

## COATING LIFE PREDICTION

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## INTRODUCTION

Hot-section gas-turbine components typically require some form of coating for oxidation and corrosion protection. These coatings are generally either aluminide 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  $Al_2O_3$  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  $Al_2O_3$  scale. In addition, cycling of a coated component results in cracking and spalling of the  $Al_2O_3$  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 aluminide 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 aluminide and LPPS NiCoCrAlY coatings (125 and 625  $\mu m$  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  $\text{Al}_2\text{O}_3/\text{NiAl}_2\text{O}_4$  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/spalling 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/spalling 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 spalling model. The spalling 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 spalling 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  $\text{Al}_2\text{O}_3$  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  $\text{NiO}$ ,  $\text{Cr}_2\text{O}_3$ , or  $\text{NiCr}_2\text{O}_4$ ) 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  $\gamma'$  and  $\beta$  phases occurs before coating failure.

The cyclic oxidation lives of the LPPS NiCoCrAlY 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/spalling 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 spalling model was an order of magnitude larger than expected for Al<sub>2</sub>O<sub>3</sub> 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 Al<sub>2</sub>O<sub>3</sub> that spalls after oxidation at 1050 to 1150 °C. Second, the spalling model assumes only external formation of Al<sub>2</sub>O<sub>3</sub>. 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 spalling model was estimated so that the predicted weight change was similar to that for the LPPS coating on U-700 (fig. 8). The spalling 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  $\gamma'$  or  $\beta$  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 spalling 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  $\zeta'$  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 improved 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

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3. Nesbitt, J. A.; and Heckel, R. W.: Modeling Degradation and Failure of Ni-Cr-Al Overlay Coatings. NASA TM-83672, 1984.
4. Nesbitt, James A.: Overlay Coating Degradation by Simultaneous Oxidation and Coating/Substrate Interdiffusion. NASA TM-3738, 1984.

# OXIDATION-LIMITED COATING LIFE PREDICTION

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

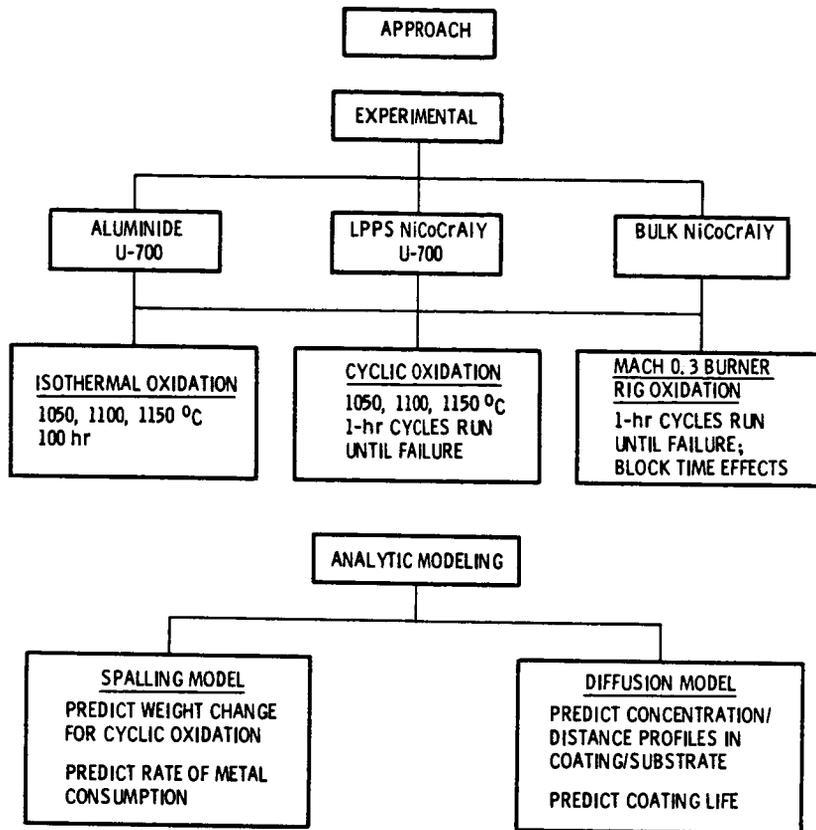
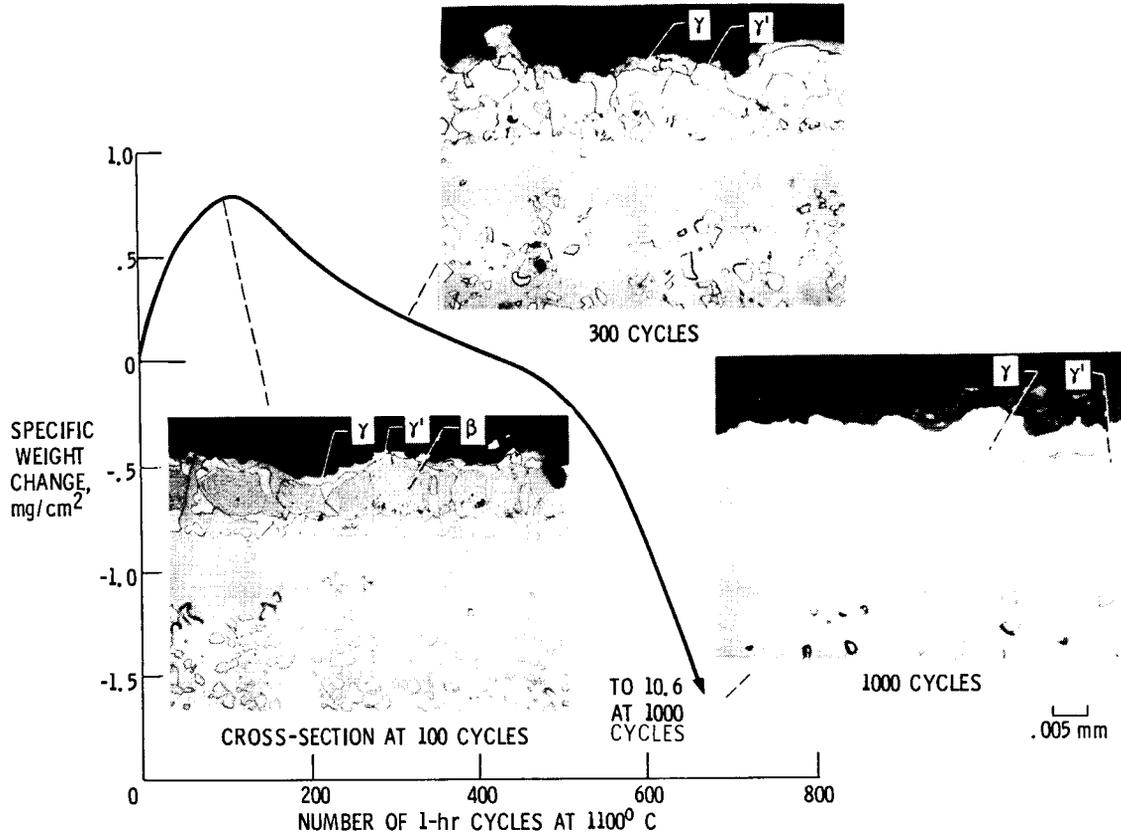


Figure 1

# CYCLIC FURNACE OXIDATION OF ALUMINIDE COATED U-700



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Figure 2

# CYCLIC FURNACE OXIDATION OF LPPS NiCoCrAlY OVERLAY COATING

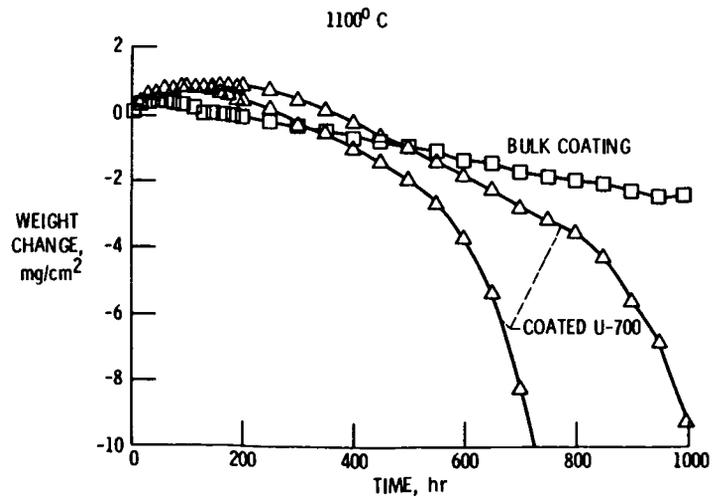
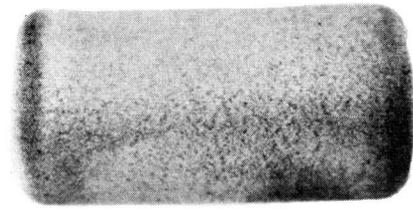


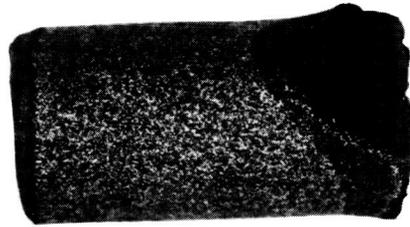
Figure 3

MACROSCOPIC COATING FAILURE

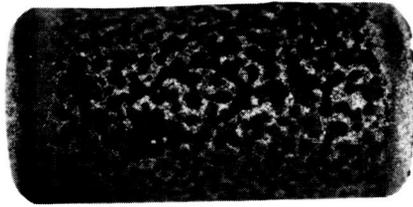
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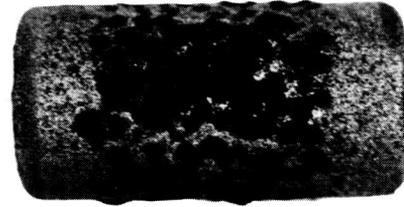
100 hr ISOTHERMAL AT 1150° C



1000 1-hr CYCLES AT 1100° C



100 1-hr CYCLES AT 1150° C



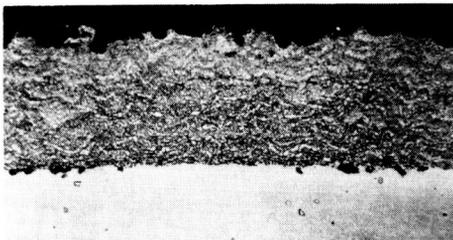
250 1-hr CYCLES AT 1150° C

Figure 4

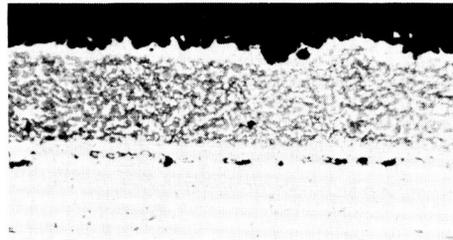
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MICROSCOPIC COATING FAILURE

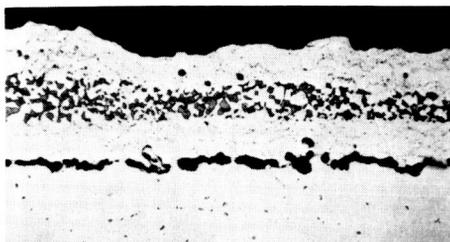
1150° C



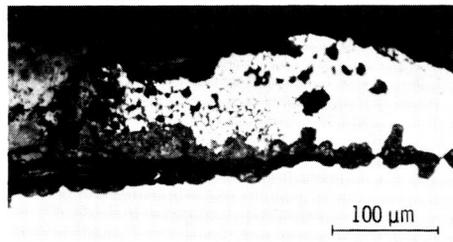
AS-SPRAYED PLUS 4-hr ARGON ANNEAL AT 1080° C



THIRTY 1-hr CYCLES



100 1-hr CYCLES



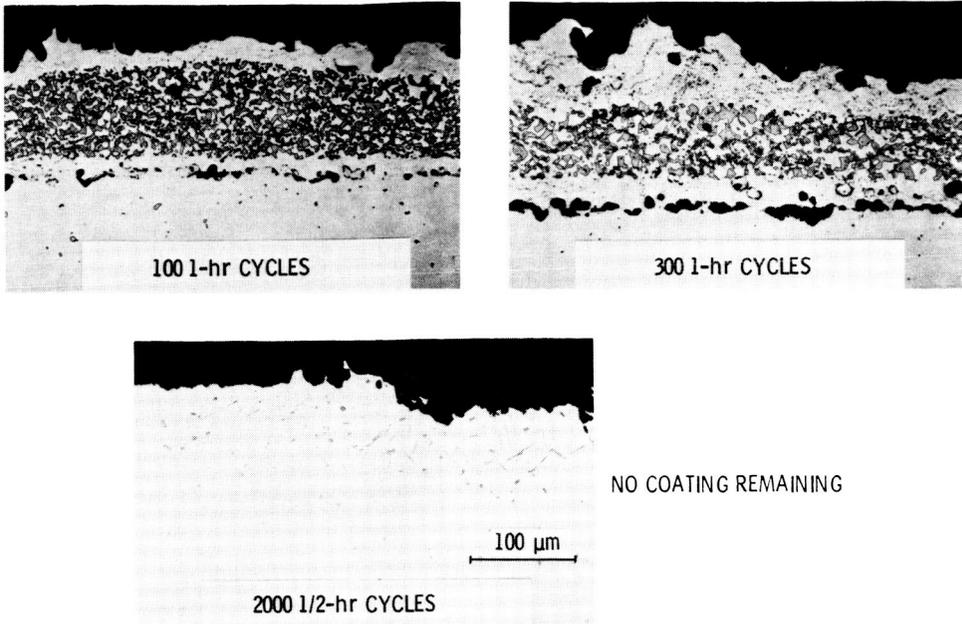
250 1-hr CYCLES

Figure 5

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# MICROSCOPIC COATING FAILURE

1100° C

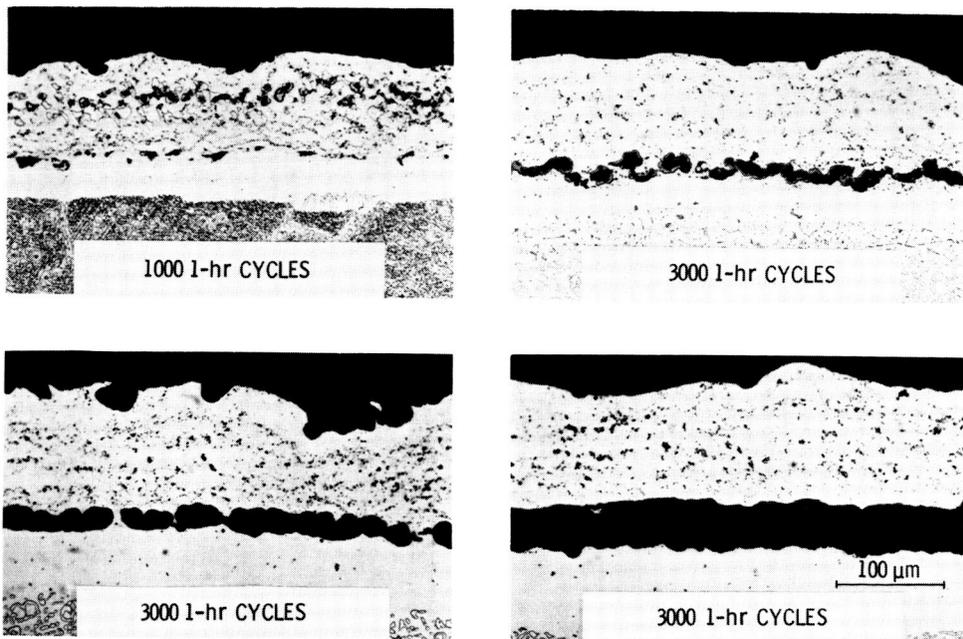


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Figure 6

# MICROSCOPIC COATING FAILURE

1050° C



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Figure 7

# CYCLIC FURNACE OXIDATION OF LPPS NiCoCrAlY COATING ON U-700

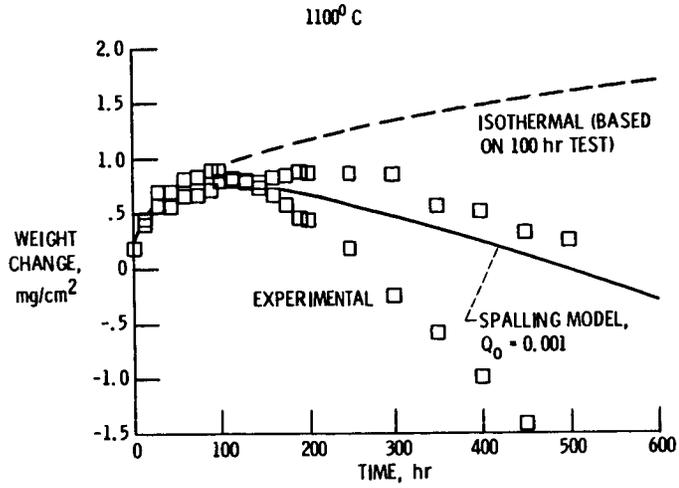


Figure 8

## ALUMINUM CONSUMPTION PREDICTED BY SPALLING MODEL

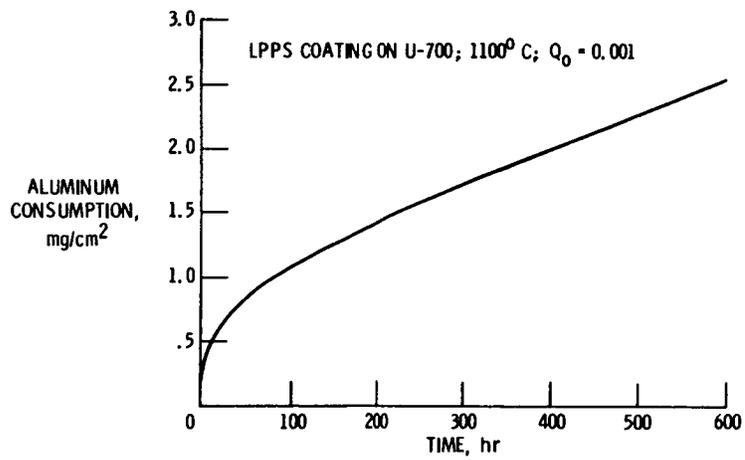


Figure 9

### RATE OF Al CONSUMPTION PREDICTED BY SPALLING MODEL

LPPS COATING ON U-700; 1100° C;  $Q_0 = 0.001$

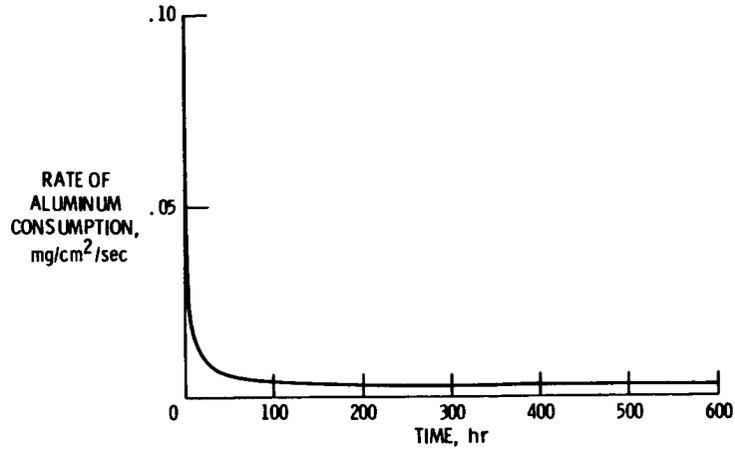


Figure 10

### ALUMINUM CONCENTRATION PREDICTED BY DIFFUSION MODEL

LPPS NiCoCrAlY COATING ON U-700; 300 1-hr CYCLES 1100° C

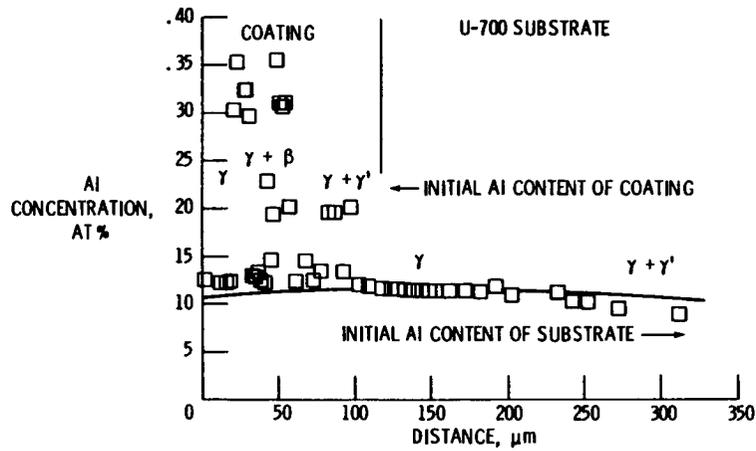


Figure 11

## DUAL CYCLE ATTACK

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

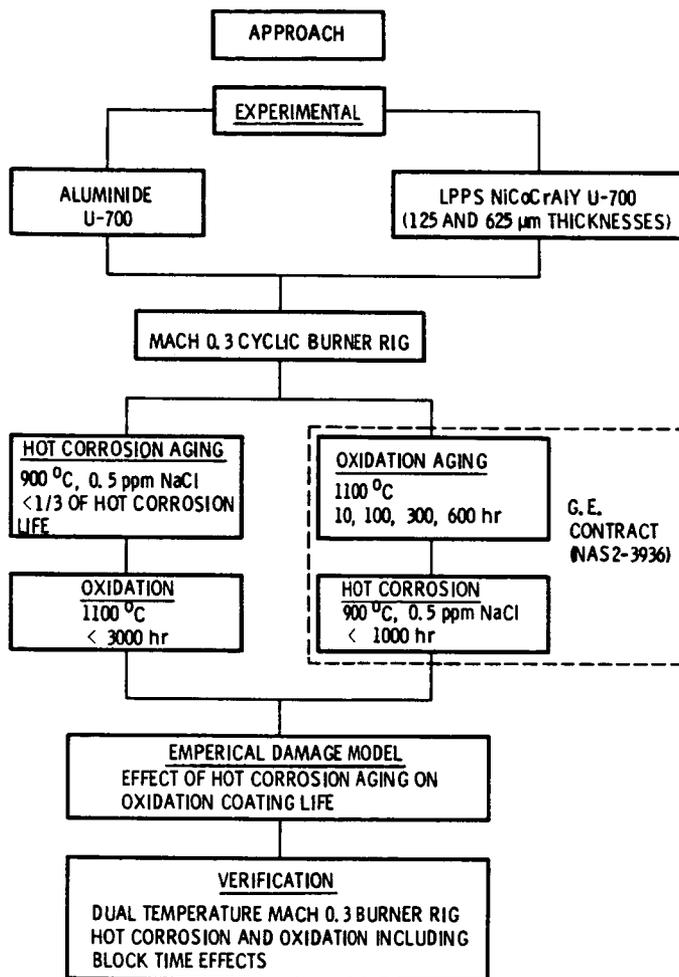


Figure 12