Failure analysis of turbo-blower blades

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Abstract

Twelve percent chromium martensitic stainless steel blades of the medium-pressure stage of a thermoelectric centre turbo-blower broke during use. The present work will investigate the possible causes for the premature failure of these blades and the condition of the high-pressure blades. The results indicated that at least one of the blades of the medium-pressure stage failed by a corrosion-fatigue mechanism, whose nucleation was associated with the presence of corrosion pits on its suction side. The high-pressure blades presented hardness bellow the specification and presence of corrosion pits and cracks. The softening of the microstructure (from 250 to 220 HV) could not be explained by microstructural instability during the intermittent use of the blade, indicating that either the material was initially tempered at a higher temperature or that the working temperature is higher than the predicted. A rough and conservative estimation of the residual life of the high-pressure blades indicated that the high-pressure blades should also replaced. The quality of the feed water (from the sea) used for the production of the vapour should be optimised and any action taken to improve the cleanliness and dryness of the equipment during its intermittent use may improve the life of the blades.

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1. Introduction

The medium-pressure stage blades of a thermoelectric centre turbo-blower broke during use. This equipment worked continuously between 1976 and 1982 and 1993 and 2002, and intermittently between 1983 and 1992, under the following conditions: vapour entry temperature = 440 °C and exit temperature = 220 °C; entry pressure = 40 kgf/cm²; exit pressure = 20 kgf/cm²; turbine rotation: 3600 rpm. The blades were manufactured in 12% chromium martensitic stainless steel. In 1997, the blades of the high-pressure stage failed prematurely and were replaced. The present work will investigate the possible causes

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for the premature failure of the medium-pressure stage blades and the present condition of the high-pressure blades.

2. Experimental procedure and results

Figs. 1(a) and (b) show the turbo-blower after dismounting, while Fig. 2(a) shows, in more detail, two broken blades of the medium-pressure stage. The fracture surfaces are relatively plane and the arrows indicate some crushing marks on the pressure side (concave) of the blade, caused by the deblading operation. An opposite view of the same blades (see Fig. 2(b)) reveals the presence of superficial deformation marks due to the impact of the blade against other detached pieces, which broke previously.

Figs. 3(a) and (b) show the microstructure of the medium-pressure blades, which is composed of tempered martensite and spheroidised carbides, featuring a hardness of 245 HV$_{0.5}$ (21 HRC). The stoichiometry of the carbide particles could not be determined by SEM microanalysis due to its small size. The micro-

Fig. 1. Turbo-blower: (a) rotor-stage of the turbine after dismounting. Only high-pressure stage blades are present; (b) details of the high-pressure stage blades.
structure of the high-pressure blades (see Figs. 4(a) and (b)) is composed of tempered martensite and spheroidised carbides, comparatively coarser, and features a hardness of 220 HV$_{0.5}$ (16 HRC).

Fig. 5(a) shows the fracture surface of a medium-pressure blade, featuring crack-arrest marks and steps, indicating multiple crack nucleation on the suction side (convex) of the blade, followed by stable crack growth. Fig. 5(b) shows the presence of a secondary cracking parallel to the fracture surface. The presence of superficial irregularities on the suction side (convex) of the blade is also observed.

In order to further examine the secondary crack and superficial cavities, a cross-section examination was carried out. Figs. 6(a) and (b) reveal the morphology of the cavities (diameter of 250 $\mu$m), featuring a layer of chromium-iron oxide with presence of chlorine. Figs. 6(b) and (c) present the morphology of the secondary cracks (length between 100 and 150 $\mu$m), also featuring a thin layer of chromium-iron oxide with presence of chlorine. The results indicate the formation of corrosion pits on the suction side (convex) of the blade and the action of stable crack growth by a corrosion-fatigue mechanism.

Visual inspection of the high-pressure stage after sand blasting operation also revealed the presence of corrosion pits on the suction side (convex) of the blades, near the root and in the middle of the face (see Figs. 7(a) and (b)). The cross-section examination reveals the presence of corrosion pits (diameter of 150 $\mu$m) and secondary cracks (length of 30 $\mu$m) nucleating at corrosion pits (see Fig. 8).
3. Discussion

The results indicate that the corrosion pits promoted the nucleation and growth of corrosion-fatigue cracks on the suction side (convex) of the blades, responsible for the premature failure of at least one medium-pressure blade. 12% chromium martensitic stainless steels (see Table 1) are corrosion resistant, but prone to severe localised corrosion and corrosion-fatigue under marine atmospheres due to the presence of chlorine [1,2].

Pitting corrosion is a localised attack, which produces small spherical cavities. Corrosion pits act as stress raisers and, additionally, present internally an electrolyte featuring chemical composition and pH different from the external environment. Previous investigations on the pitting corrosion mechanism of stainless steels under marine atmospheric environment showed that pitting is preferentially nucleated during the dry cycle, after a critical chloride concentration has been reached. The repassivation process, which begins as the specimen surface dries up, could be completed within the dry-cycle, if perfect dryness of the pit cavity takes place. The localised increase in the aggressiveness of the solution over the attacked (anodic) areas promotes further nucleation of corrosion pits and the growth of existing pits. In this sense, the humidity in the atmosphere and the presence of an aggressive environment, such as salt air, can cause the condensation of sea-salt vapour on certain parts of the turbine when they are idle. This observation

Fig. 3. Medium-pressure stage blades, microstructure. Tempered martensite (ferrite and spheroidised carbides). SEM-SEI: (a) 2500×; (b) 10,000×.
indicates that any action taken to improve the cleanliness and dryness of the equipment during its intermittent use may improve the life of the blades. Some utilities, for instance, heat their turbines, when they are not used to avoid this problem, and promote periodic cleaning (abrasive blasting, washing with water and organic solvents). The investigated turbo-blower worked 40% of its life as a replacement equipment (intermittent use) and the on site inspection of the turbo-blower confirmed, furthermore, that the feed water used for the production of the vapour was originated from the sea, facts which enhance the aggressiveness of the marine atmosphere [3–6].

The results indicated that the high-pressure stage blades showed after 5 years of use the presence of corrosion pits (150 μm) and corrosion-fatigue cracks (30 μm) on the suction side (convex) of the blades. Additionally, their tempered martensite microstructure is comparatively coarser and features a lower hardness (220 HV, equivalent to an UTS = 700 MPa, against 245 HV or UTS = 750 MPa for the medium-pressure stage blades), indicating that the fatigue life of the high-pressure stage blades should be comparatively lower.

This material is usually tempered between 110 and 160 °C above the working temperature to insure thermal and microstructural stability. As the maximum working temperature in the high-pressure stage is approximately 440 °C (vapour entry temperature), the tempering temperature should be carried out between 610 and 650 °C, resulting in a minimum hardness of 250 HV [7,8]. The softening of the micro-

![Fig. 4. High-pressure stage blades, microstructure. Tempered martensite (ferrite and spheroidised carbides). SEM-SEI: (a) 2000×; (b) 10,000×.](image)
structure (from 250 to 220 HV) could not be explained by microstructural instability during the use of the blade, as 44,000 h at 440 °C is equivalent to 0.1 h at 650 °C (same Larson–Miller parameter, see Eq. (1)), fact which would promote an insignificant softening [8]. This result indicates that either the material was initially tempered at a higher temperature or that the working temperature is higher than that predicted, by the equation

$$LMP = \frac{T(20 + \log t)}{1000},$$

where LMP is the Larson–Miller parameter; \(T\) is temperature in Kelvin, and \(t\) is the time in hours.

Previous work [9] studied the effect of manufacturing residual stresses and superficial defects on the fatigue strength and crack growth of gas turbine compressor blades, manufactured in high chromium steels. The results indicated that the presence of corrosion pits (diameter of 1.0 mm and depth of 0.3 mm) decreased the bending fatigue limit of a 12.8 Cr martensitic steel test-pieces – tempered at 983 K, as follows:
- from 420 to 310 MPa (initial condition with surface plastic deformation to induce deeper residual compressive stresses);
- from 400 to 220 MPa (initial surface condition: mechanical polishing).

Additionally, the results of the rotating bending fatigue test on new 12% Cr martensitic steel blades showed that the presence of a corrosive environment (sea-salt vapour) decreased the fatigue limit from 320

Fig. 5. Fracture surface of a medium-pressure stage blade: (a) beach marks and steps, indicating multiple crack nucleation on the suction side of the blade, associated with the presence of superficial irregularities (cavities). SEM-SEI, 20×; (b) presence of secondary cracking (see arrow) on the suction side of the blade, parallel to the fracture surface. SEM-SEI, 21×.
to 180 MPa. A rough and conservative estimation of the fatigue life of the high-pressure blades was carried out, using the data available in this investigation [9] for a 12% Cr martensitic stainless steel and the crack growth equation shown in Eq. (2)

$$N = \frac{1}{A\sigma_f^n} \int_{c_1}^{c_2} \frac{dc}{[Y\sqrt{C}]^n},$$

where the symbols are defined as follows: $A$ and $n$, parameters of the Paris equation; $\sigma_f$, nominal stress in the section of the blade where crack initiates (MPa); $c_1$ and $c_2$, initial and final crack size ($m$); $Y$, crack geometrical factor; $N$, number of cycles for the crack to grow from $c_1$ to $c_2$. 

Fig. 6. Microstructural characterisation of the medium-pressure stage blade: (a) cross-section along the cavities showing corrosion pits with an internal layer of oxide on the suction side of the blade. SEM-BEI, 129×; (b) cross-section along the cracks on the suction side of the blade, showing the presence of an internal layer of oxide. SEM-BEI, 800×; (c) typical EDS microanalysis spectrum of the oxide layer revealing the preponderance of iron, chromium and oxygen and additional presence of chlorine.
For the life estimation in sea-salt vapour (see Eq. (3)), the following assumptions were made: frequency of 60 Hz; continuous use of the turbo-blower; room temperature operation; crack nucleation at the suction side of the blade; \( Y \) value constant and equal to 0.7; \( A \) \((5.60 \times 10^{-12}) \) and \( n \) \((4.41) \) values of a 12\% Cr martensitic stainless steel (calculated for room temperature) [9]. The initial crack size of the high-pressure blade was considered as the measured crack size (30 \( \mu \)m) plus the pit cavity depth (150 \( \mu \)m), amounting to 180 \( \mu \)m, and the final crack size assumed as 3 mm (size of the region of stable fracture for the fractured medium-pressure blade). For the life estimation in air (see Eq. (4)), the same assumptions were made, but values of the Paris equation parameters of a 12\% Cr martensitic stainless steel have been changed to \( A = 3.58 \times 10^{-12} \) and \( n = 4.16 \) [9].

\[
N(\text{sea-salt vapour}) = 2.1 \times 10^{16} \times \sigma^{-4.41}, \tag{3}
\]

\[
N(\text{air}) = 1.2 \times 10^{14} \times \sigma^{-4.16}. \tag{4}
\]

The fatigue limits \( \sigma_{FL} \) for both conditions were estimated using Eq. (5) and assuming a crack length of 200 \( \mu \)m, \( \Delta K_{air} = 5.47 \) MPa m\(^{1/2} \) and \( \Delta K_{sea-salt vapour} = 4.20 \) MPa m\(^{1/2} \) [9]. The results indicated a fatigue limit of 552 MPa (air) and 424 MPa (sea-salt vapour)

\[
\sigma_{FL} = \Delta K_{th}/[Y \cdot (c_{in}^{1/2})]. \tag{5}
\]

Because fatigue is basically a slip process, any environment that affects slip will affect the fatigue crack propagation rate. In corrosive environments, the fatigue crack propagation can also be affected by the
environmental interaction with the material ahead of the crack front, resulting in brittle striation, quasi-cleavage, cleavage or intergranular decohesion fracture modes (mixed fracture mode or corrosion-fatigue). Previous investigation on 12% Cr martensitic stainless steel tempered fatigue showed that the fatigue limit in 3% sodium chlorine solution (100 MPa) is much lower than in air (500 MPa). For the same fatigue strength (500 MPa), the fatigue life decreased from $10^7$ cycles (in air) to $10^5$ cycles (in salt solution). For lower $\Delta K$ values, however, the corrosion fatigue crack growth rate in brine was slightly lower than that in air. This anomaly was believed to be due to the high testing frequency (30 Hz), which resulted in fluid not...
being able to escape from the crack tip region (wedge action of the trapped fluid reducing the stress intensity at the crack tip). When the material is susceptible to environmental attack, the effect of the environment is greater at lower frequency, when longer times allow the environment to affect the material ahead of the crack tip [10,11].

Table 2 shows the predicted life of the high-pressure blades in the Paris regime for both conditions (air versus sea-salt vapour), considering stable crack propagation even for loads below the fatigue limit. The release of some of the contour conditions (effects of working temperature of 440 °C and the intermittent use of the turbo-blower on the corrosion-fatigue crack-growth) should decrease even further the residual life of the blade in both environments. This conservative estimation indicates that the replacement of the high-pressure stage blades should be carried out during the present maintenance operation due to presence of corrosion pits (diameter of 150 μm), corrosion-fatigue cracks (length of 30 μm) and the low-hardness of the material. Finally, the life prediction results unexpectedly indicated that the residual life in sea-salt vapour is approximately fifty times greater than air, fact which could not be proved experimentally. This final observation indicates that the fatigue life prediction results should be examined with great care.

4. Conclusions

1. The results indicated that at least one of the blades of the medium-pressure stage failed by a corrosion-fatigue mechanism, whose nucleation was associated with the presence of corrosion pits on the pressure side (concave) or suction side (convex).
2. The high-pressure blades presented hardness below the specification and surface presented corrosion pits and secondary cracks on the pressure side (concave) or suction side (convex).
3. The softening of the microstructure (from 250 HV to 220 HV) of the high-pressure blades could not be explained by microstructural instability during intermittent use of the blade, indicating that either the material was initially tempered at a higher temperature or that the working temperature is much higher than the predicted.
4. A conservative estimation indicates that the replacement of the high-pressure stage blades should be carried out during the present maintenance operation.

5. The quality of the feed water (from the sea) used for the production of the vapour should be optimised and any action taken to improve the cleanliness and dryness of the equipment during its intermittent use may improve the life of the blades.

References