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Failure analysis of a gas turbine blade made of Inconel 738LC alloy

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Abstract

The failure analysis of the 70 MW gas turbine first stage blade made of nickel-base alloy Inconel 738LC is presented. The blades experience internal cooling hole cracks in different airfoil sections assisted by a coating and base alloy degradation due to operation at high temperature. A detailed analysis of all elements which had an influence on the failure initiation was carried out, namely: loss of aluminium from coating due to oxidation and coating phases changing; decreasing of alloy ductility and toughness due to carbides precipitation in grain boundaries; degradation of the alloy gamma prime (γ') phase (aging and coarsening); blade airfoil stress level; evidence of intergranular creep crack propagation. It was found that the coating/substrate crack initiation and propagation was driven by a mixed fatigue/creep mechanism. The coating degradation facilitates the crack initiation due to thermal fatigue.

The substrate intergranular crack initiation and propagation were due to a creep mechanism which was facilitated by grain boundary brittleness caused by formation of a continuous film of carbides on grain boundaries, the degradation of γ' due to elongation (rafting) and coalescence, and high thermomechanical stress level.

Keywords: Failure analysis; Creep–fatigue interaction; Gas turbine failures; Turbine blade failure; Metallurgical examination

1. Introduction

Gas turbine blades are made of nickel-base and cobalt-base superalloys principally. During the operation of power generation gas turbines, the blades and other elements of hot gas path suffer service induced degradation which may be natural or accelerated due to different causes. The degradation or damage may

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have a metallurgical or mechanical origin and results in reduction of equipment reliability and availability. It also increases risk of failure occurring. Also, due to blade material metallurgical deterioration, the material creep, fatigue, impact and corrosion properties decrease. There are different factors which influence blade lifetime as design and operation conditions but the latter are more critical.

Generally speaking, most blades have severe operation conditions characterized by the following factors:

- Operation environment (high temperature, fuel and air contamination, solid particles, etc.).
- High mechanical stresses (due to centrifugal force, vibratory and flexural stresses, etc.).
- High thermal stresses (due to thermal gradients).

Typically there are acting two or more factors simultaneously causing reduction of blade lifetime under the following damage mechanisms [1]:

- Creep
- Thermal fatigue (low cycle fatigue)
- Thermomechanical fatigue (high cycle fatigue)
- Corrosion
- Erosion
- Oxidation
- Foreign object damage

The type of damage which occurs in gas turbine blades and nozzles after a service period can be divided into:

- External and internal surfaces damage (corrosion, oxidation, crack formation, erosion, foreign object damage and fretting).
- Internal damage of microstructure as γ' [$Ni_3(Al,Ti)$] phase aging (rafting), grain growth, grain boundary creep voiding, carbides precipitation and brittle phases formation.

Surface damage produces blades/nozzles dimensional changes which result in operational stress increase and turbine efficiency deterioration.

In service, blade material deterioration is related to the high gas temperature, high steady state load levels (centrifugal load) and high thermal transients loads (trips, start-ups and slowing downs). However, the degree of deterioration in individual blades differs due to several factors such as:

- Total service time and operation history (number of start-ups, shut-downs and trips).
- Engine operational conditions (temperature, rotational speed, mode of operation (base load, cyclic duty)).
- Manufacturing differences (grain size, porosity, alloy composition, heat treatment).

The Inconel 738LC alloy commonly used for gas turbine blades is strengthened by precipitation of γ' phase. The microstructural changes due to blade operation at high temperature include irregular growing of γ' particles (rafting) and formation of carbides in grain boundaries and matrix [2]. This leads to alloy creep properties reduction [3].

In order to have an instrument for the deterioration evaluation of gas turbine blade alloy, it is necessary to correlate the influence of service induced microstructural degradation to the change in mechanical properties. This can be used for monitoring and evaluation of extent and degree of material damage and lifetime consumed and to obtain recommendations for blade rejuvenation treatments, operation and reposition [4].

Application of effective methods of material deterioration evaluation can be used for practical lifetime prediction, just in-time blade rehabilitation (rejuvenation), safe and cost-effective lifetime extension and to avoid blade catastrophic failure.

2. Background

The blade under evaluation was the first stage blade of a 70 MW combustion turbine with gas inlet temperature of 1086 °C. The evaluation was carried out after 24,000 h of blade operation period in mode of base load. The blade is made of nickel-base Inconel 738LC superalloy by means of conventional investment casting (equiaxed grains) and coated by Pt–Al (RT22) coating by diffusion process. The alloy composition (% wt) is shown in Table 1. The combustion turbine operates on natural gas and the power plant is located inland.

3. Microstructural characterization of a gas turbine blade

The microstructure evaluation of different zones of the blade shown in Figs 1–4 was carried out. The microstructure of the blade hot section (airfoil) was compared to the cold reference zone (blade root) to evaluate the degree of alloy deterioration. The comparative evaluation include the morphology change of the γ' particles, carbide precipitation and characterization of grain type and size. Also, the blade coating degradation was evaluated.

3.1. Microstructural evaluation of the blade root (reference zone)

A metallographic analysis of the microstructure in the blade root (zones indicated in Figs. 1 and 4), was carried out. The blade root is considered as a "cold zone" because it is not exposed to the hot combustion gases and microstructural changes are not considerable. This is why this zone may be used as a reference zone. The blade root surface was etched by a solution of 10% perchloric acid ($HCLO_4$) diluted in ethanol and next electro polished to reveal the microstructure. The microstructure of the transversal section of the root (see Fig. 4) is shown in Fig. 5. The microstructure consists of equiaxed grains of γ phase (alloy matrix) and fine particles of γ' precipitated within the matrix. Also, dispersed particles of carbides in the grain boundaries and matrix were found. The average grain size was determined as 365 μm approximately. This type of microstructure is common for γ' precipitation nickel-base alloys. Fig. 6 shows the morphology and size of the γ' phase. A duplex γ' structure is observed. The fine particles appears to be γ' formed during aging due to partial solution. Although the blade root is considered as a cold zone which normally does not suffer considerable microstructural changes, it is possible to observe some slight elongation of the original cubical shape of the γ'. The average size of γ' (diametral maximum dimension) was 0.9 m and the average elongation/rafting (length/width ratio) was 1.2. Also, the volume fraction of γ' in measured area was determined. The γ' volume fraction was 56%.

The grain boundary microstructure changes during service due to γ' instabilities. These changes appear in a form of transformation of carbides of MC type to $M_{23}C_6 + \gamma'$ and by formation of a continuous film of

Table 1
Chemical composition of Inconel 738LC superalloy (wt%)

Alloy	C	Cr	Ni	Co	Mo	W	Cb	Ti	Al	B	Zr	Ta
IN 738LC	0.11	16	Bal	8.5	1.75	2.6	0.9	3.4	3.4	0.01	0.06	1.75

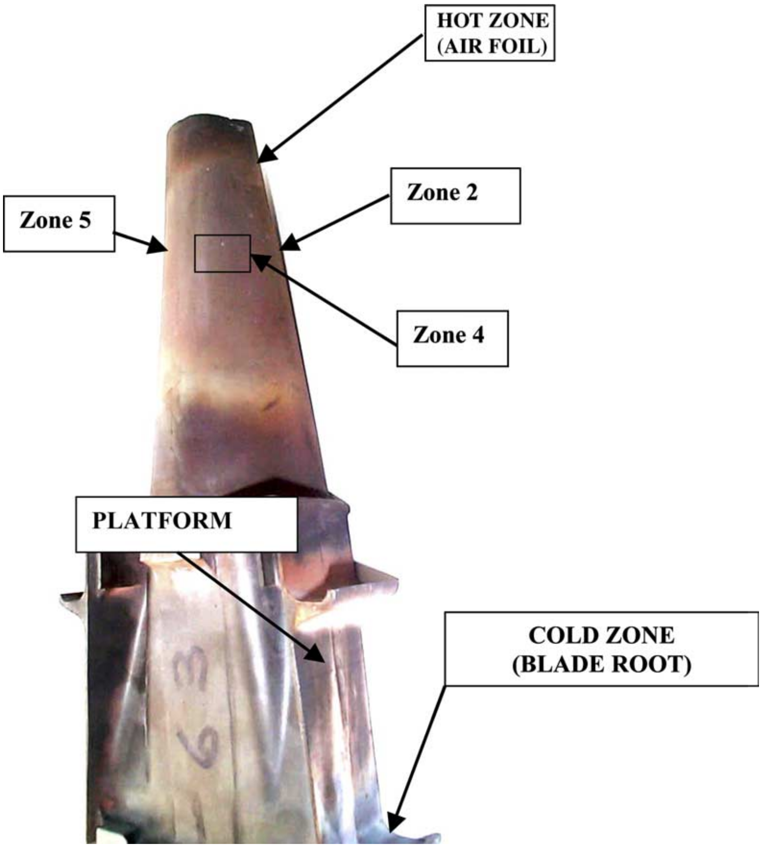

Fig. 1. Analysis regions on the blade convex side.

carbides along grain boundaries resulting in alloy brittleness [4]. In Fig. 7 is shown the original state (initial state) of MC type carbides precipitated in grain boundaries and matrix, and carbides of $M_{23}C_6 + \gamma'$ type precipitated in grain boundaries. The volume fraction in grain boundaries of MC type carbides was 45% and volume fraction of $M_{23}C_6$ was 55%.

3.2. Microstructural evaluation of blade hot section (airfoil)

Samples for microstructural evaluation were taken from the blade airfoil hottest zones as indicated in Figs. 1–3. The hot gas flow trace on the blade airfoil (oxidation pattern) corresponds to the hottest zones of the blade (see Figs. 1 and 2). In Fig. 8 the microstructure of a sample taken from zone 5 (see Figs. 2 and 3) can be appreciated after electro polishing. The microstructure consists of grains of γ and carbide particles precipitated in the matrix and grain boundaries. The average grain size was determined to be 560 μm. In grain boundaries was found a continuous film of carbides of 1.5–3 μm thickness approximately. In Fig. 9 is shown the detail of a continuous film of carbides in grain boundaries. As a result of transformation of carbides of MC type to $M_{23}C_6 + \gamma'$ type during operation at high temperature, the volume fraction of MC type carbides in grain boundaries was 5% and $M_{23}C_6$ type was 95%. These values are very different to those in the blade root. In Figs. 9–11 showed the presence of γ' particles with considerable degree of deterioration.

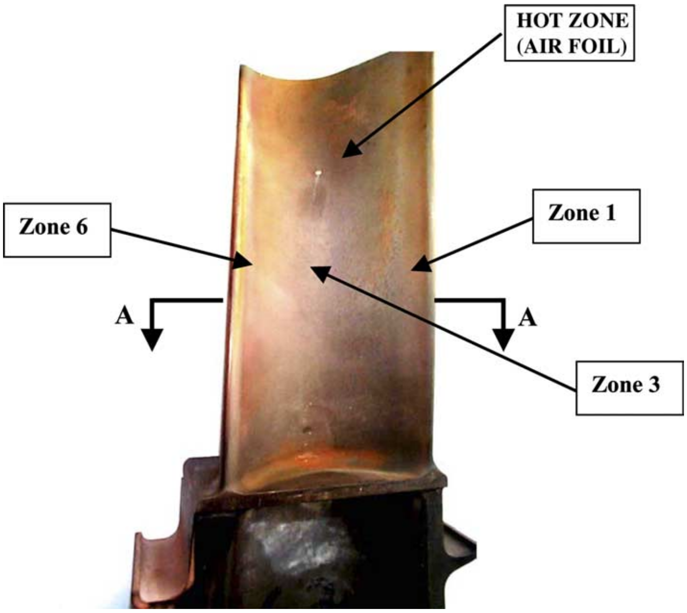

Fig. 2. Analysis regions on the blade concave side.

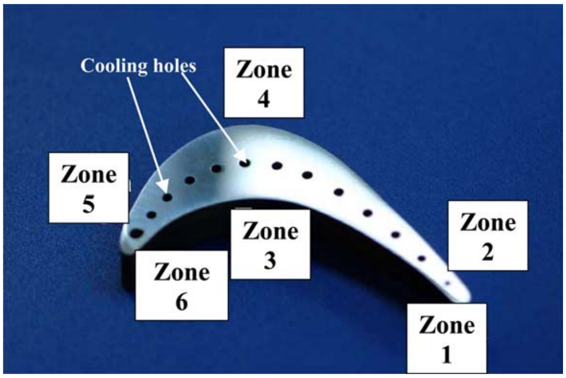

Fig. 3. Analysis regions on the blade A–A transversal section from Fig. 2.

The γ' original cubical shape changes into elongated platelets (rafts) oriented perpendicularly to the principal stress direction. After a certain time the γ' elongation reaches an equilibrium state where no further elongation takes place. Also, the coalescence (growing) of γ' particles is observed. The time to reach the equilibrium state is stress and temperature dependent so that the time to reach the maximum γ' raft length decreases with increased temperature and/or stress. The final equilibrium γ' raft length also differs, depending on applied stress and temperature level. The γ' elongation in nickel-base alloys is related to gradual decrease of the total blade life [5–7]. For the Inconel 738LC blade, the γ' elongation was quantified by measuring the length/width ratio (R ratio) for approximately 8000 γ' particles. Each particle was measured in an elliptic projection of D_{max} (length) and D_{min} (width). The average was calculated using a computerized image analysis system. The average size of γ' particles (D_{max}) in zone 5 (see Figs. 2 and 3) was 1.3 μm

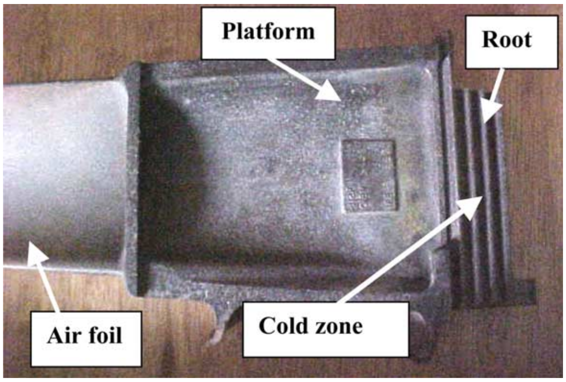

Fig. 4. General view of the blade root zone.

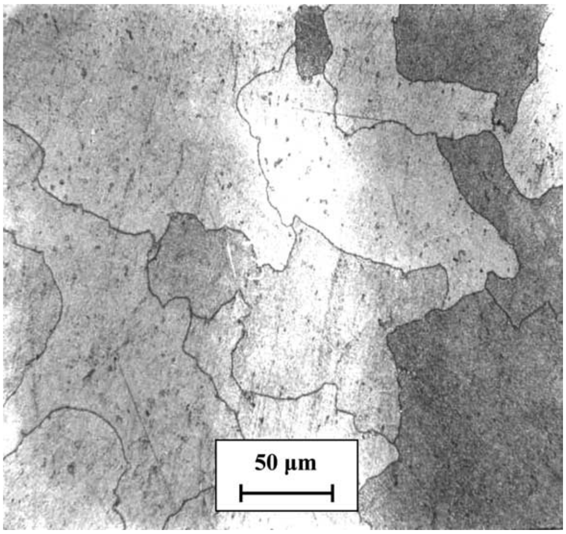

Fig. 5. Grain size in the blade root.

(Fig. 10) and the average elongation/rafting (length/width ratio) 1.8. The γ' volume fraction in the measured area was 42%.

The blade airfoil has a diffusion Platinum Aluminium coating which consists of three zones (Fig. 12):

- External zone of $PtAl_2$.
- Intermediate zone of N_iAl or β.
- Internal or interdiffusion zone.

Fig. 12 shows coating deterioration in internal cooling holes of the airfoil. Loss of the coating constituents due to oxidation can be seen. The thickness of coating in different airfoil transversal sections (zones) is shown in Table 2. Using the same method all blade airfoil zones (1–6 indicated in Figs. 1–3) were metallographically evaluated. A similar tendency of microstructural transformation was found but with some variation of degree of alloy deterioration from zone to zone related to blade temperature variation in

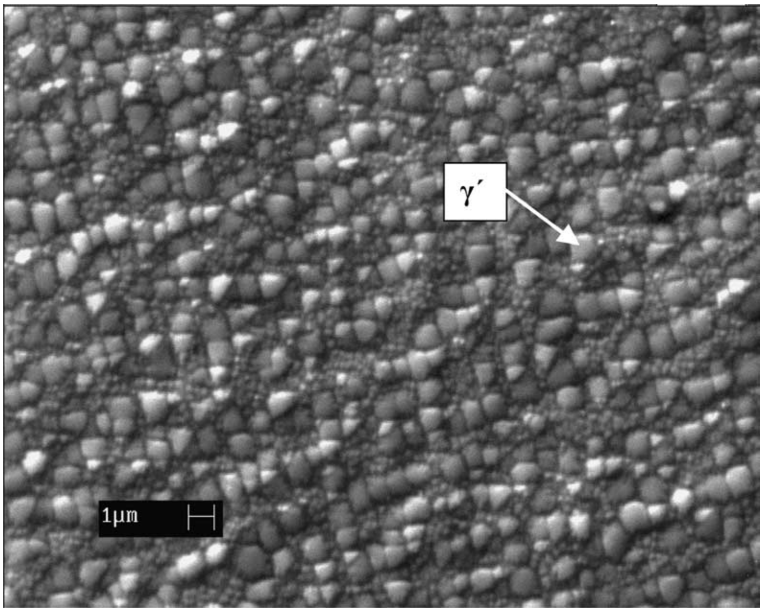

Fig. 6. Gamma prime (γ′) phase morphology in the blade root.

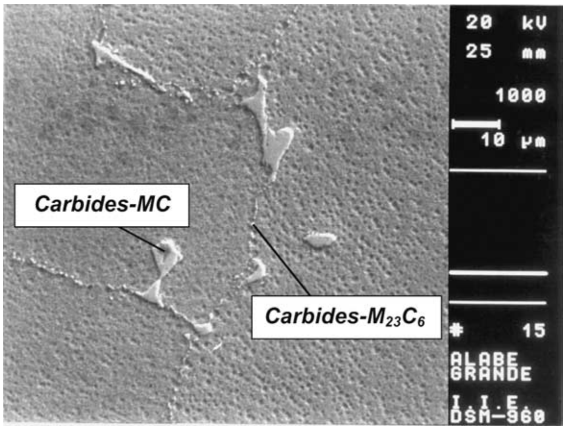

Fig. 7. Carbide particles in the matrix and grain boundaries of the blade root zone.

the airfoil. The results of the blade airfoil evaluation are showed in Table 2 compared to the reference zone (blade root).

4. Crack evaluation

The cracks in the internal cooling holes in the hottest sections of the blade (airfoil central sections) were detected (see Fig. 12). Cracks initiate in the cooling hole coating and propagate into the substrate following grain boundary trajectories. The crack size reaches 0.4 mm. At the crack tip, small creep voids were iden-

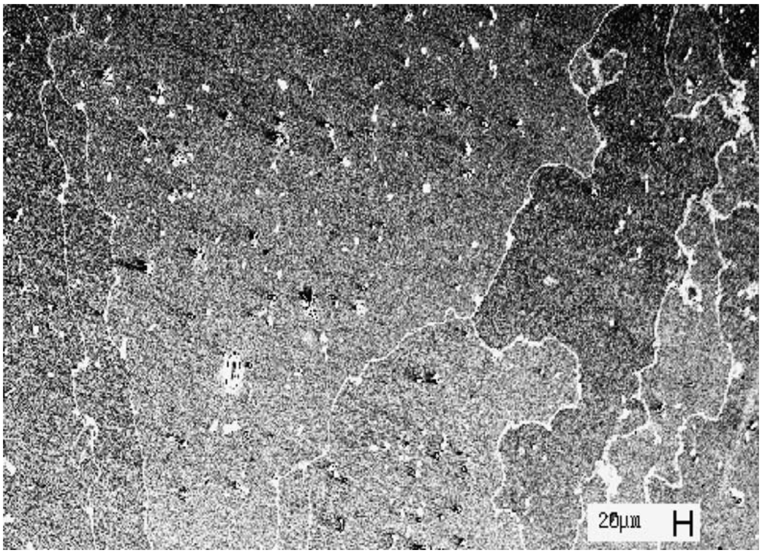

Fig. 8. Grain size and carbide particles precipitated in the matrix and grain boundaries of the blade airfoil zone 5 – from Figs. 1 and 3.

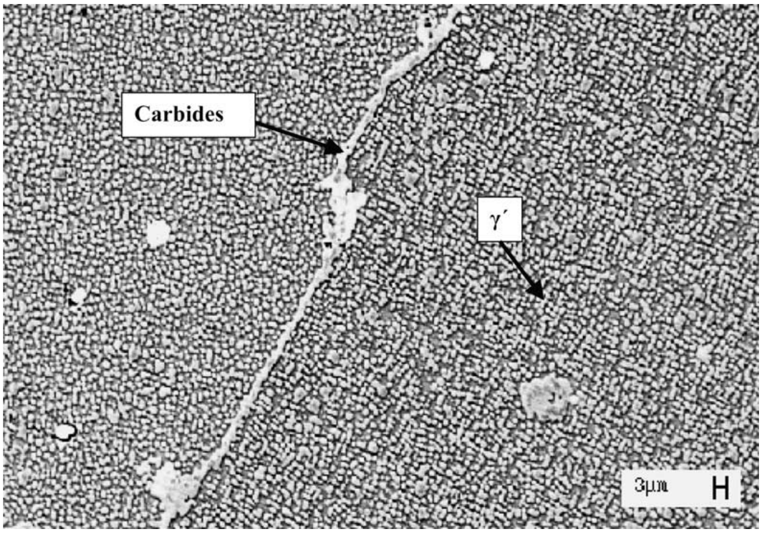

Fig. 9. Continuous band of grain boundary carbides and γ' morphology in the zone 5.

tified on the intergranular facets at the extreme tip of the crack (Fig. 13). On the basis of these evidences it was evaluated that the crack initiation/propagation was derived by a mixed fatigue/creep mechanism. The coating crack initiation was probably due to a thermal fatigue mechanism as a result of high thermal transient loads (trips, start-ups and slow-downs), and crack grain boundary initiation and propagation in the substrate by a creep mechanism (high steady state load).

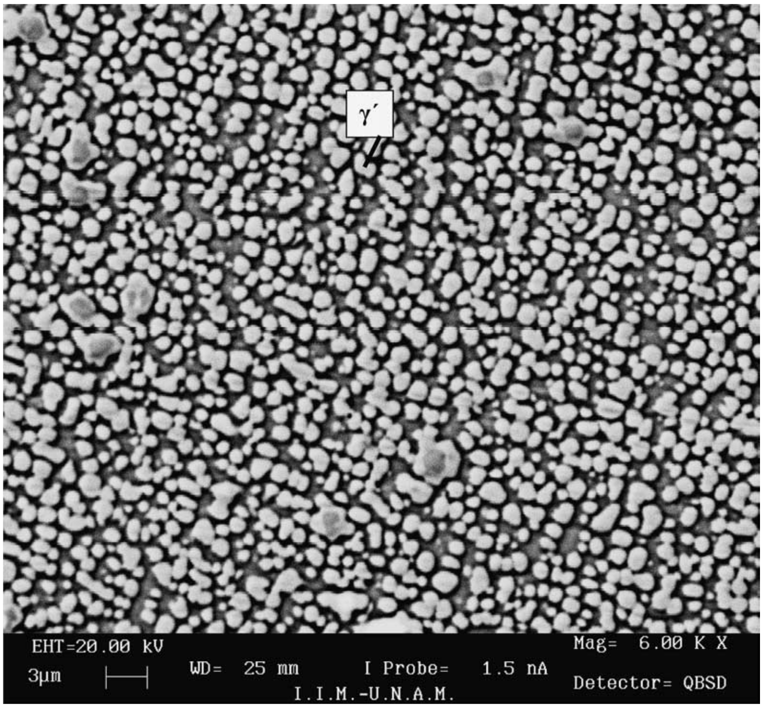

Fig. 10. Gamma prime (γ′) morphology in the zone 5.

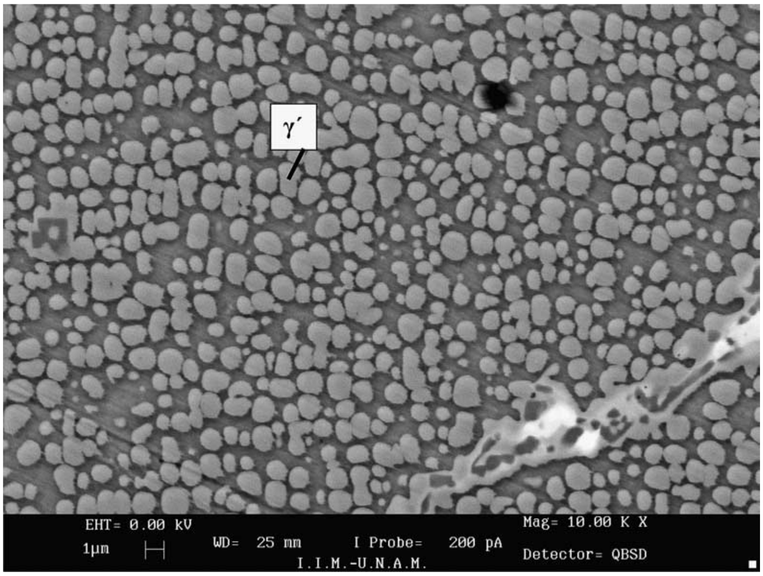

Fig. 11. Gamma prime (γ′) morphology in the zone 3.

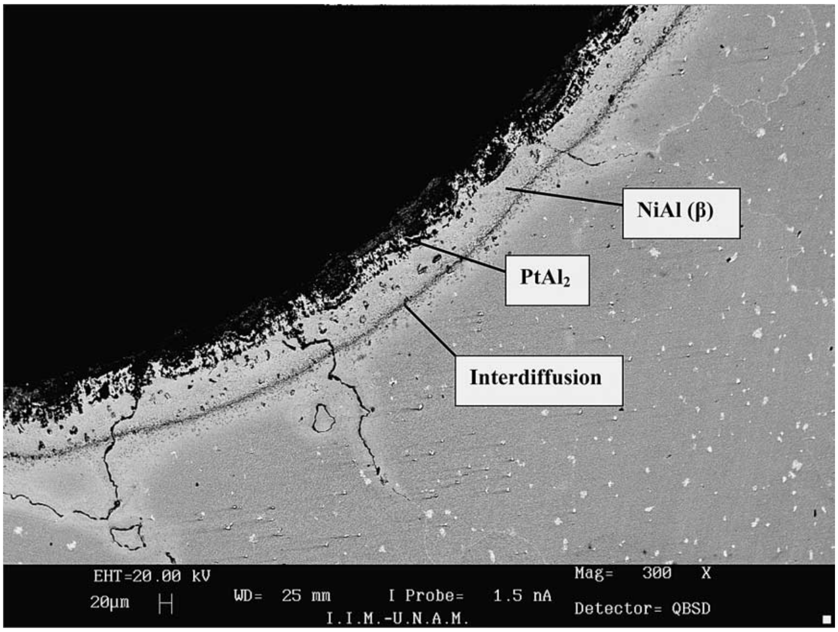

Fig. 12. Termomechanical fatigue cracks in the cooling holes.

Table 2
Quantitative microstructure characterization of a gas turbine blade

Zone	Coating thickness ext./inter. (μm)	Size of γ' phase μm	Elongation (rafting) of γ' phase	Volume fraction of γ' (%)	Carbides MC type (%)	Carbides $M_{23}C_6$ type (%)
Root	–	0.9	1.2	56	45	55
1	70/129	1.7	1.6	44	15	85
2	80/129	1.7	1.6	44	15	85
3	100/–	1.2	1.6	48	12	88
4	110/–	1.2	1.3	50	10	90
5	96/100	1.3	1.8	42	5	95
6	110/115	1.3	1.7	41	5	95

5. Stress evaluation

Thermomechanical stress analysis was carried out on the first stage blades by means of the finite element method (FEM) using the commercial code ANSYS. The maximum tension stresses in the blade airfoil were approximately 341 MPa (see Fig. 14) and were located in the cutting plane at 60% vane height on the cooling hole surface. The stresses were principally thermal stresses developed due to high temperature gradients across the airfoil wall (between vane external surface and surface of internal cooling holes). A very good agreement of the predicted stress distribution and crack location can be seen.

6. Discussion

The presence of a continuous film of carbides of 1.5–3 μm thickness in grain boundaries is a result of transformation of carbides of MC type to carbides of $M_{23}C_6$ type due to high temperature operation of the blade.

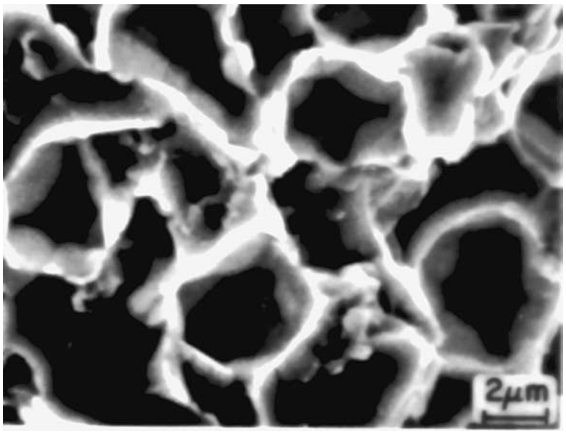

Fig. 13. The small creep voids identified at the extreme tip of the crack.

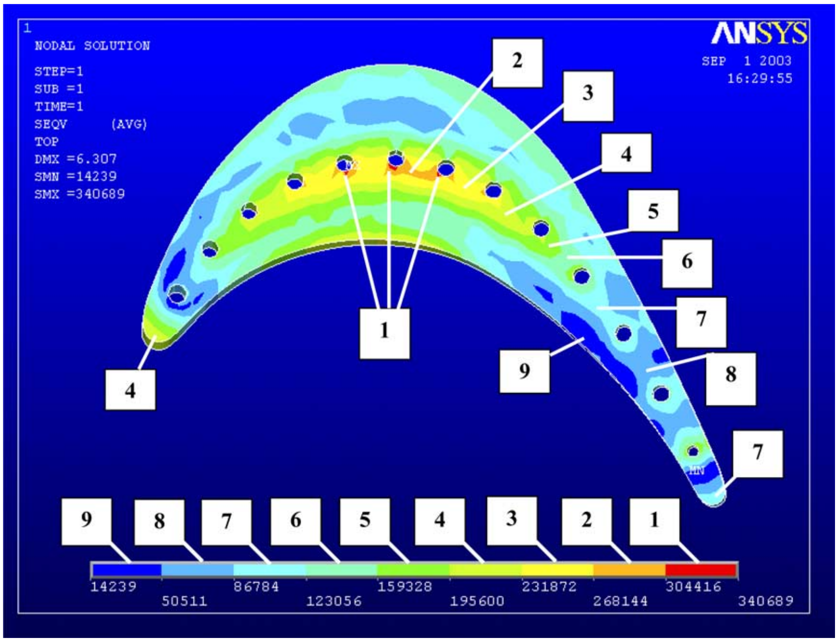

Fig. 14. Blade airfoil stress distribution in the cutting plane at 60% vane height.

This dense and continuous network of carbides reduces ductility and toughness of the alloy to 30% of initial value and facilitates crack initiation and propagation which leads to reduced lifetime [5–7]. The degradation of γ' due to elongation (rafting) and coalescence (growing) originates a reduced alloy creep lifetime. The average size of γ' particles in the airfoil (hot zone) was from 33 to 89% bigger than the same grain size in the root (reference zone) as is shown in Table 2. The average elongation (rafting) of γ' particles in the blade airfoil was from 8% to 50% bigger than in the reference zone.

Crack initiation in internal cooling holes has the following sequence:

- Coating degradation and cracking due probably to a thermal fatigue mechanism and environmental attack.
- Crack initiation in substrate.

Generally speaking the degradation of coatings resistant to high temperature proceeds in two ways:

1. Loss of coating constituents due to oxidation and corrosion, which results in loss of Al from coating to create a thin protective layer of Al_2O_3 on the surface, and
2. Interdiffusion of coating-substrate, which results in diffusion of alloy elements into the coating [9].

In this particular case degradation of the Al–Pt coating occurs due to loss of aluminium from the coating to form protective Al_2O_3 on the surface. This oxide is next spalled due to thermal cycles, erosion, etc. In turn the coating provides additional aluminium to form new aluminium oxide. According to the aluminium content (also Pt) reduction, the coating is consumed and the phases change as follows.

1. $PtAl_2$ particles present in external surface layer dissolve to NiAl to form singular phase (Ni,Pt)Al or β phase.
2. This phase discomposes to biphase structure γ' $(Ni,Pt)_3Al + \beta$.
3. Because aluminium is consumed, the β phase cannot survive and only γ' exists.
4. This discomposes to γ and γ' (Ni and other elements into solid solution).
5. Finally only γ exists and coating is consumed completely [10].

Practically, coating protects base metal when it is formed by only singular phase β and loss its protective characteristics when it transforms to a biphase structure $\gamma' + \beta$. This stage is optimum for coating restitution (stripping deteriorated coating and recoating). In the case of a gas turbine fuelled by natural gas the coating deterioration occurs due to oxidation in the blade airfoil hottest zones, which typically may be leading edge or other zones depending on blade design. There were found many different airfoil zones with different degrees of coating deterioration.

Substrate grain boundary crack initiation and propagation together with the presence of small creep voids on the intergranular facets of the tip of the crack is evidence of a creep failure mechanism. This is supported by the stress distribution. The predicted airfoil maximum stress location on the internal cooling hole surface is consistent with the crack location. Substrate crack initiation and propagation is facilitated also due to grain boundary brittleness [8] caused by the formation of a grain boundary continuous film of carbides as mentioned before (see Fig. 9).

7. Conclusions

The failure analysis of the 70 MW gas turbine first stage blade made of nickel-base alloy Inconel 738LC, after 24,000 h of operation at high temperature was carried out. The microstructural investigation of the blade hot section (airfoil) revealed the presence of a continuous film of carbides of 1.5–3 μm thickness in grain boundaries of the base material as a result of transformation of carbides of MC type to carbides of $M_{23}C_6$ type due to high temperature operation of the blade. This dense and continuous network of carbides reduces the ductility and toughness of the alloy to 30% of the initial value and facilitates crack initiation and propagation which leads to reduced lifetime. The degradation of γ' is due to elongation (rafting) and coalescence (coarse growing) and originates a reduced alloy creep lifetime. The average size of γ' particles in the airfoil (hot zone) was from 33% to 89% bigger than the same grain size in the root (reference zone). The average elongation (rafting) of γ' particles in the blade airfoil was from 8% to 50% bigger than in the reference zone. Coating deterioration due to an oxidation mechanism (loss of aluminium) was registered in the blade airfoil hot zones.

Cracks were found in the cooling holes to 0.4 mm deep. It was evaluated that crack initiation/propagation was derived from a mixed fatigue/creep mechanism. Substrate crack initiation and propagation was facilitated due to grain boundary brittleness caused by formation of a grain boundary continuous film of carbides. Because the cracks penetrate the coating and substrate significantly in highly stressed areas (airfoil), it can be concluded that the blade lifetime was consumed and it is not possible to apply a repair process (recoating, rejuvenation heat treatment, etc.) to restore blade original characteristics and extend lifetime [11–13]. To make possible refurbishing and lifetime extending blades should be retired from service before cracks initiate in the substrate. To accomplish blade lifetime extension, the blade cooling system should be improved to prevent failures by reducing the airfoil thermal gradients, minimizing the airfoil thermal stresses.

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