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The influence of high temperature on DV-2 jet engine Ni-based superalloy turbine blade degradation

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Abstract

Blades and vanes made from Ni – and Co – base superalloys that are used in the hot section of all gas turbine engines are coated to enhance resistance to hot corrosion. The most widely used coatings are those based on the intermetallic compounds NiAl and CoAl, which are formed by diffusion interaction of aluminium with surfaces of the nickel and cobalt alloys, respectively. Simple aluminide coatings resist high – temperature oxidation by the formation of protective layer of alumina and can be used up to about 1 150°C. The coatings degrade by loss of aluminium due to spalling of oxides under thermal cycling conditions. HPT blades of DV – 2 jet engines are made from Ni – based superalloy ŽS6K. For improving alloy's high temperature resistance are blades coated with Al – Si diffusion layer. Depending on jet engine flying regime an overcrossing of working temperature may occur. It may result into coarsening of the gamma prime particles, thus decreasing its strengthening effect. A real case study of DV-2 jet engine turbine blades after high temperature exposition is discussed in this article.

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Keywords: High pressure turbine blade; Ni-base superalloy ŽS6K; Al-Si protective coating; gamma prime coarsening; working temperature overcrossing; SEM observation; quantitative analysis of protective coating;

1. Introduction

The aero jet engine DV – 2 is used as a drive unit for convertible jet fighter Aero L – 39 MS or Yak – 130. There are various advanced materials, such Ti alloys and Ni – base alloys used for its construction. A turbine of this engine

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Nomenclature

DV - 2	aero jet engine used in Russian double seats training fighter YAK 130
HPT	High Pressure Turbine blades are covered with Al-Si protective layer
LPT	Low Pressure Turbine blades, there are two stages of low pressure turbine, but only 1st is with Al-Si protective layer
SEM	Scanning Electron Microscopy
FCC	Face Cubic Centered, or known as K12 lattice, typical lattice for gamma solid solution in Ni-based superalloys (Ni+Co or Ni+Cr respectively - γ)
$L1_2$	Lattice of coherent hardening phase known in Ni-based superalloys as gamma prime (Ni_3Al (Ti – respectively) - γ')
MC	Primary carbides, in Ni-based superalloys formed by Cr, Ti, Ta or Nb and situated inside the grains
$M_{23}C_6$	Secondary or complex carbides, in Ni-based superalloys formed mostly by Cr and situated at the grain boundaries

consists of two stages. The first one is high pressure turbine (HPT) and the second are two low pressure turbines (LPT). Blades of all turbine discs are made from cast Ni – bases superalloy ŽS6K and blades of HPT and 1st LPT are coated with protective Al – Si layer, which improves high temperature properties of base material. Working temperature for turbine blades are in range from 705°C at start of the engine to 750°C at regular flying mode [1]. However, during flying mode an overcrossing of working temperature may occur and each one has significant influence on base material as well as protective alitize layer. Especially, when blades have worked for 500 hours or more at regular conditions before overcrossing the optimal working temperature occurred. Of course, there is an overhauling after every 500 flying hours where blades are completely removed from turbine discs and checked for degradation.

The protective Al-Si coatings degrade by loss of aluminium due to spalling of oxides under thermal cycling conditions. Incorporation of reactive elements, such as yttrium and hafnium, by co-deposition during aluminizing [2] can significantly improve adherence of the protective alumina scales and therefore extend protective coating life. At temperatures above about 1000°C interdiffusion of the coatings with substrates contributes significantly to degradation. Practical coating service lives are limited to operating temperatures of 870°C up to 980°C with only short excursion at the highest temperatures. Additions of up to about 5 % Si improve both oxidation and hot corrosion resistance [3]. Silicon can be co-deposited with aluminium by pack cementation [2, 4] and related out – of – contact processes. So – called slurry processes wherein a liquid suspension of aluminium and silicon powders is applied to the alloy surface, then dried and fired at elevated temperatures, can also be used to incorporate silicon [5]. The oxidation and hot corrosion resistance of these coatings are more or less influenced by the composition of substrate alloys. Tantalum and hafnium improve cyclic oxidation and hot corrosion resistance, the latter element by improving the adherence of the protective layer of alumina [6]. Molybdenum and tungsten compromise hot corrosion resistance.

Because of the brittle fracture behaviour of NiAl up to temperatures of 650°C up to 775°C, all aluminide coatings exhibit such fracture below these temperatures while above these limits ductile behaviour occurs [7]. This behaviour can either compromise or enhance thermal fatigue resistance of substrate alloys depending on such complex factors as the exact nature of the thermal cycle and the structure – equiaxed, directionally solidified, or single crystal of the alloys, Fig. 1, [8].

If these effects are limiting, designers may require use of more expensive overlay coatings of the MCrAlY (M = Co and/or Ni) and/ or thermal barrier (zirconia) types.

To summarise all degradation facts, the most common degradation at turbine blades is diffusion changes forced by high temperature and abrasive wearing from particles contained in exhaust gas. This high temperature influences the protective alitize layer where Al_2O_3 oxide is formed at surface what results in increasing the thickness of layer and consequentially decreasing of Al concentration and forming carbides in diffusion area at border layer – base material. In some special areas, where Al_2O_3 concentration is high becomes a very brittle protective layer and is ripped out [9 - 10].

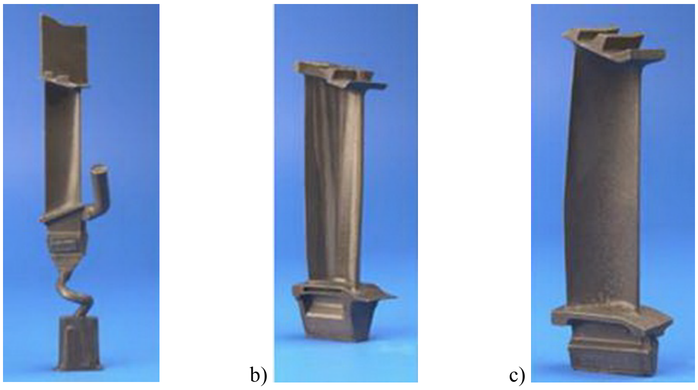


Fig. 1. Examples of nickel – base superalloy blade application: a) as monocrystalline blade; b) directionally solidified; and c) polycrystalline blade

Common degradation of the base Ni superalloy is by coarsening of strengthening γ′ - phase (gamma prime) and forming the rafts. Carbides presented in alloy have tendency to create carbide net ($M_{23}C_6$) on grain boundaries what affect creep rupture life [11 - 13]. The metallography analysis of short – time working temperature overcrossing at engine start up and its influence on protective layer and base material is discussed in this article.

2. Experimental material and methods

The turbine blades of HPT are made by precise casting method from Ni – base superalloy ŽS6K. Alloy ŽS6K is former USSR superalloy used in DV – 2 jet engine and can be roughly compared to Nimonic group alloys. Chemical composition obtained by *SPECTROMAXx* (in wt. %) is *0.2% - C, 4.76% - Co, 3.05% - Ti, 12.44% - Cr, 5.22% - Al, 5.28% - W, 3.48% - Mo, 2% - Fe, 0.4% - Mn, 0.001% - S, 0.001% - P,* and *0.198% - B*. It is used for turbine rotor blade and whole cast small sized rotors with working temperature from 800 up to 1050°C. The alloy is made in vacuum furnaces. Temperature of liquid at casting in vacuum form 1500°C÷1600°C, depends on parts shape and amount. Cast ability of this alloy is very high with only 2%÷2.5% shrinkage.

Blades made of this alloy are also protected against hot corrosion with protective heat proof alitize layer, so there are able to work at temperatures up to 750°C for 500 flying hours.

A typical microstructure of ŽS6K Ni – base superalloy as – cast is showed on Figs. 2 and 3. Microstructure of as – cast superalloy contain dendritic segregation caused by chemical heterogeneity (Fig. 2 (a)) and particles of primary MC and secondary $M_{23}C_6$ carbides (Fig. 2 (b)).

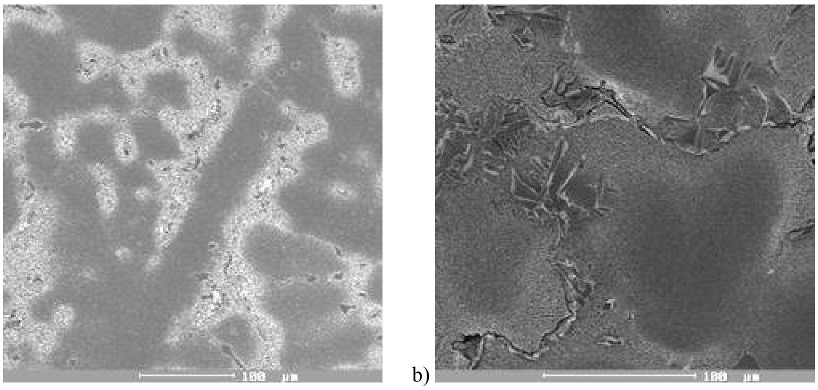


Fig. 2. Microstructure of As – cast Ni – base superalloy ŽS6K: (a) dendritic segregation, (b) primary MC and secondary $M_{23}C_6$ carbides, etch. Marble, SEM

Primary carbides (Ti, Mo, and W)C are presented as an block shaped particles mainly inside of grains. Secondary carbides are presented as a "Chinese" script shape particles on grain boundaries.

However, microstructure also contains solid solution of elements in base nickel matrix – so called γ phase (Ni(Cr, Co, and Fe), an austenitic FCC matrix phase) and intermetallic strengthening precipitates, which are a product of artificial age – hardening and has significant influence on mechanical properties and creep rupture life – so called γ′ phase (gamma prime, Ni_3(Al, and Ti) – ordered coherent precipitate phase with $L1_2$ structure), Fig. 3 (a). The γ′ precipitates are usually present in volume fractions in the range of 20÷60% depending on the alloy, with a typical precipitate size of 0.25 to 0.5 μm for high temperature applications [14]. Both of these phases, γ (gamma) and γ′ (gamma prime) are also creating γ/γ′ eutectic, Fig. 3 (b).

a) b)

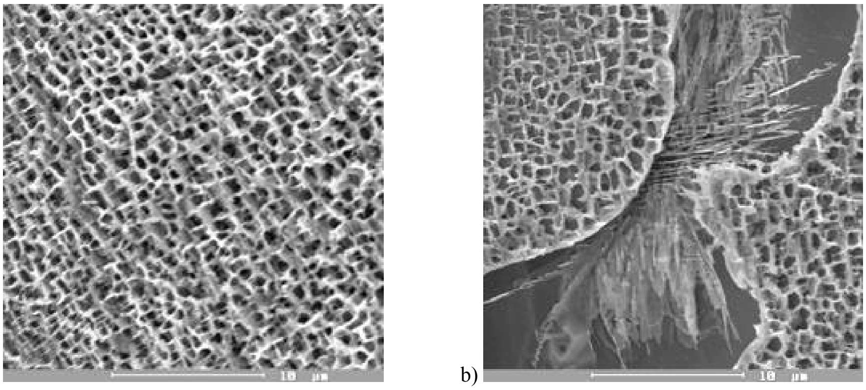


Fig. 3. Ni – base superalloy ŽS6K: (a) morphology of γ′ precipitate, as - cast, (b) eutectic cell of γ/γ′, etch. Marble, SEM

For HPT blades were applied protective Al-Si layer. The procedures of application of such diffusion layer including following annealing is reported in Table 1.

Table 1. An example of a table

Heat – treatment	Conditions
Homogenization annealing	in vacuum, temperature 1225 °C, holding 4 hrs., cooling with argon to 900 °C per 10 min.
Alitize AS2	1. Spraying of AS2 layer; (AS2 – koloxylin solution 350 ml, Al – powder 112 g, Si – powder 112 g) 2. Diffusion annealing; temperature 1000°C, 3 hrs., slowly cooled in retort

Turbine blades have worked 500 hours in regular conditions (after 500 hours blades were checked at overhauling) and after further 100 hours of short time work high temperature overcrossing has occurred. A regular temperature at start up of the engine is 705°C, measured at the end of exhaust pipe; rises for 2 seconds approximately at 751°C. Experimental evaluation consists of protective alitize layer metallography evaluation and metallography of base material. After metallography preparation (grinding, polishing and chemical etching) of specimens for alitize layer degradation was used metallography software NIS Elements [15 – 17]. Observing of alitize layer degradation was focused on four significant parts of blade as leading edge, flap pantile, flow edge, and back; see Fig. 4 for detailed description. Base material was evaluated with SEM and method of chemical mapping was used for identifying of elements in the interlayer area, layer – base material.

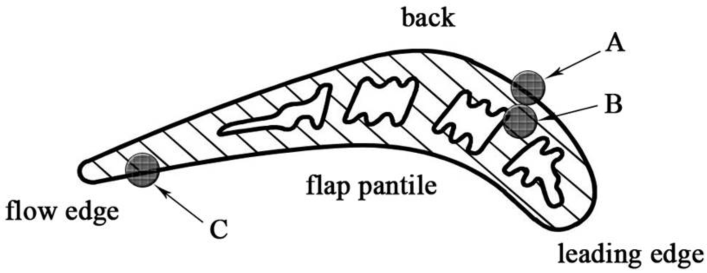
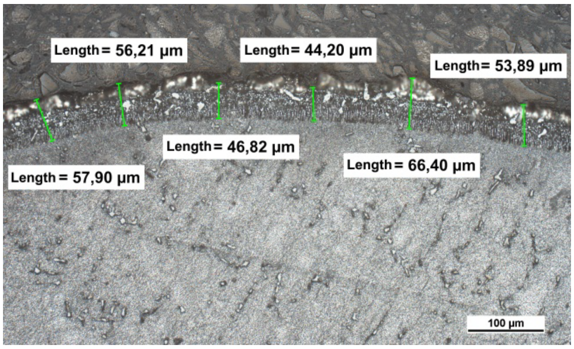


Fig. 4. The cross section of blade and description zones of interests
Area A – the uniformity and homogeneity of alitize layer from 1/4 of blade back.
Area B – morphology and amount of γ' - phase and grain boundary right under alitize layer.
Area C – degradation of alitize layer right before flow edge of blade and needles of Cr rich carbides

3. Results

A three various HPT blades made from ŽS6K with protective Al – Si layer after short – time high temperature overcrossing (751°C/2s) were selected for experimental procedure. Turbine blade analyses consist of:

- metallography evaluation of non-etched specimen for layer thickness measurement and homogeneity of layer on chemically etched specimens,
- metallography evaluation of base material.

a) 

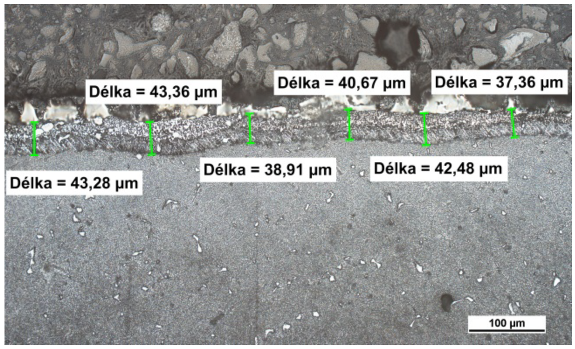
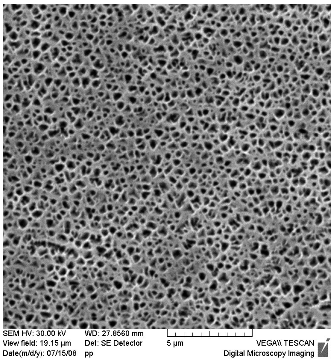
b) 

Fig. 5. Al – Si protective layer thickness measurement, non-etched specimen: (a) leading edge with average coating thickness of 50 µm, (b) flow edge with average coating thickness of 45 µm

Al – Si layer thickness h depending from spraying thickness at its application and can be easily determined from dependence $h = 0.7\delta$, where δ is for spraying thickness. Then for starting stage Al – Si layer thickness is around $h = 24.5$ µm (an optimum spraying thickness for annealing at 950°C is up-to $\delta = 35$ µm). The thickness of Al – Si layer is higher in all observed areas what indicates higher temperature of loading than reported 751°C. Results of analysis are presented on Fig. 5.

The base material, alloy ŽS6K, shows signs of high temperature degradation at grain boundaries where more secondary $M_{23}C_6$ carbides formed and lowering creep resistance of base material. The strengthening intermetallic γ' precipitate also degrades by changing its morphology from previous cubic shape like onto coarse spheroidal particles with average size 0.6 µm, Fig. 6.

The changes in shape and size of γ' precipitate cause decreasing creep rupture life of alloy. As a matter of a fact, creep at high temperatures is driven mainly by diffusion mechanism and deformation rate increases with higher amount of dislocations and their easier slip, which have been previously blocked by tide distribution of cubic morphology of γ' precipitates. Coarsening and spheroiding of γ' precipitates leads to creation of wider channels between γ' particles and dislocations can easier move, so deformation of base material can propagate.

a) 

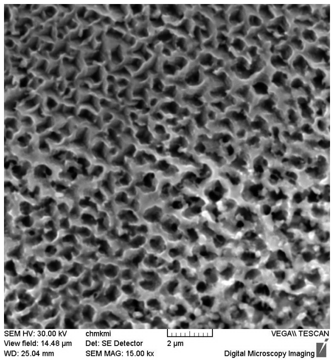
b) 

Fig. 6. Degradation of γ′ precipitate at high temperature (751°C/2s) loading compared to starting stage microstructure: a) γ′ - phase morphology, initial stage, b) after 751°C/2s for 600 hours of loading, etch. Marble, SEM

4. Conclusion

From analyses made on HPT blades of aero jet engine DV – 2 after short time working temperature overcrossing we can conclude:

- Protective Al – Si layer is significantly degraded all around the surface, its thickness on leading edge is average 50 µm, and on the back is 36 µm and flow edge with average thickness 45 µm, so Al – Si layer is inconvenient. A needle like Cr rich carbides and blocky form of primary Ti rich carbides are forming. In this area is a high expectation of crack initiation at further working cycle of blades.
- Base material, Ni – base superalloy ŽS6K, also shows significant level of degradation, especially on strengthening intermetallic precipitate γ′, which has formed rafts and changes it morphology, from cubic to spheroidal shape. The morphology changing of γ′ has influence on creep rupture life of blades due to decreasing of precipitation hardening effect and increasing of γ/γ′ mismatch.

All reported changes indicate that temperature of overheating was not 751°C/2s. This kind of degradation occurred after longer time exposition at temperatures close to 900°C.

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