Assessment and Predictions of HAZ Tensile Properties of High-Strength Steels

An investigation into the grain-coarsened HAZ of single-pass weldments reveals a means of predicting HAZ strength and ductility

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ABSTRACT. In the present investigation, the grain-coarsened HAZ tensile properties of controlled rolled accelerated cooled (CONRAC) and quenched and tempered (QT) high-strength steels (Ry<sub>0.2</sub> > 500 MPa) have been examined using the weld simulation technique. It is shown that very high yield and tensile stresses (Ry<sub>0.2</sub> > 900 MPa, R<sub>y</sub> > 1200 MPa) may be found at fast cooling rates (ΔT<sub>1/2</sub> < 5 s), with a consequent loss in ductility. For slow cooling rates (ΔT<sub>1/2</sub> > 10-20 s), however, welding of QT steels may cause a considerable loss in base plate yield strength. Based on the results obtained, empirical equations have been developed to predict grain-coarsened HAZ strength and ductility in single-pass weldments. A comparison shows that yield and tensile strength can be predicted with an accuracy of about ±50 MPa, while elongation calculations reveal a precision of ±3%.

Introduction

The major impetus for developments in steelmaking practice over the last decades has been provided by the need for materials with an improved heat-treatment response to external loading. The present investigation was undertaken with the objective of providing quantitative information on strength and ductility of the grain-coarsened HAZ of modern structural steels, based on tensile testing of weld thermal simulated specimens. From stress-strain relationships and hardness measurements, empirical equations have been developed to predict HAZ yield and tensile strength, as well as ductility from a knowledge of base metal chemical composition and welding parameters.

Materials and Experimental Procedure

Materials

In this study, one controlled rolled accelerated cooled (Steel A) and two quenched and tempered (Steels B and C) steels have been examined. It is apparent from data in Table 1 that these steels represent a broad variation in base plate chemical composition (i.e., niobium, copper, nickel and molybdenum contents). This, in turn, gives rise to significant differences in steel microstructure (Fig. 1) and strength level – Table 2.

KEY WORDS

High-Strength Steel
Grain-Coarsened HAZ
Weld Cooling Time
HAZ Transformation
HAZ Yield Strength
HAZ Tensile Strength
HAZ Ductility
HAZ Hardness
Strain Hardening
Empirical Equations

Table 1—Steel Composition (wt-%)

<table>
<thead>
<tr>
<th>Steel&lt;sup&gt;(a)&lt;/sup&gt;</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cu</th>
<th>Ni</th>
<th>Mo</th>
<th>Cr</th>
<th>Al</th>
<th>V</th>
<th>Ti</th>
<th>Nb</th>
<th>P&lt;sub&gt;cr&lt;/sub&gt;&lt;sup&gt;(d)&lt;/sup&gt;</th>
</tr>
</thead>
<tbody>
<tr>
<td>A&lt;sup&gt;(b)&lt;/sup&gt;</td>
<td>0.13</td>
<td>0.33</td>
<td>1.52</td>
<td>0.002</td>
<td>0.001</td>
<td>0.18</td>
<td>0.19</td>
<td>0.01</td>
<td>0.04</td>
<td>0.028</td>
<td>0.01</td>
<td>0.008</td>
<td>0.034</td>
<td>0.23</td>
</tr>
<tr>
<td>B&lt;sup&gt;(b)&lt;/sup&gt;</td>
<td>0.13</td>
<td>0.25</td>
<td>1.30</td>
<td>0.006</td>
<td>0.001</td>
<td>0.01</td>
<td>0.64</td>
<td>0.17</td>
<td>0.02</td>
<td>0.036</td>
<td>0.03</td>
<td>0.017</td>
<td>0.002</td>
<td>0.21</td>
</tr>
<tr>
<td>C&lt;sup&gt;(b)&lt;/sup&gt;</td>
<td>0.14</td>
<td>0.26</td>
<td>1.29</td>
<td>0.005</td>
<td>0.001</td>
<td>0.24</td>
<td>0.97</td>
<td>0.18</td>
<td>0.02</td>
<td>0.044</td>
<td>0.03</td>
<td>0.003</td>
<td>0.003</td>
<td>0.26</td>
</tr>
</tbody>
</table>

<sup>(a)</sup> Less than 0.003% N and 0.0003% B.
<sup>(b)</sup> 50-mm plate thickness.
<sup>(c)</sup> 80-mm plate thickness.
<sup>(d)</sup> P = C + Cr + Mn + Si + Mo + V. 

Weld Thermal Simulation and Tensile Testing

Assessment of grain-coarsened HAZ strength and ductility was done on the basis of tensile testing of weld simulated specimens. Bars of 8-mm (0.3-in.) diameter (50- to 70-mm/2- to 2.8-in. length) were heated by induction to a peak temperature of 1350°C (2462°F) followed by controlled cooling in helium or argon. The cooling time, ΔT<sub>1/2</sub>, which is...
an adequate index for the weld cooling cycle, was varied systematically from 2 s (as obtained by water quenching from the peak temperature) to 65 s. This was made possible by adjustments of the gas flow rate. The temperature-time cycles were recorded by chromel/alumel thermocouples spot welded at the midlength of the bar surface.

Due to small bar dimensions involved, specimens will exhibit a uniform microstructure and hardness throughout their length after weld simulation.

The tensile properties were assessed in the longitudinal direction, i.e., with the specimen length axis parallel to the plate rolling direction. Test bars were taken from the plate surface of Steels B and C, and from the center of Steel A. The thermally cycled specimens were machined down to dimensions shown in Fig. 2 prior to tensile testing, which was performed with a constant crosshead speed of 2 mm/min. Because of a low base plate content of nitrogen (and carbon), continuous yielding was observed in all cases, independent of cooling rate.

**Metallography**

One half of the tensile specimens was prepared for metallographic examination. The volume fraction of various microstructural constituents was calculated from more than 600 points counted at 500 and 1000X in the light microscope, using the following classification system:

- M - lower bainite and martensite
- SP - ferrite sideplates (i.e., Widmanstätten ferrite and upper bainite), intergranularly or intragranularly nucleated parallel ferrite laths
- AF - acicular ferrite, intragranularly nucleated separate ferrite needles of a high-aspect ratio
- P - grain boundary or polygonal ferrite and pearlite

The metallographic examination also included determination of Vickers hardness (HV) of selected specimens.

**Results and Discussion**

**Grain-Coarsened HAZ Transformations Behavior**

It is apparent from the data contained in Fig. 3 that the HAZ hardenability of Steel C is substantially higher than that of Base Plates A and B (only martensite and ferrite sideplates did form). This observation is consistent with a higher base plate alloying level (i.e., a higher Pcm value) in the case of Steel C - Table 1. Typical grain-coarsened HAZ microstructures are illustrated by the micrographs in Fig. 4. As expected, Base Plate C exhibits the highest HAZ hardness level, as shown by the results in Fig. 5. In this figure, both data for simulated specimens and bead-on-plate welds are compared with predicted values based on the Suzuki BL70 formula (Ref. 1). With the exception of very fast cooling rates, the hardness level can be predicted with excellent precision from the Suzuki equation.

**Basic Tensile Test Data**

The effect of cooling time Δt/s on the resulting HAZ strength and ductility is illustrated graphically in Fig. 6. It is seen that the yield and tensile strength are rapidly reduced by an increase in Δt/s, approaching a constant level of about 500 and 700 MPa (72.5 and 101.5 ksi) for Rp0.2 and Rm, respectively, for values of Δt/s larger than about 40 s. At fast cooling rates (Δt/s < 8 s), very high strength levels can be observed, which means that the yield and tensile strength may exceed 900 and 1200 MPa (130.5 and 174 ksi), respectively, in low-heat-input MMA or GMA welding (E < 2 MJ/m). Because of a higher hardenability, Steel C exhibits the highest strength level at intermediate and slow cooling rates (Δt/s > 8 s). In contrast, lower strength values are generally obtained for Base Plate B owing to a low Pcm value and low martensite content. Thus, the grain-coarsened HAZ yield and tensile strength are strongly dependent on the volume fraction of martensite formed on cooling.

The HAZ ductility appears to be slightly different between the steels examined when comparison is made on the basis of reduction in area, while the elongation is similar. The ductility is generally reduced at fast cooling rates (Δt/s < 8 s) because of the martensite formation.

**Effect of Weld Cycles on Base Plate Strength**

Current design criteria for welded offshore structures are either based on the

**Table 2—Base Plate Mechanical Properties**

<table>
<thead>
<tr>
<th>Steel</th>
<th>Yield strength Rp0.2 (MPa)</th>
<th>Tensile strength Rm (MPa)</th>
<th>Elong. (%)</th>
<th>Reduction in Area (%)</th>
<th>Hardness HV5 (kg/mm²)</th>
<th>CVN at −40°C (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>513</td>
<td>628</td>
<td>29</td>
<td>70</td>
<td>214</td>
<td>183</td>
</tr>
<tr>
<td>B</td>
<td>607</td>
<td>679</td>
<td>21</td>
<td>80</td>
<td>250</td>
<td>251</td>
</tr>
<tr>
<td>C</td>
<td>628</td>
<td>727</td>
<td>36</td>
<td>76</td>
<td>270</td>
<td>274</td>
</tr>
</tbody>
</table>

**Fig. 2—Geometry and dimensions of tensile test specimens (Norsk Standard NS 10 100)**
yield strength or Young’s modulus. Since the strength level within the actual HAZ is unknown, it has been assumed that this level always exceeds that of the base metal. Such assumptions, however, are not necessarily correct and may lead to considerable underestimations when it comes to high-strength steels, as will be shown by the present investigation. In Fig. 7, the basic HAZ strength data reported in Fig. 6 have been modified with respect to the base metal tensile properties and replotted versus cooling time $\Delta T_{9/5}$. It is seen from the figure that the initial yield strength has been reduced for a wide range in cooling time ($\Delta T_{9/5} > 10$ s for Steel B and $>25$ s for Steels A and C), while the tensile strength always exceeds that of the base metal. Moreover, the base plate yield strength is further reduced within the grain-refined and intercritical regions of the HAZ (Ref. 2). This means that, under high-heat-input conditions ($>2.5$ MJ/m), the HAZ may exhibit a lower yield strength than that of the base metal, which clearly will affect the mechanical integrity of the weldment. In order to clarify to what extent welding may affect the yield-to-tensile strength ratio, the data presented in Fig. 6 have been replotted versus the cooling time $\Delta T_{9/5}$ in Fig. 9. It is seen from the figure that this ratio decreases steadily with increasing cooling time, approaching a constant level of about 0.7. This ratio is increased to about 0.8 for the accelerated cooled steels. Moreover, in the case of quenched and tempered steels, the yield point approaches 90% of the tensile strength, as seen from the data for Steels B and C in Table 2. This will, in turn, increase the risk of plastic instability during service. In general, strengthening by work hardening is well described by the following relationship:

$$\sigma = K\epsilon^n$$

Here $\sigma$ denotes the true stress, $\epsilon$ is the true strain, $K$ is a constant (strain hardening coefficient) and $n$ is the strain hardening exponent.

Several engineering stress-strain curves from tensile testing have been used in calculating $n$ values for a wide range in cooling time $\Delta T_{9/5}$. The results obtained are summarized in Fig. 10. It is shown that the value of $n$ is strongly influenced by the weld cooling rate. As expected, the strain hardening exponent, $n$, is reduced.
with increasing cooling rates, particularly for \( \Delta t_{1/5} \) values below 10 s. This is due to the formation of martensite with a high dislocation density. At slower cooling rates, the formation of ferrite sideplates, in addition to autotempered martensite, gives rise to a substantial increase in the strain hardening ability.

Based on regression analysis of the data contained in Fig. 10, the following parabolic relationship between the strain hardening exponent \( n \) and the cooling time \( \Delta t_{1/5} \) has been developed.

\[
n = 0.065 (\Delta t_{1/5})^{0.17}
\]

(2)

From the present investigation, it is seen that the value of \( n \) is virtually independent of the base plate chemical composition. This is in agreement with previously published data on base metal tensile testing (Ref. 5), where a value of \( n \) close to 0.04 has been found for quenched specimens of various chemical compositions. This value is slightly lower than the value of 0.073 obtained for grain-coarsened HAZ at fast cooling rates (\( \Delta t_{1/5} = 2 \) s).

Predictions of Grain-Coarsened HAZ Yield Strength

From the literature reviewed, it is apparent that the 0.2\% offset yield strength, \( R_{p0.2} \), can be calculated with good precision from Vickers hardness measurements in agreement with the following relationship (Ref. 6):

\[
R_{p0.2} (\text{MPa}) = 3.3 \, \text{HV} \, (0.1)^m
\]

(3)

Here, HV denotes the Vickers hardness number, \( m(= n + 2) \) is the exponent in
Meyers law and n is the strain hardening exponent. The value of m is about 2.5 for fully annealed materials and approximately 2 for fully strain-hardened metals (Ref. 7).

In the case of grain-coarsened HAZ, the regression analysis gave the following relationship:

\[ R_{p0.2} \text{ (MPa)} = 3.1 \text{ HV} (0.1)^n - 80 \]  

(4)

Since n is related to the weld cooling parameter \( \Delta t_{85} \) through Equation 2, the grain-coarsened HAZ yield strength can be calculated as follows:

\[ R_{p0.2} = 3.1 \text{ HV} (0.1)^{0.065 (\Delta t_{85})^{0.17}} - 80 \]  

(5)

In the present investigation, the following expression has been developed for grain-coarsened HAZ:

\[ R_m = 3.5 \text{ HV} (1-n) \left( \frac{12.5}{1+n} \right) - 92 \]  

(7)

The value of n can be calculated from Equation 2. The tensile strength predicted from Equation 7 has been plotted versus measured values in Fig. 12. As experienced with yield strength, the tensile strength can be calculated with an accuracy of ±50 MPa. Consequently, the grain-coarsened HAZ yield and tensile strength can be estimated from measured or calculated hardness values and the weld cooling time \( \Delta t_{85} \) with sufficient precision.
Predictions of Grain-Coarsened HAZ Ductility

In Fig. 13 the total elongation at fracture has been plotted versus tensile strength. Although there is a relatively large scatter in elongation values for a given tensile strength level, the following relationship is found (with a correlation coefficient \( r^2 \) of 0.6):

\[
A_{35} \% = 5.75 \times 10^4 R_m^{-1.25}
\] (8)

Values calculated from Equation 8 have been plotted versus measured elongation in Fig. 14. In general, elongation at the fracture can be predicted with a precision of ±3%, with a small overestimation at low elongations and underestimation at high values.

Practical Implications

In a real HAZ, the yield behavior will be much more complex than that inferred from the present investigation. In particular, local yielding may be suppressed by surrounding regions of higher yield strength. This point is essential, since it has been shown by Fig. 6 that welding with medium and high heat inputs \((E > 2.5 \, \text{MJ/m}, \Delta t_{1/2} > 12 \, \text{s})\) can provide formation of HAZ regions with yield strength below that of the base plate. Under high-heat-input welding conditions, the width of the HAZ developed can be relatively large. This may, in turn, influence both fracture toughness and fatigue properties in small scale testing, from which design curves are frequently established. In particular, the calculation of the elastic component of CTOD requires knowledge of HAZ yield strength level. It has been assumed that the CTOD values are influenced by the yield strength of both the weld metal and the base plate, in addition to that of the HAZ, due to the small width of the HAZ relative to the specimen thickness (Ref. 9). For this reason, it has been suggested that the yield strength applied in calculations is the average between the base plate and weld metal strength (Ref. 10). As shown by the present investigation, this may lead to a substantial underestimation of the yield strength level within the HAZ region where crack propagation normally occurs, particularly under low-heat-input welding conditions. This implies that the elastic contribution in the formula becomes too high under circumstances where the CTOD value is low (<0.1 mm). An opposite trend may be expected for combinations of high-heat-input welding of QT steels and low CTOD values, due to the associate loss in HAZ strength relative to that of the base plate. In the case of higher CTOD values (>0.1 mm), however, the elastic contribution is negligible, indicating that the calculated CTOD value is virtually independent of HAZ yield strength level.
under prevailing conditions.

In addition, it must be recognized that, for a proper use of high-strength steels, the base plate yield-to-tensile strength may be significantly increased by welding, with an associate reduction in strain hardening exponent. This means that the plastic deformation prior to instability occurs over a narrow range in stress, with a consequent loss in overall elongation in, e.g., wide-plate testing (Ref. 11).

Conclusions

The main conclusions that can be drawn from the present investigation are as follows:

1) In the case of low heat-input welding, a very high yield and tensile strength ($R_{p0.2} > 900$ MPa, $R > 1200$ MPa) may develop within grain-coarsened HAZ, with a consequent loss in tensile ductility (elongation less than 15%).

2) For high heat input welding (represented by $\Delta u_{0.5}$ values above 10 s for Steel B and 25 s for Steels A and C), the base plate yield and tensile strength may be substantially reduced, approaching a constant level of 500 and 700 MPa, respectively.

3) The strain hardening exponent is strongly dependent of the HAZ microstructure. Low values of the strain hardening exponent $n (<0.10)$ have been found for a martensite microstructure formed during water quenching from the peak temperature of 1350°C.

4) Empirical equations have been developed for quantitative predictions of grain-coarsened HAZ strength level and ductility in single-pass weldments. At the present stage, HAZ yield and tensile strength can be calculated with an accuracy of about ±50 MPa, while elongation calculations reveal a precision of ±3%.

Acknowledgments

The authors wish to thank Mr. P. E. Kvaale (Statoil), Ms. S. Karlson and Mr. O. Handa (Norsk Jernverk A/S) for conducting the tensile tests. Thanks are also due to the Royal Norwegian Council for Scientific and Technical Research and Statoil for financial support.

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