

To the Graduate Council:

I am submitting herewith a dissertation written by Kevin Tacy Zysk entitled "The Effects of Electron Beam Rotation Upon Electron Beam Welded Copper - 304 Stainless Steel Couples." I have examined the final copy of this dissertation for form and content and recommend that it be accepted in partial fulfillment of the requirements for the degree of Doctor of Philosophy, with a major in Engineering Science and Mechanics.

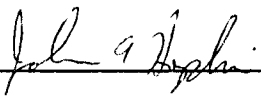

Mary Helen McCay, Major Professor

We have read this dissertation
and recommend its acceptance:

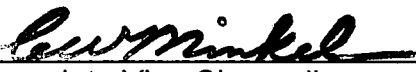








Accepted for the Council:


Associate Vice Chancellor
And Dean of the Graduate School

THE EFFECTS OF ELECTRON BEAM ROTATION UPON
ELECTRON BEAM WELDED COPPER – 304 COUPLES

A Dissertation

Presented for the

Doctor of Philosophy

Degree

The University of Tennessee, Knoxville

Kevin Tacy Zysk

December 1998

ACKNOWLEDGEMENTS

I am taking this opportunity to thank all those people who gave so much of their time to help me prepare this dissertation. Because so many people provided help to me over the past few years a blanket thank you would be woefully inadequate. For me, it is imperative that I mention all those who have contributed. If I have inadvertently left anyone out please accept my apologies.

The first order of business is to recognize and thank those committee members, who gave so much of their valuable time so that I might grow academically. These people are Dr. Mary Helen McCay, Dr. Dwayne McCay, Dr. Naren Dahotre, Dr. Ted Paluden, Dr. John Hopkins, and Dr. Remi Engles. I will always be in your debt. A note of special thanks is due Dr. Mary Helen McCay and Dr. Naren Dahotre for their patience and guidance during this ordeal.

There have been numerous people within AEDC's MetLab, Weld Shop, and Machine Shop, involved behind the scene, making completion of this dissertation possible. Their assistance in welding and machining of test coupons, performing mechanical tests, preparing samples, and taking micrographs has been greatly appreciated. They include Randall Johnson, Steve Ary, Joe McCabe, Jerry Payne, and Doyle McConnel. Thank you all.

I am very much indebted to both Arvind Agarwal and Melissa Meyer of the UTSI for their help in making the SEM micrographs and microhardness maps that I have used.

This project could not have been completed without the help and cooperation of the management team, past and present, at AEDC DOT, Robert Crook, Dave Bond, Jere Matty, and Tom Best, Carlos Tirres, and Keith Kushman. Robert Crook's help in solving my "Microsoft" formatting nightmares is greatly appreciated.

I also need to thank Fuffy Beardon, Kim Reed, and Kent Turner of ACS's graphics/photo department for producing the artwork and digitized images used in this dissertation.

The final note of thanks goes to my wife Kyung Sun. Without her support, completion of this task would have been impossible. When the days stretched into nights that continued into the next day she was always there to make sure I had warm meals, took my vitamins, and the lawn was mowed. She kept the "Big Blue Frog" going. Thanks Kyung Sun.

ABSTRACT

The United States Air Force, Arnold Engineering Development Center, has been using copper to 304 stainless steel couples made using the electron beam welding process during the fabrication of intrusive gas-path diagnostic probes for over five years. Only a limited physical analysis of the resulting welds had been done. The purpose of this investigation was to determine the effects that varying the rotation frequency of the electron beam had upon the mechanical characteristics of the copper to 304 stainless steel couples.

During welding of the samples, all controllable weld process parameters were held constant with the exception of the electron beam rotation frequency; the rotation frequency was varied from 20 *Hz* to 180 *Hz* in steps of 20 *Hz*. Samples welded without electron beam rotation provided a baseline for comparison.

Microhardness distributions showed that weld region homogeneity as evidenced by microhardness maps and optical microscopy was a function of the electron beam rotation frequency. As the electron beam rotation frequency was increased so did the uniformity of the microhardness distribution.

There was no correlation between electron beam rotation frequency and weld tensile strength since each test coupon failed in the Cu base

material outside of the weld region. The weld region material was stronger than the Cu base material regardless of the electron beam rotation frequency.

The top weld bead width and bottom weld bead width decreased only slightly with increases in the electron beam rotation frequency.

The welds made at all electron beam rotation frequencies used for this study contained cracks within the weld region, heat affected zone (HAZ), or both. The relative number, length, and location of the individual cracks changed with electron beam rotation frequency. Cracking in the HAZ due to liquid metal embrittlement (LME) was not evident in those samples welded with the electron beam rotation frequencies below 100 *Hz*. Cracking due to LME outside of the weld region in the HAZ was observed to increase with the electron beam rotation frequency above 80 *Hz*.

The relationship between weld region cracking and residual stress within the weld region was shown to be dependent on the electron beam rotation frequency. Cracking in the weld region was not observed in samples welded with the electron beam rotation frequency above 80 *Hz*. Within the weld region the maximum crack length was 0.5 *mm* with most crack lengths approximately 0.2 *mm* or less. Cracking in the baseline sample, welded without electron beam rotation, was limited to a 0.01 *mm* crack in the 304 SS HAZ.

Divorced spherical structures similar to those found in documented monotectic systems were observed. These structures were found to include regions of planar front growth that transitioned into perturbed cellular growth forms.

TABLE OF CONTENTS

CHAPTER	PAGE
I. INTRODUCTION.....	1
II. BACKGROUND	5
A. Electron Beam Welding Basics	5
B. The Cu-Fe System	12
C. The Fe-Ni-Cr System	14
D. Solidification Forms In Alloys	15
D.1. Planar front solidification	16
D.2. Cellular solidification.....	19
D.3. Dendritic solidification.....	21
E. Monotectic Solidification.....	22
F. Thermal Residual Stress in Weldments	24
F.1. Residual stress in the weld region.....	25
G. Liquid Metal Embrittlement.....	29
III. EXPERIMENTAL PROCEDURES	32
A. Sample Preparation.....	32
B. Electron Beam Welding.....	32
C. Test Specimen Preparation.....	38
D. Optical Metallurgical Examination	39
E. Microhardness Testing	40
F. Tensile Testing	41

G.	Weld Pool Width Measurement	42
IV.	RESULTS AND DISCUSSION.....	43
A.	General Weld Region Morphology	43
B.	Microhardness in the Weld Region.....	45
C.	Tensile Tests Results	53
D.	Weld Region Width.....	54
E.	Weld Cracking	60
E.1.	LME and HAZ cracking	61
E.2.	Weld region cracking	69
F.	Solidification Structures In the Weld Region	72
V.	CONCLUSIONS.....	80
	LIST OF REFERENCES.....	83
	APPENDIXES.....	89
	APPENDIX A (Tables)	90
	APPENDIX B (Figures)	95
	APPENDIX C (Derivation of the CST).....	153
	VITA.....	157

LIST OF TABLES

TABLE		PAGE
III-1	Chemical Composition of Cu and 304 SS.....	90
III-2	Key Machine Parameters	91
IV-1	Theoretical Molten Fluid Velocity Data	92
IV-2	Thermophysical Properties of Cu and 304 SS.....	93

LIST OF FIGURES

FIGURE	PAGE
II-1 A typical layout of an EB welding machine [B-1].....	96
II-2 Schematic of a keyhole weld	97
II-3 Cu-Fe phase diagram [B-2].....	98
II-4 Cu-Pb phase diagram [B-3]	99
II-5 Isothermal section at 900 °C of the Fe-Ni-Cr ternary phase diagram [B-4]	100
II-6 Artistic interpretations of the basic solidification structures: (a) planar growth, (b) cellular growth, (c) columnar dendrite, and (d) equiaxed dendrite [B-5]	101
II-7 A graphical representation of the key parameters in the constitutional supercooling process: (a) the phase diagram, (b) the solute enriched layer ahead of the solid-liquid interface, (c) solidification with a stable interface, and (d) solidification with an unstable interface [B-6]	105
II-8 Hypothetical monotectic phase diagram	109
II-9 Micrograph of a cast Cu-36 Pb alloy. Dark areas are Pb rich [B-7]	110
II-10 Micrograph of a cast Cu-Pb alloy containing 6.83 wt% Pb. Dark areas are Pb rich [B-8]	111
II-11 Reference planes used to describe the weld region	112
II-12 Changes in stress during welding: (a) start of welding, (b) time a, (c) time b, and (d) time c [B-9].....	113
II-13 Top view of a weld showing an area of fluid expulsion where the electron beam exited the weld pool prior to solidification	117
III-1 Layout of weld coupons	118
III-2 Overhead view of dog bones with bites	119

III-3	Schematic of microhardness test pattern.....	120
IV-1	Weld region area in the YZ plane versus electron beam rotation frequency. The area is based upon a trapezoid formed using the top weld bead width, the bottom weld bead width, and the thickness of the sample	121
IV-2	Composite figure, made using three limited field-of-view images, of the cross-section (YZ plane) of sample 4-3S welded at 40 Hz	122
IV-3	Optical micrograph of sample 4-3F, welded at 40 Hz, showing Cu rich interdendritic material. The light regions are Fe rich dendrites; the dark areas are the Cu rich material	123
IV-4	Optical micrograph of a portion of the top weld nugget region along the Cu wall of sample 4-3S, weld at 40 Hz, showing Fe rich particles (light colored particles) surrounded by a Cu rich matrix (dark material)	124
IV-5	Optical micrograph of sample 5-10F showing the distribution of small structures contained within the weld region, beam rotation set at 180 Hz.....	125
IV-6	Map of Knoop microhardness in the YZ plane for sample 5-1F, welded without electron beam rotation	126
IV-7	Map of Knoop microhardness in the YZ plane for sample 4-2S, beam rotation frequency set at 20 Hz.....	127
IV-8	Map of Knoop microhardness in the YZ plane for sample 4-3S, beam rotation frequency set at 40 Hz.....	128
IV-9	Map of Knoop microhardness in the YZ plane for sample 2-4S, beam rotation frequency set at 60 Hz (Includes only approximately one-half of the weld region on the Cu side.)	129

IV-10	Average Knoop microhardness, standard deviation, minimum, and maximum values of Knoop microhardness as a function of electron beam rotation frequency. Linear curve fits of the data are included to show trends. (Note that the Cu base material, 304 SS base material, and the average Knoop microhardness of the baseline sample welded without electron beam rotation have Knoop microhardness values of HK85, HK400, and HK205 respectively.).....	130
IV-11	Map of Knoop microhardness in the YZ plane for sample 4-5F, beam rotation frequency set at 80 Hz.....	131
IV-12	Map of Knoop microhardness in the YZ plane for sample 5-6S, beam rotation frequency set at 100 Hz.....	132
IV-13	Map of Knoop microhardness in the YZ plane for sample 3-7S, beam rotation frequency set at 120 Hz.....	133
IV-14	Map of Knoop microhardness in the YZ plane for sample 1-8S, beam rotation frequency set at 140 Hz.....	134
IV-15	Map of Knoop microhardness in the YZ plane for sample 5-9S, beam rotation frequency set at 160 Hz.....	135
IV-16	Map of Knoop microhardness in the YZ plane for sample 3-10F, beam rotation frequency set at 180 Hz.....	136
IV-17	Micrograph of sample 3-5F, in the YZ plane, showing ductile failure in the Cu outside of the weld region, beam rotation set at 80 Hz. The black region is mounting material.	137
IV-18	Top and bottom weld bead width in the YZ plane versus electron beam rotation frequency for: (a) all electron beam rotation frequencies and, (b) with the electron beam rotation frequencies of 120 Hz and 180 Hz excluded. The baseline sample widths are 2.44 mm on top and 1.02 mm on the bottom.....	138
IV-19	Schematic diagrams of: (a) beam directions and, (b) velocity vectors	140
IV-20	Micrograph of sample 5-6S showing penetration of Cu rich material into the grain boundary along the 304 SS side of the weld region, beam rotation set at 100 Hz	142

IV-21	Micrograph of sample 4-3F, welded at 40 Hz, showing cracks in the weld region that extend into the HAZ	143
IV-22	Micrograph of sample 4-2S, welded at 20 Hz, showing infiltration of Fe rich material into the Cu HAZ	144
IV-23	Micrograph of sample 5-10F, welded at 180 Hz, showing a post solidification crack in the 304 SS HAZ that extends just into the weld region. A smaller pre-solidification crack containing Cu rich material is also shown	145
IV-24	Micrograph of sample 3-5F showing divorced spherical structures	146
IV-25	Darkfield micrograph of sample 1-4F showing closely packed fibrous composite solidification structures within the weld region, electron beam rotation set at 60 Hz. The Fe rich material appears black while the Cu rich fibrous material appears gold in this micrograph.....	147
IV-26	Micrograph of a cast Cu-35 wt% Pb alloy containing a divorced spherical region. Dark regions are Pb rich material [B-10].	148
IV-27	SEM micrograph of sample 3-5F showing the interior of a divorced spherical structure that includes the peripheral planar growth region. The Cu rich material was removed using a solution of FeCl ₃ and water.	149
IV-28	SEM micrograph of sample 3-5F showing slightly perturbed cellular growth.....	150
IV-29	SEM micrograph of sample 3-5F showing parallel cellular growth at the edge of the planar growth region.....	151
IV-30	SEM micrograph of Sample 3-5F showing dendrite.....	152

NOMENCLATURE

Symbol	Meaning	Units
AEDC	Arnold Engineering Development Center	
Ag	Silver	
AWS	American Welding Society	
BAg-8	Brazing Alloy With Composition Ag-28% Cu	
C	Carbon	
C_L	Concentration of the Liquid	wt%
C_L^*	Concentration of Liquid at the Solid Liquid Interface	wt%
C_0	Initial Alloy Composition	wt%
C_S	Concentration of the Solid	wt%
C_S^*	Concentration of Solid at the Solid Liquid Interface	wt%
CCC	Copper Contamination Cracking	
CO ₂	Carbon Dioxide	
Cr	Chromium	
CST	Constitutional Supercooling Theory	
Cu	Copper	
D_L	Diffusion Coefficient in Liquid	mm ² /s
EB	Electron Beam	
EBW	Electron Beam Welding	
EDM	Electrical Discharge Machining	

Fe	Iron	
G	Temperature Gradient in the Liquid	$^{\circ}\text{C}/\text{mm}$
GB	Grain Boundary	
LBW	Laser Beam Welding	
HAZ	Heat Affected Zone	
Hg	Mercury	
HK	Knoop Hardness	
LME	Liquid Metal Embrittlement	
L ₁	Liquid 1	
L ₂	Liquid 2	
OFHC	Oxygen Free High Conductivity	
Pb	Lead	
R	Solidification Rate	mm/s
S/L	Solid/Liquid	
SS	Stainless Steel	
T _c	Critical Temperature	$^{\circ}\text{C}$
Ti	Titanium	
T _l	Liquidus Temperature	$^{\circ}\text{C}$
T _m	Melting Temperature	$^{\circ}\text{C}$
T _m	Monotectic Reaction Temperature	$^{\circ}\text{C}$
T _s	Solidus Temperature	$^{\circ}\text{C}$
V _f	Volume Fraction	

USAF	United States Air Force	
UTS	Ultimate Tensile Strength	<i>Pa</i>
UTSI	University Of Tennessee Space Institute	
Zn	Zinc	
fcc	Face Centered Cubic	
k	Partition Coefficient	
m_l	Slope of the liquidus line	$^{\circ}\text{C}/\text{wt}\%$
<i>wt%</i>	Weight Percent	
α	Alpha Phase of an Alloy	
β	Beta Phase of an Alloy	
ΔT	Temperature Difference	$^{\circ}\text{C}$
δ	Boundary Layer Thickness	<i>mm</i>
γ	Gamma Phase of an Alloy	
ε	Amplitude of Perturbation	<i>mm</i>
ε'	Strain Rate	$(\text{mm}/\text{mm})/\text{s}$
λ_1	Primary Dendrite Arm Spacing	<i>mm</i>
λ_2	Secondary Dendrite Arm Spacing	<i>mm</i>
θ	Angle	<i>radians</i>
σ'	Load Rate	<i>N/s</i>

CHAPTER I

INTRODUCTION

This research was initiated to determine if any correlations between electron beam (EB) rotation-frequency and the physical characteristics of the resulting welds in dissimilar weld couples, welded using the EB welding process, could be shown. Prior to the start of this work, the state of the art for EB welding copper (Cu) to 304 stainless steel (304 SS) was limited. The question of the miscibility of Cu in Fe and the high energy density required to form this weld, not generally available using conventional welding methods, effectively eliminated any demand for it. Additionally, little information on the effects of electron beam rotation on weld homogeneity and mechanical properties was available. Information that was available addressed the effects of manipulating the beam path as a cosmetic surface treatment and a means to prevent weld root spiking only. The use of the dissimilar weld couple with limited miscibilities of the major constituents provided a convenient means to visualize weld homogeneity.

The use of dissimilar welds at the Arnold Engineering Development Center (AEDC) has become a common practice since the purchase of an

electron beam welder in 1992. In particular, the Cu to 304 SS weld has become a key process in the manufacture of intrusive high-temperature gas-path probes. Without the ability to make this weld the upper limit on probe operational temperature and heat flux would be limited by silver brazes. Because the eutectic braze alloy, silver (Ag)–28 wt% Cu (commonly referred to as BAg-8), used on portions of the probes at AEDC melts at 780°C, the maximum safe operating temperature for the probes is less than 600°C, well below the exhaust gas temperature for modern gas turbine engines.

The combined requirements for structural integrity and high heat flux loading, 23 W/mm^2 , dictated the use of 304 SS probe bodies with the tips made of oxygen-free high-conductivity (OFHC) Cu. Although the 304 SS portion of the probe is not subjected to the high heat flux and temperatures experienced by the probe tip, it still must provide structural support at elevated temperatures. Hence the probes are actively cooled with water to ensure their mechanical integrity under the high heat flux loads.

Early rudimentary experiments, performed by personnel from the United States Air Force (USAF) and Sverdrup Technology, at PTR-Precision Technologies, Inc., led to the decision to attempt to fabricate the probes using the Cu-304 SS EB weld. Budget constraints limited testing to hydrostatic pressure testing only; maximum test pressures were 102 atm. As long as a

probe did not leak or burst it was considered to be usable. No attempt, however, was made to metallurgically examine the dissimilar welds or optimize the process.

The success of the intrusive probes led to considering employing the electron beam welding (EBW) process on arc heater components. The proposed high risk use mandated a metallurgical examination of the welds to ensure suitability. The original limited metallurgical examination was funded by the USAF AEDC's Directorate of Technology.

Since weld integrity under pressure was a primary concern of the arc heater group at AEDC, a series of tensile tests were performed to determine if any relationship between electron beam rotation frequency and tensile strength existed [1-1]¹. No correlation between electron beam rotation frequency and tensile strength could be determined from the results; all samples failed in the copper region outside of the weld zone. The fact that the EB welds were stronger than the base metal (Cu) was adequate justification to use the welding process in actual hardware.

No useful information on the application of the EB welding process to a Cu and 304 SS weld was available at the time of the initial experiments. The author was aware of experiments being performed at the same time at The

¹ Numbers in brackets refer to similarly numbered references located in the List of References.

University of Tennessee Space Institute (UTSI), using a CO₂ laser as a heat source to perform a similar weld [1-2]. Though the laser beam welding process, like EBW, is a high energy density beam process, there are several differences that can make direct comparison difficult.

The results of this research have shown that the use of electron beam rotation during EB welding does have a measurable influence on the resulting weld morphology and physical attributes of the weld region. The information derived from this effort can be used to assist in the optimization of the electron beam welding process for the Cu-304 SS weld couple.

CHAPTER II

BACKGROUND

A. Electron Beam Welding Basics

Fusion welding has been defined as a joining process that requires the coalescence or melting of a base material and any filler materials used [II-1]. The more traditional means used to melt the materials usually include the various arc welding processes and gas welding processes. Friction welding and reactive joining techniques are also available. These techniques will not be discussed in this paper. In all these examples, the heat input to the fusion zone is substantial when compared with the laser beam welding (LBW) process or the electron beam welding (EBW) process; both high energy density beam processes. The primary advantage EBW has over LBW and the other welding processes is the ratio of the depth of penetration to the weld region width is greater. Unfortunately this advantage does come at a substantially greater cost for equipment.

The electron beam welding process uses the transfer of kinetic energy, from a stream of extremely high velocity electrons directed at a target, to generate a high energy density beam. Although the mass of an individual electron is quite small, 9.11×10^{-31} kg, a stream of electrons traveling in

excess of 0.7 times the speed of light focused over an area of 0.008 to 0.03 mm^2 can impart a substantial amount of energy. With approximately 96% of the kinetic energy of the high velocity electrons being converted to heat, the energy density at the surface of the target can be in excess of $10^5 W/mm^2$. This is normally sufficient to melt or vaporize anything in the path of the electron beam [II-2]. The Wentgate DynaWeld EB welding machine at AEDC has a maximum power output of 15 kW at an accelerating voltage setting of 150 kV and beam current of 100 mA. In a single ampere of current flow there are 6×10^{18} electrons.

The electron beam welding process can be linked back to the x-ray tubes devised by Roentgen, Snook, and Coolidge at the turn of the twentieth century. Developing a welding process was not their intent however. One of the early problems associated with the development of x-ray tubes was melting of the target material as the supply current was increased [II-3]. Although the melting of targets by high velocity electrons may have been a problem for the early investigators of x-rays, the intentional melting of targets, also known as weldments, is the basis for EB welding. The life-time of a tube was, and still is, limited by the electron flow rate or current generated between the source and target. Increasing the current, beyond established limits, shortens the life of or destroys the tube by raising the temperature of the

target beyond its melting point; destroying the target, the tube, and the ability to generate x-rays.

Though Roentgen, Snook, and Coolidge do not appear to be credited with the invention of the EB welding process, they were among the first to deal with it.

The basic layout for a modern EB welding machine is shown in Figure II-1². The electrons used in the welding process are generated by thermionic emission occurring at the tungsten filament or ribbon housed in the electron gun assembly. The negatively charged electrons are attracted toward the positively charged plate and further accelerated and focused by the magnetic lens. Finally the electrons, now travelling at approximately 0.7 times the speed of light, impinge on the positively charged target or piece to be welded. The negative charge of the electrons provides a convenient means to manipulate the direction and velocity of the electrons.

The EB welder at AEDC has the capability to be programmed in several modes that include rotating the beam of electrons, at various rotational frequencies, in circular or elliptical paths according to the standard formula for an ellipse,

$$\frac{x^2}{a^2} + \frac{y^2}{b^2} = 1. \quad (\text{II-1})$$

² All figures and tables are located in the Appendices.

The rotation of the electron beam is driven by electromagnetic fields acting upon the electrons.

The electron beam is deflected at an angle of approximately three degrees off of normal to provide a horizontal component to the electron beam that induces rotational movement of the weld pool. The spreading of the electron beam diameter that does occur is typically less than five percent of the beam focal length [11-4]. At small angles,

$$\tan\theta \approx \theta. \quad (11-2)$$

The beam spread can be approximated by

$$\text{Beam Spread} \approx \theta \times (\text{beam focal length}). \quad (11-3)$$

Since the focal length is a control parameter the beam spread can be controlled by altering the focal length of the electron beam.

The rotation of the electron beam is controlled independently of part translation and can be done with or without translation of the item to be welded. Stationary circular welds can be easily performed by regulating input parameters for the electron beam path.

By spreading the electron beam over a larger area the power density at the surface can be readily controlled. This technique can be used to reduce porosity in welds or smooth a rough surface resulting from excessive power input or volatile components.

The actual shape that the electron beam takes is a function of the electromagnetic focussing device and the aperture opening. Although the size and mass of the electron are very small, any useful current flow for EB welding makes it impossible to focus the beam to an infinitely small point. Collisions and repulsive effects with other electrons prevent focussing to a single point. Circular electron beams of uniform energy density with diameters of 0.1 *mm* to 0.2 *mm* are common [11-4]. By diffusing the electron beam at the outer edges a near Gaussian energy distribution can be generated.

The EBW process can be used in various modes. The electron beam can be focused above the surface, for diffuse surface heating or conduction based welding, focused at the surface for a more intense power density relying on thermal conduction to distribute the energy, or focussed below the surface. This process flexibility allows the EB welding machine to be used for heat treating, brazing, or other surface modification processes. When the electron beam is focused at or below the surface, keyhole mode welding is generally the goal. Keyhole mode welding is possible with any of the three focal points, depending on power input and the part to be welded.

In keyhole-mode welding, the intensity of the electron beam vaporizes the material beneath the electron beam and melts the material adjacent to the electron beam. The vapor pressure of the vaporized metal forces the flow of molten material beneath and around the electron beam and out to the surface

behind the electron beam itself as depicted in Figure II-2, a schematic drawing of a keyhole weld. The EBW process permits deep narrow keyhole welds to be made. The advantage to this type of weld is the complete penetration of the weld material, in one pass, with minimal heat input when compared to traditional methods such as the shielded metal arc welding process (SMAW) that is sometimes referred to as "stick welding."

Residual stress and the resulting distortion are functions of non-uniform thermal gradients and total energy input [II-5]. Large energy inputs result in wide welds with large temperature variations or thermal gradients. The EBW process has the capability to produce welds with far less thermal distortion since the total energy input is low. This should not be interpreted to mean there is no thermally induced distortion.

Component design must also be considered in an overall effort to minimize distortion or residual stress. The residual stress from a circular EB weld, made while developing weld parameters for AEDC, in OFHC Cu, was capable of resisting a pressure of 245 *atm* before failure even though the entire weld region was machined away.

The typical EB weld is made in what is called a hard vacuum (10^{-6} *Torr*). A pressure of one *Torr* will support a column of mercury (Hg) 1 *mm* in height. Standard atmospheric pressure is given as 760 *mm* of Hg or 760 *Torr*. Unwanted interactions with oxygen and nitrogen are eliminated as a result of

the vacuum conditions. This is particularly important when dealing with welds of highly reactive materials such as titanium (Ti) and Cu. Care must be taken to allow adequate cooling prior to opening the vacuum chamber when dealing with materials that react vigorously with air at high temperatures. Cooling, within the vacuum chamber, is accomplished through thermal conduction or radiational cooling. Back filling the vacuum chamber with an inert gas such as argon is possible if rapid cooling is required.

The EB welding process can also be accomplished in what is referred to as partial vacuum conditions and in air. In both cases the efficiency of the process is greatly diminished due to the beam of electrons colliding with the molecules in the air. To overcome the problem with gas molecule collisions, the power level used to weld in these conditions is increased well above the level used with normal hard vacuum conditions and the distance between the electron gun assembly and surface is reduced. The net effect is a broadening of the beam that leads to increased thermal input to the weld. This comes at the expense of increased power supply sizes and increased equipment cost. These techniques are typically used on components that can resist the resulting residual stresses. Heavy duty truck axle housings are often welded using this process.

One other negative feature of the EB process is the generation of x-rays. Any time a charged particle changes velocity, electromagnetic radiation

is emitted. In the case of the EBW process, negatively charged electrons are abruptly halted; bremsstrahlung radiation (breaking radiation) is emitted as a consequence. The intensity of the x-rays is not great and these rays can normally be stopped by the walls of the vacuum chamber. Equipment that operates in atmospheric conditions does require shielding to protect workers.

B. The Cu-Fe System

Little information on the Cu-Fe is available compared with the more traditional alloy systems such as Cu-Zn or Fe-C; systems commonly used due to the existence of a substantial body of knowledge and experience. As a result, it is not surprising that the Cu-Fe system is seldom used.

The lack of commercial use of the Cu-Fe system has been attributed to ignorance of the system and preconceived notions of the inability to weld these two materials [II-6]. This belief in the inability to weld Cu to Fe is linked to the apparent immiscibility of liquid Fe in liquid Cu. It is generally accepted that there is a region of immiscibility in the Cu-Fe system, however, it does not appear to be explicitly defined. E. G. West expressed doubt concerning the complete miscibility of liquid Fe in liquid Cu and suggested that the presence of carbon (C) may be a contributing factor to the immiscibility [II-7]. Nakagawa attributed liquid phase separation to the presence of small amounts of impurities, including carbon [II-8].

Commercially available Fe-Cu alloys that are routinely welded include Corten B steel (USX, U. S. A.), Cupten (Nippon Kokan K.K., Japan), and Nicuten (Algoma Steel Corp. Ltd., Canada). These steels contain from 0.4 wt% Cu to 1.35 wt% Cu as well as other elements [II-9]. The effect of these other elements on the welding process can not be neglected; as a binary system no longer exists.

The binary phase diagram for the Cu-Fe system [II-10], shown in Figure II-3, gives no clear indication of immiscibility; it shows no monotectic region. A monotectic region is characterized by two liquids, immiscible over some compositional range, that upon cooling, transform into a solid phase and a liquid phase.

An isothermal cut, a horizontal traverse of the binary phase diagram beginning at 0 wt% of one element and ending at 100 wt% of that same element, should include a series of phase changes from a single phase region, to a two phase region, and to another single phase region [II-11]. This would require that any monotectic region be surrounded by a single liquid phase at a temperature above the monotectic reaction temperature. This is clearly not the case in the Cu-Fe phase diagram shown in Figure II-3. An example of a monotectic region, completely surrounded by a liquid, is clearly shown in Figure II-4, the equilibrium phase diagram of the Cu-Pb system [II-12].

The Cu content in the Fe phase varies from 0 wt% to 7.6 wt% at elevated temperatures and from 0 wt% to 2.2 wt% at 850°C [II-13]. Below a temperature of 600°C the solubility of Cu in the Fe is extremely small. It must be remembered that these phase diagrams being discussed are for equilibrium conditions, conditions that allowed sufficient time for all reactions or changes to occur. At the opposite end of the Cu-Fe phase diagram the amount of Fe dissolved in the Cu rich phase approaches 0 wt% below 600°C.

C. The Fe-Ni-Cr System

A ternary phase diagram, containing three elements, is used to represent this system. Often it is referred to as the "Stainless Steel" phase diagram since the primary elements of the commercial stainless steels are Fe, nickel (Ni), and chromium (Cr), and therefore it is considered applicable for this study on 304 stainless steel.

Two forms of ternary phase diagrams are in use, the constant temperature or isothermal cross-section form and the projection, onto an equilateral triangle, of liquidus or solidus temperatures represented by isotherms [II-14]. Points in equilibrium between various phases are shown with bold boundary curves. The equilibrium temperature at an intersection of phases is sometimes included in the phase diagram. The "projected isotherms" version offers greater flexibility to the user; only one diagram is

needed. The isothermal equilibrium ternary phase diagram for the Fe-Ni-Cr system, with the composition for 18-8 SS indicated by the arrow, is shown in Figure II-5.

The 304 SS used for this study had a composition of 19 wt% Cr, 9 wt% Ni, and the balance Fe. This falls in the γ Fe-Ni region (austenitic face centered cubic (fcc) grade stainless steel).

D. Solidification Forms in Alloys

During the solidification process, pure metals normally form a planar solid-liquid (S/L) interface [II-15]. This is not the case for alloys. There are four basic solidification structures associated with alloys; planar fronts, cells, columnar dendrites, and equiaxed dendrites. Interpretive sketches for descriptive purposes of each of these basic structures can be seen in Figure II-6.

The structures that form during solidification are a function of the alloy composition, temperature gradients, and cooling rates. It is not unusual for more than one of the four structures to be present in a solidified weld region. Small cellular structures adjacent to the weld region wall transition into columnar cells and then into columnar dendrites. Fine equiaxed dendrites will tend to form in the center region of the weld region

D.1. Planar front solidification

Identifiable regions of plane front solidification are usually absent in alloy welds due to very specific solidification conditions that must be satisfied. The conditions for maintaining a planar S/L interface therefore maintaining plane front solidification, based on thermodynamic considerations only, have been described by Chalmers and others in what is called the constitutional supercooling theory (CST) [II-16]. The term constitutional is included to clearly identify that the supercooling that occurs is the result of compositional effects and not temperature [II-17]. Mullins and Sekerka have included interface kinetics and heat transfer in their interface stability theory. The CST will be used to quantify the requirements for maintaining a planar S/L interface for this effort.

Mathematical considerations used to simplify the CST include the use of straight liquidus and solidus lines. Over small concentration ranges this is a reasonable approximation. This allows the equilibrium partition ratio to be defined as

$$k = \frac{C_s^*}{C_L^*} \quad (\text{II-4})$$

where C_s^* and C_L^* are the compositions of the liquid and solid respectively at the S/L interface. The slope of the liquidus, m_L , is defined as dT/dC_L . Further

assumptions include local equilibrium at the S/L interface and no diffusion in the solid phase.

The following explanation of the criterion for maintaining a planar S/L interface is based upon that of Kuo [II-18]. At some point during solidification, an alloy of bulk composition C_0 has its S/L interface at a temperature T^* as shown in Figure II-7 [II-19]. Figure II-7 is for illustrative purposes only. No scales are implied or intended.

As the alloy solidifies, solute is rejected ahead of the moving S/L interface causing the composition just ahead of the interface to increase. Recall that the solubility of a solute in a solid decreases with temperature. It is assumed that diffusion is the only transport mechanism for movement of solute away from the interface. Even if mixing by convectively driven flow was present, a very small stagnant boundary layer ahead of the S/L interface would still exist. The concentration gradient would be imposed over a smaller distance.

The equilibrium liquidus line, $T_L=f(C_L)$ in Figure II-7a, a function of liquid composition, can then be used to obtain an equilibrium liquidus-temperature, $T_L(x)$, profile in the liquid as a function of distance. In Figures II-7c and II-7d, $T_L(x)$ is shown with an arbitrary curvature. If at any distance ahead of the S/L interface the actual temperature in the liquid, $T_{actual}(x)$, is less than $T_L(x)$ then that liquid is said to be supercooled. The slope of $T_{actual}(x)$ is defined as G , the

temperature gradient of the liquid as a function of distance ahead of the S/L interface.

The criterion, given by Chalmers, for avoiding supercooling is that the gradient of the actual liquid temperature, G , must be greater than that of the gradient of the equilibrium liquidus line $T_L(x)$ evaluated at $x=0$. This requires that

$$\left. \frac{dT_L(x)}{dx} \right|_{x=0} \geq -\frac{Rm_L}{D_L} (C_L^* - C_S^*) \quad (II-5)$$

where R is the solidification rate, m_L is the slope of the equilibrium liquidus, and D_L is the thermal diffusivity of the liquid. This can be further simplified to the more usable format

$$\frac{G}{R} \geq \frac{\Delta T}{D_L}. \quad (II-6)$$

A complete mathematical derivation for the CST criterion can be found in Appendix C.

If the temperature of the liquid in front of the S/L interface exceeds the equilibrium liquidus temperature then planar growth is assured. Any perturbation that may form would be contained in a region above the melting temperature for that particular composition and would melt back. For the case where the actual temperature ahead of the S/L interface was below the equilibrium liquidus temperature any growth that formed would survive causing the breakdown of the planar growth front leading to a cellular form.

As the degree of supercooling increases the morphology of the solidification structure changes from planar to cellular and finally to dendritic. The transition from planar to cellular can be predicted using the CST but the capability to successfully predict the transition from cellular to columnar dendritic is not yet available [II-20].

Relationships based on experience, between temperature gradients and solidification rates, have been put in chart format for use in predicting solidification microstructure [II-21].

D.2. Cellular solidification

The breakdown of the planar interface due to supercooling, described above, results in the formation of surface perturbations that grow into cellular structures. The rate at which the cells grow is a function of the supercooling and concentration gradients in the liquid. The ratio of the temperature gradient in the liquid to the solidification rate, G_L/R , is often used as a figure of merit to describe conditions present during the solidification process. The thermal gradient in the liquid and the solidification rate can be varied independently through a wide range and still maintain a cellular solidification form. The change that does occur is in the size of the cellular structure. Using G_L equal to $10^\circ\text{C}/\text{mm}$ and the weld velocity (V) set at 0.1 mm/s gives a reasonable assurance of developing a cellular structure in an alloy that has a freezing range of 50°C [II-21].

The growth of the cellular structure is parallel to but in opposite direction to the heat flow. If the heat flow is out of the weld region, the cells will grow into the weld region. The axes of the cells may be aligned with the heat flow direction but do not have any preferred crystallographic growth direction [II-22].

For the case where the heat flow is from a solid into a liquid, as in the case of equiaxed dendrite structures, growth would have to be in the same direction of the heat flow. Kurz and Fisher do not include this type of growth as a "normal solidification condition" and state that cells can not grow under these conditions [II-23].

The breakdown of the planar front can be the result of thermal fluctuations, compositional changes, and surface tension effects. Of these three possibilities only surface tension effects are not explicitly included in the CST. Flow driven by surface tension effects could alter the local composition at a planar interface by bringing a fluid of slightly different composition to an already thermally perturbed surface. The change in composition could reduce the supercooling or increase the degree of supercooling ensuring planar growth of accelerating the breakdown of planar growth respectively.

The spacing cellular structures assume is determined by temperature and compositional gradients. The cellular growth process appears to optimize its growth to meet changing conditions. The spacing, λ , of the perturbed

surface can be reasonably approximated by a sinusoidal function relating the amplitude of the perturbation to cell spacing and position on the surface and has been given as

$$z = \epsilon \sin(2\pi y / \lambda). \quad (\text{II-7})$$

The amplitude is given by ϵ , λ defines the period or peak-to peak spacing of the sine wave, and y defines the position along the planar front surface [II-24].

As the tip of a slight perturbation develops, an increase in surface area results. The ability to transfer solute at the tip is enhanced and growth is accelerated. The tip also is growing opposite to the heat flux. In the region between the cells the composition of the fluid does not fluctuate greatly since the growth rate here is slower than at the tip. Any lateral growth would also not be in the proper orientation relative to the heat flux.

D.3. Dendritic solidification

As supercooling increases, the cellular form develops into columnar dendrites, a tree like structure with regularly spaced stalks and branches. The main dendrite stalk is referred to as the primary dendrite and has a regular spacing, λ_1 . The arms that grow perpendicular to the stalk are the secondary arms and also have regular spacing, λ_2 . Arms growing perpendicular to the secondary arms are sometimes found; these are called tertiary arms.

Dendrites, unlike the cellular structures, grow along a preferred axis that is closest to the heat flux direction; columnar dendrites grow away from

the heat flux direction while equiaxed dendrites grow parallel to and in the same direction as the heat flux. For the cubic crystals this direction is in the [001] direction. Competitive growth occurs early in the solidification process due to the initial cells not being aligned with the preferred dendrite growth directions. Those dendrites in the preferred orientation grow at a faster rate and dominate the resulting structure. Despite this competitive growth, regular primary dendrite spacing is retained throughout the solidification process.

The dendrite will grow in length until it impinges on another dendrite. At this point the dendrite will grow by adding mass to the stalk and arms providing material for solidification is present. The region between the dendrites will contain material rejected or segregated during the solidification process. This segregate, since it is the last material to solidify, will have a major impact on the properties of the solid.

E. Monotectic Solidification

A monotectic region of a binary phase diagram, such as Figure II-8, is characterized by two liquids, L_1 and L_2 , immiscible over some compositional range, that upon cooling below the monotectic reaction temperature, T_m , transform into a solid phase and a liquid phase [II-25]. Note that Figure II-8 is a hypothetical phase diagram for demonstration purposes only. The lack of miscibility of the two liquids is attributed to differences in the forces of attraction between the different atoms within the liquid region. If the attractive

force between L_1 atoms exceeds the force of attraction between L_1 and L_2 atoms, energy input will be required to force mixing. Without this energy input the Gibbs free energy of the system is minimized by the separation of the two liquids.

The monotectic reaction is generally expressed as



where L_2 solidifies at a lower temperature than L_1 .

During the monotectic solidification process L_1 reacts to form the solid α phase in the presence of L_2 . Density differences of the liquid phases and surface wetting of the α phase influence the resulting microstructure. If the L_2 wets the initial α phase material further growth of the α phase is impeded. The insolubility of the α in the L_2 phase prevents growth due to mass transfer by diffusion. Nucleation of new α phase particles is the only way to deplete the L_1 . Upon reaching T_m the L_1 that has not been converted to the α phase transforms to L_2 . The relative volume amount of the L_2 that ultimately solidifies is usually much less than the volume fraction of the α phase.

Some caution is recommended when attributing the apparent small volume fraction of the α phase that solidifies solely to the monotectic reaction. In the case of the Cu-36 wt% Pb alloy, the density of lead at its melting point is $10.67 \times 10^3 \text{ kg/m}^3$ while the density of Cu at its melting point is 8.00×10^3

kg/m^3 [II-26]. The Pb is 1.33 times as dense and had only 36% of the total weight to start.

Figure II-4 is the phase diagram for the copper-lead (Cu-Pb) monotectic system. A micrograph of an as cast sample with 36 *wt%* Pb is shown in Figure II-9 [II-27]. The Pb-rich material that has solidified has filled in the regions between the Cu-rich dendritic areas. A micrograph of an as-cast sample of a 6.83 *wt%* Pb sample, Figure II-10 [II-28], is included to show the variation in solidification structures that are possible. In both micrographs the light areas are the Cu-rich matrix and the dark areas are Pb-rich.

F. Thermal Residual Stress in Weldments

Residual stress in a structure has been defined as that stress that remains after all external loads have been removed [II-29]. In the case of weldments the source of the stress is variations in thermal gradients that occur in restrained portions during the cooling cycle of the welding process. Some form of physical restraint is essential in the development of residual stress.

A portion of a weldment that is uniformly heated and cooled in an unconstrained environment will not develop residual stresses upon cooling; dimensional changes that occur during heating will revert to the original values on cooling since no plastic deformation occurred. If the same portion of the weldment is constrained during heating, but not during cooling, any upset or

displacement of material, caused by compressive loading due to expansion on heating, would reduce the compressive loading to below the elastic limit. The degree of upset would be a direct function of the amount of stress caused during the heating cycle. Upon cooling the region would return to a stress free state; the upset region would not be restrained so no tensile forces would develop. There would be some dimensional change due to the plastic deformation that occurred during the heating cycle.

For the case where the sample region is restrained on heating and cooling and the elastic limit of the material, at some point, is exceeded, residual stresses will result. Expansion due to thermal growth will lead to compressive forces within the weldment. The onset of local plastic deformation will reduce the stress level and relocate material. As the heated portion of the weldment that is restrained cools, the material contracts, but since its dimensions have been altered due to displacement, residual tensile forces develop. The last portion to cool generally is in residual tension [II-30].

F.1. Residual stress in the weld region

The following explanation of the development of residual stress in the weld region has been adapted to the EB process. It is for stresses that develop in the weld region transverse to the direction of the weld, plane YZ in Figure II-11 a schematic of the reference planes used to describe welds for this study, stresses that occur on either side of the weld region. This allows

consideration of the effects keyhole welding has on the residual stress. Assumptions include weldments at room temperature, fixed welding conditions in the keyhole mode, and one homogeneous material. The transfer of energy ahead of the beam into the unmelted region is limited by thermal conduction from the weld region and is ignored. Figure II-12 provides some perspective on the sequence of events that occur during the weld solidification process. No time or distance scale is implied.

At the onset of keyhole mode welding at $time=0$ as shown in section B-B of Figure II-12a, the electron beam vaporizes a cylindrical volume of metal beneath the electron beam. Molten material begins to flow around the perimeter of the electron beam and beneath the electron beam where the energy density of the electron beam has been reduced to a level insufficient to melt the material. The vapor pressure of the vaporized metal contained in the keyhole region drives material out of the keyhole behind the electron beam leaving a characteristic herring bone pattern. Fluid expulsion is especially prominent just prior to the EB reaching the end of the material. This end result can be seen in Figure II-13, a top view (XY plane) of a sample at the end of the weld path.

As the electron beam translates a molten pool develops just behind the electron beam. This pool of fluid expands due to the drastic increase in temperature, from room temperature to the boiling point of the material, but

exerts no pressure on the confines of the weld region. The liquid metal is free to expand out of the cavity.

The molten material begins to transfer energy by convection to the weld region boundaries where it is conducted away as shown at *time=a* in section C-C of Figure II-12b. As the temperature at the edge of the weld region begins to rise the metal begins to expand. At the same time the temperature in the weld region center begins to drop. No additional energy is being input to the weld region center and there are no energy source terms to consider. A quick energy balance will show that the energy being transferred to the edge of the weld region had to come from the weld region center and in the process the weld region center's temperature had to decrease. This results in tensile forces being formed in the weld region center while compressive forces are being generated in the region away from the weld region center. The forces must be balanced otherwise the sample would have to move. Recall that for a stationary object the

$$\sum F = mA \equiv 0. \quad (II-9)$$

The continued transfer of energy away from the weld region center ultimately results in a region of high tensile stress in the weld region center that is balanced by compressive forces some distance away from the weld region shown at *time=b* in section D-D of Figure II-12c. These stresses remain after cooling.

If the residual stresses in the center of the weld region exceed the elastic limit of the material plastic deformation will occur. The weldment will develop an obvious curl. In the situation where the residual stresses exceed the yield strength of the material in the weld region, cracks develop.

Often the residual stress within the weld region exceeds the yield strength at the elevated temperatures just after solidification has occurred. At these elevated temperatures the yield strength within the weld region is substantially lower than at room temperature. The cracking within the weld region that results at the elevated temperatures is generally termed hot cracking. Cracking in the weld region will continue as long as the residual stress exceeds the yield strength of the material.

The weld region is a three dimensional volume and the residual stresses that develop between the top and bottom of the weld region need to be considered. Fortunately the same basic principles used above apply.

Shortly after the weld pool has formed the top and bottom surfaces of the weld pool begins to cool, *time=a* in Figure II-12b. The material beneath the surface is still expanding while the outer surfaces are contracting inducing tensile stress on the surface and compressive stresses internally. As the inner portion cools additional tensile loads are developed. When these tensile loads exceed the yield strength cracks parallel to the top surface form

Energy from the weld region is removed through both the top and bottom surfaces and laterally into the bulk material. If the dominant mode of heat transfer is in the lateral directions into the bulk material then the resulting tensile stresses will pull in those directions. If the tensile stress exceeds the ultimate strength of the metal, at the given temperature, a crack or cracks perpendicular to the surface will form. If the dominant heat transfer mechanism transfers energy through the top and bottom surfaces then cracks will form parallel to the top surface, as noted above, due to the tensile forces on the surfaces pulling in those directions.

G. Liquid Metal Embrittlement

Liquid metal embrittlement (LME) has been described as the reduction of fracture stress and the loss of ductility caused by liquid metal in contact with the surface of a solid metal [II-31]. This should not be interpreted to mean that simply contacting one solid surface with molten metal would automatically lead to failure. The mechanism that leads to the embrittlement must be considered.

A natural place to initiate LME is at the grain boundary (GB) level. In an alloy this material is the last material to solidify and the first to melt. The area to volume ratio for the GB material is very high; the actual volume of material in the GB is relatively small but is spread over a large area. Since the GB surface area is large the free energy of the surface is also very high. This

free energy is a measure of the number of available bonds for chemical reactions. Increasing the number of available bonds increases the potential for chemical reactivity.

The method used to get a molten metal into a GB must be considered when considering LME. The spacing between the GB is small so any movement into the GB by a liquid phase metal would need to be by capillary action. This will move material, but at what rate? If the GB could be displaced, or opened, the molten metal could flow much easier.

In a heated environment such as the weld region, the GB area adjacent to the weld region is also heated. This region is sometimes referred to as the heat affected zone (HAZ). During the heating and cooling of the weld-solidification process the weld region experiences both compressive stresses and tensile stresses. On heating, the edge of the weld region is placed in compression. Molten metal under pressure could possibly be squeezed into the GB; this is unlikely since the GB spacing under compression would not be any greater than under normal conditions. Viscosity effects at the atomic level would also hinder pressure-assisted capillary driven flow.

If tensile loads in the HAZ were sufficient to open the GB then capillary driven flow could be accelerated. This would require a source of molten material be available while the GB was in tension. It has long been

recognized, by Nippes, that tensile stresses are required to initiate LME [II-32].

The presence of low melting point and weak metal in the GB does not present a problem until the item is put into service while hot or cooling occurs. For this scenario, the current GB area has been filled with a material that is of lower ultimate strength than the original GB material. As the region surrounding the GB cools additional tensile stresses develop. The weak link is the new GB area filled with the weak material. The ultimate tensile strength (UTS) at the GB is now greatly reduced.

Another possible problem is the hydraulic force that develops within the GB during cooling around the liquid material. It is assumed that the GB area that has not been infiltrated with molten metal is still relatively warm, still in a weakened state. As the region around the GB contracts the hydraulic pressure within the liquid filled GB pushes back forcing a crack to propagate along the GB. This is very similar to attempting to seal a can of paint with the rim full of liquid, the harder one tries to close the lid the more the lid tries to open. If the surrounding region is placed in tension at any time in the future this crack can possibly open further.

CHAPTER III

EXPERIMENTAL PROCEDURES

A. Sample Preparation

Fifty test coupons of OFHC Cu and 304 SS, with chemical compositions listed in Table III-1, were cut to the dimensions shown in Figure III-1. The contact surfaces of both the Cu and 304 SS coupons were machined flat and perpendicular to the top surface of the coupons to ensure complete contact during the welding process. This procedure minimized the possibility of high velocity electrons traveling past the actual sample without imparting any energy to the weld zone.

All the test coupons were manually cleaned in acetone, mechanically abraded with a medium grade of non-woven abrasive (Scotchbrite®, green), and finally rinsed with acetone to remove any residual abrasive material. No effort was made to determine the presence of residual abrasive particles in the specimens. This cleaning procedure duplicates that normally used within AEDC's EB weld area.

B. Electron Beam Welding

Every attempt was made to minimize the number of variables influencing the welding results. Table III-2 lists the machine parameters

employed. All parameters were controlled by a simple welding program developed for this experiment. The conditions at 0 *Hz* were used as a baseline for comparison; these conditions had previously been used at AEDC on similar welds and had provided complete weld penetration.

Electron beam rotation frequency was the only intended variable. The electron beam rotation frequency was varied from 0 *Hz* to 180 *Hz* in steps of 20 *Hz*. Five sets of samples were welded at each of the ten electron beam-rotation frequency conditions. The decision to limit the step size to 20 *Hz* was based upon the assumption that examination techniques available at AEDC would not be able to detect any changes resulting from the use of a smaller step size.

Rotating the electron beam while translating the beam does not affect the total energy input to the weld. At a fixed translational velocity, the same amount of energy is input at the weld region surface whether the beam rotates or not. When the beam is being rotated, the actual area covered by the beam changes compared with the non-rotating case. The energy density distribution within the beam also changes due to rotation.

A simply analogy of the weld-pool fluid movement during the EB welding process to the dragging of a molten pool of metal during gas welding of thin sheet metal can be made. Once a molten pool of metal is established on a piece of thin metal with an oxygen-acetylene torch as a heat source, the

pool can be easily dragged along the surface. Surface tension between the hot and cold materials imparts a small amount of momentum to the region. As a result, the molten pool tends to follow the tip of the torch; stopping the torch abruptly does not necessarily stop the movement of the weld pool.

As the electron beam is rotated the molten pool within the keyhole also rotates in the direction of the electron beam rotation. Since the molten weld pool is rotating, the effects of convectively driven fluid flow will appear. Welds tend to be wider than without beam rotation and the degree of mixing within the solidified weld region changes. When the effects of fluid flow due to beam rotation are combined with the linear motion of the molten fluid, undercut at the weld region edge sometimes results [III-1]. Material that is not necessarily liquid can be stripped off or away from the weld region wall and mixed into the bulk. As the electron beam rotation frequency is increased the degree of weld region mixing is also expected to increase. Measurement of the actual weld pool rotation frequency was not done.

The ability to rotate the electron beam in an elliptical orbit provides a means to deliver a diffuse electron beam to the surface. This reduces the energy density input at the surface, reduces regions of excessive energy input, and alters thermal gradients within the region. Control of these factors by electron beam rotation can be used to reduce porosity during keyhole mode welding or to smooth the surface of rough welds. The energy-density

distribution pattern, without electron beam rotation, is a uniformly filled circular distribution while the energy-density distribution pattern resulting from the rotation of the electron beam resembles a Gaussian distribution [III-2]. It was expected that the use of electron beam rotation would affect the extent of weld penetration as the electron beam rotational frequency was increased.

Rotation of the electron beam was accomplished using beam control software provide by the EB welding machine manufacturer (Wentgate DynaWeld). The software provides control of the actual rotational frequency and the electron beam path. The actual beam path follows that of an ellipse, given previously by equation II-1, with control of the major axis and minor axis as an operator input. For this effort, a circular path, a form of an ellipse,

$$x^2 + y^2 = r^2, \quad (III-1)$$

was used.

The electron beam diameter, with rotation, used for this study was estimated to be 0.5 mm. This estimate was provided by the manufacturer of the EB welder. No actual measurements of the electron beam diameter have been performed at AEDC.

It was recognized that machine variability and human error were potential sources of variation in the weld results. At the time this work was done, AEDC's experience with the Wentgate EB weld machine was limited and no studies had been performed to quantify these two sources of variation.

Simply repeating a procedure and obtaining matching results had been considered sufficient justification to assume there was no variation due to the machine or operator.

Two sets of specimens at a time were placed on a support platform in the vacuum chamber of AEDC's Wentgate DynaWeld high-voltage 15 kVA EB welding machine and welded. The specimens were not restrained in any manner. Although it was recognized that welding two specimens at a time was quite inefficient, this procedure ensured that each sample was placed at a common starting point that could be easily referenced.

One of the samples was welded with the electron beam rotation frequency set at a specific frequency and the other sample within the vacuum chamber was welded with the electron beam rotation frequency increased by 20 Hz. For example, the sample labeled 1-1 was welded with the electron beam rotation frequency set at 0 Hz. Sample 1-2 was inside the vacuum chamber at the same time. Sample 1-2 was welded with the electron beam rotation frequency set at 20 Hz.

After a short cool down period, approximately 10 to 15 minutes, within the vacuum chamber, samples 1-1 and 1-2, were removed. Samples 1-3 (40 Hz) and 1-4 (60 Hz) were then placed in the vacuum chamber and welded. This scheme was repeated until all 10 frequency settings were used. This produced the welded sample-set 1-1 through 1-9. The process was then

repeated for another 10 samples; generating sample-set 2-1 through 2-9. A total of five sample-sets were made.

During welding, the electron beam was centered on the interface between the Cu and 304 SS sections as shown in Figure III-1 with the beam focused at the surface. Location of the center of the interface was accomplished using the external optical viewing scope that was supplied with the EB welder. The magnification of the viewing scope was 10X. The center of the cross hair of the viewing scope is mechanically aligned with the center of the electron beam during viewing. Setting the cross hair at a given position defines the electron beam center.

To define the electron beam path for an arbitrary straight line, a center point is defined at the beginning and end of the weld path. Movement of the weldment on a support table within the vacuum chamber is accomplished either by rotating manual positioners or by actuating electronically controlled drivers. Rotary encoders record the positions used to define the endpoints. The numerically controlled drive unit positions the weldment at the starting point prior to the initiation of welding and controls all movement of the part along the path during welding.

This manual process for defining the weld start and end points does introduce the potential for operator error. Fortunately, the process makes precision machining and locating of weldments prior to welding unnecessary.

This process is suitable for short-run welding operations. Location of the electron beam focal point is an input to the machine software.

C. Test Specimen Preparation

The welded coupons were machined and cut into tensile specimens similar to that recommended by the American Welding Society (AWS) specification B4.0-85. Traveling-wire electrical-discharge machining (EDM) was used to cut the samples to the dog-bone shape. The EDM technique was selected to minimize any influence excessive heating or work hardening, that would occur using more conventional cutting techniques, may have had on the weld region. An overview of the sample configuration is shown in Figure III-2.

All "dog-bites," two from each dog-bone, were retained for examination for this study; one set of samples was used for the previously referenced study by Dahotre and Zysk [III-3]. The relative position of each dog-bite with respect to the weld start position was recorded on each dog-bite. Only the samples from the weld-finish region (approximately 12 *mm* from the end of the actual weld bead) were used for the earlier study. This gave reasonable assurance that all the welds had reached steady state conditions.

The samples used in this study included dog-bites cut from the weld-start region of the same weld coupons used in the Dahotre-Zysk study (approximately the first 12 *mm* of the sample) as well as from samples not

previously used. The early samples have been inadvertently destroyed. It is believed that the 12 *mm* distance from the initiation point of welding still allows adequate time for steady state conditions to develop.

One representative sample from each frequency class was mounted, polished, and etched for examination by optical microscopy. Abrasive polishing was done using 120, 180, 220, 320, 400, and 600 grit silicon carbide abrasive disks with a constant flow of water. Final polishing was done using 6 μm diamond paste and 0.3 μm agglomerated alumina suspension. Three etchants were used; one a solution of 5 g of ferric chloride, 20 ml of hydrochloric acid, and 100 ml of ethanol, Marble's solution, and a solution of 10 g of FeCl_3 and 90 cm^3 of water. The different etchants were used to reveal different morphologies and increase contrast during examination with the scanning electron microscope (SEM).

D. Optical Metallurgical Examination

The "dogbites" from the initial 12 *mm* of the weld coupons 5-1F, 4-2S, 4-3S, 2-4S, 4-5F, 4-6F, 5-6S, 3-7S, 1-8S, 5-9S, and 3-10F were mounted, polished, and etched in preparation for producing micrographs for this research effort as described in the previous section. Micrographs were made using a Nikon Epiphot TME optical metallurgical microscope.

E. Microhardness Testing

Microhardness testing was performed using an automated LECO M-400-H Hardness Testing Machine. The test device was setup with a Knoop indenter, a mass load of 100 *grams*, and an impact time of five seconds. A rectangular grid pattern was used. A schematic representation of the test pattern is shown in Figure III-3.

The distance between rows was set at 200 μm for all samples. The distance between columns was set at 220 μm for the baseline sample and those samples welded with electron beam rotation frequencies of 20, 40, 80, and 160 *Hz*. The column spacing was set at 100 μm for the samples welded with the electron beam rotation set at 60 and 100 *Hz*. The column spacing for the samples welded with the electron beam rotation frequency set at 120, 140, and 180 *Hz* was 300 μm .

The actual number of rows and columns used was a function of the individual sample. Ten samples were tested and all beam rotation frequencies were included in the testing. All test points were located in the YZ plane as shown in Figure II-11 at least 1 *mm* beneath the top surface of the XY plane of the weld to avoid making microhardness measurements in the undercut region of the welds that were filled with mounting material. This distance varied for each sample.

False-color microhardness maps, each map containing from 300 to 800 test points, of the ten samples were generated using software available at the UTSI.

F. Tensile Testing

The top surface and the bottom surface of each test coupon (the XY plane) were milled flat to remove any surface irregularities that would induce stress concentrations during tensile testing. These surfaces had substantial weld bead undercut due to excessive energy input during welding. All 50 welds had complete penetration. The coupon surfaces were milled approximately 12 *mm* from the center of the weld interface. The depth of milling was varied as a function of the weld bead undercut. This process allowed accurate measurement to be made of the cross sectional area that would be put in tension.

All 50 samples were destructively tested using AEDC's Baldwin/Southwork Tate Emery tensile testing machine. A load-rate, σ' , of 6000 *lbf/min* was applied to each coupon until failure. The thickness of the samples to be tested and the limited equipment available at AEDC's metallurgical laboratory dictated the use of the load-rate device instead of a strain-rate test device.

To determine the approximate strain-rate the load rate is divided by both Young's Modulus and the cross-sectional area under load,

$$\epsilon' = \sigma' / (\text{area} * Y) . \quad (\text{III-2})$$

This strain-rate is approximate since it requires the use of a standard value for Young's Modulus. The standard value selected may not necessarily be the correct value for the material being tested.

G. Weld Region Width Measurement

Measurement of the weld region width was done using a Nikon Epiphot TME optical metallurgical microscope equipped with a graduated reticule. Measurements of the weld region width at both the top and bottom surface of each sample were made. The measurements were made in the YZ plane as shown in Figure II-11.

CHAPTER IV

RESULTS AND DISCUSSION

A. General Weld Region Morphology

There was a maximum variation of 32% in the weld region cross-sectional area, in the YZ plane, of the samples welded for this study. The maximum variation occurred between the samples welded at 120 *Hz* and 180 *Hz* (smallest area, both samples had equal areas) and the sample welded at 20 *Hz* (largest area). Figure IV-1 shows the decline in the weld region area versus electron beam rotation frequency with a linear least squares fit of the data included to show the general trend. The baseline sample welded without electron beam rotation had the same cross-sectional area as the samples welded at 120 *Hz* and 180 *Hz*.

The cross-sectional area used was based upon a hypothetical trapezoid formed using the top weld bead width, the bottom weld bead width, and the thickness of the weld coupon. Recall that the area of a trapezoid is given by

$$area = \frac{(a + b) * h}{2} \quad (IV-1)$$

where *a* = top weld bead width, *b* = bottom weld bead width, and *h* = coupon thickness. This represents a reasonable approximation to the actual shape of the weld region in the YZ plane as can be seen in Figure IV-2 of sample 4-3S welded at 40 *Hz*. A detailed discussion of the influence of electron beam

rotation on the top weld bead width and bottom weld bead width can be found in section IV-D.

Each weld had complete penetration and the formation of a bottom weld nugget on the 304 SS side of the weld region. This has been attributed to welding with the beam power (2.64 kW) in excess of the minimum amount required for complete penetration. Top bead undercut was observed in all samples except those welded at 20 Hz, 40 Hz, and 60 Hz. An explanation of a contributing cause of this undercut is discussed in detail in section IV-D. The reader is reminded that a schematic showing the orientation of the planes used throughout this study can be seen in Figure II-11.

Fe-rich dendritic microstructure with Cu rich interdendritic material dominated the weld regions of all samples. These morphologies can be seen in Figure IV-3, a micrograph of sample 4-3F welded at 40 Hz. At electron beam rotation frequencies of 60 Hz or less, distinct regions, within the weld region, containing phases resembling the base material were evident adjacent to the weld region walls. Figure IV-4, an optical micrograph of a portion of the top weld nugget region along the Cu wall of sample 4-3S, welded with the electron beam rotation frequency set at 40 Hz, shows Fe rich particles surrounded by a Cu rich matrix. These regions of Fe rich particles within the Cu matrix can increase the local microhardness.

Welding with the electron beam rotation frequency set at 180 Hz produced a fine distribution of Fe rich particles within the weld region as

shown in Figure IV-5. These small particles tended to reduce the average Knoop microhardness.

Divorced spherical Cu rich phases were observed in regions adjacent to the 304 SS walls of the weld region. These structures are discussed in section IV-F.

B. Microhardness in the Weld Region

The Knoop hardness values within the weld region for the ten samples tested varied between HK78 and HK407. The Knoop hardness for the Cu base material outside of the weld region was between HK80 and HK90 while the Knoop hardness for the 304 SS base material outside of the weld region approached HK400. The reader is reminded that a low Knoop hardness number is associated with a soft material and a large Knoop hardness number is associated with a hard material

A characteristic of all the samples examined was a thin zone, within the weld region, of consistent microhardness adjacent to the Cu wall. Despite the minimal mixing observed with optical microscopy in this zone, the Knoop microhardness increased from approximately HK80 in the base Cu region to HK160 in the narrow region adjacent to the base Cu and within the weld region. Alloying and strengthening due to the presence of Fe-rich second phase materials within a Cu rich matrix were likely causes for this increase in Knoop microhardness. If this zone consisted of pure Cu, the heat from the

weld region would have contributed to grain growth and softening reducing the Knoop microhardness.

Evidence of liquid Fe rich material shearing off Cu rich material along the edge of the weld region was found in several weld samples. The leading edge of the rotating fluid that surrounds the keyhole acts much like the leading edge of an end mill cutter. It shears material from the adjacent wall as it is fed into the material. The Fe rich material then solidified adjacent to the Cu wall.

Optical examination of the sample welded without electron beam rotation, sample 5-1F, showed unexpected evidence of mixing within the weld region that was most likely the result of flow driven by the vapor pressure within the keyhole and surface tension effects. The effects of flow resulting from the buoyancy of liquid or solid Fe in liquid Cu can not be ignored. The density of liquid Cu at its melting point is approximately $8 \times 10^3 \text{ kg/m}^3$ while the density of liquid Fe is approximately $7 \times 10^3 \text{ kg/m}^3$. However, the mixing or alloying of the Cu and Fe during the welding process may negate any influence density differences of the pure metals may have in a quiescent environment. The determination of the effects of flow resulting from any density differences in the liquid metals is beyond the scope of this effort.

The microhardness map, seen in Figure IV-6, for this sample shows two distinct partitions of uniform microhardness over a distance of 7000 μm in the Z direction of the YZ plane. This microhardness map is consistent with optical observations of this sample at low magnification that showed two distinct

regions or layers of material, dissimilar from that of the base Cu, adjacent to the Cu wall of the weld. This dissimilarity supports the claim of mixing due to vapor pressure and surface tension noted previously. The weld region contained a uniform dispersion of spherical Fe rich phases in a Cu rich matrix. This sample provided a baseline for comparison with those samples welded with electron beam rotation. The average microhardness and standard deviation of the microhardness within the weld region for this sample were HK205 and HK31 respectively.

The reader is advised that the microhardness scales, false-color ranges, and dimension scales on the individual microhardness maps contained in this dissertation vary from sample to sample due to software constraints. With the exception of Figure IV-9, the microhardness maps contain test points in both the Cu and 304 SS bulk material outside of the weld region. The microhardness map shown in Figure IV-9 does not contain any test points within the 304 SS bulk material. These restrictions should be kept in mind when making any direct comparisons between the individual microhardness maps.

Microhardness maps of the welds made with beam rotation frequencies of 20 Hz, 40 Hz, and 60 Hz, samples 4-2S, 4-3S, and 2-4S respectively, showed clear evidence of inhomogeneity within the weld region, having widely varying Knoop microhardness values. A completely mixed weld of uniform composition would be expected to have no variation in microhardness within

the weld region. An unmixed weld would have distinct pockets or regions with widely varying microhardness values. The microhardness maps for samples 4-2S, 4-3S, and 2-4S are shown in Figure IV-7, Figure IV-8, and Figure IV-9 respectively. (The microhardness map shown in Figure IV-9 covers approximately one half of the weld region width, the yellow region to the right of the blue area, yet contains 800 test points. It contains no test points within the 304 SS base material.)

Close to the Cu wall within the weld region of each of these three samples, 4-2S, 4-3S, and 2-4S, Knoop microhardness values of HK300 can be seen adjacent to regions with microhardness values of HK180 or less. This region is shown in Figure IV-8. This indicates that the extent of mixing, within the weld region of these samples, has changed from that of the baseline sample. The effects of electron beam rotation on the samples have not contributed to a homogeneous weld region composition but appear to have reduced the mixing when compared with the baseline sample.

The average microhardness values within the weld region for the samples welded at 20, and 40 *Hz* were approximately equal as can be seen in Figure IV-10. Sample 2-4S, welded at 60 *Hz*, had nearly identical average microhardness and standard deviation values, when compared with the samples welded at 20 *Hz* and 40 *Hz*, despite covering only one half the weld region.

The increased average Knoop microhardness for these samples above that of the baseline sample, welded without electron beam rotation, was approximately 33%. This is consistent with the observations made using optical microscopy showing the presence of numerous regions or pockets containing harder Fe rich material, surrounded by a Cu rich matrix. The presence of Fe rich materials would increase the probability of making a microhardness impact test point at an Fe rich site thus driving up the value of average Knoop microhardness.

Figure IV-10 includes the mean Knoop microhardness values within the weld region, the average deviations from the mean Knoop microhardness values, the minimum Knoop microhardness values, and the maximum Knoop microhardness values as a function of electron beam rotation frequency for each electron beam rotation frequency used in this study. A linear least squares fit to the data has also been included for each data set plotted to help visualize trends.

The samples welded with electron beam rotation frequencies between 80 *Hz* and 140 *Hz*, samples 4-5F, 5-6S, 3-7S, and 1-8S respectively, had similar microhardness maps with evidence of increased mixing as the electron beam rotation frequency increased when compared with the samples welded with electron beam rotation frequencies from 20 to 60 *Hz*. As noted previously, a completely mixed weld of uniform composition would be expected to have no variation in microhardness within the weld region. An

unmixed weld would have distinct pockets or regions with widely varying microhardness values. Distinct regions with microhardness values of approximately HK160 and HK240 were observed adjacent to the Cu wall of the weld regions indicating alloying had occurred. The microhardness maps for these samples are shown in Figures IV-11 through IV-14.

The average Knoop microhardness showed a decline as the electron beam rotation frequency increased from 80 to 140 *Hz*. This is shown in Figure IV-10. A 37% decrease in average Knoop microhardness within the weld region occurred when the electron beam rotation frequency was increased from 20 to 140 *Hz*. The average Knoop microhardness for the sample welded with the electron beam rotation frequency set at 20 *Hz* (sample 4-2S) was HK275 while the Knoop microhardness was HK172 for the sample welded with the electron beam rotation frequency set at 140 *Hz* (sample 10-8S). The decline in Knoop microhardness is attributed to increased but incomplete mixing within the weld region. The increased mixing decreases the size of the Fe rich regions that are surrounded by the softer Cu rich matrix. This both increases the probability of impacting Cu rich regions and if an Fe rich region is impacted, the soft Cu matrix surrounding it will tend to yield on impact thus reducing the average value of Knoop microhardness.

As the electron beam rotation frequency increased for this group of samples, the Knoop microhardness away from the weld region walls but within the weld region showed a decline. This is evident by comparing Figure IV-11

with Figure IV-14. The microhardness map of sample 1-8S shows substantially more area with microhardness values of approximately HK160 than does sample 4-5F. The weld region compositions though not homogeneous were comparable to that of the baseline sample. The region of exceptionally large Knoop microhardness, approximately HK400, in the center of Figure IV-12 is due to the presence of Fe rich material at that location.

The microhardness maps of the samples welded with the electron beam rotation frequency set at 160 *Hz* and 180 *Hz*, shown in Figure IV-15 and Figure IV-16 respectively, for samples 5-9S and 3-10F, had nearly uniform microhardness distributions that indicate a more homogeneous distribution within the weld region. This is supported by the average Knoop microhardness within the weld region as seen in Figure IV-10. The microhardness distributions of these two samples are quite similar to that of the baseline sample. Optical examination of these two samples verified that particle size and distributions similar to those observed in the baseline sample existed.

As the electron beam rotation frequency was increased the degree of mixing within the weld region, implied by the microhardness distributions and average microhardness values, also increased. This mixing increases alloying effects within the weld region by physically bringing distinct liquid phases into contact. Even if the liquids remain immiscible the agitation within the molten region tends to produce smaller solidification structures that can increase

microhardness. The average Knoop microhardness increased by more than 12% when the electron beam rotation frequency was increased from 160 to 180 Hz. The distribution of small structures within the weld region of a sample welded at 180 Hz can be observed in Figure IV-5.

This mixing of discrete second phase materials is the basis for the process used to create dispersion strengthened Cu products such as Glid-Cop®, a high strength Cu alloy containing dispersed second phase particles, an oxide of aluminum, in a Cu matrix [IV-1]. Plastic flow is impeded by the presence of the uniformly distributed small and hard spherical particles. A compressive load is borne by the harder particles contained within the matrix as the softer matrix is displaced by the load. This is the same principle employed in composite materials and in the construction of asphalt roadways.

According to SCM Metal Products

$$Yieldstrength = C \left(\frac{f}{r} \right)^{1/4} \quad (IV-2)$$

where C is a constant based on the shear moduli of the matrix and the second phase, and particle-dislocation interactions. The average radius of the second phase particles is given by r and f is the volume fraction of the second phase material [IV-2]. As the average size of the particles decrease the yield strength increases. Typically, material hardness increases with an increase in yield strength.

The transition from the near-uniform distribution of equal microhardness values within the weld region of the baseline sample, to varying microhardness

distributions, and finally returning to uniform microhardness distributions, as the electron beam rotation frequency was increased, provides evidence of the effect electron beam rotation has upon weld region mixing for dissimilar weld couples.

As the electron beam rotation frequency was increased from 80 *Hz* to 180 *Hz* the microhardness distribution approached that of the baseline sample. The negative effects that low-frequency electron beam rotation had upon the weld region microhardness distribution were reduced by increasing the electron beam rotation frequency. Again, evidence of the effect that electron beam rotation has upon weld region mixing for dissimilar weld couples has been demonstrated.

It is apparent that for the conditions used in this study, that while welding with the lower electron beam rotation frequencies increased the average microhardness it also increased the microhardness variation within the weld; these frequencies should be avoided if uniformity of composition is important. Welding with either high electron-beam rotation frequencies, 160 *Hz* or 180 *Hz*, or no electron beam rotation at all should be used if a uniform distribution of microhardness in the weld region is desired.

C. Tensile Test Results

There was no correlation between electron beam rotation frequency and weld strength since each test coupon failed in the Cu base material outside of the weld region. The weld region material was therefore obviously

stronger than the Cu base material regardless of the electron beam rotation frequency. The average stress at failure was 247,550 *kPa* with a standard deviation of 11,633 *kPa*. The average stress at failure for the baseline samples was 258,166 *kPa* with a standard deviation of 3,143 *kPa*. Published UTS values for the Cu base material used range from 234,000 *kPa* to 344,000 *kPa* depending on the amount of processing performed (eighth hard to hard). Tensile tests of the Cu base material were not performed. Figure IV-17 shows the region of ductile failure in a sample welded with the electron beam rotation frequency set at 80 *Hz*.

D. Weld Region Width

Figure IV-18a shows the width of the weld region in the YZ plane as a function of the electron beam rotation frequency along the top XZ plane and the bottom XZ plane surface of the weld specimens. A linear least squares fit to the data has also been included for each data set plotted to help visualize trends. The baseline sample, sample 4-1S, welded without electron beam rotation had a top weld bead width of 2.44 *mm* and a bottom weld bead width of 1.02 *mm*. If the curve fits are used as the indicators, electron beam rotation increased the width in all cases when compared with the baseline sample welded without electron beam rotation. Figure IV-18b excludes the samples welded with the electron beam rotation frequencies of 120 *Hz* and 180 *Hz*. The narrow top weld bead widths of these two samples are believed to be the result of power fluctuations to the electron beam gun during welding. Figure

IV-18b has been included to show the minimal change in the top weld bead width versus electron beam rotation frequency that has been attributed to welding with excess power.

The increase in the width of the top weld bead and the bottom weld bead for those specimens welded with the rotating electron beam, when compared with the baseline case of 0 *Hz*, was expected. The increase is partially attributed to the increase in the electron beam diameter, due to the three degrees off-axis beam deflection, and the resulting decrease in the energy density at the surface [IV-3].

The rotational flow of liquid within the weld pool also contributes to the increased width. The angular velocity of the flow within the pool changes with the electron beam rotation frequency and results in an additional transfer of energy to the walls of the weld region due to convection, more melting of the weld region walls occurs. This energy transferred due to convection is added to that energy transferred from fluid flowing around the keyhole while welding without electron beam rotation.

On one side of the weld pool the fluid will have a maximum flow velocity due to the combined effects of both the linear and angular velocity vector components of the electron beam. The flow velocity on the opposite side of the weld pool will be at a minimum since the rotational component of the electron beam velocity vector along this edge will be opposite to the linear velocity vector of the electron beam [IV-4]. The actual velocity vector, or

effective velocity, at any given time or position, is the vector sum of the two velocity components. This effect can be visualized with the help of Figure IV-19 that also shows that the Cu was always on the side with the minimum electron beam velocity vector and the stainless steel was always on the side with the maximum velocity vector.

Table IV-1 shows the minimum and maximum theoretical velocities along the periphery of the molten weld pool during welding. The values shown in Table IV-1 are based upon an electron beam radius of 0.5 *mm* that is translating at 9.7 *mm/s*. The weld pool was assumed to be rotating at the same rate as the electron beam for these calculations. This assumption ignores any viscous effects that would normally act on a molten fluid in motion. The translation distance per revolution is the distance that the weldment translates in the X direction during the time required for the beam to make one revolution. Note that the minimum velocity vector is actually negative which means the instantaneous direction of propagation is opposite of the direction of the translating weldment.

The negative velocity side of the weld developed weld bead undercut since the energy input from the electron beam was not being transferred away from the keyhole edge as rapidly as on the high velocity side. Since the hot fluid adjacent to the solid Cu is not moving as rapidly as on the 304 SS side of the weld pool, the hot fluid dwells longer near the Cu allowing more material to be melted and drawn into the weld pool. This should not be construed to

mean that electron beam rotation is the only cause of weld bead undercut; electron beam rotation contributes to the problem of weld bead undercut. Consideration of the fact that the Cu melts at a temperature 317 °C lower than the minimum temperature (1400 °C) of the melting range of the 304 SS must also be included.

The top weld bead increased in width from 2.44 *mm* to 2.74 *mm*, an increase of more than 12%, when the electron beam rotation frequency was changed from 0 *Hz* to 20 *Hz*. The increase in the width of the bottom weld bead for the same change in frequency was more dramatic; a 79% increase in width was noted.

Excluding the top weld bead widths at 120 *Hz* and 180 *Hz*, a linear curve fit indicates a small variation, -6×10^{-4} *mm/Hz*, in top weld bead width with electron beam rotation frequency. Including 120 *Hz* and 180 *Hz*, the slope of a curve fit line of the top weld bead width versus beam rotation frequency, found using linear regression, is -2.13×10^{-3} *mm/Hz*. This indicates a greater, but still slight, decrease in the width of the top weld bead with an increase in frequency.

This slight effect is expected since welding with excess power would produce welds of similar width regardless of the electron beam rotation frequency. The diffusing of the electron beam due to rotation would not be sufficient to reduce the power density of the electron beam, at the weld surface, below that minimum power density level required for complete

penetration. Once this threshold is exceeded the efficiency with which the electron beam penetrates increases. If the electron beam diameter is held constant, the addition of power to the electron beam still only vaporizes a well-defined volume; the rate at which the vaporization is accomplished is increased however.

The cause of the 16% decrease in the top weld bead width at electron beam rotation frequencies of 120 *Hz* and 180 *Hz*, when compared with samples welded at 100 *Hz*, 140 *Hz*, and 160 *Hz* has not been completely determined. Some of the variation in weld bead width can be attributed to the location in the sample where the measurements were made.

The top weld bead of sample 3-4F, welded at 60 *Hz*, showed cyclic variations in the weld region width, along the XY plane, of up to 11%. The top weld bead of this sample was ground in the XY plane to just below the top surface of the weld, polished, and etched to reveal any surface details. The exact depth in the Z direction at which these measurements were made in the XY plane was not determined. The cyclic variations in the weld region width at the surface of the XY plane may be greater than those quoted above.

It was noted that this cyclic variation in the top weld bead width occurred along the 304 SS side of the weld region. These cyclic variations have been attributed to fluctuations in the EB gun power-supply circuit. The power-supply unit failed and was replaced within six months after making the

welds for this study. No verifications of system stability were performed at AEDC prior to or after the replacement of the power supply.

The general downward trend in bottom weld bead width versus electron beam rotation frequency is attributed to the decrease in power density at the top surface while welding with a rotating electron beam, electron beam divergence, and collisions between the electrons and metallic gases within the keyhole. The electron beam was focussed at the top surface, the XY plane, of the samples. Linear regression was used to fit a straight line from the data for bottom weld bead width versus electron beam rotation frequency. The resulting slope was $-3.88 \times 10^{-3} \text{ mm/Hz}$ indicating a decrease in bottom weld bead width with an increase in the electron beam rotation frequency.

As the electrons impinge on the surface they give up most of their energy in the initial collisions. The surface material is vaporized and the continued flux of electrons causes additional vaporization of material creating more metallic vapors. As the keyhole fills with metallic vapors, the atoms within the vapor obstruct and deflect the flux of electrons. Since the electron beam periphery has been diffused by the rotation process and by beam divergence, the electron beam takes on a near-Gaussian distribution. Only the inner core of the beam has the power density sufficient to vaporize a path through the entire weld [IV-5].

Ultimately, the divergence of the electron beam and the distribution of the metallic vapors within the keyhole region will control the amount of energy

that reaches the bottom of the weld region and dictate the bottom bead width. The distribution of metallic vapors within the keyhole is influenced by the electron beam rotation, the fluid flow in and around the keyhole, the joint configuration, and the distribution of metals within the weld region. The extent of influence any one variable has on the electron collision process is beyond the scope of this research effort.

E. Weld Cracking

The welds made at all electron beam rotation frequencies used for this study contained cracks within the weld region, HAZ, or both. The relative number and the length of the individual cracks changed with electron beam rotation frequency. Cracking in the baseline sample was limited to a single 0.01 *mm* crack in the 304 SS HAZ.

Cracks were defined as piece-wise continuous openings, with a length to width ratio much greater than one, in the weld region or adjacent to the weld region and that contained no fill material. This excluded pores and regions uniformly filled with material. An example of a region in the HAZ that was uniformly filled with Cu rich material and therefore was not considered a crack is seen in Figure IV-20. Figure IV-21, sample 4-3F shows an example of cracks within the weld region and the HAZ. The presence of Cu was observed on both sides of most of the cracks within the HAZ while Cu was present in all of the cracks within the weld region. Within the HAZ the cracks were generally intergranular.

As the electron beam rotation increased the number of cracks within the weld region decreased. The maximum crack length was 0.5 *mm* with most crack lengths approximately 0.2 *mm* or less. Given that the widest weld bead examined was only 2.84 *mm* at the top surface in the XZ plane, the lengths of these cracks approached 20% of the weld region width at the top and were almost half the width of the weld region in the middle and bottom of the weld region.

E.1. LME and HAZ cracking

Within the HAZ, cracking occurred prior to the complete solidification of the weld region and after the solidification of the weld region. If the cracking occurred prior to the complete solidification of the weld region then the opportunity existed for Cu rich material, that melts at a temperature at least 317°C lower than the lowest temperature of the 304 SS melting range, to back flow into the crack. The samples that had cracks within the HAZ that penetrated into the weld region yet contained no evidence of Cu rich material within the HAZ portion of the crack most likely occurred after the weld region had solidified.

Only one case was observed where Fe rich material penetrated into the Cu base material. This is shown in Figure IV-22. The flow path of the Fe rich material appears to have melted back the Cu base-material smoothing out the Cu grain boundaries prior to solidifying.

LME has been defined as the reduction of fracture stress and the loss of ductility caused by liquid metal in contact with the surface of a solid metal in the presence of tensile stress. Nippes has indicated that cracking due to liquid metal embrittlement does not occur in situations where the sample is under compression [IV-6].

By definition LME is not limited to the HAZ yet this region was where most of the cracks attributed to LME occurred in the samples examined for this study. In several samples, cracks did extended from the weld region into the HAZ. Figure IV-21, sample 4-3F welded with the electron beam frequency set at 40 *Hz*, shows the severe cracking that occurred near the middle (in the Z direction) of the sample in the YZ plane of the sampled welded at 40 *Hz*. The weld region and HAZ obviously experienced tensile strains in excess of the yield strength of both the Cu rich interdendritic material and 304 SS base material during solidification.

The traditional view of LME takes the approach that liquid metal would have to infiltrate the GB in the HAZ while the weld region is experiencing thermally induced tensile stress. The mechanism for transport of the liquid metal along the grain boundaries has been attributed to capillarity by Mushala [IV-7].

Clear evidence of LME, with a back filled crack, is supported by Figure IV-20, a micrograph of sample 5-6S that shows a vein of Cu rich material extending into the GB of the HAZ along the 304 SS side of the weld. It is

reasonable to assume that prior to welding there were no cracks in the 304 SS. The GB within the HAZ, at some point in time, was opened sufficiently to allow infiltration of Cu rich material.

The source of material for this vein in Figure IV-20 appears to be the interdendritic material present in the partially melted two-phase region at the edge of the weld region. The possibility also exists that the pocket of material that appears to supply the vein may have been swept along with the rotating flow. The original source for this material being the Cu on the opposite side of the weld region that prior to solidification become trapped between dendrites as the Fe rich phase solidified, as opposed to interdendritic materials that are rejected at the solid-liquid interface and trapped in the solidification process. Regardless of the source of Cu rich material, the GB within the HAZ was opened sufficiently by the residual tensile stresses permitting infiltration by Cu.

The tensile stresses must be capable of producing a critical level of plastic deformation in the presence of the embrittling atoms in order for cracking due to liquid metal embrittlement to occur [IV-8]. The ratio, R , of the maximum cleavage stress ($\sigma_{\max}$) needed to separate two atoms to the maximum shear stress ($\tau_{\max}$) has been given as a guide for determining the critical level of plastic deformation needed for LME to occur by Nippes [IV-9]. According to Nippes

$$R = \frac{\sigma_{\max}}{\tau_{\max}} \leq 5. \quad (\text{IV-3})$$

The fact that not all of the Cu rich veins developed open cracks during solidification is attributed to stress relief within the sample from the cracks that did open. As the residual tensile stress builds beyond the yield point of the Cu rich vein (assuming it has solidified) the Cu rich material fails and the crack opens reducing the overall stress level. Cracking will also occur when liquid Cu infiltrates a GB that is in tension. The liquid Cu cannot resist any tensile stresses.

In the reports by Mushala and Nippes a listing of characteristics for what they call Copper-Contamination Cracking (CCC) is given. They are listed below for convenience. Nippes defines CCC as hot cracking remote from the fusion boundary line resulting from surface contamination of the base metal by Cu [IV-10]. It is important to remember that in the report by Nippes, the Cu was placed on the surface of the materials to be tested (a Cu contamination layer), the samples were heated, and then the samples were placed in tension. At least six of these conditions were present in the EB welded samples for this study. Conditions believed present in this study are preceded with an asterisk. Due to the different objectives of this study, it isn't known if other conditions were met.

The required conditions are [IV-11]:

1. *The hot cracking is restricted to portions of the weld heat-affected zone which experiences temperatures above the melting point of the embrittling Cu-contamination layer;

2. *Hot cracking can occur if the welding process provides adequate shielding protection of the weld heat affected zone to prevent oxidation of the Cu on the surface of the weldment;
3. The susceptibility to the Cu-induced hot cracking decreases with increased temperature above some transition temperature in the heat-affected zone;
4. *The crack morphologies encountered in the weld heat-affected zone of FCC metals are intergranular and are oriented perpendicular to the principal-stress direction;
5. *The cracks propagate at high velocities once a critical amount of plastic strain is introduced into the weld heat-affected zone;
6. The length of CCC cracks is proportional to the volume of liquid metal present on the base metal;
7. The yield stress and strain-hardening rate of the base metal are unaffected by CCC prior to fracture;
8. *The susceptibility to CCC increases with increasing strain rate;
9. The fracture stress of FCC materials, although normally not dependent on grain size, is grain-size dependent when grain boundaries are wet by certain metals;
10. *In materials which are not notch-sensitive, the liquid must be present at the crack tip for the continued propagation of copper-contamination-cracks.

During the heating cycle of the EB welding process, the weld pool is molten and the HAZ is free to expand. The molten material within the weld pool is also free to expand and flow out of the confines of the surrounding cool walls transferring energy from the molten pool into the HAZ in the process. The HAZ expands due to the increase in temperature yet is not physically restrained by the molten liquid. Any restraint would have to come from the base material surrounding the HAZ.

This form of restraint could lead to a state of compression within the HAZ as high temperature materials try to expand against the cooler surrounding materials. Away from the molten liquid, if the thermally induced compressive stresses are sufficient then plastic deformation, or upset, of the high temperature materials, that accommodates the expansion, will occur. The yield strength of metals is normally reduced at elevated temperatures.

Once the molten weld pool begins to solidify, implying that cooling and shrinking of the weld metal has begun, the joint is locally restrained permitting residual tensile stresses to develop that exceed the UTS of the material resulting in the formation of cracks.

During cooling, the possibility also exists that the surface of the weld region has solidified and cooled to a temperature lower than that of the material within the core of the weld region. While in the vacuum chamber this requires that the transfer of thermal energy, away from the surface due to

radiational cooling, exceeds the transfer of energy within the weld region due to thermal conduction.

As the inner portion of the weld region is cooled, residual tensile stresses develop between the top surface and the inner portion of the weld region. If these stresses exceed the UTS of the HAZ cracking will occur. The UTS of 304 SS drops with increases in temperature and the material within the HAZ adjacent to the weld region was heated to a temperature very close to that of the weld region making it susceptible to post solidification cracking.

With the exception of the samples welded at 20 Hz, the principal tensile stresses acted in the Z direction of the YZ plane; cracks within the HAZ were parallel to the Y direction. Figure IV-23, sample 5-10F welded at 180 Hz, shows a crack in the Y direction of the YZ plane that begins at the edge of the weld region and extends into the 304 SS HAZ. The HAZ was obviously still weak just as tensile stresses began to develop and act upon it. No evidence of Cu was found in this crack indicating that this crack developed after the weld region had solidified.

The intergranular crack seen above the large crack in Figure IV-23 does contain Cu rich material indicating that this HAZ crack occurred prior to the solidification of the weld region, molten material was available to back fill the crack.

Given the presence of the low melting point liquid Cu rich phase adjacent to the GB within the HAZ during the EB welding process and the

residual tensile stresses acting upon the HAZ it is entirely possible that a rapid or near instantaneous failure of the GB ensued [IV-11]. The opening of the GB would have provided local stress relief leaving the open GB available to be back filled with the liquid Cu rich phase. If it is assumed that there are no longer any residual tensile stresses acting on this section of the GB the crack will not propagate even in the presence of the liquid Cu rich material. The Cu rich material that has infiltrated the GB will solidify as the weld region and HAZ cool to below the melting point of the Cu rich material. The melting point of pure Cu (1083°C) is 317°C lower than the minimum melting range temperature (1400°C) of the 304 SS. The effects of the alloying of the Cu within the weld region will result in a maximum melting range temperature that will be less than 1083°C and will be a function of composition.

Cracking due to LME outside of the weld region in the HAZ was not observed in those samples welded with the electron beam rotation frequency below 100 *Hz*. Cracking in the HAZ was observed to increase with beam rotation frequency above 80 *Hz*. Multiple cracks 0.5 *mm* in length and 0.01 *mm* in width were observed in the HAZ on the 304 SS side of sample 3-7S welded with the electron beam rotation frequency set at 120 *Hz*.

For those samples welded with the electron beam rotation frequency less than 80 *Hz* a fine filigree pattern of Cu rich material extending approximately 1 μm into the GB of the HAZ along the 304 SS side of the weld region was observed. Some caution is advised when considering this

occurrence. The possibility of smearing Cu into the 304 SS GB region of the HAZ during grinding and polishing exists. Additional investigation into this area is needed.

E.2 Weld region cracking

Only the sample welded with the electron beam rotation frequency set at 20 *Hz*, sample 4-2S, was found to have substantial weld region cracks perpendicular to the XY plane (the Z direction). These cracks were approximately 0.5 *mm* in length and 1 μm in width and located less than 0.015 *mm* from the 304 SS weld region wall within the weld region. The principal tensile stresses in this sample were aligned perpendicular to the XZ plane indicating that upon cooling the energy was being transferred parallel to the top and bottom surfaces into the sample itself. The HAZ walls cooled faster than the top and bottom surfaces of the weld coupon resulting in residual tensile stresses developing between the walls of the HAZ and the weld region center. The sample was acting as its own heat sink. No suitable explanation has been found for this obvious exception to the cooling patterns exhibited by the other samples. Despite the extensive cracking perpendicular to the tensile load direction during tensile testing this sample failed in the region outside the weld region.

The samples welded with electron beam rotation frequencies of 40 *Hz*, 60 *Hz*, and 80 *Hz* had cracks that were parallel to the XY plane indicating that upon cooling the residual stresses that developed were perpendicular to the

XY plane (the Z direction). This was a result of substantial energy transfer out the top and bottom surfaces of the weld region due to radiational cooling, early in the solidification process. Within the vacuum chamber radiational cooling is the only mechanism for energy transfer away from the surface. Thermal transfer by conduction within the weldment is still required to get the energy to the surface.

The residual tensile stress due to shrinkage of the surfaces of the XY plane, upon initial cooling was opposed by compressive stresses due to thermal expansion within the hot weld region. As the weld region material cooled additional residual tensile stresses developed within the weld region pulling on the already solidified surface. Ultimately the residual tensile stresses exceed the yield strength of the weld region material resulting in failure. Obviously the radiational heat transfer exceeded the energy transfer due to conduction within the samples.

The extensive three dimensional network of Cu rich interdendritic material observed in the weld regions provides a weak link when the weld regions are subjected to tensile stresses during or after solidification. This weakness would be accentuated at temperatures close to or exceeding the melting range of the Cu rich interdendritic material. Pure Cu melts at 1084°C while the interdendritic material will melt at some lower temperature depending upon the composition. The melting range of the 304 SS is from 1400°C to 1450°C.

When the electron beam rotation frequency is increased, temperature gradients within the weld region decrease due to the continuous addition of thermal energy from the rotating flow into the solidifying region surrounding the keyhole. The local temperature of the solidifying region will be influenced by the temperature of the rotating pool of molten liquid that does not vary substantially with time. The continued transfer of energy by convection in the liquid over small distances will act to maintain the solidifying region at a uniform temperature over a longer period of time as opposed to the situation without electron beam rotation thus reducing the cooling rates. The amount of energy transferred due to convection will be a function of the fluid velocity that is driven by the electron beam rotation frequency. No measurements of the actual fluid rotation frequency or velocity were made.

The effect of diffusing the electron beam, a consequence of using electron beam rotation, will simply be to decrease the power density at the weld surface. However, this effect is independent of the actual rotation frequency; no changes in power density between the rotation frequencies of 20 Hz and 180 Hz will be evident since both the electron beam diameter and power are fixed.

The degree or extent of cracking within the solidified weld region will be a strong function of the cooling rate. The residual stress that develops in the solidified weld region will be reduced with a reduction in the cooling rate. As the weld region solidifies, thermal induced residual stresses develop that can

exceed the yield strength of the still hot solidified material in the weld region. Cracks that develop within the slowly cooling weld region can be back filled with molten liquid if that material is available and there is sufficient time for the process to occur. The slow cooling rates allow greater contact time with the molten metal. A negative aspect of the slow cooling rates is related to LME. The slow cooling rates allow the GB greater contact time with liquid metal. This can lead to LME described previously.

Examination of the samples showed that cracking within the weld region was eliminated as the electron beam rotation frequency was increased above 80 *Hz*. This has been attributed to the slower cooling rates at the higher beam rotation frequencies and the extended availability of liquid material, both Cu rich and Fe rich, made possible by the fluid rotation, to backfill any cracks that develop prior to complete solidification.

F. Solidification Structures in the Weld Region

The dominant microstructure observed within the weld region was a dendritic form characterized by a composition of 58 *wt%* Fe, 18 *wt%* Cu, 9 *wt%* Ni, and 15 *wt%* Cr with substantial amounts of interdendritic material present. The interdendritic material was a Cu rich mixture; its constituency was not analyzed. A typical distribution of interdendritic material can be observed in Figure IV-3, a micrograph of sample 4-3F.

The Cu-Fe equilibrium phase diagram shown in Figure II-3 does not indicate the presence of a monotectic region yet a significant number of

structures associated with monotectic alloys were identified. Divorced spherical structures, seen in Figure IV-24, sample 3-5F, and closely packed fibrous composite structures as shown in Figure IV-25, a darkfield micrograph of sample 1-4F, were found adjacent to the 304 SS wall of the solidified weld region. This form is consistent with descriptions of monotectic solidification structures provided by Elliot [IV-12]. The liquid Cu rich phase that solidified last engulfed the previously solidified α solid phase.

The divorced spherical structure microstructures can be compared with those of the known Cu-Pb monotectic alloy. The micrograph of a Cu-35 wt% Pb alloy, seen in Figure IV-26, has a very similar structure to that shown in Figure IV-24.

The Cu-Fe phase diagram generated by Nakagawa [IV-13] does show a region of limited immiscibility at 50 atomic percent Cu yet the phase diagram published in 1992 by Swartzendruber, Figure II-3, has not delineated that region. For this study however, the Cu-Fe phase diagram commonly published and given in Figure II-3 should only be used as guide. The weld region mixture that was formed in this study contains at least two other elements, Cr and Ni, that obviously will influence the solidification process. In addition, no check for impurities was made. Both Nakagawa [IV-14] and West [IV-15] recognized the influence impurities have on the Cu-Fe system.

The Fe rich region shown in Figure IV-27 includes a halo like Fe rich region surrounding what appears to be a mixture of Cu rich phases and Fe

rich phases. The periphery of the halo consisted of a region of planar front growth extending from a dendritic region. Extended cellular like growths that originated at the planar surface can be seen growing into the center. The center region contained solidification morphologies similar to the bulk solidified weld region but at a much smaller scale.

Since both planar solidification growth fronts and cells grow opposite to the direction of the heat flow, the circular planar growth front that appears to have developed required radial heat flow out of the spherical liquid mass into the surrounding dendrite field. This is the same growth pattern as observed for undercooled equiaxed dendrites.

Etching another similar sample, sample 3-5S, welded at 80 Hz, with FeCl_3 and water (10g FeCl_3 per 90cm³ H₂O) removed a substantial portion of the Cu rich surface contained within the halo region. Examination using a scanning electron microscope (SEM) showed the same clean periphery.

Some of the shading effects can be attributed to the polishing process. Since the Cu rich material was substantially softer than the Fe rich material, the inner Cu rich region wore at a faster rate. This allowed the abrasives to slide down to the lower Cu rich level and then slide back up to the Fe rich level during polishing slightly rounding the edge. This slight rounding of the periphery was not significant and did not affect the results. Chemical analysis using the SEM showed that the average composition of the periphery matched

that of the dominant dendritic composition of the solidified bulk weld region. The cellular growths were clearly originating within the clean halo like region.

The halo like region of the spherical microstructures discussed above contained plane front growth that evolved into cellular growth over a very short distance. The thickness of the planar growth region that forms the halo like periphery of the spherical region shown in Figure IV-28 was on the order of $0.7\ \mu m$. The cellular microstructures contained within the inner region were approximately $1\ \mu m$ in diameter and reached lengths of approximately $3.5\ \mu m$.

To verify the possibility of planar front growth on the periphery of the spherical structure as the cause of the halo like region, Chalmers' CST criterion was applied. This required several assumptions since critical information, cooling rates, composition, compositional gradients, temperatures, temperature gradients, and thermal properties, was not available.

The approximate diameter of the spherical central mass in Figure IV-27 is $32\ \mu m$. At the onset of planar front solidification the temperature within the spherical region was assumed to be uniform at equal distances from the center, radial isotherms, and varied according to an exponential function of radial distance from the periphery. The temperature at the periphery was assumed to be $1095^{\circ}C$ while the temperature in the center was $1460^{\circ}C$. This temperature difference approximates the freezing range for a Cu-47 wt% Fe

alloy. This matches the baseline composition for the assumed equal volume weld.

The thermal properties of the actual alloy formed during welding are not available. Values of thermal diffusivity for most alloys in the liquid phase are not available and those that are available are subject to experimental errors. The thermal diffusivity of the liquid contained within the weld region was assumed to be $6 \times 10^{-5} \text{ cm}^2/\text{s}$. The value used was a compromise between values given by Iida and Guthrie [IV-16]. Table IV-2 contains relevant thermophysical data for Cu and 304 SS.

A quick iteration showed that with $T_0 = 1095^\circ\text{C}$ at $x=0$, the outer edge of the spheroid,

$$T(x) = T_0 e^{(18000x)} \quad (\text{IV-4})$$

provided a reasonable approximation of the initial boundary conditions. At $x=0$, $T=1095^\circ\text{C}$ and at $x=16 \text{ } \mu\text{m}$, the center of the spheroid, $T=1460^\circ\text{C}$.

The temperature gradient of the liquid ahead of the S/L interface was calculated using the first derivative of $T(x)$ evaluated at $x=0$; this is referred to as G . The reader is reminded that the actual temperature of the liquid ahead of the S/L interface is used, and not the equilibrium liquidus temperature, to determine G . The actual boundary layer thickness was not needed for this calculation. A reasonable boundary layer thickness, δ , used in zone refining with Fe as the solvent would be $\delta < 0.015 \text{ cm}$ [IV-17]. This would be up to nine

times the thickness of the region being considered and probably not applicable given the differences in the process. Evaluation of

$$G = \frac{dT}{dx}, \text{ at } x=0, \quad (\text{IV-5})$$

gives

$$G = \frac{dT}{dx} = 18000T_0 e^{(18000x)} \Big|_{x=0} = 1.97 \times 10^5 \frac{^\circ\text{C}}{\text{cm}}. \quad (\text{IV-6})$$

The CST criterion,

$$R \leq \frac{GD_L}{\Delta T}, \quad (\text{IV-7})$$

is used to determine the solidification rate. This requires that

$$R \leq \frac{\left(1.97 \times 10^5 \frac{^\circ\text{C}}{\text{cm}}\right) \left(6 \times 10^{-5} \frac{\text{cm}^2}{\text{s}}\right)}{365^\circ\text{C}} = 0.032 \frac{\text{cm}}{\text{s}} \quad (\text{IV-8})$$

to maintain a planar S/L interface for the given conditions. This is about 3% of the welding speed used. This appears plausible based upon the maximum growth rate allowable relative to the weld velocity and a maximum growth rate quoted by Kuo for a substantially different set of conditions [IV-18].

Figure IV-28 and Figure IV-29 clearly show the transition from planar front growth to cellular growth. The transition from planar front to cellular growth requires a combination of supercooling and perturbation to initiate growth. The perturbation could be in the form of rapidly changing thermal gradients, local concentration changes, or surface tension effects within the divorced spherical mass.

No porosities within these divorced spherical regions were observed. If present, the porosities could influence surface tension and induce instabilities.

The divorced spherical regions did contain both Cu rich and Fe rich phases. The existence of two distinct and immiscible phases within the divorced spherical region would influence surface tension within the regions.

Thermal fluctuations and surface tension most likely were the driving forces for perturbation. The fluid within the divorced spherical region was trapped in place by the outer region that had already solidified. Any thermal disturbance in the solidifying weld region could change the surface tension between the interfaces altering local composition.

With a change in composition supercooling could increase or decrease. If a perturbation were to grow into a region of liquid that was not supercooled it would melt back. Perturbations that extend into a region that is supercooled will survive and grow at a rate that is a function of the degree of supercooling.

It can be seen in Figure IV-29 that the direction of the cellular growth for several cells approaches being parallel. There is also a difference in length as a function of position on the periphery. The cells on the right hand side are substantially shorter in length with some having barely grown beyond a perturbed surface. This is consistent with cellular growth opposite to the direction of heat flux. Perturbation spacing was estimated to be $2.5 \mu m$. Wide variations in spacing were also noted but not quantified.

The rate at which heat was removed obviously affected the microstructure since adjacent to the Cu wall of the weld region the divorced spherical structures did not form. The coefficient of thermal conductivity in

solid 304 SS is approximately $15 \text{ W/m}^{\circ}\text{C}$, quite low when compared with that for OFHC Cu, $400 \text{ W/m}^{\circ}\text{C}$.

Within the confines of the spherical region several other microstructures were found. What appears to be a fragmented dendrite can be seen at a distinct interface or grain boundary in Figure IV-30. While the secondary arms appear to be uniformly spaced, of uniform diameter, and at right angles (the [100] family of directions) there is considerable variation in composition at the tips. Points 7,6, and 4 are all rich in Cu, approximately 84 wt% Cu yet points 1,3, and 5 have only 20 wt% Cu. Point 2 does not appear to be part of this dendrite.

The distance between the tips at right angles to each other is approximately $1.5 \mu\text{m}$. In order for this dendrite to grow with the differing compositions noted the composition of the interdendritic fluid compositions would need to be radically different over a very short distance. Mixing effects would be negligible once the periphery was formed thus two bodies of fluid of different composition would have to exist in close proximity to this dendrite. This implies fluid immiscibility at the microstructure level.

CHAPTER V

CONCLUSIONS

The physical analysis performed has shown that weld region uniformity, determined by mapping the Knoop microhardness within the weld region of the Cu-304 SS couple, was a function of the electron beam rotation frequency. As the electron beam rotation frequency was increased so did the uniformity of the microhardness distribution within the weld region. The microhardness level of the weld region was not a consideration during the development phase of the intrusive probes used at AEDC yet this study has shown that microhardness can be an effective indicator of weld region homogeneity, an important factor in high pressure probe applications.

Tests that include only the actual weld region material must be devised in order to determine the true effect of electron beam rotation on the weld tensile strength. The tensile strength of the welds exceeded that of the OFHC Cu for all electron beam rotation frequencies employed during the present study.

A determination of a specific relationship between top and bottom weld-bead width and electron beam rotation frequency for the conditions used in this study could not be made, it appears that the welds were performed with

excess electron-beam power. The top weld bead width and bottom weld bead width decreased only slightly as the electron beam rotation frequency was increased from 20 *Hz* to 180 *Hz*.

Cracking within the weld region and in the HAZ adjacent to the weld region was shown to be dependent on the electron beam rotation frequency. Welding with the electron beam rotation-frequency below 100 *Hz* resulted in the formation of cracks within the weld region while cracking in the HAZ attributed to LME was observed in those samples welded with electron beam rotation frequencies above 80 *Hz*. Cracking was most severe in those samples welded with electron beam rotation frequencies of 20 *Hz* and 40 *Hz*. Since cracks were detected in all the samples produced for this study, including the baseline sample, design trade-offs, based on desired results, should be included in any future decision to use these weld couples in critical applications.

The results of this study have demonstrated that electron beam rotation during EB welding of Cu-304 SS couples does have both positive and negative effects upon the resulting weld. It has also been recognized that the conditions used for this study were not necessarily the optimum. The conditions used did provide a range of results that can be used to further refine the EB welding processes used at AEDC. Additional studies of the effects of electron beam rotation on EB welded Cu-304 SS couples, that

extend the electron beam rotation frequency range beyond the present range of 180 *Hz* and reduce the electron beam rotation frequency increment to 10 *Hz* or less are justified.

Optimization of the EB welding process, for future applications of the Cu-304 SS couples, will require that the desired weld region properties be clearly defined and prioritized with respect to their importance in actual use.

A detailed microstructural study is needed to provide a more complete understanding of the weld region morphology and the monotectic solidification process.

LIST OF REFERENCES

REFERENCES

- [I-1] Dahotre, N. and Zysk, K., "Electron Beam Welding of 304 Stainless Steel to Copper," Proceedings of the Conference "The Laser and the Electron Beam in Welding, Cutting and Surface Treatment State of the Art 1993," Reno, NV, Nov. 1993, pp. 105-118.
- [I-2] Gopinathan, S., Murthy J., McCay, T. D., and McCay, M. H., "Microstructural Evaluation of Laser Welded Copper-Stainless Steel Welds," Proceedings of ICALEO 1993, Volume 77, Laser Materials Processing, Orlando, FL, Oct 1993, pp. 661 – 671.
- [II-1] Kou, S., *Welding Metallurgy*, John Wiley & Sons, New York, NY, © 1987, p. 3.
- [II-2] Publication 62-200.02, PTR-Precision Technologies, Inc, Enfield, CT.
- [II-3] Cullity, B. D., *Elements of X-Ray Diffraction*, second edition, Addison-Wesley Publishing Company, Inc., Reading, MA, © 1978, p. 21.
- [II-4] Private communications with Don Powers, Technical Director, PTR-Precision Technologies, Inc., Enfield, CT.
- [II-5] Ref. [II-1], p. 109.
- [II-6] Vidts, J., De Cuypere, F., De Say, A., INCRA Research Project No. 122, "The Effect of Copper on the Weldability of Steel, Final Report," The International Copper Research Association, New York, NY, Sept 1968, p. 2.
- [II-7] West, E. G., *Copper and Its Alloys*, Ellis Horwood Limited, Chichester, West Sussex, England, © 1982, p. 94.
- [II-8] Nakagawa, Y., "Liquid Immiscibility in Copper-Iron and Copper-Cobalt Systems in the Supercooled State," *Acta Metalurgica*, Volume 6, Nov 1958, p. 704.
- [II-9] Ref. [II-6], p. 3.

- [II-10] *ASM Metals Handbook, Volume 3*, ASM, Metals Park, OH, © 1992, p. 2-168.
- [II-11] Van Vlack, L. H., *Elements of Material Science and Engineering*, Fifth Edition, Addison-Wesley Publishing Company, Inc., Reading, MA, © May 1985, p. 385.
- [II-12] Ref. [II-10], p. 2-175.
- [II-13] Ref. [II-10], p. 2-168.
- [II-14] Kingery, W. D., Bowen, H.K., Uhlmann, D.R., *Introduction to Ceramics*, Second Edition, John Wiley & Sons, New York, NY, © 1976, pp. 295-297.
- [II-15] Ref [II-1], p. 129.
- [II-16] Ref [II-1], p. 129.
- [II-17] Flemings, M. C., *Solidification Processing*, McGraw-Hill Inc., New York, NY, © 1974, p. 59.
- [II-18] Ref [II-1], p. 134.
- [II-19] Ref [II-1], p. 59.
- [II-20] Ref [II-1], p. 135.
- [II-21] Kurz, W., Fisher, D. J., *Fundamentals of Solidification*, Trans Tech Publications, Switzerland, © 1989, p. 88.
- [II-22] Ref [II-21], p. 66.
- [II-23] Ref [II-21], p. 66.
- [II-24] Ref [II-21], p. 56.
- [II-25] Elliot, R., *Eutectic Solidification Processing, Crystalline and Glassy Alloys*, Butterworths & Co (Publishers) Ltd., London, England, © 1983, p. 298.

- [II-26] Iida, T., and Guthrie, R. I. L., *The Physical Properties of Liquid Metals*, Clarendon Press, Oxford, England, © 1988, p. 70.
- [II-27] Smith, W. F., *Principles of Materials Science and Engineering*, McGraw Hill Publishing Company, second edition, New York, NY, © 1990, p. 447.
- [II-28] Copper Development Association Inc., 260 Madison Avenue, New York, NY, © 1996.
- [II-29] Ref [II-1], p. 109.
- [II-30] Wulpi, D. J., *Understanding How Components Fail*, American Society For Metals, Metals Park, OH, © 1985, p. 67.
- [II-31] Nippes, E. F., Ball, D. J., "Final Report, INCRA Project No. 287B, Copper Contamination Cracking: Cracking Mechanism and Crack Inhibitors," International Copper Research Association, Inc., New York, NY, June 1980, p. 5.
- [II-32] Ref [II-31], p. 22.
- [III-1] Mayer, R., Dietrich, W., and Sundermeyer, D., "New High-Speed Beam Current Control and Deflection Systems Improve Electron Beam Welding Applications," *Welding Journal*, Volume 56, Number 6, June 1977, pp. 35-41.
- [III-2] Ref [II-4]
- [III-3] Ref [I-1], p. 108.
- [IV-1] SCM advertising publication, "A Material For Today, A Technology For Tomorrow, GlidCop®," SCM Metal Products, Cleveland, OH, 1990, p. 2.
- [IV-2] Ref [IV-1], p. 3.
- [IV-3] Ref [II-4]
- [IV-4] Ref [III-1] pp. 35-41.
- [IV-5] Ref [II-4]

- [IV-6] Ref [II-31], p. 20.
- [IV-7] Ref [II-31], p. 10.
- [IV-8] Ref [II-31], p. 22.
- [IV-9] Ref [IV-3], p. 7.
- [IV-10] Ref [II-31], p. 3.
- [IV-11] Ref [IV-3], p. 5.
- [IV-12] Ref [II-25], p. 229.
- [IV-13] Ref [II-8], p. 710.
- [IV-14] Ref [II-8], p. 704.
- [IV-15] Ref [II-7], p. 94.
- [IV-16] Ref [II-26], pp. 213-216.
- [IV-17] Ref [II-17], p. 45.
- [IV-18] Ref [II-1], p. 135.
- [B-1] Ref [II-2]
- [B-2] Ref [II-10], p. 2-168.
- [B-3] Ref [II-10], p. 2-175.
- [B-4] Ref [II-10], p. 3-44.
- [B-5] Ref [II-21], pp. 63, 67, and 70.
- [B-6] Ref [II-17], p. 59.
- [B-7] Ref [II-27], p. 447.
- [B-8] Ref [II-28]

[B-9] *The Welding Handbook*, Volume 1, 7th Edition, American Welding Society, Miami, FL, © 1976, p. 224.

[B-10] Ref [II-28]

APPENDICES

APPENDIX A

TABLE III-1. Chemical Composition of Cu and 304 SS.

	Cr <i>wt%</i>	N <i>wt%</i>	Fe <i>wt%</i>	C <i>wt%</i>	Cu <i>wt%</i>
304 SS	19	9	bal	0.08	
OFHC					99.95

TABLE III-2. Key Machine Parameters.

MACHINE VARIABLE	VALUE
Traverse Velocity	9.7 <i>mm/s</i>
Beam Current	22 <i>mA</i>
Voltage	120 <i>kV</i>
Gun to Work Distance	25.4 <i>mm</i>
Beam Rotation Frequency	0 - 180 <i>Hz</i> in steps of 20 <i>Hz</i>

Table IV-1. Theoretical Molten Fluid Velocity Data

Beam Rotation Frequency (Hz)	Translation Distance per Revolution (mm)	Beam Periphery Rotational Velocity (mm/s)	Minimum Vector Velocity (mm/s)	Maximum Vector Velocity (mm/s)
0	N/A	N/A	N/A	9.7
20	0.49	31.4	-21.7	41.1
40	0.24	62.8	-53.1	72.5
60	0.16	94.3	-84.6	104.0
80	0.12	125.7	-116.0	135.4
100	0.09	157.1	-147.4	166.8
120	0.08	188.5	-178.8	198.2
140	0.07	219.9	-210.2	229.6
160	0.06	251.3	-241.6	261.0
180	0.05	282.7	-273.0	292.4

Table IV-2. Thermophysical Properties of Cu and 304 SS

Property	Units	Cu	304 SS
T_m	$^{\circ}\text{C}$	1083	1400 - 1450
T_b	$^{\circ}\text{C}$	2630	
α	m^2/s	37×10^{-6}	
D_l	cm^2/s		6.5×10^{-5}
k^*	$\text{W}/\text{m}^{\circ}\text{C}$	400	15
ρ_l	kg/m^3	8000	7003

* at 300 $^{\circ}\text{C}$

APPENDIX B

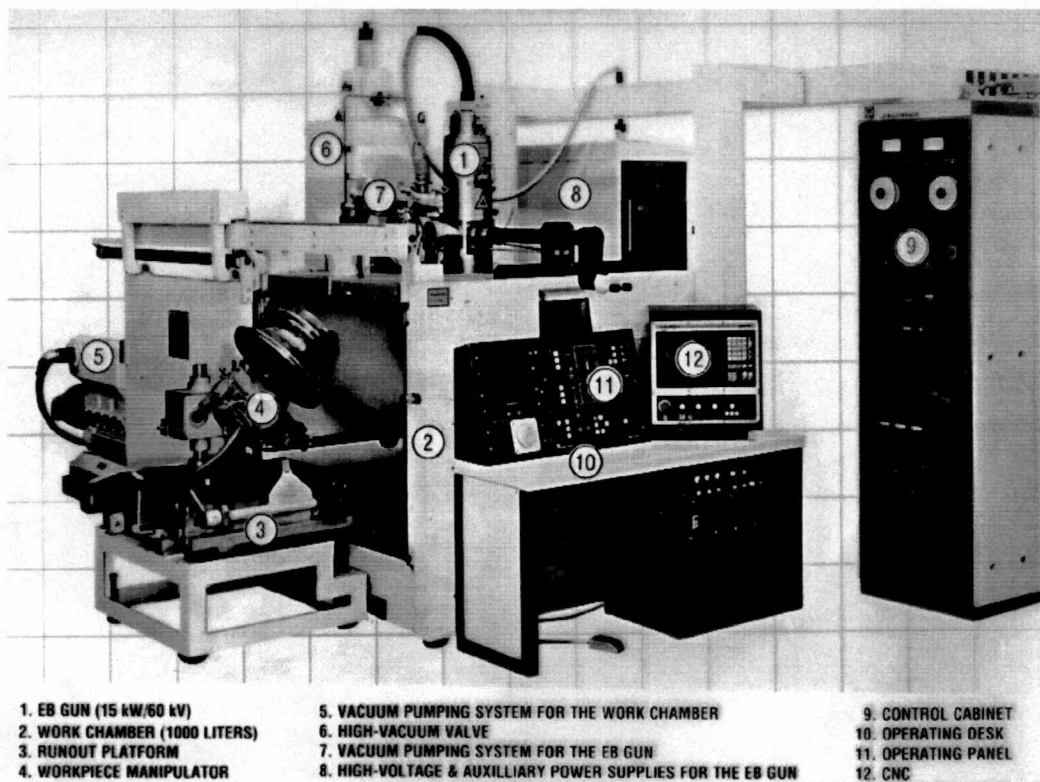


Figure II-1. A typical layout of an EB welding machine [B-1].

Adapted from: Publication 62-200.02, PTR-Precision Technologies, Inc., Enfield, CT. Used with permission.

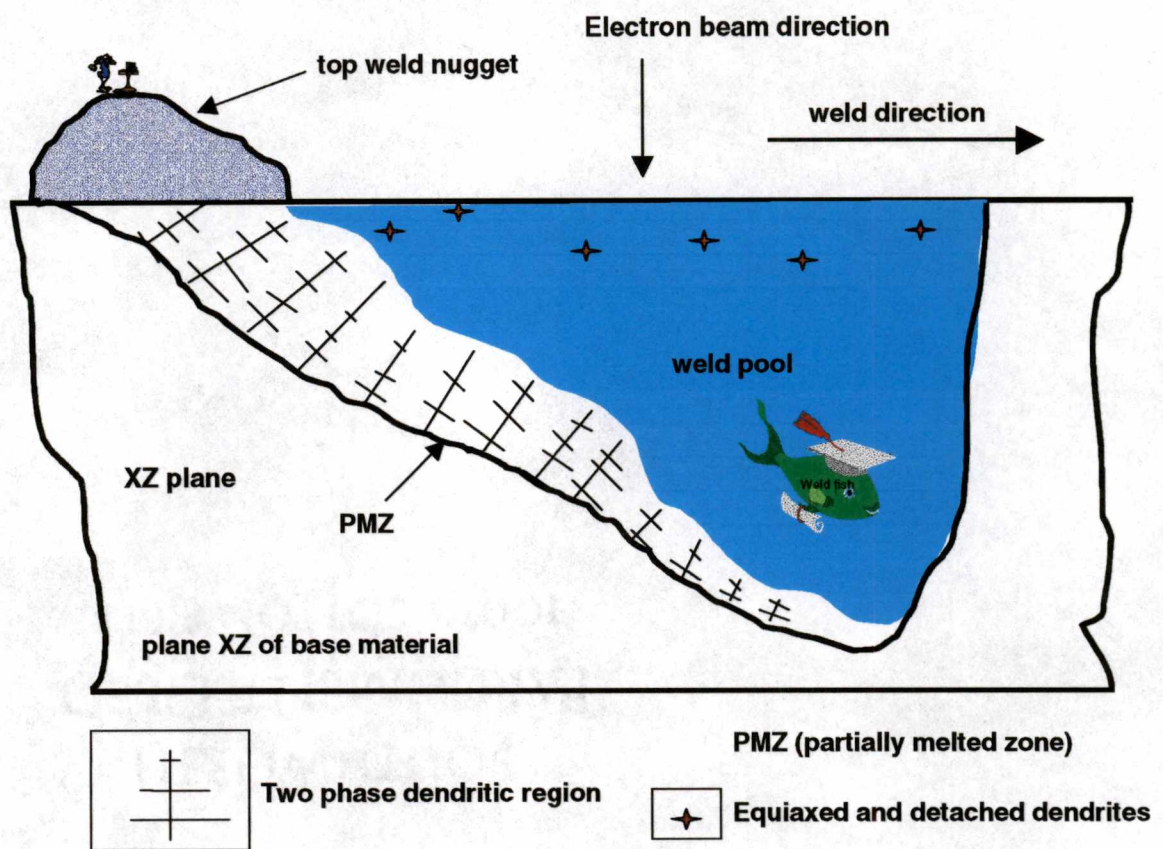


Figure II-2. Schematic of a keyhole weld.

Cu-Fe

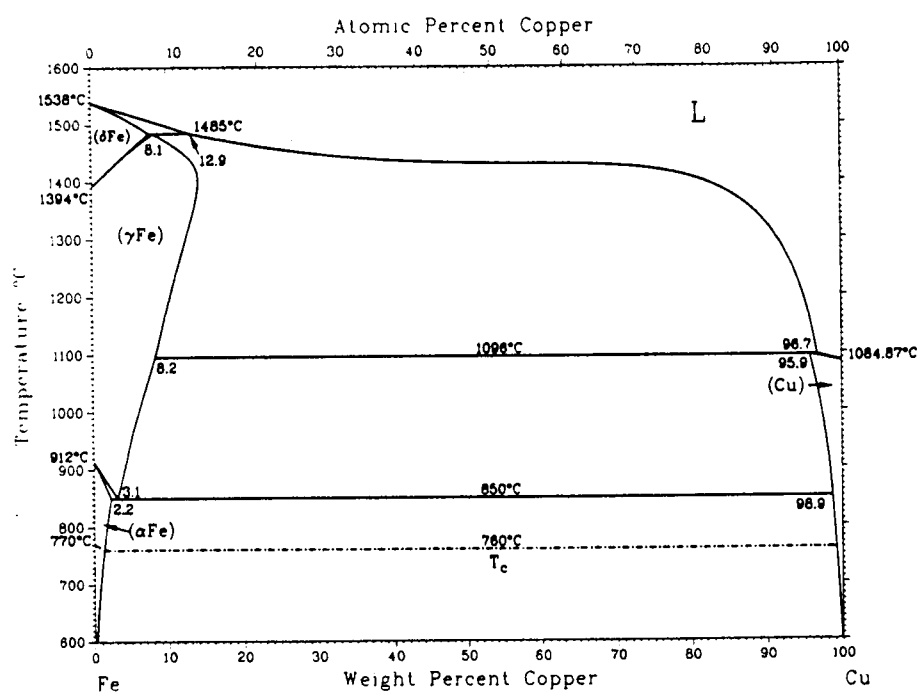


Figure II-3. Cu-Fe phase diagram [B-2].

Source: *ASM Metals Handbook, Volume 3*, ASM, Metals Park, OH, © 1992, p. 2-168. Used with permission.

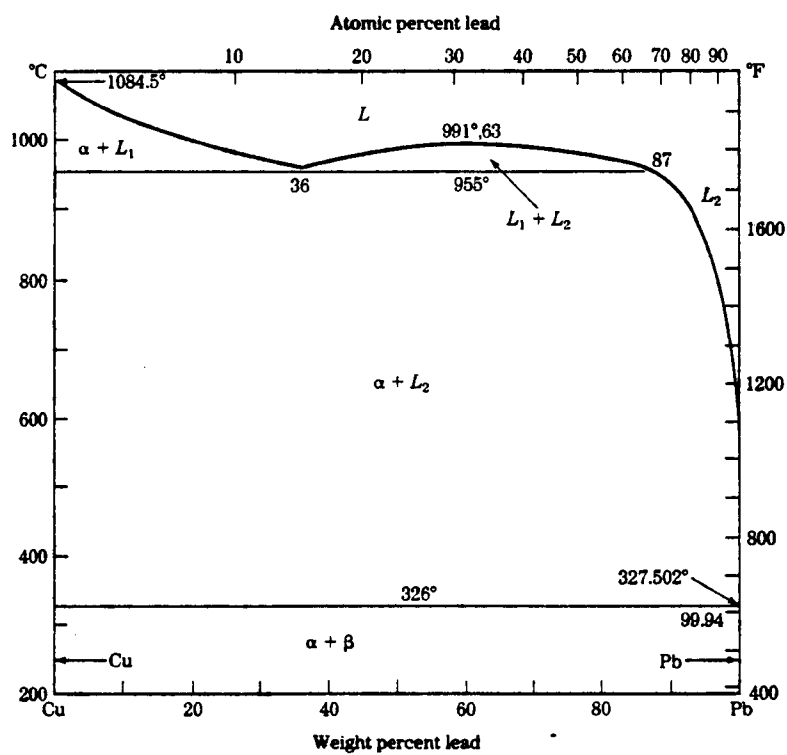


Figure II-4. Cu-Pb phase diagram [B-3].

Source: *ASM Metals Handbook, Volume 3*, ASM, Metals Park, OH, © 1992, p. 2-175. Used with permission.

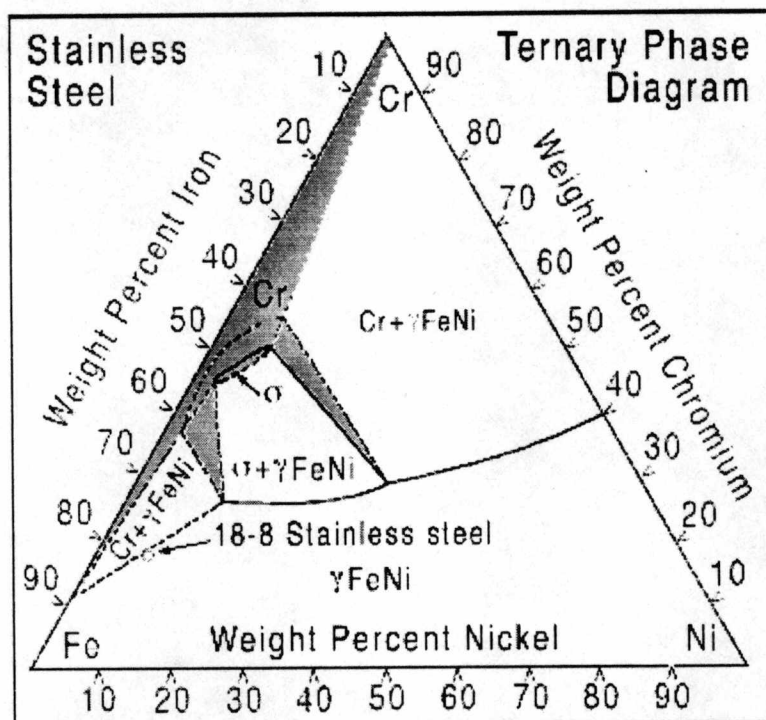


Figure II-5. Isothermal section at 900 °C of the Fe-Ni-Cr ternary phase diagram [B-4].

Source: *ASM Metals Handbook, Volume 3*, ASM, Metals Park, OH, © 1992, p. 3-44. Used with permission.

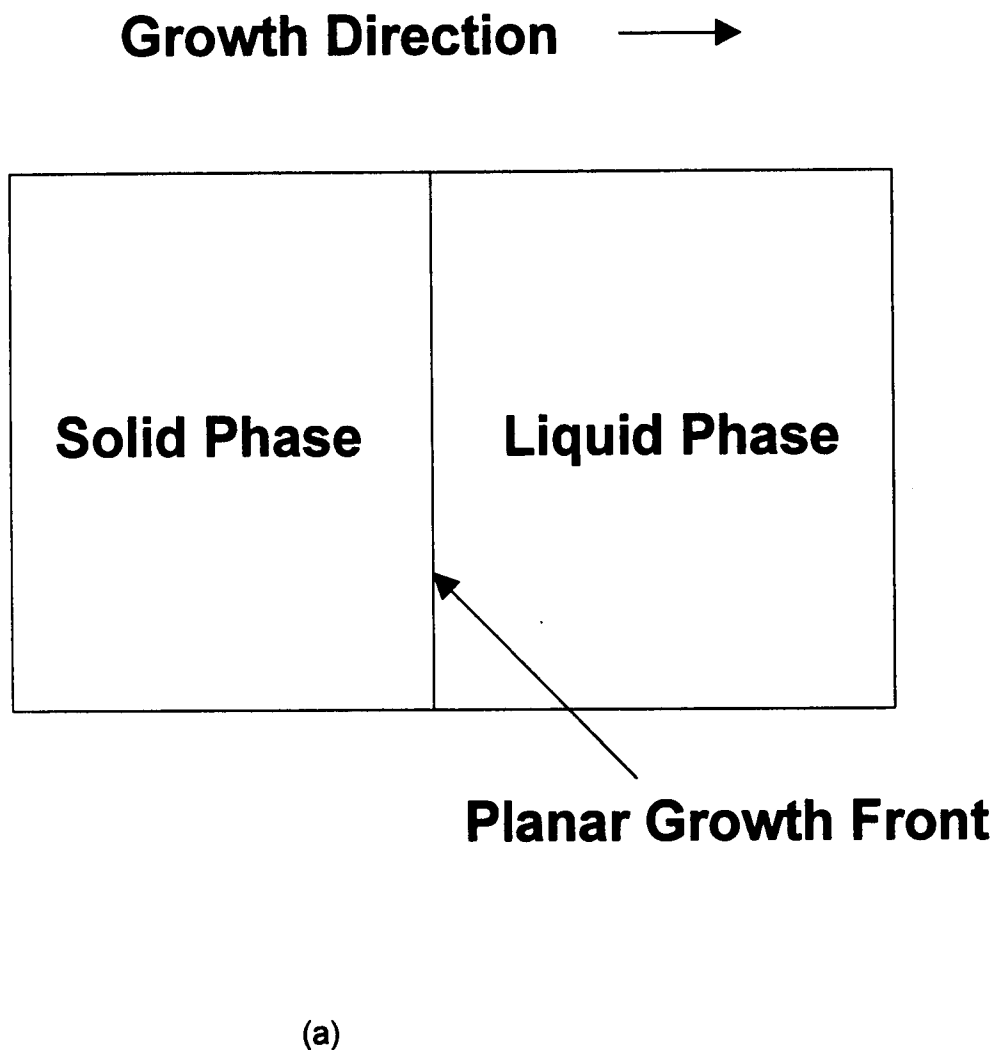
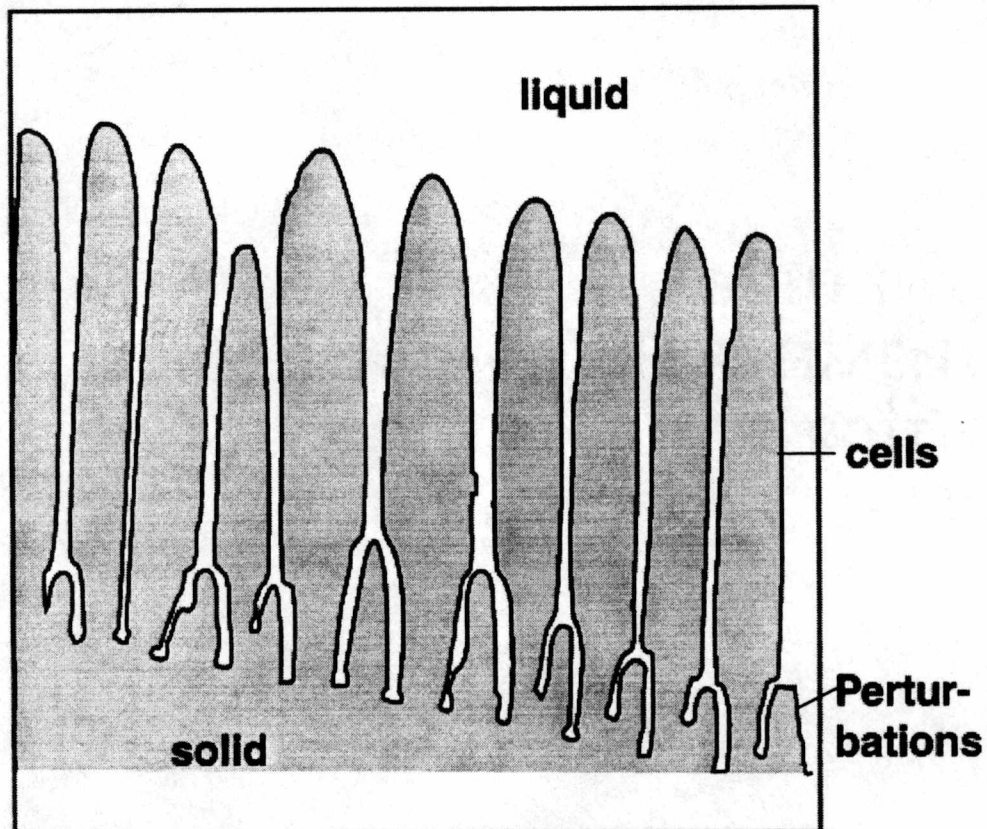


Figure II-6. Artistic interpretations of the basic solidification structures: (a) planar growth, (b) cellular growth, (c) columnar dendrite, and (d) equiaxed dendrite [B-5].

Adapted from: Kurz, W., Fisher, D. J., *Fundamentals of Solidification*, Trans Tech Publications, Switzerland, © 1988, pp. 63, 67, and 70. Used with permission.



(b)

Figure II-6. (continued).

Adapted from: Kurz, W., Fisher, D. J., *Fundamentals of Solidification*, Trans Tech Publications, Switzerland, © 1988, pp. 63, 67, and 70. Used with permission.

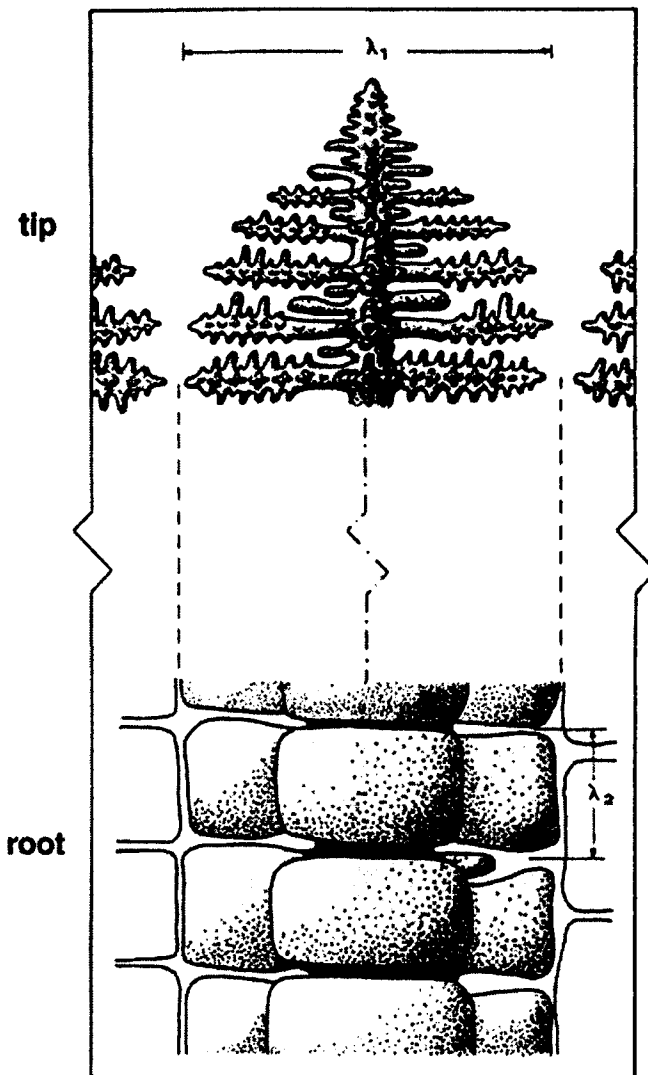
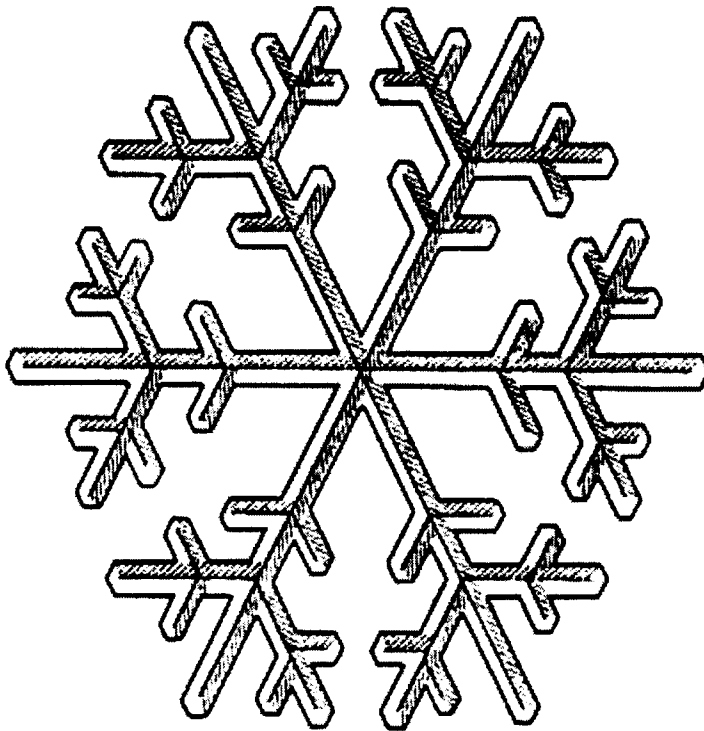


Figure II-6. (continued).

Adapted from: Kurz, W., Fisher, D. J., *Fundamentals of Solidification*, Trans Tech Publications, Switzerland, © 1988, pp. 63, 67, and 70. Used with permission.



(d)

Figure II-6. (continued).

Adapted from: Kurz, W., Fisher, D. J., *Fundamentals of Solidification*, Trans Tech Publications, Switzerland, © 1988, pp. 63, 67, and 70. Used with permission.

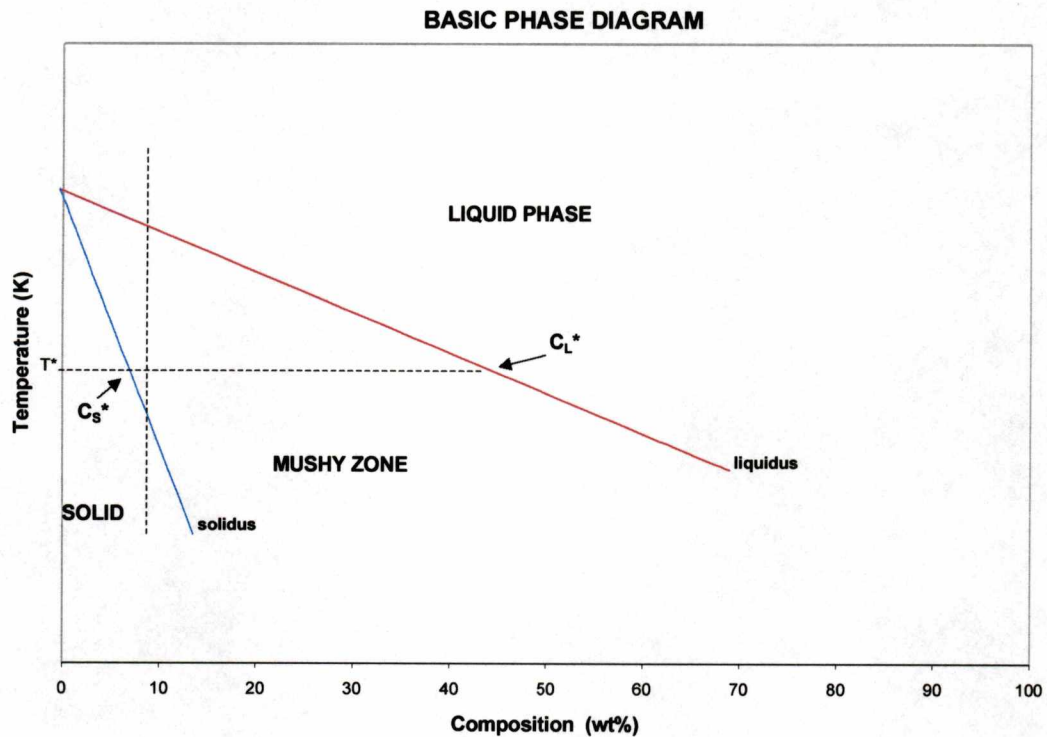
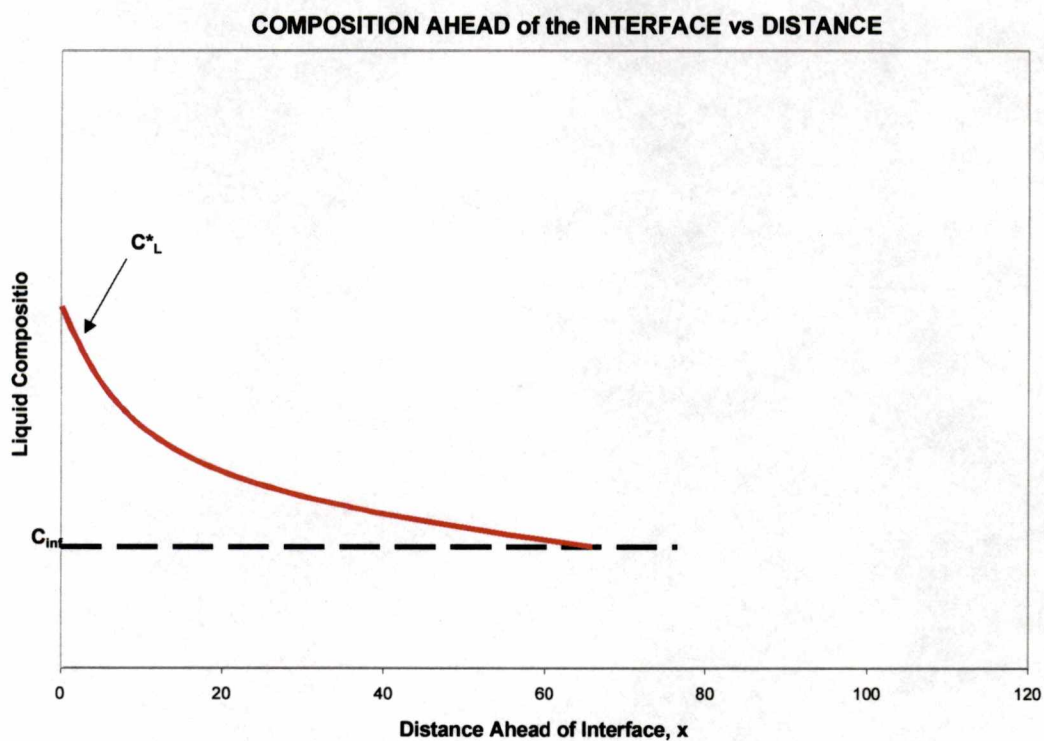


Figure II-7. A graphical representation of the key parameters in the constitutional supercooling process: (a) the phase diagram, (b) the solute enriched layer ahead of the solid-liquid interface, (c) solidification with a stable interface, and (d) solidification with an unstable interface [B-6].

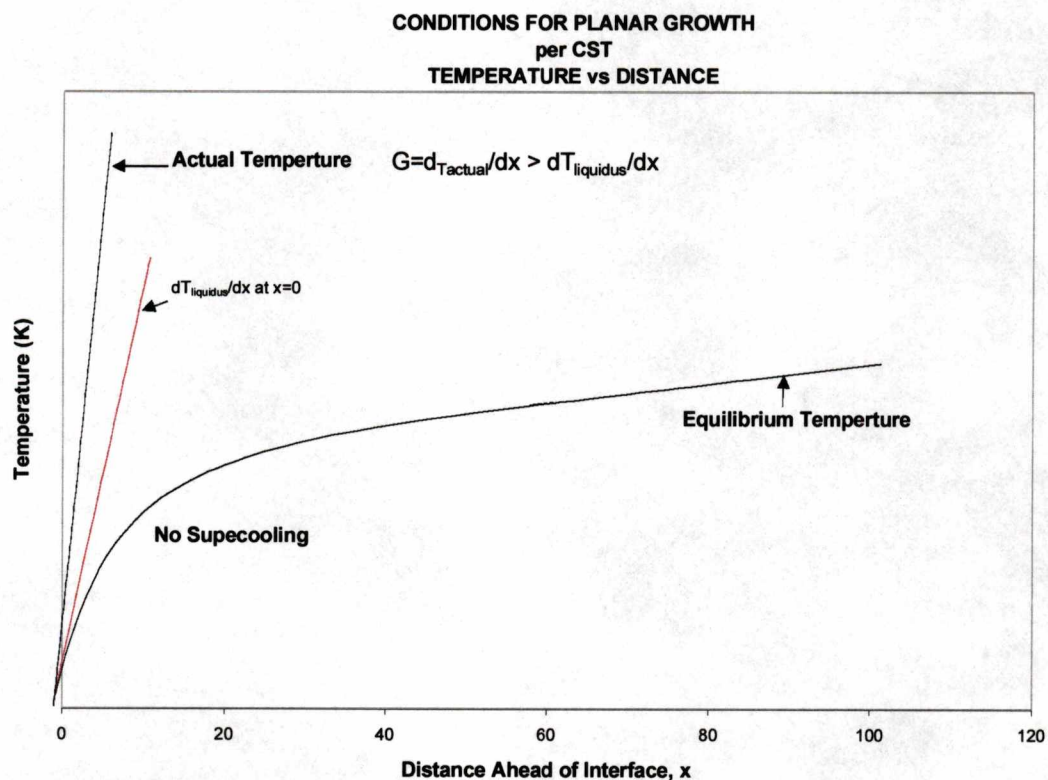
Adapted from: Flemings, M. C., *Solidification Processing*, McGraw Hill Book Company, New York, NY, ©1974, p. 59. Used with permission.



(b)

Figure II-7. (continued).

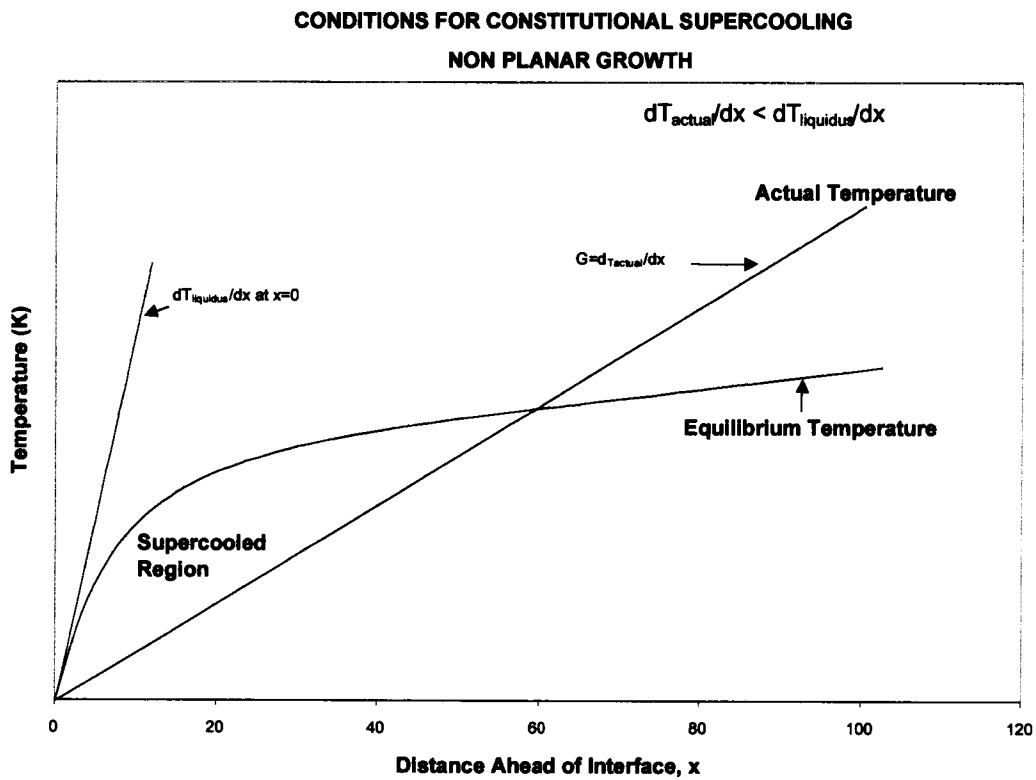
Adapted from: Flemings, M. C., *Solidification Processing*, McGraw Hill Book Company, New York, NY, ©1974, p. 59. Used with permission.



(c)

Figure II-7. (continued).

Adapted from: Flemings, M. C., *Solidification Processing*, McGraw Hill Book Company, New York, NY, ©1974, p. 59. Used with permission.



(d)

Figure II-7. (continued).

Adapted from: Flemings, M. C., *Solidification Processing*, McGraw Hill Book Company, New York, NY, ©1974, p. 59. Used with permission.

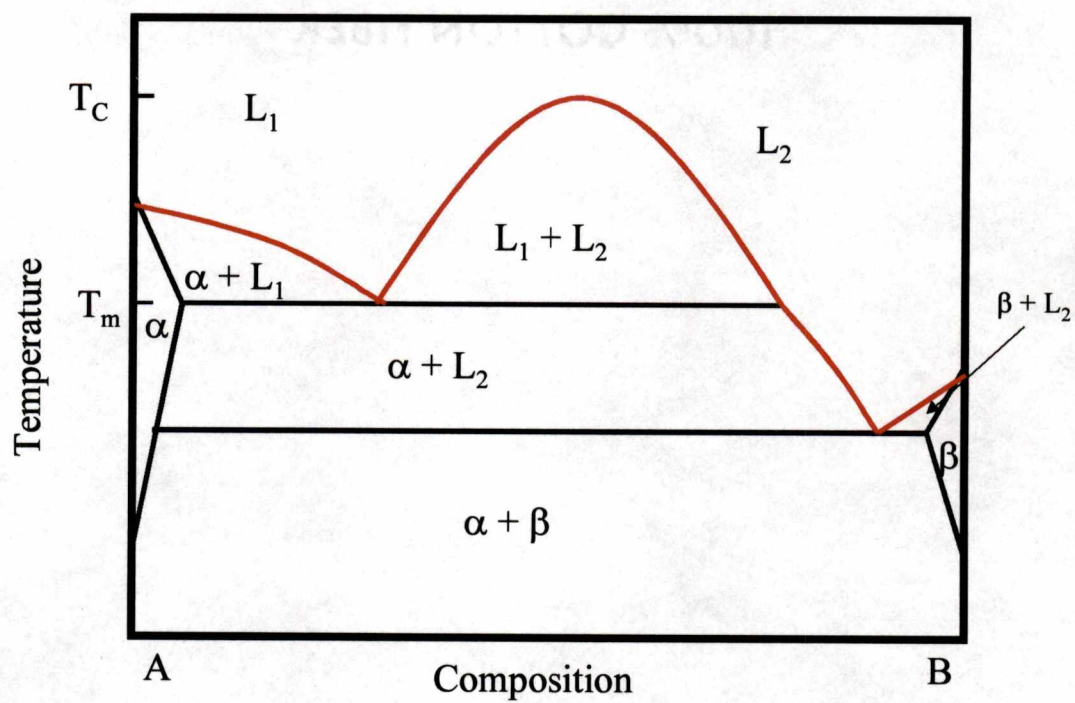


Figure II-8. Hypothetical monotectic phase diagram.

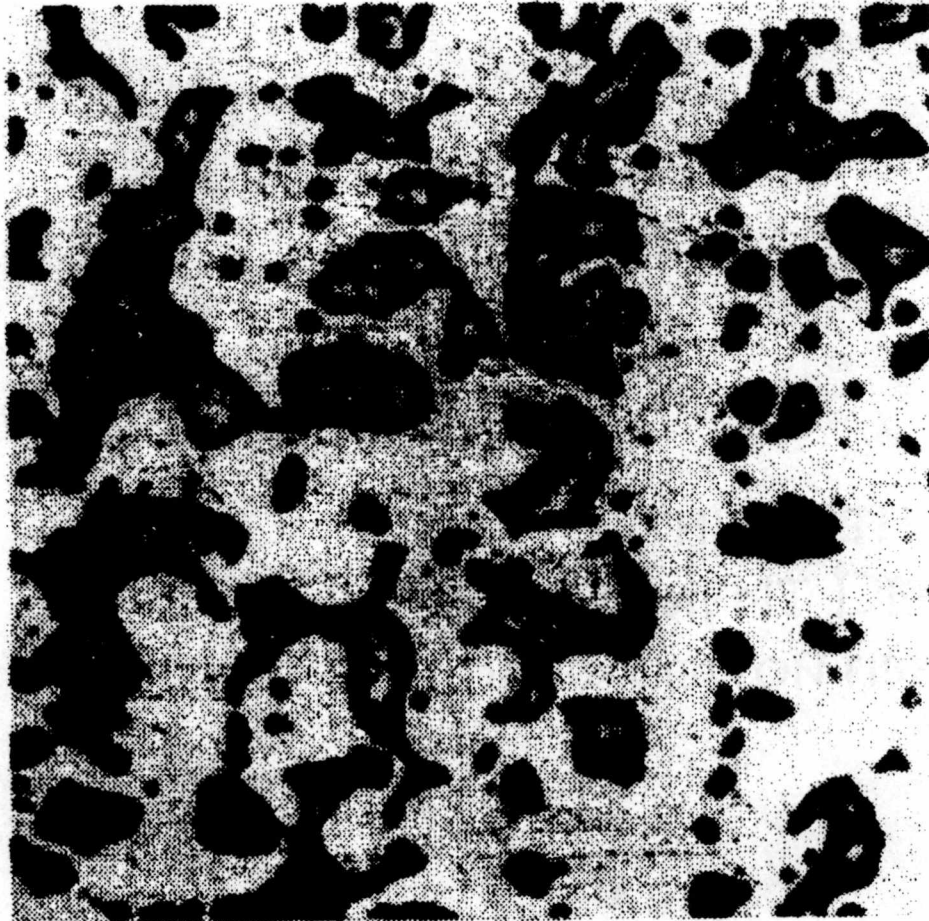


Figure II-9. Micrograph of a cast Cu-36 Pb alloy. Dark areas are Pb rich [B-7].

Source: Smith, W. F., *Principles of Materials Science and Engineering*, Second Edition, McGraw Hill Publishing Company, New York, NY, © 1990, p. 447. Used with permission.

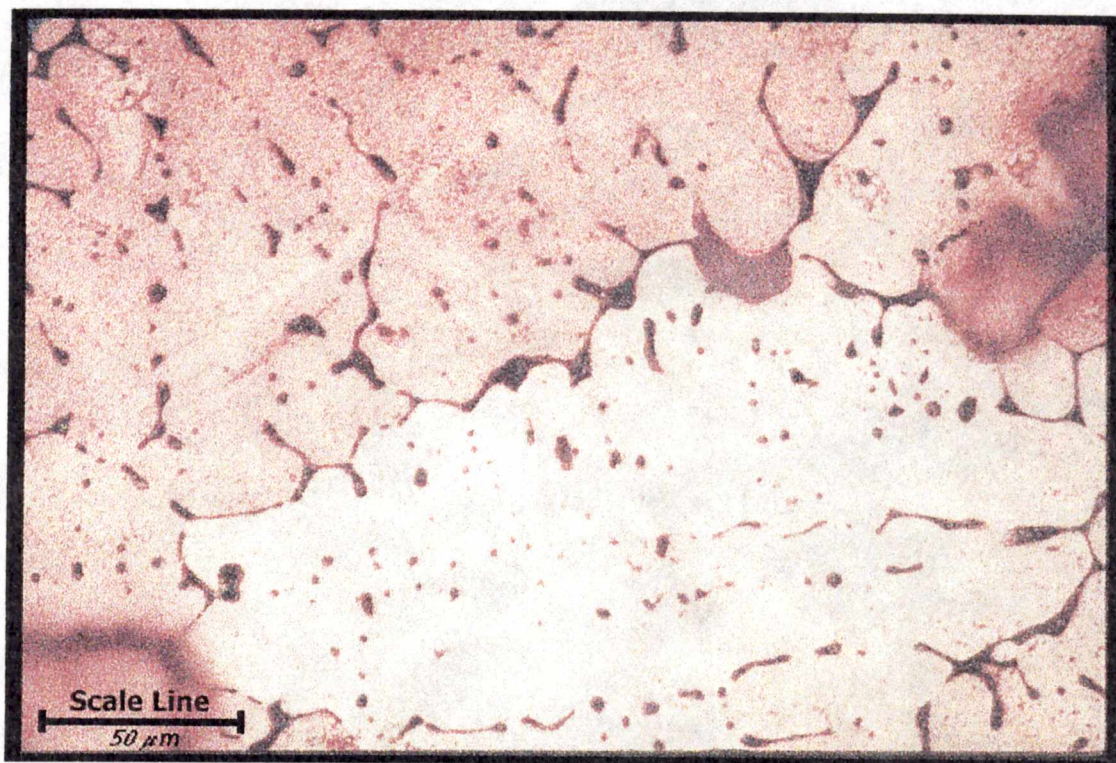


Figure II-10. Micrograph of a cast Cu-Pb alloy containing 6.83 wt% Pb. Dark areas are Pb rich [B-8].

Source: Copper Development Association Inc., 260 Madison Avenue, New York, NY, © 1996. Used with permission.

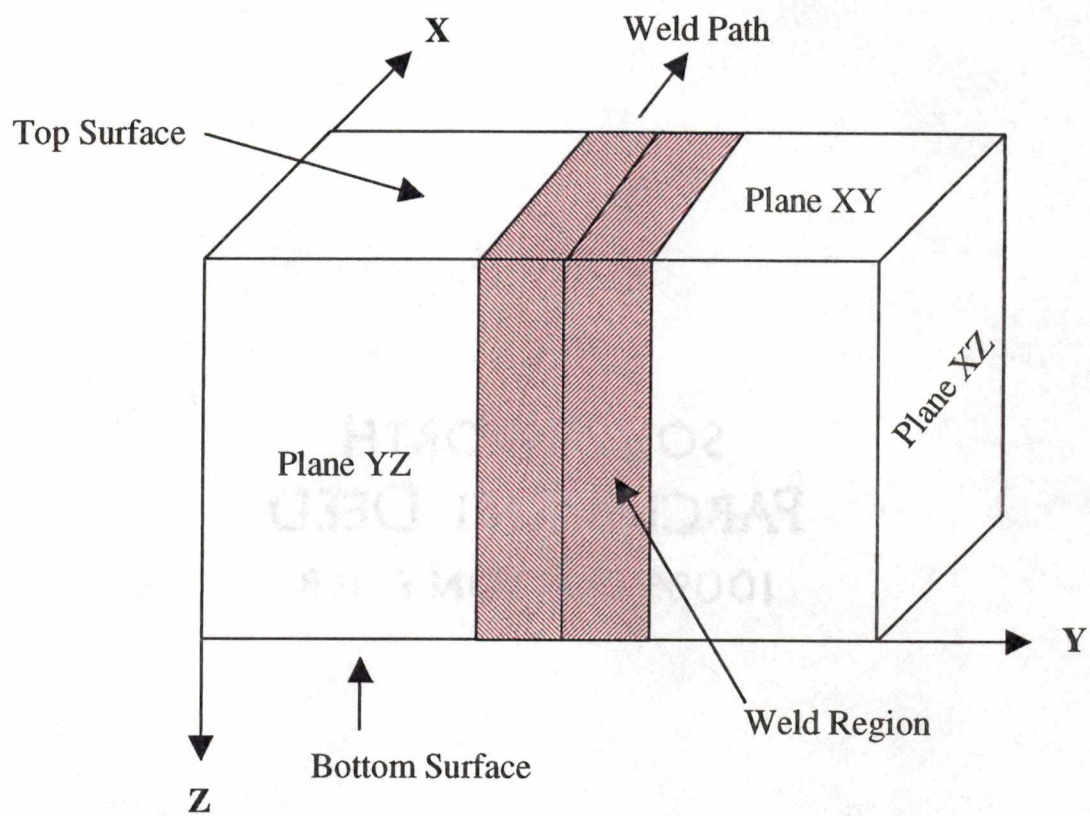
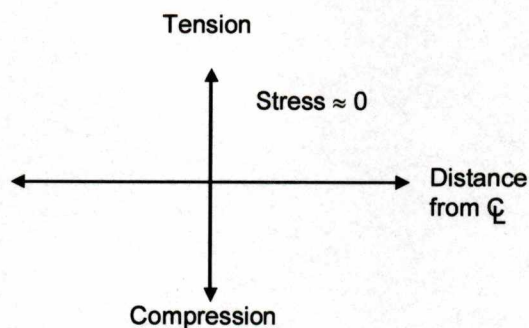
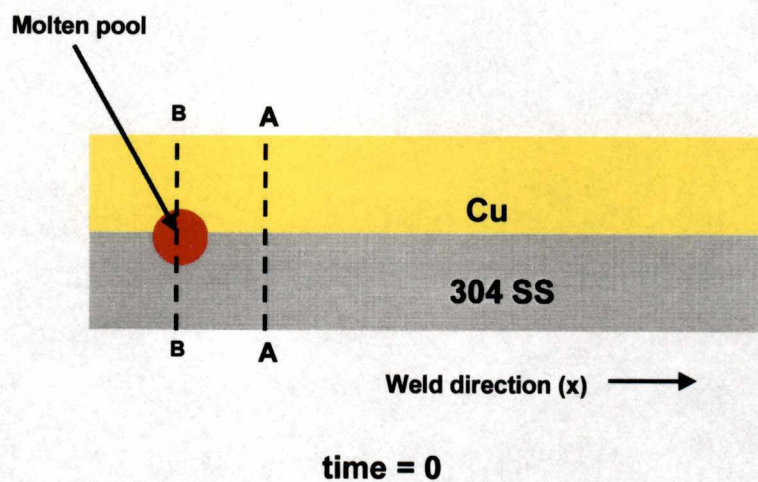


Figure II-11. Reference planes used to describe the weld region.



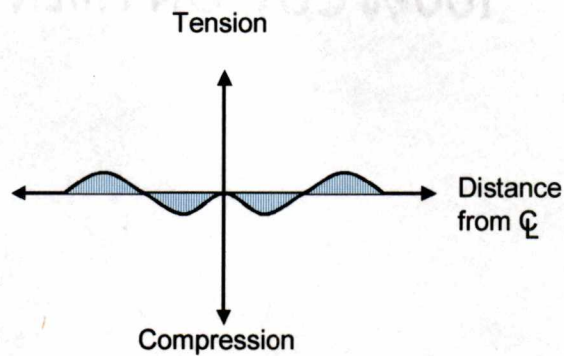
Section A - A



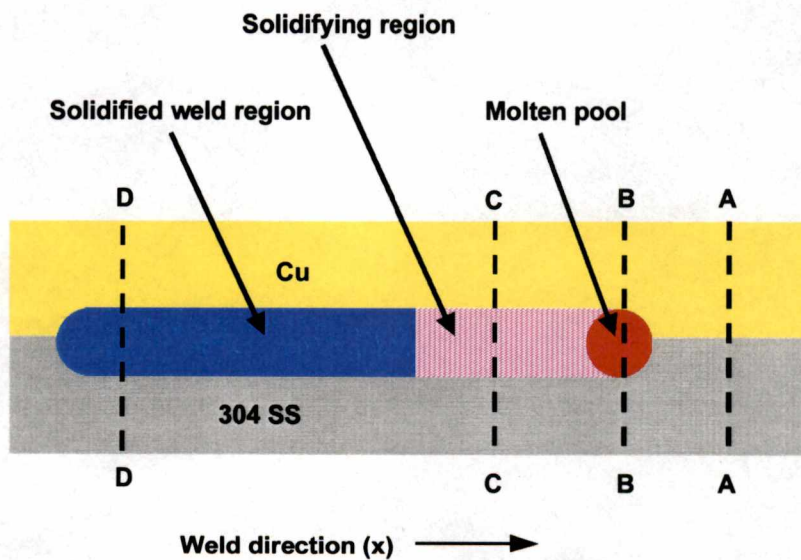
(a)

Figure II-12. Changes in stress during welding: (a) start of welding, (b) time a, (c) time b, and (d) time c [B-9].

Adapted from: The Welding Handbook, Volume 1, 7th Edition, American Welding Society, Miami, FL, ©1976, p. 224. Used with permission.



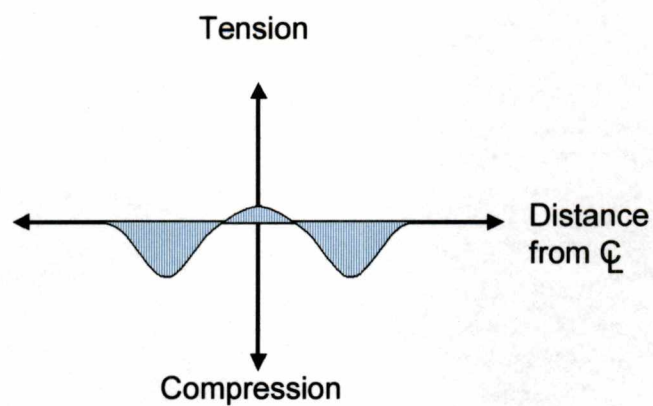
Section B - B



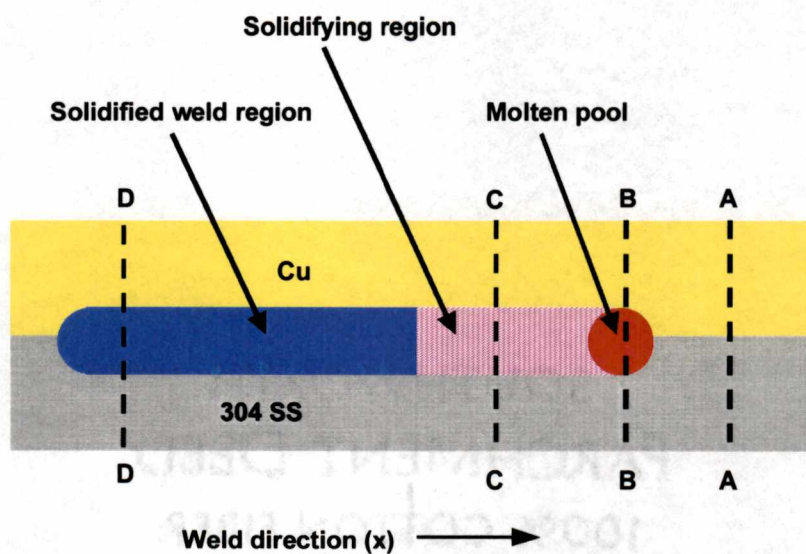
(b)

Figure II-12. (continued).

Adapted from: The Welding Handbook, Volume 1, 7th Edition, American Welding Society, Miami, FL, ©1976, p. 224. Used with permission.



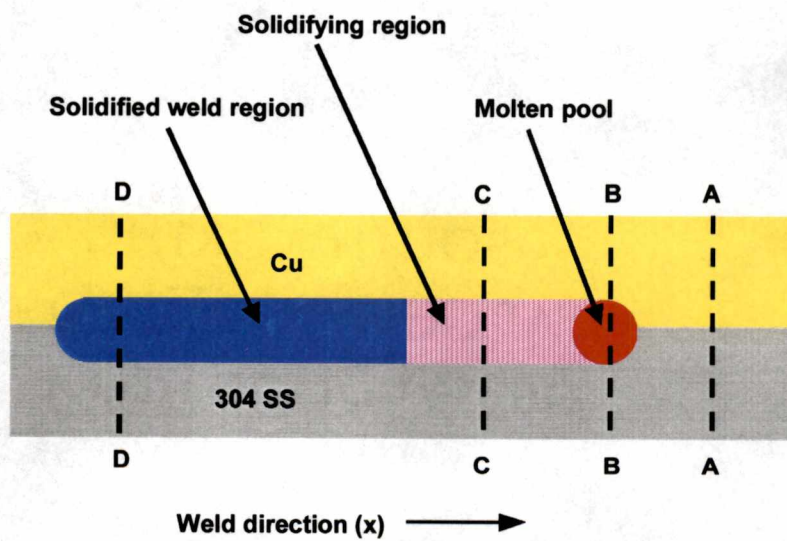
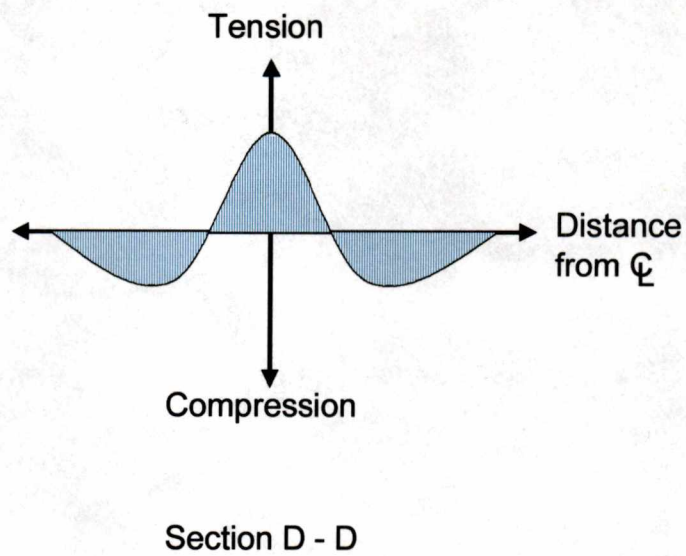
Section C - C



(c)

Figure II-12. (continued).

Adapted from: The Welding Handbook, Volume 1, 7th Edition, American Welding Society, Miami, FL, ©1976, p. 224. Used with permission.



(d)

Figure II-12. (continued).

Adapted from: The Welding Handbook, Volume 1, 7th Edition, American Welding Society, Miami, FL, ©1976, p. 224. Used with permission.

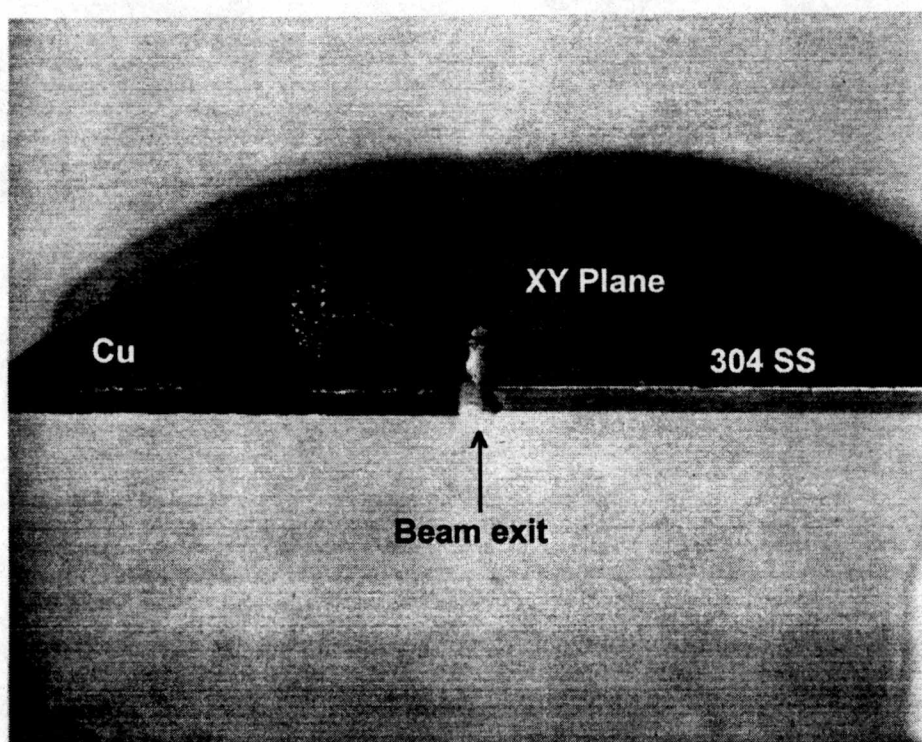


Figure II-13. Top view of a weld showing an area of fluid expulsion where the electron beam exited the weld pool prior to solidification.

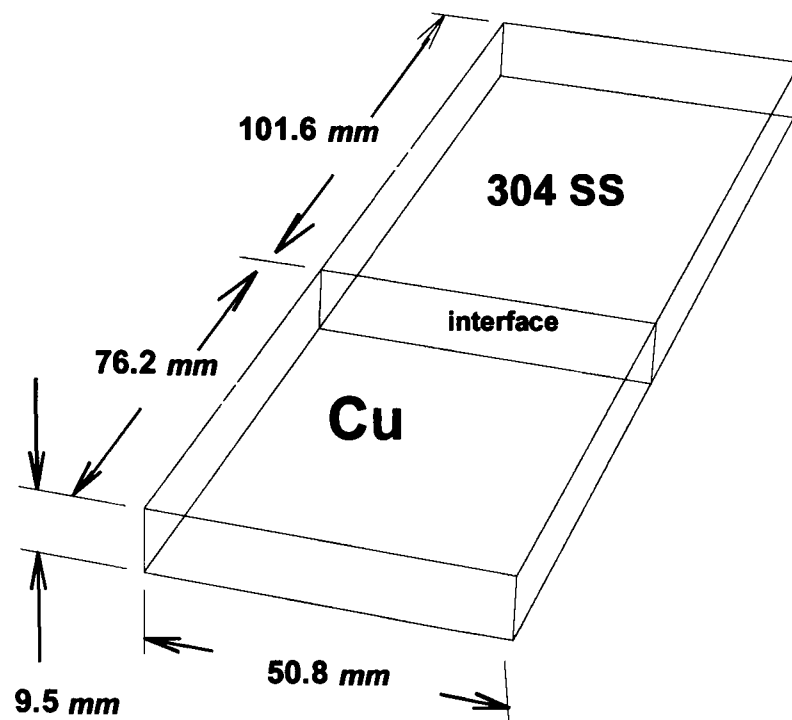


Figure III-1. Layout of the weld coupons.

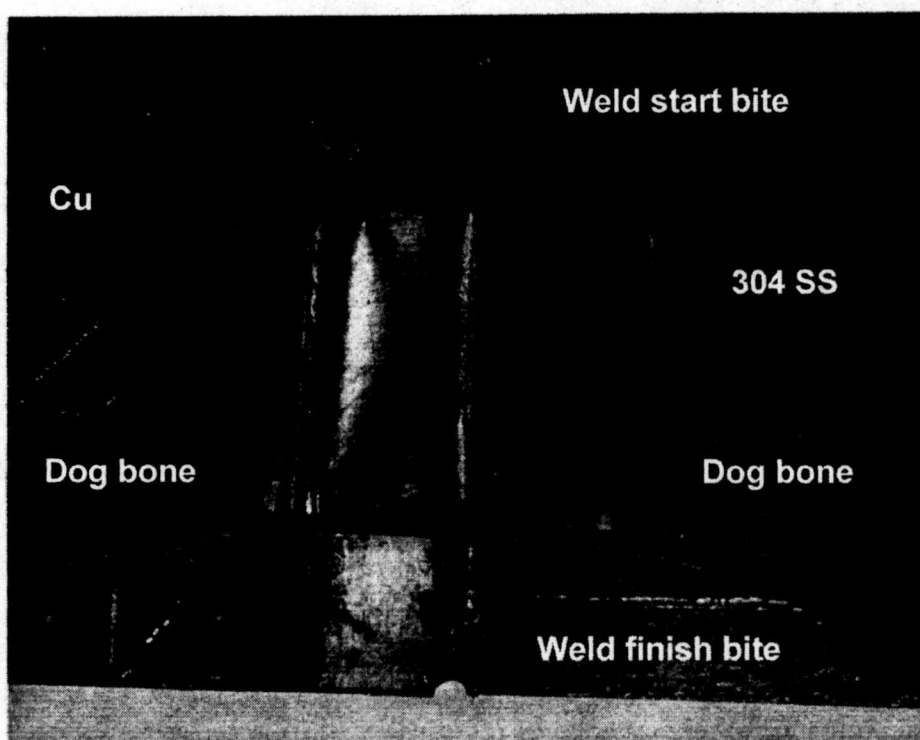


Figure III-2. Overhead view of dog bones with bites

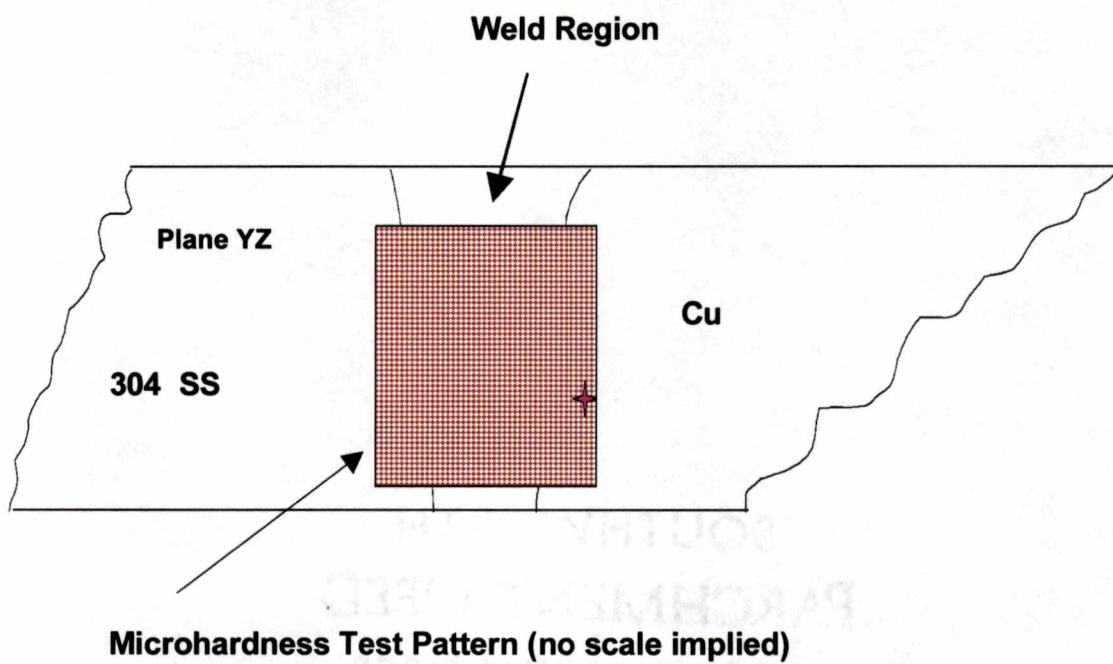


Figure III-3. Schematic of microhardness test pattern.

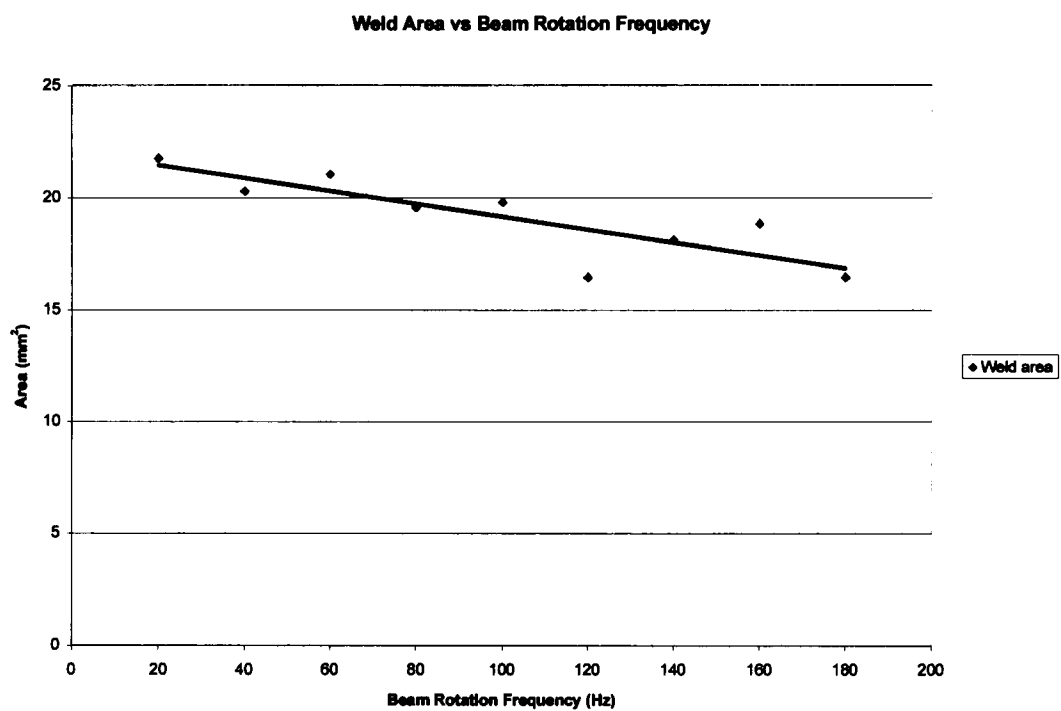


Figure IV-1. Weld region area in the YZ plane versus electron beam rotation frequency. The area is based upon a trapezoid formed using the top weld bead width, the bottom weld bead width, and the thickness of the sample.

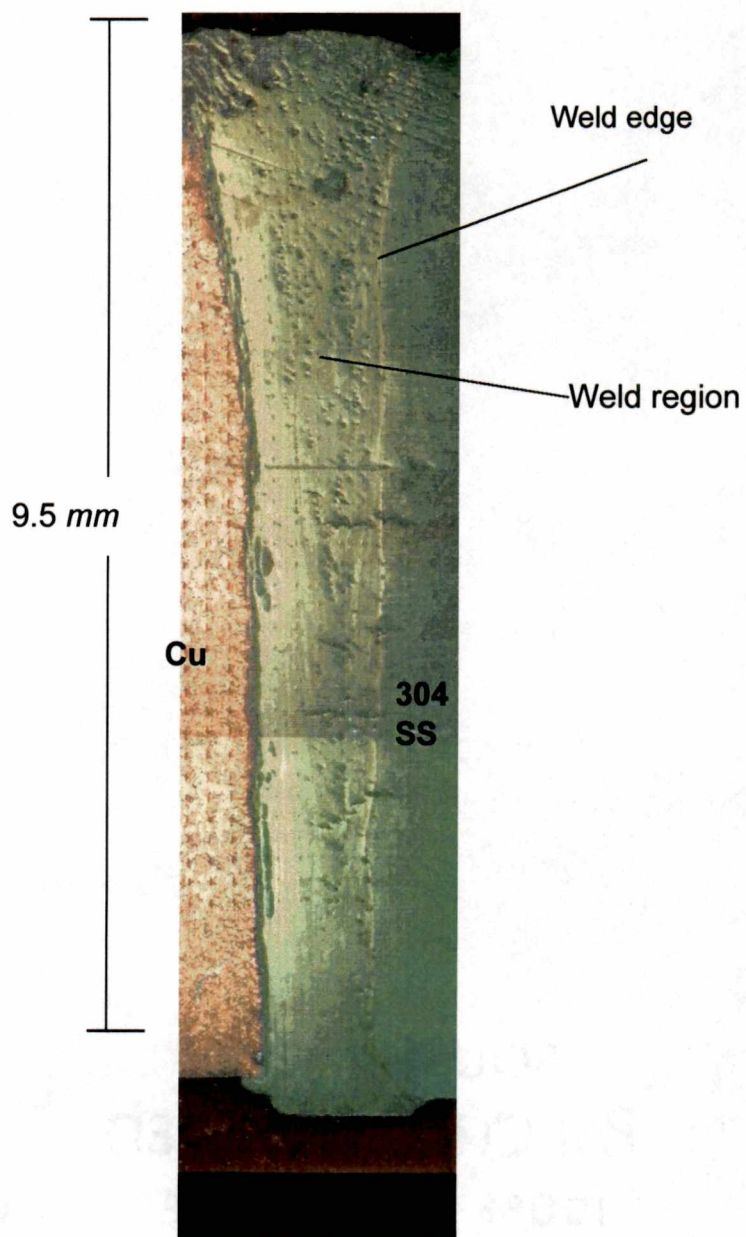


Figure IV-2. Composite figure, made using three limited field-of-view images, of the cross-section (YZ plane) of sample 4-3S welded at 40 Hz.

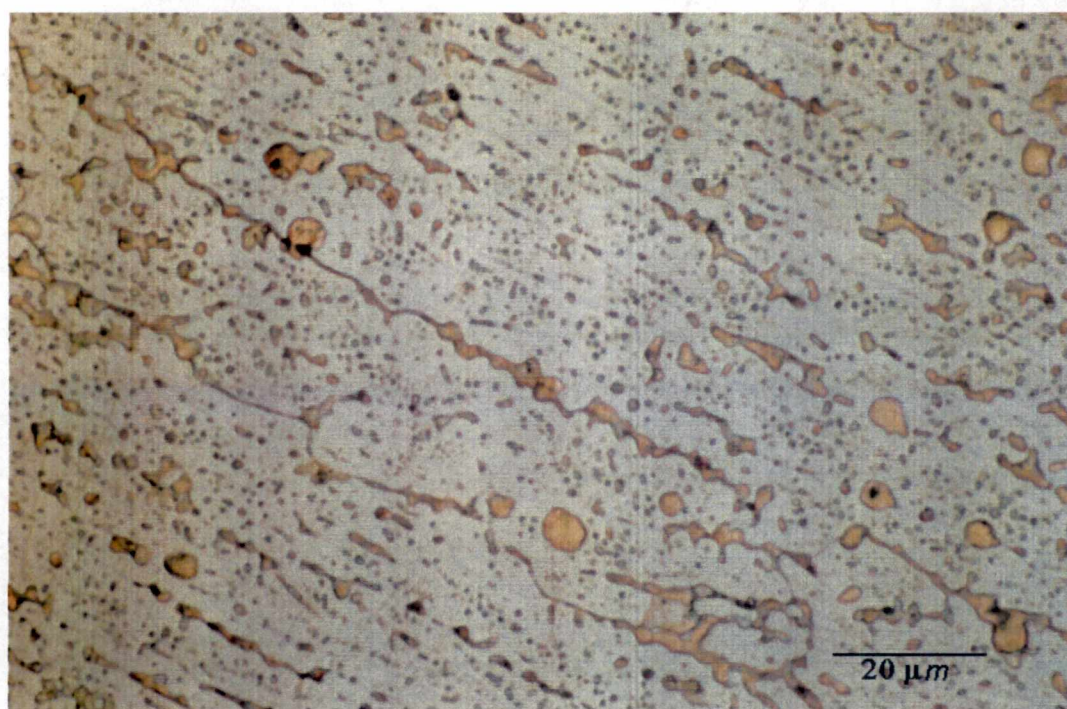


Figure IV-3. Optical micrograph of sample 4-3F, welded at 40 Hz, showing Cu rich interdendritic material. The light regions are Fe rich dendrites; the dark areas are Cu rich material.



Figure IV-4. Optical micrograph of a portion of the top weld nugget region along the Cu wall of sample 4-3S, welded at 40 Hz, showing Fe rich particles (light colored particles) surrounded by a Cu rich matrix (dark material).

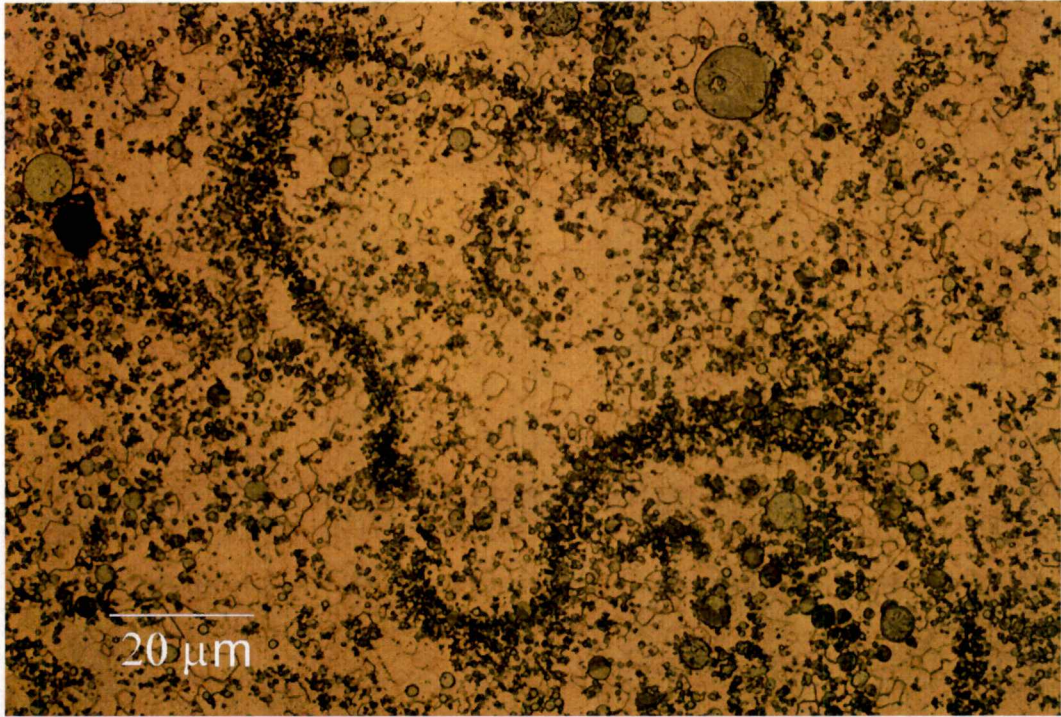


Figure IV-5. Optical micrograph of sample 5-10F showing the distribution of small structures within the weld region, beam rotation set at 180 Hz.

Zone adjacent to HAZ with
uniform microhardness

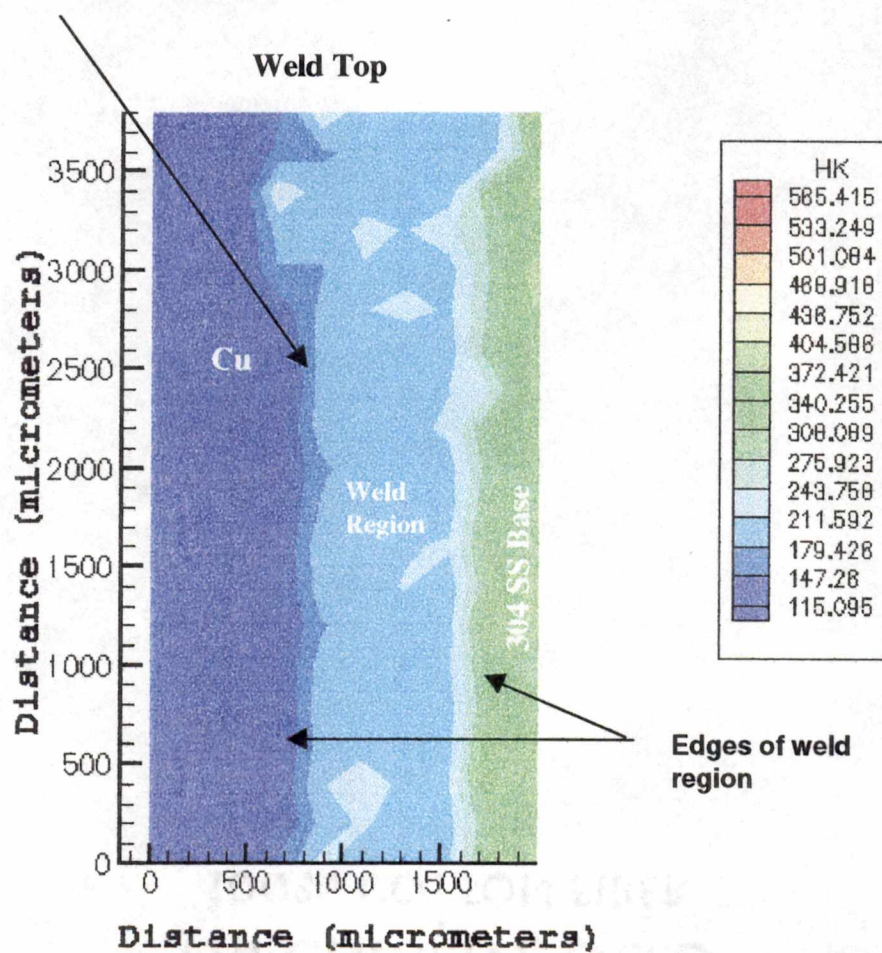


Figure IV-6. Map of Knoop microhardness in the YZ plane for sample 5-1F, welded without electron beam rotation.

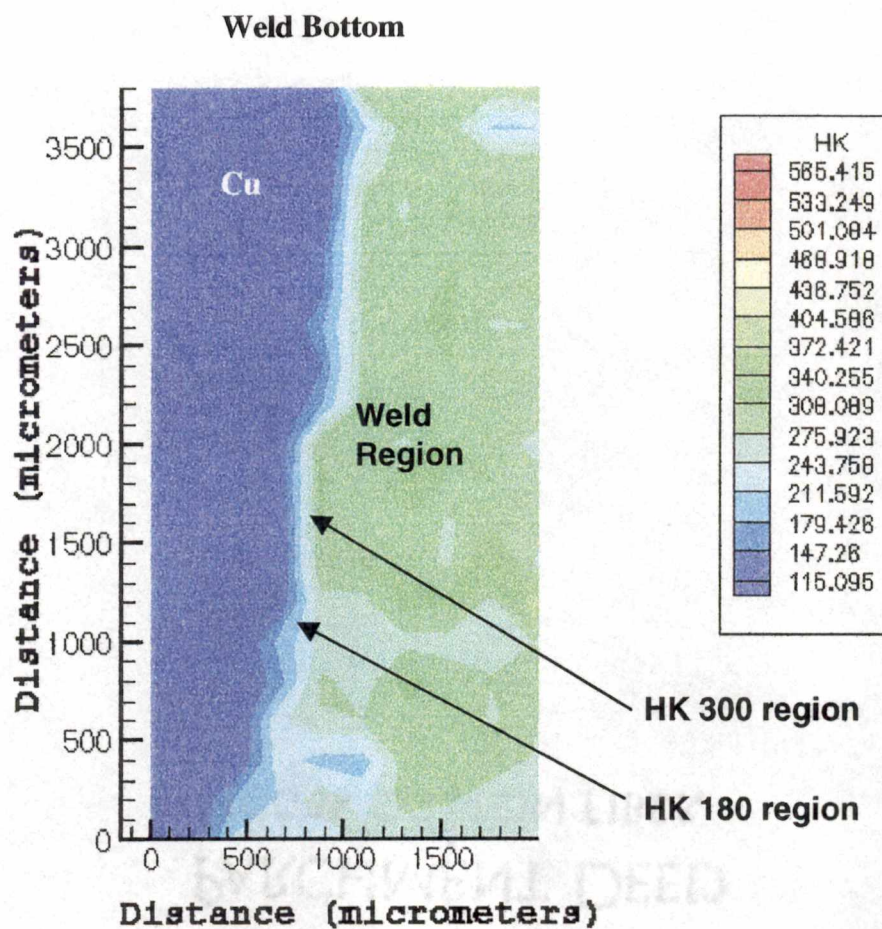


Figure IV-7. Map of Knoop microhardness in the YZ plane for sample 4-2S, beam rotation frequency set at 20 Hz.

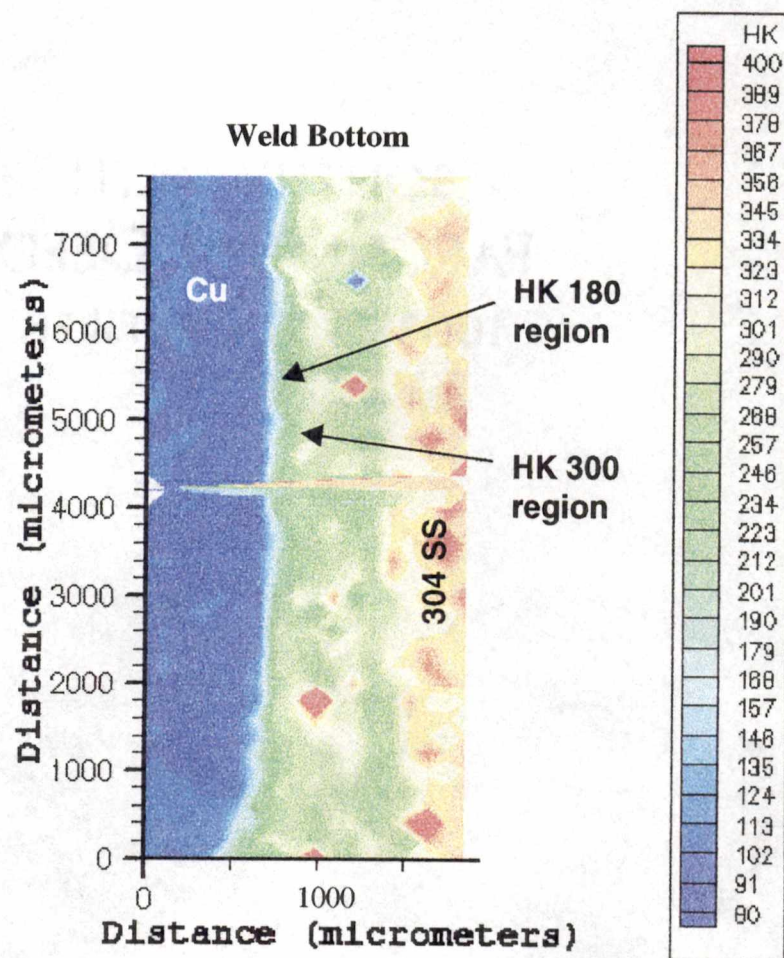


Figure IV-8. Map of Knoop microhardness in the YZ plane for sample 4-3S, beam rotation frequency set at 40 Hz.

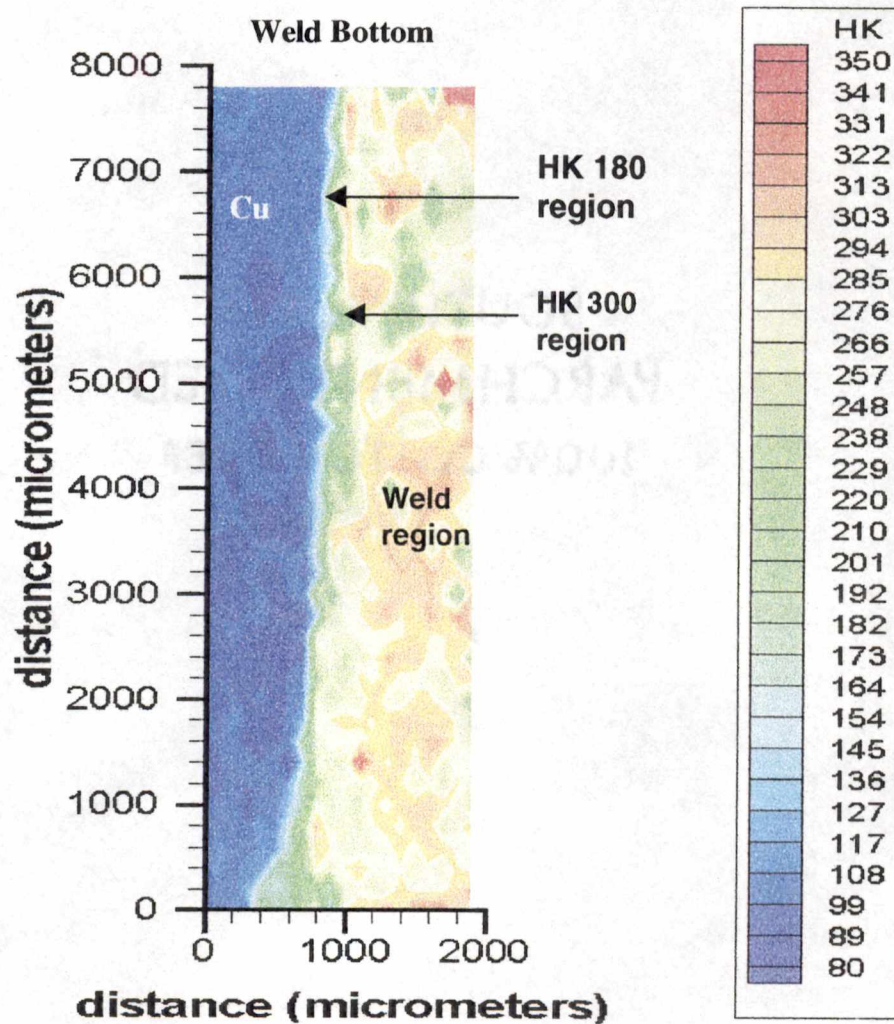


Figure IV-9. Map of Knoop microhardness in the YZ plane for sample 2-4S, beam rotation frequency set at 60 Hz. (Includes only approximately one-half of the weld region on the Cu side.)

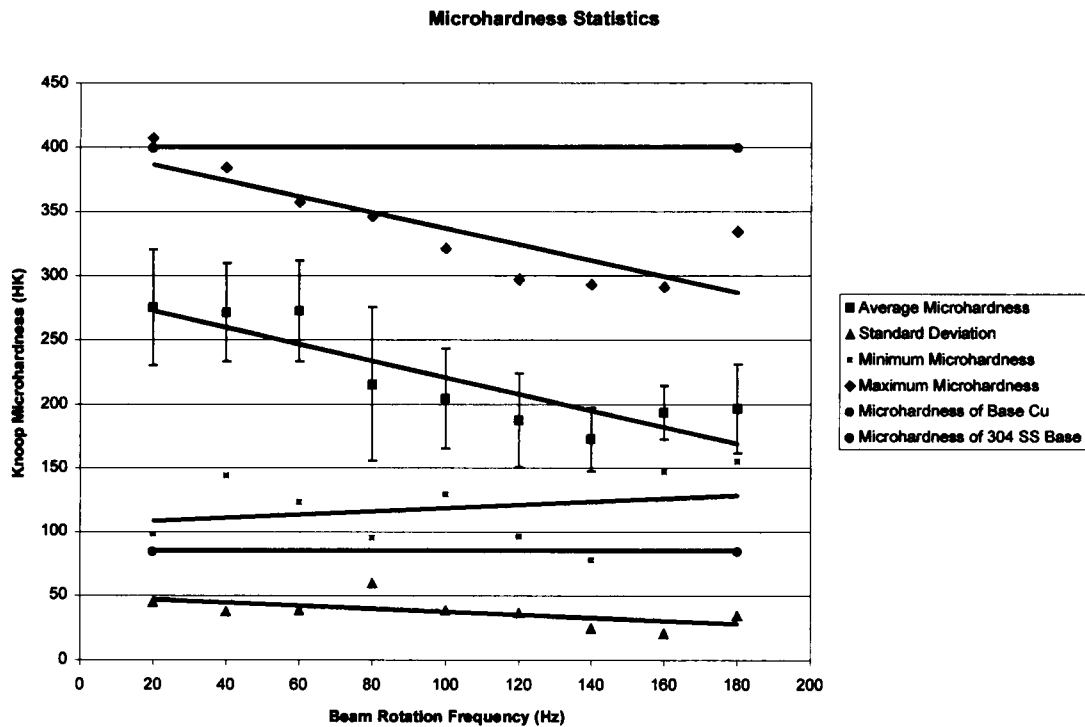


Figure IV-10. Average Knoop microhardness, standard deviation, minimum, and maximum values of Knoop microhardness as a function of electron beam rotation frequency. Linear curve fits of the data are included to show trends. (Note that the Cu base material, 304 SS base material, and the average Knoop microhardness of the baseline sample welded without electron beam rotation have Knoop microhardness values of HK85, HK400, and HK205 respectively.)

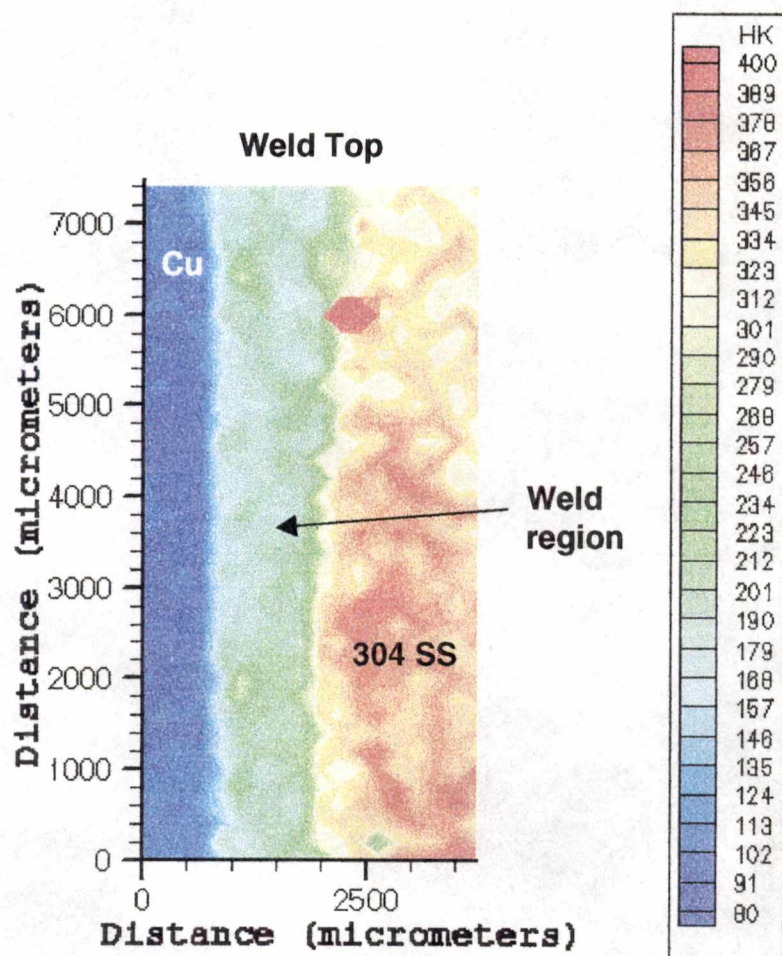


Figure IV-11. Map of Knoop microhardness in the YZ plane for sample 4-5F, beam rotation frequency set at 80 Hz.

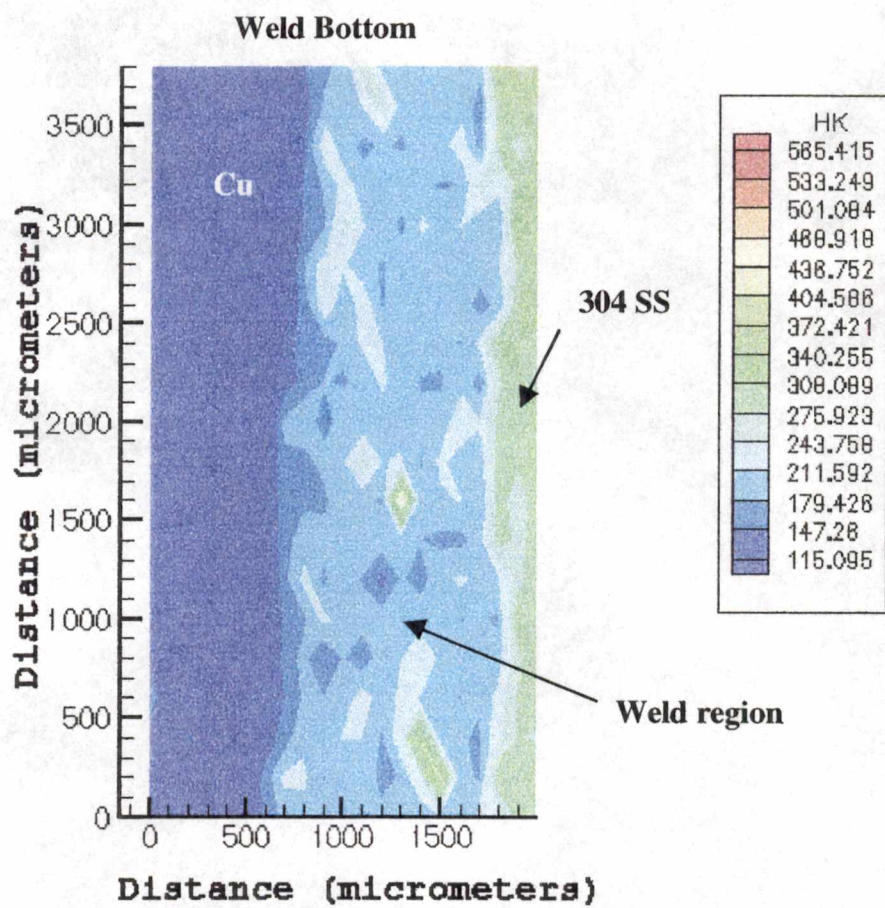


Figure IV-12. Map of Knoop microhardness in the YZ plane for sample 5-6S, beam rotation frequency set at 100 Hz.

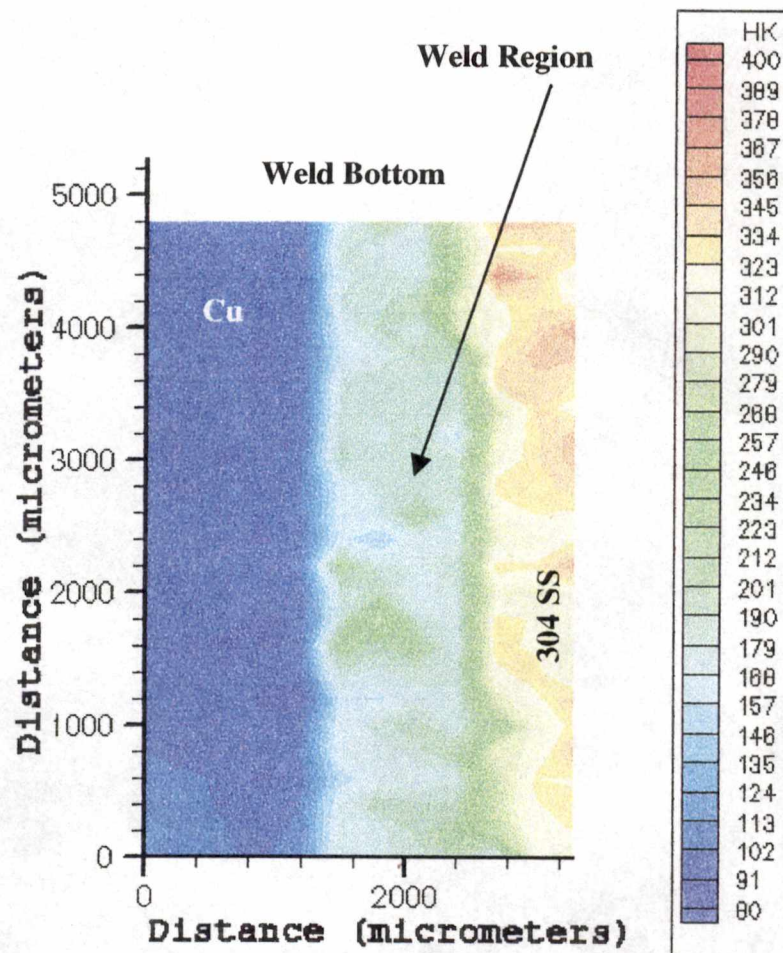


Figure IV-13. Map of Knoop microhardness in the YZ plane for sample 3-7S, beam rotation frequency set at 120 Hz.

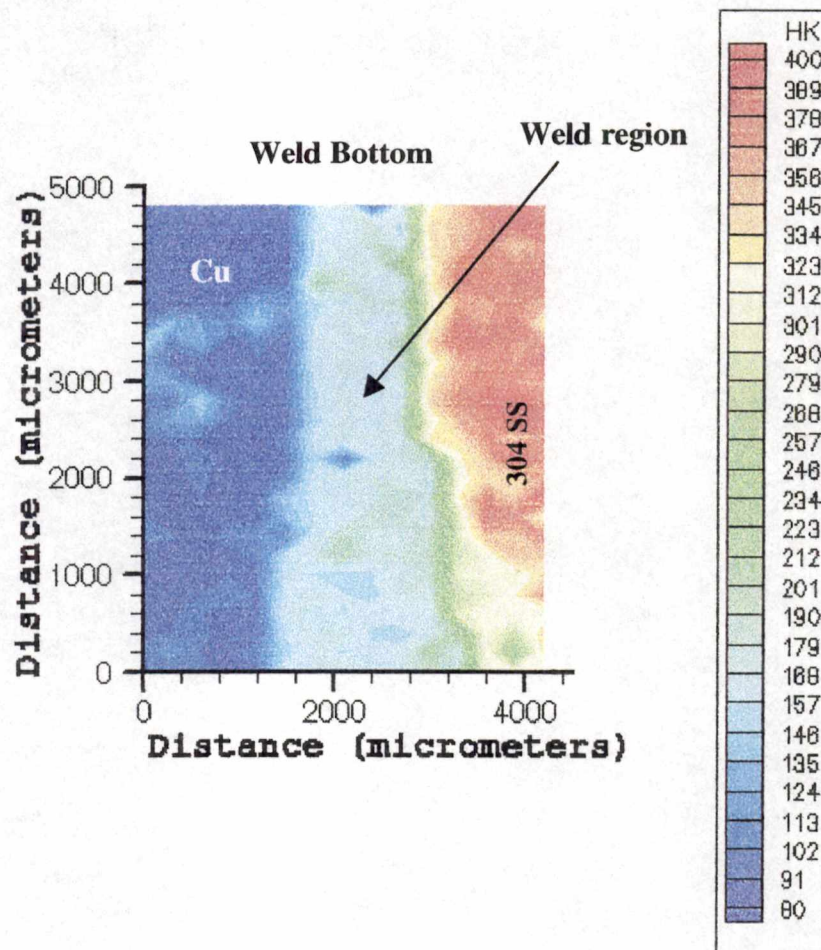


Figure IV-14. Map of Knoop microhardness in the YZ plane for sample 1-8S, beam rotation frequency set at 140 Hz.

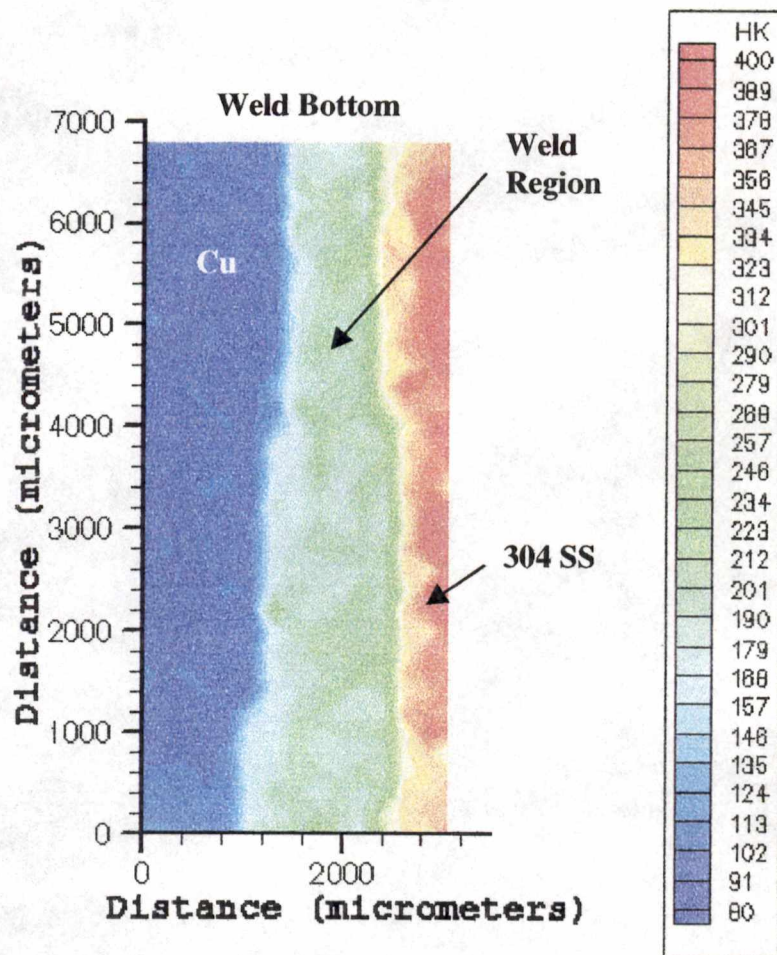


Figure IV-15. Map of Knoop microhardness in the YZ plane for sample 5-9S, beam rotation frequency set at 160 Hz.

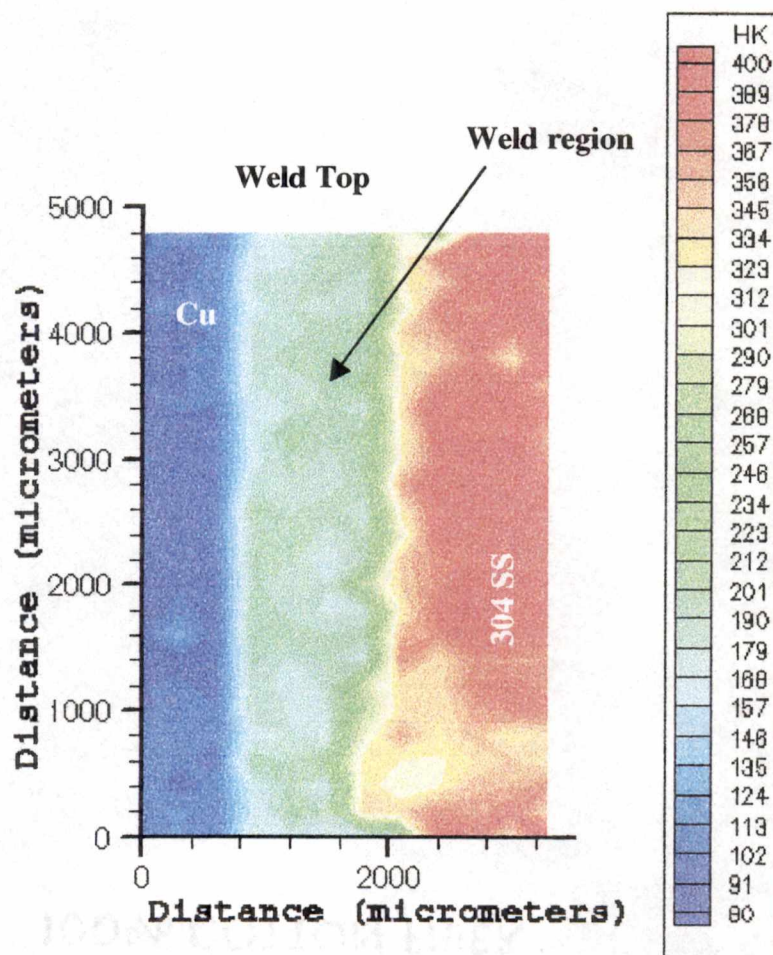


Figure IV-16. Map of Knoop microhardness in the YZ plane for sample 3-10F, beam rotation frequency set at 180 Hz.

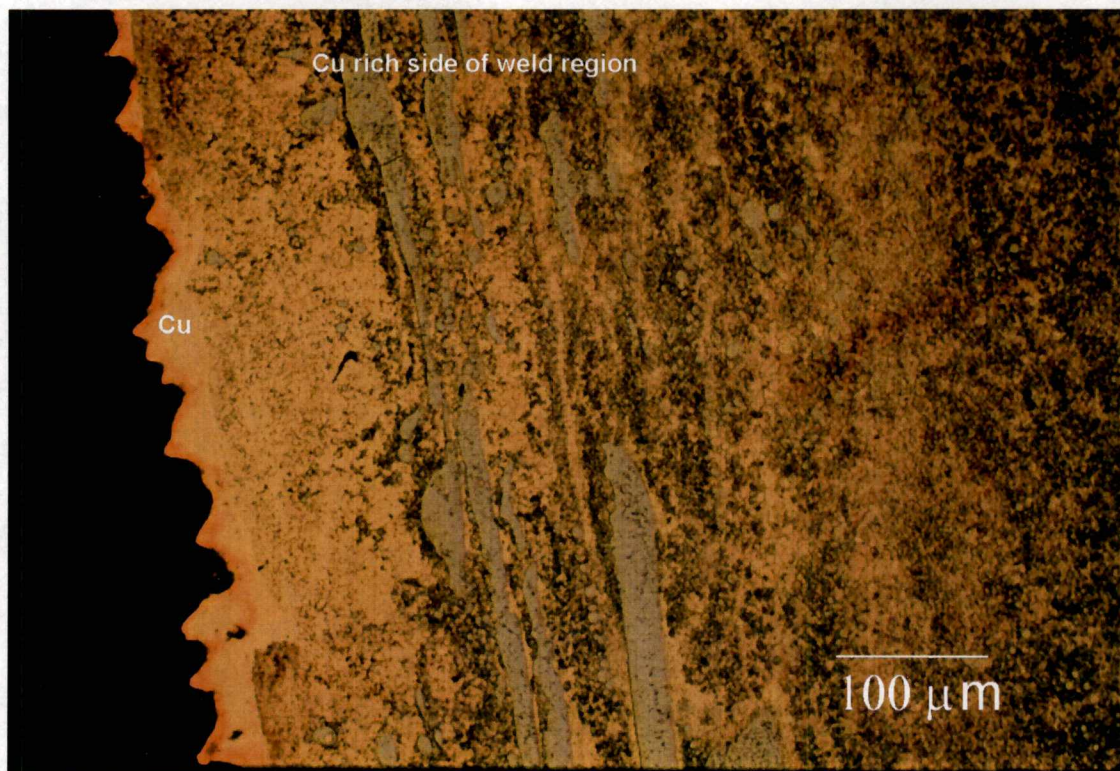
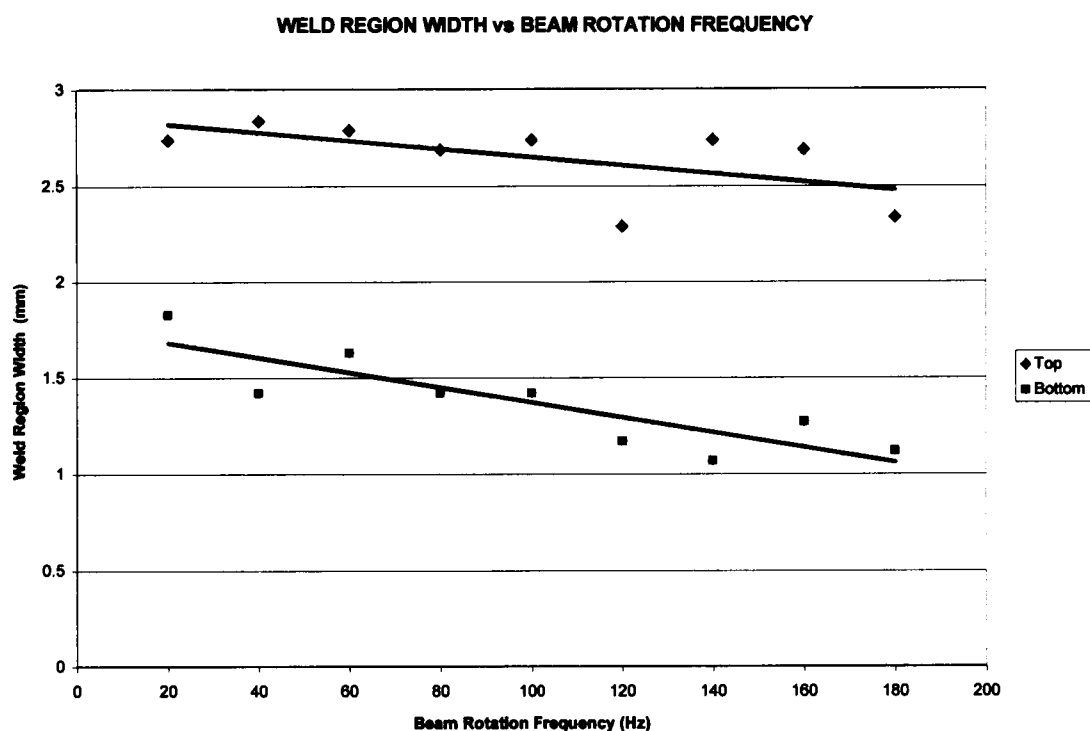
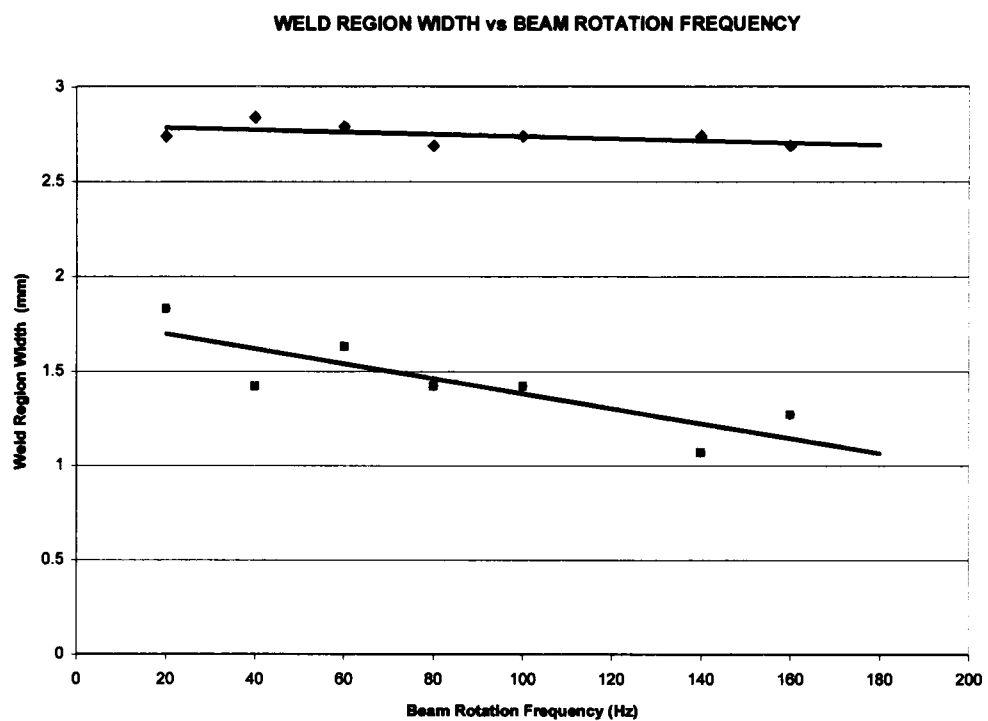


Figure IV-17. Micrograph of sample 3-5F, in the YZ plane, showing ductile failure in the Cu outside of the weld region, beam rotation set at 80 Hz. The black region is mounting material.



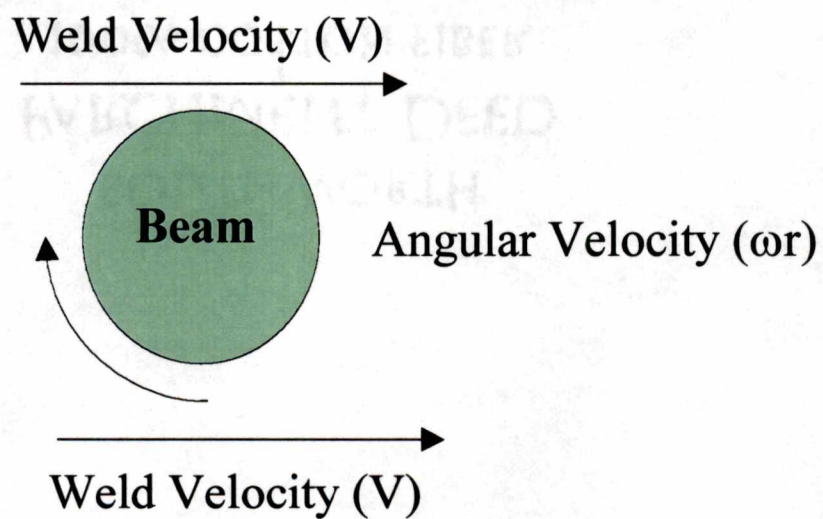
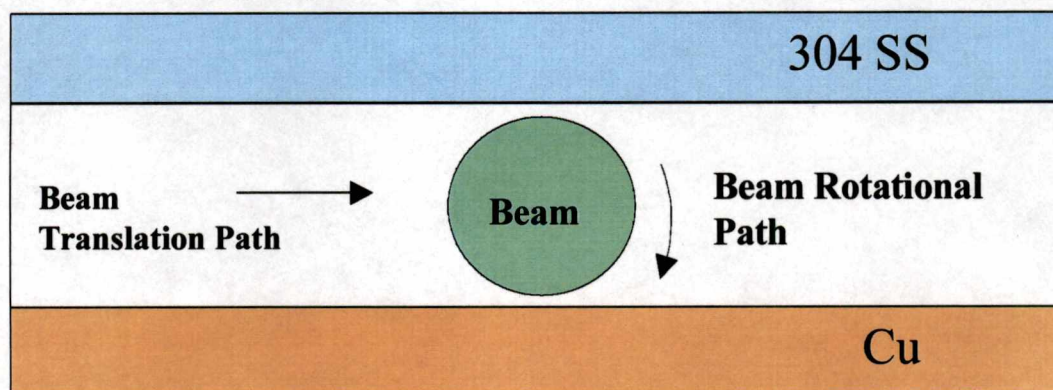
(a)

Figure IV-18. Top and bottom weld bead width in the YZ plane versus electron beam rotation frequency for: (a) all electron beam rotation frequencies and, (b) with the electron beam rotation frequencies of 120 Hz and 180 Hz excluded. The baseline sample widths are 2.44 mm on top and 1.02 mm on the bottom.



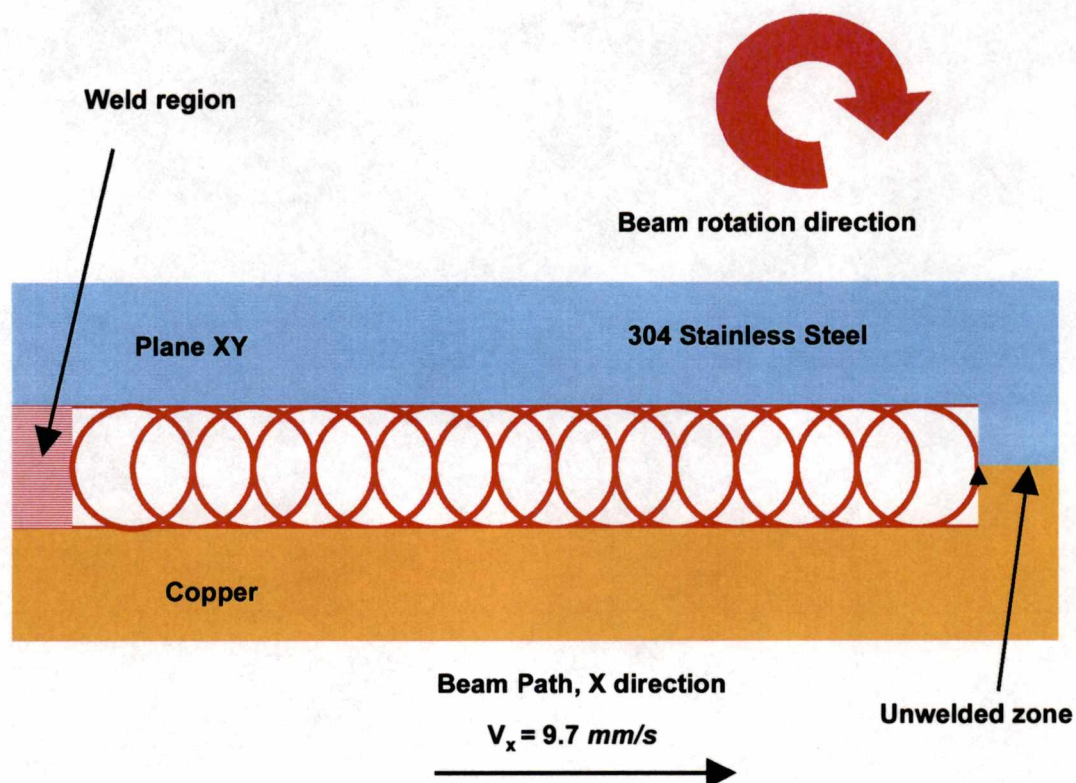
(b)

Figure IV-18. (continued).



(a)

Figure IV-19. Schematic diagram of: (a) beam directions and, (b) velocity vectors.



(b)

Figure IV-19. (continued).

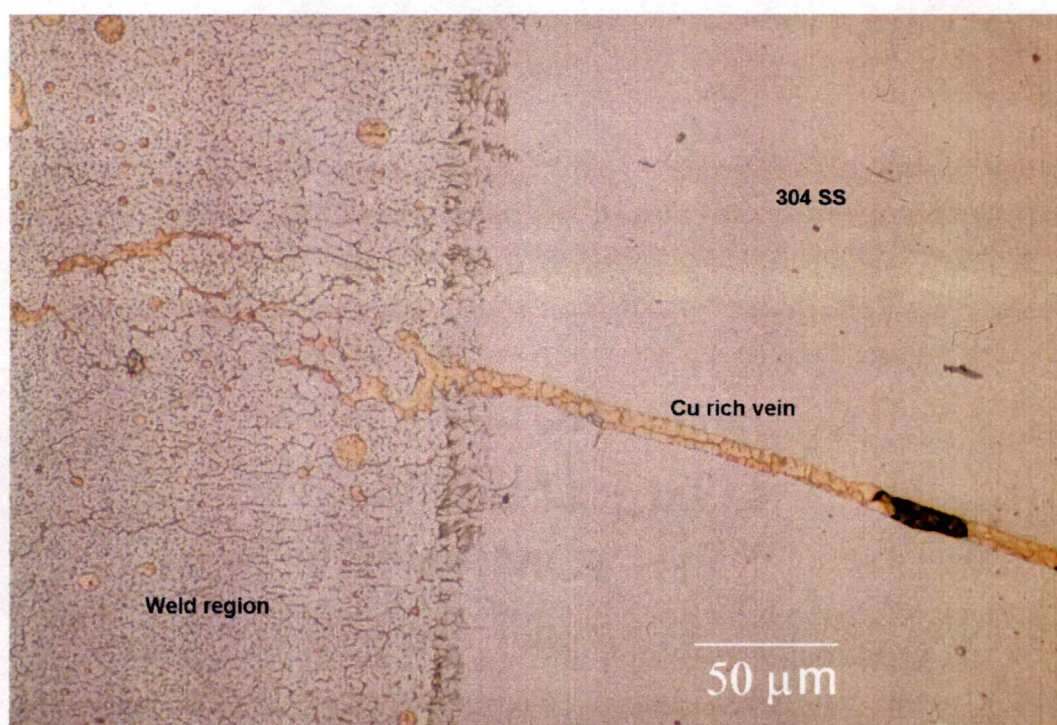


Figure IV-20. Micrograph of sample 5-6S showing penetration of Cu rich material into the grain boundary along the 304 SS side of the weld region, beam rotation set at 100 Hz.

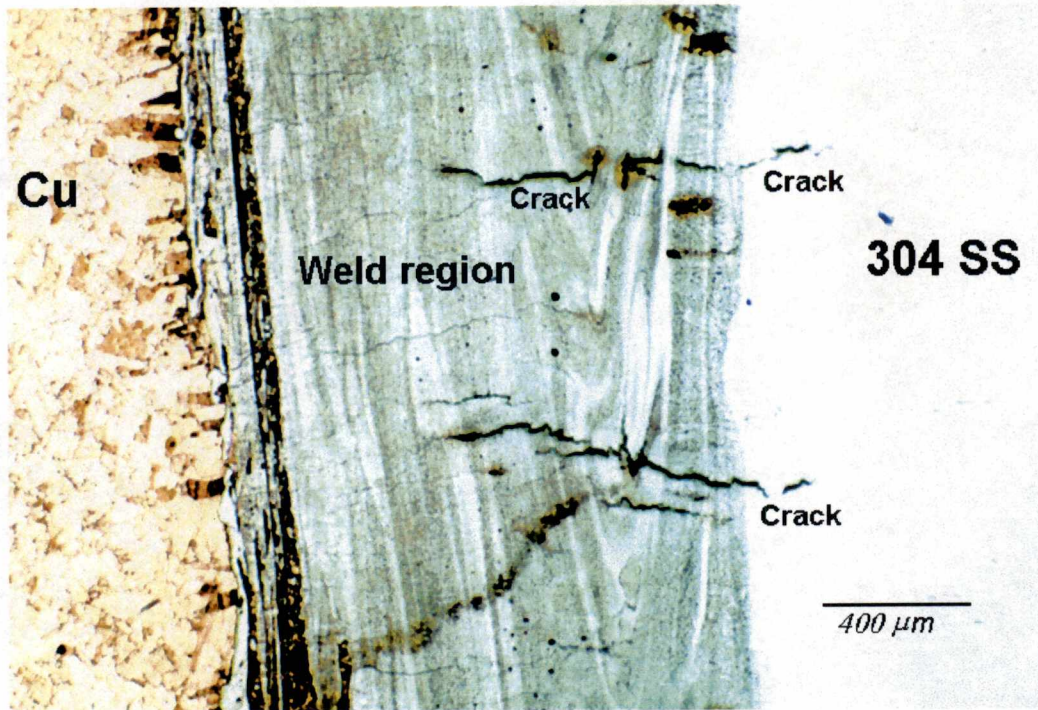


Figure IV-21. Micrograph of sample 4-3F, welded at 40 Hz, showing cracks in the weld region that extend into the HAZ.

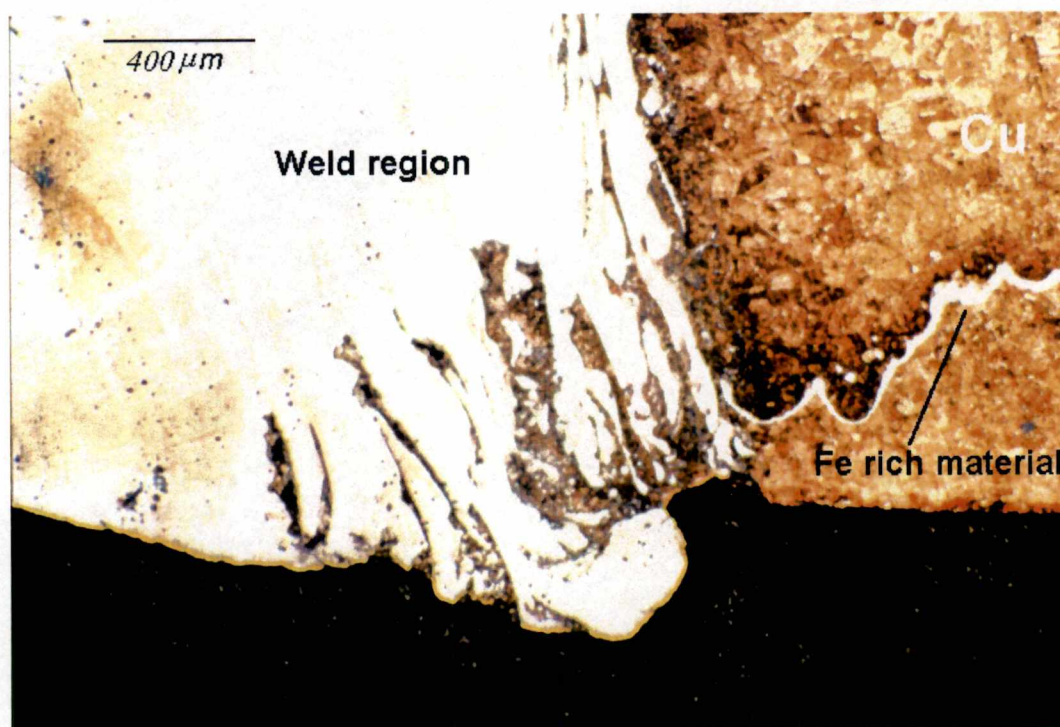


Figure IV-22. Micrograph of sample 4-2S, welded at 20 Hz, showing infiltration of Fe rich material into the Cu HAZ.

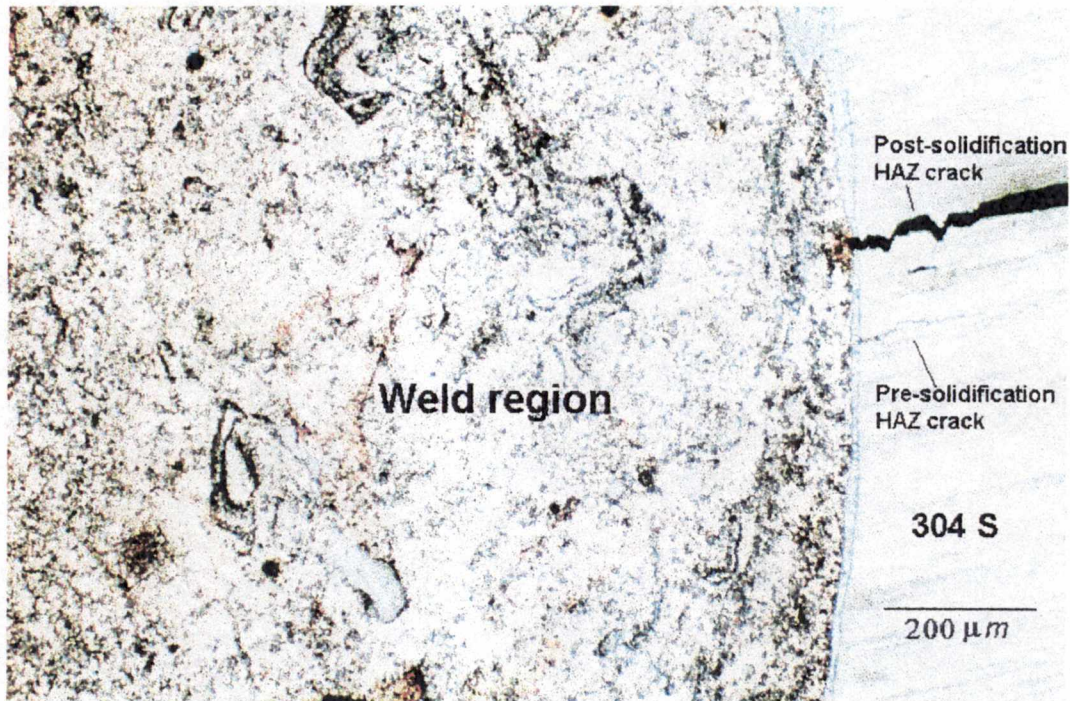


Figure IV-23. Micrograph of sample 5-10F, welded at 180 Hz, showing a post solidification crack in the 304 SS HAZ that extends just into the weld region. A smaller pre-solidification crack containing Cu rich material is also shown.

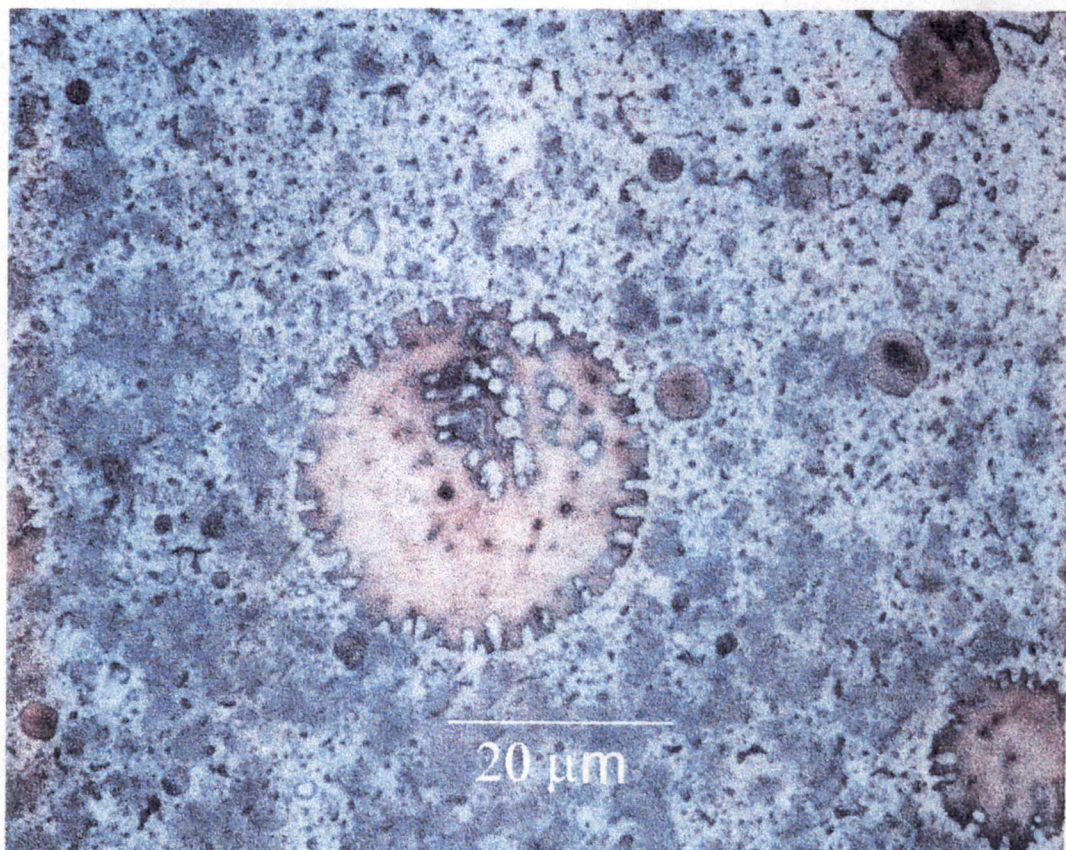


Figure IV-24. Micrograph of sample 3-5F showing divorced spherical structures.

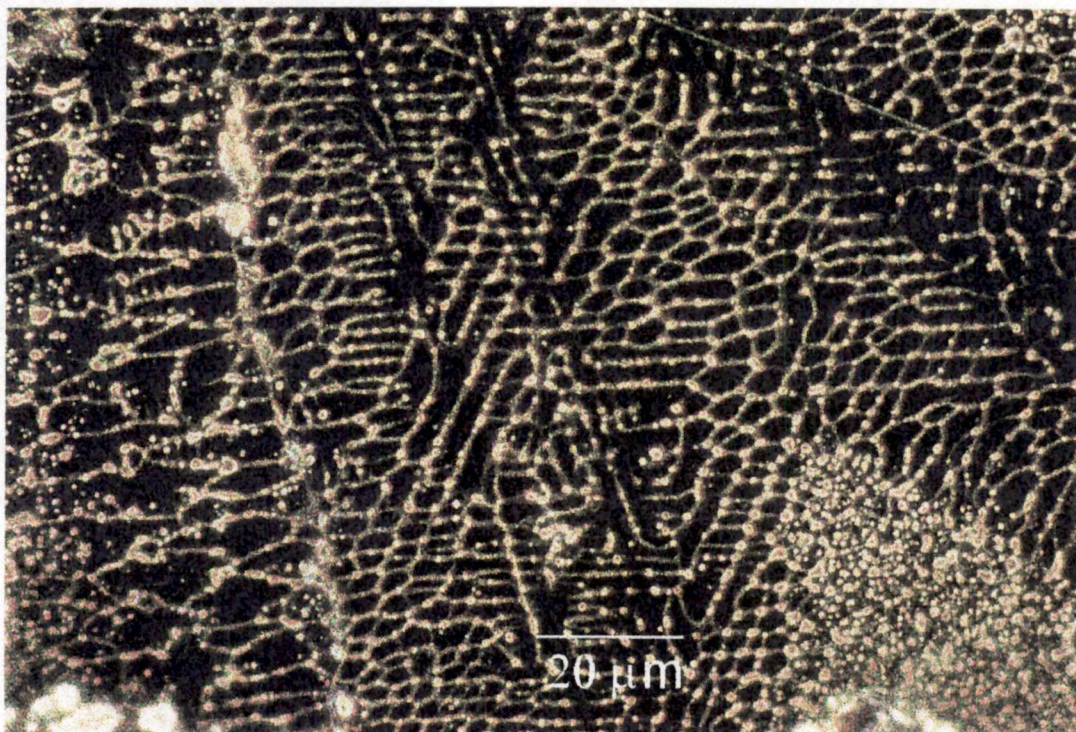


Figure IV-25. Darkfield micrograph of sample 1-4F showing closely packed fibrous composite solidification structures within the weld region, electron beam rotation set at 60 Hz. The Fe rich material appears black while the Cu rich fibrous material appears gold in this micrograph.

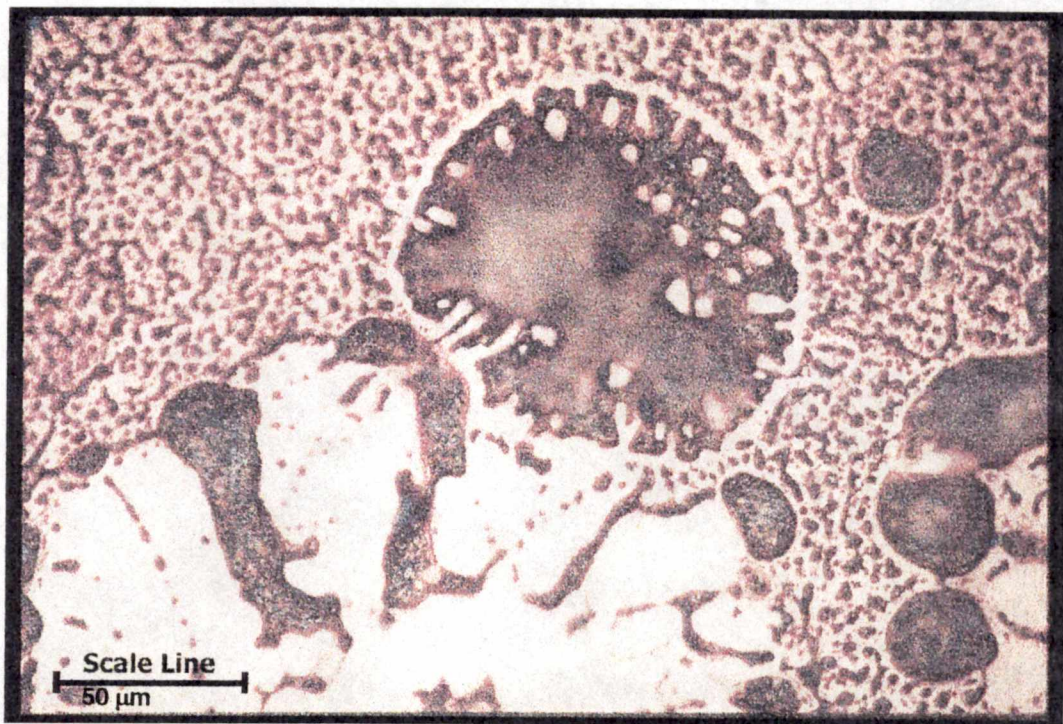


Figure IV-26. Micrograph of a cast Cu-35 wt% Pb alloy containing a divorced spherical region. Dark regions are Pb rich material [B-10].

Source: Copper Development Association Inc., 260 Madison Avenue, New York, NY, © 1996. Used with permission.

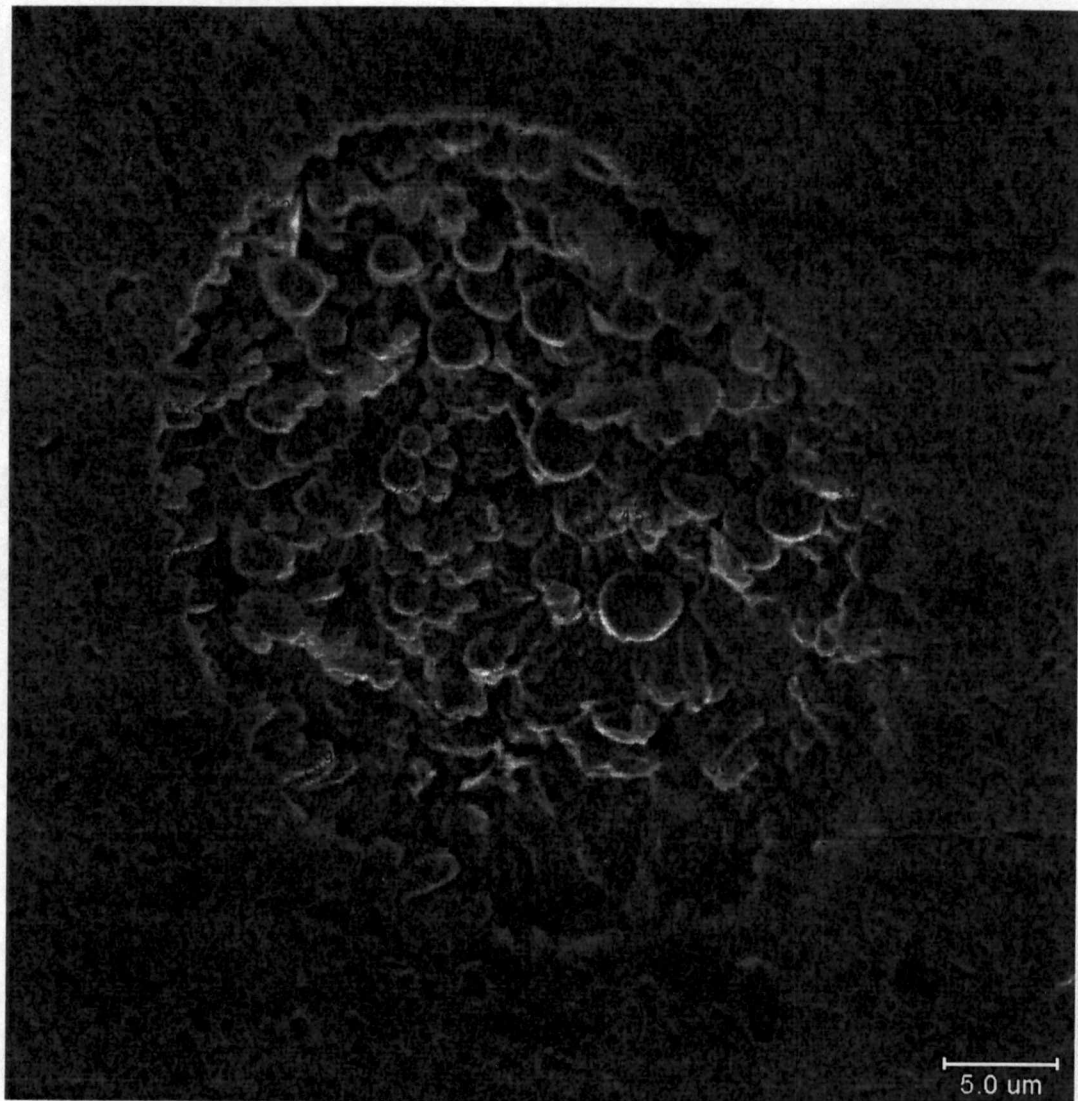


Figure IV-27. SEM micrograph of sample 3-5F showing the interior of a divorced spherical structure that includes the peripheral planar growth region. The Cu rich material was removed using a solution of FeCl_3 and water.

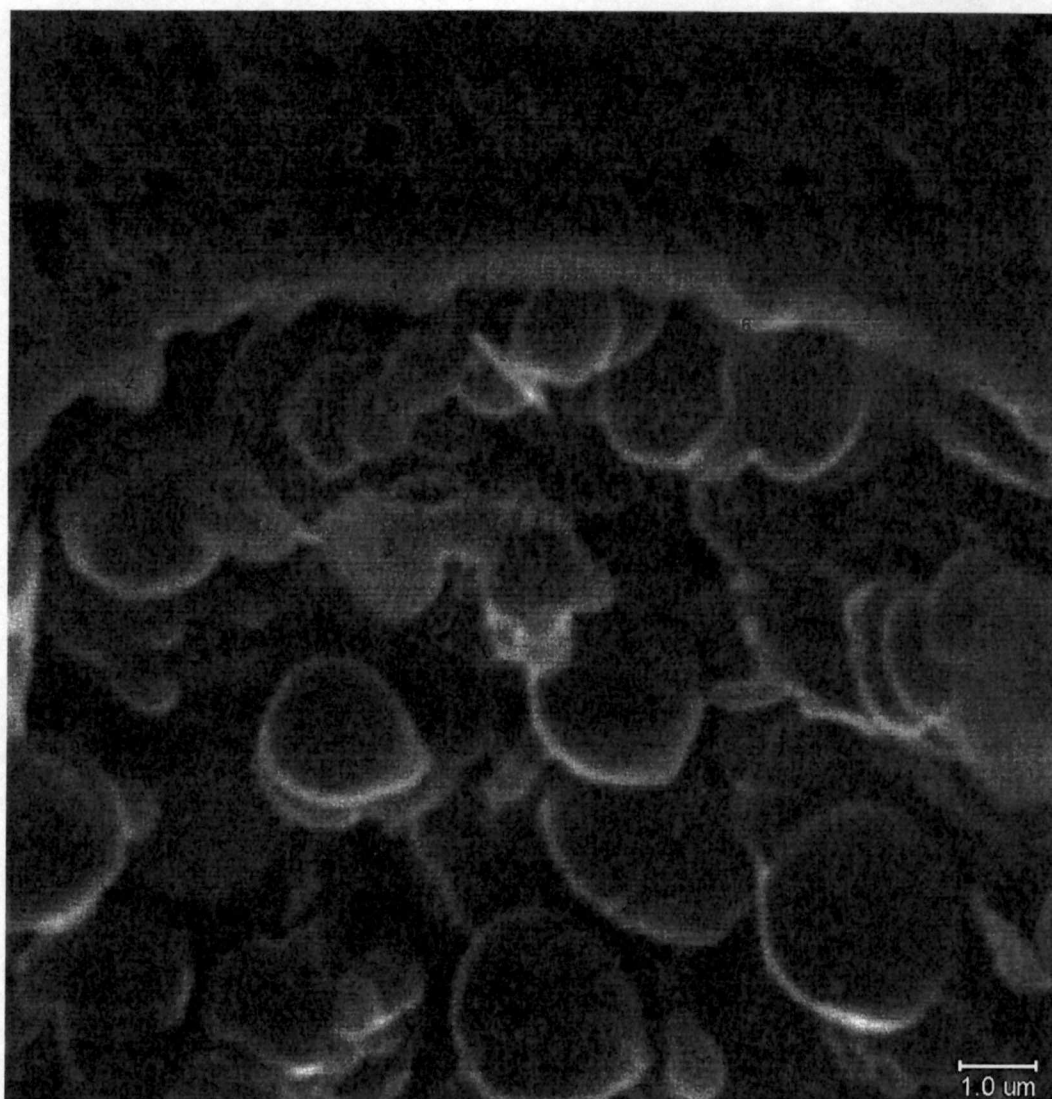


Figure IV-28. SEM micrograph of sample 3-5F showing slightly perturbed cellular growth.

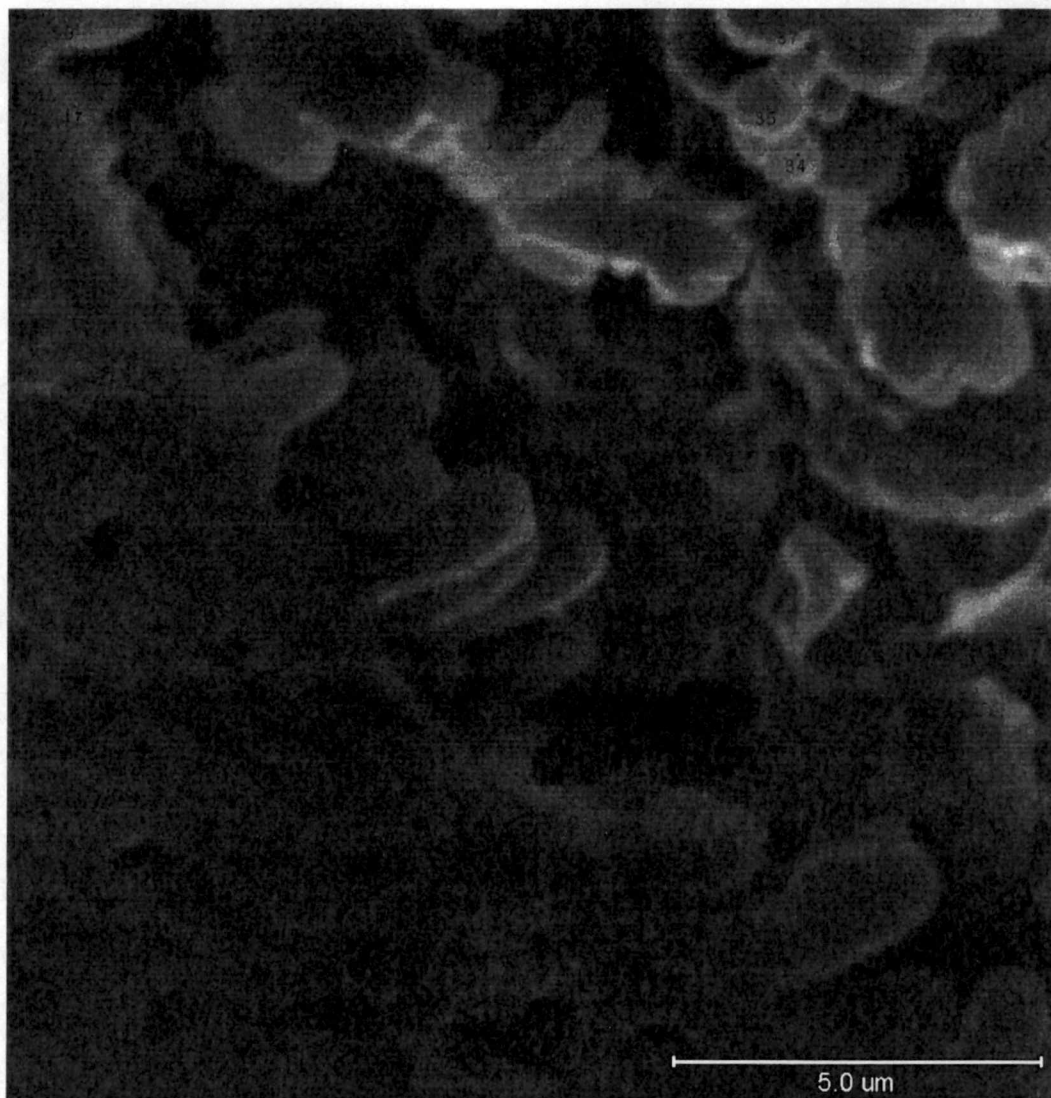


Figure IV-29. SEM micrograph of sample 3-5F showing parallel cellular growth at the edge of the planar growth region.

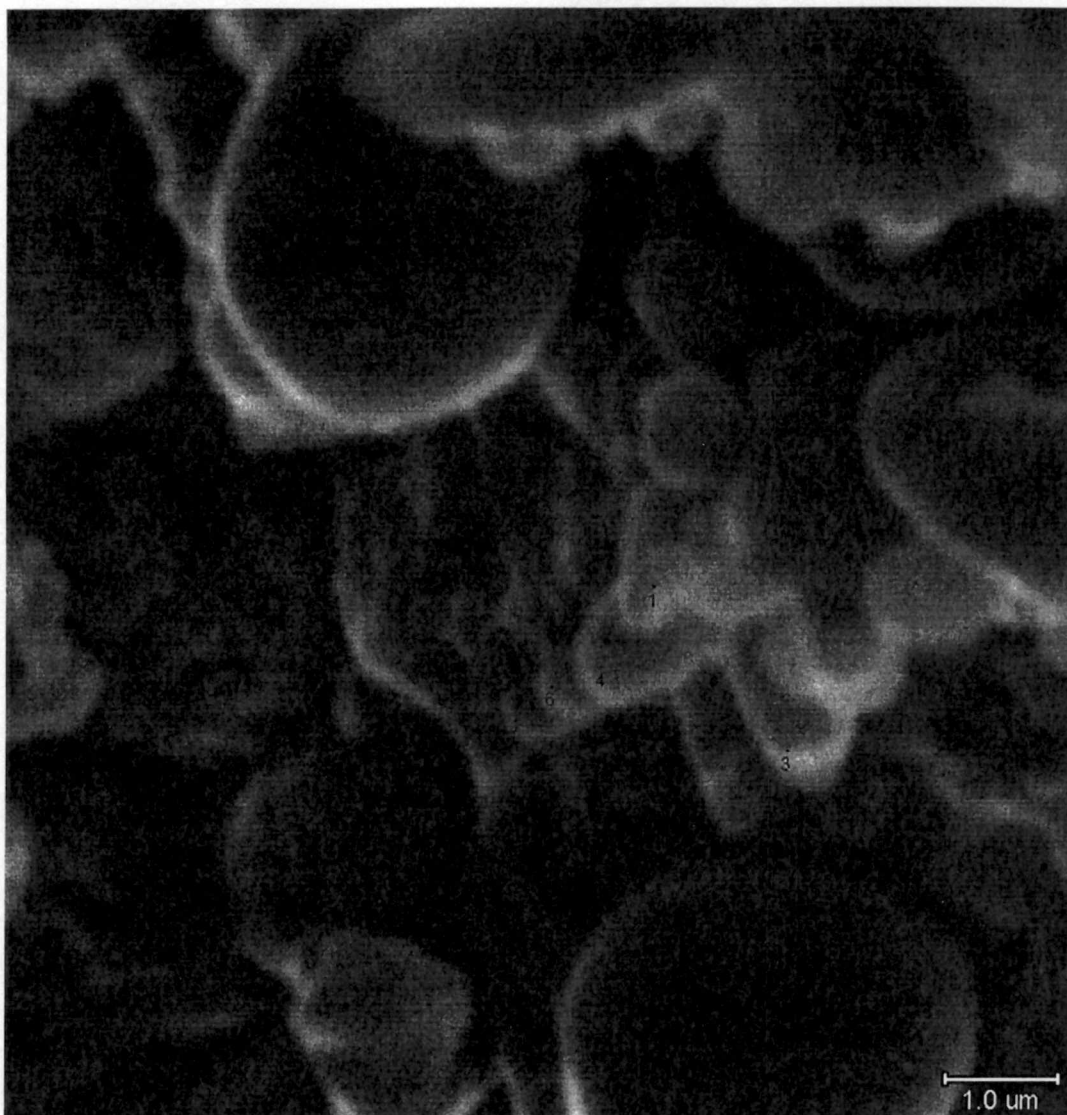


Figure IV-30. SEM micrograph of sample 3-5F showing dendrite.

APPENDIX C

DERIVATION OF THE CRITERION FOR THE CONSTITUTIONAL SUPERCOOLING THEORY (CST)

The derivation of the criterion for the CST begins with the conservation of solute at the S/L interface; a material balance across an infinitesimally thin plane of area "A." The material entering the plane due to motion less the material out of the plane due to both motion and diffusion must equal the accumulation of solute within the plane. That is,

$$A C_L^* R - A C_S^* R - (-A D_L \frac{dC_L}{dx} \big|_{x=0}) = (A \delta x) \frac{d\bar{C}}{dt}. \quad (C-1)$$

The assumption of an infinitesimally thin plan implies that

$$\delta x \approx 0 \quad (C-2)$$

The area term in equation C-1 can be factored out resulting in

$$\frac{dC_L}{dx} \big|_{x=0} = \frac{R}{D_L} (C_L^* - C_S^*) \quad (C-3)$$

From the phase diagram, the slope of the liquidus line is defined as

$$m_l = \frac{dT_L}{dC_L} \quad (C-4)$$

and the actual temperature gradient, G , in the liquid ahead of the S/L interface is defined as

$$G = \frac{dT_{actual}}{dx}. \quad (C-5)$$

The partition coefficient "k," the ratio of the composition of the solid at the interface to the composition of the liquid at the interface, is defined as

$$k = \frac{C_s^*}{C_L^*}. \quad (C-6)$$

Multiplying both sides of equation C-3 by equation C-4 results in

$$\left. \frac{dT_L}{dx} \right|_{x=0} = \frac{dT_L}{dC_L} \frac{dC_L}{dx} = -\frac{R m_L}{D_L} (C_L^* - C_s^*). \quad (C-7)$$

The CST criterion defined by Chalmers requires that the actual temperature gradient at the S/L interface, G , evaluated at $x=0$, must be greater than the equilibrium temperature gradient in order to maintain planar growth. Applying this criterion to equation C-7 yields

$$\frac{G}{R} \geq \frac{m_L}{D_L} (C_L^* - C_s^*). \quad (C-8)$$

During steady state conditions the solid composition at the S/L interface matches that of the bulk liquid, that is

$$C_s^* = C_o. \quad (C-9)$$

Substitution of the partition coefficient, k , in equation C-9 gives

$$C_L^* = \frac{C_o}{k}. \quad (C-10)$$

When equation C-10 is substituted into equation C-8 the result is

$$\frac{G}{R} \geq -\frac{m_L C_o}{k D_L} (1 - k). \quad (C-11)$$

If straight solidus and liquidus lines are assumed it can be shown geometrically that the equilibrium freezing range, ΔT , can be defined as

$$\Delta T = -m_L C_O (1/k - 1) . \quad (C-12)$$

Equation C-12 can now be substituted into equation C-11 to give a simpler form of the CST criterion,

$$\frac{G}{R} \geq \frac{\Delta T}{D_L} . \quad (C-13)$$

VITA

Kevin Tacy Zysk was born in Boston, Massachusetts on July 18, 1950. He was graduated from the Shaw Preparatory School in Boston, Massachusetts in August of 1969. He later attended Northeastern University and Lowell Technological Institute on a part-time basis while working full time.

After working in private industry for a number of years he entered the University of Tennessee at Chattanooga in August 1983. In May 1987, he received the degree of Bachelor of Science in Engineering.

He began study at the University of Tennessee Space Institute toward the degree of Master of Science in Mechanical Engineering in June 1987. This degree was awarded in May 1989. Upon graduation he accepted full time employment with the United States Air Force at the Arnold Engineering Development Center.

In January of 1990 he began part-time study for a doctorate degree while continuing to work for the Air Force. The Air Force graciously provided the funding for his study. He graduated with a Doctor of Philosophy in Engineering Science and Mechanics in December of 1998.