Characterization of FCC Al-Cu-Ni-Mn-Ag High Entropy Alloy

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CHARACTERIZATION OF FCC AL-CU-NI-MN-AG HIGH ENTROPY ALLOY

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Master’s Program in Metallurgical and Materials Engineering

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Dedication

This thesis is dedicated to those whose unwavering support and encouragement have been my guiding lights throughout this academic journey. To my family, especially my mother, whose love and sacrifices have been the bedrock of my existence. Your unwavering belief in my potential has fueled my determination to reach new heights. To my mentors and advisors, whose wisdom, guidance, and patience have shaped my intellectual growth. Your passion for knowledge has been contagious, and I am grateful for all the lessons learned. To my friends, whose friendship provided the much-needed respite from the rigors of research. Your laughter and companionship have been a constant source of joy. In loving memory of my father, whose inspiration and wisdom continue to resonate in my thoughts. Though physically absent, your spirit lives on in the pages of this work.

This thesis is a testament to the collective efforts of those who have touched my life, leaving indelible imprints on my academic and personal journey.
CHARACTERIZATION OF FCC AL-CU-NI-MN-AG HIGH ENTROPY ALLOY

by

GINA ZAVAALA ALVARADO, B.S.

THESIS

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Abstract

The effect of Ag on the microstructure developed in Al-Cu-Ni-Mn alloy has been determined. The modified Al-Cu-Ni-Mn alloy by Ag addition shows the presence of three microconstituents consisting of phases rich in (1) Cu, (2) Ni, and (3) Ag. The foregoing alloys heated for 24 hours from 600 to 1000 °C show excellent oxidation resistance. Oxide formation and microstructural changes of the alloy have been characterized by elemental mapping and X-ray diffraction (XRD). Results show that the elements of Al and Mn preferentially oxidize while Cu and Ni provide oxidation resistance to the alloy. Hardness was taken on the alloy and its microconstituents before and after oxidation. In the as-received condition, the microconstituent with the highest hardness is Ag-rich (429 HV), and the lowest with Cu-rich (392 V). After oxidation at 1000 °C, Ni-rich has the highest (447 HV) and Ag-rich the lowest (112 HV) hardness. Grain size measurement was performed to observe the change and amount of the microconstituents. Hardness and grain size measurements have been taken to make a Hall-Petch relationship.
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1. Introduction

1.1. History

The advancement of techniques for generating novel materials has played a vital role in the progress of humanity [1]. The use of metals and alloys has significantly enhanced our capabilities across various industries such as transportation, food production, agriculture, machinery, defense, etc [2]. From the first use of bronze to nanomaterials, technological improvements have been made possible thanks to the development of new materials and processes.

1.2. Background Concept

Conventional alloys are developed with a base element. They use one or two principal elements mixed with a small number of other elements, also known as trace elements, to improve specific properties [2] [3]. Many alloys are considered to be designed under this approach, such as Al alloys [4] or Fe alloys [5].

The field of material science has seen a noteworthy shift in focus toward the development of novel materials with astonishing properties. The concept of crystalline multi-principal element alloys (MPEAs) and high entropy alloys (HEAs) was first introduced almost 2 decades ago by Cantor et al. [6] and Yeh et al. [7] independently. HEAs were a sharp contrast in alloy design in comparison to the conventional method as they contain multiple elements (five or more) in an equal or near-equal atomic percentage (at.%). While MPEAs share the same concept, they do not have equiatomic proportions. HEAs were named by Yeh et al. [7] because the solid solution state exhibits greater mixing entropies compared to conventional alloys [8].

The fundamental principle behind high entropy alloys lies in the multiple elements in equiatomic proportions that result in a complex solid solution. Each element contributes different
properties and interactions in the lattice structure, which leads to a higher number of potential atomic arrangements. This increase in atomic configuration enhances disorder and configurational entropy in the alloy. This results in exceptional mechanical [9], chemical [10], and physical [11] properties in comparison to conventional alloys [12]. HEAs are currently used in a wide range of applications, from biomedical uses [13] to extreme environments [14].

HEAs may be classified based on the degree of entropy. Entropy is based on a thermodynamic system that will be in equilibrium when Gibbs free energy (G) is reduced [15]. The following equation establishes the relationship of the free energy of a given system:

\[
\Delta G = \Delta H - T \Delta S
\]  

Equation (1) shows that enthalpy (H) and entropy (S) have a direct relationship with the equilibrium state at a given temperature (T).

To predict the lowest energy state of a given alloy, the lowest mixing free energy (\(\Delta G_{mix}\)) can be determined when the free energy changes are compared to other states. The differences in free energy of mixing [15] can be related by:

\[
\Delta G_{mix} = \Delta H_{mix} - T \Delta S_{mix}
\]  

Where the enthalpy of mixing (\(\Delta H_{mix}\)) and entropy of mixing (\(\Delta S_{mix}\)) have a direct relationship between \(\Delta G_{mix}\) in equation 2.

According to Boltzmann’s hypothesis, the configurational entropy of mixing (\(\Delta S_{mix}\)) of an n-element equimolar alloy changing from an elemental to random solution state is calculated from:

\[
\Delta S_{mix} = R \ln n
\]  

Where R is the gas constant of 8.31 J/K·mol [7].
When mixed at equimolar concentrations, $\Delta S_{\text{mix}}$ increases with a larger number of alloying elements, $n$, according to equation 3. A visual representation of equation 3 is shown in Figure 1. Entropy of mixing as a function of number of elements in equimolar alloys The minimum requirement is having five elements in HEAs since it represents the threshold where the mixing entropy becomes sufficiently significant to offset the mixing enthalpy in most alloy systems. Consequently, this guarantees the formation of solid solution phases [15].

If the enthalpy remains constant, increasing the entropy with a high number of elements will consequentially decrease Gibbs free energy, stabilizing the alloy system. If the temperature increases, $G$ will be reduced even more. Hence, a high-entropy alloy system could be thermodynamically stable, even at high temperatures.

Figure 1. Entropy of mixing as a function of number of elements in equimolar alloys [15]
Alloys can then be categorized depending on the magnitude of their mixing entropy in the random solution state. Alloys can be classified by their configurational entropy ($\Delta S_{conf}$) using the following conditions [15]:

- Low-entropy alloys: $\Delta S_{conf} \leq 0.69R$
- Medium-entropy alloys: $1.61R \geq \Delta S_{conf} \geq 0.69R$
- High-entropy alloys: $\Delta S_{conf} \geq 1.61R$

Low-entropy alloys usually have one or two major elements and are considered to be the traditional type of alloys. Medium-entropy alloys have two to four major elements while high-entropy alloys have at least five major elements.

This classification organizes alloys based solely on equiatomic proportions. However, there’s the possibility that alloys might not be equiatomic. In such a case, Yeh [16] proposed an alternative method to evaluate the configurational entropy of an alloy ($\Delta S_{configuration}$). The configuration entropy of alloys that do not have equiatomic proportions can be classified with the following conditions:

- Low-entropy alloys: $\Delta S_{configuration} \leq 1R$
- Medium-entropy alloys: $1R \geq \Delta S_{configuration} \geq 1.5R$
- High-entropy alloys: $\Delta S_{configuration} \geq 1.5R$
1.3. Core Effects

Several review papers [3] [16] [17] [18] [19] [20] explain the four primary effects that dominate HEAs:

a) High entropy effect
b) Sluggish diffusion effect
c) Lattice distortion effect
d) ‘Cocktail’ effect
1.3.1. High Entropy Effect

The thermodynamic effect of high entropy is thought to be the most unique and significant effect on HEAs. This effect creates a less sophisticated microstructure and enhances the formation of solution phases [16].

In a solid state of an alloy, there are three categories of competing states: intermetallic compounds, elemental phases, and solid-solution phases. Elemental phase is the final solid solution based on just one element. Intermetallic compound refers to the stochiometric compounds having specific superlattices. Lastly, solid solution is the phase that contains the mixing of all elements in a crystal structure.

According to the second law of thermodynamics, the equilibrium state is achieved by the state with the lowest mixing free energy ($\Delta G_{mix}$) [16]. The relationship between H and S of each competing states are the following:

- Elemental phases: small $\Delta H_{mix}$ and $\Delta S_{mix}$
- Compound phases: large $\Delta H_{mix}$ but small $\Delta S_{mix}$
- Solid-solution phases: medium $\Delta H_{mix}$ and high $\Delta S_{mix}$

Following equation (1), a high $\Delta S$ and relatively low $\Delta H$ will yield a lower free energy. In the case of the competing states, a solid-solution phase would be predominant in the equilibrium phase. Based on equation 3, as the number of components increases, the change in entropy will also increase [21].

The thermodynamic effect of high entropy stabilizes single-phase solid solutions as the equimolar alloy favors them over intermetallic compounds, which are brittle and complex to
analyze. This effect also avoids the development of many stochiometric compounds and increases the formation of solution-type phases [16].

![Figure 3. Visual representation of the high entropy effect [21]](image)

1.3.2. Sluggish Diffusion Effect

Diffusion is necessary for the formation of new phases in HEAs. However, HEAs have a limited vacancy concentration for substitutional diffusion compared to traditional alloys. This is due to the vacancies being associated with an excess mixing entropy and a positive enthalpy of formation [16]. During diffusion, a vacancy is surrounded by different elements atoms. The low lattice potential energy sites then serve as locks and delay atomic diffusion, creating a sluggish diffusion effect. Tsai et al [22] showed direct evidence of the sluggish diffusion effect on a Co-Cr-Fe-Mn-Ni HEA. It was also shown that the degree of sluggish diffusion relates to the number of elements in the alloy.
The sluggish diffusion effect provides valuable advantages found in literature [23] [24] [25]: nanoparticles, increased recrystallization temperature, slower grain growth, and increased creep resistance.

![Visual representation of the sluggish diffusion effect in high entropy alloys](image)

**Figure 4.** Visual representation of the sluggish diffusion effect in high entropy alloys [21]

### 1.3.3. Lattice Distortion Effect

The structural effect of severe lattice distortion occurs from the different atom sizes inside the matrix. In the HEAs, every atom is bordered by different elements, leading to lattice stress and strain due to their size differences. The different types of bonding and crystal structure also play a role in lattice distortion [16]. In comparison with traditional alloys, HEAs have a higher lattice distortion effect due to the higher number of elements. The effect of lattice distortion increases hardness and strength due to the large solution hardening in the distorted lattice [18].
Figure 5. Visual representation of a lattice with atoms that have different sizes [21]

1.3.4. ‘Cocktail’ Effect

The cocktail effect addresses the enhanced properties obtained by alloying [18]. The term “cocktail” was first used in the alloy field by Ranganathan [26] to highlight the enjoyment from alloy design and development. Using more of a descriptive term, Ranganathan explained the positive outcome and increased properties by mixing different element combinations [16].

1.4. Types of High Entropy Alloys

Since the breakthrough of high entropy alloys, they have been classified by different categories, such as entropy level, mechanical behavior, or elements. The classifications of HEAs based on
elemental composition are refractory metal, light metal, precious metal brass and bronze, 3d transition metals, and interstitial compound HEAs [18]. HEAs have shown promising high-temperature properties when they are composed of refractory metals with a body-centered cubic (BCC) structure. Since the discovery of HEAs, research has been mainly oriented into HEAs with BCC elements [27] [28] [29]. This is due to BCC elements having a higher melting point compared to face-centered cubic (FCC) elements. The lack of research done on specifically FCC high entropy alloys has opened the opportunity to create a new HEA classification.

1.5. FCC High Entropy Alloys

FCC HEAs have been recently studied and have potential applications in various fields. The theory of strengthening FCC HEAs has been explained by C. Varvenne et al. [30] using a Ni-Co-Fe-Cr-Mn FCC. Gang Quin et al. [31] showed strengthening by Mo addition in Co-Cr-Fe-Mn-Ni FCC HEAs. Another recent study focused on the effect of strain, strain rate and temperature of Co-Cr-Fe-Mn-Ni FCC HEAs by Raturi et al [32].

Oxidation behavior and microstructural characterization of Al-Cu-Ni-Mn and Al-Cu-Ni-Mn-Si non-BCC HEA has been reported by Hitter et al. [33]. Oxidation behavior was analyzed at 600, 700, and 800 °C. Three phases were identified in the as-received condition based on chemical composition. The oxides formed on the alloy’s surface were analyzed and hardness measurements were taken on the alloys and the phases.
1.5.1. Basic Properties of FCC High Entropy Alloys

Table 1 shows basic properties of the elements used in this study including melting point, crystal structure, atomic radius, lattice parameter, and density. All metals have a FCC structure, expect manganese which has a cubic structure. The elements outlined have a melting point ranging from 660.4 to 1453 °C. Their atomic radius ranges from 1.12 to 1.445 Angstroms and 3.5167 to 8.931 Angstroms for their lattice parameter. The metal with the highest density is Silver with 10.501 g/cm³ while Aluminum has the lowest at 2.70 g/cm³.

Table 1. Properties of the metals used in this work [34]

<table>
<thead>
<tr>
<th>Element</th>
<th>Crystal Structure</th>
<th>Melting Point (Celsius)</th>
<th>Atomic Radius (Angstroms)</th>
<th>Lattice Parameter (Angstroms)</th>
<th>Density (g/cm³)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Aluminum</td>
<td>Face Centered Cubic</td>
<td>660.4</td>
<td>1.432</td>
<td>4.09</td>
<td>2.70</td>
</tr>
<tr>
<td>Copper</td>
<td>Face Centered Cubic</td>
<td>1084.9</td>
<td>1.278</td>
<td>3.6151</td>
<td>8.933</td>
</tr>
<tr>
<td>Nickel</td>
<td>Face Centered Cubic</td>
<td>1453</td>
<td>1.243</td>
<td>3.5167</td>
<td>8.912</td>
</tr>
<tr>
<td>Manganese</td>
<td>Cubic</td>
<td>1244</td>
<td>1.12</td>
<td>8.931</td>
<td>7.3</td>
</tr>
<tr>
<td>Silver</td>
<td>Face Centered Cubic</td>
<td>961.8</td>
<td>1.445</td>
<td>4.09</td>
<td>10.501</td>
</tr>
</tbody>
</table>
1.5.2. Hume-Rothery Rules

As previously mentioned, one of the aspects of high entropy alloys is that due to their high mixing entropy, solid solutions are preferentially formed over intermetallic phases. If the Hume-Rothery rules are followed, it is more likely that the alloy will create a solid solution [35]. To confirm that the high-entropy alloy will create solid solutions, the Hume-Rothery rule can be applied. The rules are the following:

I. Similar atomic radii (difference of less than 15%)
II. Similar crystal structure
III. Similar electronegativity
IV. Similar valency

Table 2. Binary combinations following the Hume-Rothery rules

<table>
<thead>
<tr>
<th>Element</th>
<th>Crystal Structure</th>
<th>Atomic Radius</th>
<th>Electronegativity</th>
<th>Valence</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>FCC</td>
<td>1.432</td>
<td>1.61</td>
<td>3</td>
</tr>
<tr>
<td>Cu</td>
<td>FCC</td>
<td>1.278</td>
<td>1.9</td>
<td>3</td>
</tr>
<tr>
<td>Ni</td>
<td>FCC</td>
<td>1.243</td>
<td>1.91</td>
<td>4</td>
</tr>
<tr>
<td>Mn</td>
<td>Cubic</td>
<td>1.12</td>
<td>1.55</td>
<td>7</td>
</tr>
<tr>
<td>Ag</td>
<td>FCC</td>
<td>1.445</td>
<td>1.93</td>
<td>3</td>
</tr>
<tr>
<td>Al-Cu</td>
<td>Same</td>
<td>12%</td>
<td>15%</td>
<td>Similar</td>
</tr>
<tr>
<td>Al-Ni</td>
<td>Same</td>
<td>15%</td>
<td>16%</td>
<td>Similar</td>
</tr>
<tr>
<td>Al-Mn</td>
<td>Not the same</td>
<td>28%</td>
<td>4%</td>
<td>Not similar</td>
</tr>
<tr>
<td>Al-Ag</td>
<td>Same</td>
<td>1%</td>
<td>17%</td>
<td>Similar</td>
</tr>
<tr>
<td>Cu-Ni</td>
<td>Same</td>
<td>3%</td>
<td>1%</td>
<td>Similar</td>
</tr>
<tr>
<td>Cu-Mn</td>
<td>Not the same</td>
<td>14%</td>
<td>23%</td>
<td>Not similar</td>
</tr>
<tr>
<td>Cu-Ag</td>
<td>Same</td>
<td>12%</td>
<td>2%</td>
<td>Similar</td>
</tr>
<tr>
<td>Ni-Mn</td>
<td>Not the same</td>
<td>11%</td>
<td>23%</td>
<td>Not similar</td>
</tr>
<tr>
<td>Ni-Ag</td>
<td>Same</td>
<td>14%</td>
<td>1%</td>
<td>Similar</td>
</tr>
<tr>
<td>Mn-Ag</td>
<td>Not the same</td>
<td>22%</td>
<td>20%</td>
<td>Not similar</td>
</tr>
</tbody>
</table>
The possible binary combinations with the elements used on this study based on the Hume-Rothery rules are shown in Table 2. The combinations of Al-Mn and Mn-Ag do not follow the rules as they have different crystal structures, atomic radii, electronegativity, and valence. However, several combinations such as Ni-Ag and Cu-Ni do follow the Hume-Rothery Rules.

1.5.3. Oxidation of FCC High Entropy Alloys

The study of oxidation is an elemental aspect of this work as it gives the alloy the potential use at higher temperatures while maintaining its properties. Some high entropy alloys have been extensively researched in their oxidation-resistant properties. For example, it was shown that HEAs in the Al-Co-Cr-Fe-Ni system are heat and corrosion resistant, and that the elements that contribute more to oxidation resistance in high-entropy alloys are Al and Cr [36].

Other two studies analyzed the microstructure [29] and oxidation behavior [28] of Nb-Cr-V-W-Ta HEA. In the as-received condition, five phases were identified based on chemical composition. The oxidation behavior was analyzed in between 600 and 1400 °C in static and cyclic oxidation. Cylindrical crystals of W and Ta and granular crystals of Nb and Cr oxides were found at 1000 and 1200 °C while whiskers of V oxide were identified at 1200 °C. Although the authors designed the HEA with BCC elements, it shows a great example of oxidation behavior analysis and microstructural characterization.
1.5.4. Hall-Petch Relationship

The Hall-Petch relation establishes that the smaller grain size results in a greater flow-stress. The flow-stress is the stress in a tension test that corresponds to a defined value of strain [35]. The empirical relationship can be expressed in the form of:

\[ H = H_0 + k_H d^{-1/2} \]  \hspace{1cm} (4)

Where the hardness is \( H \), the average grain diameter is \( d \), the slope is \( k_H \), and the intercept of the line with the ordinate axis is \( H_0 \), which corresponds to the hardness found at a hypothetical infinite grain size. [35]

When tensile data is tested at room temperature and different strains, the following relationship follows:

\[ \sigma = \sigma_o + k d^{-1/2} \]  \hspace{1cm} (5)

Where the flow stress is \( \sigma \), and \( k \) and \( \sigma_o \) are constants equivalent to \( H_0 \) and \( k_H \) from equation 4.

\( k \) is also referred to as the strengthening coefficient, which is unique to each material. This linear relationship in equation 5 was first proposed by Hall et al. [37] and Petch et al. [38], which is commonly known as the Hall-Petch equation.

Chen et al. [39] developed a Hall-Petch relationship of a Hf-Nb-Ta-Ti-Zr alloy after it was subjected to cold rolling and annealing from 1000 to 1200 °C for 24 hours. A Hall-Petch relationship was found in the form of \( \sigma = 942 + 270 d^{-0.5} \), with a friction stress of 942 MPa and a coefficient of 270 MPa/mm\(^{1/2}\). For the resulting slope, two activation energies were found: 205 kJ/mol between 1000 and 1100 °C and 401 kJ/mol between 1100 and 1200 °C. The higher activation energy relates to a greater degree of disorder as grain boundary movement requires volume
diffusion through the lattice. The rate is then defined by the slowest species, which is Ta and Nb in the case of the Hf-Nb-Ta-Ti-Zr alloy. These values show that the alloy has a low grain-boundary strengthening and a high intrinsic stress in comparison to tungsten alloys and stainless steel, respectively.

The strengthening mechanism of FCC high-entropy alloys was analyzed by Yoshida et. al [40] by investigating the effect of friction stress and their Hall-Petch relationship using subsystems of Co-Cr-Fe-Mn-Ni HEA. The Hall-Petch relationships were experimentally obtained from high and medium entropy alloys with FCC elements.
2. Research Problem

2.1.1. Problem Statement

This thesis aims to contribute to research by characterizing a new HEA: Al-Cu-Ni-Mn-Ag. Characterization of the alloy was done in its as received condition using back scattered electrons in a scanning electron microscope (SEM). Static oxidation was performed at 600, 700, 800, 900, and 1000 °C for 24 hours to analyze oxide formation, microstructural and chemical changes with respect to temperature. The oxide development was studied using SEM and x-ray diffraction. Elemental color mapping on the energy-dispersive spectroscopy (EDS) was used to determine the locations of the elements after oxidation. Microstructural characterization revealed the changes in the microconstituents after oxidation in comparison with already observed microstructures in the as received condition. An attempt has been made to use the Hall-Petch equation to describe the changes in hardness values as a function of grain sizes of the microconstituents. An in-depth grain size measurement of each identified microconstituent was also performed in the as received condition and after oxidation. This procedure was extended to include the study indicating which of the microconstituents contributed more to the hardening.

2.1.2. Significance of the Study

Research focused on non-BCC HEAs may be important for two main reasons: (a) low temperature applications and (b) conservation of metals. By designing HEAs with FCC elements, the temperature range in which the alloy can be used decreases in comparison to BCC HEAs, creating new opportunities for low temperature applications. Conservation of metals can be obtained if a non-BCC HEA substitutes a conventional alloy with one base element. For instance,
a high copper content brass could be replaced by a 25 at.% Cu HEA if the desired properties are achieved. The substitution would result in a decrease of copper needed, contributing to metal conservation.

### 2.1.3. Research Objectives

The research objectives of this research are the following:

i. Investigate a novel non-BCC high entropy alloy.

ii. Characterize the as-received alloy and identify the phases present.

iii. Evaluate the oxide layer and identify predominant oxides.

iv. Evaluate the microstructure of the alloy after oxidation.

v. Hardness measurements of the alloy and its microconstituents.

vi. Analyze grain growth after oxidation.

vii. Plot hardness and grain size values to obtain a Hall-Petch relationship.
3. Experimental Procedures

3.1. Fabrication of the alloy

The Al-Cu-Ni-Mn-Ag alloy was prepared in California by the company Plasma Materials. An electric arc furnace was used to melt the alloy multiple times to guarantee homogeneous distribution of the metals. The alloy was made with pure metals of at least 99.9% purity in equiatomic proportions. Following casting, the alloy was machined into cubes of 5mm by electric discharged machining (EDM).

3.2. Sample Characterization

3.2.1. SEM/EDS

Hitachi SU-3500 SEM, which has EDS capabilities, has been used for microstructural analysis. SEM was used to observe the microstructures in the back-scattered secondary imaging mode, while EDS enabled the identification of elemental compositions in atomic percentages through color mapping.

Elemental composition measurements with EDS were taken as an average of 6 scans through color mapping for overall composition and point spectra for phase composition.

The operation parameters are the following:

- Accelerating Voltage: 15-20 kV
- Probe Current: 65-110 µA
- Working Distance: 6.5-12 mm
3.2.2. XRD

X-Ray Diffraction (XRD) was used to analyze phase composition of the alloy. A Bruker D8 Discover X-Ray diffractometer was used with a step size of 0.02° at a rate of 1 second per step.

3.2.3. Grain Size Measurements

Measurements of grain size were performed according to ASTM E112-13 [41] following the linear intercept method.

3.2.4. Metallography

For metallography, samples were prepared using 1000 grit paper according to ASTM E3-11 [42].

3.3. As-received

After metallography was performed on the as-received sample, Hitachi SU-3500 SEM was used to observe the microstructures in the back-scattered secondary imaging mode. EDS was used to create elemental maps to identify the chemical composition of the alloy and its microconstituents. XRD was performed to analyze the phases present in the alloy.

3.4. Oxidation Experiments

The static oxidation experiments were conducted for 24 hours in a SentroTech ST Muffle Box furnace at 600, 700, 800, 900, and 1000 °C. The samples were prepared by polishing all six sides of the cube up to 1000-grit paper followed by ultrasonic cleaning in isopropanol. In the as-received condition, the average weight was calculated by measuring each sample three times. The samples were individually placed on a covered alumina crucible and then positioned below the
thermocouple, right in the center of the inside of the furnace. During heating and cooling, the furnace had a rate of 10 °C/min. The samples were left to cool inside the furnace, and the average weight after oxidation was measured. The mass gain per unit area (g/cm²) as a function of time and temperature was calculated to determine the oxidation resistance of the alloy.

3.5. Mechanical Testing

Hardness measurements were taken for the alloy using Rockwell Hardness tester on B scale (HRB) and Vickers hardness scale (HV). Microhardness testing was performed on MicroMet III at a 50-gram load 15 sec dual time. A 10-gram load was used for Cu rich phases at 600-800 °C due to their smaller grain size. For macrohardness testing, Instron Wilson Rockwell 2000 series was employed.
4. Results and Discussion

4.1. As Received Al-Cu-Ni-Mn-Ag HEA

In the as-received condition, the Al-Cu-Ni-Mn-Ag high entropy alloy’s chemical composition was determined using EDS. The chemical composition of the alloy, in atomic percent, is shown in Table 3. The table shows a comparison between measured values versus the values given to the manufacturer.

Table 3. Chemical Composition of Al-Cu-Ni-Mn-Ag in the as-received condition

<table>
<thead>
<tr>
<th>Element</th>
<th>EDS</th>
<th>Manufacturer</th>
</tr>
</thead>
<tbody>
<tr>
<td>Aluminum</td>
<td>21.9</td>
<td>20</td>
</tr>
<tr>
<td>Manganese</td>
<td>17.3</td>
<td>20</td>
</tr>
<tr>
<td>Copper</td>
<td>25.1</td>
<td>20</td>
</tr>
<tr>
<td>Nickel</td>
<td>22.5</td>
<td>20</td>
</tr>
<tr>
<td>Silver</td>
<td>13.3</td>
<td>20</td>
</tr>
</tbody>
</table>

The microstructure of the Al-Cu-Ni-Mn-Ag HEA in the as received condition was analyzed. Figure 6 shows the microstructure of the alloy with the color mapping of each element. It is evident that there are areas in the microstructure where some elements segregate. Based on the color mapping, Al, Ag, Cu and Ni have the more well-defined areas, while Mn is more distributed throughout the microstructure.
Figure 6. Color mapping of the Al-Cu-Ni-Mn-Ag HEA in the as-received condition

Based on the color mapping micrographs obtained from EDS, a total of three microconstituents were identified in the Al-Cu-Ni-Mn-Ag alloy. Figure 7 shows the three microconstituents at a high magnification, which are rich in different elements: (a) black area rich in Ni, (b) gray area rich in Cu, and (c) white area rich in Ag. The matrix of the alloy is Ag-rich, which is represented by the white area. A more or less black rounded Ni-rich is surrounded by a gray Cu rich microconstituent. However, a dendritic structure has been identified at lower magnifications as shown in Figure 8.
Figure 7. Microstructure of Al-Cu-Ni-Mn-Ag HEA in the as received condition

Figure 8. Microstructure of Al-Cu-Ni-Mn-Ag HEA in the as received condition at a low magnification
XRD was also used to characterize the microconstituents of the alloy in the as received condition. Figure 9 shows the XRD pattern in the as received condition. The peaks show the existence of Aluminum, Aluminum Manganese, and Aluminum Manganese Nickel rich phases. Table 4 shows the chemical composition of the phases found in the alloy by EDS on SEM.

Figure 9. XRD pattern of the Al-Cu-Ni-Mn-Ag HEA
Table 4. Summary of phase compositions by EDS of the Al-Cu-Ni-Mn-Ag HEA

<table>
<thead>
<tr>
<th>Phases</th>
<th>Elements</th>
<th>At %</th>
<th>Atomic Radii Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Nickel-Rich</strong></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Al</td>
<td></td>
<td>36.85</td>
<td>R(<em>{\text{Ni}}/R</em>{\text{Al}} = 0.868)</td>
</tr>
<tr>
<td>Cu</td>
<td></td>
<td>7.95</td>
<td></td>
</tr>
<tr>
<td>Ni</td>
<td></td>
<td>43.375</td>
<td></td>
</tr>
<tr>
<td>Mn</td>
<td></td>
<td>11.475</td>
<td></td>
</tr>
<tr>
<td>Ag</td>
<td></td>
<td>0.3625</td>
<td></td>
</tr>
<tr>
<td><strong>Copper-Rich</strong></td>
<td></td>
<td></td>
<td>R(<em>{\text{Cu}}/R</em>{\text{Mn}} = 1.141)</td>
</tr>
<tr>
<td>Al</td>
<td></td>
<td>12.7625</td>
<td></td>
</tr>
<tr>
<td>Cu</td>
<td></td>
<td>55.2875</td>
<td></td>
</tr>
<tr>
<td>Ni</td>
<td></td>
<td>4.875</td>
<td></td>
</tr>
<tr>
<td>Mn</td>
<td></td>
<td>24.9</td>
<td></td>
</tr>
<tr>
<td>Ag</td>
<td></td>
<td>2.1625</td>
<td></td>
</tr>
<tr>
<td><strong>Silver-Rich</strong></td>
<td></td>
<td></td>
<td>R(<em>{\text{Ag}}/R</em>{\text{Mn}} = 1.290)</td>
</tr>
<tr>
<td>Al</td>
<td></td>
<td>1.5375</td>
<td></td>
</tr>
<tr>
<td>Cu</td>
<td></td>
<td>13.95</td>
<td></td>
</tr>
<tr>
<td>Ni</td>
<td></td>
<td>1.025</td>
<td></td>
</tr>
<tr>
<td>Mn</td>
<td></td>
<td>21.25</td>
<td></td>
</tr>
<tr>
<td>Ag</td>
<td></td>
<td>62.25</td>
<td></td>
</tr>
</tbody>
</table>

It is unclear why the microconstituents identified by XRD on Figure 9 do not match with what is observed by EDS on SEM as shown in Figure 9. For the microconstituents identified, a comparison was made between the atomic radii ratio of the dominant metal and the second most dominant metal of each phase. The atomic radii ratio of all three phases are shown in Table 4. The atomic radii ratio of the Nickel-Rich phase (43.374% Ni), Copper-Rich phase (55.2875% Cu), and Silver-Rich phase (62.25% Ag) are 0.868, 1.141, and 1.290 respectively. We can conclude that the percentage of the dominant metal increases as the atomic radii ratio increases.
4.2. Oxidation of Al-Cu-Ni-Mn-Ag HEA

Oxidation testing was performed in air for 24 hours at 600, 700, 800, 900, and 1000 °C. Scaling was only seen at 1000 °C as shown in Figure 10. The mass gain per unit surface area in g/cm² was plotted in a graph with respect to temperature, which is shown in Figure 11. The Al-Cu-Ni-Mn-Ag HEA shows excellent oxidation resistance, especially at the highest temperature of 1000 °C.

![Figure 10. Samples after oxidation at A) 600, B) 700, C) 800, D) 900, and E) 1000°C](image)

In a recent study, Hitter et al. [23] examined the impact of Si on Al-Cu-Ni-Mn alloy. The results of this work indicate that the addition of Ag to the same quaternary alloy enhanced oxidation resistance, surpassing the effects of Si as previously reported by Hitter et al. [23].
The oxidized surface and interphase summary is shown in Figure 12. The identified intermetallics were Aluminum, AlMnNi₂, and Al₀.₅₅Mn₀.₅₅ according to XRD results from Figure 9. The main oxide forming from 600 to 900 °C is manganese oxide. However, at 1000 °C, aluminum oxide becomes the main oxide forming. In comparison to the other elements, Al and Mn tend to oxidize more as they segregate to the surface in contact with the oxygen layer. From 600 to 700 °C, the oxide-metal interphase section clearly shows manganese oxide formation. Aluminum and manganese oxide forms at the oxidized surface and oxide-metal interphase at 800 °C. From 900 to 1000 °C, aluminum oxide becomes the main oxide forming at the interphase. The intermetallic compounds that were identified mainly contained Cu and Ni. It seems that the alloy's resistance to oxidation could be attributed to these two metals as the main factors.

Figure 13 shows the XRD of the oxidized surfaces while Table 5 shows the summary of oxides identified. According to XRD results, Mn₂O₃, which is the higher oxide, forms at 600 and 700 °C, Ag₂O forms at 800 °C, while Mn₃O₄ and CuMn₂O₄ form at 900 and 1000 °C. These are
surprising results as (a) Al and Ag could not be identified, (b) the higher oxide (Mn$_2$O$_3$) forms at lower temperatures, and (c) manganese oxide was not detected at 800°C. It's possible that some of the oxides on the surface couldn't be identified by XRD due to a thicker layer of oxides, or due to the density of the oxides.

Table 5. Summary of oxides formed in Al-Cu-Ni-Mn-Ag as identified in XRD graphs

<table>
<thead>
<tr>
<th>Temperature</th>
<th>600 °C</th>
<th>700 °C</th>
<th>800 °C</th>
<th>900 °C</th>
<th>1000 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mn$_2$O$_3$</td>
<td>Mn$_2$O$_3$</td>
<td>Ag$_2$O</td>
<td>Mn$_3$O$_4$</td>
<td>Mn$_3$O$_4$</td>
<td></td>
</tr>
<tr>
<td></td>
<td>CuMn$_2$O$_4$</td>
<td></td>
<td>CuMn$_2$O$_4$</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Figure 12. Elemental mapping of the oxidized surface and interphase of the Al-Cu-Ni-Mn-Ag HEA after oxidation at 600, 700, 800, 900, and 1000 °C for 24 hours.
Figure 13. XRD graphs of the oxidized Al-Cu-Ni-Mn-Ag HEA at A) 600 °C, B) 700 °C, C) 800 °C, D) 900 °C, and E) 1000 °C.
Table 6 shows a summary of the chemical composition of the HEA after oxidation. The results show that Mn decreases as the oxidizing temperature increases. The highest values of Al and Ag and the lowest values of Ni and Mn were found at 1000°C. The lowest percentage of oxygen was also found at 1000°C, which is the same temperature in which the alloy is shown to have higher oxidation resistance based on the oxidation curve in Figure 11.

<table>
<thead>
<tr>
<th></th>
<th>600 °C</th>
<th>700 °C</th>
<th>800 °C</th>
<th>900 °C</th>
<th>1000 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>O</td>
<td>51.82</td>
<td>54.38</td>
<td>51.10</td>
<td>54.88</td>
<td>50.42</td>
</tr>
<tr>
<td>Al</td>
<td>0.93</td>
<td>4.67</td>
<td>10.13</td>
<td>4.68</td>
<td>11.33</td>
</tr>
<tr>
<td>Cu</td>
<td>5.65</td>
<td>1.97</td>
<td>1.43</td>
<td>3.57</td>
<td>3.22</td>
</tr>
<tr>
<td>Ni</td>
<td>0.417</td>
<td>1.50</td>
<td>1.13</td>
<td>0.52</td>
<td>0.25</td>
</tr>
<tr>
<td>Mn</td>
<td>39.20</td>
<td>37.17</td>
<td>35.77</td>
<td>32.63</td>
<td>26.15</td>
</tr>
<tr>
<td>Ag</td>
<td>1.967</td>
<td>0.32</td>
<td>0.47</td>
<td>3.73</td>
<td>8.66</td>
</tr>
</tbody>
</table>

4.3. Hardness Measurements

The work of Hitter et al [33] shows the hardness results of the Al-Cu-Ni-Mn quaternary alloy. In the as received condition, the quaternary alloy has a hardness of 74 HRB, while its microconstituents have (i) 264 HV in the Ni-rich phase, (ii) 133 HV in the Mn-rich phase, and (iii) 270 HV in the Al-Cu-Ni rich phase.

Hardness measurements of the Al-Cu-Ni-Mn-Ag alloy and its phases are shown in Table 7 and Figure 14. The HEA has a hardness value of 97 HRB in the as-received condition. After oxidation
for 24 hours, the hardness value of the alloy decreased up to 80 HRB from 600 to 900 °C. At 1000 °C, the hardness value slightly increased from 80 to 90.25 HRB. Although there was still an increase of hardness at the highest oxidation temperature, the highest hardness is in the as-received condition.

The microhardness in the as-received condition shows is similar value for the Ni-rich and Cu-rich microconstituents, while Ag-rich has the highest value at 429 HV. After undergoing oxidation testing, all three microconstituents decrease their hardness to 600 °C. Hardness values remain relatively constant up to 700 °C. The Ni rich microconstituent increases its hardness from 800 to 1000 °C. Cu and Ag-rich remain constant from 600 to 800 °C. At 900 °C, the hardness of the Cu rich increases while Ag rich decreases. At 1000 °C, the Ag-rich phase has its lowest hardness while Ni-Rich has the highest hardness of 447 HV.

The addition of Ag to Al-Cu-Ni-Mn increases the hardness of the alloy by more than 20% in the received condition. The phases present in both alloys are quite distinct, however they both share a Ni-rich phase. The Ni-rich phase present in the Ag alloy has a higher hardness than the one found in the quaternary alloy in the as-received condition.

Table 7. Macrohardness of Al-Cu-Ni-Mn-Ag and microhardness of the phases in Al-Cu-Ni-Mn-Ag

<table>
<thead>
<tr>
<th>Temp(°C)</th>
<th>Macrohardness (HB)</th>
<th>Microhardness (HV)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Ni-rich</td>
</tr>
<tr>
<td>As received</td>
<td>97</td>
<td>393</td>
</tr>
<tr>
<td>600</td>
<td>93.25</td>
<td>274</td>
</tr>
<tr>
<td>700</td>
<td>89.75</td>
<td>268</td>
</tr>
<tr>
<td>800</td>
<td>87.5</td>
<td>329</td>
</tr>
<tr>
<td>900</td>
<td>80</td>
<td>412</td>
</tr>
<tr>
<td>1000</td>
<td>90.25</td>
<td>447</td>
</tr>
</tbody>
</table>
Figure 14. Graphical representation of the hardness values of the phases found in Al-Cu-Ni-Mn-Ag HEA

4.4. Grain Size

The microstructure after oxidation was analyzed at low and high magnifications as shown in Figure 15. At 600 °C, the microstructure is similar to the as-received condition. At 700 °C, there’s still a dendritic structure at low magnifications, but at high magnification it seems that the grains are not that well defined. At 800 °C, there is a substantial increase in grain size that is observable in both low and high magnifications. Grains continue to grow and spread at 900 and 1000 °C. An interesting feature appears from 800 to 1000 °C in the Cu-Rich and Ag-Rich phase. A Widmanstätten type pattern appears to a certain degree in the Ag-Rich phase and slightly in the Cu-Rich phase.
Figure 15. Microstructures of Al-Cu-Ni-Mn-Ag HEA after static oxidation for 24 hours at 600, 700, 800, 900, and 1000 °C at high (left) and low (right) magnifications.
Table 8 shows the grain size of the high entropy alloy taken using the intercept method. It can be observed that the grain size of the Al-Cu-Ni-Mn-Ag alloy increases as the temperature of oxidation increases.

Table 8. Grain size of the Al-Cu-Ni-Mn-Ag alloy in µm

<table>
<thead>
<tr>
<th>Sample</th>
<th>Grain size (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-received</td>
<td>8.07</td>
</tr>
<tr>
<td>600 °C</td>
<td>10.53</td>
</tr>
<tr>
<td>700 °C</td>
<td>9.01</td>
</tr>
<tr>
<td>800 °C</td>
<td>22.93</td>
</tr>
<tr>
<td>900 °C</td>
<td>35.23</td>
</tr>
<tr>
<td>1000 °C</td>
<td>37.87</td>
</tr>
</tbody>
</table>

In order to analyze the amount present of each phase in the alloy, a phase percentage was calculated. Table 9 shows the summary of the phase percentages found on the Al-Cu-Ni-Mn-Ag HEA. Taking into consideration the as-received and oxidation samples, the average composition of the phases is:

a. Ni-Rich Phase (35.7%)
b. Cu-Rich Phase (34.5%)
c. Ag-Rich Phase (29.7%)

In the as-received condition, the microstructure is mainly composed of Ni-rich. After oxidation, Cu-rich is the most dominant at 600 and 700 °C. At 800 °C, there’s an increase of Ni-rich percentage. Ag-Rich is the main microconstituent at 900 °C. Lastly, at 1000 °C, all three microconstituents are more proportional to each other.
Table 9. Summary of the microconstituents percentages in each Al-Cu-Ni-Mn-Ag HEA

<table>
<thead>
<tr>
<th>Sample</th>
<th>Ag-Rich</th>
<th>Cu-Rich</th>
<th>Ni-Rich</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-received</td>
<td>27%</td>
<td>29%</td>
<td>45%</td>
</tr>
<tr>
<td>600 °C</td>
<td>25%</td>
<td>50%</td>
<td>26%</td>
</tr>
<tr>
<td>700 °C</td>
<td>22%</td>
<td>64%</td>
<td>15%</td>
</tr>
<tr>
<td>800 °C</td>
<td>32%</td>
<td>8%</td>
<td>60%</td>
</tr>
<tr>
<td>900 °C</td>
<td>43%</td>
<td>24%</td>
<td>33%</td>
</tr>
<tr>
<td>1000 °C</td>
<td>30%</td>
<td>33%</td>
<td>37%</td>
</tr>
<tr>
<td>Total</td>
<td>29.7%</td>
<td>34.5%</td>
<td>35.7%</td>
</tr>
</tbody>
</table>

Figure 16. Graphical representation of the amount of microconstituents present in the Al-Cu-Ni-Mn-Ag HEA
A graphical representation of the data from Table 9 is shown in Figure 16. A summary of the trend of microconstituents percentage at different oxidation temperatures is shown in Table 10. The following tendency has been observed:

- Cu-Rich, represented as an orange circle, increases its percentage with an increase in oxidation temperature up to 700 °C. At 800 °C, the amount decreases considerably, followed by an increase up to 100 °C.
- Ni-Rich, represented as a gray square, decreases its percentage up to 700 °C. It increases significantly at 800 °C, followed by a decrease at 900 °C where it remains relatively constant up to 1000 °C.
- Ag-Rich, represented as the blue square, stays relatively constant up to 900 °C. It slowly decreases its percentage up to 700 °C, and then increases up to 900 °C. At 1000 °C, it slightly decreases.

Table 10. Trend of dominant microconstituents with changes in temperature

<table>
<thead>
<tr>
<th>Temperature Range, °C</th>
<th>Cu-Rich</th>
<th>Ni-Rich</th>
<th>Temperature Range, °C</th>
<th>Ag-Rich</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-received–700</td>
<td>↑</td>
<td>↓</td>
<td>As-received–700</td>
<td>≈</td>
</tr>
<tr>
<td>700 – 800</td>
<td>↓</td>
<td>↑</td>
<td>700 - 900</td>
<td>↑</td>
</tr>
<tr>
<td>800 – 1000</td>
<td>↑</td>
<td>↓</td>
<td>900 – 1000</td>
<td>↓</td>
</tr>
</tbody>
</table>

From the trend outlined above, there is a reversed relationship between the Cu and Ni rich phases with respect to oxidizing temperature. As the temperature increases up to 700°C, the amount of Cu-rich phase increases while the Ni-rich phase decreases. At 800°C, the amount of Cu-rich phase decreases significantly while the Ni-rich phase increases significantly. From 800 to 1000°C, the amount of Cu-rich phase begins to increase, while the Ni-rich phase amount decreases until 900°C and then stabilizes approximately at 1000°C.
4.5. Hall-Petch Relationship

The grain size measurements and the hardness values of the Al-Cu-Ni-Mn-Ag HEA were plotted in Figure 17 to obtain a Hall-Petch relationship. The hardness values in Rockwell hardness B-Scale and the inverse reciprocal of the square root of the grain size was plotted to obtain the linear equation of $\sigma_y = 49.039x + 77.095 \, d^{-1/2}$. A value of 77.095 was measured for the strengthening coefficient ($k$), while $\sigma_y$ has a value of 49.039. It is important to note that the grain size used to calculate the Hall-Petch relationship is based on the composition of the three micro constituents found in the alloy’s microstructure.

A trend that relates the hardness and grain size to the dominant microconstituents at different oxidation temperatures is shown in Figure 17. At higher temperatures, the alloys exposed to oxidation at 800 and 1000°C are dominated by the Ni-rich microconstituent. At 900°C the alloy’s main microconstituent is Ag-rich. At lower temperatures of 600 and 700°C, the alloy is dominated by the Cu-rich microconstituent.

Figure 17. Hall-Petch graph of Al-Cu-Ni-Mn-Ag HEA in respect to grain size and hardness
It has been attempted to establish a Hall-Petch relationship in high-entropy alloys (HEAs) in various studies conducted in literature. A Hf-Nb-Ta-Ti-Zr alloy that has been cold-rolled and annealed for 24 hours at temperatures ranging from 1000 to 1200°C follows the Hall-Petch equation \( \sigma_y = 942 + 270 \, d^{-1/2} \), where \( \sigma_y \) is the yield strength and \( d \) is the grain diameter [39]. The frictional stress and strengthening coefficient are 942 MPa and 270 MPa/mm\(^{-1/2} \), respectively. Yoshida et al [40] studied the effect of frictional stress on the Hall-Petch relationship in subsystems of Co-Cr-Fe-Mn-Ni HEA to analyze the strengthening mechanism of FCC high-entropy alloys. Among the analyzed alloys, Co\(_{20}\)(CrNi)\(_{80}\) had the highest frictional stress with \( \sigma_y = 280 + 168 \, d^{-1/2} \). On the other hand, (CoNi)\(_{95}\)Cr\(_5\) and Co-Fe-Ni alloys had the lowest friction stress values with \( \sigma_y = 30.2 + 240 \, d^{-1/2} \) and \( \sigma_y = 62.6 + 366 \, d^{-1/2} \), respectively. The frictional stress and strengthening coefficient terms of the Hall-Petch equation in the Al-Cu-Ni-Mn-Ag HEA is considerably lower than what has been reported in the literature. It is important to note that the data for the alloys analyzed in the references cited in this paper do not have FCC crystal structures. Therefore, a meaningful comparison cannot be made based on the present set of data.
5. Conclusions

A high entropy alloy containing elements with an FCC and cubic structure including Al, Cu, Ni, Mn, and Ag was investigated and characterized to understand its behavior and how the addition of Ag in the Al-Cu-Ni-Mn quaternary alloy affected its properties.

The Al-Cu-Ni-Mn-Ag alloy shows three microconstituents in the as received condition, rich in: (a) Ni, (b) Cu, and (c) Ag. For each phase, a ratio of the most dominant and second most dominant element was analyzed. It was found that as the dominant metal increases, the atomic radii ratio increases as well.

Oxidation in air was done for 24 hours at a temperature range from 600 to 1000°C. The alloy showed excellent oxidation resistance as the mass gain per surface area was minimal. The oxidized surface and oxide-metal interphase were analyzed through elemental mapping. The main oxides forming are manganese oxide from 600 to 900 °C and aluminum oxide at 1000 °C. Cu and Ni were the main intermetallics, while Al and Mn showed tendency to oxidize more. The oxides identified by XRD were Mn$_2$O$_3$ at 600 and 700°C, Ag$_2$O at 800°C, and Mn$_3$O$_4$ and CuMn$_2$O$_4$ at 900 and 1000°C. Cu and Ni might give the alloys its oxidation resistance properties as they are not present in the oxide form.

Elemental composition of the alloy’s surface after oxidation showed that Mn decreases as the oxidizing temperature increases. The highest values of Al and Ag and the lowest values of Ni, Mn, and oxygen were found at 1000°C, which is the same temperature in which the alloy is shown to have higher oxidation resistance based on the oxidation curve.

Hardness was measured on the alloy and its microconstituents. In the as received condition, the addition of Ag to the quaternary alloy increases hardness by more than 20%. After oxidation,
hardness decreases up to 900°C, and at 1000 °C it slightly increases. In the as-received condition, the Ag-rich microconstituent demonstrated the highest level of hardness at 429 HV. However, following oxidation at 1000°C, the Ni-rich microconstituent surpassed it with a hardness of 447 HV, making it the hardest microconstituent.

Micrographs show that the grain size increases as the oxidation temperature increases. A Widmanstätten pattern is slightly present in the Ag-rich and Cu-rich phase from 800 to 1000°C. Grain size and phase percentage were measured and analyzed. A reversed relationship between the phase percentage and the oxidizing temperature of the Ni and Cu rich phase was identified. The hardness versus inverse square root of the grain diameter was plotted and a hardening rate of 77.095 HR/μm⁻¹/² was found by the least square fit method. The Hall-Petch plot related the hardness and grain size to the dominant microconstituent at different temperatures. At higher temperatures, the hardness is mainly controlled by Ni-rich, while Cu-rich controls the lower temperatures.
References


Curriculum Vita

Gina Zavala Alvarado completed her Bachelor of Science in Metallurgical and Materials Engineering from The University of Texas at El Paso in 2021. In the spring of 2022, she enrolled in the master’s program in Metallurgical and Materials Engineering.

Throughout her academic career, Gina worked as a teaching assistant in the Mathematical and Engineering Department. Additionally, she completed three internships with NASA Jet Propulsion Laboratory and Ellwood City Group. Currently, she is working as a research assistant at the university, focusing on her research. During her time as a student, she held the position of an officer of Alpha Sigma Mu and was a member of the MMBME student advisory board. Furthermore, she is trilingual in English, Spanish, and Japanese.

After graduating, Gina plans to move to Seattle, Washington, to begin her career in the aerospace industry. She is passionate about leading and learning from others, which she hopes to incorporate into her future endeavors.

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