INTRODUCTION

Hot-section gas-turbine components typically require some form of coating for oxidation and corrosion protection. These coatings are generally either aluminate coatings or MCrAlY overlay coatings where M represents nickel, cobalt, or a combination of these two elements. Both types of coating are protective as a result of the selective oxidation of aluminum to form an external, continuous Al$_2$O$_3$ scale. The coatings act as a reservoir of aluminum since the aluminum content of the coating is always greater than that of the substrate. In service environments, coatings degrade by several mechanisms including oxidation and hot corrosion, interdiffusion of the coating and substrate, foreign object damage, and erosion (ref. 1). Coating/substrate interdiffusion involves not only the loss of the aluminum from the coating into the substrate but also the diffusion of less-desirable elements to the coating surface where they may oxidize and hinder or prohibit formation of the Al$_2$O$_3$ scale. In addition, cycling of a coated component results in cracking and spalling of the Al$_2$O$_3$ oxide scale, which further accelerates coating degradation.

Efficient use of coatings requires reliable and accurate predictions of the protective life of the coating. Currently, engine inspections and component replacements are often made on a conservative basis. As a result, there is a constant need to improve and develop the life-prediction capability of metallic coatings for use in various service environments. The purpose of the present work is aimed at development of an improved methodology for predicting metallic coating lives in an oxidizing environment and in a corrosive environment.

APPROACH

The present study combines both experimental studies and numerical modeling to predict coating life in an oxidizing environment. The experimental work provides both input to the numerical models and verification of the model predictions. The coatings being examined are an aluminate coating on Udimet 700 (U-700), a low-pressure plasma spray (LPPS) NiCoCrAlY overlay coating also on U-700, and bulk deposits of the LPPS NiCoCrAlY coating. The approach taken in this study is shown schematically in figure 1.

Experimental Testing

The experimental testing involves isothermal and cyclic furnace oxidation at 1050, 1100, and 1150 °C. In addition, Mach 0.3, cyclic burner rig testing of the aluminate and LPPS NiCoCrAlY coatings (125 and 625 μm thicknesses) on
U-700 is also being undertaken at 1100 °C. Isothermal oxidation of the coated U-700 and bulk coating yields the growth rate of the Al₂O₃/NiAl₂O₄ scales which form on these coatings. At the conclusion of the test it is often possible to measure the weight of oxide which spalls on cooling. The fraction of oxide that spalls is used as an input to the cyclic oxidation/spall model discussed below. Cyclic furnace and burner rig oxidation yield the weight changes of the coated specimens reflecting the oxide growth and spallation which occurs during thermal cycling. Analysis of the coating after an oxidation exposure includes x-ray diffraction and polarized light metallography of the retained surface oxides, observation of microstructural changes, scanning electron microscopy, and electron microprobe analysis to measure concentration/distance profiles across the coating and substrate.

### Numerical Modeling

Two computer models are being used to predict the oxidation-limited life of the metallic coatings. A cyclic oxidation/spall model (ref. 2) predicts the oxide growth and amount of oxide spallation that occurs during cyclic oxidation. The isothermal oxide growth rate and spall fraction (the ratio of the oxide which spalls on cooling to the total oxide present before cooling) are input to the spall model. The spall model predicts the weight change of a coated specimen undergoing cyclic oxidation, the rate of metal consumption, and the total weight of metal consumption during cyclic oxidation. A diffusion model (ref. 3) simulates the diffusional transport associated with both coating oxidation and coating/substrate interdiffusion. Diffusion coefficients and the rate of metal consumption predicted by the spall model are input to the diffusion model. The diffusion model predicts aluminum and chromium concentration/distance profiles in the coating and substrate and the time for which the coating is able to supply sufficient aluminum to continue forming an Al₂O₃ scale. The diffusion model therefore predicts coating failure when there is insufficient transport of aluminum to the oxide scale.

### CURRENT STATUS AND RESULTS

Isothermal and cyclic furnace oxidation testing is nearly complete. Cyclic furnace testing of the coated U-700 specimens are being carried out to failure of the coatings for the three test temperatures of 1050, 1100, and 1150 °C. For the purpose of this study, coating failure has been defined as the occurrence of one or more of the following: accelerated weight loss during cyclic oxidation; the presence of less-protective oxides (as NiO, Cr₂O₃, or NiCr₂O₄) on the coating surface; the massive spallation of the coating; or the internal oxidation of aluminum in the coating, indicating the presence of less-protective oxides on the surface.

Degradation of the aluminide coating has previously been discussed and is reviewed in figure 2. Generally, total depletion of the γ' and B phases occurs before coating failure.

The cyclic oxidation lives of the LPPS NiCoCrAIY coated specimens were less than expected. Figure 3 shows a comparison of the weight change of LPPS-coated specimens and the bulk LPPS coating. The coated U-700 specimens exhibit a significant weight loss after 500 1-hr cycles at 1100 °C. Visual examination
of the cylindrical specimens after cyclic oxidation revealed considerable oxidation and spallation of the coating, especially near the ends of the cylinders (fig. 4). The coating failure which occurred after cyclic oxidation can be compared with a fully intact, protective coating that has undergone isothermal oxidation (also shown in fig. 4). Microstructural examination of coating and substrate cross sections before cyclic oxidation exposure revealed very little porosity at the coating/interface (fig. 5, as-sprayed plus 4-hr argon anneal at 1080 °C); whereas examination of specimens after cyclic oxidation showed extensive porosity formation at the coating/substrate interface (fig. 5 to 7). Apparently, the porosity formed and grew with increasing coating/substrate interdiffusion and is probably the result of a vacancy flux imbalance (ref. 4) caused by unequal atomic transport across the coating/substrate interface. The coatings eventually delaminated and spalled near the cylinder ends, most probably due to a lack of contact with the substrate. Extensive oxidation then occurred at the coating/substrate interface via the interconnected porosity (fig. 5 to 7) causing rapid coating degradation and massive coating spallation. As expected, the higher the temperature, the more rapid the formation of the porosity and the shorter the time to coating failure.

There was poor agreement between the observed weight change and that predicted by the cyclic oxidation/spallation model for the LPPS NiCoCrAlY coated U-700. Two causes for this poor agreement were evident. First, the fraction of spalled oxide measured after the isothermal test (i.e., approximately 0.01 at 1100 °C) and input to the spallation model was an order of magnitude larger than expected for Al₂O₃ scales (ref. 2). The discrepancy between the spall fractions has not been completely resolved due to the difficulty in measuring the extremely small amount of Al₂O₃ that spalls after oxidation at 1050 to 1150 °C. Second, the spallation model assumes only external formation of Al₂O₃. Significant oxide formation within the porosity at the coating/substrate interface, massive coating spallation (especially at the coating ends), and formation of less-protective oxides were observed experimentally. To estimate the oxidation-limited coating life using the diffusion model, the spall fraction input to the spallation model was estimated so that the predicted weight change was similar to that for the LPPS coating on U-700 (fig. 8). The spall model was then used to predict the weight of aluminum consumed (fig. 9) and the rate of aluminum consumption (fig. 10), the latter being input to the diffusion model.

The aluminum concentration measured after 300 1-hr cycles at 1100 °C and the aluminum profiles predicted by the diffusion model are shown in figure 11. The diffusion model does not predict the γ' or θ phases remaining in the coating. The predicted concentration profile contains less aluminum in the coating and more in the substrate than measured in the test specimen. The rate of aluminum consumption predicted by the spallation model was probably too high due to the poor spalling resistance of the coated specimens. In addition, the diffusion coefficients input to the diffusion model (ref. 4) were measured on single-phase Ni-Cr-Al alloys and may not be appropriate for the complex, multiphase superalloys. Based on the cyclic oxidation behavior of the LPPS coating on U-700 at shorter times, the diffusion model predicts the life of the coating to be in excess of 1500 hr at 1100 °C. Experimentally, massive coating spallation was observed between 500 to 1000 1-hr cycles at 1100 °C. Obviously, the formation of the porosity and the oxidation within the pores significantly decreased coating life.
FUTURE WORK

Measurement of concentration/distance profiles in several of the coated specimens after various cyclic oxidation exposures will permit a more extensive verification of the predictive capability of the diffusion model. Continued burner rig testing should provide oxidized specimens which, although containing porosity at the coating/substrate interface, may eliminate oxide formation within the porosity and the resultant massive coating spallation. Premature coating failure may therefore be eliminated and permit an accurate test of the life-predictive ability of the diffusion model. A computer model is also under development to simulate $\gamma'$ and $\beta$ depletion during degradation of aluminide coatings. This aluminide diffusion model should be capable of predicting coating life. Measured concentration/distance profiles after cyclic oxidation of the aluminide coated U-700 will be compared with those predicted by the diffusion model to determine the accuracy and usefulness of the model. Predicted and measured coating lives will also be compared. It is anticipated that the conclusion of this work will result in an improve methodology for predicting the oxidation life of both overlay and aluminide coatings.

DUAL CYCLE ATTACK

An experimental study has recently been initiated to investigate the effect of aging in a corrosive environment (900 °C, 0.5 ppm Na) on the oxidation life of the two coatings discussed above in this paper. An attempt will be made to develop an empirical model to relate coating life to combined oxidation/hot corrosion cyclic exposure. The approach this study will take is schematically shown in figure 12. Burner rig testing in the oxidizing environment has been initiated.

REFERENCES


OBJECTIVE: TO DEVELOP AN IMPROVED METHODOLOGY FOR PREDICTING THE OXIDATION LIFE OF METALLIC COATINGS

EXPERIMENTAL

ALUMINIDE U-700

LPPS NiCoCrAlY U-700

BULK NiCoCrAlY

ISOTHERMAL OXIDATION
1050, 1100, 1150 °C
100 hr

CYCLIC OXIDATION
1050, 1100, 1150 °C
1-hr CYCLES RUN UNTIL FAILURE

MACH 0.3 BURNER RIG OXIDATION
1-hr CYCLES RUN UNTIL FAILURE; BLOCK TIME EFFECTS

ANALYTIC MODELING

SPALLING MODEL
PREDICT WEIGHT CHANGE FOR CYCLIC OXIDATION
PREDICT RATE OF METAL CONSUMPTION

DIFFUSION MODEL
PREDICT CONCENTRATION/DISTANCE PROFILES IN COATING/SUBSTRATE
PREDICT COATING LIFE

Figure 1
CYCLIC FURNACE OXIDATION OF ALUMINIDE COATED U-700

Figure 2

CYCLIC FURNACE OXIDATION OF LPPS NiCoGaAlY OVERLAY COATING

Figure 3
MACROSCOPIC COATING FAILURE

100 hr ISOTHERMAL AT 1150° C

1000 1-hr CYCLES AT 1100° C

100 1-hr CYCLES AT 1150° C

250 1-hr CYCLES AT 1150° C

Figure 4

MICROSCOPIC COATING FAILURE

AS-SPRAYED PLUS 4-hr ARGON ANNEAL AT 1080° C

THIRTY 1-hr CYCLES

100 1-hr CYCLES

250 1-hr CYCLES

Figure 5
MICROSCOPIC COATING FAILURE

1100°C

100 1-hr CYCLES

300 1-hr CYCLES

NO COATING REMAINING

2000 1/2-hr CYCLES

Figure 6

MICROSCOPIC COATING FAILURE

1050°C

1000 1-hr CYCLES

3000 1-hr CYCLES

3000 1-hr CYCLES

Figure 7

CS-84-3973

CS-84-3974
CYCLIC FURNACE OXIDATION OF LPPS NiCoCrAlY COATING ON U-700

Figure 8

ALUMINUM CONSUMPTION PREDICTED BY SPALLING MODEL

Figure 9
RATE OF AI CONSUMPTION PREDICTED BY SPALLING MODEL
LPPS COATING ON U-700; 1100°C; O₂ = 0.001

RATE OF ALUMINUM CONSUMPTION,
mg/cm²/sec

TIME, hr

Figure 10

ALUMINUM CONCENTRATION PREDICTED BY DIFFUSION MODEL
LPPS NiCoCrAlY COATING ON U-700; 300 1-hr CYCLES 1100°C

Al CONCENTRATION, AT %

U-700 SUBSTRATE

Y + β  Y + γ'  INITIAL AI CONTENT OF COATING

Y  γ  Y + γ'  INITIAL AI CONTENT OF SUBSTRATE

DISTANCE, μm

Figure 11
DUAL CYCLE ATTACK

OBJECTIVE: TO CHARACTERIZE THE EFFECT OF COMBINED OXIDATION AND HOT CORROSION CYCLIC EXPOSURE ON COATING LIFE

APPROACH

EXPERIMENTAL

ALUMINIDE
U-700

LPPS NiCoCrAlY U-700
(125 AND 625 μm THICKNESSES)

MACH 0.3 CYCLIC BURNER RIG

HOT CORROSION AGING
900 °C, 0.5 ppm NaCl
<1/3 OF HOT CORROSION LIFE

OXIDATION
1100 °C
< 3000 hr

HOT CORROSION
900 °C, 0.5 ppm NaCl
< 1000 hr

OXIDATION AGING
1100 °C
10, 100, 300, 600 hr

G.E. CONTRACT (NAS2-3936)

EMPERICAL DAMAGE MODEL
EFFECT OF HOT CORROSION AGING ON OXIDATION COATING LIFE

VERIFICATION
DUAL TEMPERATURE MACH 0.3 BURNER RIG
HOT CORROSION AND OXIDATION INCLUDING BLOCK TIME EFFECTS

Figure 12