

Effect of substrate curvature on residual stresses and failure modes of an air plasma sprayed thermal barrier coating system

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Abstract

A set of aerofoil shaped air plasma sprayed thermal barrier coated (APS-TBC) specimens were adopted in this paper to investigate the stress distributions in the ceramic top coat (TC) and the thermally grown oxide (TGO), the mechanism of local crack generation and propagation at the TC/BC (bond coat) interface. The failure mode of the TBC system, the distribution of asperities at TC/BC interface, thickness of the TC and BC, and the TC microstructure were found to be influenced by substrate curvature. Residual stress was therefore measured across the thickness of the TC, along the undulating TGO and mapped at locations of asperities where failure tended to occur to interpret the initiation of local failure. The role of the TGO was investigated via its chemical bonding with the TC and the decohesion occurring at the TGO/BC interface. The crack propagation at the interface has been discussed with respect to the macro-failure of the TBC system.

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1. Introduction

Blades used for land-based gas turbine power plants and aero engines operate at high gas temperatures above which the parent superalloy cannot withstand without additional internal cooling.^{1,2} To protect the blades and to further promote efficiency, turbine blades with ceramic thermal barrier coatings (TBCs) were introduced into service during the early 1970s.³ The benefits these low-thermal conductivity ceramic systems offer the base material of components in the engines are significant, including extended component lives, due to a reduced creep rate and lower rate of oxidation. TBCs normally consist of three layers: an outer ceramic top coating (TC) to insulate the superalloy substrate from high gas flow by sustaining an appreciable temperature difference of up to 300 °C depending on TC

thickness^{4,5}; a metallic bond coat (BC) providing good adherence between the TC and substrate as well as oxidation and/or corrosion protection to the underlying superalloy. A thermally grown oxide (TGO), predominately Al₂O₃, develops between the TC and the BC during exposure to the service temperature.^{6,7} Within the BC, the Al present, either in solution or in the form of β -NiAl phase, creates and feeds this protective alumina layer. It constitutes the Al reservoir of the coating. Depletion of this reservoir results in compositional changes in the TGO that may form α -Al₂O₃ and/or a mixture of chromia and spinels ((Cr,Al)₂O₃) and (Ni(Cr,Al)₂O₄).⁸ The growth of the TGO between the TC and BC can be divided into three stages: a transient stage, when all thermodynamically stable oxides can form (e.g. θ -Al₂O₃, Cr₂O₃, NiO, etc.); a steady-state stage, when the high temperature stable, long-term phase (α -Al₂O₃) is established and grows; and finally a breakaway stage, when the growth of less protective oxides leading to the failure of the scale.⁹ At oxidation temperatures below 1000 °C, the transient θ -Al₂O₃ phase is also thermodynamically stable and can co-exist with α -Al₂O₃ in the TGO.¹⁰

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The failure of APS-TBC systems under oxidation is a complex mechanism and is still not fully understood.¹¹ The process involves several general phenomena including: thermal expansion mismatch stresses; the growth of the TGO at an undulating BC/TC interface; the formation of spinel/mixed oxides due to Al-depletion in the BC; the sintering of the porous ceramic TC leading to a deterioration of strain tolerance and thermal resistivity; the degradation of the metal/ceramic interface toughness; cracking, crack coalescence, and delamination of the TBC at either interface or within an individual layer (TGO or TC).¹²

The residual stresses within the TBC arise from three main sources: (i) the deposition process,^{6,7,13} (ii) thermal stress mismatch during cooling^{14,15} and (iii) growth of the TGO. This is usually associated with the nucleation and growth of micro-cracks caused by stresses developing between the various layers or within the TGO.^{16,17}

The degree of curvatures, concave or convex, that are characteristic of the geometry of turbine blades and vanes, influence the thermal/residual stress distribution within multilayer TBC system^{18,19} which greatly affects the energy release rates associated with interface delamination and crack propagation (circumferential or axial)^{18,20} during spallation. The roughness of TC/BC interface has been found to influence the nature and evolution of residual stresses in the ceramic as well as in the TGO.²¹ Protrusions of ZrO_2 into the BC are preferred sites for damage initiation. The sintering effects at high temperature in such small volumes fixed by the geometry of the BC produce localised tensile stresses. At the same time, the stresses within the TGO layer contouring such protrusions change during oxidation. Undulations of TGO into the BC usually show higher compressive stresses compared to TGO undulations into the TC.²¹ In the TGO, besides the substrate curvature variation, the interface undulation and local geometrical features can also affect the stress distribution and relaxation under cyclic behaviour. Therefore the failure modes within this layer are complex.^{22,23}

In this paper, a set of modified aerofoil-shaped specimens, with representative turbine blades' curved features, were used to study the effect of the substrate curvature. The deposition process, the resulting TC microstructure, the TC, BC and TGO thicknesses, and the TC/TGO interface undulation were considered in this study as parameters that may influence the failure modes of the TBC system. Residual stress measurements were then undertaken on cross-sections along the thicknesses of TC and TGO after extended thermal exposure to correlate the residual stress evolution to the coating degradation in terms of crack initiation and propagation. Four types of microstructures at the TC/TGO interface were found to have the potential to start and eventually lead to interfacial failure. The residual stress maps at these locations are discussed with respect to mechanical debonding at the interface and are compared with existing models.

2. Materials and methods

Modified aerofoil-shaped specimens were designed to recreate curvatures found on industrial gas turbine blades (Fig. 1).

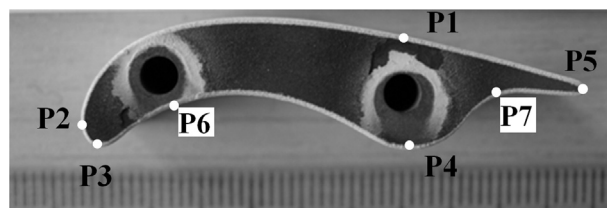


Fig. 1. Picture of a modified aerofoil-shaped specimen. P1, P2, P3, P4 and P5 are convex areas whereas P6 and P7 are concave.

Deposited on a CMSX4 Ni-superalloy substrate by high-velocity oxy-fuel (HVOF) spraying was an AMDRY 995 CoNiCrAlY BC (Table 1), over-coated with an air plasma sprayed (APS) yttria-stabilised tetragonal zirconia (YSZ) TC. The manufacturing process caused the TC and BC thicknesses to vary with the curvature. The BC was measured to be between 50 and 135 μm and the TC between 160 and 300 μm . Long cycle thermal exposure tests were carried out on a set of eight specimens in resistance heated furnaces at 925 °C in air for up to 10,000 h. After each cycle (250 h), the specimens were removed gradually from the hot furnace and cooled in air to room temperature. Selected specimens were removed for destructive examination after periods of 100, 2740, 4000, 7000 and 10,000 h. As shown in Fig. 1, the specimens were each ~ 43 mm long and ~ 22 mm wide with curvatures that changed around the periphery. Seven locations, P1–P7, the curvature of which are listed in Table 2, were chosen around the modified aerofoil-shaped samples for study (Fig. 1). The curvature was calculated as the inverse of the radius, thus, a flat surface would have a curvature equal to zero, with the sign indicating whether the surface was concave or convex.

Both scanning electron microscopy (SEM) and energy dispersive X-ray spectroscopy (EDX) were used to observe cross-sections in order to characterise the microstructure of the specimens. Line-scans of TC residual stress and mapping of residual stresses along the undulation of the TGO were undertaken on the cross-sections at each position using non-destructive laser stimulated spectroscopy, both Raman spectroscopy (RS) and photo-stimulated luminescence piezospectroscopy (PLPS).²⁴ The equipment is a Renishaw RamaScope spectrometer, model 2000, fitted with a laser source with a wavelength, λ , of 514 nm and an integrated microscope that allows the observation of the specimen surface. The residual stress measurement locations and traverses are shown in Fig. 2.

Two Raman spectra for the TC are shown in Fig. 3. Peak 6 (640 cm^{-1}), which has pronounced intensity in this case, was fitted by mixed Gaussian and Lorentzian method and chosen to calculate the stress by a conversion factor, $5.60\text{ cm}^{-1}\text{ GPa}^{-1}$, obtained previously using diamond anvil cell on this set of APS-TBC.²⁵ The other peaks located within the lower frequency range, as shown in Fig. 3, have significantly reduced intensity at many locations. This scenario limits the application of the peak centred at 465 cm^{-1} for calibration of APS-YSZ, as proposed by Limarga et al.²⁶ based on bulk dense pure YSZ material. Hence,

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