Effect of Ultrasonic Shot Peening on High Cycle Fatigue Behavior in Type 304 Stainless Steel at Elevated Temperature

by
Toshifumi KAKIUCHI*, Yoshihiko UEMATSU*, Norihiko HASEGAWA** and Eisuke KONDOH***

Ultrasonic shot peening was applied to type 304 austenitic stainless steel and four-point rotating bending fatigue tests were performed at room temperature (R.T.) and 573K up to $10^8$ cycles. The fatigue strengths at 573K were much lower than those at R.T. irrelevant to shot peening process. The shot-peened specimens exhibited higher fatigue strengths than untreated ones at both R.T. and 573K, indicating that shot peening had beneficial effect even at the elevated temperature of 573K. Clear fatigue limits were recognized in both shot-peened and untreated specimens at R.T. On the contrary, at 573K, fatigue failure occurred at high cycle fatigue (HCF) region ($N > 2 \times 10^6$ cycles) in the shot-peened specimens, while not in the untreated ones. Fatigue crack initiated at the specimen surface due to cyclic slip deformation in the untreated specimens at R.T. and 573K and in the shot-peened ones at R.T. However the shot-peened specimens showed sub-surface crack initiation with a typical fish-eye pattern in the HCF failures at 573K. Inclusions were not recognized at the center of the fish-eye. The transition of crack initiation mechanism at 573K of shot-peened specimens could be attributed to the combined effects of the surface hardening by shot peening and the softening of material at the elevated temperature.

Key words:
Stainless steel, High cycle fatigue, Elevated temperature, Shot peening, Sub-surface crack initiation

1 Introduction

Austenitic stainless steel, type 304, is widely used for the pressure piping of nuclear power plants because it has a good combination of superior creep and corrosion resistances, coupled with enhanced resistance to sensitization and associated intergranular cracking. In the pressure piping of nuclear power plants, where the service temperature is around 573K (300ºC), high cycle fatigue (HCF) failure could take place due to the high-frequency vibration of components. Kanazawa and Yoshida conducted HCF tests using type 304B and 316B stainless steels at the elevate temperatures ranging from 673K (400ºC) to 973K (700ºC)\(^1\). They revealed that HCF failure did not occur at the temperatures lower than 873K (600ºC), while HCF failure occurred at 973K where the failure stress cycle was larger than $10^7$. Recently some further researches had been performed concerning with HCF behavior of austenitic stainless steels\(^2\)-\(^7\). In the HCF behavior of ductile austenite stainless steels, surface crack initiation was dominant, while sub-surface crack initiation was generally seen in high strength steels\(^8\)-\(^9\). However, Ogawa et al. had conducted ultrasonic fatigue tests using type 316NG and revealed that surface crack initiation hardly occurred, while the highly pre-strained specimens showed sub-surface crack initiation with a typical fish-eye pattern in HCF region\(^10\)-\(^11\). Müller-Bollenhagen et al. had also revealed that sub-surface crack initiation was recognized in type 304 with very high volume fraction of strain-induced martensitic phase\(^7\). It indicates that sub-surface crack initiation could occur even in ductile austenitic stainless steels in HCF region, when the materials were highly pre-strained. But there have been very limited studies about the HCF behavior of type 304 at elevated temperature. In addition, shot peening is often applied to the surface of piping components because shot peening could introduce hardened layer, namely severely-strained layer, on the specimen surface with compressive residual stress, resulting in the enhanced fatigue properties. Thus shot peening is widely applied to austenitic stainless steels\(^10\)-\(^11\). It is practically important to clarify the effect of temperature and shot peening on the HCF behavior of type 304 at elevated temperature. Hayashi et al. had conducted fatigue tests using shot-peened type 304 in air and pure water at 561K (288ºC)\(^12\). They had concluded that the shot peening followed by surface polishing could highly improve fatigue strengths at 561K, while the fatigue limit was defined at $10^7$ cycles of fatigue loading. Consequently, the HCF behavior of shot-peened type 304 up to $10^9$ cycles at the service temperature has not been understood.

In the present study, rotating bending fatigue tests have been performed at room temperature (R.T.) and 573K up to $10^8$ cycles using ultrasonic shot-peened and untreated type 304 austenitic stainless steels, and the effect of shot peening and temperature on the HCF behavior was investigated.

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2 Experimental Procedures

2.1 Materials and Specimens

The materials used are two kinds of type 304 austenitic stainless steels having the different lot numbers, and denoted as the materials “A” and “B”. The chemical compositions (wt.%), mechanical properties and microstructures of both materials are shown in Tables 1, 2, and Fig. 1, respectively. The mechanical properties at 573K were measured using the material A only. It should be noted that the material was extensively softened at 573K. The average grain sizes are 120μm and 52μm for the materials A and B, respectively. The hourglass-shape fatigue specimens shown in Fig. 2 with a reduced section of 8mm diameter were machined and then mechanically polished by emery paper and finally buff finished before fatigue tests. The ultrasonic shot peening (USP) was applied to the gauge section after the buff finished before fatigue tests. The ultrasonic shot peening reduced section of 8mm diameter were machined and then hourglass-shape fatigue specimens shown in Fig. 2 with a respectively.

Table 1 Chemical compositions of materials (wt.%).

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>0.05</td>
<td>0.35</td>
<td>1.47</td>
<td>0.026</td>
<td>0.001</td>
<td>9.21</td>
<td>18.25</td>
<td>Bal.</td>
</tr>
<tr>
<td>B</td>
<td>0.06</td>
<td>0.26</td>
<td>1.73</td>
<td>0.034</td>
<td>0.025</td>
<td>8.01</td>
<td>18.64</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

Table 2 Mechanical properties.

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% proof stress σ0.2 (MPa)</th>
<th>Tensile Strength σB (MPa)</th>
<th>Elongation δ (%)</th>
<th>Vickers Hardness HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>A at R.T.</td>
<td>246</td>
<td>594</td>
<td>80</td>
<td>169</td>
</tr>
<tr>
<td>A at 573K</td>
<td>150</td>
<td>423</td>
<td>46</td>
<td>92</td>
</tr>
<tr>
<td>B at R.T.</td>
<td>278</td>
<td>627</td>
<td>62</td>
<td>168</td>
</tr>
</tbody>
</table>

2.2 Experimental Procedures

Rotating bending fatigue tests were performed using four-point bending type fatigue testing machine equipped with an electric furnace. The test temperatures are R.T. and 573K, and the test frequency was 60Hz. After fatigue tests, fracture surfaces were investigated by means of a scanning electron microscope (SEM) in detail. Hardness distributions were measured by a Vickers hardness tester at a load of 4.9N with holding time of 30s. To measure the hardness distributions at the elevated temperature of 573K, a small hot plate was integrated into a Vickers hardness tester. Residual stress was measured by an X-ray diffraction (XRD) method.

3 Results

3.1 Hardness and Residual Stress Distributions

Figure 3 shows Vickers hardness and residual stress distributions of the material SPA as a function of a distance from surface. The hardness was measured on the cross section of the specimen ten times at the same depth from the surface and averaged. Residual stress was measured step by step removing specimen surface. Ultrasonic shot peening produces high hardness to a depth of about 600μm, and the maximum hardness at the surface reaches 324HV. The residual compressive stress due to shot peening at the surface is about -597MPa, while a relaxation of the compressive stress occurs at 400μm from the surface. The maximum compressive residual stress (-597MPa) exceeds σ0 shown in Table 2 (594MPa). But it should be noted that the true stress-true strain curve was measured on this material, and the true stress-true strain was measured using the following equation:

\[ \sigma = \frac{F}{A_0} \]

where σ is the true stress, F is the maximum load, and A0 is the initial cross-sectional area.

Table 3 Conditions of ultrasonic shot peening process.

<table>
<thead>
<tr>
<th>Peening process</th>
<th>Shot material</th>
<th>Sonotrode amplitude</th>
<th>Sonotrode frequency</th>
<th>Specimen rotational speed</th>
<th>Peening time</th>
<th>Arc height</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ultrasonic shot peening</td>
<td>SUJ2 ø3.0mm</td>
<td>50μm</td>
<td>20kHz</td>
<td>12rpm</td>
<td>180sec</td>
<td>0.14mm C (0.35mm A)</td>
</tr>
</tbody>
</table>

Fig. 1 Microstructures: (a) Material A, (b) Material B.

Fig. 2 Configuration of fatigue specimen.

Fig. 3 Hardness and residual stress distributions as a function of distance from surface.
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strain curve was measured on this material, and the true stress at the final fracture reached 714MPa. Thus the maximum residual stress was still lower than the true tensile strength.

The fatigue tests were conducted at 573K, so that the shot-peened sample was kept at 573K for 48 hours to estimate the effect of annealing at 573K on the hardness. Vickers hardness distributions after annealing at 573K are also plotted in Fig.3. The annealing time of 48 hours is equivalent to the time of 10^6 loading cycles in the fatigue test operated at 60Hz. It is clear that the annealing of shot-peened samples at 573K does not induce any relaxation of hardness.

3.2 Fatigue Strength

The S-N diagrams at R.T. and 573K are shown in Fig. 4. The fatigue strengths of the materials A and B are nearly the same irrelevant to the test temperature and shot peening process. In the untreated specimens, the fatigue limit at 573K (solid symbol) was about 100MPa lower than that at R.T. (open symbol). HCF failure (Nf > 2x10^6 cycles) was not recognized in the untreated specimens at both temperatures. The shot-peened specimens exhibit about 80MPa and 60MPa higher fatigue limits than the untreated ones at R.T. and 573K, respectively, indicating that shot peening has beneficial effect on fatigue properties even at the elevated temperature of 573K. It implies that the compressive residual stress at the specimen surface was not fully released during fatigue loading at 573K. It should be noted that HCF failures were recognized around 10^7 cycles only in the fatigue tests conducted at 573K using the shot-peened materials, SPA and SPB.

3.3 Fractographic analysis

The typical SEM images showing crack initiation sites of the untreated specimens (material A) at R.T. and 573K are revealed in Fig. 5. Fatigue cracks initiated at the specimen surface due to cyclic slip deformation irrelevant to the test temperatures. The crack initiation sites of the shot-peened specimens (material SPA) at R.T. and 573K are shown in Fig. 6. Surface crack initiation is recognized in the shot-peened specimens similar to the untreated ones. However, sub-surface crack initiation with a typical fish-eye pattern occurred only in the fatigue tests of the shot-peened specimens at the elevated temperature of 573K. The sub-surface crack initiation, denoted by “*” in Fig. 4, corresponds to the HCF failure around 10^10 cycles, where totally five specimens showed sub-surface crack initiation in both materials SPA and SPB. The main objective of conducting fatigue tests using the materials B and SPB is to confirm whether sub-surface crack initiation could occur in the materials with different lot numbers.

Typical sub-surface crack initiation site is revealed in Fig. 7. The crack initiation site near the center of fish-eye was investigated by back-scatter electrons to identify an inclusion. Fig. 7(c) and (d) are the magnified views at the crack initiation site observed by secondary and back-scatter electrons, respectively. As shown in BEI (back-scattered electron image) of Fig. 7(d), no inclusion was recognized. Figure 8 shows the other typical fish-eye crack initiation from inclusion. The inclusion indicated in Fig. 8(b) was analyzed by an energy dispersive X-ray spectrometry (EDX) method as shown in Fig. 9, where the inclusion was identified as the intermetallic compound mainly consisting of Al and Ca. It is noteworthy that the sub-surface crack initiation from inclusion (Fig. 8) was only seen in one of five specimens. The other four specimens with sub-surface crack initiation had no inclusion at the crack initiation sites. It indicates that the
sub-surface crack initiation without inclusion was dominant in the HCF failure of the shot-peened specimens at 573K. Ogawa et al. had also revealed that sub-surface crack initiated in -20% pre-strained type 316NG stainless steel in HCF region at R.T., and inclusion was not recognized at the crack initiation site\(^4\).\(^5\).

Fig. 7 SEM micrographs showing fish-eye fracture in shot-peened material SPA at 573K (\(\sigma_a\approx250\text{MPa}, N=7.0\times10^8\)): (a) Macroscopic appearance, (b) Center of fish-eye, (c) Magnified view of crack initiation site (Secondary electron image), (d) Magnified view of crack initiation site (Back-scattered electron image).

Fig. 8 SEM micrographs showing fish-eye fracture in shot-peened material SPA at 573K (\(\sigma_a\approx240\text{MPa}, N=1.0\times10^9\)): (a) Macroscopic appearance, (b) Magnified view of crack initiation site.

Fig. 9 EDX analysis of inclusion at crack initiation site.

Fig. 10 Macroscopic appearances of fatigue fractured specimens (material SPA): (a) at R.T., (b) at 573K.

4 Discussion

4.1 Sub-surface Crack Initiation of Shot-peened Specimens at 573K

4.1.1 Effect of Thin Oxide Film at 573K The shot-peened specimens showed sub-surface crack initiation in the HCF region at 573K. The macroscopic appearances of fatigue fractured specimens (material SPA) at R.T. and 573K are revealed in Fig. 10(a) and (b), respectively. The specimen surface tested at 573K is colored bronze, indicating that thin oxide film was formed during fatigue test. There is a possibility that very thin but very hard oxide film could induce the transition of crack initiation from surface to sub-surface. In order to clarify the effect of oxide film, the specimen was statically annealed at 573K for about 48 hours, which was equivalent to the time of 10\(^7\) cycles at 60Hz. Then the fatigue tests were conducted at R.T. using the annealed specimens with oxide film. The test results are shown in Fig. 4 by semi-solid symbols. The fatigue strengths of the specimens with oxide film were similar to those without film, and sub-surface HCF failure did not occur. Therefore, it is concluded that thin oxide film could not induce sub-surface crack initiation by itself.

4.1.2 Effect of Strain-induced Martensitic Transformation Müller-Bollenhagen et al. had revealed that high volume fraction of strain-induced martensitic phase could lead to sub-surface crack initiation in type 304 stainless steel\(^1\). Thus the volume fraction of martensitic phase, \(V_m\), was measure by an XRD method\(^1\). In the untreated materials A and B, martensitic phase was not detected on the surfaces. \(V_m\) of the materials SPA and SPB are 4% and 3%, respectively, indicating that strain-induced martensitic transformation had occurred on the specimen surfaces by severe plastic deformation. The authors have reported that \(V_m\) of pre-strained type 304 could be increased by the stress cycling during fatigue test due to work hardening\(^1\). Thus the \(V_m\) was measured using the run-out specimens of the materials SPA and SPB, but \(V_m\) are the same before and after fatigue tests in both materials. Müller-Bollenhagen achieved \(V_m\) of 54% in type 304 by applying pre-stress at low temperature, and reported that very high \(V_m\) resulted in the sub-surface crack
initiation, while surface crack initiation was dominant in the specimens with $F_m$ of 27%\(\text{min}\). In this case, the materials SPA and SPB exhibit much lower $F_m$ than Müller-Bollenhagen’s samples and it could be assumed that the strain-induced martensitic phase near specimen surface was not responsible for the sub-surface crack initiation of shot-peened specimens at 573K.

4.1.3 Effect of Residual Stress  The residual stress of the material SPB was measured on the specimen surface by an XRD method. The compressive residual stress was -410MPa, and lower than that of the material SPA. Subsequently, the residual stresses of the material SPB were measured using the run-out fatigue samples at R.T. ($\sigma_a = 330$MPa, $N = 1 \times 10^8$ cycles) and at 573K ($\sigma_a = 220$MPa, $N = 1 \times 10^8$ cycles). The stresses are -281MPa and -303MPa at R.T. and 573K, respectively. It indicates that the compressive stress was partially released during fatigue loading. But the residual-stress release was nearly the same between R.T. and 573K. Consequently, it is considered that residual stress distribution and its release are not related to the sub-surface crack initiation at 573K.

4.1.4 Effect of Hardness Distribution at 573K  The hardness at 573K was measured by a Vickers hardness tester equipped with a small hotplate. The hardness distributions at R.T. and 573K are shown in Fig.11. As shown in the figure, the difference of hardness between the maximum value at the surface and the sub-surface minimum value is larger at 573K than at R.T. Furthermore, the gradient of hardness is steeper at 573K. It indicates that the softening of matrix due to the elevated temperature of 573K occurred more remarkably in the un-hardened area. Therefore, it is considered that the larger drop of the sub-surface hardness at 573K than at R.T. is mainly responsible for the sub-surface crack initiation mechanism of shot-peened type 304 at 573K. It should be noted that the hardness at the sub-surface crack initiation site reaches minimum value at 573K, while not at R.T.

4.2 Prediction of Fatigue Limit of Shot-peened Specimens  It is known that fatigue limit, $\sigma_{\text{fl}}$, could be estimated from Vickers hardness, $HV$, based on the following equation\(^{15,16}\), $\sigma_{\text{fl}}=1.6HV$. Vickers hardness of the untreated material A is 169$HV$ at R.T. and 92$HV$ at 573K. Thus the fatigue limits of specimen A could be estimated as 270MPa and 147MPa at R.T. and 573K, respectively, where the corresponding experimental fatigue limits at $10^8$ cycles are 250MPa and 140MPa. It indicates that the prediction gave slightly unconservative results, but the estimation from the hardness is reasonable because the differences between predicted and experimental fatigue limits are within 8% and 5% at R.T. and 573K, respectively. Subsequently, Fatigue limits of shot-peened specimens were estimated based on the surface residual stress using modified Goodman diagram as shown in Fig.12. The predicted fatigue limits of the material SPA by the initial residual stress of -597MPa are much higher than the experimental values at both R.T. and 573K. Subsequently, the residual stress of run-out specimen SPA at R.T. ($\sigma_a = 330$MPa, $N = 1 \times 10^8$ cycles) was measured. The stress was -277MPa, indicating that the compressive stress was partially released during fatigue loading. The fatigue limits were again estimated using the residual stress of the run-out specimen. As shown in Fig.12, the predicted value using -277MPa are close to the experimental ones. It implies that the stress relaxation occurred at the early stage of fatigue life. Consequently, it is considered that the residual stress near the surface was primary controlling factor for the fatigue limit of the shot-peened type 304. It should be noted that the
prediction of fatigue limit based on the initial residual stress could lead to the unconservative result.

5 Conclusions
Rotating bending fatigue tests have been conducted using ultrasonic shot-peened and untreated type 304 at R.T. and 573K, and the effect of shot peening and temperature on the HCF behavior was investigated. The conclusions are summarized as follows.

(1) The shot-peened specimens exhibited higher fatigue strengths than the untreated ones at both R.T. and 573K, indicating that the shot peening had beneficial effect on fatigue strengths even at the elevated temperature of 573K.

(2) Hardness was improved and compressive residual stress was introduced near specimen surface by shot peening process. The residual stress was partially released by stress cycling at fatigue limit from the initial state.

(3) Fatigue crack initiated at the specimen surface in the untreated specimens irrelevant to the test temperatures. In the shot-peened specimens, however, surface crack initiation was dominant at R.T., while sub-surface crack initiation was recognized at 573K in HCF region.

(4) Sub-surface crack initiation of the shot-peened specimens at 573K could be attributed to the steeper hardness gradient near the surface at the elevated temperature than at R.T.


References