Influence of the Strain on the Electrical Resistance of Zinc Oxide Doped Thin Film Deposited on Polymer Substrates**

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Zinc oxide (ZnO) thin films have been extensively studied in recent years, because of their low material cost, relatively low deposition temperature, non-toxicity, and stability in hydrogen plasma, if compared with indium tin oxide (ITO) and tin oxide (SnO_2). Due to their unique properties, it is an excellent candidate for many applications in optoelectronic devices such as liquid crystal displays and solar cells as well as surface acoustic wave devices, ultra sonic transducer arrays, chemisorption gas sensors, mass-loading sensors, and optical waveguides.[1]

Nowadays, there is a great interest in replacing glass substrates with polymer substrates, particularly in flat panel display and touch screens technologies, where low volume, lightweight, and robustness are important. Besides that, the added flexibility of polymer substrates opens new application fields that utilize curved surfaces such as amorphous silicon large area flexible position sensors.[2] In these flexible applications, there is a trade-off between using thick layers of conductive oxides to reduce the resistivity and using thin layers, which can withstand greater strain in the substrate. One particularly interesting application is touch screens. In this case, the point contact of the user's finger or stylus is calculated from the resistance between the point of contact and a reference position. If the resistance of the conductive oxide increases because of cracking, the device may be less reliable.

Due to these new demands, some authors start preparing ZnO:Al films deposited on organic substrates such as polyisocyanate.[3,4] Recently we presented the first results on ZnO:Al thin films deposited on Mylar substrates.[5]

In order to correlate the influence of the mechanical strain on the electrical properties, we have measured the electrical resistance as a function of mechanical strain in-situ using special electrical probes associated to a miniature tensile testing machine.

The prepared films were physically stable and show good adherence to the polymer substrates. No crack or peel off of the films was observed after deposition. Figure 1 shows a typical SEM micrograph of ZnO:Al with 140 nm thickness.

In Figure 2, the nominal stress-strain curves for coated and non-coated PET films are displayed. Both materials show a yield transition at 2 % strain followed by a slightly ascendant plastic plateau. No significant changes were observed for the PET and the ZnO:Al coated PET substrate. The only differ-

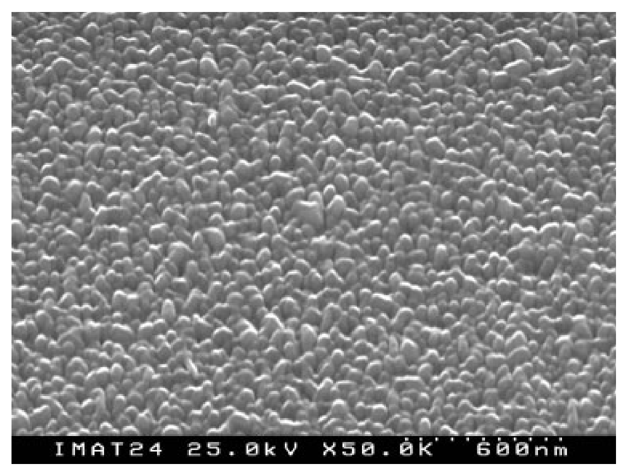

Fig. 1. Scanning electron micrograph for ZnO:Al thin film with 140 nm thickness deposited on PET substrate.

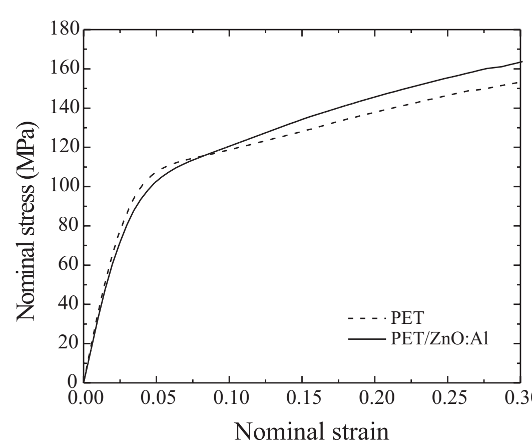

Fig. 2. Dependence of the nominal stress-strain curves in uniaxial tension of PET and ZnO:Al coated PET films.

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[**] *The authors would like to acknowledge A. Lopes for the SEM analysis and B. Munsch from Dupont for the technical assistance provided for this work. This work was supported by the "Fundação para a Ciência e a Tecnologia" via Pluriannual Contracts with CENIMAT and by the projects: POCTI/1999/ESE/35578 and POCTI/1999/CTM/35440.*

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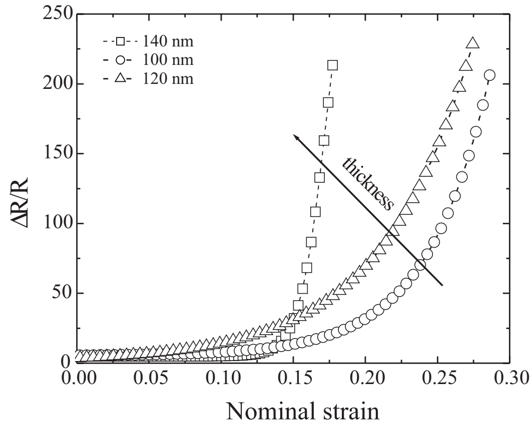

Fig. 3. Change in electrical resistance (ΔR/R) of ZnO:Al coated PET substrate as a function of nominal strain, for the three thicknesses, as indicated inside the figure.

ence is for high strain where the nominal stress is larger for the coated polymer substrate. The elastic modulus, nominal yield stress, and nominal yield strain are similar for both samples: 3100 MPa, 72 MPa, and 2.2 %, respectively.

The change in the electrical resistance as a function of the uniaxial tensile elongation are presented in Figure 3 for the three ZnO:Al films. For all samples, the resistance increases sharply at some threshold strain that depends on film thickness. The resistance of the thinner ZnO:Al film increases at the highest threshold strain while the resistance of the thick ZnO:Al film increases for the lowest strain. This is consistent with similar studies performed on silicon oxide deposited on PET substrates.[7] The increase in the resistance is due to the cracking of ZnO:Al with increasing strain.

It was observed in the optical microscope (see Fig. 4) during tensile elongation that the first cracks appear for a nominal strain of 2 % for the thinner sample. The cracks traverse the full width of the sample section but the electrical resistance remains finite. This suggests that something in the crack is responsible for the conductivity.[8] Near a strain of 8 %, a second type of crack appears (parallel to the straining direction and perpendicular to the first cracks) due to the lateral contraction of the sample. The fragments at the end of these cracks tend to overlap and enhance debonding. For values higher than 8 % of nominal strain, the density of the first type of crack remains constant while the density of the second type increases up to 10 % nominal strain. For higher values, debonding within the secondary cracks is initiated, and a deviation from linearity is observed. These results are consistent with the variations observed in the electrical resistance.

These observations are consistent with the SEM photographs presented in Figure 5. The morphology of the ZnO:Al with different thickness reveal that the thinner film exhibits the thinnest crack (around 130 nm) while the width of the thickest crack is about 670 nm. It is also possible to see that at this stage, the primary cracks parallel to the tensile direction are largely open and leave wide PET areas visible underneath. The secondary cracks initiated at 8 % strain as a result of the Poisson's ratio compression effects, also appear in the form of "tent-shaped buckles".[9] The oxide fragments are partially delaminated from the PET substrate as observed in Figure 6a. From the SEM observations it was also possible to observe the type of crack responsible for the fissures is an intergranular fracture path (Fig. 6b). This results in an interlocking of neighboring ZnO:Al fragments with the area of contact being reduced as the strain in the substrate is increased. This assumption explains the non-infinite resistance observed even after cracks were seen to have propagated across the entire gauge width length. These remarkable adhesion levels suggest the existence of covalent bonds at the interface between the polymer and the oxide.

To conclude, the uniaxial tensile strain tests and the resulting variation of the electrical resistance were reported. The increase in resistance is very important for the design/project of devices deposited on flexible substrates due to competing

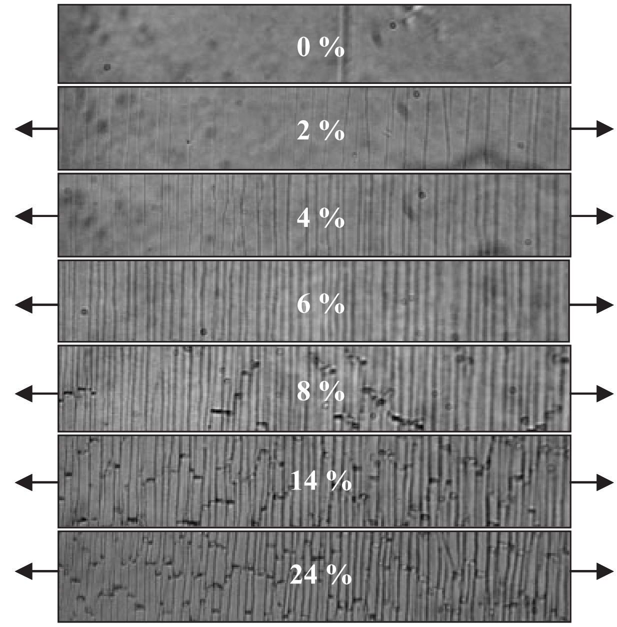

Fig. 4. Fragmentation morphology of the ZnO:Al coated PET substrate at different strains. The arrows indicate the tensile direction and the numbers inside the optical photographs are the strain values.

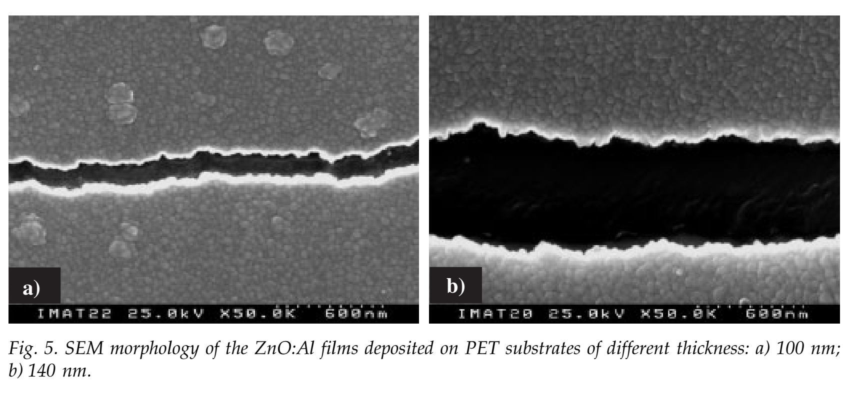

Fig. 5. SEM morphology of the ZnO:Al films deposited on PET substrates of different thickness: a) 100 nm; b) 140 nm.

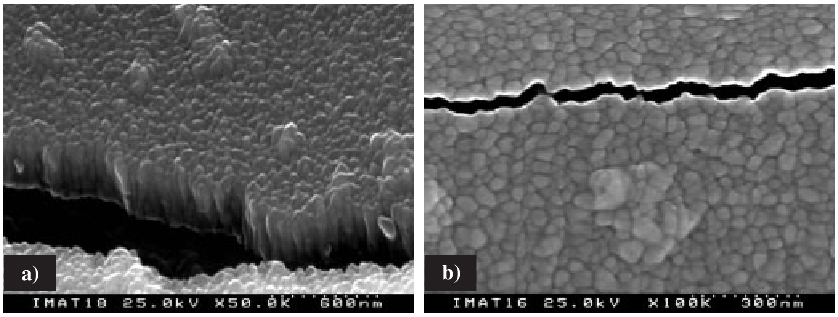

Fig. 6. a) SEM photograph (with 40 ° tilt) of a perpendicular and a transversal crack; b) High resolution SEM photograph where it is possible to see the intergranular fracture.

use of thick layers to reduce the resistance and of thin layers, which can withstand higher strain in the substrate.

The results indicate that a finite resistance of the ZnO:Al film was obtained, even after cracks had developed and propagated across the entire gauge width length. From these results, it is possible to conclude that the resistance is not directly proportional to the number of cracks but rather to the width of the cracks, which depends also on the film thickness.

Experimental

The polymeric substrates used in this work are poly(ethylene terephthalate) (PET) with a standard thickness of 100 μm from Dupont® (Mylar type D). Before depositing the ZnO thin films, the substrates were ultrasonically cleaned in a detergent bath, followed by isopropyl alcohol, and dried under nitrogen. The substrates with 10 × 10 cm dimensions were placed inside the chamber and then evacuated to a base pressure of 7×10^{-7} mbar. The ZnO:Al thin films were produced by r.f. magnetron sputtering from a commercially available sintered ceramic ZnO:Al_2O_3 (98:2) target with 99.99 % purity and 50 mm diameter from Cerac, placed at a distance of 20 cm from the substrate.

The flow rate of the sputtering gas (argon) was controlled by a mass flow controller to 10 SCCM and the deposition pressure was fixed to 1.7×10^{-2} mbar (optimized deposition conditions for this chamber, in order to guarantee uniformity over 10 × 10 $cm^{[6]}$). In order to correlate the effect of the thickness on the film properties, a set of three samples with 100, 120, and 140 nm were produced.

The growth morphologies were analyzed using a field effect S-1400 Hitachi scanning electron microscope.

The thickness of the films was measured using a Sloan Dektak 3D profilometer.

The electrical resistance as a function of uniaxial strain was measured in situ using special electrical probes. These measurements were performed with a tensile testing machine (Rheometric Scientific Minimat – Firmware 3.1) working at a constant speed of 2.5 mm/min at room temperature. Since the films were deposited on polymeric substrates, the samples were cut with a gauge length of 50 mm and a gauge width of 10 mm. For detection of the crack development during straining, the tensile tester was mounted on an Olympus BH-2 optical microscope stage. Images were recorded via a digital camera Olympus Camedia C-2020 connected to the microscope.

Received: May 01, 2002

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Dependence of the Strains and Residual Mechanical Stresses on the Performances Presented by a-Si:H Thin Film Position Sensors**

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In this communication, we describe for the first time the influence of residual stresses on the performances of large-area position-sensitive detectors produced on flexible substrates of different thickness. Two techniques were used for evaluating the residual stresses: active optical triangulation and angle resolved scattering, to measure the radius of curvature, and a technique for measuring the density of states associated with the amorphous silicon layer deposited on the polymeric substrate, the constant photocurrent method (CPM). From the results it was possible to correlate the stress-

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[**] *We want to thank Dupont and Palissy Galvani for providing Kapton type VN substrates. This work was supported by the "Fundação para a Ciência e a Tecnologia" through Pluriannual Contracts with CENIMAT and by the projects: POCTI/1999/ESE/35578 and POCTI/1999/CTM/35440.*

  1438-1656/02/0808-0612 $ 17.50+.50/0