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Modeling of the refractive index and composition of luminescent nanometric chlorinated-silicon nitride films with embedded Si-quantum dots

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The refractive index of nanometric (<100 nm) chlorinated-silicon nitride films with embedded silicon quantum dots, prepared by remote plasma enhanced chemical vapor deposition was investigated by spectroscopic ellipsometry. The complex refractive indexes and thicknesses of the films were obtained from the ellipsometric measurements using the Cauchy model. The chemical composition of the bulk of the films, including the hydrogen, chlorine, and oxygen content, was measured by elastic forward analysis (EFA) and X-ray photoelectron spectroscopy (XPS). The EFA and XPS results indicate that the films are composed of nearly stoichiometric silicon nitride (Si_3N_4) with different amounts of hydrogen, chlorine, and oxygen, and that their top surface is oxidized. Based on the composition results the refractive index of the films was fitted using the effective medium approximation considering the system: Si substrate + Si_3N_4 + Si + voids + top ultrathin oxidized surface layer.

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I. INTRODUCTION

Luminescent and down-converting silicon quantum dots (Si-QDs) embedded in silicon nitride and silicon dioxide thin films continue being of great interest to develop silicon based devices for photonic, optoelectronic, and photovoltaic applications.^{1–7} For several specific applications and investigations, the control and determination of the refractive index and composition of these films are very important. Spectroscopic ellipsometry has been widely used to measure the real and imaginary parts of the refractive index or dielectric function, and the thickness of different luminescent nanostructured silicon-rich silicon nitride films.^{5,8–13} Many of these cited works have shown that there is close correlation among the optical parameters of the films, and their composition, structure, and luminescent properties.

Based on the values of the real part of refractive index at 632 nm for pure Si (3.88), Si_3N_4 (2.0), and SiO_2 (1.45), the relationships between molar composition and refractive index for silicon-silicon nitride and silicon-silicon oxide compounds can be predicted.^{5,14} Values progressively higher than 2.0 are expected for silicon rich silicon nitride (SiN_x , $x < 1.33$) films as they become silicon richer and their atomic ratio $x = [\text{N}]/[\text{Si}]$ decreases. On the other hand, a linear decrease from 2.0 to 1.45 is expected in the values of the refractive index of silicon oxynitride as the oxygen content

increases.¹⁴ However, the values and the empirical relationship between refractive index and x depend significantly on the composition of the films and the type and amount of other impurities, which in turn depend on the deposition method and experimental conditions. For example, for amorphous silicon rich silicon nitride films deposited by radio frequency magnetron sputtering which are free of hydrogen and oxygen, the refractive index (at 632.8 nm or 1.95 eV) increases from ~ 1.83 to 3 as x , measured by Rutherford Back Scattering (RBS), decreases from 1.33 to 0.5.⁵ For silicon rich silicon nitride films deposited by chemical vapor deposition techniques, which contain significant amounts of hydrogen, the refractive index increases but in a lower range from ~ 1.8 to 2.45 as x , measured by X-ray photoelectron spectroscopy (XPS), decreases from 1.0 to 0.5.¹² In a previous work, we have found that luminescent chlorinated-hydrogenated SiN_x films with embedded silicon quantum dots, prepared by remote plasma enhanced chemical vapor (RPECVD) deposition, have $x = [\text{N}]/[\text{Si}]$, measured by XPS, in the range from 1.16 to 0.7, but, however, the refractive index of the films at 632 nm remains almost constant with low values ~ 1.82 – 1.85 .^{7,15,16} Based on the general trends mentioned above for the refractive index of silicon-silicon nitride and silicon-silicon oxide compounds, it seems contradictory that silicon rich silicon nitride films have refractive index values lower than 2.0. The explanation of the origin of the low refractive index of these films in terms of their composition was the main motivation for the present work.

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One question that complicates finding the correlation between the composition and refractive index of the films is that some of the most common techniques used to analyze the composition, such as XPS and RBS, do not detect hydrogen.^{17,18} Thus when the films contain hydrogen, other techniques, such as infrared spectroscopy (IR), elastic recoil detection analysis (ERDA) with light ions, or hydrogen forward scattering (HFS), have to be used to determine separately the hydrogen concentration.^{19,20} Although IR is a relatively inexpensive and common technique used to measure the H content, it requires a careful spectra interpretation in order to obtain reliable quantitative results;²⁰ furthermore for quantitative analysis of the nanometric films, it is less reliable due to the very poor quality (low signal to noise ratio) of the IR spectra.¹⁵ Another complication with the analysis of the composition is that the SiN_x films usually suffer oxidation when exposed to air and forms a silicon oxide layer or a graded oxynitride film.²¹ Since this oxidation reduces the amount of nitrogen at the film surface and since the XPS only identifies the chemical species at the top surface of the SiN_x films,¹⁷ the [N]/[Si] ratios determined by XPS will tend to be lower than the real [N]/[Si] ratio in the bulk of the film. On the contrary, the RBS technique provides quantitative analysis of the elements in all the bulk of the films with thickness up to 1000 nm, but the depth resolution is of the order of 10 nm, which makes difficult to resolve a thin oxide layer.¹⁸ Additionally, in order to make a good correlation between refractive index and composition it is convenient to have information on the short range structure of silicon rich-SiN_x films, to decide which of the two extreme models is the adequate to describe this structure.²² According to the random mixture model (RMM), the SiN_x film consists of separated phases of Si₃N₄ and Si phases, whereas in the random bonding model (RBM) the SiN_x film is considered as a statistical distribution of Si-Si_{4-n}N_n (n = 0, 1, 2, 3, 4) bonds.^{12,22}

In this work, we have used the elastic forward analysis (EFA) technique,²³ combined with XPS and spectroscopic ellipsometry to investigate the composition and refractive index of luminescent chlorinated-hydrogenated nanometric SiN_x films (~80–90 nm) with embedded silicon quantum dots, deposited by remote plasma enhanced chemical vapor (RPECVD) deposition. We used the XPS technique to investigate the depth of the oxide layer formed by oxidation of the top surface of the films, and EFA was used to measure the hydrogen content and bulk composition of the films. Based on the XPS and EFA results, the ellipsometric data were accurately fitted using the effective medium approximation (EMA), assuming that the bulk of the films is mainly composed of nearly stoichiometric silicon nitride, with Si and void phases embedded, and an ultrathin silicon oxide layer at the film top surface.

II. EXPERIMENTAL

For this study, a series of four silicon-rich silicon nitride films with approximately the same thickness but different size of Si-QDs were prepared using gas mixtures of SiH₂Cl₂/NH₃/H₂/Ar, and the RPECVD system reported previously.⁷

The films were deposited on the mirror polished face of single-crystalline (100) pieces of silicon wafers with resistivity of 200 Ω cm. The substrate temperature for deposition was 300 °C. The flow rates of SiH₂Cl₂, H₂, and Ar were kept constant at 5, 20, and 75 sccm, respectively. In order to change the size of the Si-QDs embedded in the films, and based on our previous experimental works,^{7,16} the flow rates of NH₃ chosen for depositing each of the four different samples were 50 (S1_50), 100 (S2_100), 200 (S3_200), and 300 (S4_300) sccm. Also based on the knowledge of the deposition rate from our previous works, the table of colors of Si₃N₄ films at various ranges of film thickness,²⁴ and the direct naked eye view of the growing films, the deposition time was controlled for each film to obtain nanometric films with thickness in the range from 77 to 93 nm (blue color). After deposition, the thickness and refractive index of the films were measured at the fixed wavelength of 632.8 of the He-Ne laser of a Manual L117 ellipsometer, using an incidence angle of 70°. Spectroscopic ellipsometry measurements were further performed at room temperature at several incident angles (45°, 55°, 65° and 75°) in the range from 300 to 1700 nm using a J. A. Woollam VASE spectroscopic ellipsometer. The simultaneous analysis of the spectroscopic ellipsometry and transmittance data was performed using the WVASE32 software (Woollam Co. Inc.). The X-ray Photoelectron Spectroscopy (XPS) measurements were made using a Jeol JS 9200 XPS. The spectra were acquired by exciting the films with an Al-K α X-ray source in the low and high-resolution regimes. Additionally, samples were characterized by elastic forward analysis (EFA);²³ in this case, a 12.0 MeV C²⁺ ion beam from a Tandem Van de Graff accelerator was used. The angle between the ion beam and the sample surface was fixed at 30° while the angle between the detected particles and the incident ion beam was fixed at 45°.

III. RESULTS AND DISCUSSION

In order to determine the refractive index (n) of the different films as a function of wavelength (λ), the ellipsometric parameters Ψ and Δ were measured for each sample at multiple incident angles in the reflection mode. The obtained data were analyzed using the WVASE32 software (Woollam Co. Inc.) that allows multiangle fitting of the spectra using the transfer matrix (Abelés) formalism and assuming a 3 layer model: Si substrate/film/surface layer. Fig. 1 shows the ellipsometric data and fit of Ψ and Δ for sample S3_200. The fit was quite good and the MSE (mean square error) was very low: lower than 7 for all the samples. The experimental data were fitted to calculate the refractive index $n(\lambda)$ of the films, considering extinction coefficient $\kappa = 0$ and using the model of Si substrate + the Cauchy dispersion relation for the film

$$n(\lambda) = A + \frac{B}{\lambda^2} + \frac{G}{\lambda^4},$$

where A , B , and C are the fit coefficients.

Figure 2 shows the refractive index calculated in this way for the four samples. When $\kappa(\lambda) \neq 0$ was taken into account in the fittings, it had very low values; lower

FIG. 1. Measured (dots) and fitted (lines) ψ and Δ for sample S3_200.

than 0.001, which are below the limit of resolution of ellipsometry.

Fig. 3 shows the XPS survey spectra of one of the samples (S3_200) without erosion and after 60 s of erosion with 2 keV argon ions. In both spectra, it can be seen the presence of Si, N, O, and Cl. It is important to note that the signal corresponding to O is reduced significantly in the spectrum of the eroded film. It is also noticeable that the C signal disappears with the erosion, indicating that the presence of carbon is due to surface contamination. It is worth to mention that the XPS spectra of different samples were very similar.

Fig. 4 shows the depth profile of oxygen for the same sample S3, which reveals a fast decay of the oxygen content in the first 60 s of erosion, which indicates that the oxygen comes from oxidation of the film, and that the oxide surface layer attains a saturation thickness, which is very small compared with the film thickness.

Figure 5 shows the high resolution XPS spectra corresponding to the O 1s peak without erosion and after 90 s of erosion. It is observed that the intensity of the O 1s peak decreases and the position of the peak is shifted towards lower energies as a result of erosion time. The peak at 532.8 eV observed in the spectrum of the sample without erosion, the surface of the film, is assigned to O1s in SiO_2 . This result indicates that the oxygen in/and close to the surface of the film is bonded to Si atoms forming SiO_2 . After

FIG. 2. Refractive index spectra for all the samples, calculated using the Cauchy dispersion relation.

FIG. 3. XPS survey spectra of sample S3_200 without erosion (top) and after 60 s of erosion (bottom) with 2 keV argon ions.

FIG. 4. XPS depth profile of oxygen for the same sample S3_200.

erosion, the peak appears at 532.1 eV attributed to bonds O-Si as well as to O 1s in $\text{Si}(\text{OH})_4$ suggesting that in the interior of the film it seems to be incorporated as OH in a very small amount.

FIG. 5. High resolution XPS spectra corresponding to the O 1s peak for the sample S3_200 without erosion and after different times of erosion.

FIG. 6. High resolution XPS spectra corresponding to the Si 2p peak for sample S3_200 without erosion and after 90 s of erosion.

The high resolution XPS spectra corresponding to Si 2p is shown in Fig. 6. From this figure, one can see that at the surface and until a depth corresponding to 90 s the Si atoms are mainly oxidized indicating again that there is a surface layer of SiO₂, and in the interior of the film, there is Si₃N₄ in higher proportion.

Figure 7 shows the resolved N 1s XPS spectra, in which it is observed that the nitrogen atoms are bonded to Si atoms forming Si₃N₄. However, the amount of N-bonded to Si is lower at the surface and increases with the depth inside the film indicating again that some N-Si bonds are replaced by O-Si bonds due to the oxidation of the film surface. Additionally, it is observed a signal around 403.4 eV attributed to NH₄, indicating the presence of H into the film.

Fig. 8 shows the EFA spectra for samples S3_200 and S4_300. Besides the signals of the elements Si, N, O, Cl, the signal due to H is clearly observed. The presence of carbon could be attributed to backscattered C ions used as projectile.

The composition of all the films calculated from the EFA measurements is given in Table I. The uncertainty of these measurements was $\pm 15\%$ for our experimental conditions.

FIG. 7. High resolution spectra of the N 1s peak for sample S3_200 without erosion and after 90 s of erosion.

FIG. 8. EFA spectra for samples S3_200 and S4_300 using projectiles of C⁴⁺ of 12.0 MeV.

According to Table I, although it is apparent that the N/Si ratio increases slightly from sample S1 to sample S4, the average of the N/Si ratio for all the films is $N/Si \approx 1.35 \pm 0.15$, which, given the uncertainty of the measurements, correspond to that of stoichiometric silicon nitride ($N/Si \approx 1.33$). Thus, the XPS and EFA results indicate that the bulk of all the films in average can be considered as composed of nearly stoichiometric silicon nitride (Si₃N₄), containing a significant amount of hydrogen (~ 12 – 17 at.%), and some chlorine (~ 4 at.%) and oxygen as OH (~ 2 at.%). Since all these elements or species form terminal bonds (Si-H, Si-Cl, Si-OH), we can assume that they generate nanovoids or nanopores inside the Si₃N₄ matrix, in a similar way as the Si-F bonds generate nanopores in fluorinated SiO₂ films.^{25,26} On the other hand, the HRTEM images of all these films indicate the presence of silicon nanoparticles.⁷ Based on all the preceding results, we assumed that the random mixture model (RMM) is the adequate to describe the short range structure of our SiN_x films. As a consequence of this and to model their refractive index, they can be considered to consist of separated phases of Si₃N₄, Si phases, and voids, as shown in Fig. 9.

Based on the XPS results that indicate that all the silicon nitride films suffer surface oxidation, the refractive index of the films were fitted using the Maxwell-Garnet expression of the effective medium approximation (EMA),²⁷ considering the system: Si₃N₄ + Si + voids + (top ultrathin oxidized surface layer). In this approach, we considered Si₃N₄ as the host material and Si and voids as the diluted phases. The refractive indices used for Si₃N₄ and Si were 2.0 and 3.85, respectively. Although by ellipsometry it is not possible to distinguish if the nanovoids are filled with hydrogen, oxygen, or air, we used the % of void in the EMA model to indicate

TABLE I. Composition of the films calculated from the EFA analysis.

Sample	Si (at. %)	N (at. %)	Cl (at. %)	O (at. %)	H	N/Si
S1_50	35	41	4	3	17	1.2
S2_100	34	43	4	2	17	1.3
S3_200	33.5	47	3.5	1	15	1.4
S4_300	33.5	48	4.5	1	12	1.5

FIG. 9. Model used to correlate the refractive index of the films with their composition using the effective medium approximation.

TABLE II. Thickness of the surface and EMA layers, and percentages of Si₃N₄, Si, and voids obtained from the EMA model. MSE is the mean square error of the fits.

	S1_50	S2_100	S3_200	S4_300
Thickness (nm) surface layer	5.9	4.9	5.1	3.3
Thickness (nm) EMA layer	77.7	91.2	76.7	84.9
%Si ₃ N ₄	74.5	77.0	79.9	80.8
%VOID	21.8	20.8	19.1	18.9
%Si	3.8	2.2	1.0	0.3
MSE	7	2	3	5

that we have a percentage of diluted material with a refractive index equal to 1.0. Under the previous assumption on the composition of the films and refractive indices of the host and diluted phases, the fit was quite good for all the samples. Table II shows the percentage of each EMA region for each film obtained from the fits. The high percentages of Si₃N₄ in this table indicates that this is the host medium of the films, in which the void and Si phases are embedded in smaller proportion. As expected, the percentage of Si₃N₄ increases with the decrease in the percentage of Si, as the flow rate of NH₃ used during deposition is increased. The at. % of Si and N was estimated using the data of Table II and the formulas: at.% Si = $3/7(\%Si_3N_4) + \%Si$ and at. % N = $4/7(\%Si_3N_4)$. The obtained results are shown in Table III.

The data of Table II also show that the % of the void content in the films decreases from sample S1_50 to sample S4_300. It should be noted that, comparing the data of Tables I and II, the void percentage is in good agreement with the percentage of (H + O + Cl) measured by EFA. On the other hand, according to the data of Table III, the relative silicon excess decreases and the nitrogen content increases from sample S1_50 to sample S4_300, and this behavior is expected since the flow rate of the NH₃ increases from sample S1_50 to sample S4_300. The apparent decrease in the thickness of the ultrathin oxide surface layer observed in Table II is consistent with the decrease in the oxygen content

TABLE III. At. % of Si and N for the films, considering the Si₃N₄ matrix and the Si phase embedded as silicon quantum dots in this matrix.

	S1_50	S2_100	S3_200	S4_300
Si	35.7	35.2	35.2	34.9
N	42.6	44.0	45.7	46.2
N/Si	1.19	1.25	1.30	1.32

in the bulk of the films observed in Table I, as going from sample S1_50 to sample S4_300, and indicates that the films are more resistant to post-deposition oxidation as the NH₃ increases.

IV. CONCLUSION

We can conclude that the low refractive index of the films, despite that they are silicon rich, is due the large percentage of voids in the bulk of the films. The invariability of the refractive index among these films, despite that they were deposited at different NH₃ flow rates, is due to the fact that as they become silicon rich they contain a higher amount of voids, so the increase in the refractive index of the films due to the increase of the Si phase content is compensated with the decrease in the refractive index due to the increase of the content of voids. Finally, it is observed that the ultrathin oxide layer formed at the surface of the films by post deposition oxidation has negligible influence on their optical properties.

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- ¹B. Sain and D. Das, *J. Lumin.* **158**, 11 (2015).
- ²F. Priolo, T. Gregorkiewicz, M. Galli, and T. F. Krauss, *Nat. Nanotechnol.* **9**, 19 (2014).
- ³O. Blázquez, J. López-Vidrier, S. Hernández, J. Montserrat, and B. Garrido, *Energy Procedia* **44**, 145 (2014).
- ⁴D. Hiller, A. Zelenina, S. Gutsch, S. A. Dyakov, L. López-Conesa, J. López-Vidrier, S. Estradé, F. Peiró, B. Garrido, J. Valenta, M. Kořínek, F. Trojánek, P. Malý, M. Schnabel, C. Weiss, S. Janz, and M. Zacharias, *J. Appl. Phys.* **115**, 204301 (2014).
- ⁵O. Debieu, R. P. Nalini, J. Cardin, X. Portier, J. Perrière, and F. Gourbilleau, *Nanoscale Res. Lett.* **8**, 31 (2013).
- ⁶W. R. Taube, A. Kumar, R. Saravanan, P. B. Agarwal, P. Kothari, B. C. Joshi, and D. Kumar, *Sol. Energy Mater. Sol. Cells* **101**, 32 (2012).
- ⁷A. Rodríguez, J. Arenas, and J. C. Alonso, *J. Lumin.* **132**, 2385 (2012).
- ⁸A.-S. Keita, A. E. Naciri, F. Delachat, M. Carrada, G. Ferblantier, and A. Slaoui, *J. Appl. Phys.* **107**, 093516 (2010).
- ⁹P. Basa, P. Petrik, M. Fried, L. Dobos, B. Pécz, and L. Tóth, *Phys. E* **38**, 76 (2007).
- ¹⁰Z. H. Cen, T. P. Chen, L. Ding, Y. Liu, M. Yang, J. I. Wong, Z. Liu, Y. C. Liu, and S. Fung, *Appl. Phys. Lett.* **93**, 023122 (2008).
- ¹¹V. E. Vamvakas, N. Vourdas, and S. Gardelis, *Microelectron. Reliab.* **47**, 794 (2007).
- ¹²J. F. Lelièvre, E. Fourmond, A. Kaminski, O. Palais, D. Ballutaud, and M. Lemi, *Sol. Energy Mater. Sol. Cells* **93**, 1281 (2009).
- ¹³V. E. Vamvakas and S. Gardelis, *Surf. Coatings Technol.* **201**, 9359 (2007).
- ¹⁴T. Bååk, *Appl. Opt.* **21**, 1069 (1982).
- ¹⁵A. Rodríguez-Gómez, A. García-Valenzuela, E. Haro-Poniatowski, and J. C. Alonso-Huitrón, *J. Appl. Phys.* **113**, 233102 (2013).
- ¹⁶A. Rodríguez, J. Arenas, A. L. Pérez-Martínez, and J. C. Alonso, *Mater. Lett.* **125**, 44 (2014).
- ¹⁷B. D. Rantner and D. G. Castner, in *Surface Analysis: The Principal Techniques*, edited by John C. Vickerman (Wiley, 2002), pp. 43–98.
- ¹⁸E. Taglaier, in *Surface Analysis: The Principal Techniques*, edited by J. C. Vickerman (Wiley, 2002), pp. 215–266.

- ¹⁹W. Bohne, W. Fuhs, J. Rohrich, B. Selle, G. Gonzalez-Diaz, I. Martil, F. L. Martinez, and A. del Prado, *Surf. Interface Anal.* **30**, 534 (2000).
- ²⁰G. Morello, *J. Non-Cryst. Solids* **187**, 308 (1995).
- ²¹S. I. Raider, R. Flitsch, J. A. Aboaf, and W. A. Pliskin, *J. Electrochem. Soc.* **123**, 560 (1976).
- ²²V. A. Gritsenko, R. W. M. Kwok, H. Wong, and J. B. Xu, *J. Non-Cryst. Solids* **297**, 96 (2002).
- ²³S. Romero, M. Fern, G. Murillo, and H. M. Berdejo, *Nucl. Instrum. Methods Phys. Res.* **194**, 164 (2002).
- ²⁴W. V. G. F. Reizman, *Solid State Electron.* **10**, 625 (1967).
- ²⁵J. C. Alonso, X. M. Díaz-Bucio, E. Pichardo, L. Rodríguez-Fernández, and A. Ortiz, *Thin Solid Films* **474**, 294 (2005).
- ²⁶J. H. Lee and C. K. U. Hwangbo, *Surf. Coatings Technol.* **128–129**, 280 (2000).
- ²⁷D. E. Aspnes, *Thin Solid Films* **89**, 249 (1982).