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Effect of Precipitation Hardening on the Fatigue Strength*

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Effect of precipitation hardening on the rotary bending fatigue strength of an austenitic heat resisting steel was investigated at room temperature. It was concluded that strengthening processes such as precipitation hardening were effective only when fatigue cracks formed were associated with slip bands, but in the presence of structural inhomogeneity such as twin boundaries the stress concentrated readily there and the fatigue strength decreased in spite of a hardness increase. In this case the work hardened layer had an important role of making the structure homogeneous at surface and it could prevent a decrease in fatigue strength. In order to improve the fatigue strength it was most important to reduce the structural inhomogeneity as well as to harden.

1. Introduction

Although it is generally accepted that an increase in the tensile strength or the hardness of metals is accompanied with an improvement of the fatigue strength, some reports show that an increase in the tensile strength does not always result in an increase of fatigue strength⁽¹⁾. This fact suggests that in order to evaluate the fatigue strength it is important to take account of the structural changes which occur in a strengthening process employed for an improvement of the tensile strength or hardness.

In this study, the effect of precipitation hardening, which is a most effective strengthening process for engineering materials, on the fatigue strength was investigated experimentally in comparison with the effect on the tensile strength or hardness. Since alloys which form complex precipitates of multi-phase are not preferable, an austenitic heat resisting steel 20-11P was used for this study. 20-11P is one of the most popular steels for exhaust valves of automobile engines and it forms only $M_{23}C_6$ carbides which are considerably stable. This steel is also favorable for this study because the dispersion of $M_{23}C_6$ can be easily changed over a wide range by heat treatment⁽²⁾.

Different morphologies of grain boundary carbides were obtained by changing the cooling rate from solution temperature, and a wide range of dispersions of matrix

carbides was obtained by employment of various aging conditions. Thus a rotary bending fatigue test was carried out at room temperature with the specimens having a wide range of hardnesses and different morphologies of grain boundary carbides, and the effect of precipitation hardening on the fatigue strength was clarified. Some discussion was also made on the basis of the fatigue test results and the microscopic observation for an initiation of fatigue cracks.

2. Test material and experimental procedures

2.1 Test material and heat treatment conditions

The chemical composition of 20-11P steel is listed in Table 1. Notable precipitation hardening can be obtained in this steel as a result of addition of both carbon and phosphorus.

Table 1. Chemical composition of steel used (wt.%)

C	Si	Mn	P	S	Ni	Cr	Mo	B	Cu
0.31	0.37	0.96	0.20	0.016	10.51	19.55	1.98	0.005	0.07

Various heat treatment conditions shown in Fig.1 were employed for hot-rolled bars of 16mm diameter. At first, solution heating was performed at 1200°C for 1 hr to dissolve precipitation hardening components. After solution heating, the specimens were cooled by two procedures, i.e., water quenching and two step cooling which consists of furnace cooling from solution temperature of 1200°C to 900°C and subsequent air cooling, in order to change the morphology of grain boundary carbides. Subsequent aging at 600~900°C for 6~200 hours provided various hardness numbers depending on the degree of dispersion of matrix carbides.

The heat treatment conditions and Vickers hardness numbers (load 20kg) of specimens are listed in Table 2. The water quenched and the two step cooled specimens are designated as WQ group and FA group re-

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Fig.1 Heat treatment procedures

spectively.

2.2 Microscopic observation

The electron microscopic observation for carbon replicas was carried out as well as the optical microscopic one to reveal the morphology of grain boundary carbides. From Fig.2 showing the microstructures of typical specimens, it is apparent that WQ group has straight boundaries with fine carbides, whereas FA group has zig-zag ones due to the formation of coarse carbides.

Transmission electron micrographs of thin films were taken to observe the matrix precipitation. An example shown in Fig.3 reveals that uniform precipitation of fine carbides is predominant in matrix of all specimens. The carbides in both matrix and grain boundaries are already reported to be $M_{23}C_6$ type.²⁾ To observe the matrix precipitation an extraction replica method was also used.³⁾

2.3 Fatigue test

Fatigue test was run at 1400 rpm with the specimens of 8mm diameter as illustrated in Fig.4 using an ONO-type rotary bending fatigue testing machine of capacity of 10 kg-m. Specimens were cooled by a strong blower when their temperatures were raised to higher level of applied stresses during test.

Since the work hardened surface layer makes an important contribution to the rotary bending fatigue strength, the microhardness distribution in mechanically finished specimens was measured from the surface to the interior. The typical results are shown in Fig.5. The depth of work hardened surface layer was no more than about 80μm. So, some specimens were electrolytically polished to 80~100μm in

Table 2. Heat treatment conditions and hardness obtained

Marks	Heat treatment conditions		Hardness (Hv:20kg)
	Solution treatment	Aging (→AC)	
WQ group	WQ-A	—	183
	WQ-B	900°C × 50hr	253
	WQ-C	1200°C × 1hr → WQ	335
	WQ-D	850°C × 6hr	369
	WQ-E	750°C × 15hr	380
FA group	FA-A	600°C × 200hr	178
	FA-B	—	263
	FA-C	1200°C × 1hr → 900°C	310
	FA-D	FC → 900°C	332
	FA-E	→ AC	362

Fig.2 Microstructures of specimens

FA-E

Fig.3 Transmission electron micrograph of precipitates in matrix

Fig.4 Specimen used for rotary bending fatigue test

Fig.5 Microhardness distribution in specimens before test

depth to remove it. Then the specimens with or without a work hardened surface layer were used in this study.

An initiation of fatigue cracks was observed by optical microscope after interrupting the test at appropriate number of cycles and taking off the specimen from the testing machine.

3. Experimental results

3.1 Relation between density of matrix carbides and hardness

Fig.6 shows the relationship between the hardness of specimens and the density of matrix carbides obtained by the microscopic observation of extraction replicas. This also includes the results for AC group air-cooled from solution temperature. It is found that Vickers hardness increment by aging ΔH_v is increased proportionally to an increase in density of matrix carbides n . From this an experimental equation of $\Delta H_v = A n^{0.4}$ is obtained. Since the carbides disperse uniformly in matrix as is shown in Fig.3, a mean particle spacing λ equals approximately to $1/\sqrt{n}$ and the following equation is obtained.

$$\Delta H_v = A \lambda^{-0.8} \quad (1)$$

An index number of this equation is so close to minus unity that a plastic deformation seems to occur by Orowan mechanism in which dislocations pass through precipitation particles leaving loops around them⁽⁴⁾.

Fig.7 shows a linear relationship between Vickers hardness H_v and 0.2% proof stress σ_y or tensile strength σ_{UTS} in 20-11P and other similar steels. From this result it is indicated that finely dispersed matrix carbides make it difficult for the deformation due to Orowan mechanism to occur and consequently they increase the proof stress and tensile strength as well as hardness.

3.2 S-N curves

Fig.8~Fig.10 represent S-N curves of specimens heat-treated by the processes as shown in Table 2. Fatigue strength of mechanically finished specimens of WQ group increases with an increase of hardness as is shown in Fig.8. A similar trend is also recognized in Fig.9 showing the results of mechanically finished FA group specimens.

In electrolytically polished specimens, however, the trend of fatigue strength differs from that of mechanically finished ones as is shown in Fig.10. In addition to overall level down in fatigue strength due to the absence of work hardened surface layer, S-N curves of WQ-C and WQ-D come close to each other at about 2×10^6 cycles and beyond this number of cycles the fatigue strength of WQ-D is lower than that of WQ-C in spite of higher hardness in the former. Similar tendency is also recognized in electrolytically polished specimens of FA group.

3.3 Relationship between fatigue strength and hardness

From above results, the fatigue strength at 10^5 cycles is plotted as a function of hardness in Fig.11. The strength at 10^5 cycles is gradually increased with an increase of hardness regardless of the pro-

Fig.6 Relationship between the density of carbides and the hardness increment by aging

Fig.7 Relationship between hardness and 0.2% proof stress or tensile strength

Fig.8 S-N curves of mechanically finished specimens of WQ group

cesses of finishing or heat treatment of specimens. Moreover mechanically finished specimens have higher fatigue strength as compared with electrolytically polished ones by about 2~5kg/mm² over the whole range of hardnesses. It is indicated that the dis-

Fig. 9 S-N curves of mechanically finished specimens of FA group

Fig. 12 Fatigue strength at 10 cycles

Fig. 10 S-N curves of electrolytically polished specimens of WQ group

Fig. 13 Relationship between surface hardness and fatigue strength at 10 cycles

Fig. 11 Fatigue strength at 10 cycles

persion of matrix carbides has a similar effect on the fatigue strength at 10^5 cycles to the effect on the static strength such as hardness, 0.2% proof stress and tensile strength.

Fig. 12 shows the relationship between the fatigue strength at 10^7 cycles and hardness. The strength at 10^7 cycles increases linearly as a function of hardness in mechanically finished specimens, whereas in electrolytically polished ones it increases with a hardness increase up to about Hv300 keeping a lower level of about 10 kg/mm^2 as compared with mechanically finished ones; reaches a maximum value at Hv320~330; and then decreases in spite of an increase of hardness. Consequently, for example, electrolytically polished specimens have much lower strength by $16 \sim 18 \text{ kg/mm}^2$ at Hv360 as compared with mechanically finished ones.

When the dependency of rotary bending fatigue behavior on the hardness of specimen is considered, it is more reasonable to think the fatigue strength in terms of surface hardness because it is more sensitive to surface hardness than to bulk one. So, the 10^7 cycles strength is plotted as a function of microhardness at surface in Fig.13. In this figure 10^7 cycles strength of specimens falls into two groups as a function of surface hardness. The strength of mechanically finished specimens belongs to a straight line A except those of WQ-A and FA-A. This line also includes the strengths of WQ-A and FA-A of electrolytically polished specimens. The strength of electrolytically polished specimens having higher hardness level, however, belongs to a curved line B. The strengths of WQ-A and FA-A of mechanically finished specimens deviate considerably from line A and seem to belong to the curve B, but the fatigue behavior of these specimens is considered to differ from that of specimens belonging to curve B⁽⁵⁾. It is suggested from Fig.13 that the fatigue strength of specimens belonging to line A depends on the dispersion of matrix carbides in the same manner as in the case of 0.2% proof stress or tensile strength, whereas a quite another mechanism determines the strength of specimens belonging to curve B.

3.4 Microscopic observation of fatigue cracks

It is generally said that the surface cracking in fatigue can occur as a result of the development of the surface notches due to slip bands and the discontinuity of strain due to grain boundaries, twin boundaries, inclusions and other structural inhomogeneities⁽⁶⁾. Surface cracks observed in this experiment belong to above two categories.

Fig.14 illustrates an example of surface cracks observed in mechanically finished specimens. The arrows designated as σ indicate the direction of stress amplitude. In mechanically finished specimens of WQ-A and FA-A, the fatigue fracture occurs generally as the result of an extension of cracks formed by broad slip bands in grain as shown with an arrow of A. The situation of surface cracks, however, is changed with an increase in hardness of specimens or a decrease in applied stress. In the specimens having higher hardness, few slip bands are formed and the fracture occurs mainly by the extension of a few cracks normal to the applied stress which is produced by the deformation just under the surface, as shown with an arrow of B. In addition to the cracks associated with slip bands, the cracks are formed in twin boundaries or grain boundaries as shown with an arrow of C or D respectively. But the latter scarcely results in a fracture.

Fig.15 shows a typical example of surface cracks observed in electrolytically polished specimens C, D and E having higher hardness. In these specimens, no slip bands are formed and twin boundaries become broader with stain as shown with an arrow of A. A wiping of the surface reveals cracks at twin boundaries⁽⁷⁾ as shown with an arrow of B. From these cracked twin boundaries,

(a) WQ-A, $\bar{\sigma} = 37.0 \text{ kg/mm}^2$, $N = 1.0 \times 10^5$

(b) FA-D, $\bar{\sigma} = 59.8 \text{ kg/mm}^2$, $N = 1.0 \times 10^5$

(c) WQ-D, $\bar{\sigma} = 58.0 \text{ kg/mm}^2$, $N = 1.5 \times 10^5$

(d) WQ-D, $\bar{\sigma} = 58.0 \text{ kg/mm}^2$, $N = 1.5 \times 10^5$

100 μ

Fig.14 Initiation of fatigue cracks in mechanically finished specimens

cracks grow normally to the applied stress as shown with an arrow of C and they result in a fracture.

After all it is indicated in general that the surface cracks associated with slip bands are predominant in the specimens which belong to A group in Fig.13, whereas in the specimens belonging to B group in Fig.13 only twin boundary cracks are observed. Coarse grain boundary carbides can be considered the site of cracking, but cracks formed in grain boundaries are observed only in mechanically finished specimen FA-E having the highest hardness value in FA group.

(a) $N=5.0 \times 10^4$ (b) $N=5.0 \times 10^4$ (c) $N=1.5 \times 10^5$ WQ-C, $\bar{\sigma} = 54.6 \text{ kg/mm}^2$ 100μ

Fig.15 Initiation of fatigue cracks in electrolytically polished specimens

3.5 Discussion

The plastic deformation in fatigue differs from that observed during static tensile test and it is characterized by the microscopic inhomogeneity. The fatigue strength is, therefore, considered to depend strongly on the following factors, namely, the resistance to plastic deformation, the strength of weaker region where cracking occurs, and the stress concentration produced by the inhomogeneous strain during plastic deformation. Since two modes of cracking associated with both slip bands and twin boundaries are observed in this investigation, the strength against two types of cracking is discussed on the basis of above three factors.

With regard to the strength $\bar{\sigma}_s$ against the cracking due to slip bands, it can be considered proportional to the resistance

to slip deformation or hardness⁽⁸⁾. So, the following equation is obtained.

$$\bar{\sigma}_s = aH_v + b \quad (2)$$

where a and b are constants.

Since McLean's formula⁽⁴⁾ is applicable to the strength to twin boundary cracking $\bar{\sigma}_{tw}$, it is given by the following equation.

$$\bar{\sigma}_{tw} = 2\sqrt{\frac{12G\gamma}{\pi L}} \quad (3)$$

where γ and L are the surface energy and the length of a twin boundary respectively. G is shear modulus.

If $\bar{\sigma}$ is the applied stress and $\bar{\sigma}_c$ is the concentrated stress caused by the inhomogeneous plastic deformation, cracking can occur at twin boundaries when the following condition is satisfied.

$$\bar{\sigma} + \bar{\sigma}_c \geq 2\sqrt{\frac{12G\gamma}{\pi L}} \quad (4)$$

If $\bar{\sigma}_t$ is a minimum applied stress which satisfies this condition, the strength against twin boundary cracking equals to $\bar{\sigma}_t$ given by the following equation.

$$\bar{\sigma}_t = 2\sqrt{\frac{12G\gamma}{\pi L}} - \bar{\sigma}_c \quad (5)$$

The first term of right side of this equation is almost independent of the hardness. And $\bar{\sigma}_c$ will increase with an increase of hardness or inhomogeneous plastic strain. Since the surface of mechanically finished specimen is subjected to severe deformation, it is more homogeneous in structure and deforms more uniformly as compared with that of electrolytically polished one. Consequently the heterogeneous plastic strain of mechanically finished specimens will be smaller than that of electrolytically polished specimens. This leads to a smaller value of $\bar{\sigma}_c$ in the former than in the latter so far as their hardness is equal level. Denoting the mechanically finished and electrolytically polished specimens by M and E respectively, the relation of $(\bar{\sigma}_t)_M > (\bar{\sigma}_t)_E$ is obtained from equation (5) and it holds at any hardness level. Since $\bar{\sigma}_c$ is increased with an increase of hardness, $(\bar{\sigma}_t)_M$ and $(\bar{\sigma}_t)_E$ are represented as the nearly parallel two lines with negative slope as a function of hardness as is shown in Fig.16. As mentioned above, $\bar{\sigma}_s$, the strength against cracking due to slip bands, increases proportionally to hardness as is shown in Fig.16. Since the

Fig.16 Schematic representation of relationship between fatigue strength and hardness

fatigue strength is determined by a weaker one of σ_s and σ_t , the relationship between the fatigue strength and the hardness can be represented by AB, C, in electrolytically polished specimens, whereas it can be represented by AB₂C₂ in mechanically finished ones. And thus the results of Fig.13 can be successfully interpreted if it is supposed that this experiment is carried out over a hatched region of hardness in this figure. The results of Fig.11 can be also well explained by supposing that the heterogeneous plastic strain decreases and σ_c becomes smaller as the applied stress increases. That is, with an increase of the applied stress, the points B₁ and B₂ in this figure will shift upwards beyond the hatched region as a result of both $(\sigma_t)_L$ and $(\sigma_t)_H$ increasing. Consequently the fatigue strength will increase simply with a hardness increase regardless of finishing process of specimens as shown in Fig.11. Further it is easily assumed that with a far higher hardness the maximum value may appear also in the 10^7 cycles-strength of mechanically finished specimens or in the 10^5 cycles-strength of both mechanically finished and electrolytically polished specimens.

From above discussion, it is concluded that all strengthening processes for materials through prevention of the movement of dislocations including the precipitation hardening are effective only when fatigue cracks leading to a fracture occur associated with slip bands. However, in the presence of structural inhomogeneity including twin boundaries which yield a heterogeneous plastic strain, the stress concentrates readily there as the hardness increases and therefore the fatigue strength decreases in spite of hardness increase. In this case the work hardened layer has an important role of reducing the structural inhomogeneity at surface and it can prevent a decrease in fatigue strength. A most important problem in order to improve the fatigue strength is how to reduce the structural inhomogeneity as well as how to harden.

4. Summary

The effect of precipitation hardening on the rotary bending fatigue strength of an austenitic heat resisting steel was investigated at room temperature.

The results obtained are summarized as follows:

- 1) In an austenitic heat resisting steel of precipitation hardening type, the dispersion of matrix precipitates is a controlling factor of the proof stress and tensile strength as well as hardness.
- 2) The trends of fatigue strength at 10^7 cycles fall into two groups as a function of microhardness at specimen surface, that is, one is represented by a straight line which increases proportionally to hardness, whereas the other is represented by a curved line which lies under a straight line and has a peak value at some hardness number. The strength of mechanically finished specimens belongs to the former and that of electrolytically polished specimens belongs to the latter except non-aged ones.

3) From the microscopic observation, it is clear that the fracture of specimens belonging to the former occurs as a result of an extension of cracks formed in slip bands, but in the specimens belonging to the latter cracks formed at twin boundaries play an important role in fracture.

4) It is concluded that strengthening processes such as precipitation hardening are effective only when fatigue cracks are initiated by the development of slip bands, but in the presence of structural inhomogeneity such as twin boundaries the stress concentrates readily there and consequently the fatigue strength decreases in spite of an increase of hardness. In this case the work hardened layer has an important role of reducing the structural inhomogeneity at surface and it can prevent a decrease in fatigue strength.

5) Little effect of morphology of grain boundary precipitates is recognized on the fatigue strength at room temperature.

6) In order to improve the fatigue strength at room temperature, it is important to reduce the structural inhomogeneity causing heterogeneous plastic strains as well as to increase the resistance to deformation evaluated by the static strength such as hardness or tensile strength.

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