EFFECT OF HEAT TREATMENT ON THE MECHANICAL PROPERTIES AND MICROSTRUCTURE OF WELDED ABRASION RESISTANT STEEL AR200

by

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Abstract

Fusion welding process is based on localized melting of materials of similar composition in order to form a permanent joint. Fusion welding is a cost efficient process widely used in the industry to join a variety of materials, weld geometries and weld characteristics.

In this research, the effect of heat treatment on the microstructure, mechanical properties and residual stresses of welded abrasion resistant steel AR200 was investigated. Standardized testing as per ASTM regulations has been carried out and included the measurement of the tensile strength, impact strength, hardness and Jominy hardenability on welded and virgin AR200 steel samples. Then, Scanning Electron Microscopy (SEM) and light optical microscopy were carried out to characterize the alloy’s microstructure as a function of heat treatment temperature and time. Finally, residual stress was measured using a neutron diffraction technique, and the elastic residual stress was mapped in the vicinity of welds in AR200 C-sections.

The tensile test results revealed a significant variation in the magnitude of yield and ultimate tensile strength between the virgin and heat treated welded AR200 samples, with the welded and heat treated samples exhibiting reduced ductility and strength. SEM microscopy revealed the entrapment of intermetallic compounds originating from weld slag, which acted as stress-concentrations and likely contributed to the initiation of fracture in the tensile specimens.

The neutron diffraction results revealed the distribution of residual strain in three orthogonal directions in the virgin AR200 material, heat affected zone and the weld itself. The results suggest that although the heat treatment decreased and homogenized the residual strains,
the duration of the heat treatment did not have a significant effect, and thus further optimization of the heat treatment parameters is required.

These research findings confirm that successful welding of AR200 abrasion resistant steel must consider both the chemical composition of the steel as well as the heat treatment, part geometry and welding operator technique variability.
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List of Symbols and Abbreviations

Symbols

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<th>Symbol</th>
<th>Definition</th>
</tr>
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<tbody>
<tr>
<td>%A</td>
<td>percent area reduction</td>
</tr>
<tr>
<td>d</td>
<td>distance between two adjacent crystallographic planes, Å</td>
</tr>
<tr>
<td>d₀</td>
<td>distance between two adjacent crystallographic planes in a stress free crystal, Å</td>
</tr>
<tr>
<td>Δd</td>
<td>difference between d – d₀</td>
</tr>
<tr>
<td>E</td>
<td>modulus of elasticity</td>
</tr>
<tr>
<td>%l</td>
<td>percent elongation</td>
</tr>
<tr>
<td>n</td>
<td>order of reflection</td>
</tr>
<tr>
<td>%wt</td>
<td>weight percent</td>
</tr>
<tr>
<td>Δ</td>
<td>wavelength, Å</td>
</tr>
<tr>
<td>λ</td>
<td>strain</td>
</tr>
<tr>
<td>ε</td>
<td>Poisson’s ratio</td>
</tr>
<tr>
<td>ν</td>
<td>stress</td>
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<tr>
<td>σ</td>
<td>stress</td>
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<td>stress</td>
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Chemical Elements

<table>
<thead>
<tr>
<th>Element</th>
<th>Description</th>
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<tbody>
<tr>
<td>Al</td>
<td>aluminum</td>
</tr>
<tr>
<td>B</td>
<td>boron</td>
</tr>
<tr>
<td>Bi</td>
<td>bismuth</td>
</tr>
<tr>
<td>C</td>
<td>carbon</td>
</tr>
<tr>
<td>Cr</td>
<td>chromium</td>
</tr>
<tr>
<td>Cu</td>
<td>copper</td>
</tr>
<tr>
<td>Dy</td>
<td>dysprosium</td>
</tr>
<tr>
<td>Fe</td>
<td>iron</td>
</tr>
<tr>
<td>Mn</td>
<td>manganese</td>
</tr>
</tbody>
</table>
Mo  molybdenum
Nb  niobium
Ni  nickel
O  oxygen
P  phosphorus
S  sulphur
Si  silicon
Sn  tin
Ti  titanium
V  vanadium
W  tungsten
Y  yttrium

**Abbreviations**

*AECL*  Atomic Energy Canada Limited
*AISI*  American Iron and Steel Institute
*AR 200/400*  Abrasion Resistant steel type 200/400
*ASTM*  American Society of Testing Materials
*AWS*  American Welding Society
*CCT*  Continuous Cooling Transformation
*CE*  Carbon Equivalent
*CNBC*  Canadian Neutron Beam Centre
*CNC*  Computer Numeric Control
*CSA*  Canadian Standard Association
*CWB*  Canadian Welding Bureau
*DBTT*  Ductile to Brittle Transition Temperature
*DCRP*  Direct Current Reversed Polarity

*Sometime designated as* Direct Current Electrode Positive DCEP
<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
</tr>
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<tbody>
<tr>
<td>DCSP</td>
<td>Direct Current Straight Polarity</td>
</tr>
<tr>
<td>FEA</td>
<td>Finite Element Analysis</td>
</tr>
<tr>
<td>FCAW</td>
<td>Flux Core Arc Welding</td>
</tr>
<tr>
<td>GMAW</td>
<td>Gas Metal Arc Welding</td>
</tr>
<tr>
<td>GTAW</td>
<td>Gas Tungsten Arc Welding</td>
</tr>
<tr>
<td>HAZ</td>
<td>Heat Affected Zone</td>
</tr>
<tr>
<td>HT</td>
<td>Heat Treatment</td>
</tr>
<tr>
<td>MCAW</td>
<td>Metal Core Arc Welding</td>
</tr>
<tr>
<td>ND</td>
<td>Neutron Diffraction</td>
</tr>
<tr>
<td>NRC</td>
<td>National Research Council</td>
</tr>
<tr>
<td>NSERC</td>
<td>Natural Sciences and Engineering Research Council (Canada)</td>
</tr>
<tr>
<td>RS</td>
<td>Residual Stress</td>
</tr>
<tr>
<td>SEM</td>
<td>Scanning Electron Microscope</td>
</tr>
<tr>
<td>UTS</td>
<td>Ultimate Tensile Strength</td>
</tr>
<tr>
<td>YS</td>
<td>Yield Strength</td>
</tr>
<tr>
<td>X-EDS</td>
<td>X-ray Energy Dispersive Spectroscopy</td>
</tr>
</tbody>
</table>
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Also would like to thank you my colleagues from Okanagan College for support this give me.

It is important to me to express my gratitude, in memory of my father Neyko Neykov, who inspired me in the Engineering profession and give me the love to aviation.
Dedication

To my son Hristo Neykov... to have inspiration in his life.
Chapter 1. Introduction

In this chapter, an overview of the fusion welding process and common problems associated with the weldability of steels will be presented. Also, tests used to evaluate the mechanical properties of welded samples will be outlined, followed by a brief description of the scope of the thesis research.

Fusion welding (FW) is one of the most frequently used methods in the manufacturing industry for the joining of metal parts, structures and semi-assemblies. Fusion welding allows for a relatively easy joining of a variety of metals at a low cost. Despite such benefits, fusion welding has technological problems. The fusion welded seam is subjected to melting and solidification, therefore the joined parts may exhibit distortion and be out of dimensional tolerance due to thermal strain. After cooling, the welding seam retains residual stresses and hot or cold cracks may readily form. Other defects include: lack of fusion penetration, undercuts, porosity, and slag inclusions in the weld. In addition, the chemical composition of some steels makes their welding difficult due to elements that form oxides, carbides and intermetallic compounds. Finally, the high heat used in the fusion welding process changes the microstructure in the heat affected zone (HAZ); therefore, the mechanical properties of the joint are often not uniform. Given the potential benefits and widespread applications of fusion welding, there is a need to determine and find methods to manipulate the quality of welds, in particular for the commonly used abrasion resistant steels, since modern industrial applications demand high quality of welds. The Canadian Standard Association (CSA) and the Canadian Welding Bureau (CWB) have implemented certification processes for industries using fusion welding processes in an effort to help guide weld quality and performance [1][2][3].
According to the American Welding Society (AWS) classification, steels with increased amounts of manganese (Mn), silicon (Si) and carbon (C) have welding problems, leading to a reduced mechanical strength, the evolution of high residual stresses and the formation of intermetallic compounds [4].

Currently, there is a paucity of literature on the weldability of AR200 steels. In particular, the evolution of residual stresses as a function of welding parameters is not reported for AR200. AR200 is relatively new steel with good mechanical properties, especially good abrasion resistance. As a result, it has gained interest of manufacturers for many industrial cost-sensitive applications. AR200 steel components were used in the mining industry for ore excavating equipment, material handling equipment, aggregate processing, road building machines and construction equipment. As a result, failure of welded sections may have significant impact on economic and human health aspects of industrial operations. [5]. Since in many of these industrial applications AR200 was used in stressed structures, the quality of welded AR200 sections must be monitored, and falls under the regulation guidelines prescribed by CAN/CSA, ASTM, as well as CWB and AWS.

Thus, the focus of this research was to study the weldability and the effect of heat treatment on the microstructure and mechanical properties of AR200 steel sections welded by a fusion welding process called the Gas Metal Arc Welding (GMAW). Neutron Diffraction (ND) technique was used to measure residual strain in the welded seam and the HAZ. Despite efforts to perform all experimental work according to established standards, limitations pertaining to sample availability, measurement technique accuracy and sampling (i.e., statistical) considerations remained. These limitations are duly discussed in relevant chapters.
This thesis document has been structured as follows: Chapter 2 of the thesis presents a literature review on the weldability of AR200. In Chapter 3, experimental procedures are described. Chapter 4 provides the discussion of the results. In Chapter 5, conclusions and recommendations are presented. Additional images are included in the Appendix. Figure 1 summarizes the scope of the work carried out in this thesis.

**Figure 1:** Scope of experimental work performed in this thesis.
Chapter 2. Literature review

This chapter provides a review of welding parameters relevant to accomplish a successful fusion welding operation. These parameters include the chemical composition of the steel, welding method and pre-heating/post-heating of the base material. Also, the effect of intermetallics on the tensile strength, impact strength and hardness properties of the welded joint is discussed. A review of mechanisms involved in residual stress evolution and the various methods for quantifying residual stresses conclude the chapter.

2.1. Steels

In the modern society, steel-based materials are used in many applications like machines, transportation vehicles, pressure vessels, kitchen appliances and many others. Steel is used primarily due to its attractive mechanical properties, such as strength and formability. Steel is an alloy of iron and carbon. The carbon content in steel varies from 0.05 wt% to 2 wt%. As the carbon content increases, the steel strength, castability and hardness generally improve. However, the increase of carbon concomitantly decreases ductility, weldability, toughness and impact strength.

2.1.1. Weldability of steels

Depending on the chemical composition of the steel alloy, its weldability can vary from excellent to poor. To achieve good weldability of the steel joint, the American Welding Society (AWS) [6] recommends that the chemical composition of the steel remains within certain compositional limits, as shown in Table 1. It can be seen that a low-carbon steel with less than 0.25wt% C, < 0.8wt% Mn, 0.1wt% Si, < 0.03wt% S and P is desirable. Table 1 also identifies
values deemed to be critical concentration limits for the most important elements, which if exceeded, result in poor weldability of steels.

Table 1: Recommended alloying levels suitable for arc welding of steels [6].

<table>
<thead>
<tr>
<th>Element (Symbol)</th>
<th>Preferred amount (wt%)</th>
<th>Too high (wt%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Carbon (C)</td>
<td>0.06 to 0.25</td>
<td>0.35</td>
</tr>
<tr>
<td>Manganese (Mn)</td>
<td>0.35 to 0.80</td>
<td>1.40</td>
</tr>
<tr>
<td>Silicon (Si)</td>
<td>0.10 or less</td>
<td>0.30</td>
</tr>
<tr>
<td>Sulfur (S)</td>
<td>0.035 or less</td>
<td>0.05</td>
</tr>
<tr>
<td>Phosphorous (P)</td>
<td>0.030 or less</td>
<td>0.04</td>
</tr>
</tbody>
</table>

In order to account for the alloying elements while using the classical Fe-C phase diagram, the carbon equivalent (CE) of the steel can be calculated using Equation 1, in order to estimate an equivalent steel composition [6] [7] [16]. This composition considers the effect of chemical elements which are known to influence steel weldability. In general, with increasing CE value, the likelihood of welding defects, such as hot and cold cracks, lack of fusion, undercuts, porosity and slag entrainment in the welding seam increase [6] [7] [16]:

\[
CE = \%C + \left(\frac{\text{Mn wt}\% + \text{Si wt}\%}{6}\right) + \left(\frac{\text{Cr wt}\% + \text{Mo wt}\% + \text{V wt}\%}{5}\right) + \left(\frac{\text{Cu wt}\% + \text{Ni wt}\%}{15}\right)
\]

Equation 1

In Equation 1, the main alloying elements with a negative impact on the weldability are included. Using this concept, a decision can be made regarding whether or not pre-heating and post-heating of the steel alloy is required in order to minimize welding defects. Figure 2 shows
recommended pre/post heating temperature for steels sensitive to cracking. The area above the lines indicates a “no cracking” zone where a welding operation can be performed safely. The area below the lines indicates temperatures where cracking is likely. If the carbon equivalent is less than 0.2wt%, pre-heating and/or post-heating is not necessary, but if the carbon equivalent is between 0.2wt% and 0.4wt%, pre-heating and/or post-heating is mandatory in the temperature range of 100°C - 150°C, as seen in Figure 2 [4]. Further, if the carbon equivalent is higher than 0.4wt%, both pre/post heating must be carried out above 150°C [6].

Table 2 summarizes the relative weldability of steel in relation to the carbon equivalent [6]. As can be seen from the data, low-carbon steels with less than 0.35wt% generally have an excellent weldability, without cracks or other defects observed in the weld. However, as CE increases, weldability decreases, and with CE above 0.4wt%, welding defects readily form. For such alloys, special welding techniques are recommended and include the use of a special welding wire, in addition to pre-heating and/or post-heating of the weld region.
Figure 2: Pre/post heat vs. carbon equivalent content [6].

Table 2: Weldability in relation to carbon equivalent [6].

<table>
<thead>
<tr>
<th>Carbon equivalent CE, wt%</th>
<th>Weldability</th>
</tr>
</thead>
<tbody>
<tr>
<td>&lt; 0.35</td>
<td>Excellent</td>
</tr>
<tr>
<td>0.36 – 0.40</td>
<td>Very good</td>
</tr>
<tr>
<td>0.41 – 0.45</td>
<td>Good</td>
</tr>
<tr>
<td>0.46 – 0.50</td>
<td>Fair</td>
</tr>
<tr>
<td>&gt; 0.50</td>
<td>Poor</td>
</tr>
</tbody>
</table>
The microstructure of abrasion resistant steels can be evaluated using the Schaeffler constitution diagram (Figure 3). In this diagram, the chromium equivalent (Equation 2) and nickel equivalent (Equation 3) are used to predict the evolution of the dominant microstructure and its effect on the weldability.

\[
\text{CrE} = \text{Cr wt}\% + \text{Mo wt}\% + 1.5 \times \text{Si wt}\% + 0.5 \times \text{Cb wt}\% \quad \text{Equation 2}
\]

\[
\text{NiE} = \text{Ni wt}\% + 30 \times \text{C wt}\% + 0.5 \times \text{Mn wt}\% \quad \text{Equation 3}
\]

Based on the Schaeffler diagram, increasing the amount of CrE will yield an alloy with increased ferrite content, while increasing NiE results in an austenitic alloy. In the case of low CrE and NiE contents, the alloy readily forms a hard and brittle martensite, which may significantly contribute to the poor ductility of the alloy in service [8][9].
Another issue hindering successful adaptation of fusion welding processes in the manufacturing industries is the formation of a Heat Affected Zone (HAZ) in the vicinity of the weld. Due to the heat generated in the welding process, the heat affected zone in the vicinity of the welding seam changes the steel’s microstructure. Depending on how far the HAZ extends from the welding seam, a gradient of microstructures with different grain sizes can be observed [8]. Different grain sizes will yield an alloy with varying localized strength and hardness. Close to the weld seam, the grains are usually hard due to rapid solidification and chilling effect. Away from the weld seam, the grains soften and the material begins to exhibit enhanced ductility. As a result, the mechanical properties of the welded material locally vary and it is desirable to homogenize (i.e., reduce variability) the material. Therefore, a heat treatment is often carried out.
to homogenize the microstructure and equalize the grain structure and strength across the welding seam [8], [9], [10].

Other problems in fusion welding involve the formation of cracks during the welding process. Formation of cracks can be divided into two groups based on the crack nucleation temperature: hot and cold cracks. Hot cracks occur during the welding process itself as a result of fast cooling of the liquid weld pool or inappropriate welding technique (e.g., very high weld speeds). Cold cracks are more difficult to detect, since they form under the surface after welding is complete. The main reason for cold cracks is inappropriate welding technique, residual stresses, hydrogen embrittlement and grain structure inhomogeneity [7][10][12].

Other problems in fusion welding involve distortion of the parts due to shrinkage of the welded seam. The dimensional variation arises due to two factors: i) volumetric shrinkage of the metal due to a phase transformation from the liquid phase to the solid phase; and ii) thermal contraction of a hot solid metal during cooling to room temperature. The largest amount of shrinkage occurs during the cooling of the casting to ambient temperature.

Distortion can be eliminated by a stress relieve heat treatment. Most effective stress relieve heat treatments are performed in a large oven and may take up to several hours to fully homogenize the microstructure and eliminate residual stresses in the weld regions.

When not alleviated, distortions can lead to the evolution of residual stresses in welded parts. Residual stresses are formed due to the differences in contraction of the welded seam and the base metal. During solidification, the welded seam contracts, while the base material resists this contraction. Thus, the weld comes under tension and the base material under compression [7][8][13][14]. This process is further discussed in detail in Chapter 2.4.
The above welding issues are strongly related to the chemical composition of the steel alloy. With the exception of specialty alloys, general steel alloys include the following elements which are added to control the mechanical performance of the alloy [8][14].

*Silicon* - increases hardness and the tensile strength, but decreases alloy toughness. Si is required and used as a deoxidizer in steel making, but is detrimental to the surface quality of rolled steel products. Silicon also contributes to the formation of undesirable intermetallic compounds, such as SiO [8][14].

*Vanadium* - stabilizes carbides and therefore enhances hardenability. However, V reduces the temperability of the steel [8][14].

*Chromium* - forms hard chromium-rich carbides, which improve hardenability. Cr also increases the steel’s resistance to corrosion, but also causes brittleness and grain growth [8][14].

*Nickel* - increases the yield and the tensile strengths and increases the toughness of the steel by refining the grains, while enhancing corrosion resistance [8][14].

*Molybdenum* - increases the yield and tensile strengths of the steel by stabilizing carbides. Mo also improves the high temperature strength of the alloys and reduces temper brittleness [8][14].

*Copper* - increases resistance to atmospheric corrosion and improves impact strength [8][14].

*Phosphorous* – is an undesirable element in most alloys as it significantly decreases ductility. However, P helps in material removal operations and is thus added in trace amounts to develop “free machining” steels [8][14].
Sulfur - is an undesirable element in the alloy as it decreases ductility and impact toughness. S is very detrimental for the surface quality of low carbon and manganese steel products [8][14].

Manganese – increases hardness, wear resistance, impact toughness and tensile strength. Manganese also contributes to the formation of MnO intermetallics, which decrease the tensile and impact strength [8][14].

Carbon – increasing the carbon content in steel increases the tensile strength, hardness, hardenability, castability and brittleness. However, an increase of the carbon content in steel generally decreases the ductility, weldability and impact strength [8][14]. Carbon forms iron carbide (Fe₃C) also known as cementite, which is hard and brittle. Thus, with increasing carbon content in the alloy, strengthening of the alloy by cementite precipitation at the grain boundaries can be achieved [8][14]. Strengthening is primarily achieved by lattice distortion by the solute atoms.

Abrasion resistant steels are used in industry primarily for applications where abrasion of the contact surfaces on moving parts is a design requirement. The chemical composition of these steels is carefully selected to withstand extensive abrasion and wear. Research on these types of steels is divided in two groups: Steels that are researched for their wear resistance and steels that are researched for their weldability. Table 3 summarizes steels that have been researched for weldability.
### Table 3: Summary of chemical composition of weldable abrasion resistant steels.

<table>
<thead>
<tr>
<th>Trade mark</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Other</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>NM 360</td>
<td>0.139</td>
<td>1.3</td>
<td>0.34</td>
<td>0.026</td>
<td>0.0078</td>
<td>Cr 0.51 Mo 0.25 Ni 0.26</td>
<td>98.18</td>
</tr>
<tr>
<td>A, B*</td>
<td>0.27</td>
<td>1.5</td>
<td>0.80</td>
<td>-</td>
<td>-</td>
<td>Ni 0.052 Cu 0.017</td>
<td>97.43</td>
</tr>
<tr>
<td>AR400</td>
<td>0.30</td>
<td>1.50</td>
<td>0.30</td>
<td>0.035</td>
<td>0.040</td>
<td>Cr 0.65 Mo 0.10 Ti 0.005</td>
<td>97.07</td>
</tr>
<tr>
<td>AR400X</td>
<td>0.16</td>
<td>1.34</td>
<td>0.42</td>
<td>0.015</td>
<td>0.005</td>
<td>Cr 0.49 Mo 0.18 Ti 0.027</td>
<td>98.06</td>
</tr>
</tbody>
</table>

* Trade mark not specified

2.1.2. **Welding of AR200**

#### 2.1.2.1. Overview of AR200

AR200 is primarily used when component abrasion resistance is required. For example, AR200 can be used for material handling and conveyor components, concrete mixer drums and screw conveyers. Table 4 lists the chemical composition of AR200 based on data taken from the SSAB data sheet (alloy manufacturer) as well as from a mill certificate pertaining to the AR200 steel used in this research [5][18]. As Table 4 shows the chemical composition of AR200 has carbon at 0.33 – 0.35wt%, manganese at 1.2 – 1.24wt% and silicon at 0.22 - 0.25wt%.

Substituting the values of alloy constituents into Equation 1 yields carbon equivalent of 0.65wt%, which suggests that weldability of this alloy may be challenging.
Table 4: Chemical composition of AR200.

<table>
<thead>
<tr>
<th>Element</th>
<th>SSAB Data sheet [5], wt%</th>
<th>AR200 mill certificate [18], wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>0.35</td>
<td>0.33</td>
</tr>
<tr>
<td>Mn</td>
<td>1.20</td>
<td>1.24</td>
</tr>
<tr>
<td>Si</td>
<td>0.25</td>
<td>0.22</td>
</tr>
<tr>
<td>Cr</td>
<td>Not listed</td>
<td>0.09</td>
</tr>
<tr>
<td>Mo</td>
<td>Not listed</td>
<td>0.03</td>
</tr>
<tr>
<td>V</td>
<td>&lt; 0.05</td>
<td>0.007</td>
</tr>
<tr>
<td>Cu</td>
<td>Not listed</td>
<td>0.30</td>
</tr>
<tr>
<td>Ni</td>
<td>Not listed</td>
<td>0.12</td>
</tr>
<tr>
<td>P</td>
<td>0.012</td>
<td>0.012</td>
</tr>
<tr>
<td>S</td>
<td>0.005</td>
<td>0.001</td>
</tr>
</tbody>
</table>

In order to successfully weld AR200 steel, SSAB recommends AWSE10018, E9018 or E90S wire for GMAW operations. These wires contain vanadium and thus the welding operation shall be carried out in a low-hydrogen environment in order to minimize the formation of intermetallic compounds in the welded seam. Further, of particular interest is also the removal of FeO and Fe₂O₃ slag, which forms during the welding process via a reaction with ambient oxygen [19], or the dissociation of CO₂ shielding gas [11].

In summary, the abrasion resistant steel AR200 is used for niche applications in the industry where excellent abrasion resistance of the material is required. However, the poor weldability of the steel presents unique challenges, mainly due to AR200’s composition leading to the formation of detrimental intermetallics.
2.1.2.2. Intermetallic compounds in AR200

Intermetallic compounds can have a dual effect on the mechanical response a steel alloy. On one hand, intermetallics can be intentionally introduced into the steel making process (e.g. via Ferro-alloys) in order to purify the chemical composition of the steel and to serve as a deoxidizer. In contrast, some intermetallics deteriorate the mechanical properties of the steel and are considered undesirable.

Oxides can readily form in the steel during the steel making process, since oxygen is blown through the molten iron to reduce carbon content from roughly 4.0wt% down to 0.1 ~ 0.5wt% [11][12]. During this process, oxygen reacts not only with the carbon in the liquid melt, but also with iron and manganese. Most intermetallics remain isolated in the slag, but some remain trapped in the liquid metal and are considered impurities and inclusions [8]. If the size of the impurities is large and they retain sharp edges, these impurities may create stress concentrations and deteriorate the steel’s strength.

The ASM Handbook [4] suggests that intermetallic compounds in welds are created by combining two or more metallic elements. The compounds can be formed by combining metalloids or a combination of metallic and metalloid constituents. The crystal structure of the intermetallic compounds is different from that of its original constituents, and in general the properties of intermetallics are between metals and ceramics. If ionic and covalent solids combine, the resulting intermetallic compound is often hard and brittle. Further, these compounds have highly directional properties such as strength, stiffness, thermal resistance and chemical inertness [8].

Formation of oxides can be explained by the affinity of oxygen to chemically react with liquid iron, manganese, titanium, silicon, aluminum and other elements present in the alloy
system. There are three main sources of oxygen ingress into the molten metal during a welding
operation: First, when no inert shielding gas is used, oxygen enters the molten metal from the
surrounding atmosphere [11]. Second, when CO₂ is used as a shielding gas, it may dissociate into
carbon monoxide and oxygen gas [11] and subsequently react with deoxidizing elements in the
flux wire and the molten metal. Third, if contact surfaces are not properly cleaned oxides may
form [11]. Additionally, the weld can be contaminated through wet/moist coated welding
electrodes and fluxes [11].

In contrast, there are desirable intermetallics which form during fusion welding and help
take up oxygen or other impurities from the molten metal pool and isolate them into the slag.
Such intermetallics are based on rutile, TiO₂, Bi₂O₃ or SiO₂ [11]. Also, using bismuth oxide
(Bi₂O₃) in fluxes improves the slag removal from the welds. However, the residual bismuth in
the solidified metal tends to isolate at the grain boundaries if the steel is heated above 600°C,
which may cause brittleness of the welded joint [11] [21].

Research done by Guimardes et al [16] suggests that although abrasion resistant steels
have very high hardness (~450 HB), with their martensitic microstructure the alloys are prone to
cold cracking during welding. Tests were performed to evaluate susceptibility to cold cracking
induced by hydrogen embrittlement. GMAW welded samples were prepared with varying heat
input. The researchers took into consideration the CE value and the thickness of test plate to
determine the preheating temperature suitable for the steel alloys. Microstructure observations
suggested that the steels did not benefit from preheating. However, steels with lower CE
coalesced better, performed satisfactorily during a bending test, and also demonstrated high
energy absorption during a Charpy V notch impact test [15]. In contrast, research done by
Adonyi et al [17] investigated the susceptibility of AR400 and AR400X to cracking in the HAZ
welded by the Submerged Arc Welding (SAW) process. Their conclusion was that AR400X is weldable if the welding procedure involves a pre-heat treatment to avoid hydrogen-induced cracking.

2.2. Residual stresses

Residual stress (RS) can be defined as “the self-equilibrating internal locked-in stress remaining within a body with no applied (external) force, external constraint or temperature gradient” [8][22][39][43]. RS can be classified in two major categories. The first category can be described as a macro, i.e. long-range RS. This type of residual stress is spread over a large area or over the entire part. The second type of RS is are a micro-residual stress, i.e. short range stress, which acts on the order of a crystal [8][23].

Residual stresses have a negative effect on the performance of the material. At the micro level, RS are initiators of crystallographic defects. On the macro scale, RS contribute to the formation of cracks [22].

The formation of residual stresses in fusion welding is due to the restriction of the movements of welded parts, and also because the heat of the welding seam is not distributed evenly [24][25]. Residual stresses are produced since the molten metal undergoes volumetric shrinkage. These thermal stresses differ at different stages of the cooling process. In Figure 4, section B – B shows that the thermal difference between the weld and the surrounding area is significant. As soon as the heat starts to dissipate (section C - C), the base material and the weld starts to shrink and thermal stresses start to evolve. At the end of the weld (section D – D), when the cooling is completed and the temperature gradients are negligible, the residual stresses reach a maximum.
Stress relieving heat treatment is the most common method used for minimize the negative impact of RS on a material [10]. In the case of steels, the heat treatment involves heating the welded part to a temperature range of 550°C - 650°C, and holding the material at this temperature to achieve relaxation of internal stresses. In this temperature range, the grain size does not change and the steel does not soften [8][10].

If RSs are alleviated from a welded component, dimensional instability over a period of time can occur. This is a serious problem for high precision parts, where distortion and internal defects are not acceptable. The solution to the problem is a secondary heat treatment.

The ASM Handbook separates methods for measuring RS into five different groups [24]. The first group uses electrical and mechanical strain gauges attached to a material; the strain is then converted to a stress using Hooke's law. This group of techniques includes sectioning, Gunnert, Mother - Soete drilling and the Stäblein successive milling [24]. The second group of techniques uses a variety of coating methods and includes Grid system dividing, brittle coating drilling and Photo-elastic coating drilling [24]. The third group of techniques includes diffraction techniques based on X-rays and neutrons. The fourth group of techniques uses
materials with stress sensitive properties and includes Ultrasonic testing, polarized ultrasonic waves, ultrasonic attenuation and a Hardness method [24]. The fifth group of techniques includes “cracking” technique methods: hydrogen – induced cracking and the Stress-corrosion cracking [24]. Some of the methods provide only surface RS analysis, such as groups 1, 2 and 5.

2.3. Measurement of residual stresses with neutron diffraction method

Neutron diffraction has evolved for the past several decades and is now used in quality control, crack detection, characterization of materials, metallurgical grain size and texture analysis, phase purity measurements, analysis of crystalline and amorphous samples, and even biological samples [26]. It can also be used for RS analysis. The theory and methods of RS measurement in welds using ND have been published in the literature and are well accepted by industry [20][22].

ND methods for RS analysis were developed by a Canadian physicist Bertram Brockhouse in 1941. Neutrons are able to penetrate deep into a material compared to X-rays, which enable only a surface RS measurement [27]. Since neutrons have no charge, they do not damage the sample, and thus ND is considered a nondestructive method (since there is no need to destroy the sample to obtain a measurement). If the wavelength of the incident neutron beam is known and the diffracted angle can be measured precisely, then the residual strain can be calculated using Bragg's law.

ND residual stress determination is based on experimental measurement of the lattice spacing for a specified crystallographic plane, as related by Bragg's law (Equation 4) and Equation 5 [8]:

\[ d = \frac{
\lambda}{2 \sin \theta
} \]

\[ \varepsilon = -\frac{1}{2} \frac{d \Delta d}{d \sin \theta}
\]
\[ n\lambda = 2d \sin(\varphi/2) \]  \hspace{1cm} \text{Equation 4}

Where:

- \( n \) = order of reflection.
- \( \lambda \) = wavelength, Å
- \( d \) = distance between two adjacent crystallographic planes, Å
- \( \varphi \) = angle of diffraction, degrees

Rearranging for the distance between crystallographic planes yields:

\[ d = \frac{n\lambda}{2\sin(\varphi/2)} \]  \hspace{1cm} \text{Equation 5}

Elastic residual strain of the crystal is calculated using Equation 6:

\[ \varepsilon = \frac{\Delta d}{d_0} = \frac{d - d_0}{d_0} \]  \hspace{1cm} \text{Equation 6}

Where:

- \( d \) = distance between two adjacent crystallographic planes in a stressed material, Å
- \( d_0 \) = distance between two adjacent crystallographic planes in a stress-free material, Å
- \( \varepsilon \) = strain of the crystal

Stress in a particular crystallographic direction can be calculated from Hooke’s law given by Equation 7 [22]:

\[ \sigma_x = \frac{E}{1+\nu} \left[ \varepsilon_x + \frac{\nu}{1-2\nu} (\varepsilon_x + \varepsilon_y + \varepsilon_z) \right] \]  \hspace{1cm} \text{Equation 7}

Where:

- \( E \) = modulus of elasticity
- \( \nu \) = Poisson’s ratio
\( \varepsilon = \) strain of the crystal in particular direction \( x, y, z \)

\( \sigma = \) stress of the crystal in particular direction \( x, y, z \)

2.4. **ND equipment**

In order to carry out neutron diffraction strain measurement, a triple-axis spectrometer coupled to a nuclear reactor is required. The nuclear reactor produces a beam of neutrons of many incoherent wavelengths (i.e., a “white” beam of neutrons). The white beam is then directed towards a monochromator, where a single wavelength is extracted and a focused beam of monochromatic neutrons is then shaped using a collimator, as shown in Figure 5. This beam subsequently becomes the “incident beam” on the specimen being measured. The volume in the specimen where neutrons interact with the lattice structure of the specimen is called the “gage volume”, and depends on the shape of the incident beam, as seen in Figure 6.

In order to measure the elastic residual strain in the material at various locations, the stage holding the sample is translated in three orthogonal directions. As a result, the gage volume is fixed in space, but the sample is translated.

Once interaction between the specimen and the neutron beam occurs, diffracted neutron beam is generated and collected in the detector. The detector counts a predetermined number of neutron interactions (i.e., neutron count). Once this count is attained, the specimen is translated to a new position and the counting begins again.
Figure 5: Triple-axis ND residual strain measurement method [22].

Figure 6: Shape of gauge volume during ND [22].
Chapter 3. Experimental procedure

This chapter discusses the experimental procedures that were used for the experiments conducted in this research. ASTM standard procedures performed include: tensile test, Rockwell hardness test, impact test and the end-quench test. Heat treatment, tempering, optical microscopy, SEM/X-EDS, chemical analysis and neutron diffraction for residual strain measurement and microstructure evolution studies were also performed.

3.1. Material used

The chemical composition of the AR200 steel donated by Nor – Mar Industries (Penticton, BC) was provided by the material supplier and is listed in the mill certificate No. 061183745, issued on Dec 12, 2009 [18] as provided in Table 5.

Table 5: Chemical composition of AR200 steel used in this research.

<table>
<thead>
<tr>
<th>Element</th>
<th>AR200 mill certificate [18], wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>0.33</td>
</tr>
<tr>
<td>Mn</td>
<td>1.24</td>
</tr>
<tr>
<td>Si</td>
<td>0.22</td>
</tr>
<tr>
<td>Cr</td>
<td>0.09</td>
</tr>
<tr>
<td>Mo</td>
<td>0.03</td>
</tr>
<tr>
<td>V</td>
<td>0.007</td>
</tr>
<tr>
<td>Cu</td>
<td>0.30</td>
</tr>
<tr>
<td>Ni</td>
<td>0.12</td>
</tr>
<tr>
<td>P</td>
<td>0.012</td>
</tr>
<tr>
<td>S</td>
<td>0.001</td>
</tr>
</tbody>
</table>
In the case of AR200 steel with CE = 0.63 wt%, Table 2 suggests that formation of welding defects will be very likely. Also, the CrE = 0.45 wt% and NiE = 10.64 wt% suggest that the microstructure is predominantly martensitic (see Figure 3).

3.2. ASTM tests

To determine the characteristics of the virgin and welded alloys, a series of ASTM tests were performed. Testing was carried out in two stages:

Stage 1 - Material characterization

Material characterization was carried out via optical microscopy, neutron diffraction, electron microscopy analysis (SEM) and chemical X-ray energy dispersive spectroscopy (X-EDS) [41].

Stage 2 – Evaluation of mechanical properties

Mechanical properties were studied via tensile testing (ASTM E8) [31], hardness testing (ASTM E18) [32], impact testing (ASTM E23) [33] and Jominy end quench test (ASTM A255 [34], ASTM A304 [35]).

3.3. Material characterization

Material characterization evaluation involved quantitative microstructure analysis, chemical composition evaluation and the study of inclusions. Also, neutron diffraction experiments were performed to quantify the residual strain in the welded samples.
3.3.1. **Microstructure analysis**

The purpose of the microstructure analysis was to identify microconstituents, type and size of the crystal grains, formation of oxides, carbides and intermetallics. This analysis was important to understand the mechanical behavior of the studied AR200 steel. The microstructure analysis was carried out using an optical microscope and a Scanning Electron Microscope with Energy-dispersive X-ray spectrometer (SEM/X-EDS) located at UBC, Okanagan campus.

3.3.1.1. **Optical microscopy analysis**

Three 1.0” diameter 4” long rods of AR200 were cut half-way along their length and butt welded. Next, the samples were cut longitudinally, sanded, polished and etched with 5% natal to reveal the internal weld structure, as illustrated in Figure 7 [28] [29]. The purpose of the optical analysis was to determine the type and the size of grains and weld regions. During the observation, flaws and defects in the welded joint were also characterized.

![Figure 7: Cross section of welded sample with weld, HAZ and base-metal regions.](image-url)
3.3.1.2. Scanning Electron Microscopy with Energy-dispersive X-ray spectroscopy

Two types of samples were examined and observed with an SEM. The first type was the previously described welded parts shown in Figure 7. The second type of samples was cut out of the fractured ends of the tensile samples, as seen in Figure 8.

Prior to performing chemical analysis, the SEM/X-EDS was calibrated using a pure copper specimen. After calibration analysis of micro-defects and the presence of inclusions, oxides and intermetallics compounds in the weld regions was carried out.

![Fractured ends of tensile test samples.](image)

**Figure 8:** Fractured ends of tensile test samples.

3.3.2. Measurement of residual stresses via neutron diffraction technique

Five samples of AR200 were prepared from 3” wide “C” channels. These samples were representative of the welded joints used to support a crane structure currently used by Nor – Mar Industries. Each sample consisted of two welded segments: one horizontal 7” long channel and one vertical 3” long channel, as seen in Figure 9. The sections were welded using GMAW and subsequently subjected to a torch heat treatment at 600°C, as seen in Figure 10. Samples with varying duration of the torch heat treatment were prepared, as shown in Table 6. During the
manual heat treatment process, every effort was made to ensure repeatable results, by trying to keep a constant distance between the torch and the sample as well as monitoring the heat treatment duration.

**Table 6:** HT schedule for ND samples.

<table>
<thead>
<tr>
<th>Sample #</th>
<th>HT duration, min</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Not HT</td>
</tr>
<tr>
<td>2</td>
<td>3</td>
</tr>
<tr>
<td>3</td>
<td>6</td>
</tr>
<tr>
<td>4</td>
<td>9</td>
</tr>
<tr>
<td>5</td>
<td>12</td>
</tr>
</tbody>
</table>

**Figure 9:** Samples prepared for residual stress measurement with ND.
ND measurements were carried out at the Canadian Nuclear Laboratories, in Chalk River, Ontario. Extraction of stress-free samples was necessary in order to determine the crystallographic distance $d_0$, which corresponds to the lattice spacing of a material when no load is applied. To obtain the $d_0$ value, four match-stick samples were cut from the “C” channels, as shown in Figure 11.

Figure 10: Torch heat treatment of the samples.

Figure 11: Stress free samples cut from “C” channel.
Prior to mapping and measuring the residual strain in the welded “C” channels, the coordinates of the sample were determined using tilted and level telescopes. Then, the first line to scan was programmed into diffractometer software. Line A75 along the welding seam was programmed, as shown in Figure 12, followed by the second line (Line A80) above the welding seam. The third line to scan was Line B which was 13.5 mm above the welding seam, followed by the last scan (Line C) in the direction perpendicular the weld. All samples of welded “C” channels were scanned in this pattern. For all lines, a minimum of 10 measurements was obtained. At each location, a gauge volume 2 mm × 2 mm × 3 mm was used.

Figure 12: Coordinates for scanning and lattice orientation.

Scanning along the X, Y and Z axis required the sample to be oriented in three different positions as shown in Figure 13 - Figure 15.
Figure 13: Sample orientation for x-strain scanning.

Figure 14: Sample orientation for y-strain scanning.
All five samples were scanned along Line A75, Line A80, Line B and Line C in three orthogonal directions X, Y and Z. The (311) crystallographic plane was used for all measurements. The wavelength of the neutron beam was $\lambda=1.55\,\text{Å}$. After crystallographic distances $d$ and $d_0$ were as measured, the magnitude of the residual elastic strain was determined from Equation 4 and Equation 6.

3.4. Mechanical testing

Classical mechanical testing was performed to determine the Ultimate Tensile Strength (UTS), Yield strength (YS), modulus of elasticity (E), percent elongation (l%), percent area reduction (A%), hardness (HRC) and impact strength of virgin and welded sections.
3.4.1. Tensile test

The tensile test was carried out according to ASTM E8 standard procedure, with 0.5” diameter round sample [30][31]. A 10” long and 1.0” round stock of AR200 steel was cut to prepare an 8” long standard sample size, as seen in Figure 16. Then, the stock bar was turned on a CNC lathe to a standard size as specified by the ASTM E8 standard. The gauge area was machined with +0.05” allowance for further final machining at the end of sample preparation. Then, the samples were cut in the middle on the lathe, as seen in Figure 17.

![Figure 16: Standard tensile test sample dimensions as per ASTM E8 [31].](image)

Traditionally, butt welds are commonly used in the industry to carry out fusion welding operations. In this work, additional weld joint types were examined, as illustrated in Figure 18.

![Figure 17: Machined tensile samples ready for welding.](image)
below. These different weld joint geometries were examined due to their industrial relevance. The FCAW butt joint represented a classical butt joint, but the welding operation was carried out with a flux-core welding wire, instead of a traditional welding wire.

Figure 18: Types of end preparation on tensile samples.

The two halves of the tensile sample were welded by GMAW using a Miller Invision 354 MP machine with operating voltage of 22.5V. The current was between 162 – 175A, positive (straight) polarity DCSP. A Lincoln electric AWS ER70S-6 diameter 0.035” (0.889mm) wire was used. The shielding gas was “blue shield” #6 consisting of 90% Ar and 10% CO₂ [47].

After welding, the tensile samples were heat treated with a torch, as seen in Figure 19, to a specific temperature for a specific amount of time. This procedure was performed according to the existing industry practice, as outlined in Table 7. As will be discussed in the following text, a second set of AR200 samples was also heat treated, but using traditional methods (i.e., using a muffle furnace).
The temperature and time during the torch heat treatment were monitored using a thermocouple attached to the tensile sample 1.75” away from the weld. The thermocouple was connected to a data acquisition unit OMEGA OMB-DAQ-56 and time - temperature profiles of each heat treated sample were recorded. After heat treating with a torch, the samples were fine machined in the gauge area of the sample to meet the surface finish requirements as per ASTM E8. Tensile testing of the welded and heat treated samples was performed on a SATEC model 120HVL tensile testing machine, as seen in Figure 20.
Table 7: Heat treatment schedule for tensile samples.

<table>
<thead>
<tr>
<th>Sample #</th>
<th>Welded</th>
<th>Defects found</th>
<th>Heat treatment applied after welding</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>No</td>
<td>No</td>
<td>No HT</td>
</tr>
<tr>
<td>2</td>
<td>No</td>
<td>No</td>
<td>No HT</td>
</tr>
<tr>
<td>3</td>
<td>No</td>
<td>No</td>
<td>870°C 2 min, air cooled</td>
</tr>
<tr>
<td>4</td>
<td>No</td>
<td>No</td>
<td>870°C 2 min, air cooled</td>
</tr>
<tr>
<td>5</td>
<td>Yes</td>
<td>Yes</td>
<td>No HT</td>
</tr>
<tr>
<td>6</td>
<td>Yes</td>
<td>Yes</td>
<td>No HT</td>
</tr>
<tr>
<td>7</td>
<td>Yes</td>
<td>Yes</td>
<td>870°C 2 min, air cooled</td>
</tr>
<tr>
<td>8</td>
<td>Yes</td>
<td>Yes</td>
<td>870°C 2 min, air cooled</td>
</tr>
<tr>
<td>9</td>
<td>Yes</td>
<td>Yes</td>
<td>No HT</td>
</tr>
<tr>
<td>10</td>
<td>Yes</td>
<td>Yes</td>
<td>700°C 2 min, air cooled</td>
</tr>
<tr>
<td>11/1</td>
<td>Yes</td>
<td>Yes</td>
<td>600°C 2 min, air cooled</td>
</tr>
<tr>
<td>11/2</td>
<td>Yes</td>
<td>Yes</td>
<td>600°C 2 min, air cooled</td>
</tr>
<tr>
<td>12</td>
<td>Yes</td>
<td>No</td>
<td>500°C 2 min, air cooled</td>
</tr>
<tr>
<td>13</td>
<td>Yes</td>
<td>No</td>
<td>400°C 2 min, air cooled</td>
</tr>
<tr>
<td>14</td>
<td>Yes</td>
<td>Yes</td>
<td>400°C 3 min, air cooled</td>
</tr>
<tr>
<td>15</td>
<td>Yes</td>
<td>Yes</td>
<td>800°C 4 min, air cooled</td>
</tr>
<tr>
<td>16</td>
<td>Yes</td>
<td>Yes</td>
<td>700°C 4 min, air cooled</td>
</tr>
<tr>
<td>17</td>
<td>Yes</td>
<td>Yes</td>
<td>600°C 4 min, air cooled</td>
</tr>
<tr>
<td>18</td>
<td>Yes</td>
<td>Yes</td>
<td>500°C 4 min, air cooled</td>
</tr>
<tr>
<td>19</td>
<td>Yes</td>
<td>Yes</td>
<td>400°C 4 min, air cooled</td>
</tr>
<tr>
<td>20</td>
<td>Yes</td>
<td>Yes</td>
<td>400°C 5 min, air cooled</td>
</tr>
<tr>
<td>22</td>
<td>Yes</td>
<td>Yes</td>
<td>Welded TIG (No HT)</td>
</tr>
</tbody>
</table>
Figure 20: Universal testing machine SATEC 120HVL.

Elongation of the samples during a tensile test was measured with an extensometer, Epsilon 35-42-0200-005-ST, attached to the gauge area of the sample, as seen in Figure 21.

Figure 21: Extensometer Epsilon model 35-42-0200-005-ST.
In order to produce the stress-strain diagrams, an ADMET digital controller connected to a computer was used. The pressure gauge of the tensile machine, which converts hydraulic pressure to an electric signal was attached to the controller. The controller recorded the pressure and displacement simultaneously.

The slope of the linear part of the stress – strain diagram was used to calculate the modulus of elasticity, E. The cordial method recommended by ASTM E8 was used for this measurement. The calculated results were sensitive to the selection of the location of the points on the linear part of stress – strain diagram. The two points selected for calculation were always the same value of 20 ksi and 35 ksi.

3.4.2. **Hardness test**

A Wilson VH3100 microhardness tester was used for Vickers hardness mapping across the weld regions. Hardness measurements were performed on the 1.0” round stock samples, which were also used for optical analysis. A grid pattern was programmed into the VH3100 software, as shown in Figure 22. The distance between indentations was 1 mm, and the separation between the lines was 5 mm.
3.4.3. Impact test

ASTM E23 [33] impact testing was carried out to determine and quantify the material’s impact strength, notch sensitivity and the Ductile to Brittle Transition Temperature (DBTT).

AR200 samples were cut to 55mm × 10mm × 10mm with 0.5mm to 0.7mm machine allowance, as shown in Figure 23. The samples were face milled on all four sides to obtain a square cross-section. At this point, samples were oversized by 0.13mm - 0.18mm to allow subsequent grinding finishing operation.
Figure 23: Standard impact test sample dimensions as per ASTM E23 [33].

After rough machining, the samples were finished on a surface grinder to meet the ASTM E23 requirements for the dimensional accuracy and surface finish. The V-notch was cut using a vertical mill and the depth and angle of the notch were cut to meet ASTM E23 requirements. Charpy V-notch impact testing was conducted with SATEC Impact Tester Model No. S1-1C, as seen in Figure 24 a). After the impact test, the samples were visually inspected and the types of the fracture were qualitatively identified either as brittle or ductile, as illustrated Figure 24 b) and c).
3.4.4. **Heat treatment test**

The purpose of the heat treatment test was to evaluate the effect of heat treatment temperature and time on the microstructure and the hardness of the AR200 steel. Ten samples were cut from raw stock material used for preparation of the tensile samples. Samples were heated to 870°C using the laboratory furnace LMF-3550/120, seen in Figure 25. Seven of these samples were quenched in water, one was air cooled, while two were furnace cooled. Six of the seven quenched samples were subsequently tempered to different temperatures ranging from 200°C to 700°C in steps of 100°C for 2 h. All of the six samples were air cooled, with details of the heat treatment in Table 8.
Figure 25: Heating the samples in laboratory furnace.

Table 8: Temperature and time schedule of heat treatment test.

<table>
<thead>
<tr>
<th>Sample #</th>
<th>HT applied</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Quenched</td>
</tr>
<tr>
<td>2</td>
<td>Tempered to 200°C for 2 h</td>
</tr>
<tr>
<td>3</td>
<td>Tempered to 300°C for 2 h</td>
</tr>
<tr>
<td>4</td>
<td>Tempered to 400°C for 2 h</td>
</tr>
<tr>
<td>5</td>
<td>Tempered to 500°C for 2 h</td>
</tr>
<tr>
<td>6</td>
<td>Tempered to 600°C for 2 h</td>
</tr>
<tr>
<td>7</td>
<td>Tempered to 700°C for 2 h</td>
</tr>
<tr>
<td>8</td>
<td>Normalized</td>
</tr>
<tr>
<td>9</td>
<td>Annealed</td>
</tr>
<tr>
<td>10</td>
<td>As fabricated</td>
</tr>
</tbody>
</table>

Observing the iron-carbon phase diagram in Figure 26 [44] for 0.35wt%C and a temperature range from 200°C - 700°C, the phases present are ~95wt% Ferrite and ~5wt% Pearlite. Ferrite is soft and ductile, while Pearlite is hard and brittle. Under equilibrium cooling conditions (i.e., slow cooling) these phases were expected to form in the heat treated samples.
After tempering, the samples were ground with a SiC paper of 240, 400, 600, 800 and 1200 grit. After grinding, the samples were polished with a diamond suspension 1µm on a medium nap cloth disk. To prepare the samples for microscopic observation, all samples were etched with 5%Nital, as seen in Figure 27 [28] [29].

After the samples were etched, they were examined using a ZEISS Axio Observer.A1m inverted microscope. The hardness was converted to Brinell scale to plot the tempering temperature vs hardness.

Figure 26: Steel section of carbon – iron phase transformation diagram [8].
3.4.5. **End quench test**

An end quench test, also known as the “Jominy” test, was used to study the hardenability of the AR200 steel. Two samples of 1.0” round stock by 4” long were placed in a furnace at 780ºC for 2 h. The samples were then pulled out of the furnace and subjected to a 20ºC stream of water at the bottom end, as seen in Figure 28. The hardness of the sample was measured in three lines longitudinally spaced 3 mm apart and 2 mm from the edge, as shown in Figure 29. The test was performed following recommendations of ASTM A255 and ASTM A304, with measurements taken every 1.5 mm for the first 25 mm distance, and every 3 mm up to 50 mm distance from the quenched end.
Figure 28: End quench test.

Figure 29: End quench test hardness measurement.
Chapter 4. Results and Discussion

In this chapter, the results obtained during the thesis research are presented and critically discussed. The results of mechanical testing and materials characterization are presented for the AR200 samples prepared under various conditions. Also, the results of residual strain measurement via neutron diffraction are presented.

4.1. Microstructure analysis

The following sections provide the results of optical microscopy and SEM observations for the AR200 steel samples prepared under different conditions.

4.1.1. Optical microscopy

4.1.1.1. Microstructure of tempered samples

The AR200 alloy is a hypoeutectoid steel. According to the phase diagram, at 0.35wt%C, the phases present in AR200 should be ~95wt% ferrite and ~5wt% pearlite. Ferrite is a soft and ductile phase, while pearlite is hard and brittle. The AR200 steel samples were tempered (as outlined in Section 3.4.4.), and the effect of the tempering process on the evolution of ferrite and pearlite is shown in Figure 30.
Observing the microstructure of Sample #1 (quenched), the microstructure consisted of martensite, ferrite and patches of pearlite. This combination of phases is commonly referred to as bainite. Martensite forms during rapid quenching and since it is a hard and brittle phase, it significantly contributes to the hardness of the alloy. Based on the distribution of the martensite, the AR200’s as quenched hardness would be expected to be in the moderate range, since the martensite crystals were relatively uniformly distributed throughout the ferrite matrix.

When the AR200 was quenched and tempered to 200°C for 2 h in the oven (sample #2), the observed martensite and ferrite phases began to fragment, but the martensite remained surrounded by the soft ferritic matrix. This trend of fragmenting martensite continued with increasing tempering temperature (samples #3, #4, #5 and #6), which resulted in a continuously
decreasing hardness of the alloy. For sample #6 tempered to 600°C for 2 hours, the martensite crystals continued to fracture, while at 700°C the martensite crystals dissolved and only fine pearlite and ferrite crystals remained.

Sample #8, which was normalized, had a microstructure consisting of fine pearlite and ferrite crystals only. Normalizing softens the steel, but yields a material that is harder than in the case of annealing. Sample #9 was annealed and contained big coarse pearlite and ferrite grains, which resulted in the lowest hardness, as will be discussed in following sections.

The microstructures observed in Figure 30 were created during rapid cooling of the AR200 steel samples from their respective heat treatment temperatures. To verify the validity of the observed phase, the phase evolution under non-equilibrium cooling conditions was examined using a Time Temperature Transformation (TTT) diagram, as seen in Figure 31. The average cooling rate for quenching was assumed to be 150°C/s [36][37]. The samples used in this study cooled from their heat treatment temperatures to room temperature (in agitated quenching water) within approximately 6 seconds. According to the TTT diagram, such cooling rate should cross the ferrite-start and bainite lines. Thus, the experimentally observed microstructures are consistent with microstructure based on thermodynamical models of phases in steels. Thus, the AR200 steel responded to heat treatment as expected.
4.1.1.2. Microstructure of welded samples

Characterization of the microstructure in welded samples was carried using butt welded joints discussed in section 4.2.2. The microstructure of the base AR200 is shown in Figure 32 and consisted of a coarse pearlite and ferrite. This type of microstructure is soft and ductile and was generated by the slow cooling of the base AR200 material.
Figure 32: Microstructure of welded joint away from the weld.

Figure 33 shows the microstructure at the welded joint. The microstructure contained fine pearlite and ferrite grains, with patches of martensite. The grains were fine due to the rapid solidification of the weld metal in the welding seam. This microstructure resulted in a harder material, as will be discussed in section 4.2.2.

Figure 34 shows the microstructure in the HAZ region. This zone exhibited a transition in the grain structure, as seen in Figure 35 on a large scale. The grain variation likely contributed to the variation of hardness in this region. The difference in grain size was likely related to the thermal profile of the material during cooling to room temperature [36].
Figure 33: Microstructure of joint in the weld.

Figure 34: Microstructure of welded joint in the transition zone between weld and HAZ.
Figure 35: Microstructure of welded joint in the transition zone.

The images in Figure 36 - Figure 38 show the relative size of the microstructure with respect to micro-indentations used for the measurement of the alloy’s hardness. As seen in the figures, the indentations spanned multiple grains in the respective regions (i.e., weld material, HAZ and base metal), thus confirming that the hardness values were affected by both the grain as well as precipitate constituents.
**Figure 36:** Microstructure of the base metal zone.

**Figure 37:** Microstructure of HAZ.
4.1.1.3. Microstructure of weld defects

Figure 39 and Figure 40 show representative defects which were frequently observed in the vicinity of the weld seam. These defects are commonly referred to as “folds” and are generally caused by a low heat input into the weld seam during the welding process. This low heat input causes poor heat penetration and low melt-through leading to insufficient fusion of liquid metal fronts. These defects often contribute to poor mechanical strength of welded joints, due to the formation of stress concentrations at the fold extremities, as well as physical discontinuity in the material (i.e., a void in the material).
Figure 39: Low melt-through defect in the junction.

Figure 40: Poor melt-through in the weld seam.

Figure 41 shows representative images of porosity defects, which mainly formed due to the entrapment of welding gases in the molten metal, since the shape of the pores was spherical.
or elliptical with round edges. However, several of the pores were also sharp with irregular shape curvature, suggesting that such pores may have formed due to solidification shrinkage of the liquid weld metal during cooling.

![Figure 41: Porosity in welding joint at the transition zone.](image)

4.1.2. SEM/X-EDS analysis of fractured tensile samples

Scanning electron microscopy was used to examine the fracture surfaces of the tensile specimens discussed in section 4.2.1. Representative tensile samples were examined. Specifically, Sample #1 (solid non-welded) and Sample #5 (welded and heat treated) were studied, as they exhibited representative behavior observed in all of the remaining tensile samples. Also, SEM microscopy was used to characterize the welding wire used during welding operations, in order to examine the possibility of weld contamination from the welding wire.

The solid (non-welded) AR200 sample fractured with a typical cup and cone fracture mode, as seen in Figure 42. The core section of the tensile sample contained voids / pores, which
likely acted as stress concentration sites and assisted in crack nucleation and material failure. Also, Figure 42 shows that an edge region fractured at an angle relative to the core region. This fracture type is consistent with the evolution of the critical resolved shear stress at the 45° angle with respect to the loading direction. It is of interest to note that the sample core region had a dimpled surface, suggesting that some level of ductility was accommodated by the material.

Figure 42: Fracture surface of non-welded sample.

Figure 43 shows a representative fracture surface from a welded specimen. In this specimen, the ductile core region was not observed. Further, the cup and cone fracture was also not evident. Instead, formation of cracks in the vicinity of intermetallic compound inclusions was
observed. These intermetallic compounds had irregular shape and sharp edges, which would induce stress concentrations in the tensile sample and contribute to material failure.

Thus, the presence of the intermetallic compounds has likely contributed to the lower tensile strength, as will be discussed in section 4.2.1. Further, the presence of intermetallic compounds in the welded samples would support the observed insensitivity of the tensile samples to heat treatment temperature or heat treatment time. As can be seen in Figure 43, cracks readily formed at the intermetallics, suggesting that the matrix material did not experience sufficient plastic loading during tensile testing. As a result, the matrix material’s properties did not significantly contribute to the tensile performance of the material. Thus, the effect of the heat treatment temperature and time (which influence the matrix microstructure) was not pronounced.

**Figure 43:** Fracture surface of welded heat treated.
Figure 44 shows a closer view of the intermetallic compound observed in the welded tensile samples. SEM-XEDS chemical analysis was conducted on the intermetallic particle and the results are plotted in Figure 45.

Figure 44: Intermetallic inclusion in the welded sample.

Figure 45: Chemical analysis of intermetallic compound.
The chemical analysis revealed that the intermetallic consisted of 4.4 wt% C, 1.6 wt%Si, 0.2 wt%Ti, 29.9 wt%Mn, 30.9 wt%Fe and 31.4 wt%O. Interestingly, oxygen was present at significant amount, suggesting that the particle was an oxide. Ingress of oxygen into the weld could be from the surrounding air or from the decomposition of the shielding CO$_2$ gas. The oxygen would react with the elements present in the steel alloy to readily form iron oxide and/or manganese oxide. Formation of these oxides is known to reduce the quality of the steel [4].

Figure 46 shows a detail of the fracture surface of tensile specimen #6. In this fracture surface, the typical ductile fracture region with dimpled deformation of the matrix grains was observed, with fine particles located in the dimples. Chemical analysis of these particles revealed that they consisted of 19.8 wt%Mn, 57.9 wt%Fe and 22.3 wt%O. Thus, the particles were likely manganese oxide compounds. In general, dispersion of fine particles in a material’s matrix results in strengthening and improved mechanical strength [8]; however, in this case it was not clear whether these particles were present in the virgin AR200 material, or if they were introduced during the welding operation. As a result, the welding wire which was used to carry out all welding operations was sectioned and examined under SEM-XEDS.

Figure 47 shows the cross section of the welding wire used throughout the welding experiments. The chemical analysis revealed that it contained: 10.9 wt%C, 0.7 wt%Al, 1.2 wt%Si, 0.2 wt%S, 1.4 wt%Mn, 80 wt%Fe and 5.5 wt%Sn. There was no oxygen observed in the wire, but manganese and other elements were present. Therefore, the manganese particles could be present as a result of the 1.3 wt% Mn in the AR200 steel, or due to the additional 1.4 wt% Mn originating from the wire. In addition to the presence of high levels of manganese, sulphur was also present in the welding wire. Sulphates are known to readily form during the welding process.
and generally contribute to lowering the quality of the weld and the tensile strength of the welded joint [8].

**Figure 46:** Sample #6 fractured area.

**Figure 47:** Welding wire chemical analysis.
4.1.3. Residual strain measurement

The following section presents the results of residual strain measurement, rather than residual stress measurement. Residual stress could not be accurately measured due to the difficulty in determining the value of $d_0$ in the three distinct regions of the weld zone. Further, information about the elastic modulus (E) of the weld regions was not known.

Figure 48 to Figure 50 show the distribution of residual strain in the x-direction measured along the “A75” linescan. As can be seen, the strain profile had a U-shaped curvature. The increase in strain magnitude at the ends was the result of connecting welds (at the ends) of the C-section.

It is evident that performing the heat treatment on the thin-wall C-section component has homogenized the residual strain in the weld zone. The magnitude of the strain near the ends has decreased, while the magnitude at the center of the weld cross section has increased (in order to maintain static equilibrium). Reduction of stress concentrations is a desirable effect of heat treatment procedures. Further, it was observed that that the x- and y-direction strains were predominantly tensile, while the z-direction strain was compressive. This localization and directionality of residual strain suggests that significant strain gradients may evolve in welded sections, highlighting the requirement to be aware of weld material anisotropy during in-service loading conditions.

The effect of the duration of the heat treatment on the magnitude of the residual strains was not statistically significant. According to the average residual strain values, the strain relaxation for the 12-min heat treated samples was more significant than for the 3-min heat treated samples. However, considering the error bars of the strain measurements, the effect of heat treatment duration could not be considered to be statistically significant. This result is
consistent with results in section 4.2.1 where the YS, UTS and E also showed insensitivity to the heat treatment duration.

**Figure 48:** Strain scan on line A75 in x-direction.

**Figure 49:** Strain scan on line A75 in y-direction.
In the case of the $z$-direction strain, plotted in Figure 50, it is evident that the residual strain was not symmetrical. The residual strain at the left of the C-section started as compressive, while on the right side it started as tensile. This trend was correlated to the physical shape of the welded section. As was explained in Section 3.3.2, and shown in Figure 9, the vertical part of the welded C-channel had one of the flanges removed. Consequently, there was only one weld present. As a result, this geometry and the associated thermal strain inhomogeneity (due to welding of only one segment of the vertical C-channel), likely contributed to a significant variation of the residual strain along the $z$-direction.

Comparing Figure 48 - Figure 50, interestingly, the residual strains in the welded sections were predominantly tensile. If the combined effect of the residual strain tension generates a significant tensile residual stress, then a crack may readily form and the weld fails. Further, the tensile residual strain may lead to distortion of the welded sections, causing
dimensional inaccuracy. Regrettably, the heat treatment did not provide sufficient energy to homogenize and relax the tensile residual strain.

Examining the magnitude of residual strain in the vicinity of the weld (5 mm above the weld line) in Figure 51 to Figure 53, revealed that the magnitude of the residual strain remained at the same level as at the weld itself. This was likely the result of the relative size of the heat affected zone (it was greater than the location of line A75, which was 5 mm from the weld). As seen in the hardness profiling (Section 4.2.5, Figure 81), the width of the weld zone and heat affected zone was nearly 30 mm wide. Therefore, the thermal strain induced by the weld was transferred into the vicinity of the weld zone, which included the region 5 mm far from the weld.

As can be seen in Figure 51 to Figure 53, the effect of the heat treatment at linescan “A80” region became obvious, as the profiles corresponding to the different heat treatment durations have separated beyond the overlap of the error bars. In particular, the effect of heat treatment was clearly visible in the z-direction residual strain profile (Figure 53). With an increase in the heat treatment time, the uniformity of the residual strain increased, and the weld line end-points were the only segments of the weld which contained a significant compressive strain. Elsewhere, the residual strain became uniformly tensile. The general trends observed for linescan “A80” were similar to those of linescan “A75”.

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**Figure 51**: Strain scan on line A80 in x-direction.

**Figure 52**: Strain scan on line A80 in y-direction.
Figure 53: Strain scan on line A80 in z-direction.

Figure 54 to Figure 56 show the residual strain profile at 13 mm above the weld line. In this region, it was noted that the effect of the heat treatment was significant. In the case of the z-direction strain, the heat treatment decreased the magnitude of the strain. However, the strain energy was translated into deforming x- and y-directions of the material, since these directions experienced a significant increase of the residual strain. These results suggest that the welded sections experienced non-uniform deformation and distortions likely occurring as a result of the post-welding heat treatment process. However, the heat treatment duration did not have a statistically significant effect on the residual strain in all three directions.
Figure 54: Strain scan on line B in x-direction.

Figure 55: Strain scan on line B in y-direction.
Figure 56: Strain scan on line B in z-direction.

Figure 57 to Figure 59 shows the residual strain distribution along a linescan perpendicular to the welded joint. These results confirm that the weld created a significant discontinuity in the material’s residual strain distribution.

The weld region (from 17 mm – 28mm) was under compression in the x-direction, but was under tension in the y- and z-directions. This trend was necessary to maintain static force equilibrium of the part. However, it was also observed that on either side of the weld region, the strain state has reversed. In the case of the x-direction strain, the compressive strain in the weld region has reversed into tension on either side of the weld material. Corresponding trends were observed in the case of the y- and z-directions.

This strain state reveals that as the liquid weld solidified (a phase transformation from the liquid state to the solid state), the volumetric shrinkage of the liquid metal has “pulled” on the AR200 steel material in the vicinity of the weld. This pull has induced tension in the surrounding
HAZ material. It is important to note, that the magnitude of the strain gradient observed in the profiles was not sufficient to nucleate cracks or any other tears (e.g., hot tears).

**Figure 57:** Strain scan on line C in x-direction.

**Figure 58:** Strain scan on line C in y-direction.
4.2. **Mechanical testing and behavior**

The following section provides the results of tensile tests, hardness and hardenability, followed by in-depth microstructure characterization.

4.2.1. **Tensile test**

Upon conducting tensile tests, the initial visual inspection revealed that all the welded samples did not fracture in the typical cup and cone fracture mode, as was observed for solid (not welded) AR200 samples shown in Figure 60 a). Instead, the welded samples fractured at 45° angle, as seen in Figure 60 b). The presence of oxides and intermetallics inclusions (slag) along the 45° fracture surfaces was the likely reason for fracture nucleation and propagation [40]. A detailed chemical analysis of these oxides was presented in section 4.1.2.
Figure 60: Tensile sample fractures.

Figure 61 shows the results of stress-strain testing. In Figure 61, the solid portion of the lines was based on direct measurement of strain using an extensometer, while the dashed portion of the lines was generated using the displacement of the gantry of the tensile testing machine.

Close examination of Figure 61 revealed that the behavior of the tested samples may be divided into three groups. In the first group, the samples exhibited high ductility with strain between $\varepsilon = 0.28 – 0.35$. Samples in this group consisted of virgin (not welded) AR200 alloy.

The second group of materials had a ductility with a fracture strain between $\varepsilon = 0.15 – 0.20$ (Samples #7, 9 and 22). These samples were seen to contain macro defects likely resulting from the welding process. The third group of samples exhibited highly brittle behavior (Samples #14, #15, #19 and #20), with a fracture strain of only $\varepsilon = 0.055 - 0.09$ [42].
Figure 61: Stress – strain results of butt-welded and heat treated samples.

Legend for Figure 61: **Number** of the sample/heating **temperature-time** for heating/Welded/Defect observed/UTS observed/YS observed/Modulus of elasticity.

Figure 61 shows that the ultimate strength (UTS) of the welded samples varied from 64 ksi (443 MPa) to 74 ksi (516 MPa), while the yield strength (YS) varied from 40 ksi (279 MPa) to 54 ksi (379 MPa). Since the lowest yield strength observed was 40 ksi (279 MPa), general engineering design guidelines for statically loaded applications suggest that a factor of safety of 2 should be used [38]. Thus, an allowable (design) stress for welded AR200 sections based on these results should not exceed 20 ksi (139 MPa).
The variation in UTS and YS was seen to be the result of micro and macro defects present in the welds. Intermetallic particles with irregular shape were seen to act as stress concentrations in the tensile test specimens. Presence of these inclusions typically induces double and triple yielding of the material in the yield zone, as was observed in Figure 62.

![Figure 62: Detail of yielding for samples with intermetallics.](image)

In addition to the butt-welded tensile samples, a second set of tensile samples with ring end preparation were also tested. The results of this second series of tensile tests are presented in Figure 63. The stress – strain curves for the tested samples exhibit fracture strain between $\varepsilon = 0.05 - 0.1$. Thus, these samples exhibit nearly four times less ductility compared to the first set of samples. Examination of the welded regions revealed inconsistency of the welded material in the gauge length zone. Specifically, a high volume fraction of intermetallic compounds, which likely affected the tensile performance of the welded samples were noted.
From the two diagrams examined, it can be concluded that the values of UTS, YS and E are strongly dependent on the amount of defects or intermetallic inclusions in the tested sample.

Figure 63: Stress – strain diagram for ring-end weld preparation samples.

4.2.1.1. Effect of heat treatment temperature and time on UTS

The effect of heat treatment time and temperature on the ultimate tensile strength of AR200 steel was examined in detail using the values obtained from the stress-strain diagrams. In the case of tests which include only two data points, the results are plotted for information purposes only. In the case of tests where more than three data points were available, a trend-line was fitted through the data points in order to help extrapolate and identify any trends.
Figure 64 shows the effect of the heat treatment temperature on the UTS. As can be seen from Figure 64, for the 4 min heat treatment durations, the heat treatment temperature did not have a significant effect on the UTS. Thus, the manual torch heat treatment was not enough to significantly alter the microstructure of the AR200 steel. Specifically, the torch heat treatment was possibly insufficient to transform the microstructure or grain structure of the AR200 steel, as was also discussed earlier (section 4.2.4).

![Figure 64: TS vs. tempering temperature 4 min heat treatment.](image)

Figure 65 shows the effect of the heat treatment time on the tensile strength of AR200. At 600°C heat treatment temperature, the UTS decreased from 69 ksi to 58 ksi. At 870°C heat treatment temperature, the UTS appeared unaffected by the duration of the heat treatment.
4.2.1.2. Effect of heat treatment temperature and time on YS

Figure 66 shows the effect of the tempering temperature on the yield strength of the AR200 steel. The time for heat treatment initially was selected at 4 min with the expectation to have a significant effect on the YS. After conducting the tensile tests, it was observed that this tempering time did not have a significant effect on the yield strength, as seen in Figure 66.
Plots in Figure 67 show that at 600°C and 780°C heat treatment temperatures, the heat treatment time did not have a significant effect on the yield strength. At 600°C, the microstructure remained as coarse pearlite and it was expected that with extended hold time, possibly recrystallization would take place, thus impacting the yield strength. However, this trend was not observed in the present study.

At 780°C heat treatment temperature, the AR200’s microstructure was that of austenite, which is softer than the pearlitic phase. Interestingly, the yield strength has not significantly decreased with the evolution of the austenite.
4.2.1.3. Effect of heat treatment temperature and time on E

The results in Figure 68 reveal that the modulus of elasticity varied between $28 \times 10^6$ psi to $33 \times 10^6$ psi. Therefore, within the heat treatment temperature range investigated (400°C - 870°C) and times, there was no significant impact of heat treatment on the elastic modulus of the AR200 steel.

Figure 68: E vs. heat treatment time.
The results of mechanical testing on heat treated AR200 steel samples have an important practical implication. The torch heat treatment, which was performed according to industrial practice was not seen to have a significant impact on the yield strength, ultimate tensile strength or the modulus of elasticity. One possible factor may be the highly localized heat input of the torch on the material. With such localized heat treatment, microstructure transformation is achieved only locally. As a result, the full material sample is not homogeneous and will fail based on the weakest microconstituent in the material. As a result, although the center section of the tensile samples was heat treated with the torch, the remaining portion of the gage area has retained the original as-rolled microstructure. It was frequently observed that failure occurred in the vicinity of the heat treated region, suggesting that the heat treated region itself was not the weakest region of the material.

An industrial implication of these results suggests that localized post-weld heat treatment with a torch is not a viable technique to ensure weld quality and material homogeneity. Thus, although it is economically attractive, a more technologically advanced heat treatment process should be implemented (e.g., heat treatment in a box oven).

4.2.1.4. Effect of weld geometry preparation on UTS, YS and E

Figure 18 showed the four conditions of weld preparation geometries investigated in this research.

As can be seen from Figure 69 to Figure 71, the butt joint resulted in the most desirable combination of mechanical response (i.e., YS, UTS and E) in comparison to other end preparation geometries. In the butt joint case, the gap between the adjoining sections was minimal, thus the possibility for ingress of impurities or welding defects was reduced. The
second highest tensile strength was achieved with a ring end preparation, where the space between the adjoining sections increased, and thus the surface area of liquid metal exposed to ambient environment and oxygen has increased. Consequently, the amount of defects would increase. In the case of the Flux Core Arc Welding (FCAW) method, the welding wire contained 1wt% chromium, which generally increases the strength of the weld [47]. However, in this study the FCAW method did not yield improvement in tensile strength, which was comparable to the last preparation geometry, the beveled end joint. In the beveled end joint geometry, the volume of the molten metal exposed to the ambient environment was the greatest, thus having a significant chance for formation of oxides. Further, the amount of liquid metal required to fill the void created by the beveled geometry was relatively large, resulting in a significant thermal gradients, which were likely associated with thermal strain gradients. Consequently, the beveled joint yielded the weakest weld joint.

![Tensile strength - all tensile samples](image)

**Figure 69:** UTS of different types of joints.
**Figure 70:** YS of different types of joints.

**Figure 71:** E of different types of joints.
4.2.2. **Hardness test**

The results of mechanical testing presented in the previous sections provide information about the welded material's bulk performance. However, it is known that welded sections exhibit localized variation in mechanical properties in the welded zone, heat affected zone and the base material. Thus, mapping across these regions was performed to study any localized variation of the hardness.

Figure 72 (repeat of Figure 22) shows the grid used for mapping the hardness profile, along with representative micrographs showing the general microstructure. Figure 72 shows that the weld region was not perfectly symmetrical.

![Figure 72: Hardness test with microstructures.](image-url)
Figure 73 and Figure 74 show the hardness evolution along the five lines of two welded samples (N1 and N2). The variation in the hardness magnitude did not significantly change between the two samples. As a result, the repeatability of the welding operation was good (average hardness was 184 HB and 203 HB for samples N1 and N2, respectively).

The microstructure of the base material near the HAZ contained coarse pearlite and ferrite, which generally have a relatively low hardness in comparison to the filler weld material. As a result, the weld material was harder than the adjoining regions, as expected [8]. In both samples, the hardness increase in the weld was ~ 15 – 20 HB relative to the base material, as exemplified in Figure 75 to Figure 77.

In general, the hardness along lines 1 and 5 (which were near the surface of the welded samples) had a more pronounced increase near the weld than in the HAZ and base material. This trend was likely related to the difference in the cooling rate near the surface of the welded section and the sample’s centerline core. At the surface, the cooling rate was higher resulting in a higher volume fraction of a hard martensite.
Figure 73: Hardness results sample N1.

Figure 74: Hardness results sample N2.
Figure 75: Hardness results sample N1 lines 1 and 5.

Figure 76: Hardness results sample N1 lines 2 and 4.
4.2.3. Impact tests

The purpose of performing impact tests was to determine the impact strength of the virgin AR200 at different temperatures. Figure 78 shows that the AR200 should not be used for low temperature applications since it had a very low impact strength at temperatures below 0°C. The majority of the samples below 25°C fractured in brittle manner, as seen in Figure 79 a). Above this temperature, the samples fractured in a ductile manner, as seen in Figure 79 b). From these results, it can be seen that the mean impact energy of $E = 67$ ft-lbs yielded DBTT of 22°C. This value of DBTT is relatively high to enable the AR200 steel to be used in low temperature environments. AR200 steel contains 0.35 wt% C and 0.25 wt% Si, which shift the S-curve to the right [8]. Therefore, the presence of a significant amount of Si possibly contributed to the high DBTT value, which makes this alloy not suitable for low temperature applications with impact loading.
4.2.4. Effect of heat treatment on hardness

Figure 80 shows the effect of heat treatment temperature on the hardness of AR200. As the tempering temperature increased, the hardness of the AR200 decreased. At 700°C, the lowest hardness was observed (165HB), which was ~2.5x lower than that of the virgin AR200 steel.
This can be explained by the fact that the higher the tempering temperature, the grain size increased leading to a reduced grain-boundary strengthening effect. Further, with an increasing tempering temperature, the microstructure evolved and the volume fraction of martensite in the samples decreased. Thus, the AR200 steel became softer with increasing heat treatment temperature.

Figure 80 also includes hardness values for as-received, normalized (with fine pearlite microstructure) and annealed (with coarse pearlite microstructure) samples. For the normalized sample, the hardness decreased to 136HB from that of the as-received sample (176HB). These values suggest that a long-term heat treatment of AR200 can deteriorate the alloy's hardness.

![Figure 80: Tempering test results after 2 h.](image)
4.2.5. **End quench test (Jominy hardenability test)**

Figure 81 shows the results of the Jominy hardenability test. As can be seen from the figure, the harden depth (case) was ~15 mm. Within that case, the hardness varied from 299HB to 195HB.

The samples in the present study were heated to 870°C, which is just 50°C above the A₃ line on the Fe-C phase diagram. During the quenching process, any delay (even of a few seconds) in transferring the samples from the furnace to the water-jet may provide sufficient time to start formation of primary ferrite. Subsequently, quenching forms martensite in already existing ferrite crystals. The AR200 potentially can achieve higher hardness, but the values observed in this study were lower than expected, as can be seen in comparison to a similar alloy (AISI 1530) in Figure 82 and Table 9.

![End quench test graph](image)

**Figure 81:** End quench test results.
Figure 82: End quench test of AISI 1530 [36].

Table 9: Comparison between end – quenched results and published results.

<table>
<thead>
<tr>
<th>Distance from the edge, mm</th>
<th>Harness range, HB</th>
<th>Hardness range, HRC</th>
<th>Distance from the edge, mm</th>
<th>Minimum hardness, HRC</th>
</tr>
</thead>
<tbody>
<tr>
<td>0 – 2</td>
<td>290 – 310</td>
<td>31 – 33</td>
<td>0 – 2</td>
<td>47</td>
</tr>
<tr>
<td>2 – 7</td>
<td>230 – 240</td>
<td>22 – 23</td>
<td>2 – 7</td>
<td>38 - 47</td>
</tr>
<tr>
<td>7 – 10</td>
<td>222</td>
<td>20</td>
<td>7 – 10</td>
<td>25 – 38</td>
</tr>
<tr>
<td>15 – 25</td>
<td>180 – 195</td>
<td>8 - 12</td>
<td>15 – 25</td>
<td>-</td>
</tr>
<tr>
<td>25 - 50</td>
<td>170 - 185</td>
<td>5 - 9</td>
<td>25 - 50</td>
<td>-</td>
</tr>
</tbody>
</table>
Figure 83 illustrates the microstructure of the Jominy sample at different locations. Close to the quenched end the microstructure contained fine crystals of ferrite and martensite. With increasing distance from the quenched end, the microstructure changed towards coarse ferrite and pearlite, leading to a decreased hardness away from the quenched end.

Figure 83: Jominy end quench microstructure.
Chapter 5: Conclusions

The key conclusions made in this work can be summarized as follows:

1. Heat treatment had only minor effect on the strength of AR200 steel samples.

2. Torch heat treatment did not eliminate RS as intended, but allowed the residual stresses to reach more homogeneous levels.

3. Impurities in the welds had a significant effect on the ductility of the steel. Large intermetallic compounds and oxides reduced the tensile strength in welded AR200 sections.

4. A thorough consideration must be made regarding the use of fluxes or flux-cored wire to improve cleanliness of the weld material. Welding wire should contain fluxes that help purify of the weld, but without a negative impact on the brittleness of the AR200 steel.

5. It is not recommended that AR200 is used for low temperature applications with impact loading conditions.

6. Heat treatment temperature should be higher than 600°C to eliminate formation of martensite.

Future work

Further experiments can be carried out with a flux cored wire (FCAW) to determine whether or not there will be a positive effect on weld purification and improvement of the weld strength. Further, the effect of the size of welding defects and their impact on the mechanical performance of welded steel sections should be examined.
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Appendices

Appendix A: Additional SEM images
Appendix B: Additional impact test images
Appendix C: Additional tensile test images