Mechanical Stability of Porous Low-k Dielectrics
K. Vanstreels, C. Wu, and M. R. Baklanov
IMEC, Kapeldreef 75, B-3001 Leuven, Belgium

This paper reviews the mechanical and fracture properties of porous ultralow-k dielectrics with the focus on chip package interaction related issues. It is shown that the mechanical and fracture properties of porous ultralow-k dielectric films are closely linked with porosity, pore morphology, network structure and deposition technology, while their fracture properties are also sensitive to reactive species in the environment. The survivability of low-k dielectrics upon integration, package assembly and subsequent reliability tests is therefore a combination of their mechanical stability, fracture properties, the specific mechanical or thermo-mechanical load and environmental effects.

© The Author(s) 2014. Published by ECS. This is an open access article distributed under the terms of the Creative Commons Attribution 4.0 License (CC BY, http://creativecommons.org/licenses/by/4.0/), which permits unrestricted reuse of the work in any medium, provided the original work is properly cited. [DOI: 10.1149/2.0071501jss] All rights reserved.

Manuscript submitted September 5, 2014; revised manuscript received October 13, 2014. Published October 21, 2014. This paper is part of the JSS Focus Issue on Advanced Interconnects: Materials, Processing, and Reliability.

For the past three decades, the semiconductor industry has been improving the performance of microelectronic devices and the functionality of advanced integrated circuits through the continuous scaling of integrated circuits and the maximization of transistor density. Consequently, this encourages the introduction of new materials, processes, chip designs, and packaging strategies into micro-/nano-electronic products. Among these material innovations are insulating materials with a dielectric constant (\(k\)) less than that of SiO\(_2\), so called low-k dielectrics. During the last 15 years, various materials and methods have been developed for fabricating low-k dielectric films, among which the plasma enhanced chemical vapor deposition (PECVD) technology of porous organosilicate glasses (OSG) is the most popular due to its better compatibility with the technological needs. The reduction in \(k\) for low-k dielectrics is being pursued through the introduction of controlled levels of porosity. However, finding a good low-k material has proven to be much more challenging than first expected due to a number of integration and reliability issues. Besides the need of being compatible with the different lithography, etching, stripping and cleaning processes that are used in state-of-the-art integration schemes, they must also have sufficient mechanical strength to withstand the high shear stresses as well as harsh chemical environments that are involved during the chemical mechanical polishing process without cohesive or adhesive failure occurring. On top of that, since low-k dielectrics exhibit intrinsic tensile stresses and have increased coefficients of thermal expansion compared to SiO\(_2\), thin film cracking and adherence are serious thermal-mechanical reliability issues for low-k dielectric materials (Figure 1). Thermo-mechanical deformation of the package during package assembly and subsequent reliability tests can induce large local stresses that can initiate and propagate cohesive and/or adhesive cracks in different back-end of line (BEOL) layers. Therefore, a careful characterization of the mechanical integrity of potential new low-k candidates is required to integrate these new materials and Cu interconnects and to assure reliability during chip packaging and under field conditions. In the microelectronics industry, the elastic modulus, fracture cohesive strength and interfacial adhesion of thin films are commonly used to assess the mechanical behavior, fracture behavior and process compatibility of interlayer dielectric materials. However, these parameters do not sufficiently represent the failure behavior and structure stability of multilevel interconnects. Therefore, the semiconductor industry needs reliable, reproducible testing methods and strategies to monitor these parameters, not only for low-k dielectric films, but also for complete interconnect stacks. This paper reviews the mechanical stability of porous ultralow-k dielectrics with the focus on chip package interaction related issues. To this end, the impact of porosity, pore morphology, network structure, UV radiation and reactive species in the environment on the mechanical stability of low-k dielectrics will be discussed.

Impact of Porosity and Pore Morphology on Mechanical Stability

The overall mechanical properties of porous low-k dielectrics are governed by the properties of the solid matrix and the pore morphology. Therefore, the successful integration of these materials requires a thorough understanding of how their mechanical stability varies with porosity and pore microstructure. Although much advancement has been made in this field over the past decades, there is still relatively little fundamental understanding of the impact of porosity and pore morphology on elastic properties and fracture behavior of porous low-k dielectrics. Extensive theoretical studies have been directed to modeling porosity effects on the elastic properties of materials, including micromechanics models,12–14 semi-empirical formulas derived from the minimum solid area model,15,16 or computationally solve the equations of elasticity for digital models of microstructure,5,17 among others. Micromechanics models can capture the effect of porosity and pore shape, but are generally limited to structures with only idealized pore geometries. On the other hand, the microstructure that corresponds to the minimum solid area model is not clearly defined, which limits the predictive and interpretative power of this method. As demonstrated by Roberts and co-authors, finite element simulations represent a very promising approach toward a more thorough understanding of the impact of porosity and pore geometry. These simulations allow various geometrical factors to be tuned separately to clarify their roles in the structure-property relationship. From the experimental point of view, investigating the impact of porosity and pore morphology is often complicated by the difficulty to separately control the porosity, pore structure and matrix properties during fabrication. Moreover, there is little fundamental understanding on how the measured elastic modulus value is influenced by the stiffness of the matrix, porosity, pore morphology, and specific technique related effects. Recently, Vanstreels and co-authors demonstrated the intrinsic effect of porosity on the elastic properties and fracture behavior of ultralow-k dielectrics by using a special curing process that allows a separate control of porosity and matrix properties (Figure 2). The observed change in elastic modulus suggest that the microstructure evolves with increasing porosity and evolves from a structure containing mesopores interconnected with microporous channels at porosities below 25%, to one with heavily interconnected mesopores at higher porosity levels, as confirmed by porosimetry and ellipsometry (PALS, ellipsometric porosimetry, and diffusion experiments. Moreover, it was found that the fracture energy decreases linearly with increased porosity. This was consistent...
Figure 1. Strain energy release rate as a function of the film density ($\rho$) and the corresponding dielectric constant ($k$) of organosilicate thin film glasses. Adhesive and cohesive fracture energies were measured in air and water using respectively 4-point bending and double cantilever beam specimens. The figure was taken from 10.

Figure 2. (a) relative Young’s modulus values as a function of porosity, compared with finite element simulations of various pore microstructures. The open dots are experimental results using the same pore template structure but different UV treatment (BB UV: broad band UV lamp, NB UV: narrow band UV lamp); the solid lines are results from simulation models (overlapping spherical pores, overlapping ellipsoidal pores or overlapping solid sphere) published in the literature; (b) effect of porosity on the relative fracture energy of PECVD SiCOH films (figure was taken from 26); The open circular dots are experimental results using 4-point bending while the open triangular dots are experimental results using the double cantilever beam technique.

Figure 3. experimental values of the function, $f(P)$, compared with finite element simulations of various pore microstructures and theoretical upper bound. The overlapping spherical pore model and solid spheres model are from 28.

with a planar through-pore fracture mechanism, in which the solid area that must be fractured changes linearly with film porosity and effects associated with pore shape, size, or distribution are screened out for a random pore microstructure. In a more recent work, various destructive and nondestructive experimental techniques, which are based on different physical principles for the calculation of the elastic properties, were compared. 27 It was shown that the measured elastic modulus clearly depends on specific technique related effects, which also appeared to be sensitive to the specific arrangement and morphology of the pores in the film. Li and co-authors have demonstrated an experimental procedure to separate out the effects caused by the matrix at different levels of porosity.28 Such a separation relies on the characterization of the density of various molecular bonds in the low-$k$ dielectric film, and the experimentally established correlation between the various bond densities (Si-O-Si, Si-CH$_2$-Si, Si-CH$_3$, among others) determined from the fourier transform infra-red (FTIR) absorption spectra and mechanical and fracture properties for similar but dense low-$k$ dielectrics. It was found that for dense OSG dielectrics, the elastic modulus correlates strongly with the densities of networking bonds (Si-O-Si and Si-CH$_2$-Si) of the materials over a broad range of process parameters.29 The elastic modulus of the matrix, $E_{\text{matrix}}$, of a porous film can be determined from the knowledge of the network bond densities in the matrix. The intrinsic porosity effect on the elastic modulus can be expressed using a dimensionless function as $f(P) = E_{\text{film}}(P)/E_{\text{matrix}}(P)$, where $P$ denotes film porosity (Figure 3). In a similar way, the intrinsic porosity effect on the cohesive fracture energy can be described by the dimensionless function $g(P)$. They found that in case of OSG low-$k$ dielectrics, the fracture energy decreases linearly with increasing porosity for porosity values up to 45%, which is in agreement with Vanstreels and co-authors.26 Alternatively, Guyer and co-authors reported a non-linear porosity dependency of fracture energy for methyl-silsesquioxane (MSSQ)-based OSG coatings.9 This was attributed to several factors including the non-uniformity of the MSSQ density through the coating thickness and contributions from frictional contact between the crack faces.9 Although increasing the total porosity of low-$k$ dielectrics leads to substantial reductions in film stiffness and fracture resistance, there is still little experimental evidence whether or not the stiffness and fracture properties of porous low-$k$ dielectrics can be substantially changed through variations in pore architecture. This is because in realistic porous microstructures, the features of the pore morphology like pore size and shape, distribution and interconnectivity have no obvious definition. Simplified microstructural models are commonly used to experimentally extract those features, however, this may lead to inconclusive trends in the measured modulus-pore size relationship. Moreover, the dependency on pore structure is often overshadowed by the simultaneous scaling of pore size with porosity. Theoretical studies offer a good solution...
to avoid any interlinkage between pore size, porosity and matrix stiffness and hence studying the intrinsic effect of pore morphology on the elastic properties of materials. Roberts and co-authors used finite element simulations to separately tune various geometrical features of different arbitrary pore structures.25,23 They concluded from their simulations that for a given porosity, the overlapping solid sphere model results in a relatively weak structure due to small solid contacts between the solid spheres and the highly interconnected porosity. Overlapping spherical pores provide high (near optimal) stiffness, while overlapping ellipsoidal pores tend to weaken a structure more than spherical pores. In a similar way, W.J. Lee and co-authors modeled and simulated different idealized porous microstructures using a random sequential adsorption algorithm and a three dimensional finite element method (Figure 4) and found that the elastic properties can indeed be substantially influenced by their pore morphologies.30 Recently, Krishatab and co-authors experimentally showed that the cohesive strength of liquid phase self-assembly (LPSA) low-k dielectrics could be improved by tailoring its pore structure.31 The main advantage of LPSA over PECVD low-k dielectrics is the ability to suppress the coupling between pore size and porosity, and thereby separating their contribution to the film strengthening. Two major distinctions between the dielectric films were found, i.e. type of bonding within the organosilica matrix and size of pores. A better cohesive strength was found for dielectric films with smaller pores despite a remarkably lower interconnectivity of its organosilica backbone, as compared to films with larger pores, thereby emphasizing the prevailing role of the structural aspects in the tested films.

Impact of Network Structure and UV Radiation on Mechanical Stability

Lowering the $k$-value of organic-inorganic hybrid low-$k$ dielectrics is achieved by introducing organic terminal groups into the silica-like backbone structure, which leads to a decrease in the film density and lowering the bond polarizability, both beneficial for the dielectric performance. However, from a mechanical perspective, these terminal groups disrupt the covalent molecular network and degrade the mechanical integrity of the materials. Experiments have demonstrated that the elastic modulus and fracture toughness of PECVD carbon-doped oxide films.36 H. Lin and co-authors used a combined computational and synthetic approach to establish fundamental scaling laws for the elastic response and fracture properties of hybrid glasses.35 The specific glasses that were investigated are ethane-bridged oxycarbosilane (Et-OCS) and methyalted ethane-bridged oxycarbosilane (Et-OCS(Me)), as illustrated on Figure 5. They show that Si-X-Si connectivity provides the best measure of network connectivity because it provides an absolute measure that is independent of the precursor structure. The scaling of bulk modulus with Si-X-Si connectivity could be described by a simple mean-field model that was based on rigidity percolation occurring at Si-X-Si connectivity (Figure 5b). The fracture energy of brittle hybrid glasses scales in a similar way as the number of bonds broken during fracture (Figure 5c). Therefore, while the elastic properties are very sensitive to changes in connectivity induced during UV curing, the cohesive crack can meander through the network on a path toward regions of low connectivity and low density within the glass network and is therefore insensitive to UV cure. This was experimentally confirmed by Gage and co-authors, who showed that UV cure exposure results in a higher degree of film connectivity in the material and significantly enhances the interfacial adhesion strength, but barely influences the fracture toughness of PECVD carbon-doped oxide films.36 H. Lin and co-authors28 and Y. Lin and co-authors29 demonstrated that the fracture energy of OSG materials, measured under mode-I condition using double-cantilever beam technique, scales linearly with the total number density of the two main networking bonds in an OSG material (Figure 6).28 This is because fracture of these materials is a bond breaking process and the energy dissipated during fracture scales with the number of broken networking bonds and with the energy of the bonds. In literature, molecular dynamics (MD) methods have been used to develop atomistic models of low-$k$ materials and the mechanical properties as well as the fracture behavior were analyzed.37-42 H. Li and co-authors used molecular dynamics simulations to investigate the fundamental structure–property relationships of OSG low-$k$ with a primary focus on the mass density and elastic properties.43 To examine the role of the bridging units and terminal groups, two idealized classes of molecular models were considered that are built out of the backbone structure of amorphous silica. The first class of models (type-I OSG) consists of a Si–O–Si and Si–CH$_3$–Si network without...
terminal groups. These models are created by replacing a fraction of the bridging oxygen atoms in the silica backbone structure with methylene groups. The second class of models (type-II OSG) have only Si–O–Si crosslinking bonds in the network, but terminal groups (–OH or –CH3) are allowed. These models were created by breaking the Si–O–Si bonds and replacing the oxygen atom with two terminal groups (either –OH or –CH3). The results show that the incorporation of –CH2– as bridging unit into the silica network considerably affects the elastic properties of the resultant OSG low-k. The greater bending rigidity of Si–C–Si relative to Si–O–Si was found to be responsible for the increased bulk modulus of type-I OSG with increasing –CH2– population (Figure 7a). Type-II OSG materials contain a number of terminal groups (–OH, –CH3) interrupting the network structure of amorphous silica. The degree of interruption can be described by the network connectivity number (\(\langle r \rangle\)), defined as the average coordination number per network forming atom.44 The MD simulations show that the bulk modulus and the shear modulus of type-II OSG are reduced significantly with increasing fraction of terminal groups or decreasing (\(\langle r \rangle\)) (Figure 7b).
Impact Reactive Species in the Environment on Mechanical Stability

Besides the intrinsic fracture properties of low-κ dielectrics, time-dependent fracture behavior is another concern for chip fabrication and long-term reliability. Moisture-assisted crack growth is a time-dependent mode of failure that occurs at loads well below those required for critical fracture. Many materials are known to be susceptible to moisture-assisted cracking including ceramics, glasses, polymers, among others. It is well known that the fracture behavior of OSG low-κ can be significantly degraded due to the ingress of water and other reactive species. This is because the strained Si-O bonds in their backbone structures at crack tip can react chemically with water molecules to assist the bond rupture process. More recently, the impact of water diffusion on the fracture behavior of nanoporous dielectric film stacks were studied for various OSG materials. Li and Vlassak studied various nanoporous OSG dielectrics following the “soak-then-test” approach. They demonstrated that the adhesion of the OSG/SiCN interface decreases over time (Figure 8), while the cohesive fracture energy of the nanoporous OSG layer remains largely constant. This was explained by the tendency of water molecules to diffuse along the OSG/SiCN interface rather than through the bulk of the OSG, and was attributed to the less hydrophobic character of the interface. Moreover, degradation of the cohesive fracture energy was observed when the OSG film stack was exposed to aqueous solutions with organic additives that enhance wetting of the OSG. Since various processing steps may cause loss of hydrophobicity to certain extent it should therefore be pointed out that common organic additives such as isopropyl alcohol or acetic acid to water could drastically change the diffusion behavior. The effect of the solution chemistry on the fracture of nanoporous glasses exposed to aqueous solutions was investigated. Crack growth rates were inhibited by acidic solutions and enhanced by basic solutions (Figure 9). They concluded that the successful implementation of nanoporous glasses depends on the selection of appropriate chemistries that inhibit crack-growth reaction kinetics.

Low-κ Dielectrics for Future Technology Nodes

Integration is the deciding factor driving the selection of the most promising materials for a given technology node. During the past decades, PECVD technology has been the method of choice for depositing silicon dioxide (SiO₂), fluorine-doped oxides (F-SiO₂), carbon-doped oxides (OSG) and porous carbon-doped oxides (p-OSG). Until now, this technology is still the most popular choice because of the good compatibility with existing ultra large scale integration (ULSI) technology and the possibility to deposit relatively good quality films with dielectric constant in the range between 3.5 and 2.4. In order to achieve dielectric constants of 2.4 and beyond, the introduction of additional artificial porosity into the low-κ dielectrics (p-OSG) becomes crucially important. However, the addition of porosity exacerbated already known integration issues due to the difficulty of combining low-κ values with other manufacturing requirements such as sufficient mechanical strength, good chemical stability, reduced pore size, and hydrophobicity. At this moment, the integration of PECVD based OSG films with dielectric constants below 2.4 are still under research. It seems however that there is a limit to values around 1.9–2.0 due to pore collapsing at higher porogen loadings. As a result, the list of potentially viable low-κ candidates has narrowed substantially over the years. For advanced and future technologies nodes below 10 nm, the PECVD technology will no longer be able to satisfy all the integration requirements due to the uncontrollable process of pore formation and the porogen residues that are formed during the UV curing process. As an alternative, spin-on self-assembling technology is considered as more promising for future generations of ultralow-κ dielectrics due to an easier removal of template and a better control over pore-organization compared to PECVD. Potential improvements of the mechanical stability could be obtained by tailoring the pore structure and/or by skeleton engineering, i.e. when the material is ordered or has high matrix connectivity. Potential low-κ candidates for future generations include periodic mesoporous organosilicas (PMO), metal-organic frameworks (MOF), among others.

PMO are ordered mesoporous silica materials of which at least one siloxane bond in the matrix is replaced by an organic bridge. PMO are usually amorphous, but their pores and pore walls are arranged in a structural order. They are similar in chemical composition compared to OSG with the advantage that low polarizable organic groups are integrated in the silica matrix. Especially, PMO with the simplest organic bridges such as ethane and methane are extremely interesting for low-κ applications because they have the lowest polarizable organic bridge. The great advantage of PMO over pure silica ordered mesoporous materials is the presence of uniformly distributed organic
functional groups in the walls and the fact that they are hydrothermally and mechanically more stable. By varying synthesis conditions, precursors and surfactants, it is possible to adjust the pore size, porosity, surface area, ordering, wall thickness and morphology of the PMO. These characteristics make them promising candidates for different applications including microelectronics. Typically PMO materials are synthesized by hydrolysis and condensation of bridged organosilanes in the presence of a surfactant which acts as structure directing agent. After an aging period, the surfactant can be removed by an extraction procedure or by a thermal treatment under an inert atmosphere, leaving behind a porous PMO material. Compared to PECVD materials, PMO materials enable a better control of the pore and skeleton structure and easier removal of the template. Moreover, the porous structures of these materials are formed without the use of sacrificial porogen and without the risk of residue formation during the UV curing. At this moment, several groups reported the papers related to PMO and low-k, in which the dielectric, mechanical and hydrophobic properties were studied. Although there is some variation in the reported $k$-values of PMO films, the microelectronic industry recognized that PMOs are one of the most promising ultralow-$k$ materials to integrate in actual devices. However, still a lot of future research is required, including plasma-damage effects, adhesion, electrical characteristics and reliability.

Metal-organic frameworks have received much scientific and technological interest because they possess a well-defined nanoscale pore size, crystallinity and extremely low dielectric constant, i.e. a combination which is barely accessible for traditional low-$k$ dielectrics. MOF are nanoporous, crystalline materials that consist of metal ions that are connected by organic linkers. By varying the metal ion or linker, various sizes and shapes of channels or cavities can be obtained. The interest in these materials first appeared after the theoretical paper by K. Zagorodny and co-authors, who calculated the dielectric constant of a series of MOF based on $\text{Zn}_n\text{O}(\text{CO}_2)_{n}$ units connected by different organic linkers and found that all the investigated MOF have $k$-values below 2.0 with good mechanical properties. However, recent experimental evaluation by S. Esilva and co-authors (ZIF-8 films) and M. Usman and co-authors (Sr-based MOF films) suggest that the theoretically estimated dielectric constants of these materials might be underestimated because of the ionic contribution that was omitted in the work of Zagorodny and co-authors. Also, the presence of metal ions in the low-$k$ material may also raise questions about the electrical reliability (leakage and breakdown) of these materials. Furthermore, appropriate deposition methods are required to provide high quality thin dense homogeneous films with minimal surface roughness and good bulk crystallinity. From this respect, surface-anchored metal-organic framework (SurMOF) films are one step closer to satisfy the integration conditions. To date, HKUST-1 (also known as Cu-BTC) is by far one of the most investigated SurMOF materials due to the relative simple preparation and the presence of open metal sites which makes it advantageous for application in catalysis and gas storage. Recently, it was demonstrated that these SurMOF materials have sufficiently good mechanical properties, with Young’s modulus close to 10 GPa, which is in very good agreement with theoretical predictions for these materials. Some concerns are related to the relatively large spread that is found in the mechanical properties across these films, surface roughness and the intrinsic poor hydrophobicity of the HKUST-1 matrix material. Even a small amount of water within the material can eliminate any advantage brought by low density and predicted low electronic polarizability of MOF, as was recently reported by Krishub. Finally, the low chemical stability of MOF materials makes it difficult to integrate them in damascene technology. Nevertheless, these materials may still find an application in subtractive technology that is explored in sub-7 nm technology nodes. Subtractive technology is an alternative Cu patterning process to that of conventional damascene technology, which includes Cu etch and low-$k$ replacement activities. Other potential candidates for future generation low-$k$ dielectrics include hyper cross-linked polymer (HCP) networks, surface-anchored covalent organic frameworks (SurCOF) materials, MOF/fluorocarbon composites, Al and Zn based MOF using solvent free synthesis and polyimide-polyoxometalate (PI-POM) nanocomposites, among others.

Conclusions
This paper reviews the mechanical and fracture properties of porous ultralow-$k$ dielectrics and provides an overview of the most commonly used experimental techniques to characterize the corresponding properties on thin films and multilevel interconnect stacks. It was shown that the mechanical and fracture properties of porous ultralow-$k$ dielectrics are intimately linked with porosity, network structure and deposition technologies, while fracture properties of low-$k$ are also sensitive to reactive species in the environment. Therefore, the low-$k$ survivability upon integration is not a simple factor of mechanical properties and fracture properties, but it is rather a combination of all these properties, together with a relatively high mechanical or thermo-mechanical load and environmental effects that will determine its future. For future generations of ultralow-$k$ dielectrics, potential improvements of the mechanical stability could be obtained by tailoring the pore structure and/or by skeleton engineering using self-assembly technology. The main potential low-$k$ candidates for future generations include PMO and MOF, among others.

References
1. K. Maex, M. R. Baklanov, D. Shamiryan, F. Iacopi, S. H. Brongersma, and Z. S. Yanovitskaya, J. Appl. Phys. 93(11), 8793 (2003).
2. D. D. Burkey and K. K. Gleason, J. Appl. Phys. 93, 5143 (2003).
3. H. Li, Y. Lin, T. Y. Tsui, and J. J. Vlassak, J. Mater. Res. 24, 107 (2009).
4. M. Baklanov, M. Green, and K. Maex, Dielectric films for advanced microelectronics, Chapter 1, Wiley, West Sussex (2002).
5. R. Sreenivasan and K. K. Gleason, Chemical Vapor Deposition 15, 77 (2009).
6. E. P. Guyer and R. H. Dauskardt, IEEE 2003 International Interconnect Technology Conference, 5-7 June 2004, Burlingame, CA, USA, 2003, 89-91 (2003).
7. E. P. Guyer and R. H. Dauskardt, IEEE 2004 International Interconnect Technology Conference, 7-9 June 2004, Burlingame, CA, USA, 2004, 236-238 (2004).
8. E. P. Guyer and R. H. Dauskardt, J. Mater. Res. 20(3), 685 (2005).
9. E. P. Guyer, M. Patz, and R. H. Dauskardt, J. Mater. Res. 21(4), 862 (2006).
10. Tack-Soo Kim and Reinhold H. Dauskardt, IEEE transactions on device and materials reliability 9(4), 509 (2009).
11. G. Wang, P. S. Ho, and S. Groothuis, Microelectronics Reliability 45, 1079 (2005).
12. Z. Hashin, J. Appl. Mech., 50, 481 (1983).
13. J. Aboudi, Mechanics of composite materials: a unified micromechanical approach, Elsevier, Amsterdam (1991).
14. R. M. Christensen, Mechanics of composite materials, Wiley, New York (1979).
15. R. W. Rice, J. Mater. Sci. 31(1), 102 (1996).
16. R. W. Rice, J. Mech. Sci., 31, 1509 (1996).
17. E. J. Garboczi and A. R. Day, J. Mech. Phys. Solids, 43, 1349 (1995).
18. J. Poutet, D. Manzoni, F. Hage-chedeg, C. G. Jacquim, M. J. Bouteca, J. F. Thovert, and P. M. Adler, J. Mech. Phys. Solids, 44, 1587 (1996).
19. R. Roscoe, British J. Appl. Phys., 3(1), 267 (1952).
20. R. M. Christensen and K. H. Lo, Journal of the Mechanics and Physics of Solids 27(4), 315 (1979).
21. Y. Benveniste, “A New Approach to the Application of Mori-Tanaka Theory in Composite Materials”, Mechanics of Materials 6(2), 147 (1987).
22. A. P. Roberts and E. J. Garboczi, Journal of the American Ceramic Society 83(12), 3041 (2000).
23. A. P. Roberts and E. J. Garboczi, Proceedings of the Royal Society of London Series a-Mathematical and Physical Sciences 458(2081), 1033 (2002).
24. A. Grill and V. Patel, Appl. Phys. Lett. 79, 803 (2001).
25. A. Urbanowicz, K. Vanstreels, D. Shamiyani, S. De Gendt, and M. R. Baklanov, Electrochem. Solid-State Lett., 12(8), H292 (2009).
26. K. Vanstreels, C. Wu, P. Verdonck, and M. R. Baklanov, Appl. Phys. Lett. 101, 123109 (2012).
27. K. Vanstreels, C. Wu, M. Gonzalez, D. Schneider, D. Gidley, P. Verdonck, and M. R. Baklanov, Langmuir 29, 12025 (2013).
28. H. Li, Y. J. Vlassak, T. Y. Tsui, and Y. J. Lin, J. Mater. Res. 24, 107 (2009).
29. Youbo Lin, Yong Xiang, Ting Y. Tsui, and Joost J. Vlassak, Acta Mater 56(17), 4932 (2008).
30. W. J. Lee, Y. J. Cho, H. S. Lee, I. M. Park, and Y. H. Park, Procedia Engineering 10, 2459 (2011).
31. M. Krishub, K. Vanstreels, T. Savage, K. Matsunaga, S. De Gendt, and M. R. Baklanov, presented at the 23rd Materials For Advanced Metallization Conference (MAM 2013), submitted to Microelectronic Engineering.
32. A. Urbanowicz, K. Vanstreels, P. Verdonck, E. Van Besien, T. Christos, D. Shamiryan, S. De Gendt, and M. R. Baklanov, J. Vac. Sci. Technol. B 29(5), 033201 (2011).
