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MELBOURNE, VICTORIA

Technical Report 27

**FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO T-56
TURBINE BLADES AFTER 2500 HOURS OF ENGINE USE**

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by

S.G. RUSSO

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**FIELD EVALUATION OF SIX PROTECTIVE COATINGS APPLIED TO T-56
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SUMMARY

This report outlines an assessment of the performance of six different aluminide coatings applied to Allison T-56 high-pressure turbine blades after 2500 hours of engine operation. The results confirm the superior resistance of the precious metal (platinum and platinum/rhodium) modified aluminide coatings to high-temperature oxidation and hot-corrosion, as previously indicated by the earlier progressive evaluations at 500, 1500 and 2000 hour intervals.

Special attention was given to the use of qualitative analysis using energy dispersive X-ray spectroscopy. This technique identified substantial residue quantities of precious metal (Pt and Pt/Rh) in their respective aluminide coated surfaces. The results show that the precious-metal modified aluminides are capable of affording protection to the T-56 substrate superalloy designated IN-738LC up to, and hopefully beyond, the desired 3000 hour interval.



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1. INTRODUCTION

The advent of the modern gas-turbine has resulted in a whole new era for materials technology. In order to increase thermal efficiency and provide greater engine thrust, the gas-turbine industry was given tremendous impetus to search for advanced materials capable of withstanding higher operating temperatures. However, the increased mechanical properties necessary in these materials could only be achieved by reducing the amount of chromium in the superalloy, resulting in a decrease in the oxidation and corrosion resistance. Other adverse effects caused by raising the operating temperature of the gas-stream include molten contaminants derived from the fuel and ingested with the air, such as sulphates and chlorides. In a high velocity oxidising gas-stream, containing sulphurous gases, these factors result in accelerated oxidation aggravated by hot-corrosion¹ (and also erosion) leading to premature failure of high-temperature engine components.

The most common approach to combat the premature deterioration by hot-corrosion of gas-turbine components, particularly turbine rotor blades and nozzle guide vanes, has been the application of protective coatings. Since their introduction in the late 1950's, nickel-aluminide coatings have provided the vast majority of aero-engines with added protection. This has been achieved through the formation of protective oxide barriers (usually Al_2O_3 and Cr_2O_3) which inhibit the subsequent reactions between the coating and the environment.

The extreme conditions that high-pressure turbine (HPT) blades are subjected to in aircraft turbine engines of today has lead to a whole new array of coating systems. Modified nickel-aluminides based on silicon, chromium and, more recently, precious metals (platinum and rhodium), have been developed and shown to improve the corrosion and oxidation resistance over the conventional nickel-aluminides.

It has been revealed [1] that the current coating used on the RAAF T-56 engines, a conventional nickel aluminide, has a greater than 60% rejection rate after 2000 hours. Accordingly, an investigation was undertaken to determine the benefits of five alternative protective coatings applied to the first-stage high-pressure turbine blades in the engines of two long range maritime patrol aircraft operated by the Royal Australian Air Force (RAAF). The excellent durability of the precious-metal modified aluminide coatings applied to the HPT blades of the RAAF Allison T56-A-14 engine, whose average inlet temperature reaches 950°C and, at times, exceeds 1077°C [2], has already been documented in two previous reports [3-4]. Other reports relating to hot-corrosion from the engines in RAAF

1 The term 'hot-corrosion' is frequently used to describe accelerated low and high temperature corrosion caused by the combined attack of high temperature exhaust gases from the combustor (which are highly oxidising and usually sulphur-bearing) and condensed phases (commonly ingested with salt from the air intake) on the surface of a component.

and RAN aircraft can be found elsewhere [5-8]. This report provides further evidence of the superior resistance of the precious-metal enriched aluminide coatings after approximately 2500 hours of engine use against other alternative coatings including (i) two nickel-modified aluminides designated A and B; (ii) a silicon-modified aluminide and (iii) an aluminide modified with chromium.

2. EXPERIMENTAL WORK

A total of twelve turbine blades were received from the RAAF Allison T56-A-14 engine (Serial No. AE 107087) after having experienced approximately 2500 hours of engine use. Six turbine blades, one from each coating system, were selected for analyses on the basis of visual inspection, with the most severely degraded blades from each coating system chosen.

Removal of unnecessary deposits including dirt on the turbine blades was achieved by ultrasonically cleaning each blade in acetone. Simple qualitative analyses using Energy Dispersive X-ray Spectroscopy (EDXS) helped to identify the oxides present on the surface of blades before they were sectioned for metallographic examination. In addition, the concave sections of the six differently coated blades were colour photographed for comparative purposes (see Fig. 1).

To guarantee the oxide coatings and any other films present on the surfaces of the blades were kept intact, the specimens were meticulously prepared. This involved, as in the two previous reports, applying an epoxy coating to the surface of the blade followed by sputtering a layer of gold. The conductivity of the gold enabled a nickel coating to be electro-deposited. Nickel, being both hard and resilient to wear, ensures the sectioned blade, upon polishing, is not 'rounded at the edges'. After mounting the blades in a soft epoxy resin, they were sectioned through the aerofoils, near the root platform, at the mid-span and near the shroud. The risk of dissolving water-soluble compounds, frequently present in either the oxide or the aluminide coating, was eliminated by using ethanol in the polishing process. After each blade was polished to a 1 μ m finish, Scanning Electron Microscopy (SEM) and EDXS equipment were both utilised to characterise differences in the coatings.

3. RESULTS

3.1 Preliminary Analyses

Visual inspection confirmed that every turbine blade examined, irrespective of coating type, had degraded in specific areas. Colour photographs presented in figure 1 illustrate the large variations in oxidation/corrosion attack for the six different coating systems and support the conclusion that the most extensive area of attack occurs at the mid-span of the concave leading edge. Areas of severe attack on the concave face, identified by regions black or green in colour, were analysed by EDXS and found to contain excessive amounts of nickel. This could be explained by the formation of NiO indicating that the protective Al_2O_3 derived from the aluminide coating and also from the aluminium in the substrate, had been exhausted. Nominal attack on the low-pressure convex side of the blades can be explained by the fact that these areas do not experience direct contact with the incoming airstream, commonly contaminated with sulfur, salt and foreign objects. Moderate erosion noted on the trailing edges of the blades may have been caused by the excessive temperatures and the high stress concentrations experienced in these regions. Temperatures reaching 1070°C are not uncommon on the trailing edge, and this could result in an increased rate of high-temperature oxidation.

The SEM was used to determine variations in degradation of the six coatings at the most severely attacked areas at the mid-span of the concave leading edge. Micrographs at a magnification of 800 times for each individual coating are shown in figure 2. Detailed micro-analyses of the six coatings, shown in figure 3, provide the following information :

- Detrimental sigma (σ) phase ² was located in the majority of the coatings at the interface of the diffusion zone and underlying matrix. Absence of any σ -rich phases in the underlying substrate of the silicon-modified aluminide can be partly explained by the fact that silicon suppresses its formation.
- Severely attacked blades contained high proportions of NiO at the surface.
- The surface of the precious-metal enriched aluminides contained appreciable amounts of platinum and rhodium-rich compounds with little sign of destruction. As indicated from laboratory crucible tests and pilot tests [9], the PtAl_2 phase resists sulphidation under gas turbine operating conditions.

² A hard and brittle phase that forms in platelets, represented by $(\text{Cr},\text{Mo})_x(\text{Ni},\text{Co})_y$ x,y b/w 1-7 x/y -1. The platelets tend to facilitate propagation of cracks and oxides through the matrix. In addition, the alloy is weakened by the removal of solid solution strengtheners including Mo and Cr.

- The diffusion zone was clearly identified in the aluminides enriched with precious metals. Although it was observed in two of the remaining coatings, the existence of such a diffusion zone was only nominal. Diffusion zones of any significance in the concave leading edges of nickel aluminide (B) and chromium-modified aluminide were not detected.

3.2 Coating Microscopy and Microanalyses

A brief description of the state of each coating after more than 2500 engine flight hours, using the micrographs of figure 2 coupled with the micro-analyses from figure 3, is presented below.

3.2.1 Nickel aluminide (Type A)

Few areas around the mid-span of the blade were found which had not been attacked. These areas included the convex side and redundant sections of the concave trailing edge. Penetration of needle-like oxides into the underlying substrate is clearly evident (see Fig. 2a). The thick oxide layer, together with the oxide needles, were predominantly Al_2O_3 with small amounts of chromium, titanium and nickel oxides. Remains of the initial β -NiAl phase are visible on the left side of micrograph 2(a). Small precipitates measuring less than $1\mu\text{m}$ diameter at the corrosion front were found to contain both chromium and molybdenum and therefore identified as the α -Cr phase. Bright precipitates containing heavy elements such as tantalum, niobium, molybdenum and tungsten were clearly evident in the diffusion zone and its surrounding areas. Although the σ -phase is not evident in figure 2(a), it was observed in numerous regions throughout the coating. Sulphides rich in chromium were located in the vicinity of the corrosion front.

3.2.2 Nickel aluminide (type B)

The mode of attack upon this coating appeared very similar to that of the previous coating. Severe oxidation was apparent over the entire area of the concave face. The extent of the attack was so harsh that there were areas, in the vicinity of the concave leading edge, where the entire diffusion zone had been consumed (see Fig. 2b). The oxide scale and its needle-like protrusions consisted of Al_2O_3 and Cr_2O_3 with smaller amounts of TiO_2 and NiO . Large blocky precipitates rich in Al_2O_3 at the oxide front typified a coating that no longer provided protection to the underlying substrate. Near to the coating/substrate interface, chromium-rich compounds including α -Cr and the σ -rich phase were in abundant supply. No explanation can be given for the lack of sulphides which normally would be expected in these seriously degraded areas.

3.2.3 Chromium-modified nickel aluminide

Figure 2(e) demonstrates an area that had suffered a high degree of degradation. It is very difficult to locate any region, on the concave face of this coating, that had not experienced severe attack. The surface of the blade had been extensively oxidised to a depth exceeding 30 μ m. This was in addition to the material already lost through spallation. The extent of degradation can be appreciated by the mere fact that tungsten and tantalum, major constituents of the diffusion zone, were both identified in the oxide scale region as tabulated in figure 3. One major difference between the oxide scale in this coating compared to the previous coatings was the abundance of light grey regions within the scale. Qualitative analyses revealed these areas to be rich in nickel and chromium, suggesting a (NiO/Cr₂O₃) spinel phase. Similarly, as with the conventional nickel aluminide coatings, large blocky Al₂O₃ phases were in plentiful supply around the surrounding areas. The original diffusion zone, containing phases rich in chromium, titanium, tantalum and tungsten, could not be located apart from a relatively small area of the trailing edge.

3.2.4 Silicon-modified nickel aluminide

The silicon-modified aluminide coating was markedly different from the previous three coatings. The absence of any needle-like penetrations suggested that this coating was capable of preventing or at least deferring the onset of high-temperature hot-corrosion. Oxides present on the surface of the coating included aluminium, chromium and nickel. Other oxides based on titanium and silicon were present in smaller quantities. However, large β -NiAl grains had been consumed by the accelerated oxidation processes occurring in the coating (see Fig. 2d). Clear evidence of a diffusion zone on this blade section was not easy to distinguish. There were nevertheless, compounds rich in chromium, tantalum and niobium in the interfacial region, in what can best be defined as a narrow diffusion zone. Degradation of this coating appeared slow, and the absence of any sulphide indicates that high-temperature oxidation was the principal mode of degradation. However, while the rate of coating degradation was low the coating itself is thin, compared with other aluminides, and is thus unlikely to provide protection beyond 3000 hours of engine operation.

3.2.5 Platinum-modified nickel aluminide

A relatively thin oxide scale, consisting almost entirely of α -Al₂O₃, was present over the entire surface of the coating. Apart from the thin oxide scale, the blade appeared to be unaffected by any form of degradation process. Even the micrograph in figure 2(e) which illustrates the area of maximum attack on this coating showed little, if any, evidence of oxidation. Figure 4(b) illustrates a more typical area of the coating microstructure, where a range of various phases are easily distinguished. Typical platinum-modified aluminide coatings contain an outer zone consisting of PtAl₂ in a hyper-stoichiometric β -(Ni,Pt)Al (aluminium rich) matrix [10]; this coating was no exception. Platinum was not detected at depths exceeding 30 μ m, suggesting the inner zone is rich in the hypo-stoichiometric β -

NiAl-(nickel rich) phase. Even though the diffusion zone had been consumed quite markedly when compared with that reported in the initial paper [3], it was still in a more than satisfactory state. The major phase present in the diffusion zone was α -Cr, identified as relatively large, dark and discrete areas. Other phases present in this region included the bright refractory metal carbides such as TaC, NbC, TiC and Mo_6C ; the γ - $\text{Ni}_3(\text{Al,Ti})_2$ and the deleterious σ -phase which is easily identifiable as bright platelets below the diffusion zone (see Fig. 2e).

3.2.6 *Platinum/Rhodium-modified nickel aluminide*

It must be emphasised that the micrograph depicting the platinum/rhodium modified aluminide (see Fig. 2f) represents an area that had experienced a maximum degree of attack. Areas resembling severe attack accounted for less than 1% of the surface. Unlike the chromium and nickel aluminide coatings, this micrograph is not characteristic of the coating as a whole. A more representative area is illustrated in figure 4(c). However, to judge each coating correctly requires comparisons to be made of the worst affected areas. Constituents in the oxide barrier were similar to those of the previous coating with the majority of phases rich in Al_2O_3 , PtAl_2 and Cr_2O_3 . Rhodium and platinum, identified by bright regions in the micrograph, were abundant over the entire concave face. Even though moderate attack was apparent in figure 2(f), it did not resemble any of the common forms of hot-corrosion. High-temperature oxidation appeared to be the major cause of attack. A two phase region consisting of α -Cr and refractory metal carbides (Mo, Ti, Ta, Nb) existed in the diffusion zone.

3.2.7 *Identification of Hot-corrosion*

Sulfur-containing compounds, abundant on any surface where hot-corrosion is prevalent, were identified in localised areas on the two conventional aluminides and the chromium-modified aluminide coating. Their presence originates from the molten Na_2SO_4 on the surface of the blades. Sodium chloride (ingested by the engine from air) at high temperatures is converted by SO_2 and SO_3 (from combustion of sulfur impurities in fuel) to Na_2SO_4 . A micrograph and corresponding EDXS analysis of an area at the corrosion front (see Fig. 4a), confirms the presence of sulfur deep within the degraded coating. Sulfur is easily transported inside the coating along grain boundaries and oxides where it maintains a self-sustaining mode of attack. Alternatively, it can penetrate oxide products and reactions are then further enhanced if molten sulfates are formed. Chloride, when present, disrupts the scale and facilitates easier access of the corrodant species to the underlying alloy surface [11].

A comparison between the well protected platinum/rhodium-modified aluminide and the extensively attacked conventional nickel aluminide is illustrated in figure 5. Although spallation was evident in the former coating, the lack of extensive substrate oxidation,

identified by dark blue/green deposits, demonstrated the superiority of the platinum/rhodium modified coating (see Fig. 5a). Alternatively, the conventional nickel aluminide illustrated in figure 5(b) shows areas where extreme oxidation had taken place. EDXS analysis confirmed the presence of large amounts of NiO on the surface³ of the blade.

4. DISCUSSION

It has been shown here, as well as in previous reports [3-4], that on their own, simple macroscopic evaluations of blade coating performance can falsely classify a particular coating. Only by complementing the macroscopic results with detailed microanalysis, including SEM, can a better appreciation of coating performance be attained. For instance, the apparent spallation that appears on the blades in figure 1(e) was found to be, after detailed SEM analysis, superficial erosion of the protective Al_2O_3 , Cr_2O_3 and PtAl_2 scale. The modified-aluminide coating was still intact with no significant indication of premature failure.

Diffusion coatings, of the kind examined thus far, provide surface enrichment of aluminium, chromium or silicon which results in the formation of protective surface oxides of Al_2O_3 , Cr_2O_3 or SiO_2 . These more stable oxides form a barrier between the environment and the coating. However, even with the added oxide scale protection, turbine blades continue to degrade by chemical and/or mechanical processes that promote scale failure.

The role of the coating, whether it be a conventional aluminide coating or a more complex modified-aluminide coating, is to provide a large reservoir of scale-forming elements such that rapid repair of stable surface oxides can occur. However, it is inevitable that a stage will eventually be reached where the reservoir is completely consumed. When this occurs, elements in the coating and substrate, whose role is to provide strengthening and not scale protection, are preferentially oxidised. These elements include nickel, titanium and the refractory metals of molybdenum, tungsten, tantalum and niobium. It has been well documented [12] that refractory metals, including molybdenum and tungsten, will readily oxidise to form volatile low melting point oxides. Additionally, titanium will readily oxidise to form rutile (TiO_2) which can be unfavourable under certain conditions by causing oxide scale breakdown due to its fast rate of growth. Every coating, except the precious-metal modified aluminides had already reached this stage of deterioration.

³ NiO has a low energy for oxide formation and thus will only form in appreciable amounts when the more thermodynamically stable oxides of aluminium and chromium have been depleted.

The superior performance of the precious-metal modified aluminides based on rhodium and/or platinum can be explained by their following unique characteristics that have been well documented:

- (i) they have high solubilities with the substrate metal;
- (ii) the difference in the a^0 spacing⁴ between the precious metal and aluminium is large, resulting in the obstruction of inward diffusion paths;
- (iii) they form high melting point compounds with good resistance to high-temperature oxidation and corrosion on the surface of the blade;
- (iv) they enhance the adherence of oxide scales [13].

The results shown in figure 4(b,c) confirm the presence of a residue of these precious metals near the surface of the blades even after service conditions had exceeded 2500 engine hours. This is a major advantage when one considers that extensive attack was already apparent on the currently employed nickel aluminide after only 1500 hours of engine usage. It is important to mention that the micrograph of the platinum-modified aluminide illustrated in figure 2(f) was not a representative area. This micrograph was based only on a small area (less than 1%) that had suffered preferential oxidation.

A simple and cost effective technique involving a silicon enriched slurry is commonly used to apply the silicon-modified aluminide coating to engine components. Consequently, the slurry can be used as a repair or 'touch-up' coating on areas of the aluminide coating that are experiencing premature degradation. This procedure has already been adopted by several civil operators to increase service life and reduce maintenance costs of engine components.

Difficulties in locating sulfur in the corrosion front may indicate that accelerated high-temperature oxidation, rather than hot-corrosion, was the dominant degradation mechanism. Even after extensive micro-analyses of a large array of corroded regions, only a small sulfur peak, shown in figure 4(a), could be recorded.

By collating all the results from the previous three reports [1,3,4] and the present one, a graph has been constructed to illustrate the gradual deterioration of each coating system. Minimum coating thickness was plotted as a function of engine operating hours and is shown in figure 6. It clearly demonstrates the superiority of the precious-metal enriched aluminides. The small discrepancy in the silicon-modified aluminide is based upon the fact that two different engines were used in this investigation and service conditions may

4 This parameter is termed the crystal lattice parameter which measures the distance between atoms in the unit cell.

not have been the same. The graph also provides a means of ranking the coatings in order of corrosion and/or oxidation resistance. Overall ranking of the six coatings is shown in Table 2.

Although the superior corrosion resistance of the precious-metal enriched aluminides compared to the conventional aluminides is beyond doubt, other parameters must be examined. A selection of these are presented in Table 1. One major drawback with the precious-metal enriched aluminides is their high cost and high ductile to brittle transition temperature (DBTT), rendering them susceptible to cracking, particularly as a result of foreign object impact. Brittle phases possessing properties in common with these types of coatings have been examined [14] and found to be prone to initiating creep or fatigue cracks. It has been suggested by many authors that such coatings be restricted to applications where radial strains or cyclic strain ranges remain small.

5. CONCLUSIONS

Based upon results obtained, both in this and previous reports [1,3,4], the following conclusions can be drawn.

1. A range of environmental degradation processes occur in aircraft gas-turbine engines. These include oxidation, corrosion and erosion, causing disruption to the integrity of high temperature components.
2. Corrosion of turbine blades and vane aerofoil components leads to decreased component lifetimes and loss of component shape (and hence aerodynamic efficiency).
3. Maximum extent of hot-corrosion occurs on the mid section of the concave leading edge of high-pressure turbine blades.
4. After 2500 hours of engine operation, the condition of the first stage high-pressure blades with the six different coatings, ranked from best to worst, was platinum/rhodium-modified aluminide >platinum-modified aluminide >silicon-modified aluminide >conventional nickel aluminide (Type A) >conventional nickel aluminide (Type B) >chromium-modified aluminide.
5. Platinum and rhodium were still providing protection to the surface of the turbine blades after more than 2500 hours of engine use.

6. There was no evidence of cracking in any of the six coating systems although spallation, especially after prolonged service, was apparent.
7. The anomalies which arose when comparing the macrographs and the micrographs of the platinum-modified aluminides confirm the importance of micro-structural and microanalysis: visual inspection alone can give misleading assessment of coating performance.

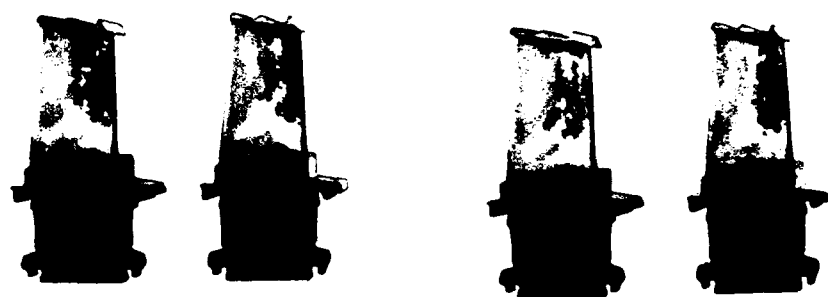
6. ACKNOWLEDGMENTS

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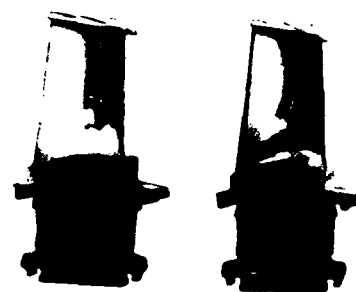
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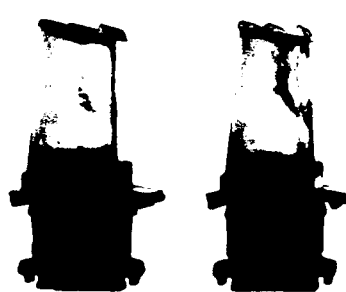


(a)

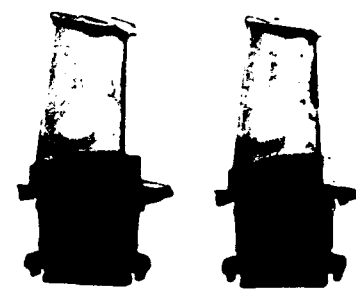
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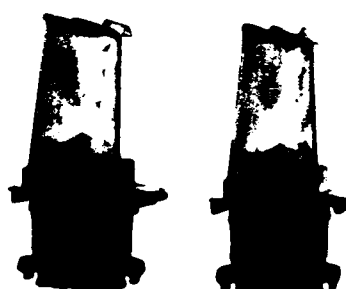
(c)



(d)



(e)



(f)

20 mm

Figure 1. Colour photographs of the concave face for the six different coatings (a) nickel aluminide (A); (b) nickel aluminide (B); (c) chromium-modified aluminide; (d) silicon-modified aluminide; (e) platinum-modified aluminide and (f) the platinum/rhodium-modified aluminide.



(a) Nickel aluminide (Type A)



(b) Nickel aluminide (Type B)



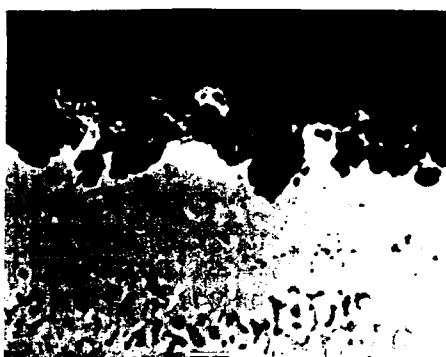
(c) Chromium-modified aluminide



(d) Silicon-modified aluminide

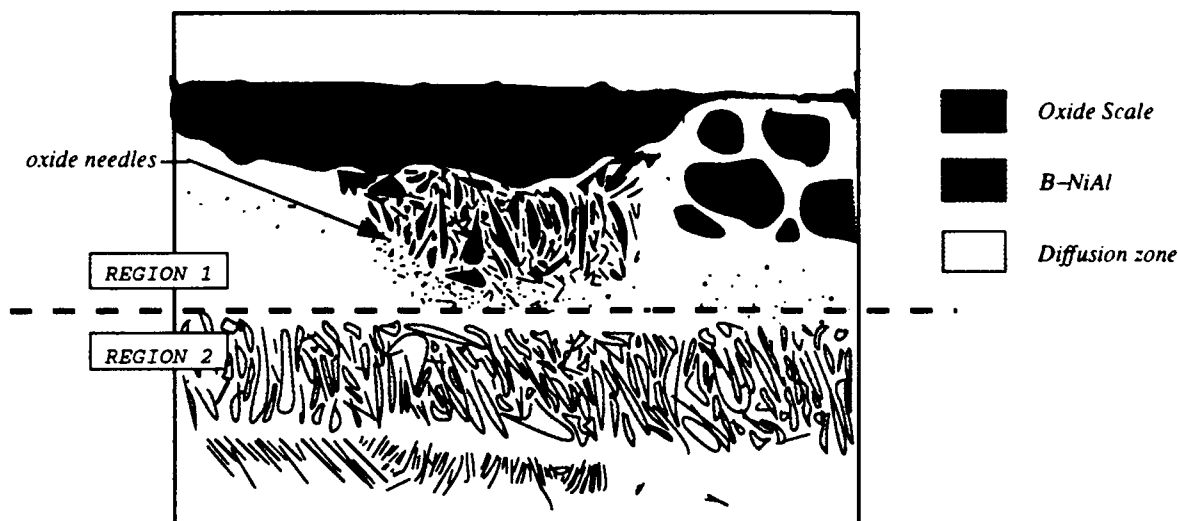


(e) Platinum-modified aluminide



(f) Platinum/Rhodium modified aluminide

Figure 2 Scanning Electron micrographs showing maximum areas of attack at the mid-span cross-section.



	REGION 1			REGION 2	
	OXIDE SCALE	OXIDE NEEDLES	SMALL PPTTES/ PHASES	DIFFUSION ZONE	UNDERLYING MATRIX
NI ALUM. (A)	VD- (Al,Cr,Ti,Ni) D- (Al,Ni,Cr,Ti)	D- (Al,Ni,Cr,Ti) B- (Ni,Cr,Ti)	D- (Cr,Mo) $\rightarrow$ α -Cr B- β -NiAl grains VB- (Ti,Ta,Nb,Al)	D- (Ni,Al,Cr,Co) B- (Ti,Cr,Ni,Ta,Nb,Al) VB- (Cr,Ni,Mo,W)	D- (Cr,Mo,Ni,Co) $\rightarrow$ σ phase
NI ALUM. (B)	VD- (Al,Cr,Ti,Ni) D- (Al,Ni,Cr,Ti)	D- (Al,Ni,Cr,Ti)	D- (Ti,Ni,Cr,Nb,Ta) VD- $\rightarrow$ α -Cr	Could not be located at CCLE area	D- (Cr,Mo,Ni,Co) $\rightarrow$ σ phase
CR- ALUMINIDE	VD- (Al,Cr) D- (Ni,Cr,Al,Ti) B- (W,Ta)	D- (Al,Ni,Ti,Cr)	VD- (Al,Ni,Ti,Cr) D- (Ti,Cr)	Could not be located at CCLE area	D- (Cr,Mo,Ni,Co) $\rightarrow$ σ phase VB- (Ta,Nb,Ti)
SI- ALUMINIDE	VD- (Al,Cr,Ni,Ti) D- (Ni,Cr,Al,Ti)	Not Present to any appreciable amount	B- β -NiAl grains	B- (Ti,Cr,Ni,Al) VB- (Ta,Nb,Ti,Cr)	VB- (Ta,Nb,Ti)
PT- ALUMINIDE	VD- (Al,Ti,Cr,Ni) D- (Al,Ni,Cr,Ti) B- (Pt,Rh)	Not Present to any appreciable amount	B- β -NiAl grains VB- β - (Ni,Pt)Al and PtAl ₂	D- (Cr,Mo,Ni) $\rightarrow$ α -Cr B- (Ti,Ta,Nb) VB- (Ta,Nb,Ti) carbide	D- (Cr,Mo,Ni) $\rightarrow$ α -Cr
PT/RH- ALUMINIDE	VD- (Al,Ti,Cr,Ni) D- (Al,Cr,Ti,Ni) B- (Pt)	Not Present to any appreciable amount	VD- inclusion/voids B- β - (Ni,Pt)Al VB- (Rh,Al,Pt)	D- (Cr,Mo,Ni) $\rightarrow$ α -Cr B- (Ti,Ta,Nb) VB- (Ta,Nb,Ti) carbide	D- (Cr,Mo,Ni,Co) $\rightarrow$ σ phase B- (Cr,Mo,W)

Figure 3. Schematic representation of a typical serviced diffusion coating divided into two distinct regions, defined as : Region 1 - containing an oxide scale, protruding oxide needles together with a collection of small precipitates and a variation of phases and Region 2 - a diffusion zone with an underlying matrix. Also shown below is a corresponding EDS (Energy Dispersive Spectroscopy) table identifying elements present in each sub-region for the six coatings. ⁵

5 [D=Dark; VD=Very Dark; B=Bright; VB=Very Bright; Ox=Oxide; pp=precipitate; $\rightarrow$ =suggests)

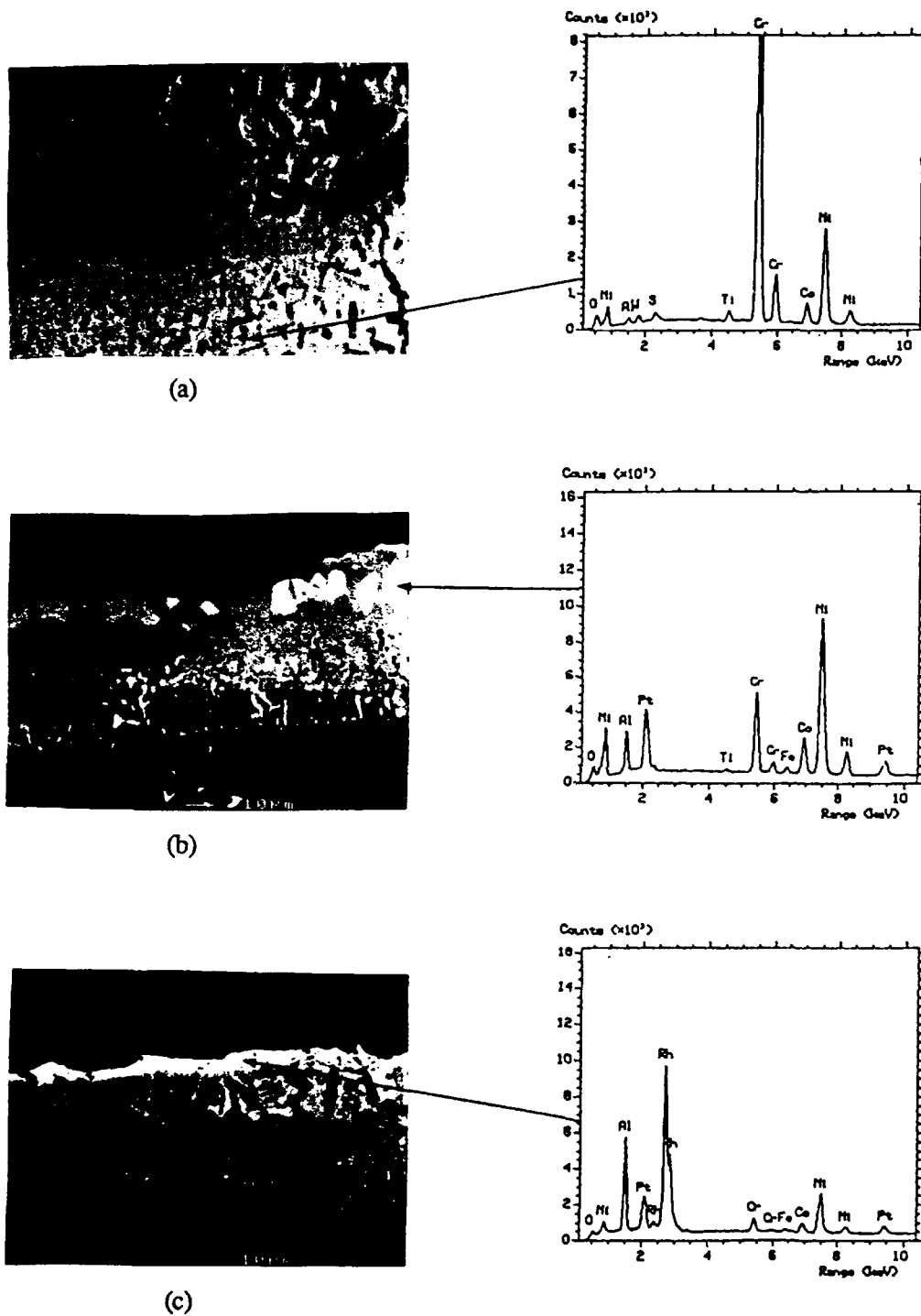


Figure 4 Micrographs and corresponding EDXS analyses of (a) an area containing sulfur located near the corrosion front of a badly attacked blade; (b) a platinum-rich area near the surface of the platinum-modified aluminide coating and (c) a rhodium/platinum-rich area near the surface of a platinum/rhodium-modified aluminide coating.

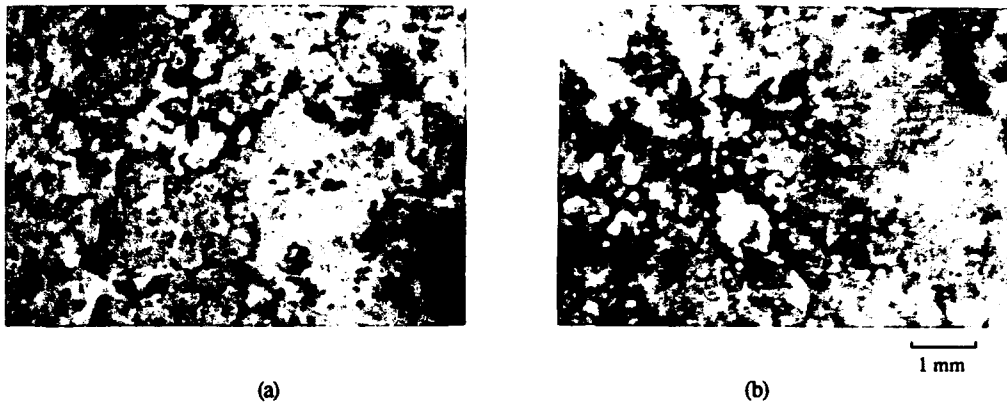


Figure 5 Differences in the surface integrity of (a) a platinum/rhodium modified aluminide and (b) a conventional nickel aluminide. The appearance of dark regions in (b) indicates that the aluminium has been exhausted resulting in the preferential oxidation of nickel. These dark regions are only superficial in (a). Micrographs were taken from areas with maximum degrees of attack near the concave leading edge.

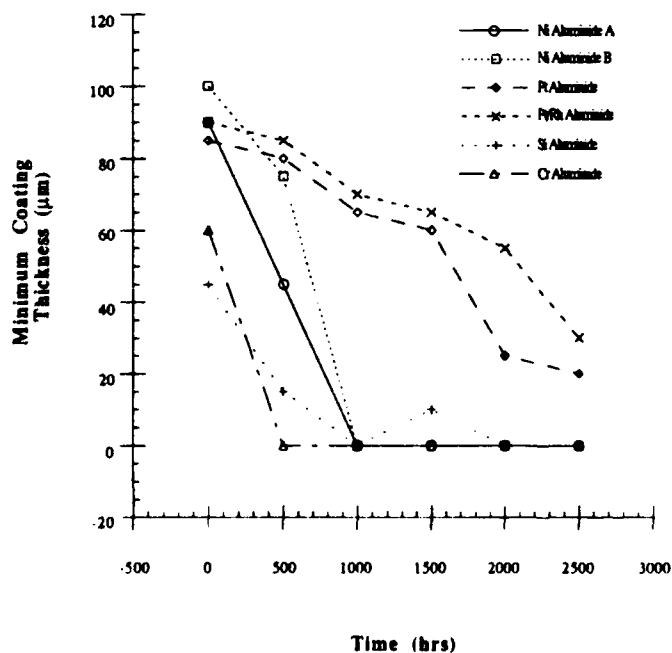


Figure 6. Minimum average coating thickness for each coating as a function of engine flight hours.

	COST EFFECTIVE	COAT COMPLEX SHAPES	CORROSION RATE	SURFACE FINISH	DUCTILITY OF SCALE	ABILITY TO MASK	SCALE GROWTH RATE	DBTT OF COATING
CONVENTIONAL ALUMINIDE (A)	LOW	YES	FAST	GOOD	FAIR	YES	FAST	MODERATE
CONVENTIONAL ALUMINIDE (B)	LOW	YES	FAST	GOOD	FAIR	YES	FAST	MODERATE
CR-MODIFIED ALUMINIDE	MODERATE	YES	FAST	FAIR	GOOD	NO	MODERATE	MODERATE
SI-MODIFIED ALUMINIDE	LOW	NO	MODERATE	POOR	POOR	YES	SLOW	HIGH
PT MODIFIED ALUMINIDE	HIGH	YES	SLOW	GOOD	GOOD	NO	SLOW	HIGH
PT/RH MODIFIED ALUMINIDE	HIGH	YES	SLOW	GOOD	GOOD	NO	SLOW	HIGH

Table 1 Coating comparisons of the six coating types studied throughout the investigation.

COATING TYPE	NO. OF HOURS OF ENGINE USE (APPROX. ONLY)						OVERALL RANK
	0	500	1000	1500	2000	2500	
Pt/Rh-mod. Aluminide	90	85	70	65	55	30	1
Pt-mod. Aluminide	85	80	65	60	25	20	2
Si-mod. Aluminide	45	15	0	10	0	0	3
Ni Aluminide (B)	100	75	0	0	0	0	4
Ni Aluminide (A)	90	45	0	0	0	0	4
Cr-mod. Aluminide	60	0	0	0	0	0	5

Table 2. Minimum Coating Thickness (μm) and overall ranking for the six coatings as a function of engine flight hours.

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16. ABSTRACT <i>This report outlines an assessment of the performance of six different aluminide coatings applied to Allison T-56 high-pressure turbine blades after 2500 hours of engine operation. The results confirm the superior resistance of the precious metal (platinum and platinum/rhodium) modified aluminide coatings to high-temperature oxidation and hot-corrosion, as previously indicated by the earlier progressive evaluations at 500, 1500 and 2000 hour intervals.</i>			