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FAILURE ANALYSIS OF JET ENGINE TURBINE BLADE

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Abstract

Jet engine turbine blade cast by investment precision casting of Ni-base superalloy, which failed during exploatation, was the subject of investigation. Failure analysis was executed applying optical microscopy (OM), transmission electron microscopy (TEM) using replica technique, scaning electron microscopy (SEM) and stress rupture life tests. On the ground of obtained results it was concluded that the failure occurred as a result of structural changes caused by turbine blade overheating above the exploitation temperature.
Keywords: Jet engine turbine blade, Ni-base superalloy, Failure, Microstructural characterization; TEM; Stress rupture life

Introduction

Exploitation in harsh environments especially at high temperatures and in highly corrosive atmosphere requires materials of superior properties. Ni-base superalloys are commonly applied at high temperatures and therefore find extensive applications in hot sections of gas turbine engines, rocket propulsion systems, and nuclear reactors. Among nickel-base superalloys, IN 100 is used mainly for jet engine parts production such as turbine blades and wheels operating in the intermediate temperature regime [1-4].

The microstructure of IN 100, as most Ni-base superalloys, is rather complex. The high temperature strength of these superalloys is due to finely dispersed particles of the γ' phase. This phase is ordered (L12) $Ni_3(Al, Ti)$ intermetallic compound with f.c.c. structure and is precipitated coherently in the Ni-rich f.c.c. matrix (the γ phase). The IN 100 mechanical properties are largely dependent on the microstructure, especially the γ' phase. Volume fraction, particle size and composition of the γ' particles have a significant influence on the strength and creep resistance of superalloys [5]. Carbides

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and borides appear as minor phases. Carbides of type MC - (Ti, Mo)C, $M_{23}C_6$ - ($Cr_{23}C_6$) and M_6C - (Ni, Co, Mo, Ti)$_6$C are detected in Ni-base superalloys. Blocky Ti-rich MC carbides are formed during solidification and are mostly distributed into the γ matrix and interdendritic regions. Due to their hardness and size they are signified as crack initiation sites. M_6C and $M_{23}C_6$ carbides usually precipitate at the grain boundaries. Being much smaller than MC carbides, these carbides are mostly distributed at grain boundaries. The role of these carbides is to impede grain boundary sliding and to improve the high temperature strength of superalloys [6]

Jet engine turbine blades are made of high temperature Ni-base superalloys obtained by the method of vacuum investment casting. The main characteristic of these alloys is their structural stability under high temperature and complex loads. If the blades are working under prescribed temperature conditions, structural changes will occur only after 1000 to 5000 h of exploitation. However, higher temperature at the inlet of combustion chamber may lead to overheating of turbine blades, which results in microstructural changes causing deterioration in engine performance and shortening of blades lifetime. In Ni-base turbine blades, high temperature exposure may produce considerable microstructural changes that might be the cause of serious deterioration of mechanical properties.

The aim of this paper was to investigate the circumstances that cause the failure of jet engine rotor blade. Fortunately, the consequences were not catastrophic since the accident occurred just seconds before take off while the plane was on the ground.

Experimental

The object of this investigation was as-cast broken turbine blade, which was a part of the second turbine stage of the „Viper“ jet engine („Rolls Royce“). The accident happened after allegedly only 187 hours of flight exploitation. The basic approach to testing was to determine whether the fracture was caused by mechanical damage of blade (impact of the outer object) or due to some microstructural changes, especially considering such a short period of exploitation. The blades were produced by investment precision casting in a vacuum from the high temperature Ni-base superalloy known under the common trade name Incoloy IN 100.

According to the producer's data („Ross and Catherall“ from England) the nominal chemical composition is given in Table 1.

Table 1 Chemical composition (in wt.%) of superalloy IN 100

Element	**Al**	**Ti**	**Cr**	**Co**	**Mo**	**V**	**C**	**Zr**	**B**	**Ni**
Amount	5.5	4.5	10	15	3	1.0	0.15	0.05	0.015	rest

Microstructural investigations were carried out by optical (OM), transmission replica (TEM) and scanning electron microscopy (SEM). The cross-sections of aero profile were cut from different parts of turbine blade (Fig. 1). Polished samples for OM were etched using Marble's reagent, *i.e.* a mixture of 10 g of $CuSO_4$, 50 mL HCl and 50 mL distilled H_2O. The standard replication technique utilizing the vacuum evaporation of carbon followed by gold shadowing, was applied to obtain replicas for TEM.

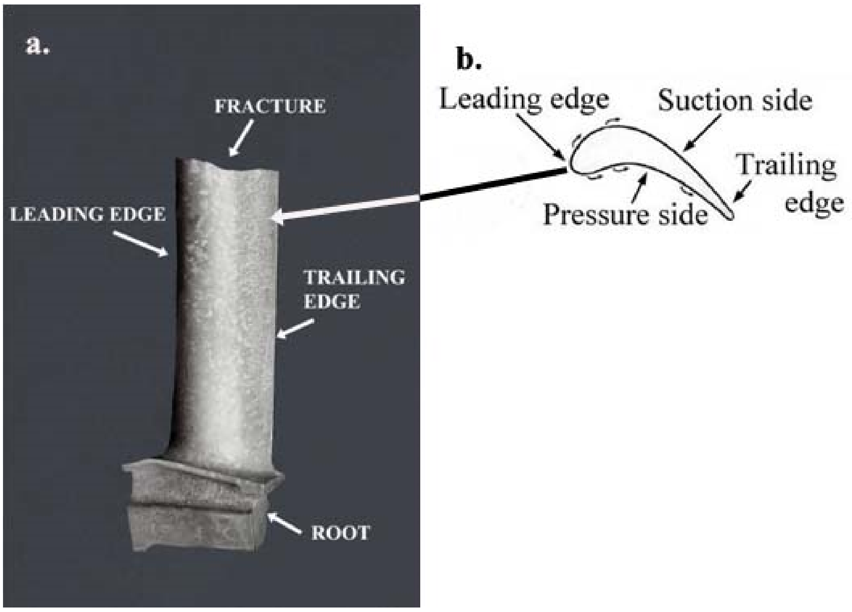

Fig. 1 a. Fractured turbine blade IN 100 of the second stage of "Viper" jet engine; b. cross-section of aero-profile

For quantitative measurement of the weight fraction of the γ' phase, anodic dissolution of sample in the 1% mixture of citric acid and ammonium sulfate in water was applied. A platinum grid served as the cathode, and the potential was maintained at 1.4 V. Duration of electrolysis was 4 h at ambient temperature. Carbides always accompanied the extracted γ' phase, and the correction for this contamination has not been made.

Stress rupture life, *i.e.* time-dependent strength that occurs under load at elevated temperature (similar to creep) was measured on facilities in "Investment Casting Foundry"-Ada, Serbia. Samples for this investigation were cut just above the blade root and were machined according to the dimensions of ASTM E 8M-94a standard rod specimens, with gage length 25.4 mm and diameter 6.35 mm. According to the internal standard prescribed for IN 100, stress rupture tests were performed at $950 \pm 3^{o}C$ under the load of 250 MPa.

For the sake of comparison some metallographic and mechanical testing were also carried out on unused blades.

Results

Microstructural characrerization

Microstructure of as-cast IN 100 turbine blade before installation in the turbine rotor, normally consists of the the matrix, which is the γ phase solid solution, small particles of intermetallic γ' phase, $Ni_3(Al, Ti)$, mostly of the cuboidal shape, lamellar γ/γ'eutectic sometimes in kidney-like appearance, and primary carbides of MC type (mostly TiC) [5, 6]. This microstructure is illustrated in Fig. 1a, b.

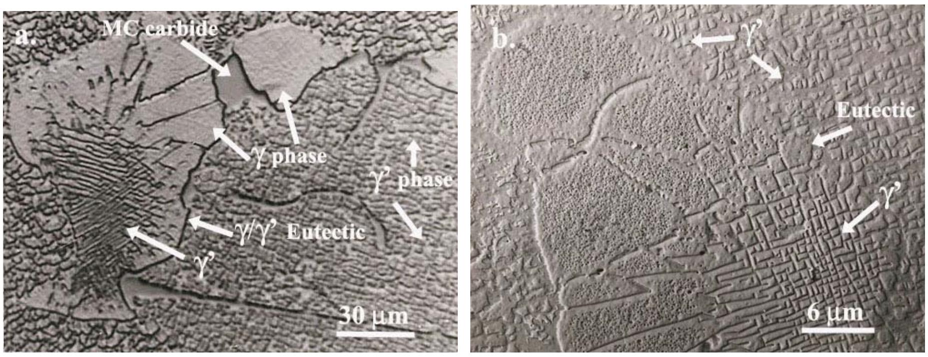

Fig. 1 Microstructure of as-cast unused turbine blade. a. OM; b. TEM replica

Fig. 2a, b shows microstructure of the cross-section of aero-profile taken a few millimeters below the fracture surface. The presence of all phases described in Fig. 1 may be seen, although it is quite visible that significant changes in the microstructure occurred. The volume fraction of the γ' particles was decreased, whereas they became coarser with rounded edges. Eutectics are completely depleted with the lamellar γ' phase, whereas in the surroundings of eutectics are depleted zones without any particles.

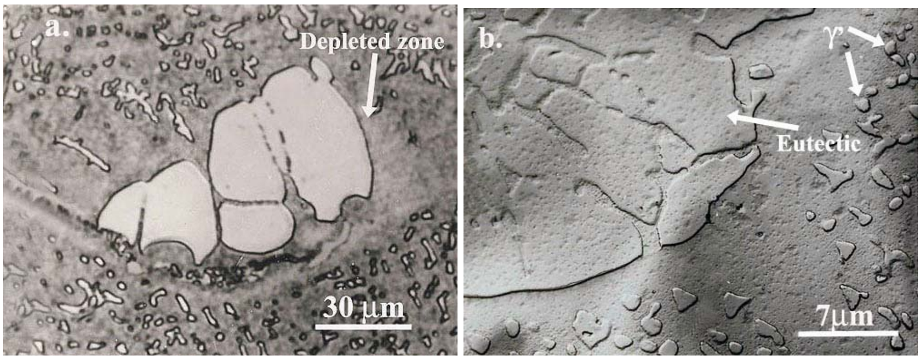

Fig. 2 Microstructure of broken turbine blade. Cross-section of a sample taken a few millimeters below the fracture surface. a. OM; b. TEM replica

Microstructure close to the leading edge of the turbine blade (same sample as in Fig. 2) is shown in Fig. 3a. γ' particles cannot be observed in the γ matrix. Also, it is highly possible that the high temperature caused incipient melting of eutectic (black fields in Fig. 3a). On the other side, the microstructure corresponding to the root of turbine blade (Fig. 3b) shows that γ' particles, although rather small, are still present in the γ matrix. Fig. 3c which also corresponds to the microstructure of the blade root, shows primary MC carbides enveloped with the chain of γ' particles. These microstructures suggest that the temperature of blade close to the fracture area was much higher than near the blade root.

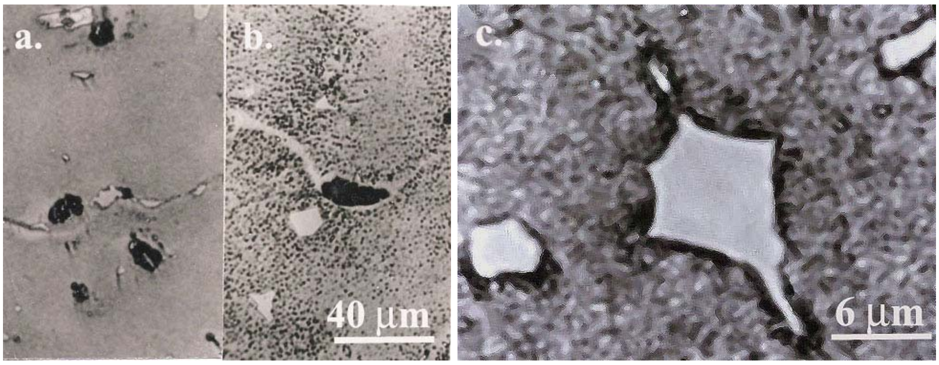

Fig. 3 Microstructure of broken turbine blade. a. OM of leading edge (same sample as in Fig. 2); b. OM of the blade root; c. TEM replica of microstructures corresponding to the blade root

Weight fraction of the γ' phase, $m_{\gamma'}$

In a number of samples taken from unused turbine blade $m_{\gamma'}$ was approximately 65 wt.%, the result corresponding to the value prescribed by "Rolls Royce" for IN 100. However, the $m_{\gamma'}$ of broken blade was much smaller, ranging between 35 and 40 wt.%, depending on the location of turbine blade. Extracted powders of unused sample of turbine blade are shown in Fig. 4a. The cuboidal shape of γ' particles are present in extracted material together with plate-like and blocky MC carbides. The size of γ' particles is approximately 0.3 μm. Extracted powders close to the fracture surface are smaller with less cuboidal form and with rounded edges (Fig. 4b).

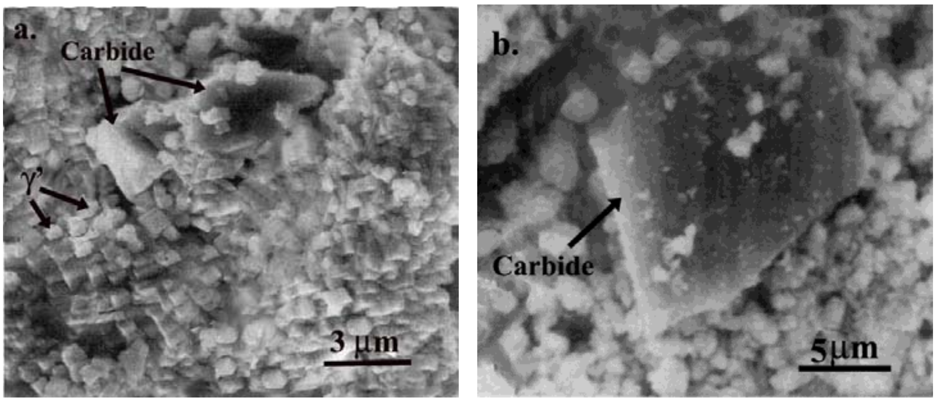

Fig. 4 SEM. Extracted powders from turbine blade. a. From unused turbine; b. close to fracture surface

Stress rupture life

According to experimental conditions the minimum value of stress rupture life should be around 53 h. The result of unused turbine blade was even higher and reached the value of 70 h. However, stress rupture life of broken blade was only 16 h. Results on weight fraction of the γ' phase, $m_{\gamma'}$, and stress rupture life are summarized in Table 2.

Table 2 Weight fraction of the γ' phase, $m_{\gamma'}$, and stress rupture life for different conditions of turbine blade

Turbine blade condition	Properties	
	Weight fraction of the γ' phase, wt.%	**Stress rupture life, h**
Unused	65	70
Broken	35-40	16

Discussion

Microstructural changes in superalloys are the result of two parameters: time and temperature. With increase in temperature these changes occur in a relatively short time, whereas at lower temperatures changes may happen at slower paste as a function of both time and temperature. According to literature data [7] changes in microstructure of IN 100 superalloy have not been detected at 800 and 900°C even after exposition as long as 1000 h, *i.e.* under these conditions γ' particles retain their morphology and cuboidal form. The coarsening of these particles occurs at temperatures higher than 900°C. Decrease of $m_{\gamma'}$, observed in all areas of turbine blade, the lowest being near the fracture surface, also documented by microstructures in Fig. 3a, b, was the result of the high exploitation temperature causing the dissolution of the γ' phase into the γ phase solid solution. The visible difference in structure between unused and broken turbine blades (Figs. 1-3) and the results of $m_{\gamma'}$ are in full accordance, and were supported by the fact that $m_{\gamma'}$ was decreased from 65 to 35-40 wt.%. According to Fig. 3a it might be supposed that $m_{\gamma'}$ very close to the fracture surface, might be even much lower. Degeneration of lamellar γ' eutectic as well as depleted zones around eutectic are the consequence of solution annealing resulting when the γ' phase particles were dissolved in the solid solution. In the areas close to the fracture surface where the temperature was highest, incipient melting of eutectic occurred (black areas in Fig. 3a, b). Under the influence of high temperature, chain-like formation of particles around massive MC carbide (Fig. 3c) is ascribed to the reaction between elements present in carbide (mostly Ti) and elements from the matrix (the γ phase) which results in formation of $Ni_3(Al,Ti)$ intermetallic compound, actually the γ' phase [8].

Taking into account very low exploitation time, 187 h, as well as the significant changes in microstructure, it is reasonable to conclude that the turbine blade was subjected to temperatures reaching approximately 1200°C, actually the γ' phase solvus temperature. Above this temperature the entire transfer of this phase into the solid solution occurs [9]. The locations of fracture, caused by incipient melting, correspond to areas where complete dissolution of the γ' phase particles happened.

Low values of stress rupture life are obviously the effect of the decreased weight fraction of the γ' phase, *i.e.* increased distance between γ' particles, and depleted zones around γ/γ' eutectic. In this way the movement of dislocations and micro-cracks should be significantly relieved.

Conclusions

During short time of exploitation significant changes in microstructure occurred resulting to low stress rupture life and the fracture of turbine blade. The results indicate that the cause of the failure was overheating of the turbine jet engine to the temperature around 1200^{o}C.

Acknowledgement

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