Microstructure controls in Gadolinium Zirconate/YSZ double layers and their properties

Kee Sung LEE,1 Dae Hyung LEE and Tae Woo KIM

School of Mechanical Systems Engineering, Kookmin University, Seoul 136–702, Korea

In this study, a double-layer thermal barrier coating was designed and deposited by an electron beam physical vapor deposition technique. The microstructure was controlled by using different target ingots during the electron beam physical vapor deposition process. The double layer coating consists of Gd2Zr2O7 for the top layer and YSZ for the bottom layer. The microstructure, hardness, and thermal shock properties of the double layer deposited using a single-ingot process were compared with the results obtained with the two-ingot process. The single-ingot process involves forming the Gd2Zr2O7/YSZ double layer using a single YSZ/Gd2Zr2O7 double-layered ingot. The two-ingot process involves coating with each of the YSZ and Gd2Zr2O7 ingots. While the XRD results indicated that the same pyrochlore structures are formed in those ingots, differences in the microstructure and composition were observed by FE-SEM and EDS analyses. These differences influence the hardness and thermal shock resistance of the coating. The results indicate that the finer singular columns produced by the two-ingot process do not improve the hardness, but delay the delamination of the top layer during thermal shock tests due to residual compressive stress and strain tolerance.

©2014 The Ceramic Society of Japan. All rights reserved.

Key-words : Thermal barrier coatings, Gadolinium zirconate, EB-PVD, Hardness, Thermal shock

1. Introduction

A thermal barrier coating (TBC) is used in high-temperature environments to protect sub-materials and to enhance the efficiency of electrical energy production in advanced turbine systems. Durability has recently become an important issue regarding TBCs because of the necessity for increased working temperatures to achieve higher levels of efficiency.1–4 Yttrium-stabilized zirconia (YSZ) is known for its low thermal conductivity, high mechanical strength, and similar thermal expansion to metal alloy substrates. In particular, the 7 to 8-wt % YSZ has a suitable composition for achieving enhanced TBC durability.5 However, the phase transformation and sintering of the YSZ composition lead to delamination of the coating at extremely high temperatures. Consequently, the durability must be improved to enable the coating to endure high-temperature environments. The unwanted phase transition and sintering during the operation of the gas turbine system ultimately lead to a deterioration in the lifetime of the coating layer.6–10

Microstructure control (grain size) is a material design method for increasing a TBC’s durability. Over several decades, studies have shown that microstructures such as grains, grain boundaries, and pores are crucial factors for determining the mechanical and thermal properties of a TBC.11–14 For example, the mechanical and thermal properties of a TBC deposited by air plasma spray (APS) are strongly dependent on the porosity and microstructure of the TBC.12–14 Some studies have shown that a TBC deposited by electron beam physical vapor deposition (EB-PVD) exhibits better durability in comparison to a TBC deposited by APS.15–17 Those TBCs deposited by EB-PVD coating and featuring singular columns and coarse grain sizes exhibit better mechanical properties.18–20 Another method for increasing the TBC durability involves selecting a new material and then redesigning the double layer according to the new selection.19–31 The use of materials with a low thermal conductivity for the TBC increases the TBC durability. This increased durability is attained because of the application of a more thermally stable layer, a property inherent to materials with low conductivity. Pyrochlore structures such as Gd2Zr2O7, for example, have exhibited resistance to phase transformations and have also shown superior thermal properties in high-temperature environments. Such structures maintain phase stability at temperatures up to 2300°C.20–23 However, the mechanical properties of Gd2Zr2O7 are inferior to those of YSZ; additionally, the thermal mismatch with the sub-material is greater.19 Although YSZ materials change their phase over long periods of operation at high temperatures, manufacturers of gas turbines have found it difficult to replace YSZ because of its superior mechanical and thermal properties. Among these properties, the thermal expansion coefficient of YSZ is the closest to that of bond coats, which are candidate ceramics for thermal barrier coatings. Therefore, many studies have concentrated on double layer designs with YSZ for the top coat.20–31 Some studies have shown that a double-layer TBC exhibits a higher level of durability and reliability relative to a single YSZ coat.27–31 However, the creation of an optimal bilayer design raises concerns related to the bilayer properties and the lifetime of the coating layer.

For producing a TBC, various processes have been studied and applied. In the gas turbine industry, for example, APS, EB-PVD, solution precursor plasma spray (SPPS) and high-velocity oxygen flow (HVOF) have been applied to the production of TBCs. In particular, the EB-PVD method has recently garnered considerable attention due to the columnar structure that it produces.15–17 The nano-sized gap between columns allows the columns to expand or shrink without touching neighboring columns, indicating that the columnar structure is highly tolerant to strain under constant stress. These columns reduce mismatches, which are caused by differences between the coefficients of thermal expansion (CTE) of the metal base substrate and the ceramic
coating layer. EB-PVD coatings are able to withstand stresses originating from thermal shock during gas turbine operation and, ultimately, this advantage enhances the lifetime of the coating layer.

In this study, we designed a double-layered TBC to improve the durability to thermal shock and mechanical disturbance. In this design, the top layer was coated with Gd$_2$Zr$_2$O$_7$. The bottom layer was coated with YSZ, which has a similar coefficient of thermal expansion to the metal alloy based bond coat (NiCoCrAlY) or Ni-based superalloy substrate. We controlled the microstructure by using different ingots during the EB-PVD coating processes. The first microstructure was coated using a single-ingot method, which contained sections of both the YSZ and Gd$_2$Zr$_2$O$_7$ materials. The second substrate was coated by a two-ingot method in which each material was applied separately.

2. Experimental procedures

2.1 Materials and EB-PVD coatings

The top and bottom layers of a double-layered TBC are composed of two or three materials: Y$_2$O$_3$ (high purity, Hermann C. Starck, Germany), Gd$_2$O$_3$ (Terio, Tgdio-3N, Japan) and ZrO$_2$ (Tosoh, Tz-25, Japan). The powders are mixed at a specific ratio, as listed in Table 1, to synthesize the 4.56-mol% YSZ-doped Gd$_2$Zr$_2$O$_7$ top-coat material. We added Y$_2$O$_3$ to improve the mechanical properties of the Gd$_2$Zr$_2$O$_7$ according to the results of our previous study. The material for the bottom coat was prepared using a commercial YSZ powder (Terio, TZY-4.5, Japan).

Two powder batches, 4.56-mol% YSZ-doped Gd$_2$Zr$_2$O$_7$ and 8-wt% YSZ, were each mixed in isopropyl alcohol for 24 h in a planetary ball-mill using zirconia balls in a polypropylene container to form a slurry. After drying, the softly agglomerated powder was crushed and then sieved through a 60-mesh screen. Two types of target ingots were prepared for the EB-PVD: a single ingot and two ingots. For the single ingot, 12.5 g of YSZ powder was stacked on 12.5 g of the Y$_2$O$_3$-Gd$_2$Zr$_2$O$_7$ powder mixture in a stainless-steel mold to form a disk with a diameter of 20 mm and then compressed to fabricate the single ingot. For the two ingots, the 12.5-g batches of YSZ powder and Y$_2$O$_3$-Gd$_2$Zr$_2$O$_7$ powder mixture were separately pressed without stacking. Each powder composition was uniaxially pressed at 50 MPa and then sintered at a specified temperature to attain a porosity of around 30%.

The TBC layer was applied to an alumina substrate while taking into consideration the main composition of thermally grown oxide (TGO). The TGO was produced during the high temperature operation of TBC. The substrate was deposited from the fabricated target ingot material, either a single ingot or two ingots, in a vacuum using a 10-kW EB-PVD apparatus. The target ingot material was placed in a crucible at the bottom of the apparatus, and the substrate was positioned towards the top of the apparatus. The deposition conditions were controlled such that the vacuum pressure was less than 10$^{-4}$ Torr, at a constant substrate temperature of 900°C, while the E-beam voltage was 7.5 kV. With the dual-section single ingot, the deposition took approximately 60 min. When the coating was performed using the two separate ingots, the YSZ bottom layer was deposited first and then, after rotating the crucible, the second Y$_2$O$_3$-Gd$_2$Zr$_2$O$_7$ top layer was deposited. The deposition time was approximately 30 min for each process.

2.2 Analysis and evaluation of EB-PVD coatings

The surface of the EB-PVD coatings deposited from the single ingot and two ingots was characterized by using X-ray diffraction (XRD). The relatively dense samples sintered at 1600°C were also characterized using XRD. The phase constitution of the EB-PVD coatings deposited from both the single ingot and two ingots was analyzed using energy-dispersive X-ray spectroscopy (EDS). X-ray mapping was also conducted to detect the compositions at the boundaries between the two coated layers. The amounts of specific elements gadolinium (Gd) and zirconium (Zr) could be clearly detected. This is because Gd is unique to the top-coat material of a double-layer coating, and the amount of the bottom-coat Zr was almost double the quantity of the top-coat Zr. The relative amounts of Gd and Zr content were calculated from EDS analysis through cross-sections of the EB-PVD coatings, from the top surface to a depth of 10 μm.

Specimen cross-sections normal to the surface were fractured to enable microstructure characterization. The as-received surface and cross-sections of the EB-PVD coatings were observed by field emission scanning electron microscopy (FE-SEM, JEOL, JSM 7401F, Japan). To compare the surface hardness of the TBC layers prepared from the two different target ingots, a micro hardness testing machine (Mitutoyo, HM-114, Japan) was used. Vickers indentations were carried out on the surface of each specimen. The top surfaces of the TBCs were ground and polished with a 1-μm diamond paste as a means of attaining similar coating thicknesses. The coating surfaces were then indented by a range of loads: 0.001, 0.002, 0.003, 0.005, 0.01 and 0.025 kgf.

The TBC specimens prepared from the two different ingots were thermally shocked from 1100°C into room temperature. The initial heating rate was 5°C/min to 1100°C. Once the temperature reached 1100°C, the samples were thermally shocked in air or water and held there for 10 min. The specimens were then again heated to 1100°C in the furnace within 1 min and held there for 10 min. A fan was directed at the samples when being cooled in air. Some specimens were dropped into distilled water at 25°C for water cooling. These cycles were repeated up to a maximum of 175 times (≥3500 min) for each sample, but were stopped when delamination occurred during the thermal shock test. Failure by delamination was defined as being that point when an area of the TBC in excess of 10 mm in any one direction spalled off. The surfaces of the TBC before and after thermal shock were examined using an optical microscope.

3. Results and discussion

To verify the main phase of each column structure in the coating layer, diffraction patterns were obtained by using XRD. The bulk ingot used for the EB-PVD coating process was also analyzed by XRD. The results of the analysis showed that all of the bulk and coating materials consisted of ordered pyrochlore-type structures with fluorite-type structures, as shown in Fig. 1. The reported XRD patterns are inset for reference, pyrochlore-(PDF 79-1146) and fluorite-type Gd$_2$Zr$_2$O$_7$ (PDF 80-0471). A. F. Fluente et al. investigated the phase transition from the pyrochlore to the fluorite structure due to cation substitution at the Gd site. According to the results of their study, Gd substitution resulting from the smaller size of the Y produces anion-deficient fluorites. In this study, similar peak shifts in the XRD data were

| Table 1. Compositions of YSZ and Y$_2$O$_3$-doped Gd$_2$Zr$_2$O$_7$ | Y$_2$O$_3$ (wt%) | Gd$_2$O$_3$ (wt%) | ZrO$_2$ (wt%) |
|---------------------|----------------|----------------|--------------|
| Y$_2$O$_3$-doped Gd$_2$Zr$_2$O$_7$ | 3.55 | 55.95 | 40.50 |
| YSZ | 8.05 | 0 | 91.95 |
observed, indicating phase transition caused by pyrochlore-to-fluorite replacement by Y in the Gd lattice. Although the structure is modified into a fluorite structure, we observed that the main phase is the same irrespective of whether the coating is formed from a single ingot or two ingots. Our previous study showed that the thermal barrier properties, including the conductivity and thermal expansion, are improved by the addition of Y to Gd, even though the phase is changed into fluorite. Therefore, we expect the pyrochlore-to-fluorite phase to be suitable for high temperature applications.

Figure 2 shows the results of an elementary EDS analysis through the double-layer TBC prepared section for both the single-ingot and two-ingot processes. The relative amounts of Gd and Zr were plotted on a graph as a function of the coating thickness. Greater amounts of Gd in the top layer resulted from the existence of gadolinium in the Gd$_2$Zr$_2$O$_7$, and the greater amounts of Zr in the bottom TBC layer were caused by the zirconium in the YSZ. As shown in the zirconium analysis in Fig. 2, the amount in the bottom coat was more than twice that in the topcoat. This is caused by the Zr composition in the top and bottom coat materials, as listed in Table 1. Boundaries appear in the EDS analysis results, regardless of the ingot type. However, the boundary is somewhat less distinct in the case of the single-ingot process. The white dots indicate the corresponding compositions in the mapping diagrams. In this figure, the composition was more distinctly separated in the case of the TBC prepared using the two-ingot process. With the single-ingot method, the coating was deposited with YSZ and Gd$_2$Zr$_2$O$_7$ concurrently. As a result, a mixed composition was observed near the boundaries between the two coating layers. The relatively clear boundary in the double-layer TBC coating was also verified from the results of X-ray mapping, as shown in the right-hand graphs. The relatively gradual gradient of the indefinite boundary produced by the single-ingot process indicates that the Gd-Zr mixed materials evaporated during the EB-PVD process using the single ingot. Alternatively, the relatively steep gradient produced by the two-ingot process indicates that the boundary condition of the double-layered coating is influenced by the ingot type.

Figure 3 shows FE-SEM micrographs of the surface of the top layer of the double-layered TBC, deposited on the substrate both by the single-ingot and two-ingot processes. Common to both microstructures in the micrographs is the facet surface morphology. However, the average grain size of the faceted grains differed considerably depending on which ingot was used during the EB-PVD process. In the top layer of the TBC prepared using the single-ingot process, the diameter of the column was relatively coarse, in comparison with that of the TBC prepared using the two-ingot process. The coarser column structures in the top layer of the
double-layer TBC prepared using the single-ingot process could be verified from the fracture side, as shown in Fig. 4(a). During TBC coating using the two-ingot process, as shown in Fig. 4(b), the surface of each bottom-coat column functioned as a base-seed for the column of the top-coat column, unlike in the single-ingot process. Therefore, fine grains based on the triangle shape with approximate 1-μm grains were formed, which explains why the surface coated by the two-ingot process possessed a fine grain structure, as shown in the surface view in Fig. 3(b).

The changes in the grain size and microstructure shown in Figs. 3 and 4 for evaporation coatings formed using different ingots are most likely caused by the deposition methodology. The single-ingot process requires a longer ingot heating time relative to the two-ingot process because the shutter is not used during the melting of the stacked ingot. Since the process must be interrupted during the two ingot evaporation coating processes, re-nucleation probably leads to the smaller surface grain size.

Figure 5 shows the hardness measured by the Vickers indentation test on the surface of the double-layered TBC samples prepared by the single-ingot and two-ingot processes. The Vickers hardness of the coating layer deposited from the single ingot was greater than that prepared using the two-ingot process even though the coating materials were the same. The higher hardness indicates that a surface prepared using the single-ingot process will be better able to resist wear than a surface prepared using the two-ingot process. This result suggests that microstructure tailoring using different ingots may significantly influence the hardness of a TBC layer. The microstructure, as shown in Figs. 3 and 4, influences the hardness. The TBC layer prepared using the single-ingot process exhibited a higher hardness of 12 GPa, owing to the denser and coarser columns. The existence of the coarse columns reduces the ability of the Vickers indenter to make an indentation on the interfaces between columns, consisting of voids. Conversely, a TBC layer prepared using the two-ingot process exhibited a lower level of hardness, 9 GPa, which was caused by its finer grains. This increases the possibility of the Vickers indenter being placed on an interface. Previous results have shown that the interfaces between columns in EB-PVD coatings exhibit weak bonding and produce an open and connected porous region. Therefore, this structure contributes to both reduced thermal conductivity and hardness. The hardness will be reduced unless the indenter is placed exactly on a column, indicating that larger porous interfaces, as in the case of fine-grained columns, will influence the hardness. A. F. Renteria et al. reported that the coarsening of the columns by post-thermal heat treatment leads to a reduction in the volumes of both the inter-
M. Anglada et al. was able to attain a high hardness value with a scattered Young’s modulus value at lower indentation depths, which indicates that the indenter-columnar void interaction is critical to determining the mechanical properties. Similarly, the higher hardness tested at lower load shown in Fig. 5 clearly indicates this interfacial pore effect on the hardness. This kind of size effect has been discussed by many EB-PVD researchers.

Optical micrographs of the surface of the TBCs prepared using the single-ingot and two-ingot processes, both before and after the thermal shock test, are shown in Fig. 6. The TBC prepared using the single-ingot process delaminated after thermal shock tests had been conducted for 6 cycles in water, and after 40 cycles in air. The delaminated regions are clearly visible in the micrographs shown in Figs. 6(b) and 6(c). On the other hand, the TBC prepared by the two-ingot method did not exhibit surface cracking or delamination under the same thermal shock conditions, indicating that the TBCs prepared by the two-ingot process exhibit better thermal stability. Since most TBC failures result from delamination, the denser and coarser columns resulting from the single-ingot process may adversely affect TBC applications.

Residual stress analysis based on a composition analysis was conducted based on Fig. 2 and the results are plotted in Fig. 7. The residual stress near the interface boundary layer between the TBC layers was calculated from the elastic modulus $E$, thermal expansion coefficient $\alpha$, temperature difference during EB-PVD processing $\Delta T$, layer thickness $h$, and Poisson’s ratio $\nu$ for each layer material.

$$\sigma_R = (\alpha_1 - \alpha_2) \cdot \Delta T \times \{E_1/(1 - \nu_1) + (2h_1/h_2)(E_1/E_2)(1 - \nu_2)\}$$  \hspace{1cm} (1)

The thermal expansion coefficient data for the top coat ($Y_2O_3$-Gd$_2$Zr$_2$O$_7$), $9.3 \times 10^{-6}\cdot K^{-1}$, and that for the bottom coat layer (YSZ), $10.7 \times 10^{-6}\cdot K^{-1}$, were used based on our previous results. The residual stress in the top-coat layer was calculated as a function of the thickness $h$ from the top surface. Figure 7 indicates the magnitude and change in the residual stress near the boundaries between the two double layers. In the Gd$_2$Zr$_2$O$_7$ TBC top layer, the residual compressive stress was formed after EB-PVD processing had been completed. However, the magnitude and stress change depended on which ingot was used during the EB-PVD processing. Apparently, the TBC top layers prepared using the single-ingot process degrade to a greater degree during thermal cycling because the lower residual compressive stress and larger change in stress in the boundary region lead to the accelerated top-coat delamination. The phase shift to a higher angle in the case of the coated samples shown in Fig. 1 also indicates that the residual stress is compressive. The shift reflects the decrease in the size of the lattice in the coating layer, supporting the existence of a compressive system.

Another factor that must be considered is the strain tolerance of the TBC as deposited by EB-PVD. An advantage of using the EB-PVD technique to deposit TBC is the improved strain tolerance resulting from the columnar microstructure.
columns expand or shrink during heating and cooling cycles, thermal expansion mismatch strains can be accommodated by the intercolumnar gaps. The result shown in Fig. 4 indicates that the double-layered TBCs deposited by the two-ingot process exhibit clearer boundaries between their coating layers and much finer columns. This means that a TBC containing high volumes of micro/nano-sized interstices will be more tolerant of strain changes during thermal cycling.

4. Conclusions

In this study, double-layered Gd$_2$Zr$_2$O$_7$/YSZ TBC coatings were designed. The microstructure of the TBCs was controlled by using single-ingot and two-ingot methods for the EB-PVD process to tailor the properties of the TBC layers.

(1) Through an XRD analysis, we confirmed that the TBC layers prepared by the single-ingot and two-ingot methods consisted of the same phases.

(2) The developed microstructures depended on the ingots used during the EB-PVD processing. Both surfaces exhibited similar facet surface morphologies; however, the coated surface produced by the single-ingot process proved to be coarser than that produced by the two-ingot process. The boundary between the two TBC layers was analyzed; we found that the boundary was less distinct in TBCs prepared by the single-ingot method.

(3) The hardness was enhanced, from between 9 and 10 GPa to between 12 and 13 GPa, by microstructure tailoring using the single-ingot method because of the coarser/denser columns and reduced inter-columnar pore fractions in the top-coat layer of the double-layered TBC. As a result, we can say that microstructure tailoring using different ingots can significantly influence the hardness of the TBC layer.

(4) On the other hand, the thermal stability was found to be superior in double-layered TBC prepared using the two-ingot process because of the residual stress and strain tolerance of the top-coat layer. While the top-coat of the TBC prepared by the single-ingot process delaminated after 6 thermal shock cycles in water, and after 40 cycles in air, the TBC prepared by the two-ingot method did not exhibit any surface cracking or delamination under the same thermal shock conditions for 175 cycles, indicating that TBCs prepared by the two-ingot method offer better thermal stability.

Acknowledgements

This work was supported by the Technology Innovation Program funded by the Korea Evaluation Institute of Industrial Technology (KEIT), Korea. We thank the Korea Institute of Energy Research (KIER) for providing the EB-PVD apparatus and Mr. Kil Ho Guahlk for his assistance with the thermal shock tests.

References

1) S. Guo and Y. Kagawa, Ceram. Int., 32, 263–270 (2006).
2) T. Patterson, A. Leon, B. Jayaraj, J. Liu and Y. H. Sohn, Surf. Coat. Tech., 203, 437–441 (2008).
3) M. Gell, L. Xie, X. Ma, E. H. Jordan and N. P. Padture, Surf. Coat. Tech., 177–178, 97–102 (2004).
4) M. Tanaka, R. Kitazawa, T. Tomimatsu, Y. F. Liu and Y. Kagawa, Surf. Coat. Tech., 204, 657–660 (2009).
5) S. Guo, Y. Tanaka and Y. Kagawa, J. Eur. Ceram. Soc., 27, 3425–3431 (2007).
6) NASA Tech. Memo, NASA, Cleveland, OH., U.S.A., No. 86905 (1985).
7) N. P. Padture, M. Gell and E. H. Jordan, Science, 296, 280–284 (2002).
8) X. Q. Cao, R. Vassen and D. Stöwer, J. Eur. Ceram. Soc., 24, 1–10 (2004).
9) A. G. Evans and J. W. Hutchinson, Surf. Coat. Tech., 201, 7905–7916 (2007).
10) R. Vassen, A. Stuke and D. Stöwer, J. Therm. Spray Technol., 18, 181–186 (2009).
11) R. B. Lawn, J. Am. Ceram. Soc., 81, 1977–1994 (1998).
12) Y. H. Sohn, J. H. Kim, E. H. Jordan and M. Gell, Surf. Coat. Tech., 146–147, 70–78 (2001).
13) A. D. Jadhav, N. P. Padture, E. H. Jordan, M. Gell, P. Miranzen and E. R. Fuller, Jr., Acta Mater., 54, 3343–3349 (2006).
14) S. I. Jung, J. H. Kim, J. H. Lee, Y. G. Jung, U. Paik and K. S. Lee, Surf. Coat. Tech., 204, 802–806 (2009).
15) R. G. Wellman and J. R. Nicholls, Tribol. Int., 41, 657–662 (2008).
16) B. K. Jang and H. Matsubara, Mater. Lett., 59, 3462–3466 (2005).
17) U. Schulz, K. Fritscher and A. Ebach-Stahl, Surf. Coat. Tech., 203, 449–455 (2008).
18) S. H. Park, S. K. Kim and K. S. Lee, Philos. Mag., 86, 5453–5463 (2006).
19) K. S. Lee, K. I. Jung, Y. S. Heo, T. W. Kim, Y. G. Jung and U. Paik, J. Alloys Compd., 507, 448–455 (2010).
20) J. Singh, D. E. Wolfe and J. Singh, J. Mater. Sci., 37, 3261–3267 (2002).
21) M. Watanabe, C. Mercer, C. G. Levi and A. G. Evans, Acta Mater., 52, 1479–1487 (2004).
22) B. K. Jang, S. Kim, Y. S. Oh, H. T. Kim, Y. Sakka and H. Murakami, Scr. Mater., 69, 165–170 (2013).
23) R. M. Leckie, S. Kramer, M. Ruhle and C. G. Levi, Acta Mater., 53, 3281–3292 (2005).
24) A. Joulia, M. Verdelle and S. Rossignol, J. Eur. Ceram. Soc., 33, 2633–2644 (2013).
25) Z. H. Hu, X. Zhou, K. Wang, J. W. Dai and L. M. He, J. Alloys Compd., 587, 126–132 (2014).
26) Z. Xu, L. He, R. Mu, S. He and X. Q. Cao, J. Alloys Compd., 492, 701–705 (2010).
27) X. Q. Cao, R. Vassen, F. Tietz and D. Stoever, J. Eur. Ceram. Soc., 26, 247–251 (2006).
28) Z. H. Xu, L. M. He, R. D. Mu, S. M. He, G. H. Huang and X. Q. Cao, Appl. Surf. Sci., 256, 3661–3668 (2010).
29) S. Zhao, L. Gu, Y. Zhao, W. Huang, L. Zhu, X. Fan, B. Zou and X. Cao, J. Alloys Compd., 580, 101–107 (2014).
30) Z. Xu, L. He, R. Mu, X. Zhong, Y. Zhang, J. Zhang and X. Cao, J. Alloys Compd., 473, 509–515 (2009).
31) W. Ma, H. Dong, H. Guo, S. Gong and X. Zheng, Surf. Coat. Tech., 204, 3366–3370 (2010).
32) J. R. Nicholls, K. J. Lawson, A. Johnstone and D. S. Rickerby, Surf. Coat. Tech., 150–151, 383–391 (2002).
33) J. A. Diaz-Guillen, A. F. Fuentes, M. R. Diaz-Guillen, J. M. Almanza, M. Santamaria and R. C. Leon, J. Power Sources, 186, 349–352 (2009).
34) Z. G. Liu, J. H. Ouyang, Y. Zhou, J. Li and X. L. Xia, J. Eur. Ceram. Soc., 29, 647–652 (2009).
35) B. A. Movchan and F. D. Lemkey, Surf. Coat. Tech., 165, 90–100 (2003).
36) Q. Wei, H. Guo, S. Gong and H. Xu, Thin Solid Films, 516, 5736–5739 (2008).
37) R. G. Wellman, H. Tourmente, S. Impye and J. R. Nicholls, Surf. Coat. Tech., 188–189, 79–84 (2004).
38) A. F. Renteria and B. Saruhan, J. Eur. Ceram. Soc., 26, 2249–2255 (2006).
39) Y. Gaillard, E. Jimenez-Pique and M. Anglada, Philos. Mag., 86, 5441–5451 (2006).
40) H. Wang and X. Hu, J. Am. Ceram. Soc., 79, 553–556 (1996).