Determination of Rolling Parameters for Numerical Simulation of Cold Rolling of Nickel Aluminum Bronze

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Determination of Rolling Parameters for Numerical Simulation of Cold Rolling of Nickel Aluminum Bronze

By

Giuseppe Iorio

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Abstract

The purpose of this work was to conduct a reverse engineering evaluation to attain state variables such as flow stress, strain, and strain rate for a finite element (FE) simulation of the rolling process. The alloy investigated included as cast and wrought Nickel Aluminum Bronze (NAB) of the similar chemistry. Chemical analyses of the materials were in conformance with the ASTM B150 and ASTM B 171 C6200 specifications. Tensile testing results of the wrought material did not show a significant difference between the transverse and longitudinal directions. Microstructural examination of the wrought material in the transverse, and in the longitudinal directions revealed minimal difference in the grain size and shape. This observation is consistent with the cross-rolling work reportedly done on the commercial plates. It was observed a typical NAB alpha (α) phase with kappa (κ_{III} (κ_{III}), and (κ_{IV}) phases. The (κ_{I}) phase was not observed possibly due to low iron content. The grain size of the wrought and the as-cast material was determined using the ASTM E112 comparison method.

The as cast (NAB) was poured into ingots and exposed to various heat treating conditions for microstructure evaluation. The metallography results of the ingots revealed that the original material prior to being cold rolled had an as cast microstructure. The study was completed performing deformation testing on a Gleeble 3500 system in the plain strain hydra-wedge to obtain flow stress curves at strain rate s^{-1} 0.1,0.3,1, 5 and 20. It was learned the as cast material can only be deformed up to 15.9 % reduction.
Chapter 1: Introduction

Nickel Aluminum Bronze (NAB) alloys are widely used in marine applications, primarily because of corrosion resistance, strength, ductility and sound weldability. Casting is used as manufacturing processes for the fabrications of propellers, and other marine applications such as plates, pumps and valves. Cast ingots are castings of a simple shape and slabs ingots can weigh from 9 to 36 metric tons and may have thicknesses up to 900 mm [1]. In order to roll strip from ingots, the ingots are rolled down to a size of a slab with thicknesses ranging from 180 to 250 mm [1]. Then they are further reduced in thickness at the roughing stands of hot strip mill down to 25 to 50 mm with subsequent reductions to the desired hot rolled thickness at the finishing mill. Final reduction in thickness can be accomplished by rolling at the cold mill [1]. Figure 1 shows the flow of rolled material through various rolling mills.

Figure 1. Flow of rolled material through several rolling mills [1].
More specifically, ingots are turned into blooms to form structural shapes and rails, in slabs to form plates and coils, and in billets to form bars and rods shown in Figure 2 [2].

![Figure 2. Products made in a rolling mill from ingots [2].](image)

The cooling rates of the NAB during the cooling of the castings are very slow on the order of $10^{-3}\, ^\circ C/s$ for the thicker sections contributing to porosity and to a coarse microstructure [3,4,5]. The coarse microstructure and porosity created in the casting process affects the material properties, including strength, ductility fracture toughness, and corrosion resistance [5, 6]. Both hot and cold rolling can lead to major improvement of material properties by refining the microstructure [7]. As-cast ingots are often characterized by large grain sizes with significant porosity and coarse second phase particles. During hot rolling, porosity can be closed up, grain size reduced by recrystallization and coarse particles broken up leading to stronger and tougher alloy [7]. Cast structure of the ingots or continuously cast alloys is broken down into a wrought structure, yielding finer grain size and improved properties [8].
1.1 Rolling

Rolling is the process of reducing the thickness or changing the cross section of a long work-piece by compressive forces applied through a set of rolls [8]. The process typically includes hot rolling warm rolling and cold rolling [9]. According to Lenard, the rolling process may be performed at temperatures above half of the melting point of the metal is termed hot rolling, or below that temperature, in which case one deals with cold rolling [9]. The temperature range, beginning below and ending above the dividing line between hot and cold rolling within the process is termed warm rolling. Lenard pointed out that this temperature for some materials is the preferred process to follow and leads to mechanical and metallurgical changes of the of the work piece, which are not possible to be achieved in either the cold or the hot flat rolling regimes [9].

1.12 Hot Rolling

Hot rolling is classified according to the temperature of the metal rolled. The term hot rolling also means hot working and is defined as working above the recrystallization temperature so that the work metal recrystallizes as it deforms [10]. The high temperatures lower the value of the flow stress and the deformation process requires lower forces to deform the material [10]. Hot rolled material is free of residual stresses, and has isotropic properties [11]. Prior to the hot rolling ingots are heated up above recrystallization temperature (Tr) having large non-uniform grains. During hot rolling, the hot metal has large grains while entering the gap between working rolls. The material is deformed, “pancaking” the grains in the direction of the material flow known as rolling direction [12]. After leaving the rolls, the grains recrystallize in a process known as static recrystallization [12,13,14,15] and begin to grow. This step is typically followed
by further deformation and the process repeats until the temperature of the material cools to a point about where recrystallization is not completed, often 50-100 °C (122-212 °F) under the normally defined recrystallization temperature (Tr). For the temperature under the recrystallization temperature a material transformation will be completed [12,16]. Figure 3a, b, and c, shows a schematic of the hot rolling process and static recrystallization with large ingot grains before it enters the roll gap, deformed grains in the roll gap and recrystallized grains after the material has left the rolls. Recrystallization during deformation (while the material is in the roll bite) can occur if enough strain has accumulated before and during deformation, and is called dynamic recrystallization [12,13,14].

![Diagram of deformation and recrystallization of the hot rolled metal](image)

**Figure 3.** Deformation and recrystallization of the hot rolled metal [11].

According to Shay, in hot rolling process the flow stresses are low which provides an advantage to require relatively low forces and power for rolling to take place so that large workpieces can be deformed with equipment of reasonable size. These can take place where friction
conditions are high and favoring high deformation for a given size of working rolls. Ductility is high which favors large deformation (up to 99%) reduction producing complex part shapes in fewer working steps, without causing large amounts of internal stresses, cracks or cold work. [17]. Pores seal up, micro-segregation is reduced or removed due to atomic diffusion which is faster at high temperatures. The metal improves mechanical properties and becomes more ductile than as cast metals due to breaking down and refinement of coarse columnar grains in the cast ingot. According to Dieter hot rolling occurs at an essential constant flow stress, and because the flow stress decreases with increasing temperature the energy required for deformation is generally less for hot rolling than for cold rolling and the structure and properties of the hot worked material are not uniformed over the cross section as in metals have been cold worked and annealed. This is due to the fact that the deformation is always greater in the surface layers causing the metal to have a finer recrystallized grain size in this region. Because the interior will be at higher temperatures for longer times during cooling than will be the external surfaces, grain growth can occur in the interior of large pieces, which cool slowly from the working temperature. Further, has been stated by Dieter the lower temperature limit for the hot rolled metal is the lowest temperature at which the rate of recrystallization is rapid enough to eliminate strain hardening in the time the metal is at temperature. For a given metal the lower hot rolling temperature depend on the amount of deformation and time when the metal is at temperature. Since the greater the amount of deformation the lower the recrystallization temperature, the lower temperature limit for hot rolling is decreased for large deformations. Therefore metal which is rapidly deformed and cooled rapidly from temperature require a higher hot working temperature for the same degree of deformation than metal slowly deformed and slowly cooled [18].
The major disadvantage experienced during deforming metals in hot rolling, Roberts pointed out, is oxidation that occurs on the surface of the metals. The thinner the work-piece or more extended their surfaces the greater the loss of yield due to scaling. The scale tends to be rolled into the metal resulting in an undesirable surface quality [19]. Due to the thermal expansion caused by the high temperatures dimensional tolerances are significant inferior to cold rolling. Hot-rolled metal is generally free of residual stresses, and has isotropic properties.

1.13 Cold Rolling

Normally it is performed at room temperature, but in general at temperature lower than 0.3 of the melting point $T_m$ measured on the absolute temperature scale. Under these conditions recovery is limited and recrystallization does not occur. The temperature at which the deformation takes place is an important determinant of the final properties. Essentially there are three major objectives in the cold rolling process, one to reduce the metal thickness further; two to increase the rolled metals’ strength by strain hardening; three to improve the dimensional consistency of the product [9]. The surface of the cold rolled material is absent of scale and generally superior to the hot rolled product. During cold rolling a metal will tend to increase its strength through work hardening, but its ductility decreases. Therefore, to remove internal stresses of cold work, it is sometimes desirable to heat treat the metal after cold working. The disadvantages of cold rolling are an increased cost due to the use of heavy equipment utilized and the cost of the wear to the equipment.
1.14 Warm Rolling

Warm rolling stands between the hot and cold rolling temperature range in the process and is not defined very closely. Lenard describes it starts somewhat below half of the homologous temperature and ends somewhat above that [9]. In the relationship between temperature and deformation, it is used a normalized parameter that is called the homologous temperature, Tm. The homologous temperature is a dimensionless parameter that is defined as the absolute temperature divided by the absolute melting temperature of the material: \( T_h = \frac{T}{T_m} \), with T for temperature and Tm for the melting temperature of the material, both in Kelvin [20]. From a manufacturing standpoint, warm working is thought to combine the advantages of hot and cold working. It requires lower loads and energies compared to cold working and has the advantages over hot working of achieving better surface finishing and dimensional precision [20]. In the steel shaping industry, it has been considered for its production advantages and lower production cost [21,22]. Madrea et al indicated that warm working compared to hot rolling leads to a finer microstructure with superior mechanical properties, better surface quality and better dimensional control, lower material losses due to decarburization and oxidation [23].
Nickel-Aluminum Bronze (NAB) is a series of copper-based alloy with the additions of 9% - 12% Al, 6% Ni and 3 to 4.5 % Fe weight percent specified by ASTM B148 [25] for cast marine propellers, ASTM B150 [26] for plates and ASTM B171[27] for bars. The high corrosion resistance makes it one of the most used alloys in marine application [28, 29, 30]. The Nickel Aluminum Bronze primarily alloying elements are Copper (Cu) with the additions of secondary elements such as Nickel (Ni), Aluminum (Al), Iron (Fe), and Manganese (Mn). The secondary elements addition percentages can vary according to the specification. The chemical composition requirement for as-cast material falls under the ASTM B148 standard and the chemical composition requirement for wrought plate material falls under the ASTM B150 standard plate specification C63200. The chemical compositions used in this research are summarized in Table 1.

Table 1. Standard Specifications of as-cast and wrought Nickel-Aluminum–Bronze

<table>
<thead>
<tr>
<th>Standard</th>
<th>Cu</th>
<th>Al</th>
<th>Ni</th>
<th>Fe</th>
<th>Si</th>
<th>Mn</th>
</tr>
</thead>
<tbody>
<tr>
<td>ASTM-B-150</td>
<td>79.3-82.5</td>
<td>8.7-9.5</td>
<td>4.0-4.8</td>
<td>3.5-4.3</td>
<td>0.10</td>
<td>1.2-2.0</td>
</tr>
<tr>
<td>ASTM-B-171</td>
<td>79.3-82.5</td>
<td>8.7-9.5</td>
<td>4.0-4.8</td>
<td>3.5-4.3</td>
<td>0.10</td>
<td>1.2-2.0</td>
</tr>
</tbody>
</table>

2.1 Nickel Aluminum Bronze Phases (NAB)

The alloy has a microstructure of Copper (Cu) rich solid solution known as alpha phase (α) with a (FCC) structure surrounded by lamellar eutectoid phase with a series of intermetallic
kappa phases (κ) and a dark etched martensitic regions beta phase (β) or retained beta prime phase (β') with a B2 high temperature phase structure [6,31, 32, 33]. Fuller et al. declared the addition of Ni and Fe enhance the formation of the κ phases from the α and β phases at the nominal values described in Table 1 and provide benefits such as increase strength, wear, abrasion resistance and increases fatigue endurance limit. The addition of Ni and Fe alloying elements to the copper–aluminum system creates new intermetallic phases and suppresses the formation of the corrosion susceptible to copper-aluminum phase [34, 35, 36, 37]. The work by K. Oh-Ishi and T.R. McNelley identified the nickel aluminum bronze microstructural changes during cooling as shown by the equilibrium cooling path in Figure 4 [38].

![Figure 4. Transformation product formation in Nickel Aluminum Bronze [36].](image-url)
Figure 4 shows that at highest temperatures only the β phase exists. It remains entirely β phase until it is cooled below 1000 °C. Then the primary α phase forms in the β phase with a Widmanstätten morphology. The NAB phase diagram in Figure 6 indicates upon cooling the κ phases begin below 930 °C with the disappearance of the β phase approximately 860 °C [6]. At temperatures below 900 °C, globular κii phase begins to form in the β phase [35]. The κi phase forms only when the iron content is greater than 5% and when a high iron to nickel ratio exists [33, 39, 40]. When the temperature falls to 860 °C and below, κiv phase starts to form in the α phase as the solubility of iron is exceeded. At 800 °C, the remaining β phase is transformed into κiii phase through the eutectoid reaction, \( \beta \rightarrow \alpha + \kappa_{iii} \). Once the material reaches room temperature, the alloy consists entirely of α phase and the κ phases, and retained β or β’ regions composed of martensite containing small particles based on NiAl intermetallic [33]. An optical micrograph of as cast wrought nickel aluminum bronze material is provided in Figure 5.

![Optical micrograph of as cast wrought nickel aluminum bronze material](image)

**Figure 5.** Optical micrographs of as-cast NAB showing α and κ phases regions [31].
2.11 Beta Phase

As shown in Figure 4 at 1000 °C only the β phase exists. The β-phase is a body centered cubic (BCC) solid solution with a lattice parameter of 0.357 nm [33,34]. The β-phase is stable only at temperatures above the eutectoid temperature, which is approximately 800 °C. Upon cooling, the β-phase transforms into the α-phase and the κ-phases. The transformation products of the β-phase depend on the cooling rate [34]. The left side of diagram in Figure 6 shows the transformations resulting from equilibrium cooling at very slow rate. The slow cooling rate result in a full β transformation to α-phase and κ-phases [40]. The κ-phase formation begins below 930 °C with the disappearance of the β-phase below approximately 860 °C. The right side of the diagram shows the transformation resulting from equilibrium at high rate. At higher cooling rate some of the β-phase remains, which has been termed retained β or β′.
The work by Howell showed that during the cooling cycle e.g., air cooling the higher temperature β phase can separate above 500 °C to produce plates of bainite in the β matrix and start forming β'. Eventually the remaining β' transforms to martensite. Howell describes retained β or β' as a dark microstructure consisting of both martensite and bainite and he points it out by the designation of α* [38,39]. This is illustrated in the optical micrograph shown in Figure 5. This structure has an ordered 3R or 2H martensitic structure with a high density of NiAl precipitates [33,34]. The retained β phase has a negative effect on the corrosion resistance of nickel aluminum bronze in sea water; therefore, to ensure that no retained β phase remains, either a slow cooling rate is necessary or a post casting tempering needs to be completed [41].

2.12 Alpha Phase

On further cooling below 1030 °C the α phase forms in the β phase with a Widmanstatten morphology [34, 35, 38]. The α phase is a face centered cubic (FCC) solid solution with a lattice parameter of 0.364 nm [34]. According to (F. Hasan et al., 1982) this phase is rich in copper with small amounts of aluminum, nickel, and iron and various precipitates, collectively referred to as K-phase, based on Fe3Al or NiAl. Supported by regions of the high temperature β phase which do not undergo diffusional decomposition on cooling (the so-called"retained") transform into a complex martensitic structure which contains a high density of NiAl precipitates [33, 40].

2.13 Kappa Phases

In high iron region, the κi phase is seen in the center of the α phase with a large, rosette shaped morphology typically 20 to 50 μm in diameter shown in Figure 7 [33].
Figure 7. As-cast NAB kappas phases distribution in the alpha metrics [30].

This phase will generally occur in NAB compositions where the Fe content is 5% or greater and may begin forming in the melt [41]. This phase is very sensitive to alloy (iron) content and cooling rate [42]. This phase is nominally (Fe₃Al, FeAl) formed around a small copper rich particle with small amounts of nickel, silicon, and manganese. The composition varies as shown in Table 2 as does the crystal structure (BCC, DO₃ or B2) shown in Figure 8 [33].
Table 2. Chemical Analysis of phases present in NAB [32,34]

<table>
<thead>
<tr>
<th>Phase</th>
<th>Elements</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Cu</td>
</tr>
<tr>
<td>Alpha</td>
<td>85.4</td>
</tr>
<tr>
<td>Beta</td>
<td>85.2</td>
</tr>
<tr>
<td>Kappa I</td>
<td>8.4</td>
</tr>
<tr>
<td>Kappa II</td>
<td>9.3</td>
</tr>
<tr>
<td>Kappa III</td>
<td>12</td>
</tr>
<tr>
<td>Kappa IV</td>
<td>2</td>
</tr>
</tbody>
</table>

Figure 8. NAB Kappa phases structures [32,39,42]
The κii phase is globular in nature and has a DO3 structure with a lattice parameter of 0.577 nm. This phase is much finer than the κi phase with a 5 to 10 μm diameter. The κii phase is nominally Fe₃ Al with small amounts of copper, silicon, nickel, and manganese. The κii phase nucleates in the β phase and is therefore found in the areas containing the κiii phase, Nelson has reported that occasionally while the α phase it is growing it enveloped some of the κii phase which had formed at the α/β interface [34]. The κiii phase has an ordered BCC (B2) structure with a lattice parameter of 0.288. This phase is nominally NiAl with small amounts of copper, iron, and manganese. It has a lamellar morphology and forms as an eutectoid decomposition product, though a proeutectoid κiii phase may also form in a globular morphology. The κiv phase is very similar to the κii phase, having a DO3 structure with a lattice parameter of 0.577 nm.
Chapter 3: Objective

This thesis is a study of reverse engineering to recreate rolling of 25 mm (1 inch) to 51 mm (2 inches) thick Nickel Aluminum Bronze (NAB) plates for marine applications. This study characterizes wrought NAB material from a remnant (NAB) slab produced by a supplier that no longer exists. The goal is to fabricate plates to the thicknesses described above using as cast scrap material of same chemistry. The full process of reverse engineering includes two stages: (1) Material data collection, and (2) FEM simulation. This study focuses only on stage 1 which provides the state variables such as flow stress, strain, strain rate, for the stage 2. This will lead to reconstruction of the rolling schedule once stage two of the overall project is completed. The final objective is to reconstruct the production process and apply it in a new facility. In this study, stage 1 is completed through chemical, microstructure evaluation and mechanical property characterization of the remnant (NAB) material as compared to a new (NAB) as cast material of the same chemistry.
Chapter 4: Procedure

4.1 Material

4.1.1 Rolled NAB

Nickel aluminum bronze wrought material was processed by a commercial facility that no longer exists. Samples of the wrought (NAB) were obtained from a remnant slab cut from a plate as shown by the sketch in Figure 8.

![Diagram of slabs](image)

**Figure 9. Wrought material location diagram**

The slab dimensions were 9 cm (3.5-inches) (W) x 42 cm (16.5-inches) (L) x 9 cm (3.5-inches) (H). Two 9 cm x 9 cm sections were saw-cut from the left end of the slab and the right side of the slab. The right side of the slab is from the vicinity of the center of the plate. The slab is shown in Figure 9.
The cuts were executed with a power hack saw. The saw used a water coolant (Trimsol) to reduce the heat generated during the cuts. The samples were macro-etched with 10% Ammonium Persulfate as shown in Figure 9 and Figure 10.

Figure 11. Left edge longitudinal view of NAB remnant
From these sections several samples were cut for metallography analysis using a Discotom 6 cut-off saw with a water cooled aluminum oxide abrasive blade. Several samples were cut in the transverse and longitudinal directions of the material sections. This can be seen in Figure 12.

Figure 12. Right edge through thickness longitudinal view of NAB remnant

Figure 13. Metallographic Specimen lay-out of NAB remnant
4.12. As cast NAB material

The as-cast material was obtained from re-melting NAB scraps of a similar chemistry to the wrought NBA material. Four sand molds were prepared to accommodate four ingots with dimensions of 51 cm (20 inches) long, 5.1 cm (2 inches) wide and 25 cm (10 inches) high shown in Figure 13.

![Figure 14. Melting operation of as-cast NAB samples](image)

Then the ingots were marked as shown in Figure 13 and received an annealing treatment for further analyses. The annealing process is tabulated in Table 3.
Table 3. Ingots annealing procedure

<table>
<thead>
<tr>
<th>Material</th>
<th>Ingot #</th>
<th>Method</th>
<th>Temperature (°C)</th>
<th>Hold Time (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AS Cast</td>
<td>1</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Annealed</td>
<td>2</td>
<td>Furnace Cool</td>
<td>663</td>
<td>6</td>
</tr>
<tr>
<td>Annealed</td>
<td>3</td>
<td>Water Quench</td>
<td>663</td>
<td>6</td>
</tr>
<tr>
<td>Annealed</td>
<td>4</td>
<td>Air Cooled</td>
<td>663</td>
<td>6</td>
</tr>
</tbody>
</table>

Ingot number 1 did not receive annealing and was left as-cast. Ingots number 2, 3 and 4 were annealed at 663 °C (1225 °F) for six hours and cooled utilizing the methods of furnace cool, water quenched and air cooled as indicated in Table 3. After the annealing procedure was completed the ingots were sectioned following the same cut procedure performed for the wrought NAB material described above. The material was monitored upon cooling with a thermal analysis sand cup set up shown in Figure 14 that included a K thermocouple connected to a data acquisition computer. A plot of the cooling data is shown in Figure 15. It can be seen that the liquidus line is approximately 1057 °C (1935 °F) and the solidus line is approximately 1038 °C (1900 °F) which is in agreement with the NAB cooling diagram in Figure 6. Around 943 °C (1743 °F) a bump appeared on the curve. This phenomenon could be caused by segregation which often is observed on large propeller castings during slow cooling, or can be caused by a lower melting point phase. The phase change at 943 °C (1743 °F) was also identified by K. Oh-Ishi and T.R. Mc Nelley [35] reported in Figure 4.
Figure 15. Thermal analysis cup for temperature cooling data

Figure 16. Nickel Aluminum Bronze cooling curve
4.2 Microstructure Analysis

4.21 Sample preparation

Specimens selected for optical microscopy were cut with a laboratory abrasive cut off saw from the front and the rear sections of the slab as shown in Figures 10 and 11. The samples were ground and polished in several steps using a Struers Abrapol-2 grinding/polishing machine. Grinding was accomplished using 120-, 240-, 320-, 600-, and 1000-grit SiC paper for 60 seconds under 150 Newton force. Polishing was completed in three steps using a Struer MOL cloth 3 µm diamond suspension for 3 minute, 200 Newton force; Struers NAP cloth– 1 µm diamond suspension for 1.5 minutes, 200 Newton force; Struers NAP cloth–OP-S 0.05 colloidal silica suspension with 1% Ferritic Nitrate addition for 45 seconds, 150 Newton force. The final step was completed etching the specimen surface using 10% Ammonium Persulfate.

4.22 Optical Microscopy Procedure

Using an Olympus PMG–3 microscope, Optical Microscopy (OM) was used to evaluate the grain structure on all the samples and used to gather information about the types of phases present in the material. The sample grain size in each of the samples was estimated in accordance with ASTM Standard Method E 112 [43].

4.3 Chemical Analysis

The chemical analysis was performed by Luvak Inc. chemical laboratory located in Boylston, MA for all elements in compliance with ASTM B150 and ASTM B171 standards.
4. 4 Mechanical Testing

4.41 Tensile Testing

The wrought NAB material tensile specimens were machined and tested at Laboratory Testing, Inc. (LTI), Hatfield, PA, in accordance with ASTM-E-8 standard. Due to the slab thickness restrictions, 2 sub-sized standard specimens with a nominal diameter of 6.35 mm (0.250 inches) were machined in the transverse direction and two standard specimens with a nominal diameter of 12.7 mm (0.505 inches) machined in the longitudinal direction. The as cast tensile specimens were all machined to a nominal diameter of 12.7 mm (0.505) inches and tested at the same laboratory.

4.42 Hardness Testing

Hardness testing was performed with a Rockwell hardness machine on the ‘B’ scale using a 100-kg weight. The Rockwell ‘B’ scale measurements for the wrought material were taken at the inner face of the 89 mm x 89 mm samples shown in Figures 10 and 11. The as cast measurements were taken from the inner face of a material section removed from the ingot. Depending on the consistency of the values obtained, either 4 to 5 hardness values were taken from each sample and averaged.

4.43 Gleeble Testing

The Gleeble thermo-mechanical simulator is used as a tool to provide judiciously controlled deformation cycles in order to study the metallurgical properties of materials. The Gleeble tests provide detailed information that can be used to learn about the thermo-mechanical history of the material. It allows developing processing maps for new materials. Furthermore,
allows to independently varying strain and temperature in ways that the data can be used to
optimize rolling schedules for rolling mill operation under specified rolling pass conditions. In
this study, the Gleeble thermo-mechanical physical simulator is used as a predictive modeling
tool. In this study the gleeble compression tests were performed in plane strain conditions. The
plane strain compression test is performed for measuring flow stress with strictly defined
thermo-mechanical parameters. In principle, the plane strain compression test is particularly
suitable for physical simulation, of rolling. Deformation testing was performed to collect data
for flow stress curves using the Gleeble 3500® system in the plain strain compression hydra-
 wedge jaw set-up. The machining and specimen preparation was accomplished in several steps.
Step one 6.35 mm (¼ inch) material was removed from the edges on both sides of the NAB as-
cast ingot to eliminate segregation and any defects encapsulated in material attributed to melting
and solidification process. Then on step two, specimens of size 10 x 15 x 20 mm were machined
from the cleaned remaining material. The schematic in Figure 17 shows the specimen geometry
and the sides of the specimen position during the deformation testing.

![Figure 17. Gleeble specimen geometry during compression testing](image)

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In step 3 a type K control thermocouple was spot welded at the center of the top surface of the specimen using a spot weld apparatus as shown in Figure 18.

![Figure 18. Gleeble specimen thermocouple installation](image)

Step four consisted sectioning a tantalum (Ta) foil and positioned it on both sides of the sample which would act as diffusion barrier/lubricant system between the specimen and the Gleeble machine grips illustrated in Figure 19.
Figure 19. Location of tantalum foil on the faces of the compression sample

Step five was accomplished by applying a thin layer of nickel-based anti-seize compound at each interface of the tantalum foil and specimen to improve lubrication reducing friction during the specimen deformation as shown in Figure 20.

Figure 20. Nickel-based anti-seize compound and where it is applied on the sample
Step 6, the final step was completed by carefully mounting the specimens into the Gleeble machine levelling the center of the samples to the machine anvils grips and connecting the thermocouples wires to the machine instrumentation panel as shown in Figure 21.

![Specimen positioning in hydra wedge grips](image)

**Figure 21. Specimen positioning in hydra wedge grips**

Testing was performed at room temperature, 22 °C (72 °F). Specimens were deformed sequentially up to a true strain ($\varepsilon$) of 0.1, 0.11, 0.16, 0.22 and a strain rate ($\dot{\varepsilon}$) of s$^{-1}$ 0.1, 0.3, 1, 5 and 20. The parameters for the true strain ($\varepsilon$) under plastic deformation conditions when the specimen is compressed was calculated using equation 2 and the percent reduction using from equation 3 [43]. Table 4 provides parameters for the testing conditions tabulated in Tables 5.
% R = \frac{(h-h_0)}{h} \times 100 \hspace{1cm} \text{(eq.3)}

Where:

h = \text{final thickness}

h_0 = \text{initial thickness}

\begin{align*}
\varepsilon &= \frac{2}{\sqrt{3}} \cdot \ln \frac{h}{h_0} \\
\text{Table 4. Gleeble initial calculation for test parameters}
\end{align*}

\begin{table}[h]
\centering
\begin{tabular}{|c|c|c|c|}
\hline
ho (mm) & h (mm) & True Strain & (% R) \\
\hline
10 & 9.9 & 0.01 & 1.01 \\
10 & 9.7 & 0.04 & 3.09 \\
10 & 9.5 & 0.06 & 5.26 \\
10 & 9.3 & 0.08 & 7.53 \\
10 & 9.1 & 0.11 & 9.89 \\
10 & 8.9 & 0.13 & 12.36 \\
10 & 8.7 & 0.16 & 14.94 \\
10 & 8.5 & 0.19 & 17.65 \\
10 & 8.3 & 0.22 & 20.48 \\
10 & 8.1 & 0.24 & 23.46 \\
\hline
\end{tabular}
\end{table}
Table 5. Gleeble tests conditions

<table>
<thead>
<tr>
<th>Material (Samples)</th>
<th>True Strain (ɛ)</th>
<th>Strain Rate (ɛ) s(^{-1})</th>
<th>True Stress (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
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<td>0.1</td>
<td>621</td>
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<tr>
<td>2</td>
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<td>639</td>
<td></td>
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<tr>
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<td>1</td>
<td>637</td>
<td></td>
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<td>20</td>
<td>643</td>
<td></td>
</tr>
<tr>
<td>6</td>
<td>0.11</td>
<td>0.3</td>
<td>660</td>
</tr>
<tr>
<td>7</td>
<td>1</td>
<td>660</td>
<td></td>
</tr>
<tr>
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<td>0.3</td>
<td>778</td>
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<td>777</td>
<td></td>
</tr>
<tr>
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<td>5</td>
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<td>20</td>
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<td></td>
</tr>
<tr>
<td>14</td>
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<td>0.3</td>
<td>840</td>
</tr>
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<td></td>
</tr>
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<td>17</td>
<td>20</td>
<td>857</td>
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</tr>
</tbody>
</table>
Chapter 5: Results

5.1 Chemical Analysis

Table 5 provides the chemical analysis results of the NAB wrought and as-cast material along with the standards ASTM-B-150 and ASTM-B-171 specifications requirements.

Table 6. Chemical analyses results.

<table>
<thead>
<tr>
<th>Specifications</th>
<th>Cu</th>
<th>Al</th>
<th>Ni</th>
<th>Fe</th>
<th>Si</th>
<th>Mn</th>
<th>Sn</th>
<th>P</th>
<th>Pb</th>
</tr>
</thead>
<tbody>
<tr>
<td>ASTM-B-150</td>
<td>79.3-82.5</td>
<td>8.7-9.5</td>
<td>4.0-4.8</td>
<td>3.5-4.3</td>
<td>0.10</td>
<td>1.2-2.0</td>
<td>-</td>
<td>-</td>
<td>0.02</td>
</tr>
<tr>
<td>ASTM-B-171</td>
<td>79.3-82.5</td>
<td>8.7-9.5</td>
<td>4.0-4.8</td>
<td>3.5-4.3</td>
<td>0.10</td>
<td>1.2-2.0</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Alloy Wrought</td>
<td>80.30</td>
<td>8.84</td>
<td>4.80</td>
<td>4.07</td>
<td>0.095</td>
<td>1.69</td>
<td>0.048</td>
<td>&lt;0.002</td>
<td>&lt;0.008</td>
</tr>
<tr>
<td>Alloy As-Cast</td>
<td>82.5-79.3</td>
<td>8.7-9.5</td>
<td>4.0-4.8</td>
<td>3.5-4.3</td>
<td>0.10</td>
<td>1.2-2.0</td>
<td>-</td>
<td>-</td>
<td>0.02</td>
</tr>
</tbody>
</table>

The results showed the samples conforming to the specification. Small traces of other elements were found in the wrought plate chemical analysis. These elements could have already been in the melting apparatus when the material was being cast.

5.2 Tensile Testing

The tensile tests were conducted in accordance with ASTM E-8 [43] standard at room temperature. The tensile testing were performed at a rate of 0.381 (mm)/(mm)/(m) 0.0150in/in/m for specimens of nominal diameter of 6.35 mm (0.250 inches) and for specimen nominal diameter of 12.7 mm (0.505 inches). The tensile testing results in longitudinal (rolling
direction) and transverse directions for wrought NAB are listed in Table 7 and Table 8. Table 9 lists tensile testing results for as-cast material. The tables include samples identification, test temperature, ultimate tensile strength, (UTS), 0.2% off-set yield strength (YS), percent elongation (% EL), percent reduction area (% RA) and material ASTM-B-171 plates specification.
Table 7. Wrought NAB material longitudinal tensile test results

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>Test Temperature (C)</th>
<th>UTS (MPa)</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of Area (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>505/1</td>
<td>22</td>
<td>662.0</td>
<td>336.5</td>
<td>18</td>
<td>19</td>
</tr>
<tr>
<td>505/2</td>
<td>22</td>
<td>665.3</td>
<td>336.5</td>
<td>21</td>
<td>24</td>
</tr>
<tr>
<td>ASTM B171</td>
<td>22</td>
<td>586.1</td>
<td>213.7</td>
<td>10</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 8. Wrought NAB material transverse tensile test results

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>Test Temperature (C)</th>
<th>UTS (MPa)</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of Area (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>350/1</td>
<td>22</td>
<td>665.3</td>
<td>355.1</td>
<td>21</td>
<td>20</td>
</tr>
<tr>
<td>350/2</td>
<td>22</td>
<td>658.5</td>
<td>338.5</td>
<td>27</td>
<td>22</td>
</tr>
<tr>
<td>ASTM B171</td>
<td>22</td>
<td>586.1</td>
<td>213.7</td>
<td>10</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 9. As-Cast NAB material tensile test results

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>Test Temperature (C)</th>
<th>UTS (ksi)</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of Area (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>505/1</td>
<td>22</td>
<td>500</td>
<td>242</td>
<td>10</td>
<td>13</td>
</tr>
<tr>
<td>505/2</td>
<td>22</td>
<td>596.1</td>
<td>240.6</td>
<td>17</td>
<td>16</td>
</tr>
<tr>
<td>505/3</td>
<td>22</td>
<td>568.8</td>
<td>242.7</td>
<td>14</td>
<td>16</td>
</tr>
<tr>
<td>ASTM B171</td>
<td>22</td>
<td>586.1</td>
<td>213.7</td>
<td>10</td>
<td>N/A</td>
</tr>
</tbody>
</table>
The wrought NAB material tensile results exceeded the minimum yield and tensile strength requirements described by wrought plate ASTM B171 specification. The results did not show a significant difference between the transverse and the longitudinal directions. This phenomenon indicated that the wrought material could have been cross rolled which is a desired application performed at commercial plants to improve mechanical properties and to make more isotropic. The as-cast tensile results showed lower values when compared to the wrought material which was expected due to the fact that the material had not received any cold work or heat treatment. The as-cast tensile results indicated the averaged ultimate tensile strength (UTS) of 562.8 MPa which did not meet the ASTM B171 specification (UTS) requirement of 586.1 MPa. The percent elongation of the cold worked material averaged 21.7 percent and the percent elongation of the as-cast material averaged 13.7 percent. Both material percent elongations were found above the ASTM E 171 set specification. Further, the fractures of the tensile specimens for both materials were analyzed with Scanning Electron Microscope (SEM) as shown in Figures 22 to Figure 26. Figure 22 through Figure 26 showed a dull fibrous fracture appearance with small dimples at the center of the specimens and shear lips around indicating of a ductile fracture. A brittle material would have shown mechanism such as cleavage or inter-granular fracture. The close up SEM micrographs in Figures 24 and 25 revealed micro-void coalescence throughout the fracture and confirmed a ductile fracture.
Figure 22. Longitudinal tensile sample 12.7 (mm) diameter wrought material

Figure 23. Transverse tensile sample 6.35 mm diameter wrought material
Figure 24. As-cast tensile sample 12.7 (mm) diameter

Figure 25. SEM micrograph of wrought tensile sample 12.7 (mm) diameter
5.3 Hardness Results

The Rockwell hardness measurements of the wrought, as-cast and annealed material were carefully measured at cross section area perpendicular to the rolling direction as indicated in the diagram in Figure 24. It is important to point out that the material’s microstructure directionality would apply primarily to the wrought material due to the cold work treatment. The through thickness direction for the wrought material in this study would represent a more consistent measurements to be compared with the annealed and as-cast material.
Figure 27. Material hardness measurement location

The Rockwell hardness results for the wrought material are listed in Table 10. Table 11 lists the annealed and as-cast results. The wrought material Rockwell ‘B’ hardness average measured value is 86. The water quenched, air cooled, furnace cool and as-cast material Rockwell ‘B’ hardness average values measured 87, 90, 88 and 77 respectively. The wrought material hardness results were higher than the as-cast material results due to the cold work conditions. However, the wrought and the as-cast material both showed lower hardness values when compared to the hardness values of the water quenched, air cooled, furnace cool conditions. The hardness results possibly indicate the wrought material prior to being cold worked could have had an as-cast microstructure obtained from slow cooling. The higher hardness values measured for the water quenched, air cooled and furnace cool could be caused by a fast cooling forming a bainitic/martensitic microstructure as indicated by previous researchers and illustrated in the diagram in Figure 4.
Table 10. Wrought material hardness measurements

<table>
<thead>
<tr>
<th>Test Number</th>
<th>Wrought NAB</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>84</td>
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<tr>
<td>2</td>
<td>87</td>
</tr>
<tr>
<td>3</td>
<td>86</td>
</tr>
<tr>
<td>4</td>
<td>89</td>
</tr>
<tr>
<td>5</td>
<td>83</td>
</tr>
<tr>
<td>6</td>
<td>86</td>
</tr>
<tr>
<td>AVE</td>
<td>86</td>
</tr>
</tbody>
</table>

Table 11. Annealed and as-cast hardness measurements

<table>
<thead>
<tr>
<th>Test Number</th>
<th>Water Quenched</th>
<th>Air Cooled</th>
<th>Furnace Cooled</th>
<th>As-Cast</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>88</td>
<td>90</td>
<td>88</td>
<td>78</td>
</tr>
<tr>
<td>2</td>
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<td>6</td>
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<td>78</td>
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<tr>
<td>AVE</td>
<td>87</td>
<td>90</td>
<td>88</td>
<td>77</td>
</tr>
</tbody>
</table>
5.4 Microstructure Analysis

Optical micrographs obtained from wrought, as-cast, air cooled, water quenched, and furnace cooled NAB samples are shown through Figures 28 to 38. The optical micrographs in Figures 28 and 29 show the wrought material longitudinal and transverse directions that appeared identical with no strong sense of grain directionality indicating the commercial wrought material plate was cross rolled which supports the tensile results obtained in longitudinal and transverse directions. In Figure 30 which is a close up of Figure 29 can be seen the presence of the $\kappa$ phases into the $\alpha$ metrics. The absence of the $\kappa_1$ phase indicated that the iron content was kept below 5 wt. % [30,31] as shown by the chemical analyses results in Table 6.

![Optical micrograph of wrought sample rolling direction](image)

Figure 28. Optical micrograph of wrought sample rolling direction
Figure 29. Optical micrograph wrought transverse direction sample 1

Figure 30. Optical micrograph wrought NAB showing $\alpha$, $\kappa$, retained $\beta$-martensite phases
The presence of the martensite bainite formation was due to a not full transformed β phase creating pockets of a dark mixture marked martensite shown in Figure 30. The measurement of the cast dendritic cell size, outlined by κ phases, and the α grain size were determined using the ASTM E112 comparison method. The α grain size for the wrought material measured value is ASTM 7.0 and the α grain size for the as cast material measured value is ASTM 5.0. The grain size is determined at a magnification of 100 X. The relationship between the Grain Size Number (GSN) and the number of grains/in² is given by the equation 4

\[ n=2^{N-1} \]  

(eq 4)

Where  
N = ASTM Grain Size Number
n = number of grains/in² at a magnification of 100 X

The evaluation of the ASTM (GSN) is carried out with the aid of standard grain size charts or with special eyepieces[46].
The as-cast material optical micrographs shown in Figures 31 and the close up Figure 32 were compared to the air cooled, water quenched, and furnace cooled material and they close-up. Gradual increase of martensite or retained β prime microstructure as the fast cooling was adopted was observed. This is in agreement with the research performed by K. Oh-Ishi and T.R. McNelley illustrated in the diagram in Figure 4 showing that the fast cooling favors martensite/bainite formation. Therefore this explains the increase of the higher tensile and hardness results of the water quenched, air cooled, furnace cool when compared to the lower tensile and hardness results of the as-cast material. From the as-cast, air cooled, water quenched, and furnace cooled NAB microstructure analysis it was learned that the as-cast microstructure is most similar to the wrought NAB microstructure. The as-cast α grain size measurement were determined using the ASTM E112 comparison method. The as-cast α grain size is ASTM 7.0.
This showed no a significant difference with the grain size measurement of wrought NAB of ASTM 5.0. The grain aspect ratio was not measured but it is expected to be different even though it is not showing a variation in mechanical properties. This is experienced during hot rolling the as-cast microstructure breaks down and replace it with recrystallized microstructure.

![Image of optical micrograph as-cast NAB showing α, κ, retained β phases](image)

**Figure 32. Optical micrograph as-cast NAB showing α, κ, retained β phases**
Figure 33. Optical micrograph NAB Furnace Cool grain microstructure

Figure 34. Optical micrograph Furnace Cool NAB showing α, κ, retained β phases
Figure 35. Optical micrograph NAB water quenched grain microstructure

Figure 36. Optical micrograph Water Quenched NAB showing $\alpha$, $\kappa$, retained $\beta$ phases
Figure 37. Optical micrograph Air Cooled NAB grain microstructure

Figure 38. Optical micrograph Air Cooled NAB showing $\alpha$, $\kappa$, retained $\beta$ phases
5.5 Gleeble Results

Figures 38 to 53 show true stress-true strain flow curve of selected true strain testing conditions of 0.1, 0.11, 0.16 and 0.22. Table 12 lists true stress values at true strain values approximately of 0.1, 0.11, 0.16 and 0.22 and percent elongation. Formations of cracks were observed on the outside edges of the specimen at true strain of 0.21 which corresponds approximately to 20% specimen reduction and not observed a rise in temperature during all the tests. As shown in Table 12 was observed that the strain rates as expected did not have a significant effect on the true stress values as the material deformed. The material percent elongation values reported in Table 12 shows evidence of the material’s ductility. It can be seen that has the material is deformed and strain harden the percent elongation increases from 7.6% up to 15.9% from an initial true strain of 0.1 to a final true strain 0.22 respectively. It can be concluded the material during deformation at room temperature 22°C it strain harden reducing ductility and the true stress become solely a function of strain due to the strain hardening phenomenon.
Figure 39. True stress true strain curve, 22 °C and 0.1 s⁻¹

Figure 40. True stress true strain curve, 22 °C and 0.3 s⁻¹
Figure 41. True stress true strain curve, 22 °C and 0.3 s\(^{-1}\)

Figure 42. True stress true strain curve, 22 °C and 0.3 s\(^{-1}\)
Figure 43. True stress true strain curve, 22 °C and 0.3 s⁻¹

Figure 44. True stress true strain curve, 22 °C and 1.0 s⁻¹
Figure 45. True stress true strain curve, 22 °C and 1.0 s⁻¹

Figure 46. True stress true strain curve, 22 °C and 1.0 s⁻¹
Figure 47. True stress true strain curve, 22 °C and 1.0 s\(^{-1}\)

Figure 48. True stress true strain curve, 22 °C and 5.0 s\(^{-1}\)
Figure 49. True stress true strain curve, 22 °C and 5.0 s\(^{-1}\)

Sample 8. \(\varepsilon = 0.11\) \(\dot{\varepsilon} = 5.0\)

Figure 50. True stress true strain curve, 22 °C and 5.0 s\(^{-1}\)

Sample 12. \(\varepsilon = 0.16\) \(\dot{\varepsilon} = 5.0\)
Figure 51. True stress true strain curve, 22 °C and 5.0 s⁻¹

Figure 52. True stress true strain curve, 22 °C and 20 s⁻¹
Figure 53. True stress true strain curve, 22 °C and 20 s\(^{-1}\)

Sample 9. \(\varepsilon = 0.11, \dot{\varepsilon} = 20\)

Figure 54. True stress true strain curve, 22 °C and 20 s\(^{-1}\)

Sample 13. \(\varepsilon = 0.16, \dot{\varepsilon} = 20\)
Table 12. Effect of strain rate on stress at true strain of 0.1, 0.11, 0.16 and 0.22 of as-cast material

<table>
<thead>
<tr>
<th>Material (Samples)</th>
<th>True Strain (ε)</th>
<th>Strain Rate (ε̇) s⁻¹</th>
<th>True Stress (MPa)</th>
<th>Reduction (%)</th>
<th>Temp deg C</th>
</tr>
</thead>
<tbody>
<tr>
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<td>0.1</td>
<td>621</td>
<td>7.6</td>
<td>22</td>
</tr>
<tr>
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<td>639</td>
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<td>22</td>
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<tr>
<td>3</td>
<td>0.16</td>
<td>1</td>
<td>637</td>
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</tr>
<tr>
<td>4</td>
<td>0.11</td>
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<td>5</td>
<td>0.16</td>
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</tr>
<tr>
<td>6</td>
<td>0.16</td>
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<tr>
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<td>0.3</td>
<td>778</td>
<td>11.7</td>
<td>22</td>
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<td>777</td>
<td>11.3</td>
<td>22</td>
</tr>
<tr>
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<td>5</td>
<td>760</td>
<td>11.2</td>
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</tr>
<tr>
<td>13</td>
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<td>20</td>
<td>773</td>
<td>11.6</td>
<td>22</td>
</tr>
<tr>
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<td>840</td>
<td>15.3</td>
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<tr>
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<td>835</td>
<td>15.6</td>
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<tr>
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<td>22</td>
</tr>
<tr>
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<td>0.22</td>
<td>20</td>
<td>857</td>
<td>15.9</td>
<td>22</td>
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</tbody>
</table>
Chapter 6. Discussion

A comparison between the rolled microstructure presented in Figure 28 and 29 and as-cast microstructure shown in Figures 31 displays that cold rolling processing of nickel aluminum bronze yields a refined, inhomogeneous microstructure. The grain size was found to be 7 for cold rolled plate and 5 for as-cast adopting the comparison method in ASTM E 112. The close analysis of the grain size evolution in cold rolling proves that is very different than it is in hot rolling as shown in Figure 3. The as cast microstructure of the ingot breaks down in hot rolling process and it allows for recrystallization mechanism to occur to generate new refined grains. The effect of these conditions on flow stress can be seen in Table 12.

Table 13. ASTM E 112 # and grain/in² relationship [46]

<table>
<thead>
<tr>
<th>ASTM E 112</th>
<th>0</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
<th>6</th>
<th>7</th>
<th>8</th>
<th>9</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>Grains/in²</td>
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<td>1</td>
<td>2</td>
<td>4</td>
<td>8</td>
<td>16</td>
<td>32</td>
<td>64</td>
<td>128</td>
<td>256</td>
<td>512</td>
</tr>
</tbody>
</table>

The cold rolled material having an ASTM grain size of 7 corresponds to 64 grains per square inch, which is much finer microstructure than the as-cast material of ASTM grain size 5 which corresponds to 16 grains per square inch. This microstructure is created during hot rolling, which is implemented to reduce ingots to blooms and eventually to slabs up-to a desired thickness. Then the plate is cold rolled to a sheet of a given final thickness with desired mechanical properties. Moreover, these grains are further deformed during cold rolling schedule. An important parameter to know would be the grain aspect ratio which was not calculated in this analysis. It would provide additional information of the grain formation after recrystallization during hot rolling and prediction of its value after cold rolling sequence designed for further thickness reduction. The large dark regions seen in Figures 30 for the cold rolled microstructure
and dark regions seen in Figure 32 for the as-cast microstructure showed a minimum presence of the martensite when compared to the heat treated microstructures. This indicates that the strength requirement of the cold rolled material is achieved mainly by work hardening than by heat treatment. Additional electron microscopy analysis would be needed to better understand the material microstructure and its influence on mechanical properties. The flow curves obtained from the Gleeble compression tests showed that material can be deformed up to 15.9 % as shown in Table 13. From the tensile values reported in Tables 7, 8 for the cold rolled in transverse and rolling directions respectively and in Table 9 for the as-cast material. It was learned that both materials even though are classified as ductile materials their ductility is limited at room temperature and it is recommended to utilize deformation at elevated temperature. However, the rolled material which was processed after hot rolling to improve its ductility through the grain recrystallization process, is superior to as-cast material. Figure 56 shows tensile data plotted in engineering stress engineering strain diagram and in Figure 57 shows the same data plotted in true stress true strain.
It can be seen that the engineering plot and the true stress strain plot have similar curve up to reaching the ultimate value (UTS) and then ending with a sudden rupture. The material strain harden just little above to the UTS and then reaches failure due to lack of ductility enhanced by high amount of Aluminum in this alloy.
The research conducted by J.A. Warton et., al. on the same alloy showed that the contribution of 9 to 12 wt % aluminum results in a higher yield strength which is attributed to a hard face center cubic (fcc) phase which enhances the properties of the castings as well as hot working in wrought alloys. The presence of Nickel up to 6 wt % also improves yield strength [47].

In terms of the rolling process, based on the results of the chemical analysis, Gleeble testing and tensile testing performed in this study they prove that due to the low ductility this material has to be hot rolled to a minimum thicknesses and finishing has to be done with a cold rolling reduction per pass lower than 15.9 %. The exact rolling and annealing schedule can be designed for the specific working rolls. This part can be done knowing what cold rolling line will be available for this study or it has to be assumed for modeling purposes.
Conclusions

The scope of this study was to determine flow stress, as a function of strain, and strain rate at room temperature. This information will be used in a finite element (FE) package as input for future work to recreate cold rolling mill schedules for the reduction of 50.8 mm (2 inches) to 25.4 mm (1 inch) to 5 Nickel Aluminum Bronze plates. This study is focused on wrought as well as-cast material and it is based on an extensive chemical, optical, mechanical, and microstructure property characterization. This work has been applied first to a wrought plate remnant and second to an as-cast material obtained by re-melting NAB scraps of similar chemistry of the wrought material after ingots were formed and analyzed at different annealing conditions. Important conclusions and observations from this study are listed below.

1. The true stress true strain flow curves can be implemented into the FE packages for deformation studies

2. The microstructure analysis performed on rolled material and on the set of heat treating condition as furnace cooled, water quenched, air cooled and as cast material revealed that the formation of martensite in the as-cast microstructure is similar to the martensite formation in rolled material.

3. The Gleeble deformation testing results reported in Table 13 shows that the as-cast microstructure can be deformed up to 15.9 % reduction. Therefore the cold rolling schedules need to be designed for a specific rolling mill with necessary annealing steps in mind.
4. Depending on the size of the working rolls, the per pass reduction as well as the load or pressure of deformation will be calculated in the second phase of this project.
References


