The Effects of Heat Treatment on Fracture Toughness and Fatigue Crack Growth Rates in 440C and BG42 Steels

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The fatigue crack growth rates, \( da/dN \), and the fracture toughness, \( K_{Ic} \), have been measured in two high-carbon martensitic stainless steels, 440C and BG42. Variations in the retained austenite contents were achieved by using combinations of austenitizing temperatures, refrigeration cycles, and tempering temperatures. In nonrefrigerated 440C tempered at 150 °C, about 10 vol pct retained austenite was transformed to martensite at the fracture surfaces during \( K_{Ic} \) testing, and this strain-induced transformation contributed significantly to the fracture toughness. The strain-induced transformation was progressively less as the tempering temperature was raised to 500 °C, and at the secondary hardening peak, 500 °C, strain-induced transformation was not observed. In nonrefrigerated 440C austenitized at 1065 °C, \( K_{Ic} \) had a peak value of 30 MPa \( \cdot \) m\(^{1/2} \) on tempering at 150 °C and a minimum of 18 MPa \( \cdot \) m\(^{1/2} \) on tempering at 500 °C. Refrigerated 440C retained about 5 pct austenite, and did not exhibit strain-induced transformation at the fracture surfaces for any tempering temperature. The \( K_{Ic} \) values for corresponding tempering temperatures up to the secondary peak in refrigerated steels were consistently lower than in nonrefrigerated steels. All of the BG42 specimens were refrigerated and double or quadruple tempered in the secondary hardening region; the \( K_{Ic} \) values were 16 to 18 MPa \( \cdot \) m\(^{1/2} \) at the secondary peak. Tempered martensite embrittlement (TME) was observed in both refrigerated and nonrefrigerated 440C, and it was shown that austenite transformation does not play a role in the TME mechanism in this steel. Fatigue crack propagation rates in 440C in the power law regime were the same for refrigerated and nonrefrigerated steels and were relatively insensitive to tempering temperatures up to 500 °C. Above the secondary peak, however, the fatigue crack growth rates exhibited consistently lower values, and this was a consequence of the tempering of the martensite and the lower hardness. Nonrefrigerated steels showed slightly higher threshold values, \( \Delta K_{th} \), and this was ascribed to the development of compressive residual stresses and increased surface roughening in steels which exhibit a strain-induced martensitic transformation.

I. INTRODUCTION

HIGH-carbon high-chromium martensitic stainless steels, such as 440C and BG42, are used in rolling element bearings where corrosion resistance is required, but problems of fatigue, fracture toughness, and dimensional stability also arise, and these factors influence the choice of heat treatments. These steels develop high strengths and exhibit secondary hardening on tempering, with a peak at about 480 °C (900 °F) for 440C and 525 °C (975 °F) for BG42. In 440C, hardness levels of about 60 HRC can be achieved on tempering at 150 °C (300 °F), even though the retained austenite content may be at about 20 vol pct. On tempering at the secondary hardening peak, the retained austenite is reduced to about 6 pct and the hardness is about 58 HRC. These properties are shown in Figures 1 through 3. Refrigeration treatments are used in order to reduce the retained austenite contents and thus to increase the hardness (Figure 1). Multiple tempering cycles are also used, particularly near the secondary hardening peak, in an effort to reduce the austenite contents and thus to improve dimensional stability. By varying the heat treating variables, such as austenitizing temperature, tempering temperature, and refrigeration treatments, different combinations of hardness and retained austenite can be produced. The influence of this interplay of retained austenite and tempered martensite on fatigue crack propagation and fracture toughness in these steels had not been established, and this was the objective of the work reported here.

It has been shown that the presence of retained austenite can improve the fracture toughness, \( K_{Ic} \), of high strength steels, and two mechanisms have been proposed to explain this improvement. Webster pointed out that the advancing crack may stop, branch, and grow around an area of stable retained austenite, thereby resulting in increased energy absorption. When retained austenite is mechanically

Fig. 1 — Effect of austenitizing temperature and tempering temperature on hardness of BG42 and 440C steels.
unstable, however, strain-induced martensitic transformation may occur in the plastic zone ahead of the advancing crack,\textsuperscript{7,8} and this also requires additional energy absorption. The beneficial effects of retained austenite and its martensitic transformation on fatigue crack growth at near-threshold levels\textsuperscript{7} and within the power law regime\textsuperscript{4} have also been reported.

The thermal decomposition of retained austenite on tempering results in the precipitation of cementite films along the martensite boundary and a consequent loss in fracture toughness. Thomas\textsuperscript{9,10} has associated the thermal decomposition of interlath retained austenite with tempered martensite embrittlement (TME). Bhadeshia and Edmonds\textsuperscript{11} reported that when 5 pct retained austenite existed, the coarsening of cementite produced by the thermal decomposition of interlath retained austenite was the major factor controlling TME in Fe-C-V steel. Horn and Ritchie,\textsuperscript{12} however, postulated an embrittlement effect of the freshly formed martensite layer induced by mechanically transformed austenite, and proposed that TME was caused by the combined effect of thermal decomposition and the mechanical instability of retained austenite. We observed TME in these steels and attempted to assess the role of strain-induced transformation by X-ray diffraction measurements of retained austenite at the fracture surfaces.

In this study, we have used vacuum induction melted-vacuum arc remelted (VIM-VAR) steels, since these materials are used in critical bearings where fatigue and fracture are potential problems. A wide range of heat treatments was used for 440C, since this material is used after tempering at low temperatures, in the vicinity of 150 °C, and after secondary hardening, at 500 °C. A more restricted set of tempering temperatures, in the vicinity of the secondary hardening peak, was used for BG42 since this steel is customarily used with these heat treatments for bearing applications where both high hardness and dimensional stability are required. Refrigeration, along with double and quadruple tempering cycles, were used for BG42, reflecting the heat treatments generally used in practice.

### II. EXPERIMENTAL PROCEDURE

The compositions of the 440C and BG42 steels are shown in Table I. Both steels were vacuum-induction melted and consumable electrode remelted (VIM-VAR) and were processed as 70 mm hot-rolled bars in the fully-annealed condition.

Test specimens of 440C steel were austenitized at 1035, 1065, and 1095 °C (1900, 1950, and 2000 °F) for one hour in an endothermic atmosphere. After quenching in agitated oil, some of the specimens were subsequently tempered for two hours in the temperature range 150 to 315 °C (300 to 600 °F) or double-tempered for two hours each at temperatures in the range 400 to 540 °C (750 to 1000 °F). Specimens of BG42 were austenitized at 1095 and 1120 °C (2000 and 2050 °F) in a salt bath for 30 minutes and oil-quenched to room temperature. All of the BG42 specimens were refrigerated at -85 °C (-120 °F) for 15 minutes, and then double or quadruple-tempered for two hours at temperatures between 495 and 550 °C (925 and 1025 °F).

Fatigue crack growth rates were measured in air of 40 to 50 pct relative humidity, using 51 mm (2 inch) wide by 6.4 mm (0.25 inch) thick compact tension specimens.\textsuperscript{13} Tests were conducted in a servohydraulic machine, operated in the load feedback mode, and programmed to provide sinusoidal loading at a frequency of 50 Hz and a ratio of minimum to maximum load (R) of 0.1.

Crack lengths were determined to an accuracy of 0.02 mm by the use of a traveling microscope, and this permitted measurements of crack growth rates as low as $10^{-8}$ mm per cycle. Fatigue cracks were initiated at a relatively high alternating stress intensity, in the range of 12 to 18 MPa m$^{-1}$.

The following procedure was used in determinations of the threshold stress intensity, $K_{th}$ was defined as the highest value of alternating stress intensity at which crack growth could not be detected after $10^{6}$ cycles. Near-threshold growth rates were approached by decreasing the alternating stress intensity, $\Delta K$, in stages until the crack growth rate fell to $10^{-8}$ mm per cycle. The decrease of the load at each step was less than 10 pct of the previously applied stress intensity. The crack increments, over which growth rates were determined at each load level, were measured at some distance (about 0.1 mm) beyond the previous position in order to avoid the plastic zone created by the application of a higher stress intensity. Following the threshold stress intensity measurement, the stress intensity was increased in steps and a similar procedure was used to measure the growth rates in the mid- and high-growth regions. Each point on the growth rate curves was obtained at a definite stress intensity, and several tests were used to confirm each observation.

Values of the plane strain fracture toughness, $K_{fc}$, were determined from compact tension specimens on which fatigue tests had been terminated and the uncracked ligaments were still long enough for the determination of $K_{fc}$. These tests were performed in accordance with ASTM standards, and each test was found to be valid by the ASTM criteria.\textsuperscript{13} Each $K_{fc}$ value is an average of at least three determinations and are probably accurate to $\pm 1$ MPa m$^{-1}$.

Microstructures were examined using optical microscopy, and the fracture surface morphology was characterized by the use of scanning electron microscopy. The austenitic grain size was about 8 to 12 μm in both steels. The amounts of retained austenite were determined by means of an X-ray technique\textsuperscript{14} using CrKα radiation. The effective depth of penetration of these X-rays is about 11 μm. This was significant in our determination of the retained austenite contents at fractured surfaces, since the radius of the plastic zone in steels with the lowest values of $K_{fc}$, about 17 MPa m$^{-1}$, was about 10 μm. The X-ray contents measured at the fracture surfaces were thus characteristic of the plastic region associated with the fracture.

### Table I. Compositions of Steels, Wt Pct

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<th></th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cr</th>
<th>Mo</th>
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<td>0.012</td>
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<tr>
<td>BG42</td>
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<td>0.011</td>
<td>0.003</td>
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<td>14.44</td>
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