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# Study of the oxygen vacancies changes in SnO<sub>2</sub> polycrystalline thick films using impedance and photoemission spectroscopies

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Changes in the concentration of oxygen vacancies within the grains of polycrystalline SnO<sub>2</sub>, due to different atmosphere exposures, were detected using impedance and photoemission spectroscopies. From measured capacitance values, variations of the potential barrier widths could be determined. It is shown that under the presence of an oxygen rich atmosphere, at relatively low temperature, the width of intergranular potential barriers increase to the point that grains become completely depleted of carriers. With subsequent exposure to vacuum, capacitance adopts a higher value, indicative of intergranular barriers and quasi-neutral regions at the center of the grains. X-ray and ultraviolet photoemission spectroscopy measurements showed that SnO<sub>2</sub> samples treated in oxidizing or reducing environments have similar barrier heights and different work functions. Results are especially relevant in the study of mechanisms responsible for metal oxide gas sensing. © 2014 AIP Publishing LLC. [<http://dx.doi.org/10.1063/1.4902150>]

## I. INTRODUCTION

Semiconducting metal oxide sensors, like SnO<sub>2</sub>, present characteristics such as sensitivity, long-term stability, robustness, and low price, which make them extremely valuable in the field of gas sensing. Gas sensors based on oxide semiconductors detect gases as electrical impedance changes upon gas adsorption and diffusion in a porous layer.<sup>1-3</sup> Consequently, the ceramic gas sensor performance depends on the porous microstructure, and on the surface reactivity with the gas. Then, as the sensitivity of these sensors depends on the film microstructure and on the substrates shape, the fabrication process is crucial in improving the sensors response.

It is widely accepted that barriers formed between particles or grains have a Schottky-type nature and that they govern the electrical behavior of the sensor. Adsorption of gaseous species at the grain boundaries can induce changes in the barriers heights and in the donor concentrations. Many oxide semiconductors have a large number of oxygen vacancies conferring their n-character. Chemisorbed oxygen species (O<sup>-2</sup> and O<sup>-</sup>) act as electron acceptors and lead to the formation of a depletion layer extending into the grains forming surface barriers.<sup>4</sup> These surface barriers play a key role as they control electron transfer between grains, affecting the overall resistance.<sup>5-8</sup>

We focus here on oxygen vacancies as they directly affect conductivity and therefore the film gas detection ability. In fact, oxygen vacancies, being the main dopant in tin oxide and other semiconducting oxides, are also relevant in other related issues. For example, it has been argued that the

degradation in metal oxide varistors is controlled by a defect electro-migration process of defects.<sup>9</sup> Degradation phenomena can appear under continuous, alternating, or pulses of voltage usually combined with thermal treatments.<sup>10,11</sup> Also, it has been shown that degraded ZnO varistors can recover their properties under an oxygen atmosphere.<sup>12</sup> These type of experiments can be explained as the result of the deformation of Schottky barriers present in the grain boundaries in which oxygen vacancies play a central role.<sup>13-18</sup>

In this work, thick films of SnO<sub>2</sub> were made using the screen-printing method onto alumina substrates on which electrodes with an interdigitated shape were deposited by sputtering. Impedance and photoelectron spectroscopies were used to characterize the intergranular potential barriers. Results indicate that the oxygen vacancy density varies within the grains as films are exposed to different atmospheres at relatively low temperatures. A detailed study of the capacitance measurements indicates that under the presence of an oxygen rich atmosphere, the dopant density drops to the point that barriers overlap and then grains become completely depleted of carriers. X-ray and UV Photoelectron Spectroscopy (XPS and UPS) measurements allow us to determine changes in barrier heights and in work functions of samples with different gas pre-treatments.

## II. EXPERIMENTAL

Thick, porous film samples were made by painting onto insulating 96% dense alumina substrates on which electrodes with an interdigitated shape were deposited by sputtering. Alumina substrates were cleaned with acetone and isopropyl alcohol, blown with dry nitrogen, and placed in a sputtering chamber. After pumping the chamber to its base pressure

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( $10^{-6}$  mbar), an adhesion layer consisting of 25 nm of titanium was deposited. Without breaking vacuum, a platinum film 200 nm thick was then deposited over the Ti layer. For defining the electrodes, the substrates with the metal films were placed in a home-built micromachining system consisting of three translation stages and a 20 W, frequency tripled Nd:YVO<sub>4</sub> laser ( $\lambda = 355$  nm). The interelectrode distance was 20  $\mu\text{m}$ . Fig. 1(a) shows a picture of the used type of substrate and Fig. 1(b), a detail of the interelectrode.

Commercial high-purity SnO<sub>2</sub> (Aldrich 99.99%, medium particle size 0.4  $\mu\text{m}$ ) was ground until a medium particle size of 0.130  $\mu\text{m}$  was obtained as determined using the intersection method from SEM images. To image the tin oxide surfaces, a JEOL JSM 6460-S scanning electron microscope was employed. In Fig. 2, we present a micrograph of the film showing a porous structure as required for sensors. The grain size is crucial for the electrical properties, as previously discussed in detail. In order to evaporate the binder, the resulting powder was heated up to 380 °C and maintained at this temperature for 1 h, using a heating rate of 1 °C/min. Later, the powder was cooled to 25 °C and a paste was prepared with an organic binder (glycerol). The used solid/organic binder ratio was 1/2, and no dopants were added.

After painting, samples were thermally treated for 1 h in air at 100 °C. This treatment was performed in order to evaporate the organic binder and to improve the adhesion of the films on the alumina substrate. Later, samples were thermally treated up to 380 °C using a heating rate of 1 °C/min and exposed to an air atmosphere at 380 °C during 1 h. After thermal treatment, the film was cooled down to 25 °C. Finally, some of the samples were kept at 380 °C for 4 h in a N<sub>2</sub> atmosphere with 5% H<sub>2</sub>. Samples were labeled S<sub>O2</sub> (samples only exposed to dry air) and S<sub>H2</sub> (samples exposed to hydrogen). Samples were placed in the measuring cell, which was then evacuated. The average thickness of the film

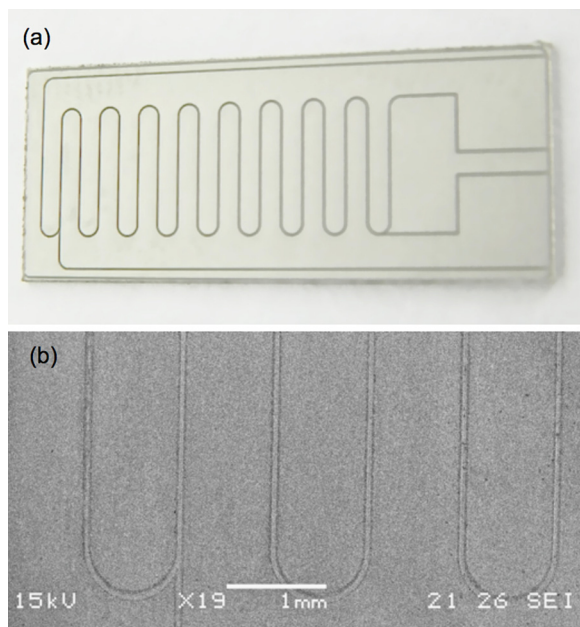


FIG. 1. (a) Substrate of 20 mm length and 10 mm wide used in the experiments. (b) Detail of the interelectrode of 20  $\mu\text{m}$ .

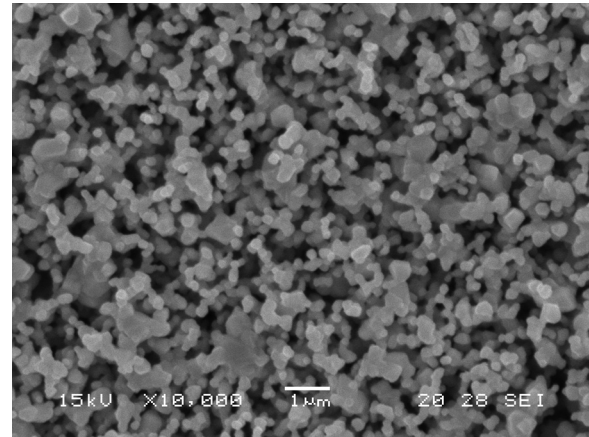


FIG. 2. SEM micrograph of the film showing a porous structure with an average grain size of 0.130  $\mu\text{m}$  as determined using the intersection method.

was 0.105 mm as shown in Fig. 3 in which a micrograph of the cross section is presented.

An X-ray powder diffraction (XRD) analysis was carried out with a Philips (PW1830) diffraction system employing CuK $\alpha$  radiation at 40 kV and 40 mA, Fig. 4. Samples were scanned between 20° and 80° with a step size of 0.01°. The XRD data were analyzed using the X'Pert PRO HighScore software. Our samples showed a single phase corresponding to tin oxide reference code 01-072-1147 of the International Centre for Diffraction Data (ICDD) (1998) Powder diffraction file database, Newtown Square (USA).

The resistance was measured, while raising and then decreasing the temperature from room temperature up to 350 °C at a rate of  $\sim 1$  °C/min with the sample kept in vacuum. Measurements were made after reaching steady state at different temperatures of the experiment range. An Agilent

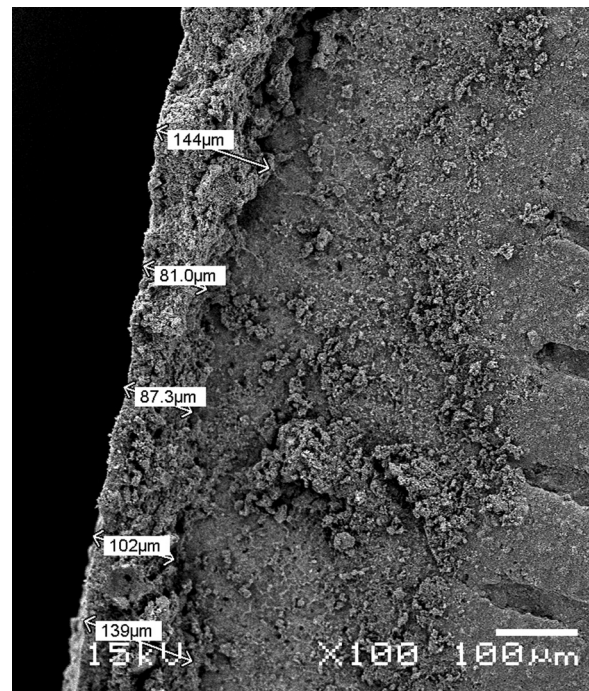


FIG. 3. Micrograph of the cross section of the film that presents an average thickness of 0.105 mm.



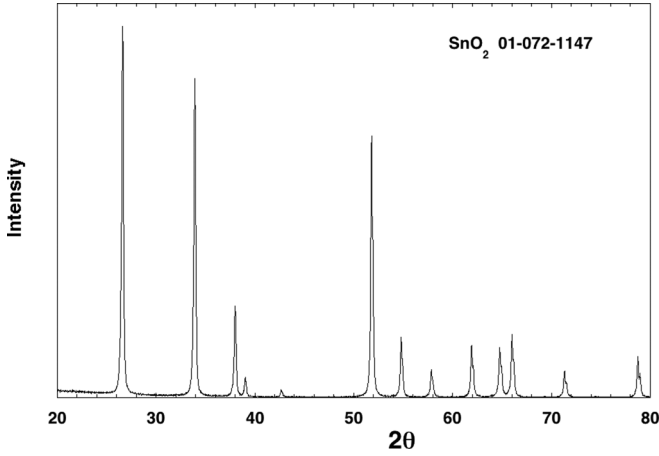


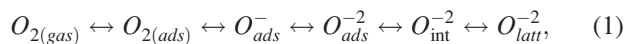
FIG. 4. High quality X-ray diffraction pattern in the range  $20^\circ$ – $80^\circ$  ( $2\theta$ ). A tin oxide tetragonal crystalline system is only observed in the powder used in our samples.

3440 A multimeter was used for the electrical conductance measurements. Finally, capacitance vs. frequency, measurements were carried out with a Hewlett Packard impedance analyzer model 4184 A with samples exposed to vacuum at different temperatures.

XPS and UPS measurements were performed under ultra high vacuum conditions (base pressure  $< 5 \times 10^{-10}$  mbar) in a SPECS spectrometer system equipped with a 150 mm mean radius hemispherical electron energy analyzer and a nine channeltron detector. XPS spectra were acquired at a constant pass energy of 20 eV using an un-monochromated  $\text{MgK}\alpha$  (1253.6 eV) source operated at 12.5 kV and 20 mA and a detection angle of  $30^\circ$  with respect to the sample normal on grounded conducting substrates. The binding energies quoted are referred to the  $\text{Au } 4f_{7/2}$  emission at 84.0 eV. UPS spectra were acquired using a He I radiation source (21.2 eV) operated at 100 mA with normal detection at constant pass energy of 2 eV. Samples were biased  $-8$  V in order to resolve the secondary electron cutoff in the UPS spectra. Work function values were determined from the width of the UPS spectra as discussed below.

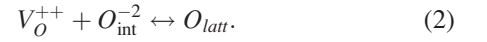
### III. RESULTS AND DISCUSSIONS

As we have proposed previously, at temperatures greater than  $\sim 200^\circ\text{C}$  changes in conductivity could be attributed to a reduction in donor concentration implying oxygen diffusion into the grains with the consequent annihilation of oxygen vacancies. Conversely, if oxygen diffuses out of the grains, vacancies are generated. On the other hand, Electron Spin Resonance measurements indicate that adsorbed oxygen transfers electrons from the grains to the chemisorbed oxygen.<sup>19,20</sup> Consistently, the relevant reactions from the gas phase to the bulk of the grain for oxygen have been proposed to be<sup>19</sup>



where  $\text{O}_{2(\text{gas})}$  refers to oxygen in the gas environment,  $\text{O}_{2(\text{ads})}$  to an oxygen molecule adsorbed at the grain surfaces,  $\text{O}_{\text{ads}}^-$  and  $\text{O}_{\text{ads}}^{-2}$  to singly and doubly ionized monatomic oxygen at

the grain surface,  $\text{O}_{\text{int}}^{-2}$  to interstitial oxygen, and  $\text{O}_{\text{latt}}^{-2}$  to oxygen at the tin oxide lattice. In the last part of Eq. (1), interstitial oxygen migrates from the surface to the bulk annihilating oxygen vacancies,  $\text{V}_\text{O}^{++}$ , *i.e.*



Therefore, exposing the  $\text{SnO}_2$  sample to an oxygen rich atmosphere at high enough T results in a decrease in  $\text{V}_\text{O}^{++}$ . On the other hand, under vacuum  $\text{V}_\text{O}^{++}$  is expected to increase. The donor density, at which grains become completely depleted, due to their small size, can be relatively high. Assuming a spherical grain, this takes place when

$$V_s = \frac{eN_d R^2}{6\epsilon_0 \epsilon_r}, \quad (3)$$

where  $V_s$  is the band bending,  $R$  is the grain radius,  $N_d$  is the dopant concentration,  $e$  is the electron charge,  $\epsilon_0$  is the vacuum permittivity, and  $\epsilon_r$  is the relative permittivity. In our case, we can estimate that complete depletion occurs for  $N_d \leq 7 \times 10^{23} \text{ m}^{-3}$ .

The presence of undesirable contaminants was checked with X-ray photoemission spectroscopy (XPS). Figure 5 shows XPS spectra corresponding to the sample oxygen before (a) and after (b) thermally treated up to  $380^\circ\text{C}$  in air. We chose this temperature because at higher temperatures, the films lose adhesion. Anyway, this temperature is enough to end up with a clean surface. Indeed, broad scans after the treatment show the presence of only Sn and O related peaks with no detectable amounts of any contaminants. Conversely, note the presence of the C 1s signal at approximately 285 eV before the treatment.

By means of XPS core level analysis, comparative values of the interface Fermi level position in the gap can be determined as it is directly related to the XPS peak binding energy position. This can be accomplished using core levels peaks. In Fig. 6, we present results for O 1s showing a

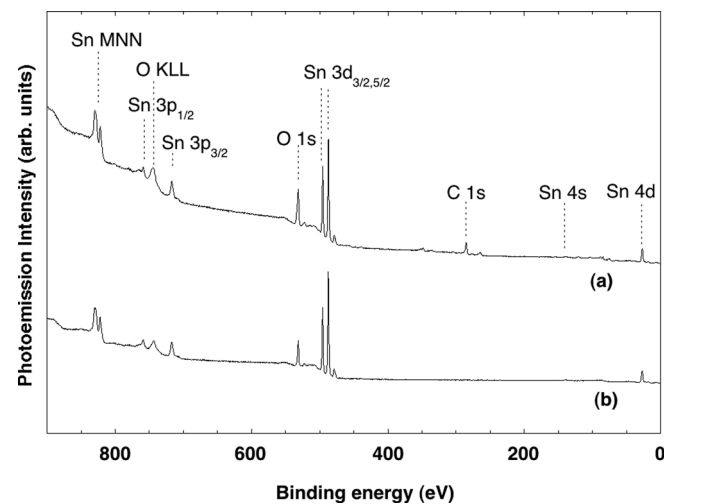


FIG. 5. (a) XPS broad scans for sample thermally treated for 1 h in air at  $100^\circ\text{C}$  and (b) after thermally treated up to  $380^\circ\text{C}$  using a heating rate of  $1^\circ\text{C}/\text{min}$  and exposed to an air atmosphere at  $380^\circ\text{C}$  during 1 h. Note that in spectrum (b), only Sn and O related XPS peaks are observed; while in (a), an important amount of C is detected.

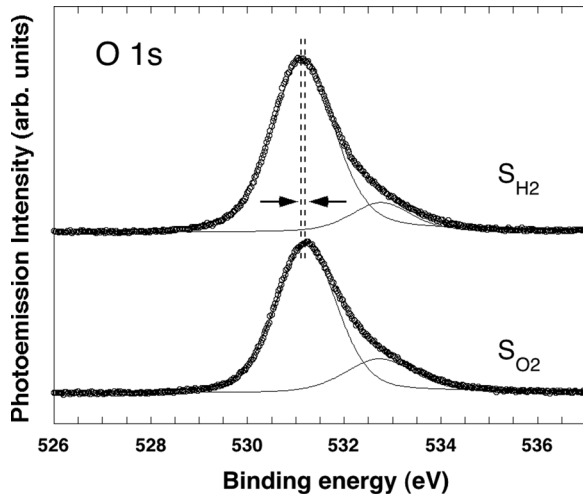


FIG. 6. O 1s core level energy distribution curves for the  $S_{H_2}$  and the  $S_{O_2}$  samples. Similar binding energy indicates that barrier heights do not differ considerably. The shoulder peak is 25% of the main peak in the  $S_{O_2}$  sample and 14% of the main peak in the  $S_{H_2}$  sample.

Schottky barrier difference of 0.07 eV with a slightly higher barrier for the  $S_{H_2}$  sample, confirming our previous findings.<sup>21</sup> These results indicate that, regardless of the sample treatment, which leads to very different conductivities, surfaces under vacuum at room temperature present the same Schottky barrier height.

The O 1s XPS signal of the films was fitted with two peaks located at 531.2 and 532.7 for the  $S_{O_2}$  sample, 0.07 eV less for the  $S_{H_2}$  sample. The high intense peak can be assigned to the O 1s core level of the oxygen anions in the  $SnO_2$ . The shoulder peak has been proposed to be due to defect sites within the oxide crystal, chemisorbed oxygen, hydroxide species, or carbonaceous compounds.<sup>22,23</sup> This last option seems not possible, as carbon is not detected in the XPS spectrum. Hydroxide species should increase after treating the sample with hydrogen. However, the shoulder peak reduces from 25%, for the  $S_{O_2}$  sample, to 14% of the intense peak, for the  $S_{H_2}$  sample. It is known that polycrystalline materials present an interfacial component at the grain boundaries or at their surfaces. It has been proposed to have several atomic layers in which atoms do not keep the exact position corresponding to the crystal structure. Thus, we are prone to assign the shoulder peak to nonstoichiometries and disorder present at the grains surfaces due to oxygen vacancies and the surface itself. The action of hydrogen, then, would be to eliminate some of the unstable oxygen at or close to the surface responsible for the shoulder peak.

Figure 7 shows the total capacitances vs. frequency during three stages of a thermal treatment carried out under vacuum for the sample  $S_{O_2}$ . In order to analyze the film capacitance behavior, it is necessary to consider the shape of the electrodes. According to the geometry of our sensors, the electrical capacitance can be related to that of semi-infinite electrodes, coplanar with a gap separating them<sup>24</sup>

$$C_0 = \frac{2\epsilon_0\epsilon_r}{\pi} l \ln \left[ \left( 1 + \frac{w}{a} \right) + \sqrt{\left( 1 + \frac{w}{a} \right)^2 - 1} \right], \quad (4)$$

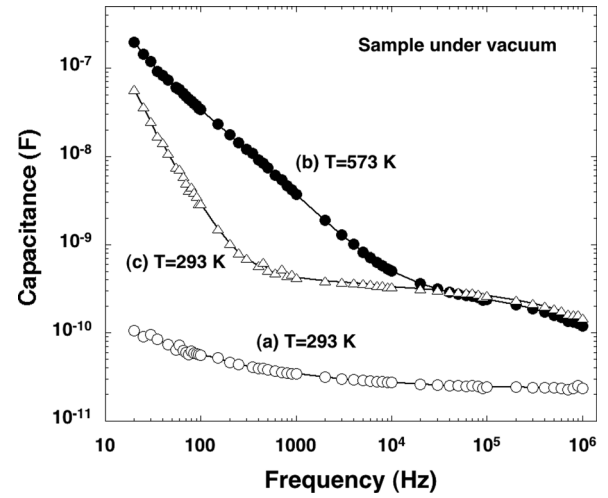


FIG. 7. Total capacitances vs. frequency during three stages of a thermal treatment carried out under vacuum. (a) corresponds to the sample initially at room temperature, (b) at  $T = 573$  K, and (c) after cooling back to room temperature.

where  $w$  is the width of the platinum interdigitated electrodes,  $2a$  is the width of the inter-electrode pathway and,  $l$  is the length of the electrode (See Fig. 8). For the dimensions of our setup  $w = 0.5$  mm,  $l = 11$  cm,  $a = 10$   $\mu$ m, with Eq. (4), we can calculate that  $C_0$  is 2.88 pF for  $\epsilon_r = 1$ .

Experimentally, we found that the capacitance before the incorporation of the sensor film for 1 MHz,  $C_1$ , is 12 pF. We have to consider that we have two capacitors in parallel, as on one side of the electrodes, there is air, with an associated capacitance  $C_{air}$ , and, on the other, the alumina substrate. Thus, the total capacitance  $C_1$  is the sum of both sides' contributions.

Due to the symmetry of the arrange, we can write

$$C_{air} = C_0/2. \quad (5)$$

The contribution to the total capacitance due to the alumina of thick  $b_{alum}$ , can be expressed, as a good approximation<sup>25</sup>

$$C_{alum} = \frac{\epsilon_0\epsilon_{alum}}{\pi} l [\sinh^{-1}(b_{alum}/a) + 0.234], \quad (6)$$

where  $\epsilon_{alum}$  is the alumina relative permittivity and the measured value of the alumina thickness is  $b_{alum} = 0.41$  mm. Using Eq. (6) and correcting for the electrode width,<sup>25</sup> we could determine that  $C_{alum}/\epsilon_{alum} \approx 1.41$  pF. This value would

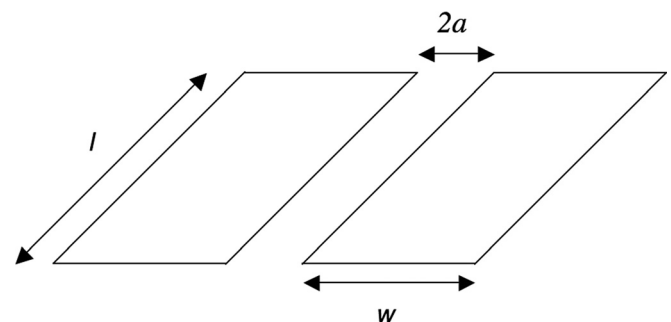


FIG. 8. Geometrical configuration of coplanar band electrodes.

be 1.44 pF for an infinitely thick alumina substrate implying that the contribution of the air beyond the alumina substrate is  $C_a = 0.03$  pF.

Experimentally, we found that the capacitance after the incorporation of the sensor film,  $C_2$ , was 23.5 pF, value measured for 1 MHz [curve (a) of Fig. 7]. The contribution to the total capacitance due to the film of thick  $b_{film}$ , can be expressed, as a good approximation, as done before for the alumina side contribution

$$C_{film} = \frac{\epsilon_0 \epsilon_{film}}{\pi} l \left[ \sinh^{-1}(b_{film}/a) + 0.234 \right], \quad (7)$$

where  $\epsilon_{film}$  is the film relative permittivity and the measured value of the film thickness is  $b_{film} = 0.105$  mm. With Eq. (7), we determined that  $C_{film}/\epsilon_{film} = 1.018$  pF, the correction for the electrode width is negligible in this case. As above, this value would be 1.44 pF for an infinitely thick film implying that the contribution of the air beyond the film is  $C_b = 0.42$  pF.

First, the measured capacitance  $C_1$ , without the sensor film, can be expressed as

$$C_1 = C_{air} + C_{alum} + C_a. \quad (8)$$

Thus, we can determine  $C_{alum} = 10.53$  pF and then the alumina permittivity  $\epsilon_{alum} = 7.5$ , value consistent with those reported in the literature.<sup>26</sup>

Second, the measured capacitance  $C_2$ , with the sensor film, can be expressed as

$$C_2 = C_{alum} + C_{film} + C_a + C_b. \quad (9)$$

With Eq. (9), we can determine  $C_{film} = 12.52$  pF and then the film permittivity  $\epsilon_{film} = 12.3$ . This value represents an effective permittivity consistent with the values 9.6 and 13.5 for the parallel and perpendicular values of the permittivity,  $\epsilon_1$  and  $\epsilon_2$ , to the regularly named *c*-axis in rutile.<sup>27</sup> Following Ref. 28, the effective permittivity for the given values of  $\epsilon_1$  and  $\epsilon_2$  can be calculated and results 12.4. Note that the found effective dielectric constants validate the performed calculations and the value found for the film permittivity indicates that grains are completely depleted of electrons.

Figure 9 shows a temperature cycle for a sample under vacuum in which the temperature was raised up to 295 °C and then back to room temperature. These results cannot be explained within the frame of a thermionic mechanism for conduction. With heating, the conductivity starts increasing with a slope corresponding to activation energy of 0.37 eV and then, above 150 °C, the slope reduces to 0.18 eV. We could assume that oxygen desorbs as the temperature is increased and then the activation energy would decrease with T. However, with a reducing barrier height, the conductivity should increase faster than at low temperatures, as a low activation energy facilitates conduction. On the other hand, we could assume that the activation energy increases with T. This could be due to the adsorbed oxygen being in the form of  $O_2^-$  below 150 °C and of  $O^-$  or  $O^{2-}$  above this temperature.<sup>27</sup> However, after cooling down, the conductivity adopts a much larger value than at the beginning of the cycle, result that is incompatible with a higher barrier.

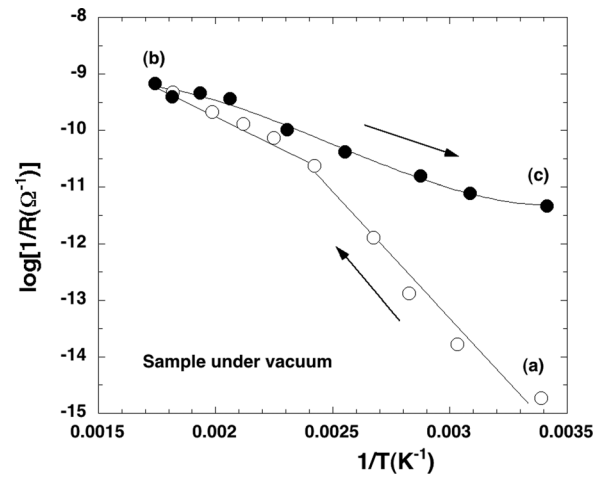


FIG. 9. Conductance in temperature cycling experiments. Temperature was raised up to 295 °C and then back to room temperature. After cooling down, the conductivity adopts a much larger value than at the beginning of the cycle. Lines are a guide to the eye.

As already mentioned, it has long been recognized that Schottky barriers at grain surfaces determine the conductance of this type of films. Consequently, adsorption of different gases is accompanied by changes in these barriers, which affects electron transport between grains and thus the resistance of the device changes. Conduction mechanisms have been interpreted in analogy to those in metal-semiconductor contact diodes. Accordingly, the electrical properties of polycrystalline semiconductors are usually described with a simple one-dimensional model representing the interface between two grains. It is regularly considered that a thermionic mechanism is responsible for the sample conductivity. However, it has been pointed out that a thermionic-field emission or tunneling contributions are unavoidable for the usual barriers characteristics.<sup>29</sup> In short, two models have been mostly used in the literature to rationalize electron conduction in polycrystalline SnO<sub>2</sub>: an activated thermionic mechanism that depends only on intergranular barrier heights, and an electron tunneling mechanism that depends both on intergranular barrier heights and widths.

We propose that the present findings can be the consequence of oxygen intragranular diffusion, which takes place when heating the samples to a high enough temperature, affecting oxygen vacancies concentration and, then, Schottky barrier widths.<sup>14,30</sup> The thermionic contribution to conductivity would not be strongly modified as long as barrier heights remain constant, as XPS results indicate, but the thermionic-field emission contribution would be especially affected as tunneling depends on both height and width of the barriers. As temperature is raised, oxygen diffuses out of the sample increasing the donor density, because oxygen vacancies are the dominating donor. Thus, tunneling is favored and, after cooling down, the sample presents a much higher conductivity. In a low-doped sample, grains can be completely depleted and, for a doping high enough, grains can have a bulk region in their centers. Thus, the slope change in conductivity at about 140 °C is an indication that oxygen diffusion led to a doping that narrowed the space charge regions to the point at which grains are not completely depleted.<sup>15,18,31</sup>





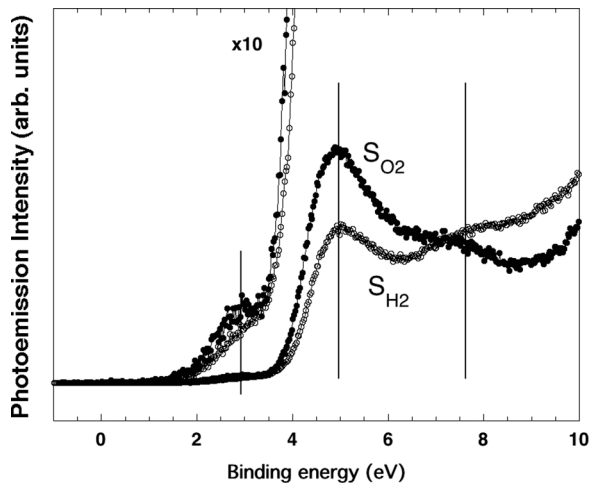


FIG. 11. UPS spectra for the  $S_{H_2}$  and  $S_{O_2}$  samples. Long lines indicate contributions of O 2p, at about 5 eV, and of Sn 5p with O 2p hybridization, around 7 eV. Gap states are observed around 3 eV, short line, as shown in the amplified spectrum.

$(E_C - E_F)$ , and/or by modifying surface electric dipoles that control the difference between the vacuum level and the conduction band bottom, *i.e.*, the electron affinity  $\chi = E_{vac} - E_C$ . Thus, the work function can be expressed as the sum of two terms

$$WF = \chi + (E_C - E_F). \quad (11)$$

For a fixed value of the electron affinity, any shift in the position of the Fermi level with respect to the conduction band will result in an equal and opposite change of the work function. However, the electron affinity can be modified by gas treatments. Indeed, large changes in electron affinity (of about 1 eV) have been attributed to surface termination after oxidizing or reducing annealing.<sup>27</sup>

Photoelectron spectroscopies offer a direct manner to determine work functions. Indeed, the secondary electron onset provides a straight measurement of the specimen's

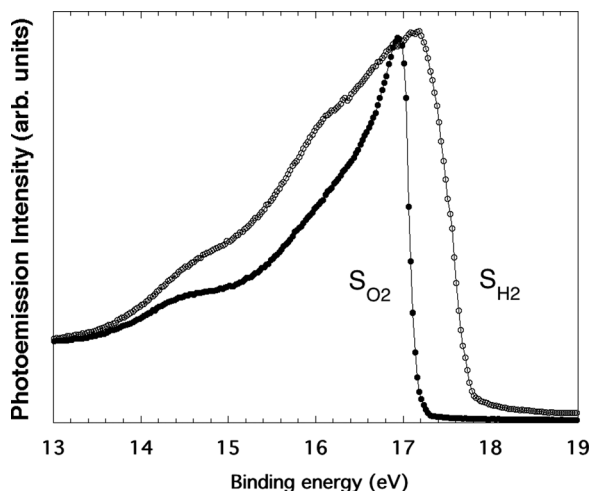


FIG. 12. Normalized secondary electron cutoff for the  $S_{H_2}$  and  $S_{O_2}$  samples. As observed, the exposure to hydrogen reduces the work function by 0.7 eV. Interestingly, this is not a consequence of changes in band bending but due to changes in the electron affinity.

work function. Then, with the help of this technique, we can investigate the influence on the work function of the treatments under oxygen and hydrogen. The results of the measurements are shown in Fig. 12. The derived work functions are 4.1 eV and 3.4 eV for  $S_{O_2}$  and  $S_{H_2}$  samples, respectively. A reduction in the work function, 0.7 eV in the present case, would be expected after the action of a reducing gas. At odds with Ref. 33, we found that this is not due to a change in the band bending, as derived from the results of Fig. 6, but from a change in the electron affinity.

#### IV. CONCLUSIONS

Gas detection with metal oxide is based on the presence of intergranular barriers that are modified with the exposure to different gaseous species. Thus, it is crucial to determine the basic mechanisms that alter those barriers responsible for the film conductivity. In this paper, we showed that polycrystalline tin oxide films present intergranular barriers that can be affected by thermal treatments at moderate temperatures. Interestingly, XPS results show that samples with different conductivities can present similar barrier heights. Thus, a thermionic mechanism cannot be responsible for the observed differences. Experiments indicate that oxygen diffusion into and out of the grains can change the doping to the point that they can present fully developed Schottky barriers, with quasi-neutral regions at their centers, or be completely depleted of carriers. In this paper, we specially focused on capacitance values, which were analyzed in detail considering the geometry involved to quantify doping changes due to in- and out-diffusion that leads to barriers overlapping. Also, we presented UPS results that show that exposing the  $SnO_2$  samples to oxidizing or reducing environments modify the work function without modifying the band bending.

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