PERGAMON

Engineering Failure Analysis 8 (2001) 521–528

ENGINEERING FAILURE ANALYSIS

www.elsevier.com/locate/engfailanal

Failure analysis of high temperature studs

Sandip Ghosh Chowdhury *, Pravesh Kumar, Swapan K. Das, D.K. Bhattacharya, N. Parida

Materials Characterization Division, National Metallurgical Laboratory, Jamshedpur 831007, India

Received 4 July 2000; accepted 30 July 2000

Abstract

Studs in the interceptor valve of a 110 MW unit failed after a service life of 148,700 h. The studs were operated under a steam pressure of 35 kg/cm^2 and a temperature of 535°C. The studs were fractured at one end of the threaded end. Various techniques were employed to analyse the failure of the studs. It has been concluded that the failure of the studs was due to reverse temper embrittlement. The failure was delayed due to the presence of Mo and V. To reduce the tendency to this kind of failure, the following steps were recommended: (a) reduce the phosphorus content in the steel to a low level or (b) reduce the grain size to about 10 μm. © 2001 Elsevier Science Ltd. All rights reserved.

Keywords: Temper embrittlement; Fasteners; Heat treatment; Intergranular fracture; Impact energy

1. Introduction

A steam turbine consists of many bolted joints. The main requirement of these bolts at high temperature is to maintain steam tight joints without relaxing or fracturing. Thus, the stress-relaxation characteristic of the bolt material is an important criterion because for a given joint, the load required to exceed the steam load is applied to the flange area by tensile loading of the bolt. The stress-relaxation characteristics must be such that even when the stresses are relaxed, they are still in excess of the design stress. The elastic strains produced by initial tightening of the bolts are progressively converted to creep strain, thus reducing the effective load on the joint. This leads to leakage in the turbine. The fracture of a bolt resulting from metallurgical causes or poor design can lead to severe damage.

Generally, the bolts fail due to creep damage or brittle fracture. During every tightening due to stress-relaxation, creep strain will be accumulated and it will reach the maximum creep ductility of the material. Brittle fracture of bolts can occur if the fracture toughness of the material is too low. The bolt materials are made of high strength steel which has inherently low toughness. The toughness can be further degraded in service by temper embrittlement. This is also known as reversible temper embrittlement (RTE). RTE refers to the decrease in notch toughness when tempered low alloy steels are heated in or slowly cooled through a

* Corresponding author. Tel.: +91-657-271709; fax: +91-657-270527.
E-mail address: sgc@csnml.nml.res.in or sandip-gc@hotmail.com (S.G. Chowdhury).

PII: S1350-6307(00)00019-4

critical temperature range i.e. 300–600°C. This form of embrittlement is of major concern to the integrity of engineering components that operate within the critical temperature range. This is manifested as an increase in the ductile-brittle transition temperature together with a change in the low temperature fracture mode from cleavage to intergranular. The problem of temper brittleness of CrMoV steels was detected during disc failures at Hinckley-Point in England in 1969 [1] and subsequently a rotor failure at Gallatin, Tennesse in 1974 [2].

High temperature turbine bolts in fossil-fired power plants are subjected to temperature at or beneath a maximum steam temperature of 540°C. These bolts and studs are generally made of Cr–Mo–V steels strengthened by stable V_4C_3 precipitates. Various failures of bolts of this grade of steel have been reported to be due to temper brittleness [3–6].

In the present paper, failure of studs in the interceptor valve of a 110 MW unit which were found broken during overhauling was investigated [7]. The position of the studs in the valve is shown in Fig. 1. The studs completed a service life of 148,700 h. The studs were operated under high quality superheated steam subjected to maximum pressure of 35 kg/cm^2 and temperature of 535°C. Studs were fractured from one end of the threaded end as shown in Fig. 2.

2. Experimental procedure

The chemical composition of the stud material was analysed by wet chemical analysis. Optical metallography and scanning electron microscopy (SEM) examinations were carried out to ascertain the microstructure of the material. Hardness measurements was made on the Vickers scale with 20 kg load. Impact specimens were made and tests were carried out as per ASTM E-23. The notch was made perpendicular to the stud axis. After the tests, fractography was carried out under SEM.

To compare the degradation of microstructure and mechanical properties, one new stud was taken and similar tests were conducted on it.

3. Results

3.1. Chemical analysis

Chemical analysis of the failed stud material is shown in Table 1. The chemical composition of the new stud was tabulated for comparison.

3.2. Microstructure

Samples were taken from the fractured end of the failed stud for metallographic investigation. As mentioned earlier, the fracture took place from one of the threaded ends. Samples of the new stud were also taken just below similar thread ends. Standard metallographic technique was employed for sample preparation and they were etched with Nital reagent. SEM micrograph shows tempered bainitic structures in both cases (Fig. 3). Some of the prior austenite grain boundaries have carbides rich in Cr and Mo. This kind of microstructure is quite common for this grade of steel.

3.3. Mechanical properties

Average hardness values of three indentations were obtained. Both studs have hardness values in the range of 240–250 VPN.

Impact tests were carried out at room temperature in a Wolpert Impact Testing Machine. Two samples were tested for each condition and the average value was taken. The impact energy values for the failed and the new stud were 9 and 188 J, respectively.

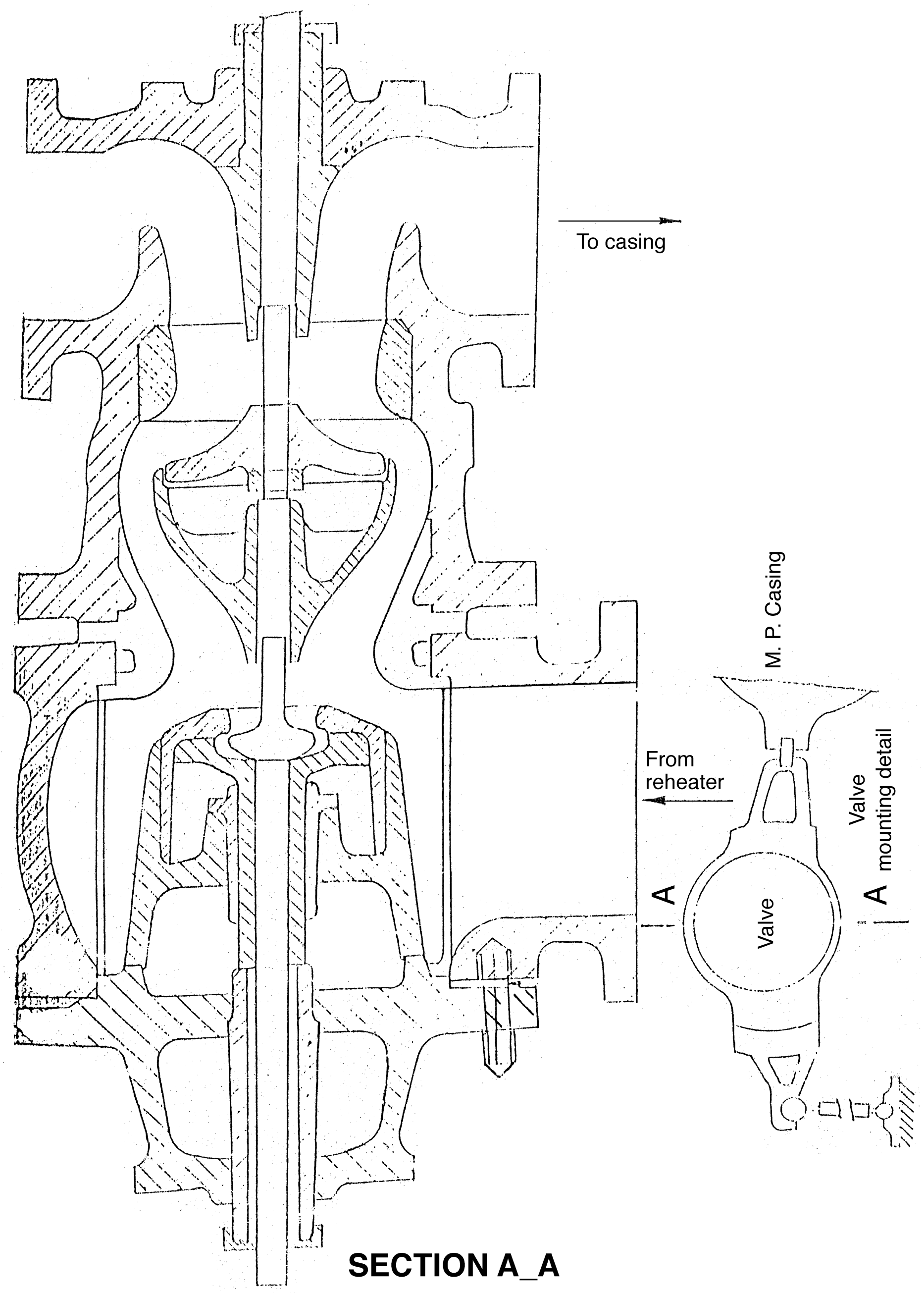


Fig. 1. Schematic diagram showing the position of the stud on the valve.

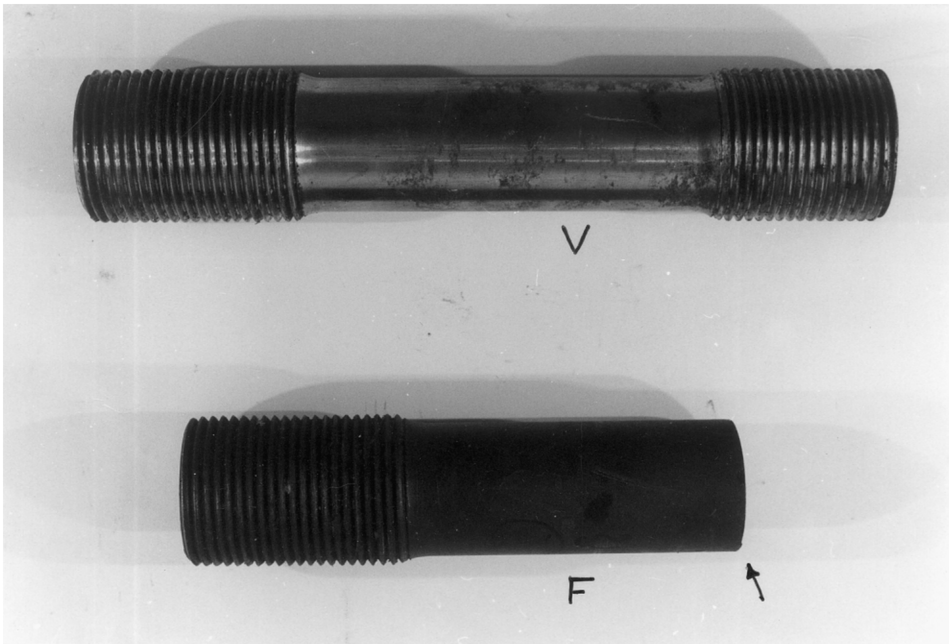


Fig. 2. Photograph of the virgin stud (V) and the failed stud (F) (arrow indicates the point of fracture).

Table 1
Chemical composition (wt.%)

	C	Cr	Mo	V	Si	Mn	P	S
Failed stud	0.20	1.56	0.78	0.36	0.19	1.24	0.038	0.035
New stud	0.21	1.82	0.88	0.35	0.22	0.81	0.033	0.031

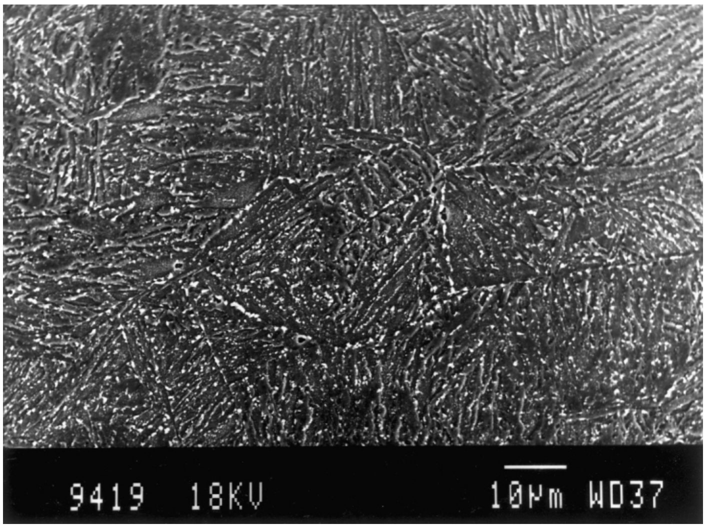


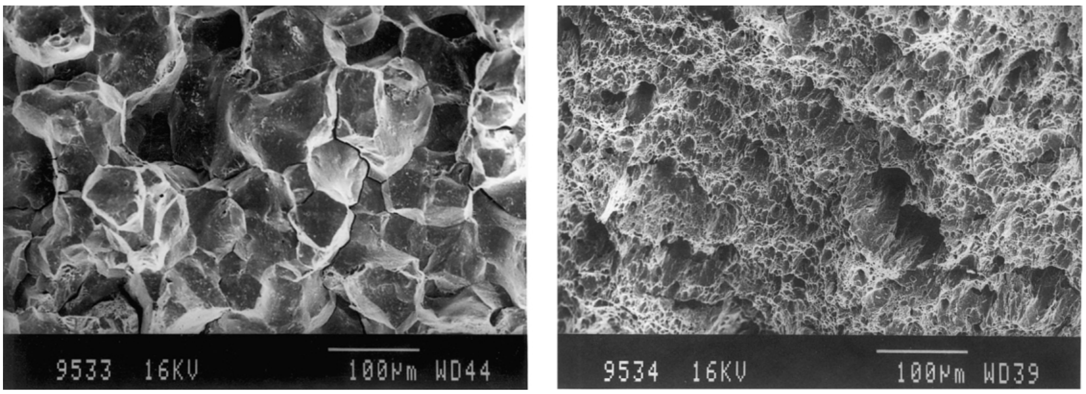
Fig. 3. Micrograph of the failed stud.

3.4. Fractography

The fracture surface of the failed stud was fully covered with iron oxides. Thus, evidence of fracture morphology could not be obtained. However, the oxide scales on the fracture surface were removed with active reagents (pickling solution). The macrophotograph of the fracture surface is shown in Fig. 4. The total fracture area can be divided into two zones: slow propagation zone (A) and fast fracture propagation zone (B). Within zone A, it can be visualised that cracks might have initiated from more than one point.

Fracture surfaces of the impact tested specimens were examined in SEM. The samples made from the failed stud showed complete intergranular brittle fracture morphology (Fig. 5a); whereas the sample from the new stud showed ductile dimple morphology (Fig. 5b). The differences in the impact energy values in conjunction with the different fracture morphology indicates that the failure took place due to intergranular embrittlement.

Fig. 4. Macrograph of the fracture surface of the failed stud.



(a) (b)

Fig. 5. (a) Fractograph showing the fracture surface of the impact tested specimen of the failed stud; (b) fractograph showing the fracture surface of the impact tested specimen of the virgin stud.

4. Discussion

From the above experimental results, it is apparent that the stud material undergoes a kind of embrittlement the rate of which is very slow. It ultimately leads to a delayed failure. In tempered steels, there are various kinds of embrittlement which take place due to structural changes. These are classified as: tempered martensitic embrittlement, temper embrittlement, hydrogen embrittlement and liquid metal embrittlement.

The studs were in contact with superheated steam at around 500°C. Thus, the possibility of hydrogen embrittlement can be ruled out, because at this temperature hydrogen will diffuse out of the specimen. It is pointed out that the studs failed after 148,700 h of service as well as no low melting point element such as Cd was found. Thus the possibility of liquid metal embrittlement can also be ruled out; if it is not so, the stud could not have completed so many hours of service at this temperature.

It is well known that temper martensitic embrittlement occurs at temperatures between 260 and 370°C. The phenomenon takes place due to decomposition of retained austenite to cementite in the interlath region of martensite plates along the prior austenite grain boundaries. In the present case, the microstructure and the hardness pattern show that the material was properly hardened and tempered before service. Thus the possibility of tempered martensitic embrittlement can also be ruled out; if it is not so, it would have failed long before.

Temper embrittlement of this category of steel takes place due to impurity segregation at the grain boundaries and finally decohesion of the grain boundary. This leads to intergranular fracture morphology. It is signalled by a material toughness loss. It is pointed out that grain boundary segregation depends on the alloying elements of the steel [8]. In Cr–Mo steel, phosphorus is the major embrittling element. However, the kinetics of phosphorus segregation in this steel is different from that of other steels because of Mo. Molybdenum had a strong effect in lowering the solubility of phosphorus in iron [9]. This was interpreted as due to a molybdenum–phosphorus scavenging reaction; so, if molybdenum is free in solution, it appears to either prevent phosphorus from segregating to the grain boundaries or to reduce the embrittling potency of phosphorus at the grain boundary [10–12]. It is also observed that the molybdenum effect disappears after very long ageing times at elevated temperatures due to precipitation of molybdenum carbides resulting in the ultimate release of phosphorus which was then free to segregate to grain boundaries [13]. Generally, molybdenum carbide forms quite late in the tempering process. It is reported that if the steel is tempered at low temperatures for a short period of time, very long (thousands of hours) ageing time is required at lower aging temperatures for the embrittlement to be effective; whereas if tempering is carried out at high temperature (~600–700°C) for a very long time, molybdenum carbides will precipitate out fully and embrittlement will be very rapid after subsequent ageing [14].

Although molybdenum is an effective element to reduce the susceptibility for temper embrittlement, the precipitation of molybdenum as carbide must be taken care off. (*It is observed by McMahon et al. [10] that the addition of 0.6% Mo in a P-doped NiCr steel resulted in the delay in embrittlement by several hundreds of hours during aging at 500°C.*) To avoid that, it is observed that vanadium is added in this grade of steel [15]. Vanadium is strong carbide former compared to Mo and Cr. V initially forms MC type carbide; this changes the Mo:C and Cr:C ratios. The increase in Mo:C ratio is favourable for Mo_2C type carbides and that of Cr:C ratio is favourable for Cr_7C_3 carbides in this grade of steel. These changes in carbide formation sequences basically slow down the precipitation of Mo as carbides. When the Mo in solid solution in the ferrite matrix is fully removed, P is free to segregate and the material thus gets embrittled.

In the present steel, the level of phosphorus is 0.038, Mo is 0.84 and V is 0.35. The presence of V alters the carbide formation schedule and allows the material to be in service for 148,700 h before failing by embrittlement. Thus, the failure of the studs was delayed due to presence of Mo and V in the steel which did not allow P to segregate to the grain boundary. With slow and gradual changes in carbide composition, Mo was removed from the solution. Finally P was free to move to grain boundaries and fracture took place by grain boundary decohesion.

Three principal factors which affect RTE in this kind of steel are:

i. bulk P content
ii. grain size
iii. accumulated strain

Recently an equation has been proposed based on studies on a series of Cr–Mo–V turbine bolts [16]

$$d \times (\%P) \times (\%\varepsilon)^{0.64} = 0.772$$

where d is the grain size, P is the phosphorus content and ε is the accumulated strain. For the present grade of material, d is 50 μm and $\%P$ is 0.038. From the above equation, it can be found out that strain is 0.006 which is very low.

The observed effect of strain on the extent of grain boundary phosphorus segregation could possibly be the result of some change in the localised diffusion activity or alteration in the type of diffusion path caused by an increase in lattice energy as a result of an increase in the dislocation density. It must be emphasised here that at this point of strain, the service conditions of these bolts should have caused the full equilibrium amount of phosphorus to diffuse to the grain boundaries and thus embrittled the bolts.

5. Conclusions

Molybdenum which acts against temper embrittlement of low alloy steel must be in solution in the ferrite phase for components which undergo long high temperature exposure. Mo basically scavenges the P from the grain boundary and prevents embrittlement. To increase the effect of Mo, V is added. V which is a strong carbide former delays the precipitation of Mo rich carbide. Thus, the removal of Mo from the ferrite phase gets delayed by several thousands of hours and this reduces the susceptibility of the material to being embrittled. Thus, the present case of failure is a case of delayed temper embrittlement of this grade of steel.

It is therefore recommended: to use low phosphorus steel; to use 2.25CrMoV steel; to replace the studs after any one of the same batch has failed.

Acknowledgements

The authors wish to thank Professor P. Ramachandra Rao, Director, National Metallurgical Laboratory, Jamshedpur for his kind permission to publish the paper.

References

[1] Kalderon D. Proc Inst Mech Eng London 1972;186:341.
[2] Kramer LD, Randolph DD. Analysis of the Tennessee Valley Authority, Gallatin unit No. 2, Turbine Rotor Burst, Presented at Winter Annual Meeting, ASME, New York, 1976.
[3] Low Jr. JR, Stein DF, Turkalo AM, Laforce RP. Trans-AIME 1968;14:242.
[4] Kaneko H, Nishizawa T, Tamaki T, Tanifuji AJ. J Japan Inst Met 1965;29:166.
[5] Yu J, McMahon Jr. C.J. Metall. Trans. A 1980;11A:291
[6] Yu J, McMahon Jr. C.J. Metall. Trans. A 1980;11A:277
[7] Briant CL, Banerji SK. Treatise on Materials Science and Tech 1983;21:25.
[8] Qu Z, McMahon Jr. CJ. Metall Trans A 1983;14A:1101.

[9] Qu Z, Kuo KH. Metall Trans A 1981;12A:1333.
[10] McMahon Jr. CJ, Cianelli AK, Feng HC. Metall Trans A 1977;8A:1055.
[11] Guillon R, Guttmann M, Dumoulin Ph. Metal Sci 1981;15:63.
[12] Viswanathan R, Joshi A. Metall Trans A 1975;6A:2289.
[13] Baker RG, Nutting J. J Iron Steel Inst London 1959;192:257.
[14] Seah MP. Philos Trans R Soc London series A 1980:265.
[15] Briant CL, Banerji SK. Inter Met Rev 1987;23:161.
[16] Bulloch JH. Int J Press Vess Piping 1999;76:79.