiO<sub>x</sub> films grown at 250 W, 30% oxygen partial pressure and different total pressures (set C).



**Figure 6.** Sheet resistance  $(\bigcirc)$  and transmittance  $(\blacksquare)$  of NiO<sub>x</sub> films (set C) versus total pressure.



**Figure 7.** (a) Tauc's plot of  $NiO_x$  samples (set B) to extrapolate the band gap energy. (b) Band gap energy versus oxygen partial pressure.

atmosphere the ageing was much faster, with up to 100 times an increment of resistivity and about 25% increment in transmittance. The simplest explanation of these results is that the excess oxygen contained in the near-surface region can be released even at room temperature allowing the films to recover a more stoichiometric composition and therefore more insulating characteristics. However, after one year the resistivity of all the samples remained below 11  $\Omega$ cm, which is still a low value as compared to many data reported in the literature for NiO films. The ageing effect was considerably slowed down, but not completely eliminated, when such films were coated by an ultrathin (2 nm) Ni film as a capping layer. From these results, it is clear that the very thin films cannot be used as an outer layer exposed to the ambient atmosphere, but they could be used inside a multilayer stack, able to block the oxygen loss. To test the possible film degradation due to a subsequent overlayer deposition we heated them in a high vacuum at 190 °C for 2 h. The samples did not show any dramatic degradation and their sheet resistance showed an increase of less than 20%.

#### 3.2. Structural and microstructural properties

For all the measured thin samples, the XRD diffraction patterns show an amorphous structure with early stage weak and broad peaks (figure 8(a)). To better investigate the crystallinity of the films, the XRD measurements were performed on the corresponding set of thicker samples (figure 8(b)). These samples show an increased crystallinity, with varying intensities of (200), (111) and (220) peaks (JPCD 78-0643 corresponding to cubic NiO). In particular, the sample grown at 1.5 Pa and 250 W shows a high crystallinity and a preferential (111) orientation, while the sample grown at 400 W shows well-defined peaks but no preferential orientation. The higher the oxygen content, the lower the crystallinity, in accordance with the literature data [16, 38]. The study of XRD peaks is often used to reveal the presence of point defects which deform the lattice structure, but unfortunately in our case the low crystallinity of the thin films did not allow a reliable peak analysis. Peak analysis was therefore performed only on the thicker films, and showed for the analyzed samples an increase of the lattice constant with respect to the



Figure 8. X-ray diffraction patterns of the representative 35 nm thick (a) and 200 nm thick (b)  $NiO_x$  films grown in different conditions.

Sample	$p(O_2)/p(Ar + O_2)$ (%)	$R_{\rm s}~({\rm k}\Omega/{\rm sq})$	ρ (Ωcm)	$p ({\rm cm}^{-3})$	$\mu \ (\mathrm{cm}^2  \mathrm{V}^{-1}  \mathrm{s}^{-1})$
$NiO_5_q$	30	91.7	0.32	$6.6 \times 10^{19}$	0.3
$NiO_6_q$	50	47.1	0.16	$4.7 \times 10^{20}$	0.1
$NiO_3_q$	66	27.7	0.096	$3.1 \times 10^{20}$	0.2
$NiO_7_q$	100	13.1	0.046	$1.1 \times 10^{21}$	0.1

standard value (4.176 Å), indicating the presence of residual compressive stresses on the films' surfaces.

The SEM micrographs in figure 9 show that the 35 nm thick  $NiO_x$  samples are characterized by a fine structure consisting of a homogeneous distribution of small grains of a dimension of less than 20 nm, which in most of the samples are round shaped. No significant influence of the investigated sputtering parameters on the films' morphology was detected.

#### 3.3. XPS studies

XPS analysis was conducted on fresh samples to limit the influence of surface contamination and/or of the release of oxygen from the films' surfaces. Due to the preferential sputtering of oxygen, we decided not to sputter our films to avoid changes in their stoichiometry. For all the investigated films, the XPS spectra of Ni 2p clearly shows two components due to Ni<sup>2+</sup> (NiO) at 854.5 eV and to Ni<sup>3+</sup> (Ni<sub>2</sub>O<sub>3</sub>) at 855.8 eV [39] (figures 10(a) and (b)).

The Ni<sup>3+</sup> peak can be associated both to the formation of a Ni<sub>2</sub>O<sub>3</sub> phase and to a large density of point defects. It has to be noted that no peaks related to the Ni<sub>2</sub>O<sub>3</sub> phase were detected in the XRD spectra suggesting that, if present, Ni<sub>2</sub>O<sub>3</sub> occurs in an amorphous form.

As shown in figures 10(a) and (b), the Ni 2p background spectra is very large and may cause uncertainty in the peak-fitting procedure, therefore O 1s fitting was preferred, even if it may overestimate the oxygen percentage in the film due to oxygen contamination. The fitting of O 1s peaks (figures 10(c) and (d)) reveals an increasing Ni<sup>3+</sup>/Ni<sup>2+</sup> ratio by increasing the oxygen partial pressure during the growth process (figure 11). The oxygen pressure seems to influence the concentration of Ni<sup>3+</sup> more than the sputtering power, as confirmed by the comparison between the XPS data obtained from two samples grown at oxygen partial pressures of 66% and 250 W (square in figure 11) and 400 W (triangle).

#### 4. Discussion

The low resistivity of the films grown under oxygen-rich growth conditions is obviously due to their non-stoichiometry.

Many authors claim that the point defects responsible for the non-stoichiometry are nickel vacancies and not oxygen interstitials. For instance, Lany *et al* justify the assertion by the much larger formation energy of interstitials with respect to the vacancies [40]. However the highly-energetic sputtering process can produce defects which would be negligible in equilibrium conditions, such as oxygen interstitials. In a simplified picture the formation of one Ni vacancy is often described as being accompanied by the transition of two neighbor  $Ni^{2+}$  lattice ions into two  $Ni^{3+}$  ions [41, 42]:

$$2Ni^{2+} + \frac{1}{2}O_2 \to 2Ni^{3+} + O_0^{2-} + V_{Ni}, \qquad (1)$$

where  $V_{\text{Ni}}$  is a nickel vacancy,  $O_{\text{O}}^{2-}$  is a lattice ion and two additional holes required to maintain the charge balance are localized on Ni sites. A similar reaction could describe the interstitial formation by simply removing the  $V_{\text{Ni}}$  term.

In this simplified picture the holes coming from the ionization of vacancies or interstitials are supposed to be localized on Ni sites. If the holes localization length is of the order of the lattice constant we have a 'small polaron' strongly bound and the conduction proceeds by hopping [43]. If the hole localization length is much larger than the lattice constant we have a 'large polaron' which can be thermally excited producing a band-like transport. In the former case the mobility is extremely low ( $\mu \ll 0.1 \text{ cm}^2/(\text{Vs})$ ) and it is thermally activated [44], while in the latter case the mobility is larger and the temperature dependence of the conductivity is mainly due to the carrier concentration. It is still controversial which model applies to NiO. Some works [45] point to the small polaron model while others, performed on more conductive samples, found evidence for the large polaron one [46].

Our measurements give mobility values larger than  $0.1 \text{ cm}^2/(\text{Vs})$ , suggesting that our  $\text{NiO}_x$  films are band conductors, and not small polaron conductors. The holes density is much lower than the density of Ni vacancies which could be deduced from the XPS data as a result of a small ionization probability, but it is high enough to give a degenerate behavior. The low-mobility values could be due to a 'large polaron' hole effective mass, to the amorphous/nanocrystal-line nature of the film and/or to a high effective surface area of films, as suggested by the SEM analysis. This explains why a very high p-type conductivity cannot be obtained even with a high concentration of free carriers.

The increase in the oxygen partial pressure leads to an increase of the carrier density measured by the Hall effect (see table 2) and to a decrease of the sheet resistance (see figure 4), in agreement with the increase of the acceptor Ni vacancies measured by XPS and in accordance with the literature [34, 47]. Also the band gap narrowing at a higher oxygen partial pressure shown in figure 7 can be explained by the higher doping level due to an increase in point defects at higher oxygen content.

However the  $Ni^{3+}/Ni^{2+}$  ratio is not the only factor influencing the film properties. In fact the increase of visible

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**Figure 9.** SEM micrographs of representative 35 nm thick NiO<sub>x</sub> samples grown at the different sputtering conditions. Films grown at: (a) 30% oxygen partial pressure (NiO\_5); (b) 66% oxygen partial pressure (NiO\_3); (c) 100% oxygen partial pressure (NiO\_7); (d) 1.5 Pa sputtering pressure (NiO\_10); (e) 50 W RF power (NiO\_1); (f) 400 W RF power (NiO\_4).

transmittance and resistance with increasing RF power shown in figures 1 and 2 would suggest a decrease of defects in the films. However the XPS results reported in figure 11 do not show a significant decrease of the  $Ni^{3+}/Ni^{2+}$  ratio when comparing the film grown at 400 W and the film grown at 250 W. This suggests a possible role of other kinds of defects or structural disorder.

The same conclusion is drawn looking at figures 5 and 6, where the transmittance and sheet resistance of the films show opposite trends which cannot be simply correlated to the



Figure 10. XPS spectra of samples grown at different oxygen partial pressures: (a) and (b) Ni 2p peaks and (c) and (d) O 1s peaks and fitting.



**Figure 11.** Ni<sub>2</sub>O<sub>3</sub>/NiO ratio versus oxygen partial pressure in selected samples grown at 250 W ( $\blacksquare$ ) and 400 W ( $\triangle$ ).

variation of the number of Ni vacancies inside the samples; this matter is still under investigation.

The higher and unstable electrical resistivity of the sample grown at the highest pressure (sample NiO\_10) may be instead correlated to the (111) preferential orientation, observed both in thick and thin films (figure 8), in accordance with data from the literature which describe the (111) planes as thermodynamically unstable [37, 48].

#### 5. Conclusions

Transparent and highly-conductive p-type NiOx films were fabricated by reactive RF sputtering starting from a Ni target at room temperature. The influence of different growth parameters on the films' performances was investigated. Results indicate that high electrical conductivity can be obtained with high oxygen partial pressure, low RF power and high sputtering pressure (up to 1.2 Pa), the oxygen partial pressure being the dominant factor. An optimal combination of growth parameters led to highly conductive and transparent films. A minimum resistivity of  $1.6 \times 10^{-2} \Omega$ cm was achieved for a film with  $T_{550nm} = 40\%$  grown in a pure oxygen atmosphere, while the highest optical transparency  $(T_{550nm} = 54\%)$  and a resistivity of  $1.1 \times 10^{-1} \Omega$ cm were obtained for a film grown at an oxygen partial pressure of 30%. The main source of the p-type conductivity of the films has been attributed to Ni vacancies, but other factors (oxygen interstitials, additional defects, disorder, grain size) are certainly involved. The mobility values suggested band-like conductivity rather than small polaron hopping.

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