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Published in:
Materials Science and Engineering A

Link to article, DOI:
10.1016/j.msea.2020.139651

Publication date:
2020

Document Version
Peer reviewed version

Citation (APA):
On the fatigue behavior of low-temperature gaseous carburized 316L austenitic stainless steel: Experimental analysis and predictive approach

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Abstract

Low-temperature gaseous carburization is a surface modification method for austenitic stainless steels. In order to investigate the effects of low-temperature gaseous carburization on the fatigue behavior of AISI 316L, fully reversed axial fatigue tests were performed at room temperature on specimens with various remaining case depths. The fatigue performance of AISI 316L could be significantly improved; a 15\% higher endurance limit is achieved after low-temperature gaseous carburization. After removal of the outer, brittle part of carburized case by electropolishing, the improvement of the fatigue performance is reduced.

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Fractography showed that for untreated specimens, fatigue cracks always initiated on the surface. For the carburized specimens, however, the locations of crack initiation sites depend on the applied stress levels. Compressive residual stresses in the case move the crack initiation site to the sub-surface; the lower the applied stress, the deeper the initiation site. A quantitative analysis of the effect on fatigue behavior forms the basis for a life prediction model, which can accurately predict the fatigue life of AISI316L steel after low temperature carburization.

**Keywords:** Low-temperature gaseous carburization; Compressive residual stress; Brittleness; Fatigue behavior; Life prediction model; AISI 316L austenitic stainless steel
1. Introduction

In metallic materials, fatigue failure under cyclic loading often initiates at the surface. Surface strengthening and compressive residual stresses can obviously improve the fatigue performance. Mechanical surface modification technologies [1-6], such as shot peening, mechanical attrition, rolling and extrusion, as well as thermochemical surface treatments [7-10], such as carburizing, nitriding and nitro-carburizing, are effective approaches to achieve better fatigue performance of metallic materials and have been applied for decades.

Austenitic stainless steels are widely used because of good structural integrity and machinability in combination with excellent corrosion resistance. Unfortunately, the low strength and surface hardness allow the initiation and easy growth of fatigue cracks, implying relatively poor fatigue performance. Low-temperature surface carburization (LTSC), firstly introduced in the mid eighties of the previous century and for a certain process variant commercialized under the tradename Kolsterizing®, has received more and more attention in the past decade as a means to strengthen the surface-adjacent region of AUSSs. During LTSC a zone of expanded austenite develops, which essentially is a supersaturated solid solution of carbon in austenite. Expanded austenite formation is associated with an important increase in hardness (and thus, the yield strength) and wear resistance, without negatively affecting the corrosion resistance. Moreover, the lattice expansion of austenite associated with supersaturated solid solution of carbon is accommodated by compressive residual stresses as well as plastic deformation in the expanded austenite zone [11-17].
Several studies have demonstrated that LTSC can improve the fatigue performance [7, 18-20]. Agarwal et al. [7] found that the tension-compression fatigue performance of 316L AUSS is greatly improved by carburization and showed that the crack nucleation site changes from the surface for non-treated samples, to the interior for carburized specimens. Tokaji et al. [18] studied the fatigue performance and fracture mechanism of carburized 316 AUSS by rotary bending fatigue tests. Compared to the untreated material, the endurance limit increases by 30%. Fatigue cracks in surface-treated specimens often nucleate near the transition from the case to the untreated substrate. Most of these studies on low-temperature carburization concluded that the improvement in fatigue properties is related to high compressive residual stresses generated during treatment. Actually, the outer part of the carburized case experiences plastic deformation (and work hardening) as a result of plastic accommodation of the lattice expansion, along with enhanced brittleness [20-22]. Peng et al. [20] showed that the outer 10 μm of the carburized case is brittle. Hitherto, only little attention has been devoted to the influence of this brittle part of the expanded austenite case on the fatigue performance. In addition to the experimental investigations, the fatigue response prediction is also important. Up to now, no attempts to predict the fatigue response of the AUSSs subjected to low-temperature carburization have been published.

In the present work, low-temperature gaseous carburization (LTGC) treatment was used to produce the carburized case. Part of the expanded austenite case was removed by electropolishing to investigate the influence of the brittle part of the case
on the fatigue performance. A quantitative analysis of the compressive residual stress effect on fatigue behavior was performed to establish a model for fatigue life prediction.

2. Experimental methods

2.1 Materials and LTGC treatment

An AISI 316L plate with 14 mm thickness is used in the present research, which is in solution-annealed state, and the chemical composition (in wt.%): C 0.022, Cr 17.36, Ni 10.25, Mo 2.06, Mn 1.13, Si 0.489 and Fe balance. The specimen dimensions for uniaxial tensile tests and fully reversed axial fatigue tests are given in Figs. 1 (a) and (b), respectively. The surface of specimen was ground with metallographic abrasive paper to grid 1200 and cleaned ultrasonically in ethanol in order to degrease the surface before the LTGC treatment.

![Fig. 1. Dimensions of the specimens (unit: mm): (a) uniaxial tensile specimen, (b) fatigue specimen.](image)

The LTGC treatment was performed in two steps in the same furnace: surface activation and gaseous carburization. For surface activation, hydrogen chloride (HCl)
was used to remove the chromium oxide passive film from the specimen surface to enable carbon atoms to diffuse into the steel. Gaseous carburization was carried out at 470 °C in a gas mixture of H₂, CO and N₂ for 30 hours. The detailed description of LTGC process applied in this paper could be found in the previous work [13].

2.2 Removal of the carburized case by electropolishing

In order to accurately and uniformly remove a specific depth range from the carburized case, electropolishing was used. The carburized specimens were electrolytically polished in a solution of (in vol.%) 14% H₃PO₄, 18% H₂SO₄, 15% H₂O and 53% C₃H₆(OH)₃ [14] at 45 °C for 38 min and 75 min, corresponding to a removed thickness of 6 and 11 µm, respectively. The specimens were labelled after their treatments as follows: carburization treatment (CT), carburization and removal of the outer 6 µm from the carburized case (CR₆), carburization and removal of the outer 11 µm from the carburized case (CR₁₁) and untreated specimen (UT).

2.3 Mechanical tests

A servo-hydraulic testing machine (SHIMADZU EHF-EV101K2-020-1A,) was used to perform the uniaxial tensile tests at room temperature. The strain rate was controlled at $6.7 \times 10^{-4}$ s⁻¹ by a contacting extensometer for the first 30% engineering strain during the test. Then, the extensometer was removed and the specimen was tested at a rate of 1.5 mm/min. The fully reversed axial fatigue tests were also operated on this machine at room temperature under stress-control mode with stress ratio $R = -1$ ($R = \sigma_{\text{max}}/\sigma_{\text{min}}$) and a frequency of 15 Hz. The stress levels applied in the fatigue test range from 230 MPa to 300 MPa with intervals of 5 MPa and each
stress level was tested at least twice.

2.4 Materials characterization

An optical microscopy (AXIO Imager. Alm) was used to study the cross-sectional microstructures of the untreated and carburized specimens after grinding, polishing and electrolytic etching in Marble’s reagent. The carbon concentration-depth profiles were measured by an electron probe micro-analyzer (SHIMAZU, EPMA-1610) [13]. Nano-indentation testing (Hysitron TI Premier) was operated to obtain the profiles of nano-hardness along the depth of the carburized case. A Berkovich diamond indenter was loaded to 12 mN and then held for 2 seconds before unloading; the loading/unloading times were 5 s for each. The residual stress analyzer (PROTO-iXRD) was operated to measure the residual stresses, using the sin^2\psi method with Mn Kα X-radiation (\(\lambda = 0.210314\) nm). Thin layers of the carburized case were successively removed by the electropolishing method (see above) to obtain the depth profiling of residual stress. The detailed method for measuring the residual stress in carburized case was described in the previous work [23]. Fractography of the tested specimens was performed with the scanning electron microscopy (SEM-Jeol JSM-6360).

3. Results

3.1 Materials characterization

Fig. 2 shows the cross-sectional micrographs of different specimens. After carburizing, etching reveals the microstructure in the substrate, while the carburized case appears as a uniform, featureless bright white zone, typical for expanded
austenite [24-26]. The thickness of the as-carburized case is 30 ± 1 μm. After electropolishing, the carburized case remains uniform, and the morphology is unaltered. The thickness of the carburized case is 24 ± 1 μm and 19 ± 1 μm for CR6 and CR11 specimens, respectively.

Fig. 2. Cross-sectional optical micrographs of the specimens in different conditions:

(a) UT, (b) CT, (c) CR6, (d) CR11.

Fig. 3 shows the carbon concentration-depth profiles over the carburized case of the specimens in different conditions. For the CT specimen, the carbon concentration reaches approx. 2.71 wt.% at the surface and decreases monotonically along depth. The depth range of enhanced carbon content is approx. 30 μm, which is consistent with the case thickness established by the microstructural characterization as shown in Fig. 2. After polishing, the surface carbon content is reduced to 2.27 wt.% and 1.83 wt.% for the CR6 and CR11 specimens, respectively. As expected, no changes
occurred in the remaining concentration-depth profiles.

Fig. 3. Carbon concentration-depth profiles of the specimens in different conditions.

Fig. 4. Hardness-depth profiles (top right) and residual stress-depth profiles (bottom) of the specimens in different conditions.
Fig. 4 shows the hardness-depth profiles (top) and the residual stress-depth profiles (bottom) over the carburized case of the specimens in different conditions. For the CT specimen, a maximum hardness value (12.2 GPa) and a maximum compressive residual stress value (2.5 GPa) are attained on the surface. Similar to the carbon concentration-depth profiles, the values of surface hardness and compressive residual stress are reduced after polishing, but no changes have occurred in the remaining parts.

3.2 Tensile results

The uniaxial tensile testing curves of the specimens in different states are given in Fig. 5 and the tensile properties for different specimens are shown in Table 1. For the UT condition, the yield strength (0.2% offset) is 283 MPa and the ultimate tensile strength is 638 MPa. Correspondingly, yield strength and ultimate tensile strength for the CT specimens are 265 MPa and 656 MPa, respectively. Furthermore, after treated by LTGC, the elongation has reduced from 82% to 63%, which indicates that the carburized case reduces the overall ductility of the specimen. For the CR6 and CR11 specimens, the mechanical properties are in-between those of the UT and CT specimens. After removing 6 μm and 11 μm from the carburized case surface, the yield/ultimate tensile strength are 274 / 650 MPa, 278 / 645 MPa, respectively. Also, the elongation of the carburized specimens after electropolishing removal is in-between the extremes for UT and CT: 77% for the CR11 specimen and 70% for the CR6 specimen.

Table 1. The tensile properties for different specimens
<table>
<thead>
<tr>
<th>Specimens</th>
<th>Yield strength $\sigma_y$ / MPa</th>
<th>Ultimate tensile strength $\sigma_b$ / MPa</th>
<th>Elongation / %</th>
</tr>
</thead>
<tbody>
<tr>
<td>UT</td>
<td>283</td>
<td>638</td>
<td>82</td>
</tr>
<tr>
<td>CT</td>
<td>265</td>
<td>656</td>
<td>63</td>
</tr>
<tr>
<td>CR6</td>
<td>274</td>
<td>650</td>
<td>70</td>
</tr>
<tr>
<td>CR11</td>
<td>278</td>
<td>645</td>
<td>77</td>
</tr>
</tbody>
</table>

Fig. 5. Engineering stress - engineering strain curves of the specimens in different conditions.

3.3 Fatigue testing results

The $S$-$N$ curves of the various specimens are shown in Fig. 6. After LTGC, the endurance limit of AISI 316L has increased from 235 MPa to 270 MPa, i.e. an improvement of about 15%. The improvement decreases after removal of part of the carburized case: the endurance limit drops to 260 MPa and 250 MPa after polishing off 6 µm and 11 µm, respectively. The ratios of the endurance limit to the ultimate tensile strength ($\sigma_{-1}/\sigma_b$) is given in Table 2. On the one hand, the endurance limit is very sensitive to the surface conditions [27]. As shown in Table 2, the carburized
Specimens have higher endurance limits. On the other hand, the carburized case accounts for a very small proportion of the cross sectional area; a case depth of 30 μm for the CT specimen compared with a sample radius of 2500 μm of the radius corresponding to 2.4% of the cross sectional area. Consequently, the improvement in ultimate tensile strength is limited, actually less than 3%. As a result, the ratio of the endurance limit and UTS scales with the endurance limit.

Table 2. The endurance limit (σₐ⁻) and the endurance limit to ultimate tensile strength ratio (σₐ⁻/σₜ) for different specimens.

<table>
<thead>
<tr>
<th>Specimens</th>
<th>σₐ⁻ / MPa</th>
<th>σₐ⁻/σₜ</th>
</tr>
</thead>
<tbody>
<tr>
<td>UT</td>
<td>235</td>
<td>0.37</td>
</tr>
<tr>
<td>CT</td>
<td>270</td>
<td>0.41</td>
</tr>
<tr>
<td>CR6</td>
<td>260</td>
<td>0.40</td>
</tr>
<tr>
<td>CR11</td>
<td>250</td>
<td>0.39</td>
</tr>
</tbody>
</table>

Fig. 6. S-N curves of the specimens in different conditions.
The SEM fractographs of UT specimens after fatigue testing are shown in Fig. 7. For UT specimens, fatigue cracks always initiate from the surface, consistent with previously published research results [7, 18-20], independent of the levels of the applied stresses. However, the number of initiation sites varies with the applied stress levels. For relatively high applied stresses, several initiation areas contain multiple initiation sites at the surface (see macro fracture in Fig. 7 a). For relatively low applied stresses, the number of initiation sites is reduced to one or two.

Fig. 7. SEM fractographs of the UT specimens: (a) $\sigma_a = 285$ MPa, (b) $\sigma_a = 260$ MPa.

The SEM fractographs of CT specimens are shown in Fig. 8. For the CT specimens, the initiation sites for the fatigue cracks depend on the applied stress levels. When the applied stress is relatively high, several locations of fatigue initiations can be identified at the surface. Furthermore, branch cracks appear near the fatigue crack initiation sites of the CT specimens (see Fig. 8b). For lower applied stress (equal to or below 290 MPa), the fatigue cracks initiate below the surface and fewer initiation sites are found. Moreover, the distance from the surface to the fatigue initiation site is related to the applied stress level. Sub-surface crack initiation was also observed in AUSSs after mechanical surface modification technologies [4, 28]. Furthermore, within a certain depth below the surface, the carburized case displays cleavage
fracture with brittle facets (see Figs. 8 d and 8 f).

Fig. 8. SEM fractographs of the CT specimens: (a) low-magnification, $\sigma_a = 300$ MPa, (b) high-magnification, $\sigma_a = 300$ MPa, (c) low-magnification, $\sigma_a = 285$ MPa, (d) high-magnification, $\sigma_a = 285$ MPa, (e) low-magnification, $\sigma_a = 275$ MPa, (f) high-magnification, $\sigma_a = 275$ MPa.
Fig. 9. SEM fractographs of the CR6 specimens: (a) low-magnification, $\sigma_a = 300$ MPa, (b) high-magnification, $\sigma_a = 300$ MPa, (c) low-magnification, $\sigma_a = 280$ MPa, (d) high-magnification, $\sigma_a = 280$MPa; and the CR11 specimens: (e) low-magnification, $\sigma_a = 290$ MPa, (f) high-magnification, $\sigma_a = 290$ MPa, (g) low-magnification, $\sigma_a = 275$
MPa, (h) high-magnification, $\sigma_a = 275$ MPa.

The SEM fractographs of CR6 and CR11 specimens are shown in Fig. 9. The depth range with obvious brittle appearance is reduced after removal of part of the carburized case. At high-level stresses, for example, the stresses beyond 285 MPa for CR6 specimens and 280 MPa for CR11 specimens, the fatigue initiation sites are still located at the surface of the specimens. For lower applied stresses, the initiation sites are located sub-surface.

4. Discussion

4.1 Fatigue fracture process and mechanisms

Fig. 10 gives the depth of the fatigue crack initiation sites below the surface for the specimens tested under different conditions. Obviously, the locations of the crack initiation sites depend on the applied stress levels and the (remaining) thickness of the carburized case. For the UT specimens, the crack initiation sites are always at the surface of the specimens. For the carburized specimens (including CT, CR6 and CR11) under relatively high stresses, the crack initiation sites are located at the surface. Evidently, surface crack initiation changes to the sub-surface crack initiation when the applied stress is reduced. For a certain applied stress and sub-surface crack initiation, the crack initiation sites are deeper below the surface for a thicker carburized case, and better fatigue life. Sub-surface crack initiation sites shift deeper into the case for reduced applied stresses and eventually locate near the case-core transition, before a further applied stress reduction reaches the endurance limit.
Fig. 10. Depth from the fatigue crack initiation sites to the surface for different specimens.

Based on the experimental observations, the fatigue damage mechanisms corresponding to the different specimens are schematically illustrated in Fig. 11. For the UT specimens, the stress concentration is highest at the surface so that the fatigue cracks always initiate on the surface, regardless of the applied stress level (see Fig. 11a). For the carburized specimens, the location of the initiation site depends on the stress level, as well as the compressive stress-depth distribution and the ductility of carburized case. The residual stress distribution (cf. Fig. 4) has the steepest gradient in the middle of the expanded austenite case. Excessive residual compressive stress gradients can cause severe stress concentrations and strain inconsistencies and, consequently, are more prone to defects during cyclic loading. Therefore, the distribution of local fatigue strength in the carburized case is not necessarily analogous to the distribution of compressive residual stress. If the applied stress exceeds the local fatigue strength at the surface, a fatigue crack will initiate at the surface of the specimen. Jiang et al. [21] found that the surface of the carburized case
would crack at a strain of 1.4%, which indicates that the surface of the carburized case is indeed brittle. When the applied stress is reduced to the local fatigue strength in the brittle case, the crack initiates there, but as a consequence of the brittleness, the crack will immediately extend to the surface, from where it will continue to develop as the source of fatigue (as shown in Fig. 11b). Tokaji et al. [18] also confirmed that microcracks immediately reach the surface because the surface region of the carburized case was very brittle. If the applied stress is further reduced, crack initiation moves to the sub-surface. Along with a reduction of the magnitude of the compressive residual stress with depth, the ductility improves. In addition, near the case-core transition, in the tensile part of the loading cycle the applied stress exceeds the compressive residual stress and stress reversal occurs, causing strain inflection (see Fig. 11c). Under cyclic loading, defects can initiate at this location of alternating stress reversal on either side of the case-core transition (see Fig. 11d). Furthermore, because of the overall state of compressive residual stress in the case, there is a compensating overall state of tensile stress in the core. Consequently, removal of part of the case reduces the magnitude of the overall tensile stress in the core. When the applied stress level is near the endurance limit, fatigue is initiated in the core. Similarly and consistently, lower applied stress or longer fatigue life, the initiation site shifts to larger depth.

It may be noted that Agarwal et al. [7] and Peng et al. [20] reported the origin of fatigue cracks at internal inclusions in the core of a 316L austentic stainless steel. No such initiation was observed in the present study, which can be attributed to a cleaner
steel quality with a low density of inclusions.

It is well-accepted that compressive residual stress in the surface region is the main reason for improving fatigue performance [7-9, 18]. It seems that the very high compressive residual stress in the carburized case, especially in the near-surface region, has indeed an important influence on the fatigue life, despite its brittleness.

Fig. 11. Schematic diagrams of fatigue fracture processes: (a) the fatigue cracks always initiate on the surface regardless of the applied stress level for untreated specimen; (b) crack initiates at the surface when the high applied stress exceeds the local fatigue strength of the carburized surface; (c) crack initiation moves to the
sub-surface when the applied stress is further reduced because of the strain inflection;

(d) crack initiates at the case-core transtion region when the stress is near the

endurance limit.

4.2 Prediction of fatigue response

The relationship between the fatigue life, $N_f$ (i.e. the number of cycles to failure) and the stress range $\Delta \sigma (\Delta \sigma = \sigma_{\text{max}} - \sigma_{\text{min}})$ can be described by Basquin’s equation [29]:

$$\sigma_a = \frac{\Delta \sigma}{2} = \sigma_f' \left(2N_f\right)^b \quad (1)$$

where $\sigma_a$ is the stress amplitude, $\sigma_f'$ is the fatigue strength coefficient and $b$ is known as Basquin’s exponent or fatigue strength exponent. The fatigue strength coefficient, $\sigma_f'$, scales with the tensile strength of the material while the Basquin’s exponent $b$ represents the increase in fatigue life with lower applied stress. The lower the absolute value of $b$, the greater is the improvement in fatigue life for a stress reduction [29]. Eq. (1) was fitted to the fatigue test data (cf. Fig. 6) and resulted in the values for $\sigma_f'$ and $b$ listed in Table 3.

Table 3. Parameters fatigue strength coefficient, $\sigma_f'$, and Basquin’s exponent, $b$, evaluated from fitting Basquin’s equation to experimental data in Fig. 6.

<table>
<thead>
<tr>
<th>Specimens</th>
<th>$\sigma_f'$/MPa</th>
<th>$b$</th>
</tr>
</thead>
<tbody>
<tr>
<td>CT</td>
<td>368</td>
<td>-0.022</td>
</tr>
<tr>
<td>CR6</td>
<td>395</td>
<td>-0.032</td>
</tr>
<tr>
<td>CR11</td>
<td>425</td>
<td>-0.040</td>
</tr>
<tr>
<td>UT</td>
<td>469</td>
<td>-0.055</td>
</tr>
</tbody>
</table>

Below the relation between fatigue life and compressive residual stress after
LTGC is quantified. Since both the stress gradient and the case depth influence the fatigue performance, quantitative analysis was performed using the (remaining) accumulated (i.e. depth-integrated) compressive residual stress (cf. Farrahi et al. [30]). To this end, the compressive residual stress-depth profile of the carburized case was described with an arbitrary fit function. The dependence of the parameters $\sigma_f'$ and $b$ (in Eq. (1)) on the depth-integrated compressive residual stress distribution is shown in Fig. 12; both $\sigma_f'$ and $b$ depend linearly on the accumulated residual stress in the case. Accordingly, Basquin’s equation can be rewritten as:

$$\sigma_a = (\alpha_1 A + \alpha_2)(2N_f)^{\left(\alpha_3 A + \alpha_4\right)}$$  \hspace{1cm} (2)

$$A = \int_0^Z \sigma_R(z) \, dz$$ \hspace{1cm} (3)

where $A$ is the depth-integrated compressive residual stress, $\sigma_R(z)$ is the compressive residual stress distribution along depth, $Z$ is the case depth. Coefficients $\alpha_1$, $\alpha_2$, $\alpha_3$ and $\alpha_4$ represent the linear dependence observed in Fig. 12.

![Fig. 12. Dependence of the fatigue strength coefficient, $\sigma_f'$, Basquin’s exponent, $b$, on the depth-integrated residual stress, $A$.](image_url)
Obviously, also the endurance limit depends linearly on the depth-integrated compressive stress and the relationship between the (logarithmic) fatigue life and depth-integrated compressive stress in the case is given in Fig. 13 and shows that the (logarithmic) fatigue life increases linearly with the depth-integrated compressive stress. Accordingly,

\[ \sigma_{-1c} = \beta_1 A + \sigma_{-1u} \]  

\[ \lg(N_{fc}) = \beta_2 A + \lg(N_{fu}) \]

where \( \sigma_{-1c} \) and \( N_{fc} \) are the endurance limit and fatigue life of the carburized specimen, respectively, and \( \sigma_{-1u} \) and \( N_{fu} \) are the endurance limit and fatigue life of the untreated specimen, respectively. \( \lg \) is the logarithm to the base 10. The coefficients \( \beta_1 \) and \( \beta_2 \) are coefficients representing the slopes in Fig. 13.

![Fig. 13](image)

**Fig. 13.** Dependence of the fatigue life, \( N_f \), endurance limit, \( \sigma_{-1} \), on the depth-integrated residual stress, \( A \).

The values of parameter \( \beta_2 \) for different applied stress values are collected in Fig. 14. Also, \( \beta_2 \) depends linearly on the applied stress amplitude.
\[ \beta_2 = \beta_3 \sigma_a + \beta_4 \]  \hspace{1cm} (6)

with \( \beta_3 \) and \( \beta_4 \) are constants. Then, Eq. (5) can be written as:

\[ \lg(N_{fc}) = (\beta_3 \sigma_a + \beta_4)A + \lg(N_{fu}) \]  \hspace{1cm} (7)

Therefore, taking into account the effects of compressive residual stress, Basquin’s equation can be rewritten as Eq. (2) and Eq. (7), consistent with the S-N curve of AISI 316L as shown in Fig. 6. The set of equations formed by Eq. (2) to Eq. (7), describes the effect of (depth-integrated) compressive residual stress on fatigue performance quantitatively in terms of Basquin’s equation, as well as the Wöhler relationship. In addition, a quantitative prediction of fatigue life is obtained with \( \beta_3 = -8.68 \times 10^{-7} \) and \( \beta_4 = 2.67 \times 10^{-4} \).

![Graph](image)

**Fig. 14.** Relationship between parameter \( \beta_2 \) and stress amplitude.

In Fig. 15, the predicted fatigue lives of the CT, CR6 and CR11 specimens are compared with the experimental results, that were used for the assessment of the quantitative parameters. Evidently the predicted data fall within a narrow band for all investigated specimens and there is no significant difference in the accuracy of fatigue life predictions for the three specimens. The experimental data of Agarwal et al. [7]
also are in very good correspondence with the fatigue lives predicted by this model, thereby validating the model for AISI 316L after low temperature carburization.

Fig. 15. Comparison of the fatigue lives between the predicted results and experimental results including the data of Agarwal et al [7].

5. Conclusions

In this work, fully reversed axial fatigue tests were performed with four types of specimens, which were prepared by a combination of low temperature gaseous carburizing (LTGC) and subsequent electropolishing. The effect of the LTGC on the fatigue behavior of 316L austentic stainless steel was investigated experimentally and quantified to arrive at a prediction of the fatigue performance in relation to the accumulated compressive residual stress in the obtained case. The main conclusions are:

(1) The fatigue performance of 316L austentic stainless steel can be significantly improved after LTGC. The increase in endurance limit is 15% (from 235 MPa to 270 MPa). After removing 6 μm and 11 μm from the (brittle part of the) case, the
improvement in endurance limit is 11.1% and 10.6%, respectively.

(2) For the untreated specimens, fatigue cracks always initiate on the surface of the specimen. For the carburized specimens (including full case depth and after removal of 6 and 11 µm), the location of the fatigue initiation site depends on the applied stress levels. The fatigue crack initiation mechanism changes from a surface crack for relatively high-level stress to a sub-surface crack for relatively low-level stress.

(3) The total compressive residual stress in the carburized case plays a major role in improving the fatigue performance after LTGC. Compressive residual stress in the carburized case appears to have a greater impact on fatigue life than limited ductility. This applies especially for the surface-adjacent part of carburized case, which has high compressive residual stress (up to -2.5 GPa) and is brittle.

(4) The data can be described quantitatively by Basquin’s equation, considering the accumulated (depth-integrated) compressive residual stress in the case. This quantitative description is consistent with independent literature data and accurately predicts the fatigue life of 316L austenitic stainless steel after LTGC.

Acknowledgements

This work is supported by National Natural Science Foundation of China (No.51475224 and No.51975272) and Innovation Project for College Graduates of Jiangsu Province (No. KYCX19_0849).

Data availability
The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

Reference


