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ABSTRACT

A new concept for enhanced protection of superalloys consists of adding an oxidation- and diffusion-resistant cermet layer between the superalloy and the outer oxidation resistant metallic alloy coating. Such a duplex coating was compared with a physical-vapor-deposited (PVD) NiCrAlY coating in cyclic oxidation at 1150° C. The substrate alloy was MA 754 - an oxide-dispersion-strengthened superalloy that is difficult to coat. The duplex coating, applied by plasma spraying, outperformed the PVD coating on the basis of weight change and both macroscopic and metallographic observations.

SUMMARY

A new coating concept for the high-temperature protection of superalloys is presented. The concept consists of initially applying an oxidation-resistant cermet (e.g., mechanically alloyed NiCrAlY + Y_2O_3 powders) on the substrate and then applying a top coating of an oxidation-resistant alloy (e.g., a NiCrAlY alloy). A method (but not the only one) for applying the duplex coating is arc-plasma spraying. The composition of the two coatings may be tailored to the substrate in order to reduce concentration gradients, and the top coating may be partly aluminized to improve its oxidation resistance.

To test this concept, the difficult-to-coat, oxide-dispersion-strengthened superalloy MA 754 was selected as the substrate. The performance of the diffusion-inhibited, plasma-sprayed duplex coating and a physically vapor-deposited (PVD) NiCrAlY coating was evaluated in cyclic furnace and Mach 0.3 burner rig oxidation tests at 1150° C. The duplex plasma-sprayed coating outperformed the PVD coating in both tests. Life of the duplex coating in the burner rig test was limited to about 200 hours because of localized wearout, as compared with less than 100 hours for the PVD coating. Post-test metallographic evaluations showed that, although the PVD coating was completely separated from the substrate after only 10 hours of furnace testing, the diffusion-inhibited coatings were well bonded after 500 hours of furnace testing as well as after 300 hours of burner rig testing.

INTRODUCTION

In advanced gas turbine engines, coatings are used to protect superalloy components from oxidation and hot corrosion. The useful lives of these components, however, are often limited by coating wearout at high temperatures.

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Wearout results from depletion of aluminum in the coating by aluminum oxide scale spallation and from dilution by interdiffusion with the substrate. The latter mechanism is often predominant at temperatures of 1100° C and above (ref. 1). Inhibiting interdiffusion thus offers a means for significantly extending coating life. Many attempts to improve the diffusional stability of coatings have been made. One approach has been to interpose an alloy layer, thereby grading the aluminum concentration profile (refs. 2 and 3). In an alternative approach a refractory metal or alloy layer is interposed (refs. 4 to 6). Neither of these approaches appeared to be successful for use at temperatures above 1100° C. This report presents a new approach involving the interposition of a cermet layer between the metallic top coating and the substrate. This approach has recently become feasible as a result of developments in mechanical alloying (refs. 7 to 9) and low-pressure arc-plasma spraying (ref. 10) that permit fabrication of the cermet and the coating application, respectively.

To most effectively demonstrate the feasibility of this concept, an alloy that has a severe coating diffusional instability problem was selected for this study. Mechanically alloyed (MA), oxide-dispersion-strengthened (ODS) superalloys have the strength potential for thousands of hours of use at hot-spot temperatures to 1200° C (refs. 8 and 9). Even though the endurance of some ODS alloys has been improved by coating processes that use either aluminide diffusion coatings, aluminum-bearing overlay coatings, or a combination of these, the coatings offer only interim, short-term solutions to the problem (refs. 11 to 14). The protection ability of these coatings rapidly degrades at high temperatures, primarily because of diffusional instability between the coating and the ODS substrate. For example, results obtained in cyclic oxidation testing for 200 hours at 1150° C in a Mach 0.3 burner rig show that NiCrAlY and aluminide coatings on ODS MA 755E offered, at best, only marginal protection (ref. 11). Post-test, cross-sectional metallographic analysis showed considerable porosity (Kirkendall voids), especially in the high-activity-processed aluminide coating specimens. Some porosity was observed only in the aluminide coating before testing. However, ODS alloys (and to a lesser extent the non-ODS superalloys) that contain minor amounts of aluminum ($\lesssim 1$ wt %) are either very difficult to coat or the coating life is very short. The tendency for Kirkendall void formation in ODS alloys during either aluminide coating (high and low activity processing) or subsequent exposure at high temperatures was shown to be greater for low-aluminum-content alloys (ref. 12). This porosity formation along the coating-substrate interface eventually leads to delamination and spalling of the coating.

At NASA Lewis various coatings on MA 754 and MA 6000E are being evaluated in cyclic furnace and Mach 0.3 burner rig oxidation tests at 1150° C in order to gain a more complete understanding of coating - ODS alloy interactions and to identify promising new coating concepts. In this investigation the low-aluminum-content alloy MA 754 has proven to be far more difficult to protect than MA 6000E. Thus MA 754 was chosen as the substrate for demonstrating the feasibility of the diffusion-inhibited coating concept.

MATERIALS, APPARATUS, AND PROCEDURES

Specimens of MA 754 (table I) were coated with either NiCrAlY or NiCrAlY plus NiCrAlY-Y_2O_3 by arc plasma spraying or with NiCrAlY by electron beam

PVD. The plasma-sprayed coatings were manually applied by using a specially designed Mach 2.0 gun with a CO_2 shroud. The chemical analyses of the spray powders are given in table II. The target thicknesses were 0.12 mm for the single coatings and 0.06 mm for each base and top coating in the duplex coatings. Mechanically alloyed powders (refs. 6 to 8) were used for the plasma-sprayed base coating on MA 754, Ni-13Cr-7Al-0.3Y-27Y_2O_3 for the furnace coupons (25.4 by 19.1 by 5.1 mm), and Ni-16Cr-8Al-0.5Y-8Y_2O_3 for the burner rig bars (12.7 mm diameter by 82.6 mm long - see ref. 3 for design and other dimensions). A dry-atomized Ni-18Cr-12Al-0.3Y alloy powder, however, was used in applying the plasma-sprayed top coating. After spraying, the specimens were glass-bead peened and then heat treated at 1080° C for 4 hours in argon. The microstructure of a plasma-sprayed NiCrAlY plus NiCrAlY-Y_2O_3 cermet coating on MA 754 is shown in figure 1(a).

The PVD NiCrAlY coating (Ni-19Cr-11Al-0.45Y), which is a commercial coating used on many current turbine blades and vanes, was applied to a target thickness of 0.12 mm by a coating vendor. After coating, the specimens were glass-bead peened and subsequently heat treated by the vendor at 1080° C for 4 hours in vacuum. A PVD NiCrAlY coating on MA 754 is shown in figure 1(b).

The coatings were evaluated in cyclic furnace and Mach 0.3 burner rig oxidation tests at 1150° C. The procedures and apparatus used in these tests are fully described in references 15 and 16. Weight changes were recorded during testing and post-test visual, X-ray diffraction (XRD), and cross-sectional metallographic analyses were performed.

RESULTS AND DISCUSSION

As-Deposited Coatings

Pretest XRD and metallographic analyses were performed on selected specimens. The XRD data shown in tables III and IV indicate that β-NiAl was a major phase only in the PVD coating; β was not detected even as a minor phase in the plasma-sprayed coatings. The $M_{23}C_6$ (minor) phase in the latter coatings is probably due to dissociation of the CO_2 shroud gas and formation of $Cr_{23}C_6$ in the plasma arc during spraying. The $M_{23}C_6$ phase, detected as a major phase in the PVD coating, is believed to have formed during the postcoating vacuum heat treatment. XRD patterns obtained on recently coated specimens showed no evidence of $M_{23}C_6$ after PVD coating but indicated the presence of $M_{23}C_6$ as a major phase (strong pattern intensity) after the postcoating vacuum heat treatment. The effect of $M_{23}C_6$ on coating performance has not been studied. Carbide formation, however, may be desirable on carbide-type directional eutectic alloys. It is suggested that, in the future, inert environments and environmental monitoring be employed during coating and postcoating heat treatments.

Cross-sectional photomicrographs of plasma-sprayed and PVD-coated MA 754 specimens are shown in figure 1. Since the plasma-sprayed coatings (fig. 1(a)) were manually applied, considerable variation in coating thicknesses was noted in these specimens. The globular particles observed throughout the PVD coating (fig. 1(b)) are believed to be $Cr_{23}C_6$. The dark areas on the substrate side of the bond interface in the PVD-coated MA 754 are areas of porosity. It is believed that this porosity occurred primarily during the postcoating heat treatment as a result of interdiffusion (i.e., nickel diffused faster into the coating than aluminum into the substrate). This porosity, however, has not been observed in PVD-NiCrAlY-coated MA 755E (ref. 11) and MA 6000E (fig. 2) before exposure at temperatures of 1100° C or higher. It is felt that this absence of porosity in MA 755E and MA 6000E is due primarily to their higher aluminum concentration and perhaps secondarily to their higher refractory metal content relative to MA 754 (table 1). This reasoning is analogous to that proposed in reference 4 with respect to aluminide coatings on ODS alloys. The plasma-spray-coated MA 754 coupon in figure 1(a), however, shows no comparable porosity. Electron microprobe (EMP) profiles of nickel, chromium, and aluminum across the coating-substrate interface indicated very little, if any, interdiffusion as compared with significant nickel and aluminum interdiffusion for the PVD-coated MA 754.

Cyclic Furnace Oxidation

The specific weight change results obtained on bare and coated MA 754 coupons in cyclic furnace oxidation at 1150° C are shown in figure 3. After an initial weight gain, bare MA 754 experienced rapid losses in weight on subsequent cycling. In contrast, the PVD-NiCrAlY-coated MA 754 coupons showed weight gains through 100 hours of testing. The latter coupons were removed from further testing primarily because they exhibited severe blistering and apparent separation of the coating from the substrate. The duplex plasma-spray-coated coupons also showed weight gains, but only to about 75 to 100 hours and then lost weight thereafter. The plasma-sprayed coupons showed no evidence of blistering, even after 500 hours of testing. The weight losses experienced by these coupons are believed to be due mainly to coating distress, which began at the corner that had an attached support tab (fig. 4). Although the support tab (a low-aluminum-content NiCrAlY alloy) is required for PVD coating, the support tab on the plasma-sprayed coupons was used only to facilitate handling during spraying.

The XRD results shown in table III indicate that the β-NiAl present in the PVD coating before testing decomposed in 100 hours or less at 1150° C as a result of spalling of the protective α-Al_2O_3 oxide and/or coating-substrate interdiffusion. It is believed that the loss of the β phase is primarily due to coating-substrate interdiffusion. The basis for this belief is that cross-sectional metallographic analysis of similarly coated coupons after only 10 oxidation cycles (fig. 5) or 10 hours of isothermal oxidation at 1150° C showed that what little β-NiAl phase remained was located just below the oxide scale. No β-NiAl phase was detected in the coating near the substrate. If spalling of the protective α-Al_2O_3 oxide had been the primary cause of the β decomposition, the remaining β phase would have been located at the coating-substrate interface. Post-test metallographic examination of the PVD coating after 100-hour cyclic oxidation showed no evidence of β-NiAl.

The superiority of the plasma-sprayed coating over the PVD coating is clearly shown in figures 5 and 6, which show the coating separated from the substrate. In figure 5 the PVD coating completely separated from the substrate (except at the corners) after only 10 hours. It is believed that the separation was a result of interdiffusion between the coating and substrate that eventually forms a continuous network of Kirkendall voids along the bond interface. The plasma-sprayed duplex coating (fig. 6(b)), however, shows very little evidence of Kirkendall voids along the bond interface after 500 hours.

This is attributed primarily to the cermet base coating, which acted as a diffusion inhibitor, and secondarily to the shallower concentration gradients. EMP profiles of nickel, chromium, and aluminum across the coating-substrate interfaces of the specimens shown in figure 6 indicated very little interdiffusion in the plasma-spray-coated coupon after 500 hours as compared with that in the PVD-coated coupon after only 100 hours. Therefore the plasma-sprayed cermet interlayer appears to be an effective diffusion inhibitor.

Cyclic Burner Rig Oxidation

Bare and coated MA 754 were oxidized at 1150° C for as long as 300 hours in a Mach 0.3 burner rig. The weight changes for all specimens are plotted in figure 7. Bare MA 754 experienced severe weight losses in less than 20 hours. The PVD-coated specimens, however, showed nonreproducible weight change behavior: One specimen showed a very high weight gain after 100 hours; the other exhibited very little weight change up to 100 hours but then began to lose weight rapidly. The behavior of the PVD-coated specimens is attributed to erratic blistering, separation, and cracking of the coating. In contrast, the behavior of the plasma-sprayed specimens was quite reproducible. Based on the time required to experience a significant rate of weight loss, the lives of the plasma-sprayed coatings were about 160 to 230 hours. However, the life of the duplex coating was only about 15 to 70 hours greater than that of the single plasma-sprayed coating. The duplex plasma-sprayed coating was only slightly diffusion inhibited with a cermet base coating containing about 8 wt % Y_2O_3. Although the conclusion is not yet verified, furnace test results seem to show that a coating system containing about 27 wt % Y_2O_3 in the base coating would perform significantly better. Furthermore aluminizing the top coating should be highly beneficial. XRD analyses (table IV) indicated that β-NiAl (a prerequisite for a good oxidation-resistant NiCrAlY) was not present in the top coating before testing, probably as a result of aluminum depletion by oxidation during plasma spraying. Thus application by low-pressure (inert gas) plasma spraying should improve coating performance (ref. 10).

Figure 8 shows coated MA 754 specimens after burner rig testing. Coating failures occurred either on or next to the back surface of the specimens, which were at a higher temperature in the burner flame than the front surface (ref. 17). The two PVD-coated specimens showed blistering and cracking of the coating (fig. 8(a)), which resulted in either a high or very low weight gain after 100-hour exposure, as discussed previously. Although the two plasma-sprayed coatings (figs. 8(b) and (c)) were adherent for at least 300 hours, they started to fail by local wearout and then by catastrophic substrate attack after about 200 hours, but only in one area of each specimen. The incidence of coating wearout is in agreement with the gravimetric data in figure 7.

After burner rig testing, selected specimens were cross sectioned through the maximum attack zone and also near the bottom. The cross section near the bottom was considered to be typical of the "as-coated" condition. Photomicrographs of the coated specimens in the as-coated and as-tested conditions are shown in figure 9. A cross section of a PVD-coated MA 754 specimen after 100 hours of exposure at 1150° C (not presented in fig. 9) showed blistering and separation of the coating from the substrate that was similar to that presented in figures 5 and 6(a), but more severe. Comparison of the microstructures in figures 9(a) and (b) indicates some loss of the two plasma-sprayed coatings after 300 hours of exposure at 1150° C. The duplex, slightly diffusion-inhibited, plasma-spray-coated microstructure after testing (fig. 9(b)), however, shows slightly less porosity on the substrate side of the bond interface and within the substrate than does the single-layer, plasma-spray-coated specimen (fig. 9(a)). This suggests that the cermet base coating (fig. 9(b)) was somewhat instrumental in reducing coating-substrate interdiffusion during exposure in the burner flame and that higher volume fractions of ceramic phase, such as tested in the cyclic furnace tests, are preferred. As noted above, localized coating wearout and severe substrate attack (fig. 9(c)) were observed on the plasma-spray-coated specimens.

CONCLUDING REMARKS

Coatings currently used to protect superalloys wear out by oxide scale spallation and by coating dilution as a result of interdiffusion with the substrate. A new coating concept involving the application of a duplex coating system wherein the inner layer is a diffusion-inhibiting cermet and the outer layer is a conventional overlay coating composition has been identified. Its performance was compared with that of a conventional physical-vapor-deposited (PVD) NiCrAlY overlay coating on the difficult-to-coat, oxide-dispersioned-strengthened (ODS) alloy MA 754 in cyclic furnace and Mach 0.3 burner rig oxidation at 1150° C.

The new coating concept consists of initially coating the substrate with an intermediate cermet layer (base coating). In the present study the cermet was prepared by mechanical alloying. In that concept an oxidation- and hot-corrosion-resistant alloy (top coating) is then deposited on the cermet. To improve coating and bonding, the as-coated alloy is glass-bead peened and heat treated at 1080° C for 4 hours in an inert atmosphere. The metallic-element composition of the two coatings may be tailored to reduce the concentration gradients between the top coating and the substrate. Also, the top coating may be partly aluminized to improve its oxidation resistance. A method for applying the top and base coatings is arc-plasma spraying the coating powders in ambient air by using an inert-gas shroud and/or in a chamber with an inert-gas or low-pressure environment. In the present work the sprayed NiCrAlY coatings were applied in air by using a Mach 2 gun with a CO_2 shroud. These coatings were not aluminized before testing. For furnace and burner rig tests the sprayed base coatings were $Ni\text{-}13Cr\text{-}7Al\text{-}0.3Y\text{-}27Y_2O_3$ and $Ni\text{-}16Cr\text{-}8Al\text{-}0.5Y\text{-}8Y_2O_3$, respectively. The sprayed top coating for both tests was Ni-18Cr-12Al-0.3Y.

In the furnace tests the duplex plasma-sprayed coating outperformed the PVD Ni-19Cr-11Al-0.45Y coating. The PVD coating separated from the MA 754 substrate in 10 hours or less as a result of interdiffusion between the coating and the substrate that led to the formation of Kirkendall voids. The plasma-spray coated MA 754 showed very little evidence of Kirkendall porosity after a 500-hour cyclic exposure. This was attributed primarily to the cermet base coating acting as a diffusion inhibitor.

The life of the duplex plasma-sprayed coating in the burner rig test was limited to about 200 hours because of localized wearout of the coating. However, in areas in which the coating was not completely worn out after 300 hours, the coating was adherent even though some Kirkendall porosity was observed in the MA 754. The PVD coating, however, blistered and cracked in less than 100 hours of cyclic exposure in the burner rig.

The new coating concept was tried on ODS MA 754, which contains only 0.3 wt % aluminum. The concept is also applicable to higher-aluminum-content ODS alloys (e.g., MA 6000E (4.5 wt % Al)) and to non-ODS superalloys. Since coating-substrate interdiffusion is the primary contributor to the wearout of metallic protective coatings, this coating concept offers a means for significantly extending coating life. Furthermore opportunities for grading thermal expansion are available by judicious selection of the ceramic constituent and its concentration in the cermet layer. Thus this concept may also be applicable to low-thermal-expansion superalloys such as directionally solidified eutectics and powder-metallurgy alloys fabricated from rapidly solidified powders.

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TABLE I. - NOMINAL COMPOSITION OF OXIDE-DISPERSION-STRENGTHENED SUPERALLOYS

Element	Superalloy		
	MA 754	MA 755E	MA 6000E
	Composition, wt %		
Cr	20.3	15.0	15.0
Al	.3	4.5	4.5
W	-----	5.5	4.0
Mo	-----	3.5	2.0
Ta	-----	2.5	2.0
Ti	.4	3.0	2.5
Y_2O_3	.6	1.1	1.1
Ni	Major	Major	Major

TABLE II. - CHEMICAL ANALYSES OF PLASMA-SPRAYED POWDERS AND ELECTRON-BEAM, PHYSICAL-VAPOR-DEPOSITED (PVD) COATING

Element	Spray powders			PVD Ni-19Cr-11Al-0.45Y coating[a]
	Ni-18Cr-12Al-0.3Y	Ni-13Cr-7Al-0.3Y-27Y_2O_3	Ni-16Cr-8Al-0.5Y-8Y_2O_3	
Ni	Major	Major	Major	Major
Cr	18.0	13.16	16.20	15.6 to 21.8
Al	12.4	6.60	8.21	10.5 to 10.8
Y	.29	21.38	6.70	.43 to .47
B	.002	(c)	<.01	(b)
C	.018	.089	.085	↓
Ca	<.001	(c)	<.01	
Co	.19	<.01	.035	
Cu	.009	.008	.027	
Fe	.075	.28	1.04	
Hf	<.01	(b)	<.01	
Mg	<.001	(c)	<.01	
Mn	.002	<.01	.027	
Mo	<.005	.011	.045	
Nb	<.01	(c)	(c)	
O_2	.011	(b)	(b)	
P	<.001	<.001	<.001	
S	<.001	.005	.004	
Si	.048	.015	.035	
Ta	<.05	ND	<.05	
Ti	<.005	<.01	.021	
V	<.005	<.01	.015	
W	<.01	(c)	(c)	
Zn	<.005	(c)	(c)	
Zr	<.01	(c)	(c)	

[a] X-ray fluorescence of coating on witness tabs (range of analyses).
[b] Not determined.
[c] Not detected.

TABLE III. - PHASES IDENTIFIED BY X-RAY DIFFRACTION BEFORE AND AFTER CYCLIC FURNACE OXIDATION OF COATED MA 754 AT 1150° C

Nominal coating composition, wt %	Major phases (relative intensities)[a]	Minor phases (relative intensities)[a]
Before oxidation		
Plasma-sprayed Ni-18Cr-12Al-0.3Y (top) plus Ni-13Cr-7Al-0.3Y-27Y_2O_3 (base)	γ or γ/γ' (VS)	$M_{23}C_6$ (W) α-Al_2O_3 (VVW)
PVD Ni-19Cr-11Al-0.45Y	γ or γ/γ' (S) β-NiAl (M) $M_{23}C_6$ (M)	α-Al_2O_3 (W)
After 100-hour oxidation[b]		
Plasma-sprayed Ni-18Cr-12Al-0.3Y (top plus Ni-13Cr-7Al-0.3Y-27Y_2O_3 (base)[d]	γ or γ/γ' (S) α-Al_2O_3 (S) Spinel (M) 0.810[c]	-------------
PVD Ni-19Cr-11Al-0.45Y	γ or γ/γ' (S) α-Al_2O_3 (S)	

[a] Relative pattern intensities: VS, very strong; S, strong; M, moderate; W, weak; VVW, very very weak.

[b] Cycles: 1 hr at 1150° C and 20 min or more cooling.

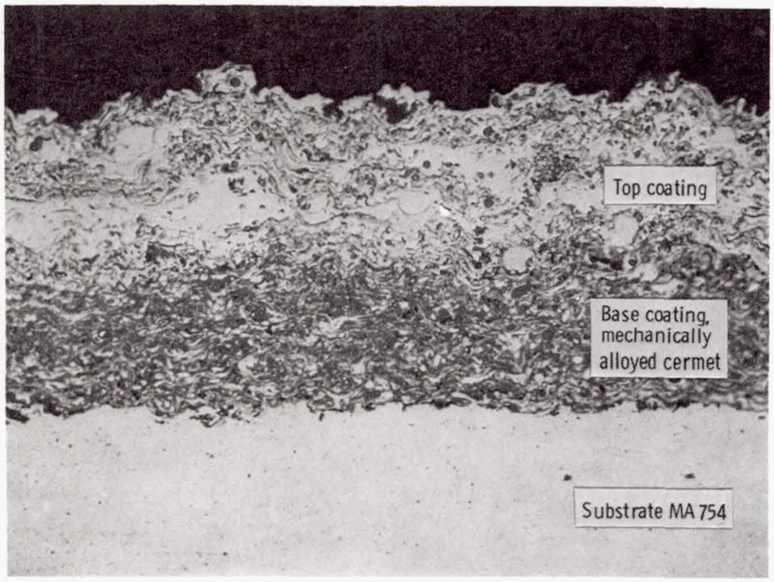
[c] Estimated lattice spacing a_0 of spinel in nanometers (±0.05 nm).

[d] Same results after 500 hr of oxidation.

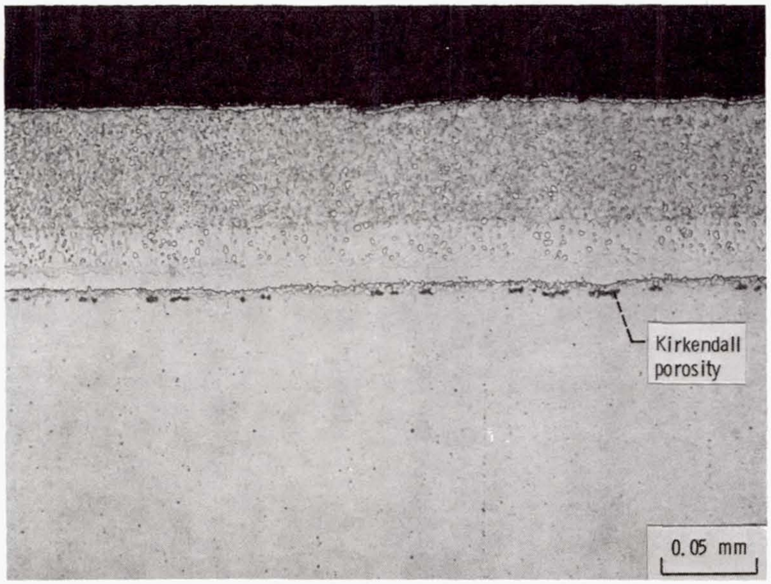
TABLE IV. - PHASES IDENTIFIED BY X-RAY DIFFRACTION ON COATED MA 754 SPECIMENS BEFORE BURNER RIG TESTING

Nominal coating composition, wt %	Major phases (relative intensities)[a]	Minor phases (relative intensities)[a]
PVD Ni-19Cr-11Al-0.45Y	$M_{23}C_6$ (S) β-NiAl (M)	γ or γ/γ' (W) α-Al_2O_3 (W)
Plasma-sprayed Ni-18Cr-12Al-0.3Y	γ or γ/γ' (S)	$M_{23}C_6$ (W) α-Al_2O_3 (VW)
Plasma-sprayed Ni-18Cr-12Al-0.3Y (top) plus Ni-16Cr-8Al-0.5Y-8Y_2O_3 (base)	γ or γ/γ' (S)	$M_{23}C_6$ (W) α-Al_2O_3 (W)

[a] Relative pattern intensities: S, strong; M, moderate; W, weak; VW, very weak.



(a) Plasma-sprayed Ni-18 Cr-12Al-0.3Y (top) and Ni-13 Cr-7Al-0.3Y-27Y_2O_3 (base) coating heat treated at 1080° C for 4 hr in argon.



(b) Physical-vapor-deposited Ni-19 Cr-11Al-0.45Y coating heat treated at 1080° C for 4 hr in vacuum.

Figure 1. - Photomicrographs of as-coated MA 754.

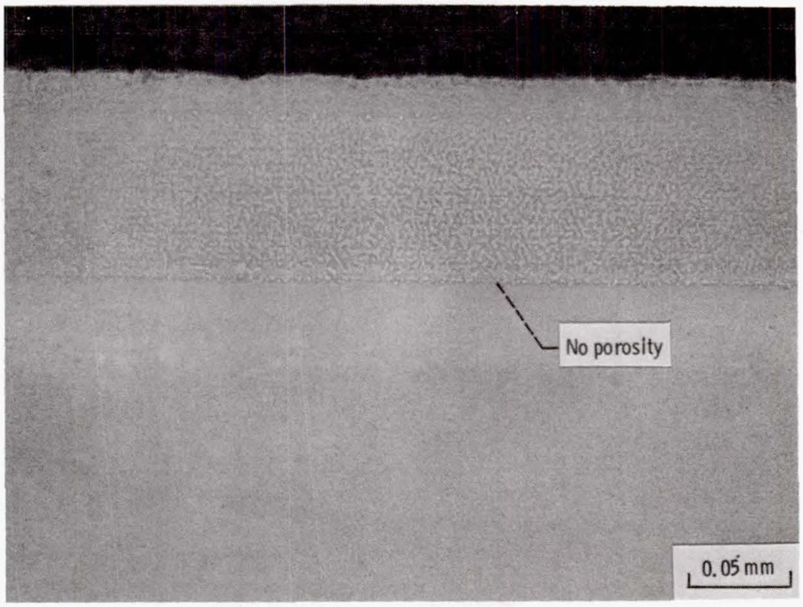


Figure 2. - Photomicrograph of physical-vapor-deposited Ni-19Cr-11Al-0.45Y-coated MA 6000E heated treated at 1080° C for 4 hr in vacuum.

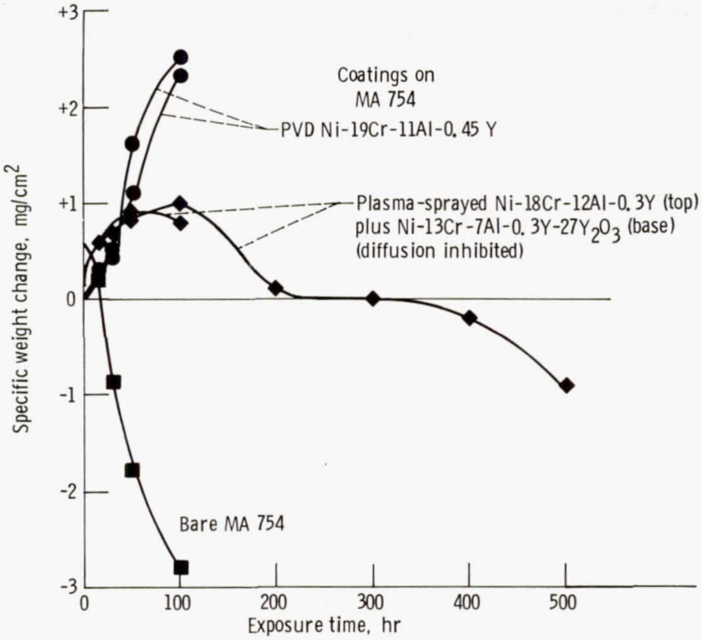
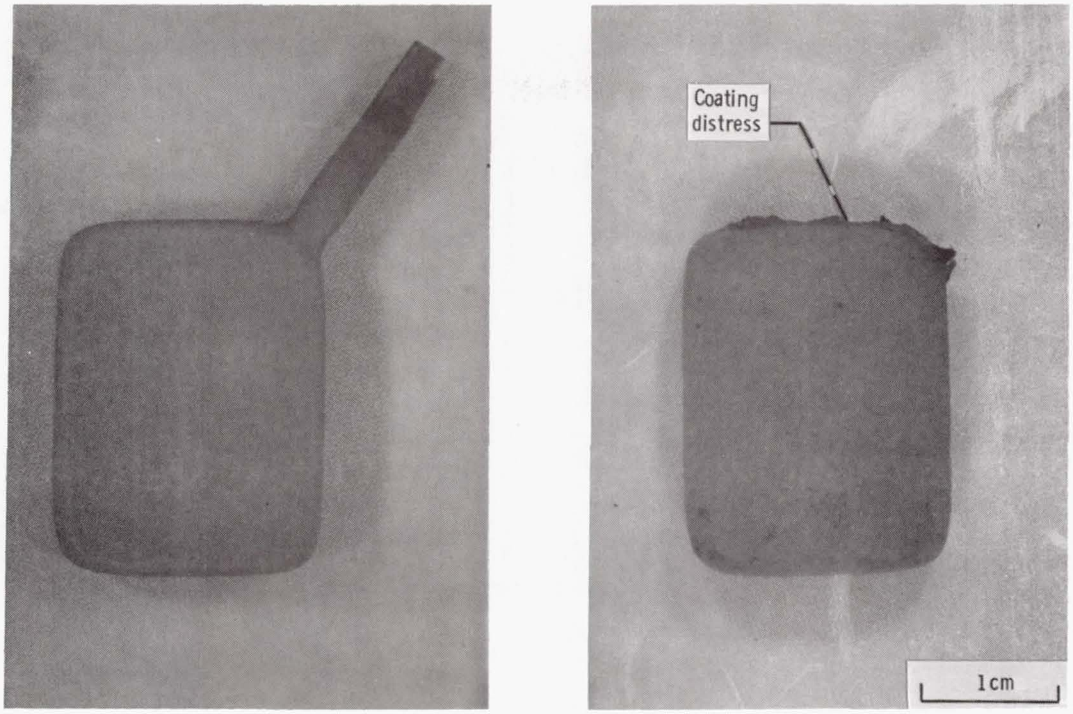


Figure 3. - Cyclic furnace oxidation of bare and coated Ma 754 at 1150° C using 1-hr cycles.



(a) As-coated coupon with support tab attached.

(b) Coated coupon after 500 hr of cyclic (1-hr cycles) furnace oxidation at 1150°C. Support tab removed before furnace oxidation.

Figure 4. - Plasma-spray-coated MA 754 coupons before and after cyclic furnace oxidation. Coating, Ni-18 Cr-12Al-0.3Y plus Ni-13 Cr-7Al-0.3Y-27Y_2O_3.

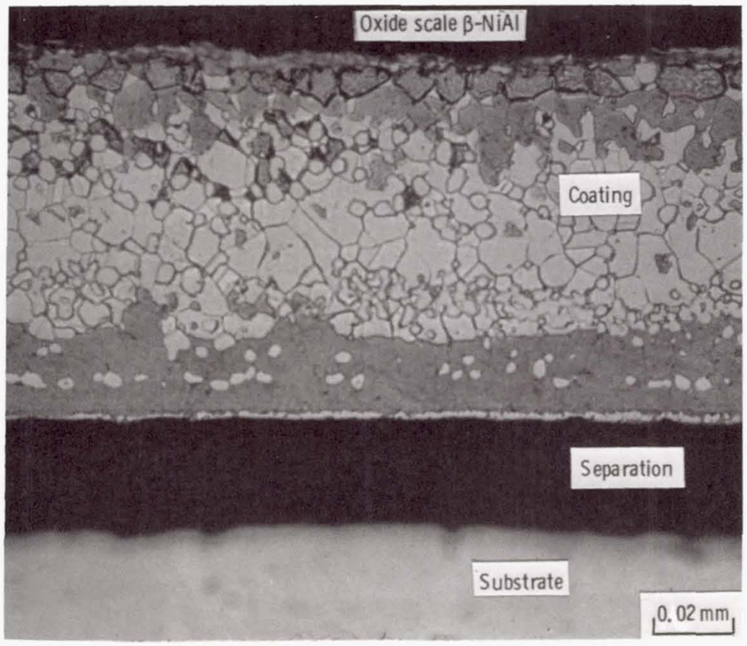
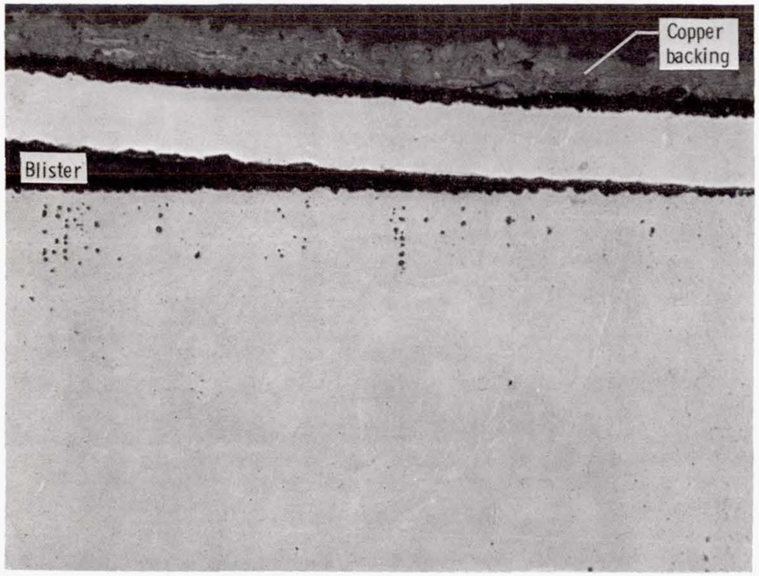
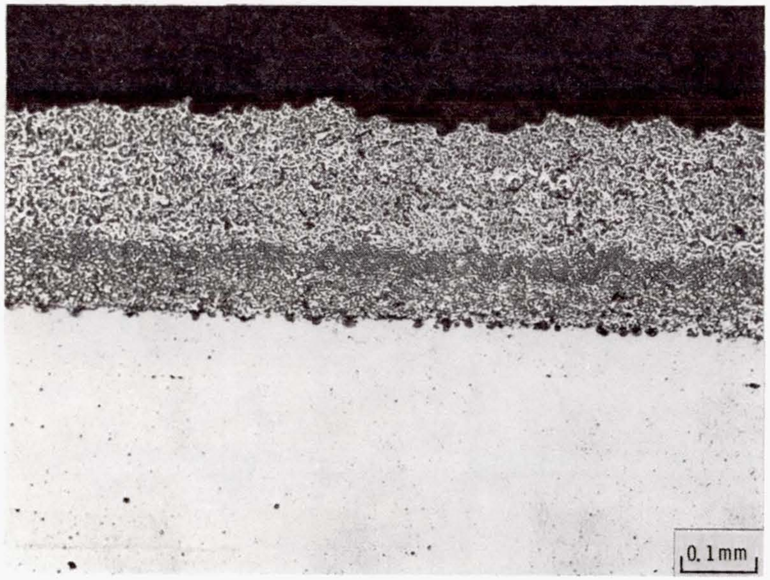


Figure 5. - Photomicrograph of physical-vapor-deposited Ni-19Cr-11Al-0.45Y on MA 754 after ten 1-hr cycles of furnace oxidation at 1150° C. Etched to reveal β-NiAl.



(a) Physical-vapor-deposited Ni-19 Cr-11Al-0.45Y on MA 754 after 100 hr.



(b) Plasma-sprayed Ni-18 Cr-12Al-0.3Y plus Ni-13 Cr-7Al-0.3Y-27Y_2O_3 on MA 754 after 500 hr.

Figure 6. - Photomicrographs of coated coupons tested in cyclic furnace oxidation at 1150° C; 1-hr cycles.

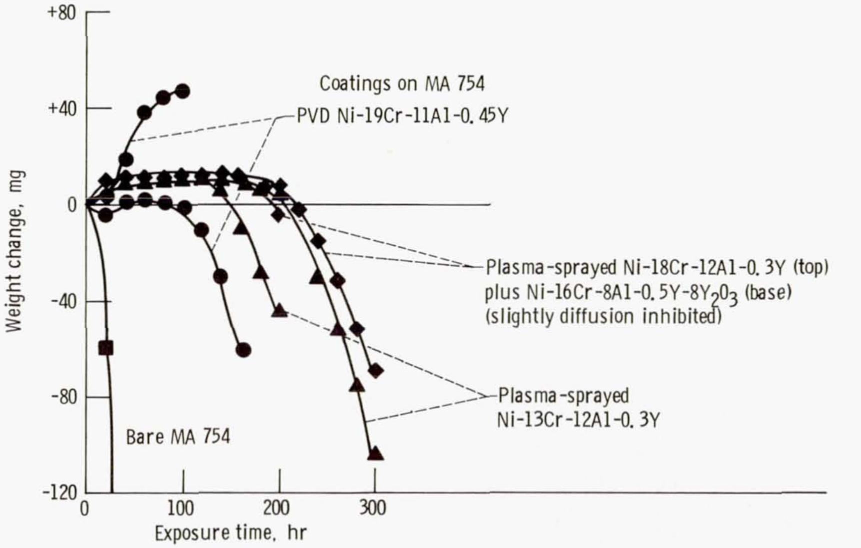
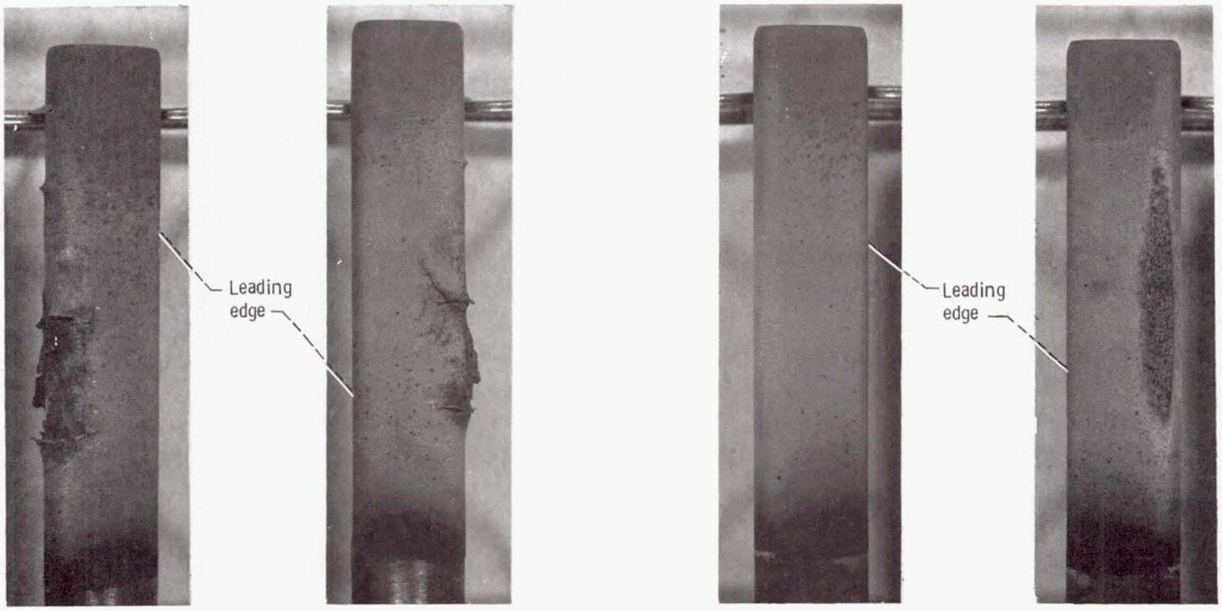
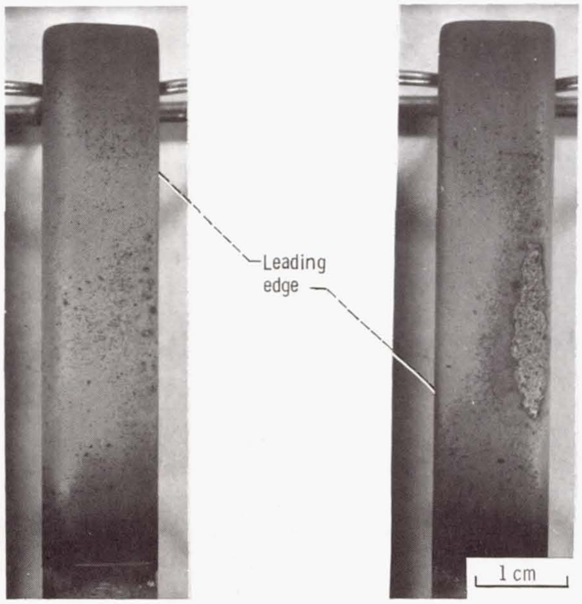


Figure 7. - Cyclic burner rig oxidation of bare and coated MA 754 at 1150° C and Mach 0.3; 1-hr cycles.



(a) Physical-vapor-deposited Ni-19CR-11Al-0.45Y-coated MA 754 after 160 hr.

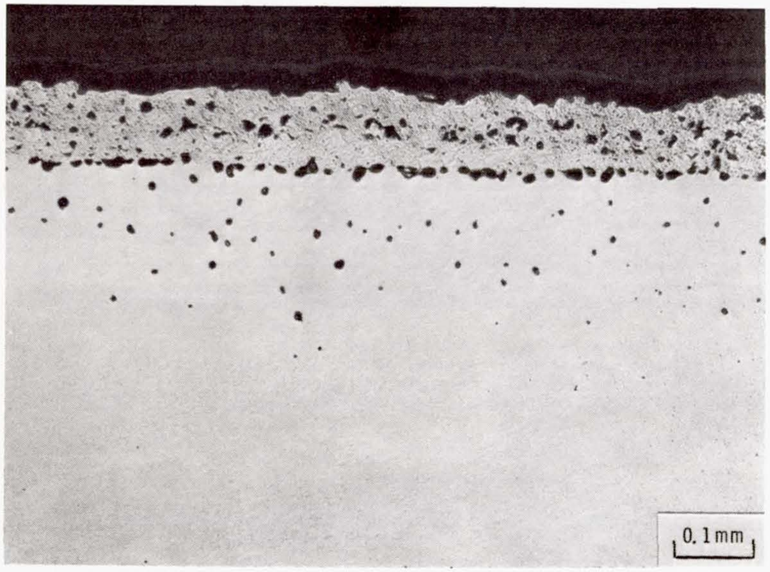
(b) Plasma-sprayed Ni-18Cr-12Al-0.3Y-coated MA 754 after 300 hr.



(c) Plasma-sprayed Ni-18Cr-12Al-0.3Y (top) plus Ni-16Cr-8Al-0.5Y-8Y_2O_3 (base) coated MA 754 after 300 hr.

Figure 8. - Coated MA 754 after cyclic Mach 0.3 burner rig oxidation at 1150° C; 1-hr cycles.

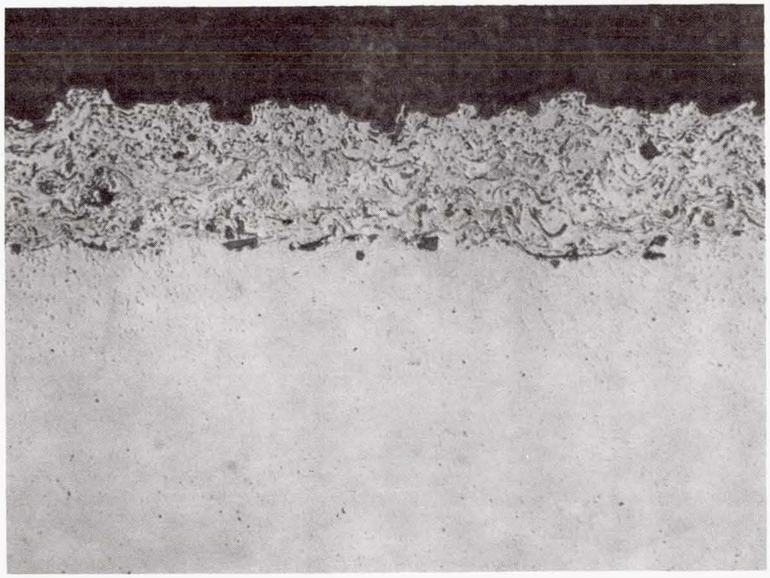
(a-1) As coated.



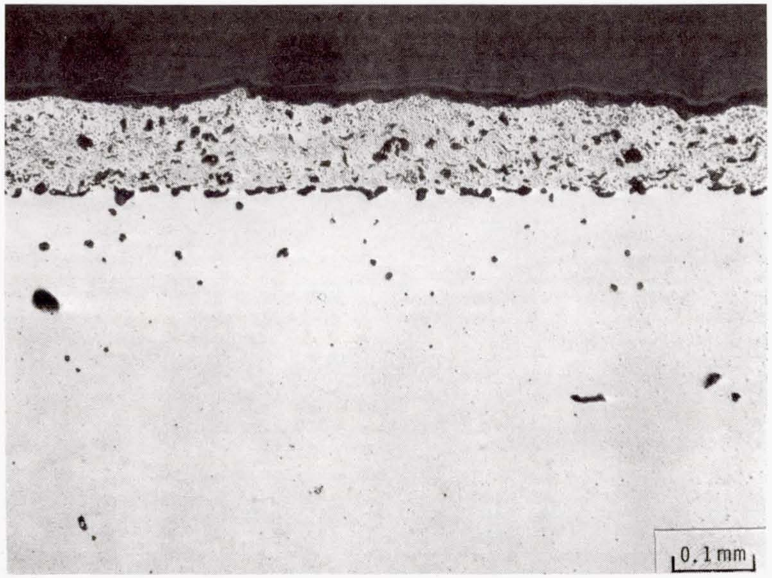
(a-2) As tested along one flat surface shown in figure 8 (b).

(a) Ni-18 Cr-12Al-0.3Y-coated MA 754.

Figure 9. - Photomicrographs of plasma-spray-coated MA 754 before and after 300 hr of Mach 0.3 burner rig oxidation at 1150°C; 1-hr cycles.



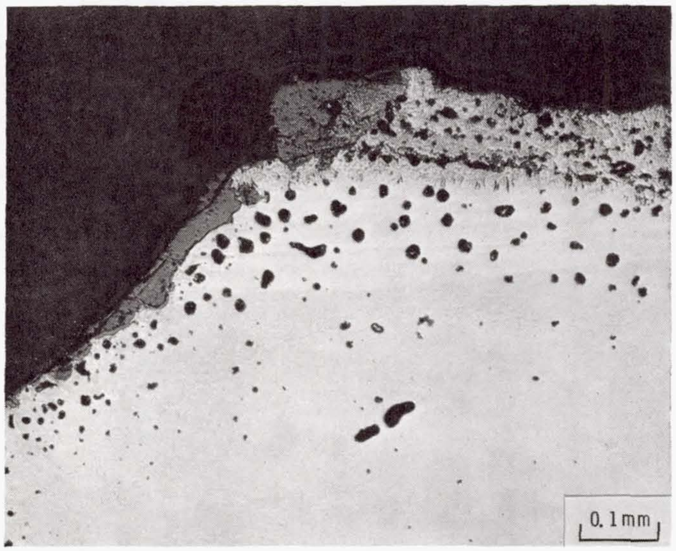
(b-1) As coated



(b-2) As tested along one flat surface shown in figure 8 (C).

(b) Ni-18 Cr-12Al-0.3Y (top) plus Ni-16 Cr-8Al-0.5Y-8Y_2O_3 (base) coated MA 754.

Figure 9. - Continued.



(c) Ni-18 Cr-12Al-0.3Y (top) plus Ni-16 Cr-8Al-0.5Y-8Y_2O_3 (base) coated MA 754 as tested along flat surface near trailing edge (fig. 8(c)), showing coating wearout.

Figure 9. - Concluded.

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16. Abstract

A new concept for enhanced protection of superalloys consists of adding an oxidation- and diffusion-resistant cermet layer between the superalloy and the outer oxidation-resistant metallic alloy coating. Such a duplex coating was compared with a physical-vapor-deposited (PVD) NiCrAlY coating in cyclic oxidation at 1150° C. The substrate alloy was MA 754 - an oxide-dispersion-strengthened superalloy that is difficult to coat. The duplex coating, applied by plasma spraying, outperformed the PVD coating on the basis of weight change and both macroscopic and metallographic observations.

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