The influence of coating compliance on the delamination of thermal barrier coatings

Hengbei Zhao*, Zhuo Yu, Haydn N.G. Wadley

Department of Materials Science and Engineering, University of Virginia, Charlottesville, VA 22903, USA

A R T I C L E   I N F O

Article history:
Received 12 October 2009
Accepted in revised form 15 January 2010
Available online 21 January 2010

Keywords:
Yttria stabilized zirconia coatings
Thermal oxidation kinetics
Delamination buckling

A B S T R A C T

Low relative density yttria stabilized zirconia (YSZ) thermal barrier coatings have been deposited on NiCoCrAlY over-lay bond-coated Hastelloy-X substrates by an electron beam-directed vapor deposition (EB-DVD) method. The coatings have been applied to both roughened and smooth bond coat surfaces. During thermal cycling, those deposited on roughened surfaces spalled by delamination with the failure located in the YSZ layer close to the thermally-grown oxide (TGO). The cracks were initiated at “corn kernel” defects in the YSZ layer by a bond coat rumpling mechanism. Coatings applied to smooth bond coat surfaces had much longer (factor of two) spallation lifetimes and delamination occurred at the bond coat/TGO interface by the nucleation and coalescence of interfacial voids. In both cases, the delamination lifetime decreased with coating density and significantly exceeded the lifetimes of higher density coatings made by conventional electron beam deposition methods. The enhanced life of the more porous coatings is consistent with an increased compliance. This reduces the YSZ layers stored strain energy contribution to the driving force for interface delamination.

© 2010 Elsevier B.V. All rights reserved.

1. Introduction

Thermal barrier systems are now widely used to protect airfoils within the hottest sections of gas turbine engines. Their benefits are achieved by inhibiting the airfoil degradation mechanisms caused by the high combustion temperature [1,2]. These thermal barrier coating systems have three components; (i) a 100–200 μm thick oxide ceramic top coat usually consisting of porous yttria stabilized zirconia (YSZ) that retards thermal transport to the airfoil, (ii) a 75–150 μm thick MCrAlY (M=Ni, Co) [3] or Pt-modified (Ni, Pt, Al) nickel alumide bond coat [4] deposited on the airfoil to provide oxidation and hot corrosion protection and, (iii) a thermally-grown oxide (TGO) layer consisting largely of α-Al2O3 that forms on the bond coat during engine operation and results in a slow bond coat oxidation rate [5,6].

The bond coats are applied by either diffusion aluminization [7,8], vacuum plasma spraying [9,10] or by vapor deposition (over-lay coatings) [11]. In MCrAlY type bond coats, low concentrations of reactive elements such as yttrium are added to improve the adhesion of the TGO, primarily by gettering interface embrittling elements such as sulfur that diffuses up from the airfoil alloy [12] and suppression of the outward scale growth [13]. The YSZ layer is deposited either by electron beam-physical vapor deposition (EB-PVD) [14] or by air plasma spray (APS) [15,16]. A recent variant of the EB process uses inert gas jets to focus the vapor plume and direct it to a substrate [17–21]. This EB-directed vapor deposition (EB-DVD) process can be operated in a way that results in YSZ coatings of lower density (higher pore content), and therefore lower thermal conductivity than EB-PVD counterparts [21].

In aero-turbines, the coating system is subjected to many thousands of thermal cycles. During take-off, the bond coat temperature can reach 1000 °C or more [22], while the outer surface of the top coat is 100–200 °C hotter. At these temperatures, significant densification of the YSZ (sintering) [23] and TGO layer growth occurs [6]. Moreover, thermal expansion misfit between the TGO and bond coat results in large compressive stresses in the TGO layer during cooling [24,25]. These stresses drive rumpling of the bond coat at a rate that depends upon multiple parameters, including the creep strength of the bond coat alloy and the initial amplitude of the surface roughness [26]. In systems with low creep strength bond coats such as NiPtAl, the rumpling amplitudes are sufficiently large that the YSZ coating delaminates, just above the TGO layer, and the eventual coalescence of these delaminated areas results in coating spallation [27]. In systems susceptible to rumpling, the bond coat surface condition can affect the spallation life and the failure pathway. Polish bond coats sustain many more thermal cycles than their grit-blasted counterparts because the reduced initial imperfection amplitude slows the rumpling rate [28–30]. However, the surface roughness could also affect durability in other ways since it has an influence on the morphology of vapor deposited YSZ layers.

For all thermally-driven delamination mechanisms, the cyclic life is believed to be strongly affected by the thickening rate of the TGO layer. The thickening occurs predominantly by inward diffusion of oxygen anions along the grain boundaries [31]. When yttria is incorporated in the TGO, it can act as a preferred channel for rapid
inward diffusion of oxygen, resulting in local TGO thickness imperfections [32]. These imperfections (pegs) may induce local zones of tensile stress that can nucleate small cracks [33]. Controversially these pegs are also thought to improve the adhesion by fastening the TGO to the bond coat [34,35]. The role of the YSZ layer, and especially its pore content, in the delamination is less clear and is the focus of the work described here.

An electron beam-directed vapor deposition (EB-DVD) technique has recently been used to deposit 7YSZ coatings [36]. This deposition process uses a gas jet to entrain the electron beam created vapor and then efficiently deposit it upon a substrate. The pore volume fraction and morphology can be widely varied by adjusting the jet flow conditions, by rotating the substrate or by changing the substrates temperature [20]. Coatings deposited at relatively low substrate temperatures (near 1000 °C) using jet flow conditions that result in a broad angle of incidence distribution of the flux result in coatings with controllable pore volume fractions between 25 and 45% [37]. This can be significantly higher than state of the art EB-PVD coatings where the pore fraction is in 20–25% range [38]. This deposition capability now allows a systematic assessment of the influence of coating porosity on the foregoing mechanisms of delamination. The investigation is conducted on 7YSZ-DVD coatings deposited on a Hastelloy-X substrate with an overlay NiCoCrAlY bond coat that we show to be prone to rumpling during thermal cycling.

The article is organized as follows. In Section 2, the coating and bond coat characterization is described. In Section 3, the thermal cycling response of coatings deposited on the rough (grit-blasted) bond coats is presented. We find a significant decrease of the cyclic life as the coating density increases. In Section 4, the corresponding response of the system with a polished bond coat is described. While the lifetime is dependent upon the initial condition of the bond coat surface prior to deposition, it still decreases with increasing coating density. The spallation lifetimes of the EB-DVD coatings are shown to be 1.5–2 times greater than those for EB-PVD coatings deposited on similar bond coats. In Section 5 we show that the dependence of the thermal cyclic life upon coating density arises from the density dependent coating compliance which controls the elastic strain energy stored in the 7YSZ layer.

2. Coating deposition and characterization

2.1. Deposition

Yttria stabilized zirconia coatings were deposited on rotated substrates using an electron beam-directed vapor deposition (EB-DVD) technique, Fig. 1 [21]. Briefly, a 12.5 mm diameter 7 wt.% Y2O3–ZrO2 source rod (supplied by TCI Ceramics Inc., MD) was positioned in the throat of a water-cooled copper nozzle and an annular oxygen doped helium gas jet was formed around the periphery of the evaporation source. A high voltage electron beam was used to evaporate the source. For the study reported here, the evaporated flux was entrained in a 10.0 slm helium, 1.0 slm oxygen gas flow. The pressure upstream of the nozzle was 140 Pa and the in the deposition chamber was 16 Pa. The precursors were coated with an 80 µm thick 7YSZ layer at a substrate temperature of 1000 °C using a range of rotation rates between 0.5 and 30 rpm at a deposition rate of 3–4 µm/min. The coatings were applied to 25.4 mm diameter Hastelloy-X substrates (provided by GE Aircraft Engines, Inc.) which had been overlay coated with a 100–200 µm thick Ni0.475Co0.22Cr0.2Al0.1Y0.005 (weight percentage) bond coat by an EB-PVD technique. The bond coat surfaces were either grit-blasted using 220 grade silica grit at a pressure of 30–40 psi or polished to a 1 µm diamond finish. Both sets of substrates were ultrasonic cleaned with acetone prior to deposition of the ceramic layer.

2.2. Property measurements

The coating density (ρ) was obtained by direct measurement of the weight of the substrate with and without the coating, Table 1. The experimental error was affected by the measurement of the coating thickness and the weight, (both were about + or −2.5%). The coatings deposited on rough bond coats had a slightly lower density due to their many growth defects. This was consistent with recent assessments of atomic assembly upon rough surfaces which have shown that surface asperities create incident flux shadows during vapor deposition and this results in very low density columnar structures [39].

The elastic modulus of porous materials usually decreases with pore volume fraction but is also pore morphology dependent [40]. The elastic modulus of representative coatings was deduced from indentation measurements using a Nanoindent Nanovea Micro Hardness Tester with a Vickers diamond indenter. The shallow cone angle ensured that the indenter contacted a lateral region of ~25 µm and sampled the stiffness of many growth columns. The indenter load and displacement were recorded during loading and unloading in the indentation process. The reduced elastic modulus was evaluated from the load-displacement data using the Oliver and Pharr method [41], and the Young’s modulus of the coatings then derived from the reduced modulus. Each of the results presented later represent the average of 5–10 indentation measurements, with error bars indicating standard deviation in the measurement. Measurements were made on the polished cross-sectional sides of the coatings.

Table 1
Coating density dependence upon rotation rate.

<table>
<thead>
<tr>
<th>Rotation rate (rpm)</th>
<th>Rough BC</th>
<th>Smooth BC</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.5</td>
<td>3.30</td>
<td>3.38</td>
</tr>
<tr>
<td>6</td>
<td>3.55</td>
<td>3.63</td>
</tr>
<tr>
<td>20</td>
<td>3.98</td>
<td>4.10</td>
</tr>
</tbody>
</table>
were also made on a sample with an independently measured modulus. These calibration experiments indicated that the nanoindentation modulus measured upon the cross-sectional surface was very similar to that reported for the reference coating.

2.3. Structure characterization

The grit-blasted bond coat has a rough surface with many impact craters and mean roughness amplitude of 2.6 µm. The polished surface with the mean roughness of 0.02 µm contained pores that originated during deposition. Some of these pores were sectioned by the polishing process. The bond coat had a three-phase microstructure consisting of β-NiAl, γ-Ni and a minor amount of an α-Cr (Co) precipitate [32]. The bond coat composition was characterized by EDS (Table 2). The grit-blasted surfaces had a significantly higher concentration of Si, Mg and Ca, presumed to have originated from contaminants on the grit.

The cross-sectional structures of TBC’s on rough bond coats deposited at 0.5, 6 and 20 rpm, Fig. 2, reveal primary columns incorporating multiple smaller diameter columns. They have a wavy morphology resulting from the gradual change in angle of incidence of the vapor flux as the samples were rotated. As the rotation rate increased, the primary columns became less wavy and wider. “Corn kernel” growth defects formed above concavities in the bond coat by the nucleation of small columns normal to the sides of the cratered surface. The coating surfaces comprised pyramidal tips with sharp facets (Fig. 3). The sides of the columns contained feathery pore structures inclined at about 45° to the growth axis. Coatings on polished surfaces contained narrower columns with smaller intercolumnar pores and a more uniform overall structure, Fig. 4. They contained no “corn kernel” growth defects.

3. Thermal cycling of coatings deposited on a rough bond coat

Thermal cycling tests were performed in a top-loading furnace. Each cycle consisted of 1-hour hold at 1100 °C followed by 10 min of cooling in forced flowing laboratory air to a temperature of ~50 °C. The specimens were examined visually every 20 cycles to detect the onset of delamination, with “failure” defined to have occurred when ~20% of the TBC had spalled. To better understand the growth kinetics of the TGO layer, some samples were taken out for observation and analysis before failure. The effects of thermal cycling upon the surface morphology and the cross-sectional structure of a coating deposited on a rough bond coat (at 6 rpm) are depicted on Fig. 3. Surface sintering clearly began during the first few thermal cycles, Fig. 3b and e. The pyramidal column tips gradually became more rounded and the primary columns locally fused as thermal cycling continued, Fig. 3b to c. The cross-sections, Fig. 3e–f, indicate that the feathery pores within the columns experienced rapid sintering and were transformed to spheroids. Significant inter-column porosity remained even after exposure to 630 thermal cycles.

The coatings eventually failed by buckling delamination, starting at the edge of coupons and propagating inward across the surface. The number of thermal cycles to “failure” is plotted as a function of sample rotation rate in Fig. 5. It can be seen that the lifetime slowly decreased as the rotation rate was increased. A data point for the spallation lifetime of EB-PVD coatings deposited on the chemically similar bond coat and subjected to similar thermal cycling [28] is included in Fig. 5. It reveals that DVD coatings have a significantly greater durability. It is also noted that the superalloy used in [28] is a single crystalline Rene NS, different from Hastelloy-X in this work. The different superalloy composition and bond coat manufacturing might affect the thermal mismatch stress and thereby the TBC lifetime [42,43]. The lifetime difference between PVD and DVD samples could be partly ascribed to this.

The evolution of the delamination near the TGO layer with cycling is shown in Fig. 6. A thin, alumina-rich TGO layer had formed during deposition of the coatings, Fig. 6a. Thermal cycling caused the TGO layer to thicken, and the amplitude of the surface undulation to increase by a rumpling mechanism [30]. This rumpling eventually resulted in the formation of small cracks in the TBC above local depressions in the bond coat (Fig. 6b). These regions frequently contained “corn kernel” defects which adhered poorly to the adjacent TBC. As the undulation amplitude increased, these cracks propagated laterally and interlinked, causing eventual spallation of the coating (Fig. 6c and d). The preferred fracture path was located within 1–2 µm

<table>
<thead>
<tr>
<th>Bond coat elements</th>
<th>Ni</th>
<th>Co</th>
<th>Cr</th>
<th>Al</th>
<th>Y</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fraction (wt.%)</td>
<td>47.3</td>
<td>20.3</td>
<td>21.0</td>
<td>11.0</td>
<td>0.4</td>
</tr>
</tbody>
</table>

Si, Mg and Ca have been detected by EDS on the grit-blasted bond coat surface. The surface contained 0.5 wt.% Si, 0.2 wt.% Mg and 0.2 wt.% Ca. These elements were undetectable on the polished surface.
above the TGO layer, through planes of pores created in the TBC when
the vapor incidence angle was most oblique to the surface. Samples
subjected to the greatest number of thermal cycles also had small,
isolated cavities at the bond coat/TGO interface (Fig. 6d) but these
were not responsible for the spallation. The TGO growth rate was
independent of the rotation rate during deposition (Fig. 7) and had an
approximately parabolic dependence upon time: $h^2 = 2kp_t$, where $h$
is the TGO thickness, $t$ is the hot-time and $k_p \sim 0.025 \text{ µm}^2/\text{h}$. The critical

![Fig. 3. Evolution of the surface and cross-sectional morphology of a 7YSZ coating deposited at 1000 °C on a roughened bond coat surface. The coatings were grown using a rotation rate of 6 rpm and thermally cycled by heating to 1100 °C for 1 h.](image)

![Fig. 4. Evolution of a TBC system deposited on a smooth bond coat surface at 6 rpm as it was thermally cycled. An increase in the TGO thickness and TBC intercolumnar gaps with cycling can be seen.](image)
thickness at “failure” was $\sim 5.0 \, \mu m$, but decreased with increase in coating density as discussed below.

4. Thermal cycling of coatings on a polished bond coat

The coatings applied to polished bond coats were much more durable than their counterparts deposited on roughened surfaces.

![Graph](image1)

**Fig. 5.** Spallation lifetimes of 7YSZ coatings as a function of the rotation rate used for their growth. Data are also shown for a 7YSZ coating deposited on the chemically similar rough bond coat using conventional EB-PVD [28].

![Graph](image2)

**Fig. 7.** TGO layer thicknesses versus number of (1 h) thermal cycles at 1100 °C for the DVD 7YSZ coatings. The data are well fitted by a parabolic law with a parabolic rate constant $k_p = 0.025 \, \mu m^2 / hr$.

**Fig. 8.** The number of cycles to “failure” ranged from $\sim 1200$ to 900 (with increasing rotation rate), compared with 550 to 400 for the rough bond coat samples. The datum for an EB-PVD coating grown on the chemically similar polished bond coat system is also plotted in Fig. 8 [28]. The DVD samples all failed by delamination at the bond coat/TGO interface. Significant sintering occurred with several small

![Images](image3)

**Fig. 6.** Evolution of the TGO layer region for a coating deposited on a roughened bond coat surface at a rotation rate of 6 rpm. A significant increase in rumpling amplitude upon thermal cycling can be seen. b shows that microcracks were initiated in the TBC layer at the TBC–TGO interface. These formed first in regions where deep rumpsles were present in the bond coat. c shows that these microcracks developed into delaminations that preferentially penetrated the high porosity bands formed during each sample rotation during TBC deposition. d shows the spalled interface and the first appearance of a crack at the bond coat–TGO interface.
columns coalescing to form larger columns with wider intercolumnar gaps, Fig. 4.

During thermal exposure, a $\beta$-depleted region developed near the TGO interface, and increased in width from $\sim 0.5$ $\mu$m in the as-deposited condition to $\sim 20$ $\mu$m at “failure.” The TGO developed local variations in thickness attributed to the oxidation of the initial surface-connected voids in the bond coat, as well as the emergence of YAG domains (pegs), Fig. 9. An intermixed zone of YSZ and alumina existed above the TGO. The critical TGO thickness at “failure” was $\sim 5.5$ $\mu$m. It again decreased as the coating density was increased (Fig. 14). The parabolic rate constant $k_p$ was 0.014 $\mu$m$^2$/h, Fig. 10; half that for the grit-blast bond coat. This is presumed to have occurred because of the presence of the contaminants (Si, Mg, and Ca) on the roughened surface [29].

5. Mechanisms governing durability

The foregoing measurements indicate that the cyclic lifetimes of the 7YSZ coatings are: (i) highly sensitive to the bond coat surface condition. Those grown on smooth bond coats have approximately twice the durability of those deposited on grit-blasted surfaces and the failure mechanisms are quite different. (ii) The EB-DVD coatings have significantly greater life than those exhibited by EB-PVD coatings deposited on the chemically similar bond coat/substrate system. (iii) The durability of the EB-DVD coatings is dependent upon the rate at which it was rotated during deposition.

The mechanism governing the durability of TBCs grown on rough bond coats is summarized in Fig. 11. As the bond coat rumples, concavities are amplified. The displacements induce tension in the superposed TBC, causing cracks to form (often next to “corn kernel” defects) which then propagate along planes of lowest local toughness (through planes of closely spaced small pores). Coalescence of a sufficient number of contiguous cracks results in a delamination that exceeds the critical length for buckling [6]. Most systems with MCrAlY bond coats are not prone to such rumpling [6] except upon stepwise cycling [44]. The thermal expansion and creep strength properties of this particular overlay composition must be sufficiently different to allow rumpling. The specifics remain to be ascertained.

Upon polishing, rumpling is suppressed, as anticipated from the Balint and Hutchinson model [26] and previous observations [27,30]. Instead, delamination occurs at the TGO/bond coat interface (Fig. 12). Small cracks initiate at the TGO thickness heterogeneities, Fig. 9c. For these to grow and coalesce, the toughness of the intervening interface
must be lower than that on pristine interfaces \cite{45,46} and this appears to be due to the appearance of small flat voids in the intervening regions, possibly in conjunction with sulfur segregation and the embrittlement of this interface \cite{12,47}. We note that Hf is found to effectively getter S and inhibit its segregation to the interface, thereby enhancing the adhesion between the TGO and bond coat \cite{48}. However, no Hf was doped into the bond coat studied here and this may have contributed either to the formation of the small flat voids or embrittlement of the interface between them.

The increase in spallation life as the rotation rate during deposition was decreased is similar for coatings applied to either bond coat surface condition. The lifetime of the DVD coatings on both surfaces are also significantly greater than EB-PVD coatings applied on the chemically similar bond coat system. While there are many morphological differences in the coatings that might be invoked to account for these observations, the most fundamental is the significant difference in coating density and thus elastic modulus. We note that the elastic modulus of the as-deposited coatings increases with coating density, Fig. 13. If the spallation data are plotted against coating density, Fig. 14b, it is apparent that the lifetime

![Fig. 10. TGO layer thickness dependence upon cyclic high temperature exposure.](image)

![Fig. 11. Spallation failure mechanisms for DVD 7YSZ coatings deposited on roughened bond coats susceptible to rumpling. a shows initial (corn kernel) growth defects near depressions in the surface of the as-deposited coatings. b shows rumpling and TBC microcrack initiation near initial depressions in the bond coat. c shows microcrack coalescence and coating spallation. Note that the TGO layer thickness and rumpling amplitude both increase with cycling for this bond coat.](image)

![Fig. 12. Schematic illustration of spallation mechanism for DVD TBC coatings deposited on a polished bond coat. The absence of initial undulations greatly slows the bond coat rumpling rate. Fracture was initiated at surface intersecting voids whose interior oxidation produced locally thickened TGO. These coalesced with simultaneously forming flat voids at the bond coat–TGO interface to cause delamination failure of the coating. No rumpling occurred when the bond coat surface was polished.](image)
decreases for both bond coats and the data from the EB-PVD experiments falls on the same trend line. The thickness of the TGO layers at which spallation occurs also decreases as the density increases (Fig. 14a). The enhanced durability relative to EB-PVD coatings, as well as the trend with rotation rate, is attributed to the lower coating density (Fig. 14).

If the spalling is deemed to occur by edge delamination at the TGO/bond coat interface (as for the polished specimens), the process is governed by the steady-state energy release rate. Mumm and Evans [32] and Spitzberg et al. [30,49] have shown that for a TGO/TBC bilayer, the steady-state energy release rate is given by:

$$G_{ss} = \frac{(E_1h_1(1 + \nu_1)e_{1r} + E_2h_2(1 + \nu_2)e_{2r})^2}{2(E_1h_1 + E_2h_2)} - \frac{M\Delta \kappa}{2}$$

with moment

$$M = \frac{E_1E_2h_1h_2(h_1 + h_2)(1 + \nu_2)e_{2r} - (1 + \nu_1)e_{1r}}{2(E_1h_1 + E_2h_2)}$$

and curvature

$$\Delta \kappa = \frac{6[(1 + \nu_2)e_{2r} - (1 + \nu_2)e_{1r}]}{h_1h_2}\left(\frac{h_1}{\xi} + \frac{h_2}{\xi}\right)$$

with

$$\xi = \frac{E_1'}{E_2'}\left(\frac{h_1}{h_2}\right)^2 + \frac{E_1'}{E_1}\left(\frac{h_2}{h_1}\right)^2 + 4\left(\frac{h_1}{h_2}\right) + 4\left(\frac{h_2}{h_1}\right) + 6$$

and

$$E_1' = \frac{E_1}{1 - \nu_1^2}, \quad E_2' = \frac{E_2}{1 - \nu_2^2}$$

The subscripts 1 and 2 refer to the TGO and TBC layers respectively. The quantities $\nu$, $h$, $e_r$, and $E$ are the Poisson ratio, thickness, residual strain and Young’s modulus, respectively. The residual strain is dominated by the thermal expansion mismatch, $\Delta \alpha$, between the coating and substrate and by $\Delta T$, the temperature drop from the peak temperature; $e_r = \Delta \alpha \Delta T \approx -0.0026$. The critical TGO thickness can be calculated by equating $G_{ss}$ to the mode II interfacial toughness. If this toughness is assumed to be the same value for all of the coatings (80 J/m²) [32,49], and assuming $\nu_1 = 0.2$, $\nu_2 = 0.3$, $e_{1r} \approx -0.0076$, $E_1 = 380$ GPa and $h_2 = 80$ µm, the relationship between the critical TGO thickness and TBC modulus can be plotted, Fig. 15a. These estimated thickness levels are comparable to those measured at “failure” for the coatings on the polished samples and they predict the same critical thickness dependence upon TBC density found in the experiments, Fig. 14a.

The result in Fig. 15a indicates that the critical TGO thickness (and therefore the spallation lifetime) has a maximum value (of ~6 µm) of a coating modulus of ~25 GPa. For the coatings grown here, it increases as the TBC density is decreased and this trend is evident in the experimental data shown in Fig. 14. The analysis also highlights the importance of the coating thickness, Fig. 15b and interface toughness, Fig. 15c. It is interesting to note that even modest increases of toughness (from 80 to 90 J m⁻²) are predicted to increase the critical thickness of a TBC with a 30 GPa Young’s modulus by about 1 µm and therefore increase the lifetime of a coating on a polished bond coat from 1330 to 1770 cycles. While changes to the coating thickness are predicted to have little effect upon the life, they do affect the optimum density (modulus) for maximum life: thicker coatings require the use of a lower coating density (modulus). The results above appear consistent with Xu et al. [50] who have argued that a larger in-plane modulus or greater coating thickness increase the
energy release rate $G$ and thus should reduce the durability. This trend is clearly evident in Fig. 15.

6. Conclusions

Low density 7YSZ coatings have been deposited on overlay NiCoCrAlY bond-coated Hastelloy-X substrates using an electron beam, directed vapor deposition method and their thermal cycling behavior has been investigated. We find that:

1. Coatings grown on grit-blasted bond coats have only half the lifetime of those deposited on polished bond coats. The bond coat surface condition also governs the mechanism of spallation failure. The lifetimes of coatings applied to rough bond coats are governed by bond coat rumpling whereas those for the polished bond coat coatings is dictated by interface delamination.

2. The spallation lifetime of 80 µm thick coatings applied to both bond coats decreases as the coating density is increased. The trend extends to coatings made by the EB-PVD method which exhibit significantly reduced failure times compared with DVD counterparts. Measurements of the coatings elastic modulus confirm a reduction in coating modulus as the pore content of the coatings increases and therefore a reduction in the stored elastic strain energy available to drive interfacial failure.

3. The critical TGO thickness at coating failure is shown to decrease with increase in coating density consistent with a steady-state strain energy release model. The lower critical TGO thicknesses of the rough bond coat surfaces are consistent with a reduced interfacial toughness which might result from contaminant elements S, Mg and Ca.

4. The results and strain energy release rate models indicate that spallation lifetimes can be optimized by (i) controlling the deposition process so that the in-plane Young's modulus of an 80 µm TBC coating is 25–30 GPa (the optimum modulus decreases as the coating thickness is increased), (ii) reducing the amplitude of initial undulations on the bond coat surface and (iii) using bond coat chemistries that have reduced propensity for void formation in the bond coat adjacent to the TGO.

Acknowledgements

We are very grateful to Anthony Evans for many helpful discussions of the interpretation of this work. We also thank David Wortman of GE Aviation for kindly providing the bond-coated substrates and Brian James for making some of the nanoindentation measurements at Nanovea (Affiliate of Micro Photonics Inc., CA). We are grateful to the Office of Naval Research who supported the research (contract # N00014-03-1-0297) under the program direction of Dr. David Shillier.

References


Fig. 15. (a) Predicted critical TGO thickness for TBC buckling-delamination failure as a function of 7YSZ Young’s modulus assuming an interfacial toughness of 80 J/m² and a coating thickness of 80 µm. The coating density–Young’s modulus correlation of Fig. 14 was used to obtain the modulus values. b shows the effect of varying the TBC thickness while maintaining the toughness constant at 80 J/m². c the effect of interfacial toughness upon the critical TGO thickness for 80 µm thick coatings.