Failure Analysis of Two Sets of Aircraft Blades

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Abstract. This paper analyses the failure of two blades, coming from different aircraft engines. The first one corresponds to the high pressure compressor manufactured in a 718 nickel base superalloy. The failure analysis carried out on this blade points towards foreign object damage (FOD). The second set belongs to the high pressure turbine of another engine. Scanning electron microscopy attributes the first fail to the premature failure by a thermo-mechanical fatigue mechanism of one blade with an inadequate microstructure. The remaining blades of this set, which possess a correct microstructure, failed due to the impacts of the debris generated by the fracture of the first one.

Introduction

The use of nickel base superalloys on various gas turbine components which operate under high temperature and high pressure conditions is widely extended. These alloys can work satisfactorily under load at temperatures near their melting point (about 85% of this figure in Kelvin degrees), clearly higher than those offered by alternative materials [1]. Nevertheless, a constant rise of both temperature and pressure is observed to improve the turbine performance, leading to harder working conditions and increasing the probability of failures.

Modern gas turbine engines for aircraft applications are generally considered to exhibit a high level of reliability, and in service failures are rare. The components most commonly rejected are the blades from both the compressor and the turbine, and the turbine vanes. The two principal causes are damage caused by ingested materials and that due to the high temperature of operation [2], that as previously indicated is becoming continuously higher. These two principal mechanisms of failure are briefly discussed. Gas turbine engines can be subject to ingestion of small, and no so small, hard particles, which induce the so called foreign object damage (usually designed as FOD). This damage takes the form of sharp V notches in the leading edges of the blades [3]. Dimensions of these notches vary from few micrometers to tens of millimeters, depending on the size, the nature and the severity of impact which induces them. Sources of FOD can be as diverse as sand particles and birds. Any gas turbine engine ingests large amounts of air when in operation, either sucked in by the compressor or rammed in by the forward motion of the aircraft. In either event any solid material entrained with the air will cause damage through either erosion or impact [2]. The importance of this failure mechanism cannot be rejected as according to the Boeing web page it causes losses of millions dollars every year to the airlines and airports [4].

The turbine blades operate at very high temperatures, very near of the edge of metallurgical alloy development. This implies that, additionally to be required to resist high mechanical loadings, three possible damage mechanisms affect these blades, being creep, multiaxial fatigue (associated with the interaction of low cycle fatigue in their longitudinal direction and vibrations induced by the gas flow in the perpendicular one) and high temperature corrosion. Under normal conditions, blades should never be operated at excessive temperatures for long enough periods to cause microstructural damage although some elevated temperature exposure for very limited periods is permitted. Even if FOD is not usual in turbine blades a potential origin of the failure must be found in the mechanism known as domestic object damage (DOD), which arises from a dislodged debris or component from another location of the engine [5]. It must be remarked that frequently the failure of one component unleashes the fracture of other ones.

The aim of the present paper is analyzing the root causes for the failure of two sets of blades. The first one consisted in a set of blades from the high pressure compressor of an engine. The second set was constituted by various turbine blades from another engine. Both engines failed prematurely.

Experimental Procedure

As indicated above, the failure analysis was performed on two sets of blades from two different aircraft engines which have failed prematurely. First one consisted in four high pressure compressor blades. Second one was constituted by a set of blades from the high pressure turbine.

The first step in both studies consisted in a visual examination of all the failed blades, task which was performed by the naked eye or with the help of a small stereoscopic microscope (x50) which allows detecting some facets that could have passed unattended. A special attention was paid to their fracture surfaces but without forgetting other aspects which could help identifying the origin of the failure or the operating mechanism. Moreover, this examination facilitates the decision about the areas most interesting to be analysed by scanning electron microscopy or where the metallographic samples will be obtained.

Once this visual examination was finished the fracture surfaces of the blades were cut allowing their fractographic analysis in the scanning electron microscope. The energy dispersive spectrometer, incorporated to this equipment, facilitates identifying those phases which have contributed to the failure.

This fractographic analysis allowed defining the most plausible origins of the failure and selecting those zones where metallographic samples would be obtained. That means destroying the blades and, consequently, was only carried out when it was considered that no more information could be obtained from the fractographic analysis. These metallographic samples were firstly examined in the unetched condition to reveal some features, such as secondary cracks, which could be hidden by the metallographic etching, making more difficult their detection. Afterwards, samples were etched and examined in both optical and scanning electron microscopes.

Results

Compressor blades: These blades were manufactured using a 718 nickel base superalloy which is the presently most widely used nickel alloy. This material was strengthened by γ'' particles. Visual examination revealed that most marked damage is sited between the top of the blades and their leading edge. A marked deformation and even a significant loss of material that was torn away, was observed in this areas. A first analysis of the morphology of this damage points to impact(s) as responsible of this damage. A second feature clearly observed during this visual examination was that the whole fracture surfaces of the blades were covered by a yellowish substance firmly adhered to them. Using the stereoscopic microscope allowed confirming the presence of this layer on the fracture surfaces. Most plausible origin of this layer was foreign material which hit against the leading edge of the blade, inducing the damage and remaining adhered on the fractured surfaces [6]. Nevertheless, the possibility that this substance would have been deposited once the failure has occurred cannot be completely discarded at this stage of the research.

Fractographic analysis of the fracture surfaces showed that the whole fracture surfaces were covered by thick deposits. Analysis of this layer by X- ray energy dispersive spectrometry (EDS) leaded to the presence of significant peaks of silicon and oxygen. The origin of these elements is associated to silica (sand) particles, which hit against the blade or was deposited on the fracture surface once it was broken. Other elements (magnesium, calcium and aluminium) which appear in small amount in this spectrum confirmed the foreign origin of this deposit. Only aluminium, and no in so large amount, entered in the composition of the alloy. Small peaks of nickel, chromium, iron and niobium arising from the base material of the blade are barely detected as their X-ray radiations are mostly absorbed by the foreign deposit. Small peaks of tungsten, cobalt and titanium, whose origin is later discussed, are also observed in this spectrum

Unfortunately, this thick layer precluded observing the actual fracture surface and detecting the fractographic facets which are present. Consequently cleaning was required to eliminate this layer but this operation can rub out these facets. In order to reduce this risk, firstly, a kindly cleaning was used but most of the deposit remained still adhered. A second cleaning, using a stronger agent, damaged some zones of the fracture surfaces, rubbing out the finest facets, but allowed detecting the ductile dimples which were present in the fracture surface pointing towards microvoid coalescence as the operating mechanism supporting the hypothesis which attributes failure to impact(s). Figure 1 (left) exhibits a low magnification micrograph of the damaged zone of one blade. The morphology of this damage is very similar to that reported in a work where small hard projectiles were fired at high speed against blade like fatigue specimen [7]. This represents a new support for the hypothesis that associates the failure to the impacts of foreign hard objects. On the right hand side of this figure it is presented a micrograph, taken in the same area, confirming the above commented presence of ductile dimples, and the operation of a microvoid coalescence failure mechanism.

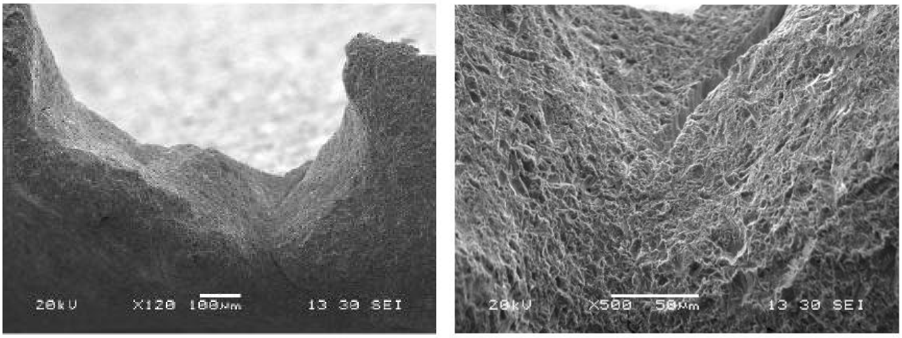

Figure 1. Damaged zone of one blade showing ductile dimples (left). Higher magnification of the same zone (right)

Even if the hypothesis which attributes failure to the impact(s) of foreign substances (FOD) is nearly confirmed metallographic samples of the selected areas were prepared. Metallographic etching revealed a marked deformation of the blades in the failure zones, being associated to the impact of hard particles against them. The examination of these samples in the optical microscope revealed the existence of an external coating, near fully disappeared in the proximity of the fracture. X-ray spectrometry of this external coating allowed being identified as cobalt bonded titanium carbide, explaining the presence of these elements in the previously commented spectrum.

One new support to this hypothesis which blames the impacts of foreign objects for the failure of the blades is given by the measurement of Vickers hardness, carried out on these metallographic samples. Values recorded in the impact zone are significantly higher (40HV0.5 more) than those obtained in the material away from the failure zone, this increase being associated to the strong deformation caused by the impacts. A more detailed discussion of this point is offered in reference [6]. Consequently, a conclusion attributing the failure of these blades to FOD was reached.

High pressure turbine blades: Material used for manufacturing these blades consisted in a Rene 142, nickel base superalloy, belonging to the second generation of directionally solidified materials, hardened by fine γ´ particles. These particles are coherent with the solid solution of alloy elements in the crystallographic lattice of nickel which constitutes the matrix. The precipitation of the γ´ phase is carefully controlled during manufacture to give a controlled size and morphology, resulting in a fine, ordered quasi-cubic precipitate to optimise high temperature strength by maximising the resistance to both diffusion creep at lower temperatures and power-law creep at higher temperatures. Moreover, the better creep behaviour of directionally solidified materials is linked to the alignment of grain boundaries in the loading direction and decreasing the number of boundaries in the transverse one.

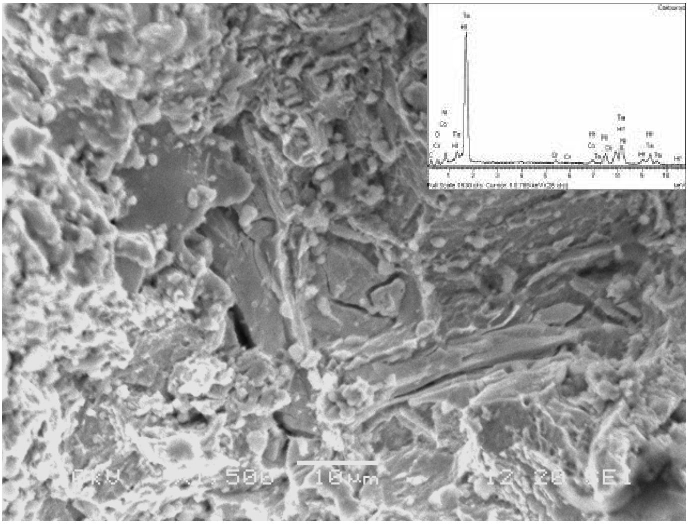

Figure 2. Carbides on the fracture surface of the presumably first failed blade and chemical analysis of one of these carbides.

Once again, visual examination revealed that the fracture surface was covered by a layer making difficult to observe the fractographic facets which are present below. Fortunately, in this case, it was possible observing the different morphology of one blade when compared with the other ones. One of them, which presumably was broken in the first place, exhibited a morphology pointing towards thermo-mechanical fatigue as the operating mechanism. A more thorough examination using the stereoscopic microscope allows identifying the periphery of one of the cooling holes as the most probable origin of the failure. Nevertheless, this point must be confirmed or denied by the posterior fractographic and metallographic analysis. On the other hand, the remaining blades showed the topography typical of impact fractures, without the presence of so large carbides.

Once finished the visual examination the fracture surface of this blade and those of two other ones were analysed in the scanning electron microscope. As it has been commented above, these fracture surfaces were covered by a layer of oxide that must be taken away in order to allow observing the fractographic facets. This cleaning operation was kindly carried out to preserve these facets to be rubbed out. Fortunately, this labour was successful. Fractographic analysis of the blade presumably broken by fatigue revealed a noticeable number of primary carbides on the fracture surface indicating that they played an important role in the fracture process. One good example of these large carbides is shown in figure 2 micrograph. X-ray energy dispersive spectrometry allows identifying these large particles as hafnium and tantalum carbides. This analysis is also included in the figure. Carbon is intentionally added to the alloy to form carbides which retard grain boundary sliding and improve the creep performance of the component. However this is achieved by a fine, homogeneous distribution of small carbides and no by these very large ones heterogeneously distributed across the matrix. These large carbides not only do not improve the creep performance but induce a significant stress concentration promoting cracking and decreasing the life of the blade.

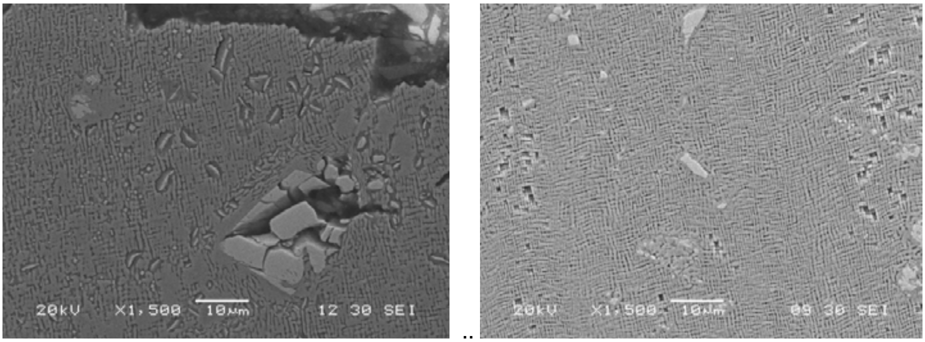

Figure 3. Microstructure of the presumably firstly broken blade (left) and one of those broken by impact (right)

When it was considered that this fractographic analysis would not yield additional information cuts, including the presumably origins of the failures were performed for obtaining the metallographic samples. An interesting microstructural difference was found between the presumably firstly broken blade and the other ones. Figure 3, obtained in the scanning electron microscope examination of the samples clearly shows this difference. On the left hand side of this figure the microstructure of the firstly broken blade can be seen. This microstructure is constituted by a matrix of solid solution and a large number of small γ' phase particles but also very large, broken, primary carbide and some worm shaped white particles are detected. These particles are clearly differenced from the cubic γ' phase, which is aligned along their axis. X-ray energy dispersive analysis leaded to a large peak of tantalum, another one, clearly lower, of nickel and a small peak of chromium. Taking into account this composition the identification of these particles as Laves phases or other of the so called topologically closed packed (TCP) intermetallic brittle compounds looks highly reasonable. On the opposite hand, those blades presumably broken by impact exhibited much finer carbides and no sign of brittle phases. However, γ′ distribution is very similar in all the blades pointing towards identical working conditions and rejecting the hypothesis that would have attributed the failure of the first blade to higher temperature associated to a deficient cooling. The micrograph on the right hand side of figure 3 which was obtained at the same magnification, exhibits this microstructure.

These large carbides, which constitute stress concentration points, are easily broken when subjected to the service stresses. This fracture nucleates a microvoid, which under successive loading fatigue cycles, growths and can coalesce with those at other nearby sited carbides. The final result is a crack which can lead to the fracture. Another aspect that must be conveniently remarked is the presence of the above commented TCP particles, which is considered unacceptable as they negatively affect to the performance of the components. In order to prevent their formation the iron content of the alloys which operate at the highest temperature iron is restricted to a very low percentage. Nevertheless, other alloy elements, such as tantalum, niobium, molybdenum, tungsten or chromium also promote the formation of these phases. The negative effect is not only associated to the nature of these phases but also their geometry plays a role, having being claimed that those of acicular shape promote the failure [8]. In the present case the particles have not an acicular shape but they are elongated enough to promote the failure.

Nevertheless, the most interesting result of the examination of these metallographic samples was confirming that at the origin of the failure of the presumably firstly failed blade a large, broken, primary carbide was present (figure 4, left). Moreover, some other large, primary, carbides can be seen at or near the fracture. Consequently, the hypothesis attributing failure to thermo-mechanical fatigue promoted by the presence of the large carbides is strongly supported. Moreover, it can be observed in this same micrograph the phenomenon designed as alloy depletion, which is the loss of some alloying elements, at the periphery of the blade. However, it was considered that this loss did not contribute significantly to the failure process.

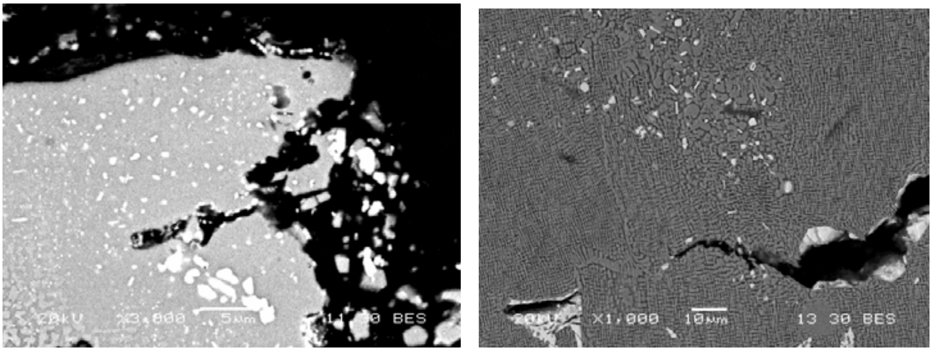

Figure 4. Large size carbides broken at the origin of the failure (left) or along a secondary crack (right)

On the right side of this figure a micrograph showing a secondary crack is presented. Once again the large primary carbides have affected negatively to the in service performance of the blade. This micrograph clearly exhibits that the crack has preferentially progressed along this large broken carbides and some TCP compounds have also contributed to the cracking process. Consequently, a conclusion linking the failure of this blade to an incorrect microstructure is reached.

Once the failure is associated to these large primary carbides it must be found a reason for their presence in the microstructure. Primary carbides are formed, as their own name indicates, during the solidification of the alloy, being later hardly solved as this process requires reaching a temperature in the range between 1200 and 1260º C to achieve it. The temperature of γ´ solution treatment is not high enough to solve this carbides, labour that is even harder when their large dimensions are taken into account [9]. The use of higher temperatures trying to solve them is risky as localized melt can be induced [10]. As a consequence, most of the primary carbides present in the as cast component are not significantly modified by the subsequent heat treatments or in-service operation, even if some reactions converting these carbides in some different ones have been claimed [1, 8].

As previously commented, metallographic examination of the other blades revealed a completely different microstructure. As it is evident in the micrograph exhibited on the right hand side of figure 2, the microstructure of these blades is constituted by the solid solution of the alloy elements in the crystallographic lattice of nickel, the γ´ phase, and some primary carbides. Nevertheless, the size of these primary carbides is significantly smaller than those observed in the firstly broken blade, this fact being easily confirmed by a comparison between the micrographs of figure 2. Moreover, no sign of brittle phases is observed in these samples. Due to all these factors these blades operated much more satisfactorily and they would have not failed if the broken pieces of the firstly failed one had not impacted against them.

Metallographic study of the samples obtained from these blades not only revealed this difference in the microstructure but also constitutes a good help for confirming that they failed by the impacts of the loose fragments. Additionally to the marks left by these impacts the strong deformation suffered by the material as a consequence of them can be also observed. This was easily noticed, as the γ´ phase lost their original perpendicular distribution and passed to be aligned following the direction of deformation induced on the material [9]. Even if the possibility of FOD as responsible of these failures cannot be categorically rejected, the position of these blades, clearly inside the engine, and the absence of foreign deposits on their fracture surfaces pointed towards domestic object damage (DOD) associated to the impacts produced by the broken pieces of the firstly failed blade, as the operating mechanism. Consequently, the logical sequence of the fracture process of the whole set can be defined. The blade which possesses a deficient microstructure failed prematurely by thermo-mechanical fatigue and its fragments induced the fracture of the other blades ones by impacting against them.

However, it rested finding a logical explanation to these differences in microstructure. As it has been indicated above the γ´ distribution is very similar in all the blades and higher temperature of this blade due to an incorrect cooling cannot be blamed for it. Moreover, a deficient cooling can induce a more advanced ripening of the particles of γ´ phase or even in the precipitation of the brittle intermetallic compounds, but no for the presence of these so large carbides. Considering all these results, the possibility that this blade, even when it was inserted in the same disc than the other ones belonged to a different heat is supported. No clear reason for this abnormal situation was found.

Conclusions

a. The fractographic analysis of the root causes of the premature failure of two sets of blades has been performed.
b. Compressor blades failed by the so called foreign object damage (FOD) mechanism due to the impact(s) induced by the ingested sand and stones.

c. The foreign substance present on the fracture surfaces of these blades was so firmly adhered and encrusted that even after a second strong cleaning was not fully removed. This fact supports the above formulated hypothesis associating failure to the impacts induced by this foreign substance, mainly silica.
d. Thermo-mechanical fatigue was blamed as the mechanism responsible for the premature failure of the first turbine blade, fracture being initiated at one of the cooling holes.
e. The microstructure of this blade presented a noticeable number of large, primary, hafnium and tantalum carbides which acted as stress concentration points promoting cracking. These carbides were formed during the solidification of the alloy and were not solved during the posterior process.
f. The origin of these so large carbides must be found in a marked segregation of the alloy elements in the cast component. Moreover, a noticeable number of worm-shaped particles, whose analysis identified as brittle intermetallic (topologically close packed TCP) phases were observed in this blade. This deficient microstructure cannot be admitted.
g. The other blades of this set, which presented a correct microstructure, constituted by the γ' phase and small carbides, without the presence of this TCP compounds, failed by a domestic object damage (DOD) mechanism due to the impacts produced by the fragments of the firstly broken one.
h. Microstructural differences between the fatigue broken and the other blades points towards a different manufacturing route and hardly belonging to the same heat. It was found no reason for this discrepancy.

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