Effect of Bridging and Terminal Alkyl Groups on Structural and Mechanical Properties of Porous Organosilicate Films

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Chemical composition, pore structure and mechanical properties of porous organosilicate low-k films with terminal and bridging organic groups are studied. It is shown that BTMSE based low-k films with alkyl bridge between the Si atoms have non-uniform pore structure with internal voids (ink-bottle like pores) while the films with terminal alkyl groups have pores with cylindrical shape. All studied spin-on deposited films have better mechanical properties than PECVD films with similar chemical composition and porosity. The best mechanical properties showed alkyl bridged low-k films. Moreover, the Young’s modulus increases with BTMSE content because of higher concentration of bridging bonds.

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Porous organosilicate glasses (OSG) have many different applications such as catalysts, adsorbents, trapping agents, drug delivery agents, stationary phases in chromatography and chemical sensors. One of the most important applications of porous OSG films is related to interconnects in advanced ULSI (ultra large-scale integration) devices where they are used as insulators with low dielectric constant. Modern ICs can be made very compact, incorporating up to several billions transistors and other electronic components in an area of about 1 cm². All transistors and other IC components have to be electrically interconnected to provide the proper functionality. The width of the conducting lines that connect different transistors in a circuit is becoming very small; in 2008, it dropped below 100 nm, and now is of the order of a few tens of nm. In parallel, the interconnect delay is becoming an increasing limitation factor of the overall signal propagation delay. The low dielectric constant materials are needed to reduce capacitance (C) between the metal (Cu) conductors. Together with low resistivity (R) of metal wires, low-k materials improve ULSI device performance by decreasing RC delay, cross talk noise and power consumption in interconnects.

Different types of organosilicate films deposited by plasma enhanced chemical vapor deposition (PECVD) and spin-on glass (SOG) technology have been developed during the last 20 years. Using of appropriate precursors allows to control the chemical composition and porosity of these films. PECVD is currently the method of choice in microelectronic industry because this technique is easily integrated in the existing device manufacturing process. PECVD precursors (normally different types of organosilanes) are transferred to the reaction chamber where Si wafer is located on a heated pedestal. Quality of the deposited films depends on plasma characteristics and pedestal temperature. In the modern interconnect technology, the dielectric layer is deposited by PECVD and patterned first before the metal deposition using lithography and plasma etching. Lithography and plasma etching define the structure of interconnect wiring by forming certain patterns in the dielectric film. Then copper should fill the patterns in the dielectric film by superfilling electrodeposition and after chemical-mechanical polishing (CMP) forms one level of metallization structure where the deposited Cu wires are isolated by low-k dielectric. This technology is an additive technique and termed as damascene technology. However, the plasma patterning of low-k dielectrics is becoming a big challenge. Plasma species (F, O radicals) deeply diffuse into pores and degrade the low-k properties. The depth of plasma species penetration is described by diffusion-recombination or random walk models and increases with pore size. Therefore ultralow-k materials (k < 2.5) developed for future technology nodes strongly suffer on plasma damage because of larger porosity and pore size. Large pore size makes also low-k films badly compatible with deposition of thin diffusion barriers that are required for advanced technology nodes. This is the reason why alternative integration schemes with metal patterning (subtractive approach) are attracting more and more attention. In this case, the dielectric deposition after metal patterning must have a good gap filling capability but the plasma etch step of low-k dielectric is excluded. The gap filling is fundamentally a challenging task for PECVD and therefore other deposition techniques must be evaluated. In the case of spin-on OSGs, the synthesis occurs by means of a sol-gel reaction. As starting precursors, different organosilanes can be used. By adding a solvent, water and a catalyst (acid or base) the precursor starts to hydrolyze and will condensate with another hydrolyzed precursor molecule forming an inorganic polymer film. The films can also be made by using just one starting precursor or by co-condensation of different precursors. One of the most important advantages of spin-on deposition using sol-gel reactions is the perfect gap filling capability. Sol-gel processing also allows to shape materials into complex geometry in a gel state and to deposit well organized solid product with predictable properties. These advantages make sol-gel technology potentially important for future ULSI technology because of controllable possibility to scale porosity and dielectric constant, and also possibility of modification of chemical composition and mechanical properties.

Different classes of low-k materials can be deposited by using spin-on technique. Organic polymers were popular in the beginning of CuLow-k interconnect technology (1996-2003) because of low polarizability of chemical bonds. However, their integration into ULSI device was too difficult because of not sufficiently good mechanical properties. Particularly, the high coefficient of thermal expansion (CTE) in comparison with other components of interconnect structure has been a critical challenge for the device reliability.

OSG materials selected as the most important low-k materials for ULSI devices have silica type matrix but part of bridging oxygen atoms in siloxane structure (=Si-O-Si=) are replaced by terminal organic groups (=Si-CH₃) (Figure 1, A,B). The organic terminal groups make OSG materials hydrophobic and allow to avoid adsorption of water molecules that have very high dielectric constant (k ≈ 80), increases the leakage current and reduces breakdown field. Depending on the properties of introduced organic groups they can be terminal or crosslinking (bridging) (Figure 1). The materials with bridging alkyl groups can also form ordered porosity, so called periodic mesoporous organosilicates (PMO). There are several publications where OSG low-k materials with terminal and bridging alkyl groups are compared...
from the point of view of their properties important for ULSI devices. Improvement of mechanical properties of the materials with carbon bridges between Si atoms has already been reported.14-15 The pore ordering can also bring additional improvement.16-18 Better resistance to plasma species and VUV light of low-k films with carbon bridges was predicted, but recent studies show that resistance to VUV light is not so obvious and depends on the length of the bridging carbon groups and their UV absorption spectra.16

Both plasma resistance and mechanical properties improvement are important for practical implementation of porous OSG films. However, precise comparison of the mechanical properties is a challenging task because their change due to the porosity and pore structure variation might be critical. Additionally, successful implementation of porous materials into integrated circuit design requests optimal balance of different properties. Improvement of mechanical properties is certainly very important but also we need to be sure that other properties (structural and electrical properties, reliability) still meet the necessary requirements.

It is necessary to mention that the matrix of completely cured OSG low-k materials is generally silica like and the terminal organic hydrophobic agents are mainly located on the pore wall.17 Introduction of the bridging alkyl groups is replacing bridging oxygen atoms in siloxane structure. Therefore improvement of mechanical properties is related to difference in the bond strength between Si-O-Si and Si-CH3 structures. On the other hand, these 2 structures have different affinity to adsorption of water molecules. The alkyl bridged group is more hydrophobic that the siloxane group and therefore less terminal CH3 groups might be needed to keep the films hydrophobic. Some of these questions are still open and this is the reason why alkyl bridged OSG materials have not been implemented into ULSI production and the interest to the correlation between their different properties still remains.

In this work we studied effect of terminating and bridging alkyl groups by comparative evaluation of porous low-k materials with careful study of their chemical composition and porosity. The total concentration of carbon in these films was also kept close to each other in different films.

Experimental

Two groups of porous samples were prepared to study the effect of network structure modification on the film properties. The first group includes classical sol-gel materials prepared by co-hydrolysis in alcoholic solution of tetraethoxysilane (TEOS) and methyltriethoxysilane (MTEOS): methyl-modified silicate (see e.g. Ref. 18), whose structure contains silicon-oxygen network with terminal methyl groups (Samples Nos: C35, C45, see Table I), Brij 30 (C12H25(OCH2OCH2)4OH with molar mass 362 g/mole, Sigma-Aldrich) was used as a surfactant to obtain porous structure by a self-assembly process.19

In the next group of samples (No: 4, 8, 12, see Table I), a part of Si-O bonds in silicon oxide network was replaced by Si-ethylene bonds.20-23 Methyltrimethoxysilane (MTMS, 98%, Fluka) and 1,2-bis(trimethoxysilyl)ethane (BTMSE, 96%, Sigma-Aldrich) were used as precursors. BTMSE precursor already contains -Si-(CH2)2-Si-groups and therefore an increase of its content leads to increase of C-C bonds content in a final polymer film (see Table I).

The films were deposited on 150 mm diameter silicon wafers (0.005 Ohm cm and 80 Ohm cm) at 2500 rpm rotating speed and cured on the hot plate at Tc = 150°C, 30 min (soft bake to remove solvents), and final curing at Tc = 400°C, 60 min.

Infrared Fourier transform spectroscopy (FTIR) was performed on a Nicolet 6700, Thermo Scientific. Porosity of low-k films were characterized by using heptane adsorption isotherms measured on ellipsometry based apparatus at 24°C. Before the measurements, the samples were annealed at 250°C to remove physically adsorbed water molecules. The films open porosity is calculated as volume of adsorbed liquid (adsorptive) from the values of refractive indexes (RI) measured during the heptane adsorption by using Lorentz-Lorenz equation:

\[
\frac{n_{eff}^2 - 1}{n_{eff}^2 + 2} = \frac{n_{ads}^2 - 1}{n_{ads}^2 + 2} + (1 - V) \frac{n_i^2 - 1}{n_i^2 + 2}
\]

where n_{eff} is measured RI of porous film when the pores partially or completely filled by adsorptive, n_{ads} is RI of liquid adsorptive, n_i is skeleton RI and V is the volume of the condensed adsorptive (open porosity).25 The porous films may have closed (isolated) and open (interconnected) pores. The relative amount of closed pores increases with the decreasing porosity. Normally, materials with V < 3% have only isolated pores.26 If the open porosity is measured from adsorption isotherms, the full porosity can be calculated from the measurement of RI value when pores are empty:

\[
V = 1 - \frac{(n_i^2 - 1)}{(n_i^2 + 2)} + \frac{(n_i^2 - 1)}{(n_i^2 + 2)}
\]

Table I. Film-forming solutions and preparation details.

| No | TEOS/MTEOS ratio | BTMSE/MTMS ratio | Solvents | Si Content (wt. %) | Water/alkoxy group ratio | Me/Si ratio | C-C/Si ratio | Brij 30 content (wt. %) |
|----|------------------|------------------|----------|-------------------|-------------------------|-----------|-------------|-----------------------|
| 4  | -                | 25/75            | THF*     | 4                 | 0.6                     | 0.6       | 0.2         | 0.2                   |
| 8  | -                | 33/67            | THF      | 4                 | 0.6                     | 0.5       | 0.25        | 0.30                  |
| 12 | -                | 47/53            | THF      | 4                 | 0.7                     | 0.36      | 0.32        | 0.30                  |
| C35| 40/60            | -                | ethanol, | 5.5               | 1.16                    | 0.6       | -           | 13.6                  |
| C45| 40/60            | -                | 2-propanol | 4.7               | 1.16                    | 0.6       | -           | 19.3                  |

*tetrahydrofuran.
where $n_0$ is the film RI before adsorption (when the adsorptive pressure $P = 0$ and $n_s = n_{eq}$). Calculation of the full porosity needs knowledge of skeleton RI ($n_s$) which normally can be estimated.\(^{25}\) In the case of OSG films, the skeleton RI is normally close to refractive index of amorphous SiO\(_2\) (1.46). Calculation of the pore radius distribution (PRD) is based on analysis of adsorption isotherms and hysteresis loops that appear due to the difference in the curvature radius of condensed liquid meniscus during the adsorption and desorption. The analysis is based on Kelvin equation that describes dependence of the relative pressure ($P/P_0$) on the meniscus curvature.\(^{27}\)

The Young’s modulus and hardness of the low-k dielectric films were measured by nanoindentation (NI). The tests were carried out using a nanoindenter (MTS Nano Indenter XP System) with a Berkovich tip, in continuous mode, to obtain the reduced modulus ($E_r$). The Oliver–Pharr method\(^{28}\) was then used to determine the elastic modulus ($E$) of various low-k dielectric films. To eliminate the effect of the Si substrate, the indentation depth was maintained at less than 20% of the film thickness.

**Results and Discussion**

The chemical composition of the films was analyzed by using FTIR. The presented FTIR spectra (Figure 2) show vibrations assigned to the $O\text{-Si-O}$ skeleton ($1250$–$970$ cm\(^{-1}\)). The terminal Si–CH\(_2\) bonds are important components of OSG low-k films and their peak is located in the range of $1280$–$1270$ cm\(^{-1}\). These groups are requested to keep low-k films hydrophobic and therefore to avoid adsorption of water molecules. The adsorbed water molecules degrade dielectric constant and electrical reliability of low-k materials.\(^{29}\) Absorption peaks from residual organic template and CH\(_2\) groups from bridging organic groups are located between $3000$ and $2750$ cm\(^{-1}\).

One can see that in the case of materials with terminal methyl groups (C35 and C45) the films porosity increases with the template (Brij 30) concentration. The principal feature of the films with alkyl bridge (4, 8, 12) is appearance of pronounced hysteresis loop in their isotherms. In the case of thin films, the adsorption branch reflects the size of internal voids while the desorption curve reflects the size of interconnecting necks. The size of interconnecting necks remains constant (Figure 5). Therefore, the presence of carbon bridge introduces certain kind of “non-uniformity” between the necks and internal voids. Such behavior sometimes is called as the effect of ink-bottle porosity.

Mechanical properties of these films are shown in the Table II and Figure 6. The elastic properties of porous materials depend on the geometrical nature of the pore space and solid phase, as well as the value of porosity. Relevant aspects of porous materials may include pore shape and size as well as the size and type of the interconnections between solid regions. These features, which generally lack precise definition, comprise the microstructure of the material.\(^{31}\) In this work we targeted to compare mechanical properties of OSG materials with different structures. For comparative purpose Figure 6 also shows mechanical properties of PECVD low-k film that was cured with broadband UV light with wavelength $> 200$ nm at $T = 400^\circ C$ (Figure 8 in the Reference 32). It is known that broadband curing with wavelength $> 200$ nm is “safe” to respect to SiCH\(_3\) groups and therefore the results can be compared with thermally cured materials.\(^{33}\) Moreover, these films were deposited by using a special technology allowing to avoid presence of so called “porogen residues” that may

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**Table II. Measured properties of studied films.**

| Sample number | Refractive Index (±0.01) | Thickness (nm, ± 5) | V(open)/V(full) (%) ± 1 | Pore radius, nm (±0.2) | Skeleton refractive index (±0.01) | Young’s Modulus (GPa ±0.2) | Hardness (GPa ±0.1) |
|---------------|--------------------------|----------------------|--------------------------|------------------------|---------------------------------|----------------------------|---------------------|
| 4             | 1.28                     | 607                  | 36/36                    | 1.61                   | 1.46                            | 6.12                       | 0.87                |
| 8             | 1.29                     | 514                  | 34/34                    | 1.61                   | 1.46                            | 6.90                       | 1.00                |
| 12            | 1.27                     | 500                  | 37/39                    | 1.61                   | 1.45                            | 7.66                       | 0.89                |
| C35           | 1.31                     | 783                  | 28/31                    | 1.10                   | 1.45                            | 6.83                       | 0.96                |
| C45           | 1.26                     | 673                  | 39/39                    | 1.56                   | 1.46                            | 4.85                       | 0.69                |
PECVD films reported in the Ref. 32 were prepared with different porosity (Figure 6a). Dependence of Young modulus on porosity is well described by empirical power-law relationship proposed by Phani and Niyogi:

$$E = E_0 \left(1 - \frac{V}{V_c}\right)^f$$

where $E$ is the effective Young’s modulus of porous material with porosity $V$; $E_0$ is Young’s modulus of solid material, $V_c$ is the porosity at which the effective Young’s modulus becomes zero and $f$ is a parameter dependent on the grain morphology and pore geometry of porous material. In fact, $f = 1$ in an ideal (hypothetical) case when a porous material made by “extraction” of certain part of solid material without destructing of the material skeleton. In all other cases, this coefficient reflects degree of skeleton disordering in comparison with non-porous prototype.

Figure 6b shows comparative data. A fragment of the data presented in Figure 6a is compared with the films deposited by spin-on method. The dashed curve (fitting C35 and C45) was plotted in assumption that they have same $E_0$ value as PECVD films. This assumption is based on similarity of chemical composition and skeleton refractive index, which reflects the density of non-porous prototype. Figure 6b shows that the spin-on deposited films with terminal methyl groups have better mechanical properties than similar films deposited by PECVD. The values of the coefficient $f$ for PECVD and spin-on films with terminal methyl groups (C35 and C45) were equal to 3.35 and 2.85, respectively. The best mechanical properties show OSG films with bridging alkyl groups. Moreover, the films deposited with higher concentration of BTMSE have higher Young’s modulus, probably, because of higher concentration of cross-linking bonds.

The samples C35 and C45 have the same carbon concentration, but were prepared by using different concentration of Brij 30 surfactant. Therefore, the surfactant defined the films porosity: the ratio of the Table III. Relative concentration of bridging and terminal carbon groups in the deposited films in comparison with the relative concentration of these groups in initial solutions. The columns 2 and 3 reflect concentration of these groups in Sol. The columns 4 and 5 give the values extracted by deconvolution of FTIR spectra and ratio of total intensity of CH$_2$ groups (both symmetric and asymmetric) to intensity of Si-X-Si (Si-O-Si and Si-CH$_2$-CH$_2$-Si).

| Sample No | Me/Si Ratio | C-C/Si Ratio | \((\Delta_{C=CH} + \Delta_{C=CH})\) | \((\Delta_{C=CH} + \Delta_{C=CH})\) |
|-----------|-------------|--------------|---------------------------------|---------------------------------|
| 4         | 0.6         | 0.2          | (±0.0002)                       | (±0.0002)                       |
| 8         | 0.5         | 0.25         | 0.0021                          | 0.0082                          |
| 12        | 0.36        | 0.32         | 0.0047                          | 0.0062                          |
| C35       | 0.6         | -            | 0.0004                          | 0.0076                          |
| C45       | -           | -            | 0.0007                          | 0.0071                          |
Figure 4. Adsorption/desorption isotherms and pore shape type of a) carbon bridged OSG films with different BTMSE content (Nos 4, 8, 12) and b) porous methyl-terminated OSG films (Nos C35, C45).

Figure 5. Pore size distribution calculated from adsorption (A) and desorption (D) isotherms. (a) the low-k films containing both terminal and bridging alkyl groups. These films have the same size of interconnecting necks (A) different than the size of internal voids (A) (b) the low-k films with only terminal CH₃ groups. No difference in pore sizes calculated from adsorption (A) and desorption (D) curves.

Surfactant concentration during the films deposition is nearly equal to the ratio of porosities in these films. The fact that Me/Si ratio in the final films is the same (Table II) and FTIR spectra are identical suggests that surfactant was completely removed during the curing and films composition is defined by TEOS/MTEOS precursor.

More complicated situation with the samples 4, 8, 12 is observed because of simultaneous change of concentration of both bridging and terminal carbon groups. Young’s modulus clearly increases with concentration of bridging carbon groups (Figure 7a). However, the increase of bridging groups concentration happens simultaneously with reduction of terminal groups. Effect of terminal groups is shown in Figure 7b. The Young’s modulus decreases with the increasing of terminal methyl groups concentration (Tables I–III). As a comparison, Figure 7b also shows Young’s modulus of sample C45 that as the same concentration of methyl groups as the sample 4. The measured E values is significantly lower (4.85 against 6.12), but C45 is also more porous. For more reasonable comparison we corrected E value using Equation 3. The corrected E value is shown as C45* and the difference with the sample #4 has become quite small. In fact, these 2 samples have the same concentration of terminal methyl groups but the matrix is cross-linked via oxygen bridge in the case of C45. In the case of sample No 4 some oxygen bridges have been replaced by -C-C- groups. The samples 8 and 12 show more significant increase of E values, because of higher concentration of bridging carbon groups and reduction of concentration of terminal groups. Therefore, these observations suggest that the carbon based bridging groups are more efficient for Young’s modulus improvement than the oxygen bridges. This fact is in agreement with the results of theoretical analysis carried out by using force-field potential and ab initio calculations based on density functional theory.36,37 It was demonstrated that the bulk modulus gradually increases from the value...
of pure silica (≈39.5 GPa) to 59.3 GPa as half of the oxygen atoms are replaced by methylene (-CH2-) bridging groups. The presented data confirm that the bridging alkyl groups improves the E values even in comparison with SiO2. However this effect is becoming remarkable at sufficiently high concentration of bridging carbon groups. It is known that E values of low-k materials can be increased by curing with 172 nm UV light, which reduces concentration of terminal methyl groups and generates a risk of the films hydrophilicity. Advantage of compositions based on replacement of terminal methyl groups by bridging carbon groups allow to achieve the same (or even stronger) effect keeping material hydrophobic which is important for various practical applications.

Conclusions

Chemical composition, pore structure and mechanical properties of porous low-k films with terminal and bridging carbon groups are studied. In BTMSE based carbon bridged low-k films, the presence of -Si-(CH2)2-Si- groups is visible from -CH2- absorption near 2900 cm\(^{-1}\). These films have non-uniform pore structure with presence of internal voids (ink-bottle like pores) while carbon terminating films have pores having cylindrical shape. All studied spin-on deposited films have better mechanical properties than PECVD films with the same composition and porosity. The alkyl bridged low-k films show better mechanical properties than the films with methyl terminating groups. Moreover, the Young’s modulus increases with BTMSE content because of higher concentration of bridging bonds.

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References

1. P. Van Der Voort, D. Esquivel, E. De Canck, F. Goethals, I. Van Driessche, and F. J. Romero-Salgueiro, Chem. Soc. Rev., 42, 3913 (2013).
2. K. Maex, M. R. Baklanov, D. Shamiryan, F. Lacopi, S. H. Brongersma, and Z. Sh. Yanovitskaya, J. Appl. Phys., 93, 8793 (2003).
3. A. Grill, J. Vac. Sci. Technol. B, 34, 020801 (2016).
4. V. Dubin, “Copper Electroplating for On-Chip Metallization”, In: M. R. Baklanov, P. Ho, and E. Zschech, (Eds.) Advanced Interconnects for ULSI Technology, Wiley & Sons, 2012, pp. 1751.
5. M. R. Baklanov, J-F. de Marnelle, D. Shamiryan, A. M. Urbanowicz, H. Shi, T. V. Rakhimova, H. Huang, and P. S. Ho, J. Appl. Phys., 113(4), 044101 (2013).
6. X. Wang, L. Liu, P. He, X-P. Qu et al. J Physics D: Applied Physics, 50, 405306 (2017).
7. Liping Zhang, J-F. de Marnelle, N. Heylen, G. Murdoch, Z. Tokei, J. Boemmels, Stefan De Gendt, and M. R. Baklanov, Appl. Phys. Lett., 107, 092901 (2015).
8. D. J. Michalak, J. M. Blackwell, J. M. Torres, A. Sengupta, L. E. Kreno, J. C. Clarke, and D. Pantuso, Journal of Materials Research, 30(22), 3363 (2015).
9. G. Dubois, R. D. Miller, and W. Volksen, Spin-on Dielectric Materials. In: M. Baklanov, M. Green, and K. Maex, (Eds.) Dielectric Films for Advanced Microelectronics., 2007 John Wiley & Sons, Ltd, page. 33.
10. W. Volksen, R. D. Miller, and G. Dubois, Chem. Rev., 110, 56 (2010).
11. S. Kim, Y. Toivola, R. F. Cook, K. Char, S-H. Chu, J-K. Lee, D. Y. Yoon, and H-W. Rhee, J. Electrochem. Soc., 15(1), F37 (2004).
12. G. Dubois et al., J. Sol-Gel. Sci. Technol., 48(1-2), 187 (2008).
13. G. Dubois et al., Adv. Mater., 19(22), 3989 (2007).
14. H. Miyoshi, H. Matsuo, Y. Oka, H. Tanaka, K. Yamada, N. Mikami, S. Takada, N. Hata, and T. Kikkawa, Jpn. J. Appl. Phys., 43(1), 498 (2004).
15. H. Miyoshi et al, Jpn. J. Appl. Phys., 44(11), 5982 (2005).
16. M. Redzheb, S. Naumov, L. Prager, M. Kristtabel, S. Armini, P. van Der Voort, and M. R. Baklanov, Appl. Phys. Lett., 108, 012902 (2016).
17. A. P. Palov, E. N. Voronina, T. V. Rakhimova, D. V. Lopaev, S. M. Zyrjanov, Y. A. Mankelevich, M. B. Krishtab, and M. R. Baklanov, J. Vac. Sci. Techn., B34, 041205 (2016).
18. H. Schmidt, “New type of non-crystalline solids between inorganic and organic materials,” J. Non-Cryst. Solids, 73, 681 (1985).
19. J. J. Brinker, C. J., MRS Bulletin, 29(9), 631 (2004).
20. B. R. Kim, J. M. Son, and M. J. Ko, Journal of Materials Science, 42, 5381 (2007).
21. B. R. Kim, J. W. Kang, K. Y. Lee, J. M. Son, and M. J. Ko, Journal of Materials Science, 42, 4591 (2007).
22. S. Kim, Y. Toivola, R. F. Cook et al. J. Electrochem. Soc., 151, F37 (2004).
23. Y. Toivola, S. Kim, R. F. Cook et al. J. Electrochem. Soc., 151, F45 (2004).
24. K. P. Mogilnikov, CSTIC 2017 – China Semiconductor Technology International Conference, March 17–18, 2017.
25. M. R. Baklanov, K. P. Mogilnikov, V. G. Polovinkin, and F. N. Dultsev, J. Vac. Sci. Technol. B, 18, 1385 (2000).
26. J. Kovačik, J. Mater. Sci. Lett., 18, 1007 (1999).