Oxidation and interfacial fracture behaviour of NiCrAlY/Al_2O_3 coatings on an orthorhombic-Ti_2AlNb alloy

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1. Introduction

Titanium aluminide alloys have been considered as promising candidates for applications in aircraft engines and land-based engine components due to their low density and attractive mechanical properties [1]. In titanium aluminide alloys, orthorhombic Ti_2AlNb alloys exhibit better thermo-mechanical processing (TMP) and creep resistance as well as higher ductility and fracture toughness than γ-TiAl and γ-TiAl based alloys [2,3]. However, the disadvantages of the lack of high-temperature oxidation resistance above 700 °C and severe interstitial embrittlement due to the dissolution of nitrogen and oxygen in the subsurface above 550 °C impede the exploitation of Ti_2AlNb alloys as structural components at elevated temperatures [4–6]. It follows that an improvement in the resistance to oxidation and interstitial embrittlement is a challenge for developing orthorhombic Ti_2AlNb alloys for use at elevated temperatures. In order to solve these problems, the use of surface coatings seems an efficient way.

MCrAlY (M = Ni, Co, or Ni + Co) coatings have been widely used on superalloy gas-turbine components as protective coatings at high-temperature [7–12]. When MCrAlY coatings were applied on O-Ti_2AlNb and other Ti–Al alloys, serious interdiffusion and interfacial reactions occurred [13–15]. In previous studies [16,17], we have investigated the effect of an Al_2O_3/Al multi-layer and Al–Al_2O_3 cermet films deposited by arc ion plating as diffusion barriers between the MCrAlY coating and the O-Ti_2AlNb alloy. Excellent inhibiting properties of the diffusion barriers and a much improved oxidation resistance of the multi-layer coatings were observed in both isothermal and cyclic oxidation tests. However, some important factors, such as the thickness and the mechanical properties of the diffusion barriers, have not been investigated. Also in arc ion plated Al_2O_3 diffusion barriers, contamination of the macro-particles may influence the inhibiting properties.

Relatively few works have been reported on the mechanical properties and interfacial fracture behaviour of coated specimens with diffusion barriers. Good interfacial stability under mechanical or thermal loading is one of the important requirements for the coated components to be in service, especially for multilayer coatings. Three-point or four-point bend tests are an effective method to investigate the interfacial fracture behaviour of multi-layer coated systems [18–22].

In this study, we have investigated the oxidation and interfacial fracture behaviour of Al_2O_3 films as diffusion barriers between NiCrAlY and orthorhombic Ti_2AlNb alloy. By varying the thickness of the Al_2O_3 interlayer, we expected that not only interdiffusion between the coating and substrate could be well suppressed but also the mechanical properties of the coating–substrate interface would be less influenced. To avoid the influence of macro-particles, we synthesized the Al_2O_3 diffusion barriers using a filtered arc ion plating technique. The interfacial strengths of the coating systems before and after thermally exposure were measured by a three-point bend test. The interfacial fracture behaviour of the multilayer coated systems is also discussed.
2. Experimental details

2.1. Specimen preparation

An orthorhombic Ti2AlNb alloy with nominal composition of Ti–22Al–26Nb (at.%) was prepared in the form of a cast ingot by the consumable arc melting method. The ingot was cut into coupons of size 10 × 15 × 2 mm³. After grinding to 800-grit, the coupons were grit-blasted (200-mesh glass ball) in a wet atmosphere, ultrasonically cleaned in ethanol and dried in cold air. The coupons were then loaded into an arc ion plating facility to deposit the Al2O3 diffusion barriers and the NiCrAlY coating. Table 1 shows the deposition conditions of the Al2O3 interlayer used in this study. After deposition, the coated specimens with and without diffusion barriers were heated in a vacuum-annealing furnace at a heating rate of 5 °C/min to 600 °C and annealed for 20 h, followed by annealing at 800 °C for 5 h.

Intermittent isothermal oxidation tests were conducted in static air at 800 °C in alumina crucibles placed in a muffle furnace. The specimens were taken out of the furnace at various intervals for mass measurement using a micro-balance with a sensitivity of 10⁻⁵ g. The total mass of a specimen together with the crucible was recorded. During each test, three parallel samples were tested to ensure experimental accuracy and reproducibility.

The phase structures of the as-deposited Al2O3 coatings were examined by a transmission electron microscopy (TEM, TECNAI G20). The oxidized specimens were analyzed using an X-ray diffraction (XRD, Cu Kα D/max-2200 VPC) from the specimen surfaces. The scanning range was 20°–85°, and the scanning speed was 0.02°/s at a step of 0.02°. Cross-sectional samples were prepared by electro-plating the oxidized specimens with a nickel coating, mounting them in epoxy, followed by grinding/polishing. The microstructure of the duplex coatings after oxidation was observed on cross-sectional samples using a scanning electron microscope (SEM Hitachi S-3400N) equipped with an energy-dispersive X-ray spectrometer (EDS). To investigate the interdiffusion behaviour in the coating-substrate system, the chemical compositions of both the coatings and the substrates near the coating-substrate interface were analyzed by EDS. In order to obtain information on the average chemical composition, the EDS analysis was conducted by carrying out window-selection measurements. Each window used in the measurements was about 40 × 4 μm².

2.2. Three-point bend tests

The three-point bend tests were performed using a servo-hydraulic testing machine (MTS A/T). Fig. 1 shows the experimental setup used for the three-point bend tests. The length of the span was 16 mm. The samples were tested in a configuration which placed the coatings in tension. The tests were performed in air at room temperature at a constant cross-head speed of 50 μm/min. Before a test, the samples were polished with 0.5 μm diamond paste to remove the coatings and the oxides on the bottom surfaces and lateral sides. During a test, when the load dropped, an in situ observation with an optical microscope was carried out to decide when to terminate the test. No less than three samples were tested for each load–displacement curve to ensure experimental accuracy. After the tests, the specimens were examined by scanning electron microscopy (SEM).

3. Results and discussion

3.1. Oxidation and interdiffusion behaviour of NiCrAlY coatings with and without Al2O3 diffusion barriers on Ti2AlNb alloys

By controlling the oxygen flux rate and substrate bias voltage (Table 1), dense and stoichiometric Al2O3 films were synthesized by a filtered arc ion plating technique. In the XPS spectra (not shown here), only the Al 2p peak at 74.7 eV and the O 1s peak at 530.6 eV corresponding to the Al2O3 phase were identified. The TEM image shown in Fig. 2(a) as well as the selected area electronic diffraction (SAED) image shown in the inset revealed that the as-deposited Al2O3 film had an amorphous microstructure. By varying the deposition time, the Al2O3 films with different thicknesses were fabricated. For the convenience of discussion, the films are termed according to the thickness of Al2O3 interlayer as: (1) 1Al2O3 (~1 μm); (2) 3Al2O3 (~3 μm); (3) 5Al2O3 (~5 μm). Fig. 2(b) shows a cross-sectional TEM image of the as-deposited NiCrAlY/3Al2O3/Ti2AlNb specimen. A dense and adherent Al2O3 interlayer with a thickness of about 3 μm was fabricated between the NiCrAlY coating and the Ti2AlNb substrate. Such a structure was designed so that the NiCrAlY coating (33–37 μm) can provide oxidation protection and the Al2O3 film can block interdiffusion between the coating and the substrate. Fig. 2(c) and (d) displays cross-sectional SEM images of vacuum-annealed coated specimens with and without the Al2O3 diffusion barrier. For the NiCrAlY/Ti2AlNb specimen, an apparent interdiffusion zone was detected (Fig. 2(c)). For the NiCrAlY/3Al2O3/Ti2AlNb specimen, very slight interdiffusion was observed (Fig. 2(d)), which means that the coating system was chemically stable by addition of a 3Al2O3 interlayer. Fig. 2(e) shows the XRD patterns of vacuum-annealed coated specimens with and without the 3Al2O3 diffusion barrier and an as-deposited NiCrAlY coating. The coatings consisted mainly of γ-Ni and γ’-Ni3Al phases. The as-deposited coating possessed wide diffraction peaks, indicating amorphous structure or nano-grains. After annealing, the width of the diffraction peaks decreased, while β-NiAl and α-Cr phases appeared. The diffraction peaks of the β-NiAl and α-Cr phases in the annealed coated specimen without Al2O3 diffusion barrier were apparently stronger than those in the specimen with 3Al2O3 diffusion barrier. This can be attributed to the relative enrichment of Cr and Al by inward diffusion of Ni from the coating, which resulted in the formation of the β-NiAl and α-Cr phases [23].

Table 1
The deposition parameters of the Al2O3 diffusion barriers.

<table>
<thead>
<tr>
<th>Diffusion barrier</th>
<th>Gas-flow rate (sccm)</th>
<th>Chamber pressure (Pa)</th>
<th>Bias voltage (V)</th>
<th>Filtered voltage (V)</th>
<th>Deposition time (min)</th>
<th>Thickness (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1Al2O3</td>
<td>80</td>
<td>10</td>
<td>0.5</td>
<td>−300</td>
<td>30</td>
<td>−1</td>
</tr>
<tr>
<td>3Al2O3</td>
<td>90</td>
<td>10</td>
<td>3</td>
<td>−3</td>
<td>90</td>
<td>−3</td>
</tr>
<tr>
<td>5Al2O3</td>
<td>150</td>
<td>10</td>
<td>5</td>
<td>−5</td>
<td>150</td>
<td>−5</td>
</tr>
</tbody>
</table>
Fig. 2. (a) A TEM image of the as-deposited Al₂O₃ coating, (b)-(d) cross-sectional SEM images, and (e) XRD patterns of the coated specimens with and without a diffusion barrier. (b) As-deposited NiCrAlY/3Al₂O₃/Ti₂AlNb; (c) NiCrAlY/Ti₂AlNb after vacuum annealing; (d) NiCrAlY/3Al₂O₃/Ti₂AlNb after vacuum annealing.

Fig. 3. (a) Mass gain data and (b) XRD patterns of the coated Ti₂AlNb specimens with and without Al₂O₃ diffusion barrier after isothermal oxidation for 500 h at 800 °C. The error bars represent one standard deviation from the mean measurements. Where not visible, the error bar is within the size of symbol.

Fig. 3(a) shows results for the interrupted isothermal oxidation of the coated Ti₂AlNb specimens with and without an Al₂O₃ interlayer at 800 °C. The results indicate that the coatings with diffusion barriers exhibit better oxidation resistance as compared with the coatings without the diffusion barrier. For the NiCrAlY/Ti₂AlNb specimens, the slope of the oxidation curve became higher after
100 h, which means that the oxidation rate increased. For the coated specimens with Al₂O₃ diffusion barriers of different thicknesses, oxidation resistance was at a similar level in the first 300 h. After 300 h, the NiCrAlY/1Al₂O₃/Ti₂AlNb specimens showed a slightly larger mass gain. Both NiCrAlY/3Al₂O₃/Ti₂AlNb and NiCrAlY/5Al₂O₃/Ti₂AlNb specimens exhibited good oxidation resistance in the whole oxidation experiment. During interrupted isothermal oxidation, no spallation occurred for all the coated specimens.

Fig. 3 shows the XRD patterns of the coated specimens with and without Al₂O₃ diffusion barriers after isothermal oxidation for 500 h at 800 °C. The oxidation products on the NiCrAlY/Ti₂AlNb specimen surface consisted of TiO₂ (Rutile), Al₂O₃ and Cr₂O₃. Meanwhile, the peaks of γ and γ' disappeared, which means that the coating had been largely exhausted by serious oxidation and interdiffusion. On the coated specimen with the 1Al₂O₃ diffusion barrier, the major presence of Al₂O₃ and Cr₂O₃ and the minor presence of TiO₂ (Rutile) were detected. Only an α-Al₂O₃ phase was identified as the oxidation product on the coated specimens with the 3Al₂O₃ and 5Al₂O₃ diffusion barriers. For all specimens with diffusion barriers, the main constituents of the metallic phase, such as γ and γ', remained. The NiCrAlY coatings had been consumed only to a limited extent due to inclusion of the Al₂O₃ diffusion barriers.

Fig. 4 shows the surface SE-mode SEM observations of the coated specimens with and without the Al₂O₃ diffusion barriers after isothermal oxidation for 500 h at 800 °C. Generally, the high temperature oxidation resistance of alloys depends on the formation of stable oxides of Al and Cr on the surface, because both metal and oxygen atoms have low diffusion coefficients in these oxides. Al₂O₃ had better protective effect than Cr₂O₃, while TiO₂ was non-protective oxide [14,24]. The polyhedral Ti-rich oxides in Al₂O₃ scale destroyed the compactness and continuity of the thermally grown surface oxide layer, which resulted in deterioration of oxidation properties. Therefore, the NiCrAlY/Ti₂AlNb specimen exhibited worst oxidation resistance due to the formation of TiO₂ oxides while the NiCrAlY/3Al₂O₃/Ti₂AlNb and NiCrAlY/5Al₂O₃/Ti₂AlNb specimens exhibited much better oxidation resistance due to the formation of pure Al₂O₃ scales (Fig. 3(a)). For the NiCrAlY/1Al₂O₃/Ti₂AlNb specimen, the thickness of the 1Al₂O₃ diffusion barrier was too thin to inhibit much Ti from diffusing into the NiCrAlY coating from the O-Ti₂AlNb substrate, which will be described later. Therefore, a slight increase of mass gain occurred in the last 200 h (Fig. 3(a)).
apparently suppressed, with no holes being found at the interfaces. The coating microstructures were influenced by the thickness of the Al2O3 interlayer. First, the proportion of the Cr phase in the NiCrAlY coatings increased with decreasing thickness of the Al2O3 diffusion barriers. The reason for this is that the content of Ni in the coatings decreases with decreasing thickness of the Al2O3 diffusion barrier due to a less inhibited interdiffusion. Thus, more Cr phase separates in the NiCrAlY coatings. Second, the interdiffusion affected substrate (IAS) zone in the substrate becomes thicker with decreasing thickness of the Al2O3 diffusion barriers. In this zone, Nb enrichment occurred and much Ni was found according to EDS. According to our former studies [16,17], new intermetallic phases, such as Ni2(Ti, Nb)Al and r-Nb2Al, might form. Tang et al. [25] also found that Ni2TiAl phase formed at the interface of MCrAlY/TiAl. Finally, thicker Al2O3 diffusion barriers result in more stable coatings and more suppressed interdiffusion.

Fig. 6(a)–(d) shows the distribution of elements on the cross-sections of the coated specimens with and without Al2O3 diffusion barriers after isothermal oxidation for 500 h at 800 °C.

For discussing and comparing the effects of diffusion barriers on inhibiting the interdiffusion of elements, in a previous paper [17], we proposed a coefficient of diffusion inhibition \( \theta \) and a factor for reaction hindering \( \varphi \), and \( \theta, \varphi = 1 - \frac{J_i}{J_i^0} \), where \( J_i \) denotes diffusion flux of element \( i \) migrating though the diffusion barrier; \( J_i^0 \) is diffusion flux of element \( i \) diffusing through the coating–substrate interface in the absence of a diffusion barrier. \( R \) denotes the reaction rate for the duplex coating with a diffusion barrier and \( R_0 \) is the reaction rate in the absence of a diffusion barrier. By the calculation method described in the earlier paper [17], values for \( \theta \) and \( \varphi \) were obtained and are shown in Fig. 6(e) and (f), respectively. According to Fig. 6, the order of ability of the inhibiting elements interdiffusion and the hindering interfacial reaction for different thicknesses of diffusion barriers was 5Al2O3 > 3Al2O3 > 1Al2O3.

3.2. Interfacial fracture behaviour during three-point bend tests

3.2.1. Load–displacement curves from the three-point bend test

Table 2 lists the data of displacements and loads of the coated specimens with and without Al2O3 diffusion barriers during a three-point bend test. Fig. 7 shows the corresponding typical load–displacement curves. The load initially increases linearly with the applied displacement. Once a critical load is reached, a sharp drop of load occurs in the load–displacement curve. At this time, a crack propagates and the rupture of the overlay coatings takes place along the coating–substrate interface, or the whole specimen fractures along some cracks.

Different load–displacement curves were observed for the as-deposited, vacuum annealed and oxidized NiCrAlY/Ti2AlNb specimens (Fig. 7(a)). For the as-deposited NiCrAlY/Ti2AlNb specimen, the maximum in the load–displacement curve was observed at 301 N, corresponding to the rupture and peeling of the NiCrAlY coating. The annealed and oxidized specimens fractured suddenly at 309 N and 270 N during the bend test, respectively.

For the coated specimens with Al2O3 diffusion barriers, the values of the maxima in the load–displacements curves are different depending on the thickness of the Al2O3 diffusion barriers and thermal history (Fig. 7(b)–(d)). Both as-deposited and vacuum annealed NiCrAlY/1Al2O3/Ti2AlNb samples show a linear elastic load–displacement relationship, with critical loads of 165 N and 219 N (Fig. 7(b)), respectively. For the oxidized samples, no sudden macro-fracture occurred. Instead, the rupture of the overlay coating occurred in the transition stage during the bend test, with a critical load for bending at 393 N. The load–displacement curve of the NiCrAlY/3Al2O3/Ti2AlNb specimen shown in Fig. 7(c) is very
similar to that shown in Fig. 7(b). The fracture of the as-deposited and vacuum heat-treated specimens occurred in the elastic stage, with the critical loads being at 121 N and 197 N (Fig. 7(c)), respectively, while a critical load of 313 N was observed for the oxidized samples. These results reveal that long-term oxidation does not reduce, but rather improves the interfacial adhesion strength and the resistance to crack propagation at the coating–substrate interfaces. For the NiCrAlY/5Al2O3/Ti2AlNb specimens, as shown in Fig. 7(d), the as-deposited, vacuum-annealed and oxidized samples show the linear elastic load–displacement relationship. The critical loads are lower than those shown in Fig. 7(a)–(c).

Since the coated surface was placed on the bottom side during the bend tests, the coating section of the specimen was only subjected to tensile-shear forces. Consequently, the main failure in the coatings occurs owing to tensile-shear deformation, particularly at the coating–substrate interface. Because the displacements...
are very small, the tensile force at the coating–substrate interface can be approximated to be the same as the applied load. Based on the above analysis, a higher critical load for delamination means a better interfacial adhesion of the coating–substrate.

Two factors related to the coating–substrate interdiffusion affect the interfacial fracture behaviour of the multi-layer coating systems: (1) interdiffusion between coating and substrate improves the interfacial adhesion by a diffusion bond mechanism [15,21] and (2) the formation of a brittle intermetallic phase due to coating–substrate interaction deteriorates the mechanical properties [26]. Combination of these two factors influences the delamination strength at the coating–substrate interface. Therefore, as illustrated in Table 2, the critical load decreases after thermal exposure for the NiCrAlY/Ti2AlNb specimens and increases after thermal exposure for the NiCrAlY/1Al2O3/Ti2AlNb and NiCrAlY/3Al2O3/Ti2AlNb specimens. For the NiCrAlY/5Al2O3/Ti2AlNb specimens, inclusion of a thicker brittle Al2O3 ceramic layer may lead to deterioration in the interfacial fracture properties. In addition, the coefficients of thermal expansion (CTE) of NiCrAlY, Al2O3 and O-Ti2AlNb are 17 × 10^-6 K^-1 [27], 8.3 × 10^-6 K^-1 [28] and 8.8–10.5 × 10^-6 K^-1 [2], respectively. The CTE mismatch might cause extensive stresses at the interfaces during oxidation, which affects the interface adhesive strength.

3.2.2. Crack propagation in the three-point bend test

After the three-point bend tests, the cracked surfaces of the specimens were observed by SEM to investigate the crack propagations. Figs. 8 and 9 show the crack propagations in the as-deposited, vacuum-annealed and oxidized NiCrAlY/Ti2AlNb specimens, respectively. For the as-deposited specimen, a microcrack is formed and propagates from the coating surface (opposite to the loading indenter point) through the entire coating thickness to the NiCrAlY/Ti2AlNb interface (Fig. 8(a)), which would initiate the fracture and cause delamination of the coating. In Fig. 8(b) and (c), the fracture and delamination of the NiCrAlY coating can be observed. No rupture of the Ti2AlNb substrate was detected. Instead, the cracks only reached and extended along the coating–substrate interface. From the enlarged micrograph of the crack tip zone in Fig. 8(d), it can be seen that the interface fracture may proceed by the nucleation, growth and coalescence of microcracks, occurring by plastic flow in both the NiCrAlY coating and the Ti2AlNb substrate adjacent to the interface.

For the vacuum annealed NiCrAlY/Ti2AlNb specimen, the crack propagation is similar to that in the as-deposited coating. The main vertical crack extends along the thickness direction. However, when the crack reaches the NiCrAlY/Ti2AlNb interface, unlike the crack propagation in the as-deposited specimens, it penetrates this interface and continues to advance in the interdiffusion affected substrate (IAS) zone in the same direction (Fig. 9(a)). When the bend test continues, this crack extends along the same direction and breaks off the substrate (Fig. 9(b)). A magnified SEM image of a microcrack, propagating along the IAS zone/Ti2AlNb interface, is shown in Fig. 9(c). For the oxidized specimens (Fig. 9(d)), fracture behaviour during bend test is basically similar to that of the annealed specimens. According to our previous study [16,17], the IAS zone consisted mainly of the ternary Ni2(Ti, Nb)Al phase, which is a brittle intermetallic phase. Cracks often form and propagate within these phases [26].

Fig. 10 shows SEM micrographs of the cross-sections of the vacuum heat-treated NiCrAlY/3Al2O3/Ti2AlNb specimens after the bend test. A crack is observed to initiate from the top surface and reach the Al2O3/Ti2AlNb interface (Fig. 10(a)). As the bending moment increases, the crack deflects into the Al2O3/Ti2AlNb interface and propagates along it, causing delamination of the coating.
The crack spacing is large and delamination of the coating (peeling off from the substrate surface) occurs along the Al2O3–Ti2AlNb interface (Fig. 10(c)), which means that this interface is the weakest. From the high-magnification SEM image micro-cracks with small spacing along the interface of the Al2O3/Ti2AlNb can be observed in the crack tip zone, which indicates that the sliding and splitting deformation occurred at this interface.

In the case of the isothermally oxidized specimens, the cracking processes are basically the same as those in the heated vacuum specimens. However, it is noteworthy that some microcracks occurred within the substrate near the interface (Fig. 11(a)), which means a higher interfacial adhesive strength of the coating/substrate. Near the crack tip (Fig. 11(b)), the microcrack path trajectory in the Al2O3 barrier layer often shows branch cracks.
These branch cracks grow across the Al₂O₃ barrier layer in some zones, and along the Al₂O₃/Ti₂AlNb interface elsewhere. With regard to the crack propagation path, it is interesting to note that the main crack is accompanied by frequent coalescence of the branch cracks. Consequently, the interfacial adhesive strength is improved for the NiCrAlY/3Al₂O₃/Ti₂AlNb specimens after oxidation.

In the three-point bend tests, the cracks nucleate from the weak interface, the brittle precipitates or the defects at the interface or on the coating surfaces and propagate in the brittle phases or along the interfaces [21,26]. Therefore, the interfacial adhesive strength and the mechanical properties of the layers affected the damage behaviour of the whole system. In the as-deposited NiCrAlY/Ti₂AlNb specimen, a low interfacial adhesion and brittle overlayer can be expected, which resulted in the fracture and delamination of the NiCrAlY coating (Fig. 8(a) and (b)). After vacuum annealing or oxidation, the interfacial adhesion is much improved. However, brittle intermetallic phase formed in the IAS zone, which resulted in the cracks extending through the IAS zone and breaking off the substrate (Fig. 9(a) and (b)). For the NiCrAlY/3Al₂O₃/Ti₂AlNb specimen, the Al₂O₃ diffusion barrier is brittle. In the vacuum-annealed status, the interface was also a weak site because of low interfacial strength due to not enough interdiffusion. Therefore, the cracks deflect into the Al₂O₃/Ti₂AlNb interface and propagate along it, causing delamination of the coating (Fig. 10(a)–(d)). After a long-time oxidation, both interfacial bonding and the overlayer ductility are improved. In this situation, the brittle Al₂O₃ diffusion barrier is the weakest site. Therefore, the microcracks form and propagate in the Al₂O₃ layer and the adjacent zones (Fig. 11(a) and (b)).

4. Conclusions

(1) Without a diffusion barrier, interdiffusion between the NiCrAlY coating and the Ti₂AlNb substrate is very serious and results in fast coating degradation and serious oxidation.

(2) With Al₂O₃ diffusion barriers, interdiffusion and interfacial reaction are effectively suppressed. Improved oxidation resistance of the multi-layer coatings is observed. A thicker diffusion barrier results in better inhibiting properties.

Fig. 10. Typical cross-sectional SEM micrographs of the vacuum-heated NiCrAlY/3Al₂O₃/Ti₂AlNb specimens after the bend test. (a) Initiation of a crack in the coating; (b) extension of the crack along the interface; (c) enlarged image of (b); (d) the crack tip zone.

Fig. 11. Typical cross-sectional SEM micrographs of the oxidized NiCrAlY/3Al₂O₃/Ti₂AlNb specimens after the bend test. (a) Extension of the crack along the interface; (b) the crack tip zone.
(3) For the coated specimens with and without a diffusion barrier, the observed mechanical behaviour is different during a three-point bend test. Appropriate interdiffusion improves interfacial adhesion. However, serious interdiffusion deteriorates the mechanical properties due to the formation of brittle intermetallic phase. These two factors influence the critical load during three-point bend tests.

(4) The different damage behaviour between the coated specimens with and without a diffusion barrier is attributed to the different interfacial strength and formation of brittle phase. For the NiCrAlY/3Al2O3/Ti2AlNb specimens, after vacuum annealing, the cracks are deflected into the interface and propagate along the interface of the Al2O3/Ti2AlNb. After oxidation, both interfacial bonding and the overlay ductility are improved. The propagation of the main crack is accompanied by the coalescence of branch cracks.

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References


