International Journal of Fatigue 70 (2015) 186–195

Contents lists available at ScienceDirect

International Journal of Fatigue

journal homepage: www.elsevier.com/locate/ijfatigue

Bending fatigue failure of atmospheric-plasma-sprayed CoNiCrAlY + YSZ thermal barrier coatings

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ARTICLE INFO

Article history:
Received 8 June 2014
Received in revised form 17 September 2014
Accepted 21 September 2014

Keywords:
Thermal barrier coatings
Fatigue
Crack path
FEM

ABSTRACT

Bending fatigue failure of conventional atmospheric-plasma-sprayed CoNiCrAlY + ZrO_2–8 wt.% Y_2O_3 thermal barrier coatings with/without the thermally grown oxide layer generated between the bond coat and the top coat was experimentally studied at room temperature. Microscopical and profilometrical characterization of as-received and fractured specimens and a simplified finite element study of cooling thermal stresses show that the same fatigue strength of both the as-coated and the oxidized specimens (i.e. its insensitivity to the presence of the thermally grown oxide) is most likely caused by a preferential through-the-thickness cracking of the thermally grown oxide layer. Moreover, the bond-coat/substrate interface is identified as the weakest part of the studied thermal barrier system under both low and high crack growth rates.

1. Introduction

Thermal barrier coatings (TBCs) are advanced material systems used in propulsion and power generation applications for protecting key engineering components against high-temperature oxidation and hot-corrosion in chemically aggressive operational environments [1–6]. In the as-coated state, the TBCs comprise at least two following layers: a ceramic top coat that shields the underlying material from heat and a metallic bond coat the main purpose of which is to protect a substrate against high-temperature degradation and improve adherence of the top coat. The bond coat is very often of *M*CrAlY type, where *M* stands either for Co, Ni, or combination of both, and it is deposited by some of the plasma spraying methods. The yttria stabilized zirconia (YSZ) is a current industry standard for the top coat as it has low thermal conductivity, a relatively high coefficient of thermal expansion and toughness and high melting temperature [4–6]. The top coat is deposited either by plasma spraying or by the electron beam vapor deposition method resulting, respectively, in a lamellar or a columnar microstructure. Although the design of thermal barriers and their working conditions differ widely, it is agreed that a structural integrity of any such system is, if there is no substantial erosion, ultimately limited by failure related to a thermally-grown oxide (TGO) layer that is gradually formed between the top coat and the bond coat during in-service high-temperature exposure [1–3,7,8]. The growth of the TGO, which protects the substrate from oxidation and hot-corrosion and thus is an integral part of coating's functionality, induces significant local stresses which inevitably lead to spallation of the top coat and subsequent rapid failure of unprotected remains of the system. In general, such degradation of TBCs is an extremely complicated process that is still not well understood. Depending on composition and working conditions of a particular system, various interrelated processes such as inter-diffusion of elements including gradual depletion of aluminum and other oxide formers from the bond coat and segregation of impurities at the top-coat/bond-coat interface, the formation of undesirable oxides and spinels along the protective α-alumina TGO layer, creep of the bond coat and the TGO, or sintering, microcracking and phase transformations of the ceramic top coat which cause changes in its thermomechanical properties might play more or less important role [3,7–13].

The stability and degradation of TBCs under a broad range of mechanical, thermal and thermomechanical loading regimes have been investigated by many researches, e.g. [9–11,13–18], and it is doubtless one of the most extensively studied topics in material sciences and engineering. Due to an enormous variety, however,

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http://dx.doi.org/10.1016/j.ijfatigue.2014.09.009
0142-1123/

the results are often related to a very specific coating system, employed deposition conditions and a particular loading regime, on all of which only limited information is often provided simultaneously. In the present work, room-temperature bending fatigue failure of conventional atmospheric-plasma-sprayed CoNiCrAlY + ZrO_2–8 wt.% Y_2O_3 thermal barrier coatings, both without/with the TGO layer formed between the metallic bond coat and the ceramic top coat, is investigated by means of various techniques. The *S–N* data obtained from fatigue experiments are studied in combination with microscopical and profilometrical measurements of as-received and fractured specimens in order to assess the role of individual layers and interfaces in the fatigue crack initiation and propagation stages. Furthermore, the acquired profilometrical data are employed in a simplified finite element model in order to estimate the distribution of thermal residual stresses upon cooling so as to rationalize the observed complex cracking mode of the geometrically irregular TGO layer. Atmospheric-plasma-sprayed CoNiCrAlY + YSZ and the bending fatigue used in this study represent a conventional TBC system and a relatively simple loading with well-defined fatigue crack initiation sites and a stress gradient similar to what may be often expected during operation. The results are of interest for rotating or moving components that might experience relatively high bending stresses due to misalignment and generally for components thermally cycled with a short dwell time at high temperature in which case the failure is expected to occur predominantly due to cracks initiated in the ceramic top coat [19] that subsequently propagate under thermal strains resulting from high thermal expansion mismatch following the shutdown of the engine. The temperatures at which the highest stresses are to be experienced by defects and cracks are located near the end of transient cooling [7] and as these temperatures are less than the ductile to brittle transition temperature of the bond coat (~650 °C) it is reasonable to assume that the micromechanisms active during the crack propagation are affinitive to that at the room temperature, similarly to what was already observed for specimens with protective diffusion coatings [20].

2. Materials, fatigue experiments and used techniques

2.1. Specimens

Specimens (Fig. 1) were manufactured from cast rods of nickel-base superalloy Inconel 713LC (Ni 12.08Cr 5.91Al 4.58Mo 2.02Nb 0.75Ti 0.1Fe 0.1Zr 0.05Co 0.05Cu 0.05Mn 0.05Si 0.05Ta 0.05C 0.01B wt.%) by the company PBS Velká Bíteš, a.s., Czech Republic. Conventional CoNiCrAlY + ZrO_2–8 wt.%Y_2O_3 thermal barrier coatings were deposited by atmospheric plasma spraying (APS) at S.A.M. – Metallizing Company, Ltd., Czech Republic. Before spraying, all specimens were grit-blasted by 22 grit mesh aluminum oxide particles by using Oteco OT 1500 facility equipped with a 9.5 mm diameter nozzle. Grit-blasting was conducted at a pressure of 5 bar using a blasting distance of 100 mm and an incidence angle of 90°. This operation, which is employed to achieve a better mechanical interlocking with the bond coat, induced a compressive residual stresses in the substrate and caused a surface roughening resulting in an increase of measured specimen's diameter of ~20 μm in comparison with the as-machined specimens.

Commercially available powders AMDRY 995 (Co 32Ni 21Cr 8Al 0.5Y wt.%) and GTV 40.23.1 (Zr 8Y wt.%) were deposited by using the Sulzer Metco GTV MF-P1000 equipment and the F4 MB-XL plasma gun during a single gun pass. Deposition parameters were set in compliance with powder suppliers' guidelines and are shown in Table 1. The average thickness of the bond coat and the top coat was approximately 210 μm and 190 μm respectively, with the local thickness varying on the order of tens microns. After plasma spraying, several of coated specimens were isothermally oxidized in air at 1050 °C for 200 h in a tubular furnace Heraeus using the heating rate of 7 °C/min and cooled to ambient air temperature after the removal from the furnace.

The microstructure was observed using an optical microscope Olympus GX-51 and a scanning electron microscope Philips XL30 equipped with an energy-dispersive spectrometer. The substrate was typical of a cast nickel-base superalloy with the average dentritic grain size of approximately 1 mm. Its microstructure was composed of a γ matrix, a strengthening γ' phase and of primary and secondary carbides with the highest population found at the interdentritic regions. As a result of APS deposition, both the top coat and the bond coat consisted of flattened "splats" formed when molten particles solidified after impacting onto surface, closed pores, microcracks generated in order to relieve the tensile quenching stresses induced due to a rapid contraction of the molten droplets and a small fraction of partially unmelted powder particles resulting from variation in particle temperature and velocity during plasma spraying. The metallic bond coat was more compact and contained numerous segmented non-uniform Al_2O_3 veins located between the splats but also at the top-coat/bond-coat interface as a result of partial oxidation of metallic powder and the bond coat during the coating deposition [21].

In the course of isothermal oxidation, the medium-thick TGO scale was grown between the bond coat and the top coat. Moreover, both the coatings and the substrate underwent microstructural changes. These included a partial closing of pores in both coatings and a mild increase of density of microcracks within the top coat, see Fig. 2. The most pronounced changes in the substrate were observed in a vicinity of Al_2O_3 grit particles that were occasionally locked in a surface during grit-blasting and caused a localized transformation of initial γ'/γ' microstructure due to aluminum diffusion into substrate, Fig. 3a and Table 2. The TGO scale was composed of ~2.5 μm thick continuous alumina layer and a more

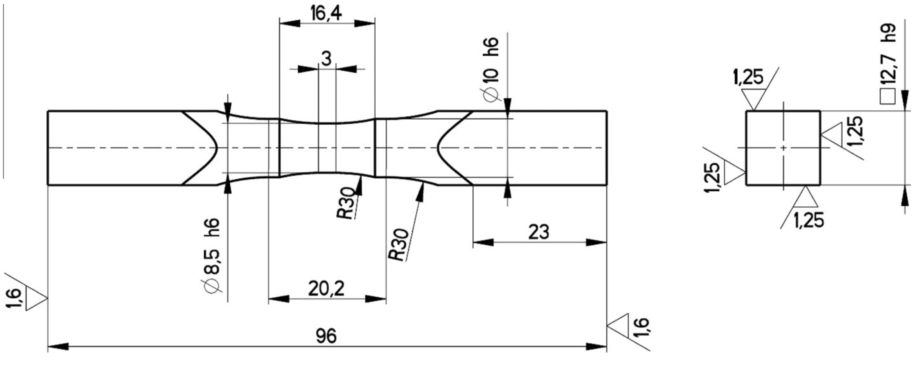

Fig. 1. Specimen geometry.