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B. Cosson, A.C. Akué Asséko, M. Lagardère, M. Dauphin

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# Investigation of the densification mechanisms and corrosion resistance of amorphous silica films

Simon Ponton <sup>a, b</sup>, Franck Dhainaut <sup>a, b</sup>, Hugues Vergnes <sup>b</sup>, Diane Samelot <sup>a</sup>, Daniel Sadowski <sup>a</sup>, Vincent Rouessac <sup>c</sup>, H el ene Lecoq <sup>d</sup>, Thierry Sauvage <sup>d</sup>, Brigitte Caussat <sup>b</sup>, Constantin Vahlas <sup>a</sup>

\*

<sup>a</sup> CIRIMAT, Universit e de Toulouse, CNRS, Toulouse, France

<sup>b</sup> LGC, Universit e de Toulouse, CNRS, Toulouse, France

<sup>c</sup> IEM, Universit e de Montpellier, CNRS, ENSCM, Montpellier, France

<sup>d</sup> CEMHTI, CNRS, Orl eans, France

\*Corresponding author: [constantin.vahlas@ensiacet.fr](mailto:constantin.vahlas@ensiacet.fr)

## Abstract

The barrier properties of the technologically attractive amorphous silica films depend on their structural characteristics at the atomic level, **which, in turn are strongly influenced by the deposition conditions**. In this paper, we propose an investigation of the **poorly investigated** densification mechanism of amorphous SiO<sub>2</sub> films processed by CVD from TEOS and O<sub>2</sub> between 400 and 550 C. Based on literature survey and **our** original experimental results, **we show** that the densification process of these films, occurring with increasing the deposition temperature, is highlighted by a decrease of the water and silanol content, probed by transmission FTIR. **We discuss** the evolution of Si-O-Si related vibration **signatures** and **we**

**use** the central force model to correlate the LO<sub>2</sub> and LO<sub>3</sub> shifts **with** the decrease of the Si-O-Si bond force **constant**, when the deposition temperature increases. Nuclear **analysis reveals** that films processed below 525 °C present hydrogen content between 5±0.3 and 7±0.3 %at. Ellipsometry measurements **attest** that films processed at 550 °C are close to O/Si silica stoichiometry and hydrogen free. **We show that application of the P-etch test results in particularly low erosion rate of 10 Å.s<sup>-1</sup> for dense films processed at 550 °C.**

**Keywords:** silica, TEOS, CVD, densification, FTIR, P-etch **test**

# 1. INTRODUCTION

Thin and dense SiO<sub>2</sub> films processed from tetraethyl orthosilicate (Si(OC<sub>2</sub>H<sub>5</sub>)<sub>4</sub>, TEOS) by chemical vapor deposition (CVD) have long been considered to solve mainly microelectronic issues, such as copper diffusion barrier, dielectric capacitor and intermetallic dielectric layers for multilayer metallization systems [1]. **Nowadays**, such films remain key enabling materials in innovative applications and devices [2][3][4][5][6]. These applications have in common a low thermal budget process requirement due to thermally sensitive, or 3D-complex geometry substrates. Addition of oxygen [7] or ozone O<sub>3</sub> [8] in the gas phase or the use of plasma assistance (PECVD) [9][10] are well known actions **resulting in lower thermal budget processes, yielding silica films at moderate temperatures**, typically lower than 600°C. More recently, atomic layer deposition (ALD) of SiO<sub>2</sub> thin films has also been performed at low temperature to solve optical applications issues [11] or to cap nanoporous anodic alumina membranes employed in biosensor **devices** [12]. The correlation between the effects of such low temperature deposition **and** the atomic arrangement of the silica structure is not well understood, especially considering that the densification process is complex and multi-parameter dependent [13]. **However**, the atomic arrangement of the material is the key to understand its targeted barrier properties when such silica films are used as barrier coatings [4][5] or to understand the special selectivity of gases when it is used in gas membranes [6].

The structure of the amorphous SiO<sub>2</sub> films has been defined by Sen and Thorpe [14] and Galeener [15] as a short range **organized continuum** built-up from tetrahedral entities centered on a silicon atom. **Each oxygen atom at the corner of a tetrahedron** is shared by another tetrahedral unit and cross-links the entire network. Twofold-coordinated bridging oxygen is more mobile than fourfold-coordinated silicon and has been considered as the main contributor to the atomic vibration of this system [16]. Vibration modes of this oxygen are intimately related to the SiO<sub>2</sub> structure **and, consequently**, the spectral changes under densification **should** reveal

information on the atomic distribution of the network. Three vibration modes of the oxygen atom linked with two silicon atoms are assigned in the mid infrared (IR) region between 400  $\text{cm}^{-1}$  and 4000  $\text{cm}^{-1}$ . These are the transversal optical rocking ( $\text{TO}_1$ ), bending ( $\text{TO}_2$ ) and asymmetric stretching ( $\text{TO}_3$ ) modes, which are observed respectively at around 450  $\text{cm}^{-1}$ , 800  $\text{cm}^{-1}$  and 1070  $\text{cm}^{-1}$  [16]. **The  $\text{TO}_1$  mode is sensitive to structural changes [17] but its frequency shift is too weak to be monitored by FTIR, i.e. the shift is in the same order of magnitude with the measurement uncertainties.** An additional vibration mode, the  $\text{TO}_4$ , is assumed by some authors between 1050  $\text{cm}^{-1}$  and 1250  $\text{cm}^{-1}$  [18–20]. It has been attributed to the out of phase vibrations of the  $\text{TO}_3$  mode [20,21], to the **longitudinal,  $\text{LO}_3$  mode** in this area [22], to the presence of residual TEOS molecules [23] and to strained 3 to 6-fold Si-O rings [19,24,25]. The debate **around** the attribution of a physical phenomenon to the  $\text{TO}_4$  broad band reveals the difficulty to correlate the **existence and the characteristics** of this vibration with the short-range organization of the network. This difficulty may be attributed to the intrinsic nature of the material, which is amorphous, with randomly crosslinked tetrahedral entities [15]. **Thus, the overall characteristics of CVD  $\text{SiO}_2$  films, such as stoichiometry [26][27], porosity [28], impurities [29][30] and mechanical strain [24] which depend on the process conditions, will affect the FTIR signature, the evolution of the refractive index and consequently they may bias the interpretation of the densification process.**

This literature review reveals that, despite uncertainties on the origin of some vibrations, FTIR is an appropriate technique to monitor the evolution of the silica network, and for this reason it will be used in the present work. We will adopt the central force model proposed by Sen and Thorpe [14] and applied to glasses by Galeener [15] in order to link FTIR signatures to structural changes. This model connects  $\text{SiO}_2$  vibrations to the Si-O-Si intertetrahedral bond angle,  $\theta$ , and the Si-O force constant,  $\alpha$ , of the  $\text{SiO}_2$  network [31]. Nonetheless, the model assumes perfect stoichiometric, continuous  $\text{SiO}_2$  network and it takes into account neither

composition deviations, namely the offset with regard to the nominal O/Si ratio, nor the presence of heteroatoms such as carbon impurities, nor the porosity, nor the mechanical strain that can be observed in CVD SiO<sub>2</sub> films. Consequently, and in order to elaborate a scenario of the evolution of the structure under different process temperatures (T<sub>d</sub>), we perform complementary characterizations to get insight on these parameters: we quantify by nuclear analysis the concentrations of silicon, oxygen and hydrogen; we determine both the refractive index by reflection spectroscopic ellipsometry (SE) using the Sellmeier model, and the porosity by ellipsometric porosimetry (EP) measurements using the Cauchy model. Finally, we apply selective chemical corrosion tests proposed by Pliskin [13][32] in order to correlate the evolution of the structure with its intrinsic resistance towards aggressive medium.

## 2. EXPERIMENTAL

Depositions were performed in a horizontal, hot wall tubular CVD reactor, presented in the supplementary material (Figure S1). For each experiment, six 30 × 10 × 0.2 mm<sup>3</sup> Si coupons cut from 4" Si(100) wafers (Sil'tronix ST) were degreased in sequential acetone and ethanol ultrasound baths for 5 min each, dried in Ar flow, positioned on a stainless steel planar substrate holder and immediately introduced in the reactor. TEOS vapors were introduced in the reactor by bubbling 58 standard cubic centimeters (sccm) of nitrogen, N<sub>2</sub> (99.9999%, Messer) through the liquid precursor heated at 51 °C. 520 sccm of O<sub>2</sub> (99.999%, Messer) are added to the mixture just before the reaction chamber. Experiments were performed at five different T<sub>d</sub>, namely 400°C, 450°C, 500°C, 525°C and 550°C, at a constant pressure of 730 Torr. An annealing at 800°C during 30 min under ambient atmosphere was performed on one sample processed at 550°C. The thickness profiles of the SiO<sub>2</sub> films on all coupons were measured by reflectometry and served to deduce mechanistic and kinetic information for each experimental condition.

More information about the deposition setup and the kinetic results obtained is available in our recent companion paper which deals with the development of a kinetic model of this process [7].

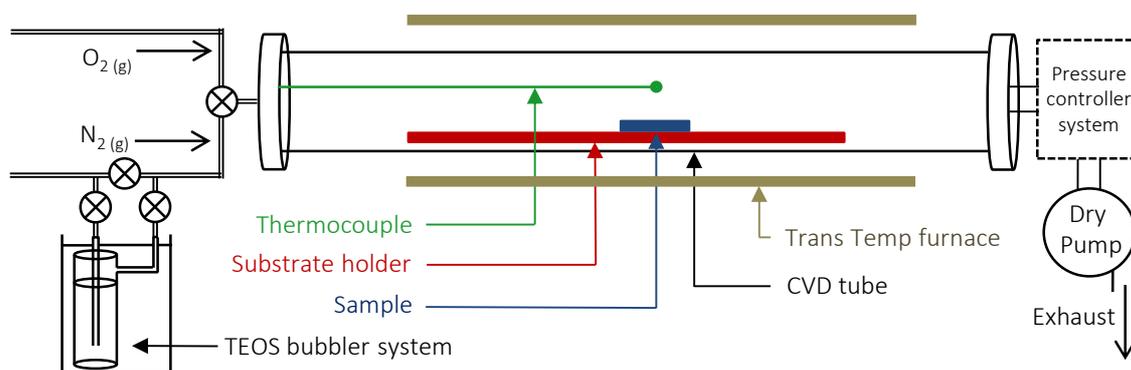


FIG. S1. Schematic view of the CVD Reactor [7].

FTIR spectrometry was performed in transmission mode with a Frontier FT-IR / NIR instrument. The 400-4000  $\text{cm}^{-1}$  spectral range was probed with a 2  $\text{cm}^{-1}$  spectral resolution. Fifty spectrum accumulations were performed for each experiment. Raw data were processed in order to remove light interferences and substrate signature. The position of SiO<sub>2</sub> peaks are affected by the thickness of the film in transmission mode [33]. Films with similar thicknesses were investigated, ranging from 338 nm to 416 nm and the absorbance was normalized by the film thickness. A homemade substrate holder was used in order to apply an angle between the incident beam and the sample, to observe the TO-LO splitting.

Two instruments were used for ellipsometry investigations. First, EP measurements were used to evaluate the open nanoporosity of the films. A Semilab GES5E spectroscopic ellipsometer was used on purpose, operating between 250 nm and 1000 nm, at a fixed incidence angle of 70°. This ellipsometer is completed with a lab made chamber set up for automatic

vapor adsorption-desorption investigation [34,35]. Porosity was estimated by probing the evolution of the refractive index due to the intake of the ethanol by the structure. **Before any acquisition of the refractive index and thickness of the layer, the sample was maintained under vacuum at 150 °C to desorb water or other volatile adsorbed compounds. Refractive index and thickness calculation were monitored at variable ethanol pressures by adding gradually the solvent in the chamber. The filled pore fraction of the open pores is a function of the partial pressure of the solvent above the sample. Ellipsometric data  $\tan(\psi)$  and  $\cos(\Delta)$  are collected then simulated and fitted on the 390-1000 nm spectral range with a goodness of fit over 0.99, using the Cauchy model (Winelli2<sup>®</sup> software).**

In addition, a Semilab SE-2000 instrument operating in the 250 nm to 1000 nm spectral range with fixed incidence angle of 70 ° was used to evaluate the refractive index by SE. The spectroscopic ellipsometry data were simulated and fitted using the Semilab SEA software. The Sellmeier model was used as a reference to determine the refractive index and the films thickness.

The O/Si atomic ratio of film was determined by coupling the Rutherford Backscattering Spectroscopy (RBS) and Nuclear Reaction Analysis (NRA) techniques for Si and O analysis, respectively. RBS was performed at a 166° detection angle with 2 MeV  $\alpha$  particles. The oxygen concentration was measured through the use of the  $^{16}\text{O}(d,\alpha_0)$  nuclear reaction with  $\text{D}^+$  ion beam of energy 0,9 MeV and at a 166° detection angle of emitted alphas. The hydrogen content of  $\text{SiO}_2$  film was measured by ERDA with 2,8 MeV alphas beam. The samples were positioned at 15° grazing incidence and the recoiled hydrogen atoms were collected at scattering angle of 30°. The global composition of films is obtained through the simulation of the RBS, NRA and ERDA spectra by using the SIMNRA software [36].

The corrosion resistance of the films was evaluated by using the P-etch solution test, proposed by Pliskin [13]. The test consists in immersing the sample for 30 s in a stirred solution

maintained at 25°C. The solution is composed of 3 parts hydrofluoric acid (49 wt%), 2 parts of nitric acid (70 wt%), and 60 parts of water. The P-etching rate for each sample corresponds to the thickness loss for a given etching time, determined by SE and is given in  $\text{\AA}\cdot\text{s}^{-1}$ ,

### 3. RESULTS AND DISCUSSION

#### 3.1 FTIR

Figure 1 presents a typical 400-4000  $\text{cm}^{-1}$  FTIR survey spectrum of a  $\text{SiO}_2$  film processed at  $T_d = 400$  °C. **According to the literature [16]**, the  $\text{TO}_1$ ,  $\text{TO}_2$  and  $\text{TO}_3$  vibration modes are detected at 450  $\text{cm}^{-1}$ , 798  $\text{cm}^{-1}$  and 1074  $\text{cm}^{-1}$ , respectively. The  $\text{TO}_4$  broad band appears around 1100-1200  $\text{cm}^{-1}$ . Water content is observed between 3300-3600  $\text{cm}^{-1}$  and silanol fingerprints are probed at 925  $\text{cm}^{-1}$  and 3650  $\text{cm}^{-1}$ . Neither carbon (Si-C) [37] nor  $\text{CH}_3$  [38] nor Si-H impurities related vibrations [29,39–41], expected respectively at 600, 1300 and 2000  $\text{cm}^{-1}$ , are present.

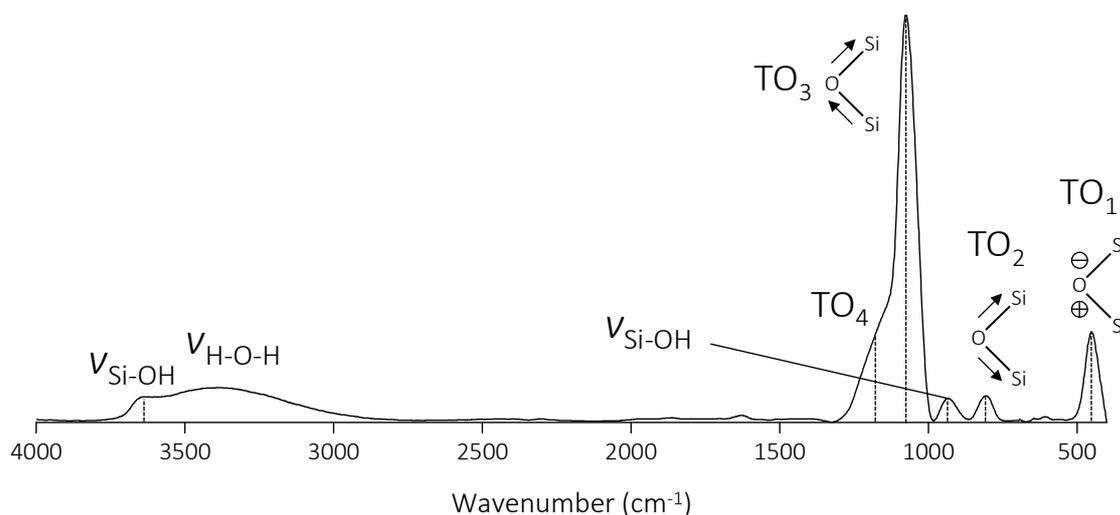


FIG. 1. FTIR survey spectrum of a  $\text{SiO}_2$  film processed at  $T_d = 400$  °C.

Figure 2 presents a zoom on the 700 – 1300  $\text{cm}^{-1}$  region of the FTIR spectrum of five samples processed at different  $T_d$ , from 400 °C to 550 °C. This spectral domain includes the Si-OH, and SiO<sub>2</sub> vibrations modes (TO<sub>2</sub>, TO<sub>3</sub> and TO<sub>4</sub>). A color code is applied for each  $T_d$  and it will be used in all forthcoming illustrations. **Several informations are highlighted in this figure. First**, a slight, though noticeable evolution of the position of the TO<sub>3</sub> peak, corresponding to the asymmetric stretching vibration mode is observed with increasing  $T_d$ . At the same time, the TO<sub>3</sub> peak broadens, from 72.5  $\text{cm}^{-1}$  at 400 °C to 81  $\text{cm}^{-1}$  at 550 °C as shown by the diagram in the insert of figure 2. **Finally, a decrease of Si-OH intensity is observed when  $T_d$  increases. These three informations will be discussed hereafter in order to have an insight into the reorganization of silica films when  $T_d$  increases.**

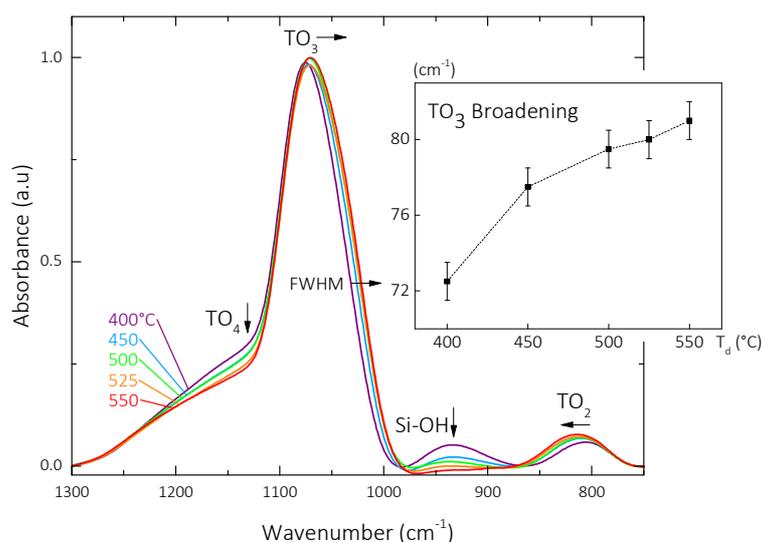


FIG. 2. Zoom in the 700 – 1300  $\text{cm}^{-1}$  zone of the FTIR spectra containing the Si-OH, TO<sub>2</sub>, TO<sub>3</sub>, TO<sub>4</sub> vibrational modes, and their evolution as a function of  $T_d$ . The evolution of the TO<sub>3</sub> peak broadening is plotted as a function of  $T_d$ .

### 3.1.1 Evolution of TO<sub>3</sub> frequency with the increase of $T_d$

It has been reported that the information contained in the evolution of the position of the  $\text{TO}_3$  peak, provides insight in the densification process of the  $\text{SiO}_2$  network [13][42]. According to these authors, the shift to higher wavenumbers (blue shift) of the  $\text{TO}_3$  peak position indicates the densification of the  $\text{SiO}_2$  network and conversely, a shift to lower wavenumbers (red shift) is assigned to a decrease of the density.

Figure 3 presents the evolution of the position of the  $\text{TO}_3$  peak as a function of  $T_d$  of the CVD samples of the present work (mauve diamond plots). The point at the highest temperature corresponds to a sample deposited at 550 °C and annealed at 800 °C for 30 min. **We observe a red shift of the  $\text{TO}_3$  peak position** with increasing  $T_d$ , while the peak position of the annealed sample shifts to a higher value. For sake of comparison, a compilation of literature data is also reported in Figure 3, as a function of deposition or post deposition annealing temperature [23,43–48][49]. Empty symbols correspond to sol-gel, room temperature processed, then annealed  $\text{SiO}_2$  films. Full symbols correspond to plasma enhanced CVD (PECVD)  $\text{SiO}_2$  films, processed between room temperature and 250 °C. Two regimes are illustrated in the diagram. For deposition, or post-deposition annealing temperatures equal to, or lower than 600 °C, the  $\text{TO}_3$  position presents a red shift or remains stable with increasing temperature. Above 600 °C, it presents a blue shift with increasing temperature.

Literature and our own data present a remarkable coherence. The global behavior of all chemical systems used for the processing of  $\text{SiO}_2$  films indicates that the assumption following which densification is illustrated by a blue  $\text{TO}_3$  shift, is not relevant for films processed and/or annealed below 600°C. This inconsistency is attributed to films defects such as porosity, impurities or mechanical strain resulting in less dense organization of the tetrahedral network [13]. The previously described  $\text{TO}_4$  broad band is the representation of the disorder of the network and it is relatively high for every silica films processed at temperature lower than 600°C. This peak can strongly impact the position of the  $\text{TO}_3$  peak. At and above 600 °C, the

Si-O-Si asymmetric stretching vibration mode is less impacted by the TO<sub>4</sub> broad band because fewer defects might be observed in the structure. Consequently, the TO<sub>3</sub> peak position presents a blue shift with increasing temperature [44].

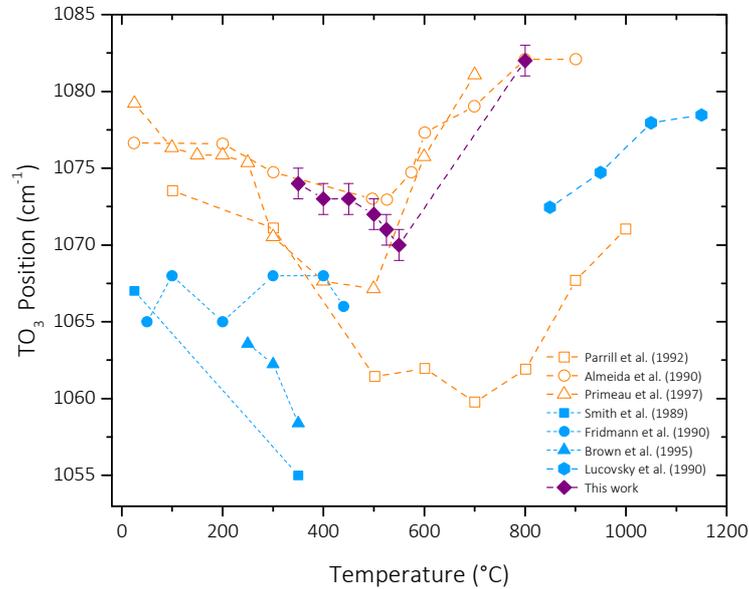


FIG. 3. Overview of TO<sub>3</sub> position as a function of post-annealing temperature for SiO<sub>2</sub> films. Empty symbol data: sol-gel films processed at room temperature. Full symbol data: PECVD films processed between room temperature and 250 °C. Empty squares [25], empty dots [43], empty triangles [23], full squares [48], full dots [49], full triangles [46], full pentagons [44]. Full diamonds correspond to samples of the present work positioned as a function of the deposition temperature. The point at the highest temperature corresponds to a sample deposited at 550 °C and annealed at 800 °C for 30 min.

### 3.1.2 Broadening of TO<sub>3</sub> peak with the increase of T<sub>d</sub>

The broadening of the TO<sub>3</sub> peak revealed in Figure 2 was also observed for PECVD SiO<sub>2</sub> films [48]. This behavior is attributed to a disorganization of the network [50] and thus it cannot be linked to the densification of the SiO<sub>2</sub> structure, either. Nonetheless, a close observation reveals that this broadening is not related to the increase of the full width at half maximum (FWHM) of the peak but rather to a change in its shape. **At high T<sub>d</sub>, the peak can no longer be fitted by only one gaussian or lorentzian function. The peak is shifted due to**

**the asymmetric, one-side increase of the peak area. An addition of a second contribution had to be assumed. The existence of two contributions below the  $TO_3$  main peak has been already suggests by authors [51,52]. To explain this behavior, they assigned the contribution of two asymmetric stretching mode vibrations from two different phases.**

### **3.1.3 Decrease of hydrated species specific peak intensities with the increase of $T_d$**

Figure 2 reveals a simultaneous decrease of the intensities of the  $TO_4$  vibration shoulder and of the Si-OH vibrations located at  $925$  and  $3650\text{ cm}^{-1}$  with increasing  $T_d$ . Above  $T_d = 500\text{ }^\circ\text{C}$ , almost no Si-OH located at  $925\text{ cm}^{-1}$  is probed. In order to confirm this trend, it is useful to focus on the  $2750 - 3800\text{ cm}^{-1}$  spectral domain, also containing Si-OH and H-OH vibrations. Figure 4 presents the FTIR spectra in this domain for the samples processed in the five investigated  $T_d$ . A common baseline has been adopted for all five spectra. The insert in figure 4 presents the evolution, with increasing  $T_d$ , of the peak intensities of the Si-OH and H-OH vibrations located at  $3650\text{ cm}^{-1}$  and at  $3400\text{ cm}^{-1}$ , respectively. As is the case for the Si-OH vibrations located at  $925\text{ cm}^{-1}$ , their intensity in the presented spectral domain also decreases with increasing  $T_d$ , and the same holds for the H-OH vibration. Moreover, it is observed in the insert of the figure that this decrease is slightly accelerated above  $500^\circ\text{C}$  for both bonds.

When  $T_d$  increases, the structure is less hydrated as shown from the decrease of the  $H_2O$  and Si-OH vibrations in Figure 4. At the same time, the intensity of  $TO_4$  decreases and the  $TO_3$  peak area increases. Moreover, the attenuation of the  $TO_3$  broadening above  $450\text{ }^\circ\text{C}$  observed in Figure 2, can be correlated to the reduced Si-OH population above this temperature. Based on this assumption, we suggest that at moderate temperature; i.e. below  $600\text{ }^\circ\text{C}$ , the  $SiO_2$  network is composed of cross-linked Si-O-Si units with impurities which impact Si-O-Si asymmetric stretching “out of plane” mode vibrations ( $TO_4$  shoulder) [19],[23],[24]. These impurities can be assigned to hydrogen species ( $H_2O$  and Si-OH) or structural disorder such as space among  $SiO_4$  tetrahedra [53]. At higher temperature, additional Si-O-Si bonds are created

and the amount of impurities decreases. This scenario results in the presence of two populations of Si-O-Si bonds, the classic Si-O-Si bond population which vibrates at the  $TO_3$  vibration frequency and another one represented by the  $TO_4$  vibration mode **or by another asymmetric vibration mode suggested by authors [51,52]**. Si-O-Si groups that vibrate at the  $TO_4$  frequency are the one impacted by impurities [23] or porosities [19], [24]. As soon as the signature of hydrogen species decreases when  $T_d$  increases,  $TO_4$  decreases and the position and shape of the  $TO_3$  change as described. In order to check this assumption and to link the FTIR signature to structural changes, it is essential to further investigate the deconvolution of each contribution, especially in the  $TO_3$ - $TO_4$  region.

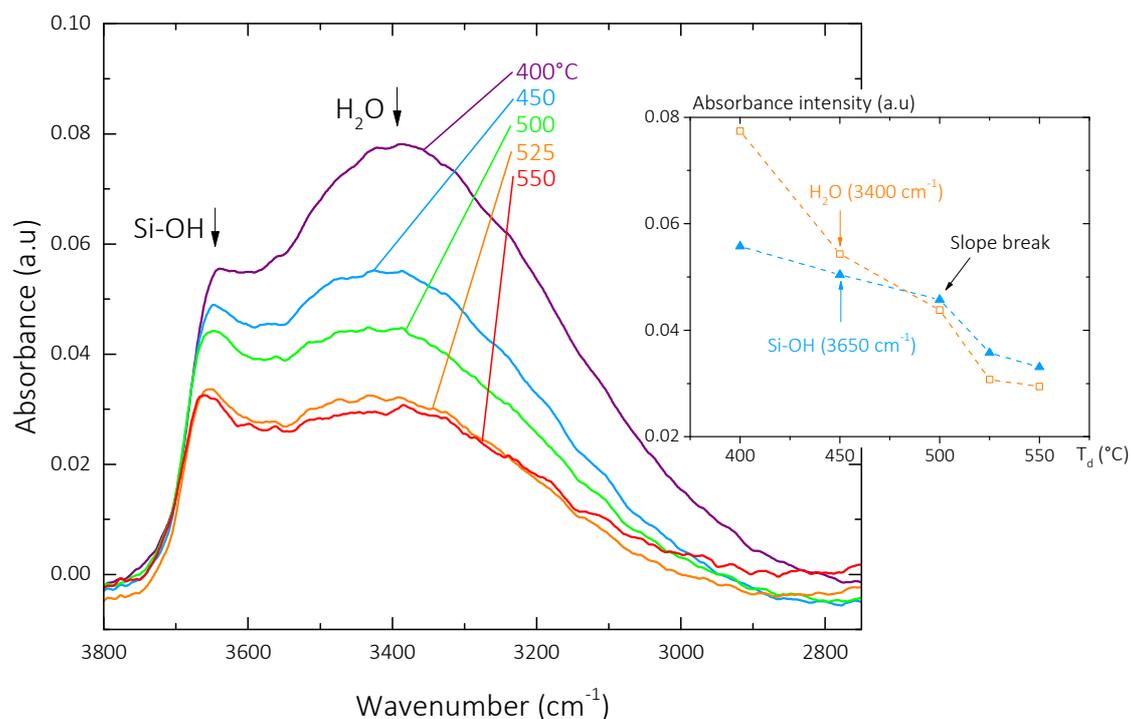


FIG. 4. Zoom in the  $2750 - 3800\text{ cm}^{-1}$  zone of the FTIR spectra containing Si-OH and  $\text{H}_2\text{O}$  vibrational modes, for films processed at different  $T_d$ . Insert: Evolution of the absorbance intensity of  $\text{H}_2\text{O}$  and Si-OH peaks as a function of  $T_d$ .

### 3.1.4 TO<sub>3</sub>-LO<sub>3</sub> splitting

By applying an oblique incidence of the infrared beam on the sample, the interaction of the electromagnetic field with the film is enhanced and the asymmetric stretching mode is split into two components [54], the transverse optical excitation (TO) and the longitudinal optical one (LO), the latter called the Berreman effect [55][56]. Figure 5 presents the evolution of each component of the asymmetric stretching mode for the five T<sub>d</sub>, for an incidence angle of 60° with the sample surface. A similar behavior to the one described hereafter is observed for all other tilt angles. The shape of the TO<sub>3</sub> peak varies when tilted, with non-Gaussian type evolution when T<sub>d</sub> increases, as previously reported. This is attributed to the existence of two populations of the Si-O-Si group. At the same time, the TO<sub>3</sub> position presents a red shift, while the position of the longitudinal component (LO<sub>3</sub>) presents a blue shift. Similar observations were obtained on mesoporous SiO<sub>2</sub> films [57]. Due to TO<sub>4</sub> splitting, Lange *et al.* [54] attributed TO<sub>4</sub> and LO<sub>4</sub> frequencies at 1200 cm<sup>-1</sup> and 1170 cm<sup>-1</sup> respectively. While TO<sub>3</sub> and LO<sub>3</sub> probe the same asymmetric stretching vibration mode, other types of vibration may less bias the LO<sub>3</sub> response. Consequently, the blue shift observed for the LO<sub>3</sub> peak is predominantly attributed to the asymmetric stretching vibration [58],[54].

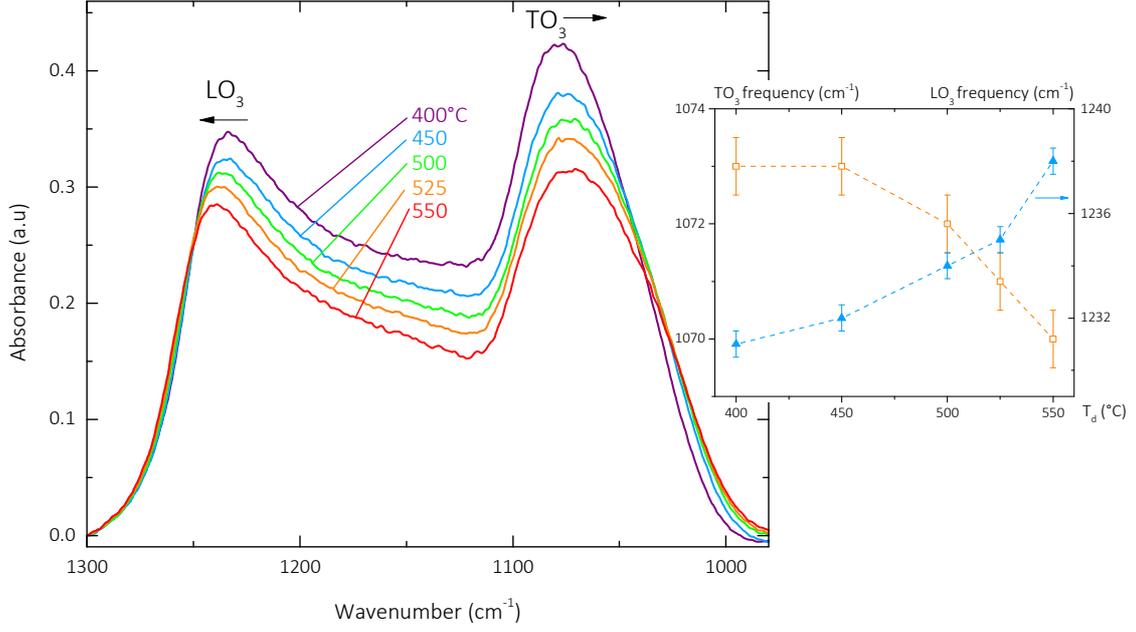


FIG. 5. Zoom in the 1000 – 1300  $\text{cm}^{-1}$  zone of the FTIR spectra, obtained for a  $60^\circ$  tilt of the incident beam with the sample surface, of  $\text{SiO}_2$  films processed at different  $T_d$ . The two  $\text{TO}_3$  and  $\text{LO}_3$  peaks are obtained from the split of the asymmetric stretching mode in these probing conditions. Insert: Shift of the  $\text{TO}_3$  and  $\text{LO}_3$  peaks as a function of  $T_d$ .

In order to confirm the effective densification of the  $\text{SiO}_2$  network with increasing  $T_d$ , as assumed by the blue shift of  $\text{LO}_3$ , we now apply the central-force model which correlates the different Si-O-Si vibration frequencies ( $\omega$ ,  $\text{cm}^{-1}$ ) with the effective force constant ( $\alpha$ ,  $\text{N.m}^{-1}$ ) and the angle ( $\theta$ ) of the bond [15]:

$$\alpha = \frac{1}{2}(\omega_{\text{LO}_2}^2 + \omega_{\text{LO}_3}^2)M_{\text{O}}(1 + 4M_{\text{O}}/3M_{\text{Si}})^{-1} \quad (1a)$$

$$\cos \theta = (\omega_{\text{LO}_2}^2 - \omega_{\text{LO}_3}^2)(\omega_{\text{LO}_2}^2 + \omega_{\text{LO}_3}^2)^{-1}(1 + 4M_{\text{O}}/3M_{\text{Si}}) \quad (1b)$$

where  $M_{\text{O}}$  ( $16 \text{ g.mol}^{-1}$ ) and  $M_{\text{Si}}$  ( $28 \text{ g.mol}^{-1}$ ) are the molar masses of oxygen and silicon. The evolution of the effective force constant and of the angle of the Si-O-Si bond versus  $T_d$  can be determined by measuring the evolution of the  $\text{LO}_2$  and  $\text{LO}_3$  frequencies,  $\omega_{\text{LO}_2}$  and  $\omega_{\text{LO}_3}$ . Since the  $\text{TO}_2$ - $\text{LO}_2$  splitting is negligible,  $\text{TO}_2$  and  $\text{LO}_2$  frequencies are merged and the

evolution of the LO<sub>2</sub> frequency is assumed identical to that of the TO<sub>2</sub> one. Table 1 details the LO<sub>2</sub> and LO<sub>3</sub> shifts, the evolutions of the force constant and of the intertetrahedral bond angle as a function of T<sub>d</sub>. The model predicts that with increasing T<sub>d</sub>, the effective force constant increases and the intertetrahedral bond angle remains stable. It is concluded that, by differentiating the TO<sub>3</sub>-LO<sub>3</sub> pairs, it is possible to isolate the asymmetric stretching contribution and to confirm the densification of the SiO<sub>2</sub> network, despite the red shift which is observed for the TO<sub>3</sub> peak.

TABLE I. TO<sub>2</sub> and TO<sub>3</sub> peak shifts, effective force constant ( $\alpha$ ) and angle ( $\theta$ ) versus T<sub>d</sub>.

T <sub>d</sub> (°C)	$\omega_{LO2}$ (cm <sup>-1</sup> )	$\omega_{LO3}$ (cm <sup>-1</sup> )	$\alpha$ (N.m <sup>-1</sup> )	$\theta$ (°)
400	798	1231	570.2	134.8
450	806	1232	572.1	134.4
500	809	1234	574.2	134.3
525	811	1235	575.7	134.1
550	813	1238	578.1	134.3

### 3.2 Nuclear Analysis

The red shift of the TO<sub>3</sub> vibration mode is assigned to films defects in general, including impurities. We perform ion beam analysis in order to access the composition of the films processed between 400 °C and 550 °C. The results are resumed in Figure 6 in terms of the evolution of the O/Si atomic ratio and of the atomic percentage of hydrogen atoms as a function of T<sub>d</sub>.

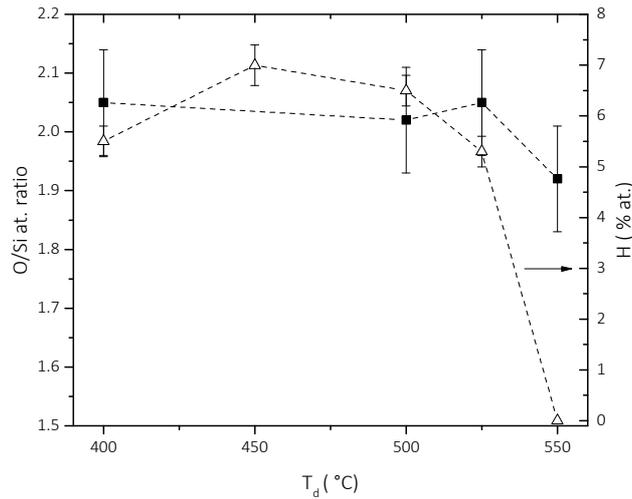


FIG. 6. Evolution of the O/Si atomic ratio (filled squares) and of the atomic percentage of hydrogen atoms (open triangles) as a function of  $T_d$ , probed by ion beam analysis. Dotted lines are guide to the eye.

Between 400 and 525 °C the O/Si atomic ratio remains stable at a value slightly higher than 2, attributed to the presence of H<sub>2</sub>O molecules in the network as was revealed by FTIR. This ratio decreases to ca. 1.93 at 550 °C. However, this evolution remains within the uncertainty of the measurements. At  $T_d$  equal to or lower than 525 °C, the hydrogen concentration varies between  $5.0 \pm 0.3$  and  $7.0 \pm 0.3$  at%. At  $T_d = 550$  °C, it is below the detection limit of ERDA, namely 0.5 at%. These results are consistent with a decrease of the concentrations of water and silanol groups at 550°C. Silanol groups were not probed at 925 cm<sup>-1</sup> by FTIR at this temperature and a low signal was probed for water and silanol groups at 3650 cm<sup>-1</sup>. However, the decrease of water content and silanol groups between 400 and 525 °C, observed by FTIR, is not probed by nuclear analysis. This can be attributed to the conditioning of the samples under secondary vacuum during nuclear analysis, resulting to water desorption from the structure. It can also be attributed to the current density received by the sample under the ion beam which may result in a preferential desorption.

It is worth noting that hydrogen-containing silica films have often been reported in the literature. For example, films processed by plasma enhanced ALD and thermal, ozone involving ALD from various silicon precursors present a relatively high hydrogen concentration, in the range 10 to 13 at.% [59]. Also, films processed by PECVD from hexamethyldisiloxane (HMDSO) and O<sub>2</sub> present a T<sub>d</sub> dependent hydrogen concentration, decreasing from 18.5 at.% at 50 °C to 5.3 at.% at 300 °C [60].

The significant amount of hydrogen probed by nuclear analysis and a non-negligible amount of water probed by FTIR can be assigned to the deposition process since it is a by-product of the CVD reaction between TEOS and oxygen [61]. It can also be due to the post processing adsorption of water in the SiO<sub>2</sub> structure, as reported by some authors [62–64]. For example, Lucovsky *et al.* observed a significant intake of H<sub>2</sub>O after 5 days storage at ambient atmosphere which has a strong effect into the TO<sub>3</sub> band [64].

Also, thermal desorption and FTIR studies of the gas evolution and the microstructure of PECVD silicon oxide grown between 70 and 700 °C from TEOS and oxygens show that between 100 and 300°C, gas desorption is due to adsorbed water in the film during air exposure, originating from liquid water and hydrogen-bonded water molecules at the macropore site [65]. At higher temperature, from 350 °C to 650 °C, gas desorption is related to the isolated silanol bonds at the macro and micropore sites formed during film growth.

### **3.3 Ellipsometry**

The densification of the films involves an evolution of their porosity, taken in the broader sense of the free space available among the SiO<sub>4</sub> tetrahedra. In order to get insight in this parameter, two films processed at 400 and at 550 °C; i.e. the highest and lowest T<sub>d</sub> were characterized by EP. The refractive index, n<sub>550</sub>, of the film processed at 550 °C, determined using the Cauchy model, was shifted from 1.450 to 1.423 after the warm up stage up to 150 °C.

Taking into account that this temperature is too low to induce structural modifications in the film, the evolution of the refractive index gap can be assigned to water removal, contained in the film ( $n_{\text{void}} = 1 < n_{\text{H}_2\text{O}}$ ). Then, ethanol was introduced in the chamber. The value of  $n_{550}$  remained unchanged to 1.423. EP measurements were not sensitive enough to determine a possible porosity in the films. In the literature, ultra-microporous  $\text{SiO}_2$  films exhibit a porosity smaller than 1 nm [66]. Studies were done in order to understand the adsorption of water molecules inside 1.04, 1.96 and 2.88 nm diameter silica pores [67]. The proposed scenario considers the coverage of the surface of each pore by silanol groups. A first stage, defined by a preferential adsorption of the water molecule near silanol groups, is followed by the formation of a water monolayer over the entire pore surface. In the presence of excess water molecules, pores are completely filled with water.

The evolution of the refractive index versus  $T_d$  using the Sellmeier models is presented in Figure 7. Two regimes are observed; below  $T_d = 500$  °C the refractive index remains stable. Above 500 °C, the refractive index decreases. Two scenarios are reported in the literature for the correlation between the characteristics of the  $\text{SiO}_2$  films and their refractive index: for stoichiometric films, the refractive index increases when the network becomes denser due to the decrease of porosity content that contain void or water with a refractive index of the environment lower than that of  $\text{SiO}_2$  [68]. For non-stoichiometric  $\text{SiO}_x$  films, the refractive index decreases when the  $\text{SiO}_2$  network becomes denser [69]. This behavior is assigned to the incorporation of oxygen in the structure, this enrichment goes with a decrease of the refractive index because  $n_{\text{Si}} > n_{\text{SiO}} > n_{\text{SiO}_2}$ .

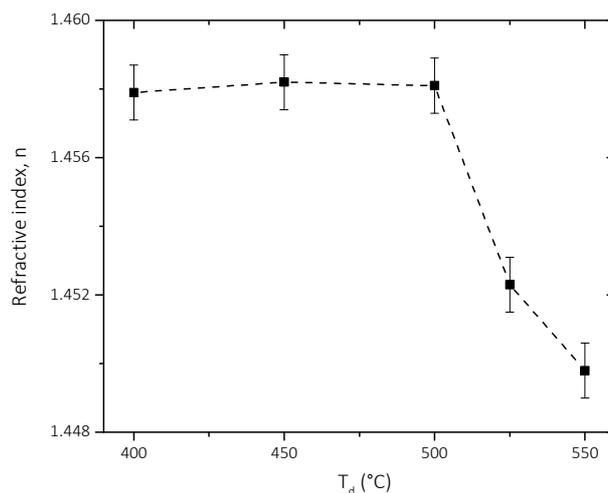


FIG. 7. Evolution of the refractive index of the SiO<sub>2</sub> films processed between 400 °C and 550 °C determined by the Sellmeier model.

At  $T_d$  up to 500 °C, the stability of the refractive index can be attributed to a film composition that remains stable as suggested by nuclear analysis. However, this conclusion does not fit FTIR results, which reveal a decrease of the H<sub>2</sub>O and Si-OH FTIR vibration modes when  $T_d$  increases. To counterbalance the expected decrease of the refractive index due to the dehydration of the layer, the incorporation of oxygen in the network is considered when  $T_d$  increases above 500 °C. At this temperature range, the incorporation of oxygen can indeed be assigned to the generation of new Si-O-Si bonds. This result suggests an increase of the structure densification when  $T_d$  increases.

### 3.4 Corrosion resistance

Pliskin proposed in 1964 an etching solution in order to comparatively investigate the resistance of various glass and oxide film systems [32]. Since then, the Pliskin etch or P-etch has been commonly adopted in order to evaluate the impact of the type of deposition technique, of the chemistry and of the different process parameters on the resistance to the involved corrosive solution of the SiO<sub>2</sub> films. Figure 8 presents the P-etch rate (Å.s<sup>-1</sup>) of the films

deposited between 400 °C and 550 °C. The resistance to the corrosive solution is clearly improved with increasing  $T_d$ , from 18 Å.s<sup>-1</sup> to 10 Å.s<sup>-1</sup>. These results favorably compare with those of previously reported SiO<sub>2</sub> films processed below 600 °C by PECVD [70] or using ozone [71], for which the P-etch rate varies between 8 and 53 Å.s<sup>-1</sup>. An annealing at 800 °C during 30 min of the films (independently of  $T_d$ ) presents a P-etch rate of 2 Å.s<sup>-1</sup>; *i.e.* similar to the denser silica film obtained by pyrolysis at temperature higher than 975°C [13]. This behavior is indirectly linked to the porosity, the composition, the stoichiometry and ultimately to the density of the films [13].

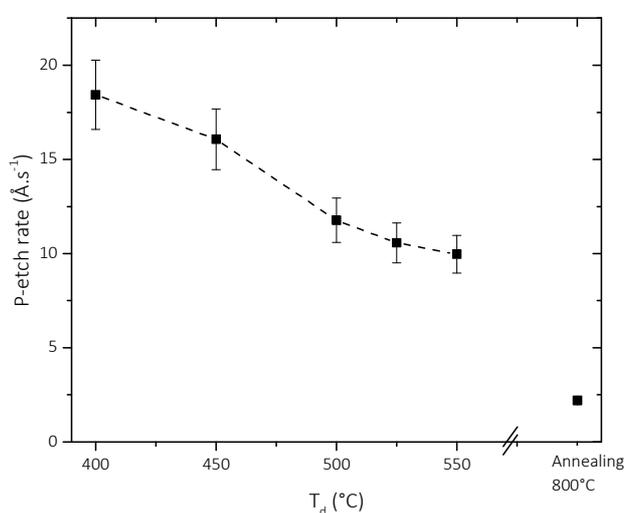


FIG. 8. P-etch rate versus  $T_d$  and post-annealing at 800°C for the SiO<sub>2</sub> films.

## 4. CONCLUSIONS

An original methodology is developed in order to propose a comprehensive scenario for the densification mechanism of CVD SiO<sub>2</sub> films processed from TEOS and oxygen in the temperature range 400 – 550 °C. The combination of different characterization techniques allows assigning the densification process first to a dehydration of the film based on the decrease of the water and silanol peaks, revealed by FTIR. A significant hydrogen content

between  $5 \pm 0.3$  %at. and  $7 \pm 0.3$  %at. is determined by nuclear analysis for films processed below  $T_d = 525$  °C. Films processed at  $T_d = 550$  °C are hydrogen free and present a slightly substoichiometric structure. Ellipsometric porosimetry allows concluding that if the films present porosity, the diameter of this porosity might be smaller than the detection limit of EP measurements. The corrosion resistance, probed by the P-etch rate test, is improved from 18 to  $10 \text{ \AA}\cdot\text{s}^{-1}$  with increasing  $T_d$ ; *i.e.* with the Si-O-Si network densification. Upon annealing at 800 °C for 30 min, the P-etch rate is further improved, decreasing to  $2 \text{ \AA}\cdot\text{s}^{-1}$ , the lowest reported value for SiO<sub>2</sub>, similar to the one of films obtained from the pyrolysis of TEOS at almost 1000 °C. This test indirectly confirms the densification of the film with increasing  $T_d$ . The integration of the reported various experimental information confirms the proposed mechanism and indicates that amorphous SiO<sub>2</sub> films processed from TEOS and oxygen at 550 °C, are dense and thus present high normalized corrosion resistance.

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