International Journal of Earth, Energy and Environmental Sciences
ISSN: 2517-942X
Vol:9, No:5, 2015

The High Temperature Damage of DV – 2 Turbine Blade Made from Ni – Base Superalloy

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***Abstract*—**High pressure turbine (HPT) blades of DV – 2 jet engines are made from Ni – based superalloy. This alloy was originally manufactured in the Soviet Union and referred as ŽS6K. For improving alloy's high temperature resistance are blades coated with Al – Si diffusion layer. A regular operation temperature of HPT blades vary from 705°C to 750°C depending on jet engine regime. An overcrossing working temperature range causes degradation of the protective coating as well as base material which microstructure is formed by the gamma matrix and strengthening phase gamma prime (forming small particles in the microstructure). Diffusion processes inside the material during exposition of the material to high temperatures causes mainly coarsening of the gamma prime particles, thus decreasing its strengthening effect. Degradation of the Al – Si coating caused its thickness growth. All the microstructure changes and coating layer thickness growth results in decreasing of the turbine blade operation lifetime.

***Keywords*—**Alitize coating layer, gamma prime phase, high temperature degradation, Ni – base superalloy ŽS6K, turbine blade.

I. 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 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. Of course, there is an overhauling after every 500 flying hours where blades are completely removed from turbine discs and checked for degradation, but anyway what happens with material or protective layer when temperature overcrossing comes after overhauling.

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 influence 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 protective layer very brittle and is ripped out [2], [3]. 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 [4]-[6].

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.

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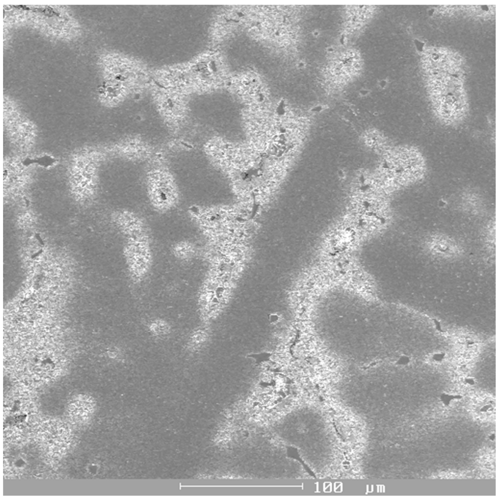
II. Experimental Material and Methods

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. 1 and 2. Microstructure of as – cast superalloy contain significant dendritic segregation caused by chemical heterogeneity (Fig. 1 (a)) and particles of primary MC and secondary $M_{23}C_6$ carbides (Fig. 1 (b)). 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. 2 (a).



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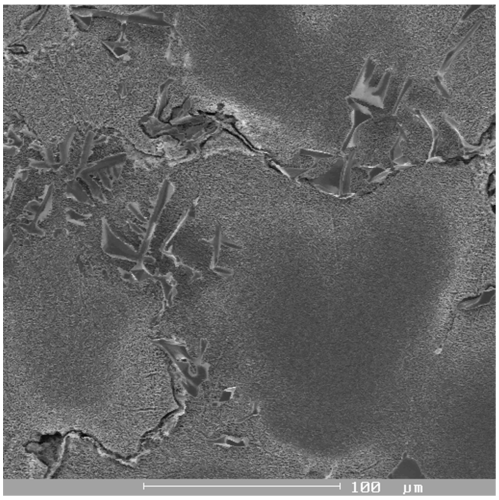

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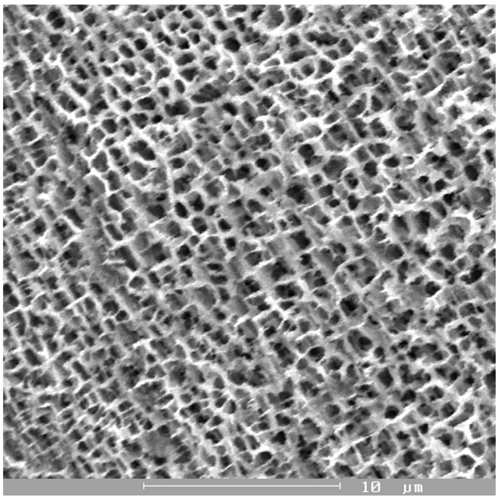
Fig. 1 Microstructure of As – cast Ni – base superalloy ŽS6K: (a) dendritic segregation, (b) primary MC and secondary $M_{23}C_6$ carbides, etch. Marble, SEM

The γ′ precipitates in precipitate strengthened nickel - base superalloys remain coherent up to large precipitate sizes due to the small lattice mismatch between the matrix phase γ and the γ′ precipitates. 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 [7]. Both of these phases, γ (gamma) and γ′ (gamma prime) are also creating γ/γ′ eutectic, Fig. 2 (b).

Turbine blades have working 500 hours of regular conditions (after 500 hours were blades checked at overhauling) and after further 100 hours of work short – time high temperature overcrossing has occurred.



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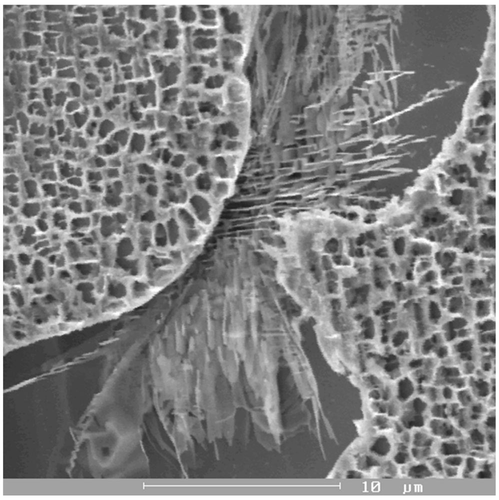


(b)

Fig. 2 Ni – base superalloy ŽS6K: (a) morphology of γ′ precipitate, as - cast, (b) eutectic cell of γ/γ′, etch. Marble, SEM

A regular temperature at start up the of 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. Observing of alitize layer degradation was focused on four significant parts of blade as leading edge, flap pantile, flow edge, and back; see Fig. 3 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.

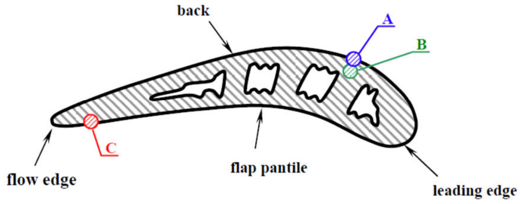
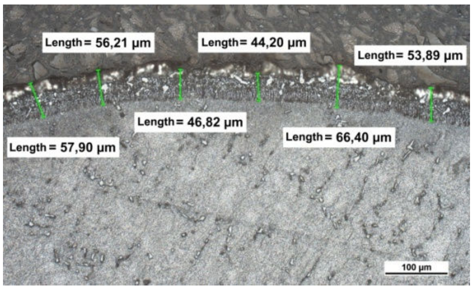


Fig. 3 The cross section of blade and description zones of interests A – checking of the uniformity and homogeneity of alitize layer from 1/4 of blade back B – evaluation of morphology and amount of γ' - phase and grain boundary right under alitize layer C – checking of alitize layer degradation or presence of Cr rich carbides needles.



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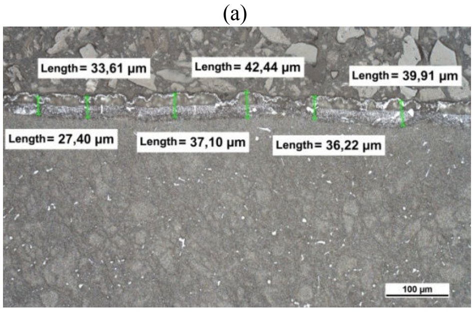

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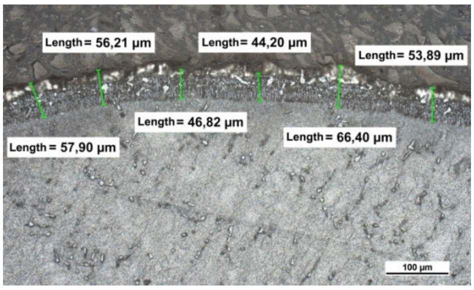


(c)

Fig. 4 Al – Si protective layer thickness measurement, non-etched specimen: (a) leading edge with average coating thickness of 50 µm, (b) turbine blade back with average coating thickness of 36 µm, (c) flow edge with average coating thickness of 45 µm

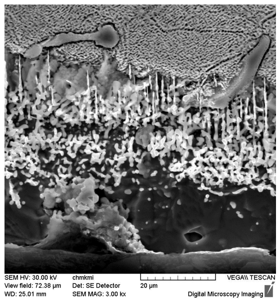
III. Results and Discussion

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.

The Al – Si layer on the surface of blade is significantly coarsened and lost its homogeneity due to high temperature degradation and abrasive wearing. Degradation as itself shows in layer thickness increasing. 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 δ = 35 µm). The thickness of Al – Si layer is higher in all observed areas what indicates higher temperature of loading than reported 751°C. The highest degree of degradation was observed at three areas from all four observed, namely leading edge with average thickness 50 µm, Fig. 4 (a); turbine blade back with average thickness 36 µm, Fig. 4 (b); and flow edge (in some parts of flow edge was not Al – Si layer identified at all, a huge degradation due to abrasive wearing at high temperatures) with average thickness of 45 µm, Fig. 4 (c).

The next problem with Al – Si layer degradation is forming a needle like Cr rich carbides and blocky form of primary Ti rich carbides in layer – base material zone. Fig. 5 reports such carbide formation confirmed also by chemical SEM mapping.



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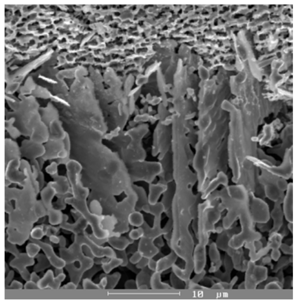

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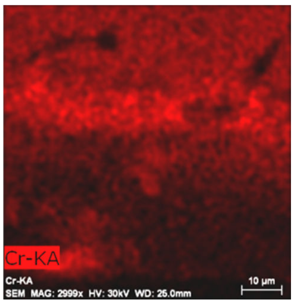

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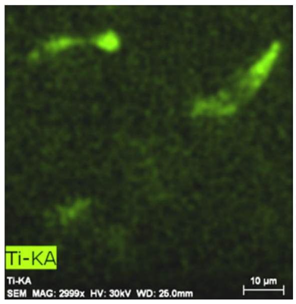
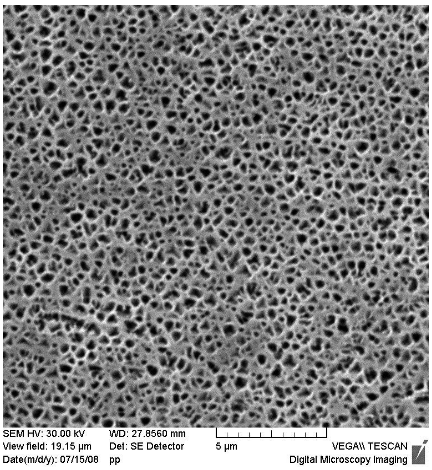


(d)

Fig. 5 Chemical analyses of layer – base material zone with Cr and Ti rich carbide formation: (a) overall view, (b) detail of Cr rich carbides, (c)-(d) SEM mapping, etch. MARBLE, SEM microprobe Bruker

The base material, alloy ŽS6K, also 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.



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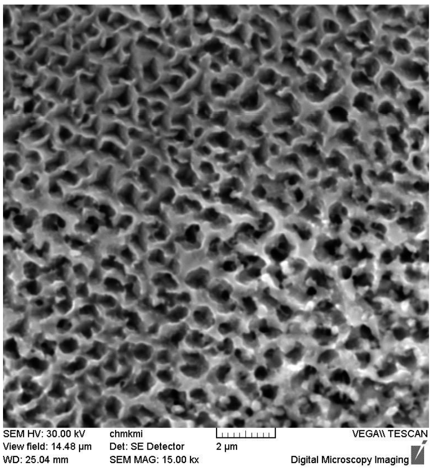

(b)

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

This shape and size changes cause decreasing creep rupture life of alloy, Fig. 7 [8]. 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.

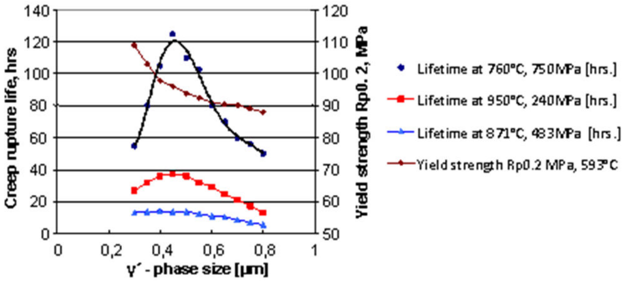

Fig. 7 Influence of γ′ - phase size on mechanical properties [8]

IV. 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.

Acknowledgment

This work has been supported by Scientific Grant Agency of Ministry of Education of Slovak Republic and Slovak Academy of Sciences, No. 1/0533/15 and project EU ITMS 26110230117.

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