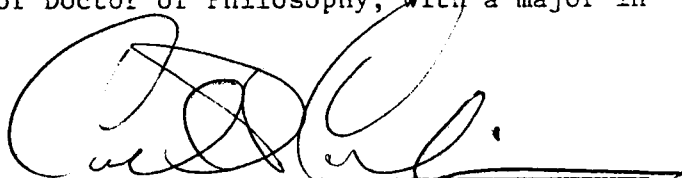


To the Graduate Council:

I am submitting herewith a dissertation written by Chang-Pin Dick Chou entitled "The Effects of Alloying Elements and Thermal History on Microstructure and Hot Cracking Resistance of Austenitic Stainless Steel Weld Metals." I recommend that it be accepted in partial fulfillment of the requirements for the degree of Doctor of Philosophy, with a major in Metallurgical Engineering.



Carl D. Lundin, Major Professor

We have read this dissertation
and recommend its acceptance:

E. E. Stansbury
Gen J. Clever
A. Mathews
C. R. Brooks

Accepted for the Council:



Vice Chancellor
Graduate Studies and Research

Thesis
B3b
.C468
cop.2

THE EFFECTS OF ALLOYING ELEMENTS AND THERMAL HISTORY
ON MICROSTRUCTURE AND HOT CRACKING RESISTANCE OF
AUSTENITIC STAINLESS STEEL WELD METALS

A Dissertation

Presented for the

Doctor of Philosophy

Degree

The University of Tennessee, Knoxville

Chang-Pin Dick Chou

March 1983

3065993

ACKNOWLEDGEMENTS

The author wishes to express his sincere gratitude to his major professor, Dr. Carl D. Lundin, for his continued encouragement, advice, patience, and understanding during the course of this investigation. Appreciation is expressed to Dr. E. E. Stansbury, Dr. B. F. Oliver, Dr. C. R. Brooks, and Dr. A. Mathews for their kind guidance and help during this work.

The author wishes to thank Mr. H. B. Thompson, Mr. Ray Bellomy, Mr. Mike Neal, and Mr. Bob McGill for their contributions to the success of this work.

The author would like to express his appreciation to the support of this research by the High Alloys Committee of the Welding Research Council in this investigation.

The author wishes to express his deepest appreciation and thanks to his wife, Pin-Ying, for her patience, understanding, and many hours of typing.

ABSTRACT

The primary purpose of the present research was to investigate the influence of the alloying elements and multiple thermal cycling on the hot cracking resistance of austenitic stainless steel weld metals. In addition, the effect of composition and multiple thermal cycling on the hot ductility was also determined and correlated with the hot cracking tendency. To more clearly define the variables, the following experimental methods were developed and experiments conducted : (1) a method to develop laboratory-made alloys with variations in composition, (2) a method to evaluate the hot ductility and the hot cracking behavior of commercial weld metals in addition to experimental compositions, and (3) a test to separate the influence of strain and thermal history on the hot cracking tendency of the alloy.

Both the Fissure Bend Test and the Gleeble Hot Cracking Test were utilized to evaluate the hot cracking resistance of austenitic weld metals. Optical metallography, SEM fractography and energy dispersive X-ray analysis were employed to define the cause and mechanism of hot cracking in the austenitic weld metals.

It was found that fissures occur primarily along grain boundaries in the HAZ from the previous weld deposit, and fissuring is enhanced by multiple thermal cycling in the HAZ. Increasing Mn and Mo content and reducing Si content can reduce the hot cracking tendency of fully austenitic stainless steel weld metals. Liquation of the Si-segregated grain boundary regions produced during thermal cycling is a dominant factor which leads to an enhanced fissuring propensity.

A theory of "Reheat Hot Cracking" is proposed for both ferrite-containing and fully austenitic welds. A region, called the "Hazard HAZ", exists near the fusion zone of subsequent bead upon multipass or repair welding. This "Hazard HAZ" region exhibits a relatively low ductility and fissures occur in this region upon welding under conditions of sufficient restraint. The cause of the low ductility in the "Hazard HAZ" is the embrittlement of austenite grain boundaries which are enriched with "harmful" elements such as Si, P, and S. The extent of the "Hazard HAZ" region depends primarily on the compositions of weld metal as well as the Cr/Ni ratio and Mn plus Mo level.

TABLE OF CONTENTS

CHAPTER	PAGE
I. INTRODUCTION	1
II. LITERATURE REVIEW	5
1. Properties of Austenitic Stainless Steels.	5
2. Weldability of Austenitic Stainless Steels	13
3. Solidification Behavior of Austenitic Stainless Steel Weld Metals.	17
4. Mechanism of Hot Cracking in Austenitic Stainless Steel Weld Metal	25
5. Effect of Multiple HAZ Thermal Cycle on Fissuring Tendency	30
6. Role of Ferrite in Austenitic Stainless Steel Weld Metals.	33
7. Factors Controlling the Hot Cracking Tendency of Fully Austenitic Stainless Steel Weld Metal.	39
8. Evaluation of Hot Cracking and Hot Ductility of Austenitic Stainless Steel Weld Metals	42
9. Effect of Weld Discontinuities on the Behavior of Austenitic Stainless Steel Weld Metals	52
10. Closure	73
III. EXPERIMENTAL PROCEDURE	75
1. Materials.	75
2. Welding Equipment and Fixtures	77
3. New Alloy Development.	81
4. Ferrite Measurement.	81
5. Fissure Bend Test.	84
6. Gleeble Hot Ductility Test	85
7. Gleeble Hot Cracking Test.	89
8. Optical Metallography.	90
9. Scanning Electron Microscopy	90
IV. RESULTS AND DISCUSSION	92
1. Introduction	92
2. Effect of Nitrogen on the Microstructure and Ferrite Content of Austenitic Stainless Steel Weld Metals.	98
3. Development of Techniques to Produce Modified Alloy Compositions	111
4. Hot Cracking Sensitivity Evaluation Using Fissure Bend Test.	126
5. Hot Ductility Behavior and Effect of Multiple Thermal Cycles on Ferrite Content.	166
6. Hot Cracking Sensitivity Evaluation Using the Gleeble Technique.	225
7. Theory of Reheat Hot Cracking.	282

CHAPTER	PAGE
V. CONCLUSIONS.	287
LIST OF REFERENCES	292
VITA	304

LIST OF TABLES

TABLE	PAGE
1. Tensile Properties and Transverse Charpy V-Notch Impact Strength of Some Austenitic Stainless Steels.	11
2. Ferrite Number Needed for Fissure Elimination	34
3. Characterization of Weld Discontinuities.	53
4. Chemical Composition and Measured Ferrite Number of Austenitic Stainless Weld Metals Used in Fissure Bend Test Evaluation.	76
5. Chemical Composition of Austenitic Stainless Steels Used in Studying the Nitrogen Effect	78
6. Chemical Composition and Measured Ferrite Number of Austenitic Stainless Steel Weld Metals Used in the Gleeble Hot Cracking Test Evaluation.	79
7. Welding Parameters for Fissure Bend Test.	80
8. Welding Parameters for Autogenous GTA Welding	100
9. Ferrite Number in Austenitic Stainless Steel Weld Metals S1 and S6	105
10. Effect of Nitrogen in Shielding Gas on GTA Autogenous Welds of Austenitic Stainless Steel Sheet	106
11. Calculation of Chemistry of Mixed Weld Deposit.	114
12. Predicted and Actual Composition of Weld Metals Produced by Dual-Wire Feeder Technique	119
13. Fissure Bend Test Results and Measured Ferrite Number of Austenitic Stainless Steel Weld Metals	129
14. Welding Conditions of Fully Austenitic Stainless Steel Weld (Consumable W1).	136
15. Results of the Fissure Bend Test on Fully Austenitic Stainless Steel Weld Metal W1	137
16. Chemical Analysis on the W1 Bare Wire, GTAW Deposited Weld Metal, and GMAW Deposited Weld Metal	138

TABLE	PAGE
17. Fissure Bend Test Results of GTA Remelting of Previously SMA Deposited Weld Metals.	141
18. Welding Parameters for As-Deposited Weld Metal and GTA Remelting.	149
19. Effect of Weaving on Fissure Bend Test Results	164
20. Chemical Composition and Measured Ferrite Number of Austenitic Stainless Steel Welds Used for Gleeble Hot Ductility Testing.	169
21. Three Groups of Weld Metal Used for Evaluating the Effect of Specific Alloying Elements With the Gleeble Hot Cracking Test.	257
22. Gleeble Hot Cracking Test Parameters for Evaluation of the Effect of Specific Alloying Elements on Hot Cracking Sensitivity	258

LIST OF FIGURES

FIGURE	PAGE
1. Vertical Section of Very Low Carbon Fe-Cr-Ni Alloys at 18% Chromium.	6
2. Effect of Temperature on (a) Linear Thermal Expansion and (b) Thermal Conductivity of the Three Main Grades of Stainless Steels.	8
3. Hot Strength Characteristics of Stainless Steels.	9
4. Results of Charpy Keyhole Impact Tests on Types 302, 304, and 304L in the Annealed and Annealed-and-Sensitized Conditions.	12
5. Effect of Growth Rate and Temperature Gradient on Constitutional Supercooling	19
6. Schematic Representation of the Influence of the Parameters C_0 , K_0 , D , G , R on the Pattern of Solidification.	20
7. Vertical Section (at 70 wt.% Fe) of Ternary Equilibrium System Fe-Cr-Ni	23
8. Effect of Constitutional Features on Cracking Susceptibility in Binary Systems	26
9. Fissure Bend Test Fixture	44
10. Schematic Diagram of the Gleeble Apparatus.	46
11. Illustration of the Development of Hot-Ductility-Test Curves.	48
12. Schematic Representation of Gleeble Hot Cracking Testing Sequence Used by Krantz	51
13. The Effect of Discontinuity Size on the Tensile Strength of Weld Metals	58
14. The Dependence of the Impact Toughness of the Weld in Steel 03Kh20N16AG6 on the Discontinuity Area and Notch Radius . . .	63
15. Fatigue Design Curve of Stainless Steel, ASME III, 1974 . . .	67
16. The Fatigue Life of Welded Joints of 12Kh18N10T with Discontinuities	69

FIGURE	PAGE
17. Recommended Clamping Fixture for Fissure Bend Test.	82
18. Weld Fixture for Welding Thin Sheet Materials	83
19. Schematic Representation of Hot Ductility Testing Sequence. .	86
20. The Geometry and the Temperature Distribution for Unwelded and Transversely Welded Test Samples.	87
21. Dimensions and Temperature Distribution in the Longitudinal Composite Weld Specimen	88
22. Measured Ferrite Number vs % N_2 in Shielding Gas for Austenitic Stainless Steel Autogenous GTA Welds S1 and S6	101
23. Nitrogen Content in Weld Metal vs % N_2 in Shielding Gas for Austenitic Stainless Autogenous GTA Welds	102
24. Microstructures of Autogenous GTA Welds of Austenitic Stainless Steel S6 (308L)	103
25. Microstructures of Autogenous GTA Welds of Austenitic Stainless Steel S4 (304).	107
26. Measured Ferrite Number vs % N_2 in Shielding Gas for Austenitic Stainless Autogenous GTA Welds	109
27. Effect of Nitrogen on the Microstructure of Autogenously Fused Austenitic Stainless Steel P2 (304LN)	110
28. Schematic Diagram of Dual-Wire Feeder	113
29. Welding Fixture Used with Dual-Wire Feeder Technique.	115
30. Welding Set Up Showing Torch, Fixture, and Dual-Wire Nozzle.	116
31. Microstructures of Austenitic Stainless Steel Weld Metals B1, B2, and D3.	117
32. Strip Appearance After Welding, Grinding, and Rolling/ Annealing	120
33. Microstructures of New Alloy D2 Manufactured by the Dual-Wire Feeder Technique.	121
34. Steps in Preparation of Chemistry-Modified Specimen with Implant Technique	123

FIGURE	PAGE
35. Microstructures of Weld Metal B3 (317LM) and Its Modified Alloys.	125
36. Microstructures of Fully Austenitic Stainless Steel Weld Metal	131
37. Microstructures of As-Deposited Fully Austenitic Stainless Steel Welds.	132
38. Calculated Weld Metal Compositions on Delong Diagram	134
39. Microstructures of Weld Metal W1, Transverse Section	140
40. Microstructures of Fully Austenitic Stainless Steel Weld Metal (E4).	142
41. Surface of SMA Deposited Weld Metal.	144
42. EDAX of Spot Weld of Fully Austenitic Stainless Steel Weld Metal E4.	145
43. Number of Hot Cracks as a Function of Energy Input for GTAW and GMAW Processes (Weld Metal W1).	146
44. Typical Weld Pad Cross-Sections Showing Bead Sequence for Multiple HAZ Study	148
45. Fissure Count as a Function of Multiple HAZ Thermal Cycles .	151
46. Microstructures of GTA Remelted Fully Austenitic Stainless Steel Weld Metal	152
47. Effect of Multiple Thermal Cycles on Austenitic Stainless Steel Weld Metal	153
48. Fracture Surface of W1 Weldment.	155
49. Fracture Surface of Deposit Region from W1 Weldment.	156
50. Fracture Surface of Multiple HAZ Exposure Region from W1 Weldment.	157
51. Fracture Surface of Austenitic Stainless Steel Weld Metal (W1) HAZ	158
52. EDAX of Fissure of W1 Weld Metal HAZ	160

FIGURE	PAGE
53. Weaving Pattern and Welding Parameters Used in GTA Weave Bead.	161
54. Weaving Pattern and Welding Parameters Used in SMA Weave Bead.	162
55. Fissures Resulting From Weaving Techniques of GTAW Remelted W1 Consumable.	165
56. Hot Ductility Test Results for Unwelded Material "A1"	170
57. Hot Ductility Test Results for Unwelded Material "A2"	171
58. Hot Ductility Tested Specimens of Unwelded Material "A2". . .	172
59. Hot Ductility Test Results for Welded Material "A1" (Transverse Composite Sample)	173
60. Hot Ductility Test Results for Welded Material "A2" (Transverse Composite Sample)	175
61. Effect of Peak Temperature and Number of Thermal Cycles on the Hot Ductility of Sample "A1"	177
62. Effect of Peak Temperature and Number of Thermal Cycles on the Hot Ductility of Sample "L".	178
63. Effect of Peak Temperature and Number of Thermal Cycles on the Hot Ductility of Sample "H".	179
64. Hot Ductility Specimens of Longitudinal Composite Welded Material "A1"	181
65. The Outer Surface of Hot Ductility Tested Specimen of Unwelded "A1" Tested On-Heating to 2300°F	183
66. Hot Ductility Tested Specimen of Unwelded "A1" Tested On-Heating to 2300°F.	184
67. The Outer Surface of Unwelded "A1" Tested On-Cooling to 2100°F After Heating to 2400°F	185
68. Outer Surface of Unwelded "A1" Tested On-Cooling to 2100°F After Heating to 2400°F	186
69. The Fracture Surface of Unwelded "A1" Tested On-Cooling to 2100°F After Heating to 2400°F	187

FIGURE	PAGE
70. The Fracture Surface of Unwelded "A1" Tested On-Heating. . .	189
71. The Outer Surface of Weld Metal of a Longitudinal Composite Welded "A1" Tested On-Heating to 2100°F After Five Thermal Cycles with Peak Temperature of 2100°F.	190
72. The Fracture Surface of a Longitudinal Composite Welded "A1" Tested On-Heating to 2100°F After Five Thermal Cycles with Peak Temperature of 2100°F	191
73. EDAX of a Longitudinal Composite Welded "A1" Tested On-Heating to 2100°F After Five Thermal Cycles with Peak Temperature of 2100°F.	193
74. The Outer Surface of the Longitudinal Composite Weld "A1" Tested On-Heating to 2300°F.	194
75. Longitudinal Welded Composite "A1" Tested On-Heating to 2300°F	195
76. The Fracture Surface of a Longitudinal Composite Welded Specimen "A1" Tested On-Heating to 2300°F.	196
77. A Longitudinal Composite Welded Specimen "A1" Tested On-Heating to 2300°F After Five Thermal Cycles with Peak Temperature of 2300°F	197
78. The Fracture Surface of Material "H" Fractured at a Peak Temperature of 2300°F After Five Thermal Cycles to 2300°F	199
79. Ferrite Number Distribution Along the Gage Length of Sample "H" After Five Thermal Cycles to a Peak Temperature of 2100°F	201
80. Ferrite Number Distribution Along the Gage Length of Sample "H" After Five Thermal Cycles to a Peak Temperature of 2400°F	202
81. Effect of Peak Temperature on the Ferrite Number of Austenitic Stainless Steel Weld Metal "H".	204
82. Effect of Peak Temperature on the Ferrite Number of Austenitic Stainless Steel Weld Metal "L".	205
83. Microstructures of Specimen "L" Thermal Cycled Five Times to a Peak Temperature of; (a) 1800°F, (b) 2100°F, and (c) 2400°F	206

FIGURE	PAGE
84. Fe-Cr-Ni Pseudo-Binary Phase Diagrams for 70% Fe and 65% Fe.	207
85. Effect of the Number of Thermal Cycles on the Ferrite Number of the Weld in Alloy "H".	208
86. Microstructures of Alloy "H" Thermal Cycled Five Times to a Peak Temperature of 2300°F.	210
87. Effect of the Number of Thermal Cycles on the Ferrite Number of the Weld in Alloy "L".	211
88. Microstructures of Alloy "L" Thermal Cycled Ten Times to a Peak Temperature of 2400°F.	212
89. EDAX of Ferrite and Austenite in Specimen "L" Thermal Cycled Five Times to a Peak Temperature of 2400°F.	214
90. Schematic Diagram Showing the Helium-Blast Cooling Technique.	215
91. Effect of Peak Temperature and Cooling Rate on the Ferrite Number of Austenitic Stainless Steel Weld Metal "L".	216
92. Effect of Peak Temperature and Cooling Rate on the Ferrite Number of Austenitic Stainless Steel Weld Metal "H".	218
93. Schematic Diagram Showing the "Hazard HAZ" Region	219
94. Generalized Pseudo-Binary Diagram Showing Relative Positions of Low Ferrite Alloy and High Ferrite Alloy	222
95. Effect of Reheat Temperature on the Fraction of Ferrite Remaining in Austenitic Stainless Steel Weld Metals	224
96. Schematic Representation of Testing Sequence Used in Gleeble Hot Cracking Test	227
97. Effect of the Magnitude of Extension Strain on the Fissuring Sensitivity of Weld Metal "Al".	229
98. Specimens of "Al" Thermal Cycled to a Peak Temperature of 2300°F with Total Extension Strain of; (a) 10.7% and (b) 16.0%	231
99. Effect of the Peak Temperature on the Fissuring Sensitivity of Weld Metal "Al".	232

FIGURE	PAGE
100. Interrelation of Strain and Temperature for the Three Methods Used in the Gleeble Hot Cracking Test	234
101. Schematic Thermal Cycle Used in the Gleeble Hot Cracking Test.	236
102. Effect of the Peak Temperature on the Fissuring Sensitivity of Weld Metal "A2".	238
103. Microstructure of As-Welded "A2".	239
104. Microstructures of Fully Austenitic Stainless Steel Weld Metal "A1".	240
105. Fissure Surface of "A1" Thermal Cycled to 2300°F with 10.7% Total Extension Strain Applied.	242
106. Fissure Surface of "A1" Thermal Cycled to 2400°F with 10.7% Total Extension Strain Applied.	243
107. EDAX Results of Fissure Surface of "A1" Thermal Cycled to 2400°F with 10.7% Total Extension Strain Applied.	244
108. Effect of the Number of Cycles, Accumulated Strain, and Peak Temperature on the Fissuring Sensitivity of Weld Metal "A1"	246
109. Effect of Prior Thermal Cycles on the Fissuring Sensitivity of "A1"	248
110. Microstructures of Fully Austenitic Stainless "A1".	249
111. Microstructures of Weld Metal "A1" Thermal Cycled Twice to 2100°F Peak Temperature.	251
112. Microstructures of Weld Metal "A1" Thermal Cycled Twice to Peak Temperatures of; (a) 2200°F, (b) 2300°F, and (c) 2400°F.	252
113. Energy Dispersive X-Ray Analysis of an As-Welded Specimen of "A1"	253
114. Energy Dispersive X-Ray Analysis of a Specimen of "A1" Thermal Cycled Six Times to 2100°F.	254
115. Gleeble Hot Cracking Test Results of Weld Metals in Group I	259

FIGURE	PAGE
116. Gleeble Hot Cracking Test Results of Weld Metals in Group II.	261
117. Gleeble Hot Cracking Test Results of Weld Metals in Group III	263
118. General Cracking Trends of Austenitic Stainless Steel Weld Metals	264
119. Fe-Cr-Ni Ternary Diagram with Hypothetical Alloys - Showing the Effect of Cr/Ni Ratio and Total Cr + Ni Content	265
120. Effect of Alloying Elements on the Hot Cracking Behavior of Austenitic Stainless Steel Weld Metals	267
121. Microstructure of Weld Metal "D3" Tested to 2300°F with Three Thermal Cycles.	269
122. Microstructures of Weld Metals "B3" and "B3Ni" Tested to 2300°F with Three Thermal Cycles	270
123. Schematic Representation of Specimen Preparation for Fractography.	272
124. Fracture Surface of Weld Metal "B1" Tested to 2300°F with Three Thermal Cycles.	273
125. Energy Dispersive X-Ray Analysis Results of the Fractured Surface of "B1"	274
126. Fractured Surface of Weld Metal "AlSi" Tested to 2300°F with Three Thermal Cycles and Their EDAX Results.	275
127. General View of Fractured Surface of Weld Metal "B3Ni" Tested to 2300°F with Three Thermal Cycles.	277
128. High Magnification of Areas A, B, and C in Figure 127	278
129. EDAX Results of Fractured Surface of Weld Metal "B3Ni" Tested to 2300°F with Three Thermal Cycles.	279

NOMENCLATURE

FN = Ferrite Number

HAZ = heat-affected zone

SMAW = shielded metal-arc welding

GTAW = gas tungsten-arc welding

GMAW = gas metal-arc welding

OLM = optical light microscopy

SEM = scanning electron microscopy

EDAX = energy dispersive x-ray analysis

L = liquid

γ = gamma

δ = delta

Fe = iron

Cr = chromium

Ni = nickel

Mn = manganese

Mo = molybdenum

Si = silicon

P = phosphorus

S = sulphur

N = nitrogen

CHAPTER I

INTRODUCTION

Recent interest in the high-temperature properties of austenitic stainless steel weld metals is in large measure due to the increased utilization of nuclear power (1). In such advanced applications the quality of the welds is of paramount importance. Fully austenitic stainless steel weld metal often exhibits hot cracking and microfissuring, especially when applied under conditions of high weld joint restraint (2). In order to limit the extent of microfissuring in austenitic stainless steel weld metal, it has become common practice to control the chemical composition and welding process so as to provide a small quantity of the ferrite phase in the as-deposited weld metal (3,4,5). However, for some corrosive environments, the presence of ferrite can be detrimental as it may be attacked preferentially (6). This could result in failure, particularly if sufficient ferrite were present to form a continuous network. Also, in a duplex structure, if working conditions involve temperatures between 550°C and 850°C, the formation of sigma phase from the ferrite may lead to embrittlement when the structure is at ambient temperatures (7). At cryogenic temperatures it has been found that the ferrite content in weld metal deposit adversely affects toughness (8). Thus, for critical applications, fabricators are frequently required to use fully austenitic stainless steel weld metal.

Although austenitic stainless steels are generally regarded as readily weldable, with considerable tolerance for variations in welding conditions without risk of cracking, fully austenitic weld metals are more prone toward cracking during deposition of a single pass and in underlying weld runs reheated by subsequent passes. It is known that the composition of the consumables as well as the welding procedure has a significant effect on the hot cracking tendency of fully austenitic stainless steel weld metal.

A need exists for a more basic understanding of the relationship between alloy content and the hot cracking tendency of fully austenitic stainless steel weld metals. It is the object of this research to study both the practical and fundamental aspects of hot cracking in the fully austenitic stainless steels. The practical aspects involve the use of minimum alloy content for economical reasons and the fundamental aspects involve the understanding of the role of specific alloying elements as to their effect on solidification and hot cracking tendency.

It has been long recognized that reheating of the material to temperatures in the vicinity of the solidus increases the fissuring tendency. However, no quantitative evaluation of the effect of multiple thermal cycling on fissuring tendency has been investigated until recently by Lundin and Spond (9). With the increasing use of heavy section weldments, repair after fabrication and repair in service become more common. Thus, the understanding of the effect of multiple thermal cycles on weld metal fissuring becomes more important and draws more concerted attention.

It is recognized that the presence of ferrite in the room temperature microstructure indicates a reduced fissuring tendency for austenitic stainless steel weld metals. The ferrite level needed to virtually eliminate fissuring is dependent upon the composition of weld metal as determined by Lundin et al. (10). More recently Masumoto et al. (11) proposed that the determination of weld metal fissuring should be based on composition alone and not on ferrite content. Using the Fe-Cr-Ni phase diagram and weld metal of different compositions, they claim that fissuring depends on whether austenite or delta ferrite is the primary phase solidifying from the liquid. The alloys whose compositions nucleate austenite as the primary phase have a high sensitivity to fissuring. However, for the weld metals that nucleate delta ferrite as the primary phase, the sensitivity to fissuring is low even with high sulphur contents (≈ 0.15 percent). Masumoto believes that when the delta ferrite nucleates as the primary phase, it is able to dissolve larger amounts of sulphur and other fissuring-inducing elements thus preventing the segregation of these elements to the grain boundaries where they are most influential in causing fissuring.

It has been confirmed that manganese has beneficial effects in reducing hot cracking (12,13) and that the cracking susceptibility increases as the Si, P, and S content increases (14,15,16) in the fully austenitic stainless steel weld metals. The role played by Mn and Mo in hot cracking is not fully understood because of the lack of fundamental investigations and researches. With the improved technology in

the refining process (14), low P and S content can be obtained and hot cracking resistance improved. It is considered that silicon contents above 0.5% in 25/20 alloys cause a significant increase in cracking (17-21), not only because the liquidus-solidus temperature interval is widened, but also because intergranular silicate films may form thus embrittling the grain boundaries. To some extent, the effect of silicon can be compensated for by increasing the manganese content.

A primary purpose of the present research was to investigate the influence of the alloying elements and multiple thermal cycling on the hot cracking resistance of austenitic stainless steel weld metals. In addition, the effect of composition and multiple thermal cycling on the hot ductility were also evaluated to determine and correlate with the hot cracking tendency. To more clearly define the variables, the following experimental methods were developed and experiments conducted; (1) a method to develop laboratory-made alloys with variations in composition, (2) a method to evaluate the hot ductility and the hot cracking behavior of commercial weld metals in addition to experimental compositions and (3) a test to separate the influence of strain and thermal history on the hot cracking tendency of the alloy.

CHAPTER II

LITERATURE REVIEW

1. Properties of Austenitic Stainless Steels

The austenitic stainless steels are alloys derived from the Fe-Cr-Ni system. Their chromium content is generally above 16%, and they contain a sufficient quantity of nickel for the austenitic structure to be stable at room temperature, or to be retained by fast cooling from elevated temperatures. Figure 1 (22) represents a schematic section of a constitutional diagram for Fe-Cr-Ni alloys containing 18% Cr and less than 0.03% carbon. Above 3.5% and up to 10% Ni, the alloy solidifies as delta ferrite, or at least with a very high proportion of this phase. And there is a temperature range in which the alloy is purely austenitic, and which widens as the nickel content becomes higher. With increasing nickel, the M_s temperature is lowered, but as long as the nickel content does not exceed about 7-8%, the M_s of the austenite remains above room temperature; the steel is then of a martensitic type. An increase in the proportion of chromium extends the temperature range over which delta ferrite exists, and consequently requires an increase in nickel content to obtain a fully austenitic structure below a specific temperature.

The austenitic stainless steels can not be hardened by heat treatment but can be strengthened by cold work, and thus they exhibit a wide range of mechanical properties (23). At room temperature, austenitic stainless steels exhibit yield strengths between 30 and 200 ksi, depending on composition and amount of cold work. They also exhibit good ductility and toughness even at high strengths, and this good ductility and toughness is retained to cryogenic temperatures (24).

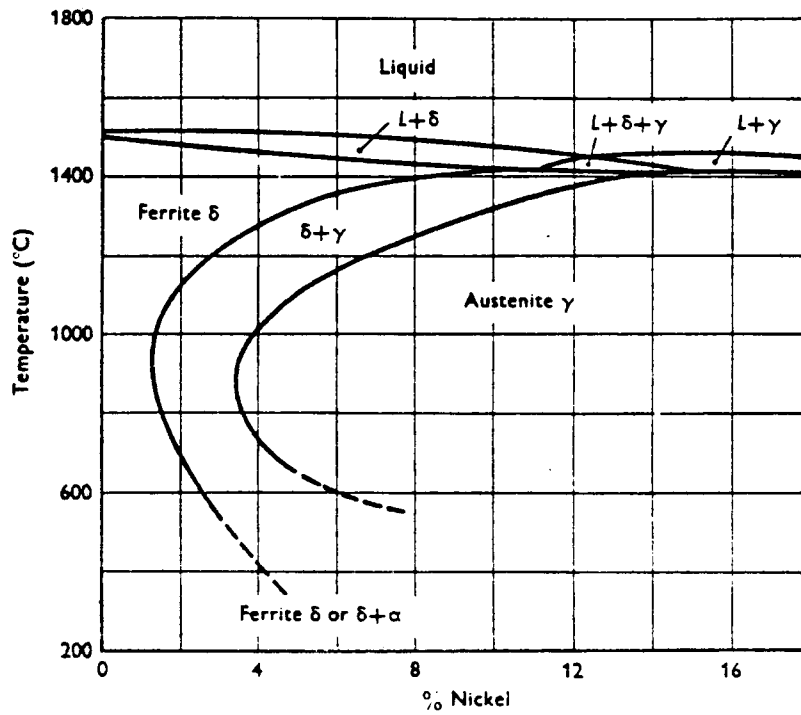


Figure 1. Vertical Section of Very Low Carbon Fe-Cr-Ni Alloys at 18% Chromium.

Because the austenitic stainless steels are often used for structural applications, particularly when cold worked, the toughness and fatigue strength of these steels are of interest. At room temperature in the annealed condition, the austenitic stainless steels exhibit exceptional toughness. Charpy V-notched energy absorption values in excess of 100 ft-lb are generally obtained. The fatigue or endurance limits of austenitic stainless steels are about one-half their tensile strength and do not differ significantly from one another (25).

The physical properties of austenitic stainless steels of most interest are linear expansion and thermal conductivity. Figure 2a (26) shows austenitic stainless steels to have greater thermal expansion than martensitic or ferritic grades. This should be considered when joining dissimilar metals. Lower thermal conductivity of austenitic grades is shown in Figure 2b (26). As a result of their low thermal conductivity and high thermal expansion, the austenitic stainless steels of the 300 series are more sensitive to thermal fatigue than steels of the 400 series.

Physical properties such as modulus of elasticity and density are also of interest. These properties are similar to those of the martensitic and ferritic stainless steels. A modulus of 28×10^6 psi and a density of 0.29 lb/in^3 can be considered typical of annealed austenitic stainless steels (25). Cold working reduces these values slightly.

Austenitic stainless steels are used at temperatures up to about 2000°F because they have good strength at elevated temperature and good resistance to corrosion and oxidation. Figure 3 (25) gives a broad

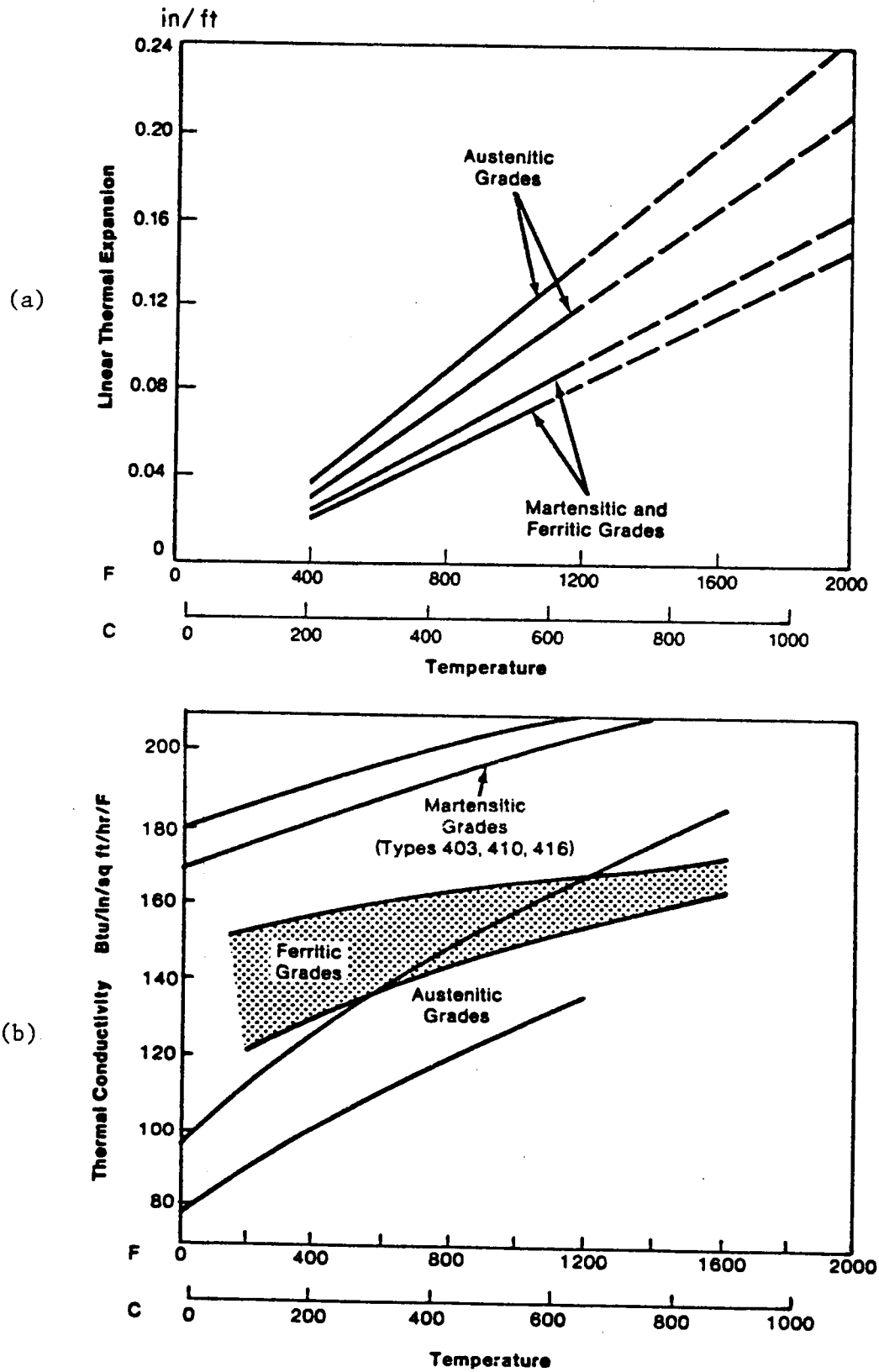


Figure 2. Effect of Temperature on (a) Linear Thermal Expansion and (b) Thermal Conductivity of the Three Main Grades of Stainless Steels.

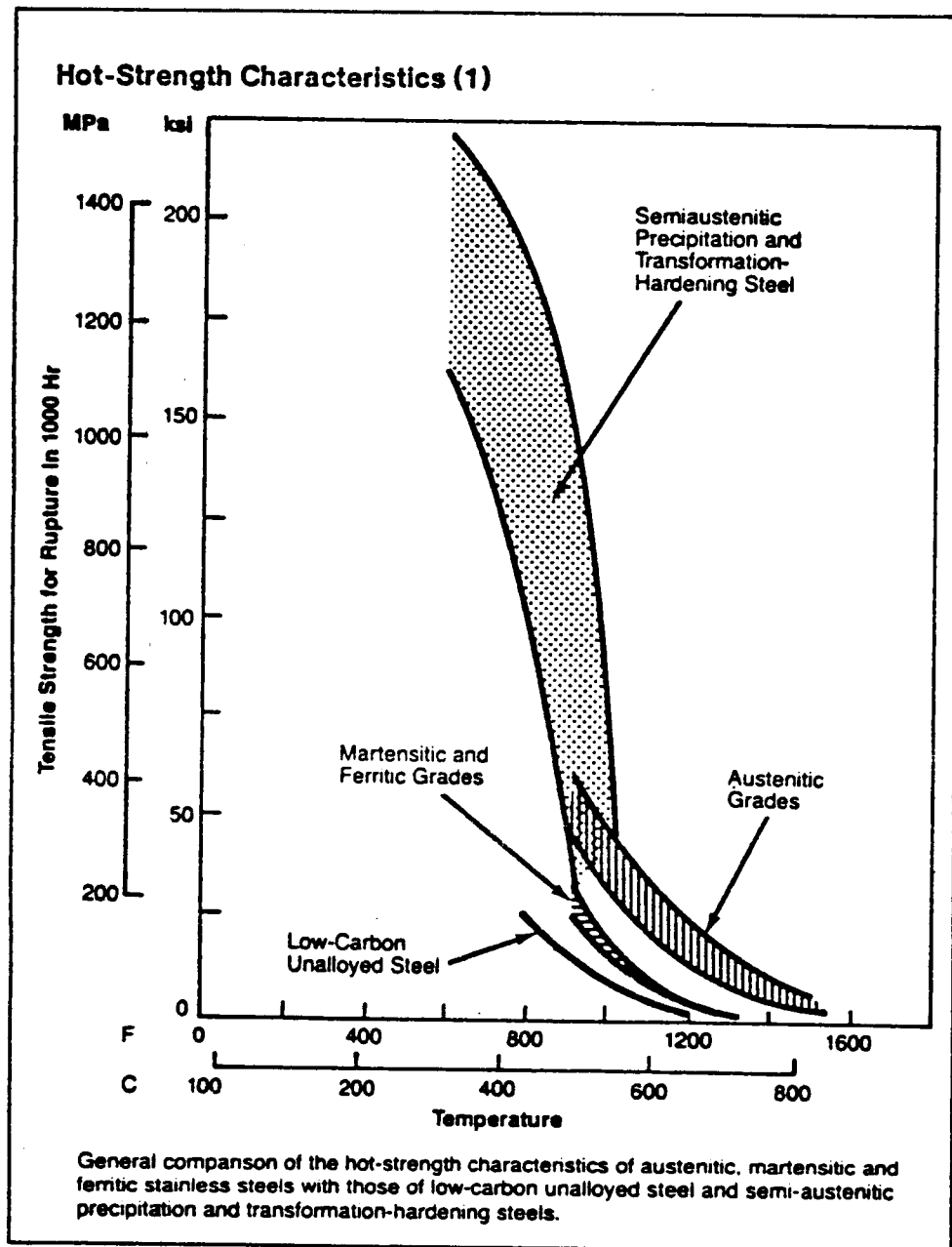


Figure 3. Hot Strength Characteristics of Stainless Steels.

concept of the hot-strength advantages of austenitic stainless steels in comparison to low-carbon unalloyed steel and other grades of stainless steels.

For the austenitic stainless steels exposure to temperatures from 800° to 1600°F can result in carbide precipitation which can cause a loss in toughness and except for the stabilized grades can lead to intergranular corrosion (27). If these alloys are exposed to temperatures from 1200° to 1400°F, intermetallic compounds such as sigma or chi might form which can also cause a loss in toughness (28).

Austenitic stainless steel is of great interest for cryogenic machinery, particularly in superconducting magnets, because of its excellent combination of strength and toughness. Table 1 (24) shows tensile properties and Charpy V-notch impact strength of some austenitic stainless steels at low temperature. Note that the high ductility (elongation and reduction of area) of the austenitic stainless steels is retained at cryogenic temperatures. Note also that the yield and tensile strengths of these steels increase as the temperature is decreased, with the larger increase occurring in the tensile strength. The toughness of these steels is also excellent at cryogenic temperatures.

Figure 4 (24) shows the results of impact tests conducted on Types 302, 304, and 304L in the annealed and in the sensitized condition. Note that in the relatively high carbon Type 302 the sensitization treatment has markedly reduced the toughness of the steel at -300°F. However, the toughness of the lower carbon Type 304 and 304L steels remains at relatively high levels after sensitization. Thus, for cryogenic equipment that is welded, low-carbon austenitic stainless steels, such as Type 304,

Table 1. Tensile Properties and Transverse Charpy V-Notch Impact Strength of Some Austenitic Stainless Steels.

AISI Type	Testing Temp., °F	Yield Strength ksi	Tensile Strength ksi	Elongation in 2 in., %	Reduction in Area, %	Energy Absorbed ft-lb
304	75	33	85	60	70	154
304	-320	57	206	43	45	87
304	-425	64	245	48	43	90
304L	75	28	85	60	60	118
304L	-320	35	195	42	50	67
304L	-425	34	220	41	57	67
310	75	45	95	60	65	142
310	-320	85	158	54	54	89
310	-425	116	178	56	61	86
347	75	35	90	50	60	120
347	-320	41	186	32	32	66
347	-425	46	211	50	50	57

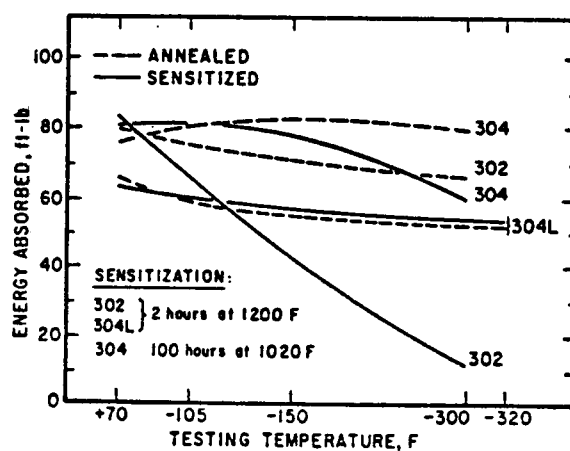


Figure 4. Results of Charpy Keyhole Impact Tests on Types 302, 304, and 304L in the Annealed and Annealed-and-Sensitized Conditions.

appear to be a better material of construction than the high-carbon austenitic stainless steels, such as Type 302.

It was found by NBS that the nitrogen-strengthened austenitic stainless steel generally exhibit higher strength but lower toughness than the standard 300 series (29). It also found that ferrite-free weld deposits have acceptable CVN values except when the nitrogen content is greater than 0.15 wt% (30). The low-temperature fracture toughness of austenitic stainless steel SMA weldments decreases sharply with increasing delta ferrite content over the range 1-10% delta ferrite. Raising the electrode carbon content from 0.03 to 0.06% decreased the low-temperature fracture toughness of E316L weldments but increased their strength (31).

The cryogenic toughness of SMA austenitic stainless steel weld metals was studied by Szumachowski and Reid (8,32). Although effects of individual factors varied from grade to grade, the following rules established by the authors govern and set the maximum toughness at cryogenic service temperatures, particularly at -320°F :

- (1) Low in ferrite - preferably 0 FN.
- (2) Low in nitrogen - preferably $< 0.05\%$.
- (3) Low in carbon - preferably $< 0.04\%$.
- (4) High in nickel - maximum permissible.
- (5) Deposited by a lime-type electrode.

2. Weldability of Austenitic Stainless Steels

The austenitic stainless steels are considered the most weldable of the high alloy steels. This favorable appraisal is mainly based upon the good toughness of the welded joints - even in the as-welded condition. Comparatively little trouble is experienced in making satisfactory welded joints if the inherent physical characteristics and mechanical properties

of an austenitic steel are given proper consideration. For example, the austenitic steels have a coefficient of expansion approximately 50% greater than plain carbon steel, while the thermal conductivity is about one-third as great. This calls for more attention to the factors that control warpage and distortion.

An important part of successful welding of austenitic stainless steels is the control of composition and microstructure through proper selection of type, welding procedure, and postweld treatment. As the steels become more complex in composition, heavier in section, or the service conditions more demanding, a greater degree of knowledge of stainless steel metallurgy is needed to avoid difficulty. Although problems may occur only infrequently, the cost of materials and importance of the weldment usually call for an early solution. Three important objectives in making weld joints in these steels are (33): (a) preservation of corrosion resistance at weld joints, (b) prevention of cracking in or near the welded joints, and (c) avoidance of harmful effects from sigma-phase formation.

When intergranular chromium carbides are precipitated adjacent to welded joints, corrosion resistance decreases. While intergranular carbide precipitation is generally regarded as taking place over the temperature range of 800°F to 1600°F, the rapid heating and cooling that accompany welding restrict sensitization to a somewhat narrower temperature range. The base metal immediately adjacent to the weld is annealed or solution treated by the heat of welding, and since the cooling rate is usually adequate to hold the dissolved carbides in solution, this zone exhibits normal resistance to corrosive attack.

However, a short distance from the weld, in the order of about 1/8 in., the narrow zone is encountered in which intergranular carbide precipitation takes place.

Carbide precipitation can be prevented by three methods: (a) solution heat treatment, (b) use of an extra low carbon (0.03% max.) alloy, and (c) use of a stabilized alloy containing special carbide-forming elements, such as titanium or columbium-plus-tantalum.

Two general forms of cracking have been observed to occur in welded austenitic stainless steels (34). These are both related to hot cracking;

1. In the weld metal.
2. In the base metal near the weld joint.

The occurrence of cracking is related to (a) the microstructure of the weld metal as solidified; (b) the composition of the weld metal, especially the content of certain residual or trace elements; (c) the amount of stress developed in the weld as it cools; (d) the ductility of the weld metal at high temperatures.

Susceptibility to hot cracking is highly dependent on the microstructure of the weld metal. Weld metal with a wholly austenitic microstructure is considerably more susceptible to hot cracking than weld metal with a duplex structure of delta ferrite in austenite. The content of alloying elements and residual elements strongly influences the susceptibility of fully austenitic stainless steel weld metal to hot cracking. Susceptibility can be reduced by a small increase in carbon or nitrogen content or by a substantial increase in manganese content. It is important to minimize the stress imposed on austenitic stainless weld metal as it cools from the solidus down to about 1800°F. It is in this

temperature range that the weld metal is most susceptible to hot cracking and if the stress level is high, fissures will form and propagate to form visible cracks. In order to obtain a weld metal of duplex structure that has a controlled ferrite content (4 to 7% generally), a filler metal of suitable composition may be selected by using a DeLong constitutional diagram.

Although the effects of residual stress on the properties of austenitic stainless steels are not well known (in comparison with the effects of cold working), it is apparent that residual stress may significantly affect the mechanical behavior of an engineering structure fabricated from stainless steel. Stress relieving can be performed over a wide range of temperature, depending on the amount of relaxation required (35). The stress-relieving temperature selected must be compatible with the extent of carbide precipitation acceptable, and with the corrosion resistance desired. Because of the high coefficient of expansion and the low thermal conductivity of austenitic stainless steels, cooling from the stress-relieving temperature must be controlled.

Exposure to temperatures in the 1100^o-1700^oF range is known to produce sigma embrittlement in chromium -nickel stainless steels (36). Both wrought materials and weld deposits are susceptible to deterioration but, given the same conditions of exposure, weld metal may be more rapidly and more severely embrittled than wrought materials (37).

Sigma phase is a hard, brittle, nonmagnetic intermetallic compound of complex structure with a BCT lattice. Sigma in weld metals may form from ferrite in duplex structures or directly from austenite in fully austenitic structure (37). Embrittlement by sigma formed from ferrite occurs relatively fast with the degree of embrittlement depending on

original ferrite content, its distribution, and amount of ferrite transformed to sigma. Sigma formation directly from austenite occurs more slowly.

The effects of alloying elements on the formation of sigma phase in stainless steels are broadly summarized by stating that ferrite stabilizers, such as Cr, Si, Mo, Al, Ti, and Nb, promote sigma. On the other hand, the austenite stabilizer, C, N, Ni, and Mn, generally retard the rate of sigma formation (38).

One obvious method of controlling sigma embrittlement of weld metals is to limit ferrite content. This can be done by careful balancing of the austenite-forming and ferrite-forming elements in the weld metal. A closely related method involves the use of fully austenitic weld metals. Since sigma forms very slowly in such metals, they offer a considerable degree of protection when exposure time in the 1100°-1700°F range is limited.

A postweld heat treatment in the 1925°-2000°F range is effective in reducing sigma embrittlement in duplex weld metals (39,40). The beneficial effect is apparently due to solution of some ferrite into the austenite and to agglomeration or spheroidization of the remaining ferrite. Sigma formed later is then reduced in amount and is present in a discontinuous pattern.

3. Solidification Behavior of Austenitic Stainless Steel Weld Metals

During the usual solidification process, the transformation of a liquid phase to a solid normally occurs by a process of nucleation and growth. The creation of solid nuclei and the growth of the solid

from these nuclei can proceed at finite rate only if the system is under-cooled below the equilibrium reaction temperature. In the fusion welding processes, however, the nucleation event is not significant since a solid-liquid interface is always present and epitaxial growth may occur from the unmelted base metal (41,42). Consequently, in welding, the growth mechanism is of primary interest.

Typical alloys always demonstrate a melting range between the liquidus and the solidus. As the alloy cools through the solidification range, solute is rejected at the solid-liquid interface. Since very little mechanical mixing of the liquid occurs in the immediate vicinity of the advancing interface, the rejected solute must be redistributed in the liquid by diffusion. The freezing process is so rapid that diffusional processes cannot effectively remove the excess solute near the interface. Hence, solute enrichment occurs at the moving interface until a dynamic equilibrium is reached. The resulting dynamic equilibrium provides an excess of solute in the liquid near the interface with the solute content decreasing to the nominal liquid composition at some distance from the interface. As a result, the effective liquidus temperature varies with distance from the interface as shown in Figure 5 (43). The actual temperature gradient in the liquid is also depicted in Figure 5. The region within which the actual temperature is less than the effective liquidus temperature is said to experience constitutional supercooling.

The interface may vary from being planar, through a cellular, a cellular dendritic, a columnar dendritic, to being equiaxed dendritic mode as the degree of supercooling increases. The extent of constitutional supercooling in a given alloy depends on four factors (44,45):

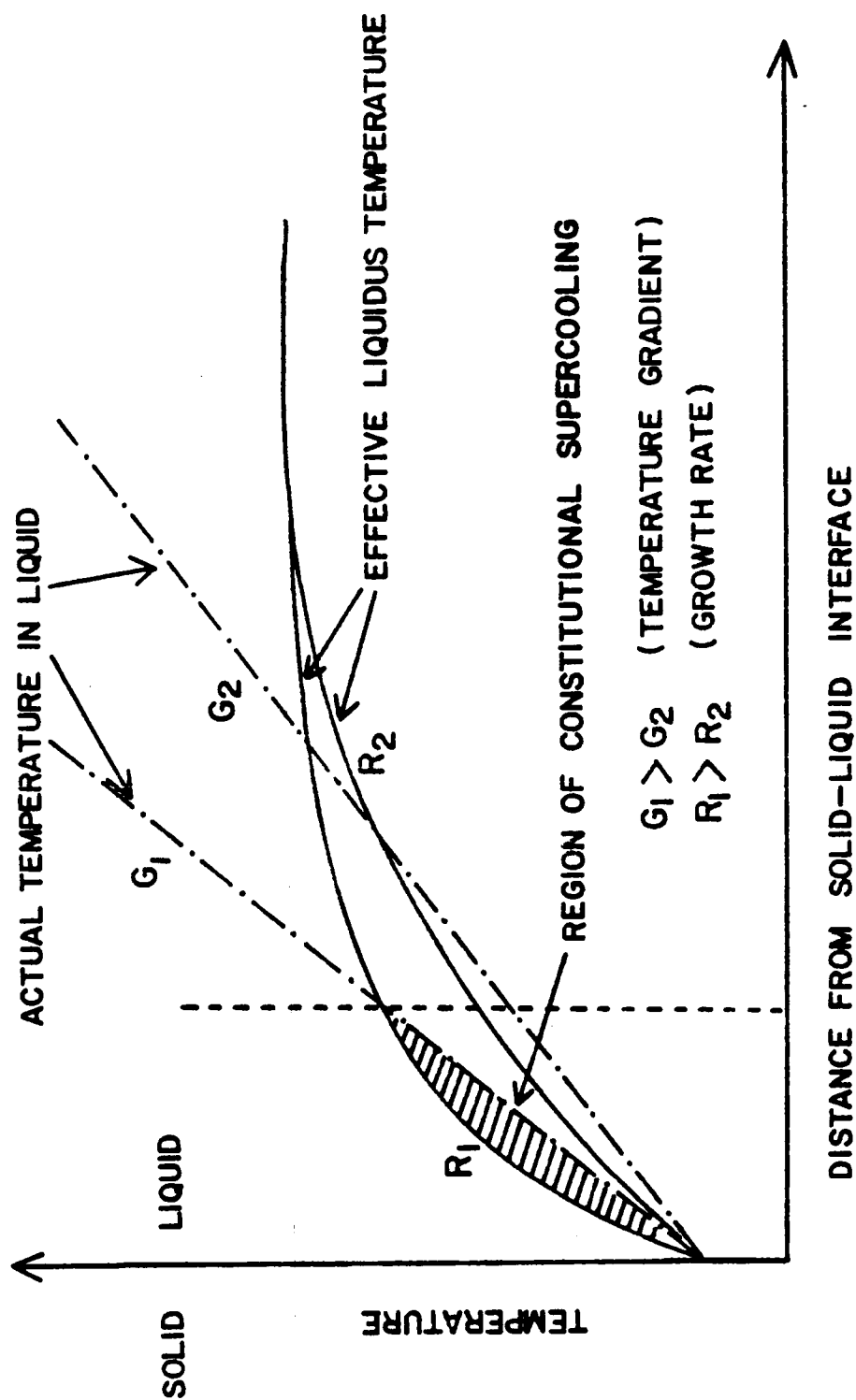
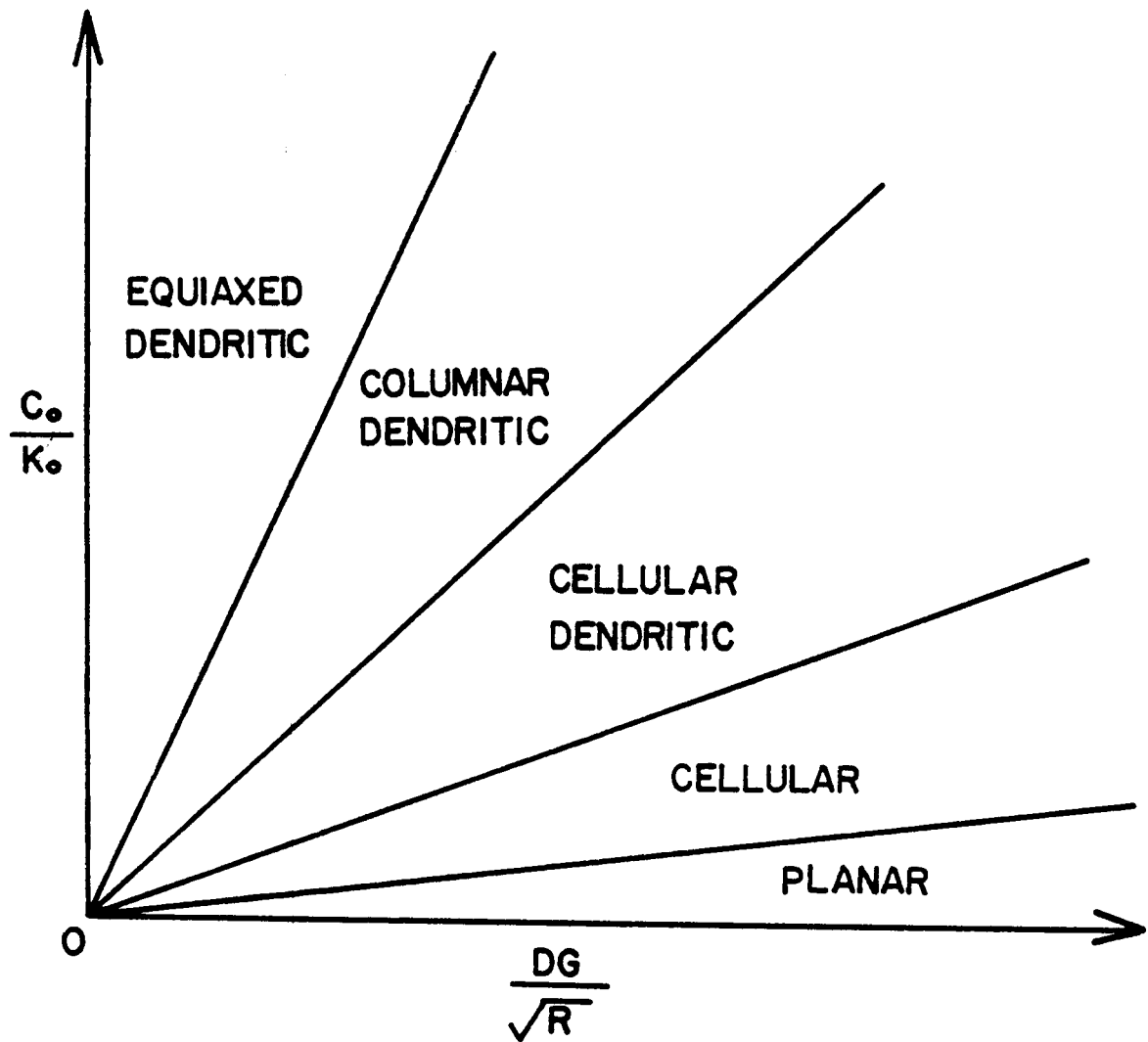


Figure 5. Effect of Growth Rate and Temperature Gradient on Constitutional Supercooling.



C_0 - NOINAL SOLUTE CONTENT
 K_0 - DISTRIBUTION COEFFICIENT
 D - DIFFUSION COEFFICIENT OF LIQUID
 G - TEMPERATURE GRADIENT IN LIQUID
 R - GROWTH RATE AT SOLID LIQUID INTERFACE

Figure 6. Schematic Representation of the Influence of the Parameters C_0 , K_0 , D , G , R on the Pattern of Solidification.

1. The temperature gradient in the liquid, G .
2. The rate of growth at the solid-liquid interface, R .
3. The nominal solute content of the alloy, C_0 . (Actually the parameter C_0 should take into account the distribution coefficient K_0 given by the ratio of the solute concentration in the solid at a given temperature to the corresponding solute concentration in the liquid at the same temperature.)
4. The diffusion coefficient of solute in the liquid, D .

For low degrees of supercooling the interface is cellular. As the degree of supercooling increases a dendritic interface develops as shown in Figure 5 and Figure 6. Dendritic growth is strongly crystallographic and the primary arms and side branches lie parallel to specified crystallographic directions, e.g. the $[100]$ directions in FCC and BCC metals and alloys (41,43). These are the rapid easy-growth directions.

As alloys solidify there are marked effects resulting from the concurrent redistribution of solute. These are not only responsible for the changes in growth morphology, but also lead to solute segregation. Segregation can be classified as either microsegregation (extending over distances of grain diameter or less) or macrosegregation (extending over more than several grain diameters). In turn microsegregation is subdivided into cellular, dendritic, and grain-boundary segregation. The short-range effects occurring when an interface is growing in a cellular form produce compositional variations which although large can readily be eliminated by heat treatment. Dendritic segregation either as coring or by the accumulation of solute in interdendritic spaces is more difficult to eliminate. Grain boundary segregation occurs either by solute accumulation in the grain boundary or by the

impingement of two interfaces growing with a growth component normal to each other. This latter type of segregation is more properly considered as macrosegregation and center-line solute accumulations can be very marked when columnar growth predominates. The detailed nature of the segregation pattern is dependent on the growth conditions and morphology of the growing dendrites and is largely a manifestation of liquid flow in the semi-solid region.

With appropriate solidification conditions and/or suitable post-weld heat treatment weld metals deposited from the E308, E316 types would be fully austenitic at room temperature. However, due to the solidification conditions which apply during welding a certain amount of δ - ferrite is generally present in such weld metals.

A vertical section through the Fe-Cr-Ni diagram at 70 percent Fe is shown in Figure 7 (11). In the vicinity of the compositions representative of austenitic stainless steels (20% Cr, 10% Ni, 70% Fe), both the δ and γ phase may coexist at room temperature depending on the thermal history of the alloy. Note that in addition to the $\delta + L$, $\gamma + L$, $\delta + \gamma$, δ and γ phase fields, there exists a three-phase region, $\delta + \gamma + L$, in the immediate vicinity of the austenitic alloys of interest. For alloys to the right of the three-phase reaction area, the primary solidifying phase is δ , and subsequent transformation results in the formation of γ from the δ by a solid-state, isomorphous reaction. The relative amounts of δ and γ in the room temperature microstructure will be dependent on cooling rate as well as composition. Alloys to the left of the three-phase reaction field solidify to γ and remain so upon cooling. Segregation naturally occurs during solidification for alloys both to the left and

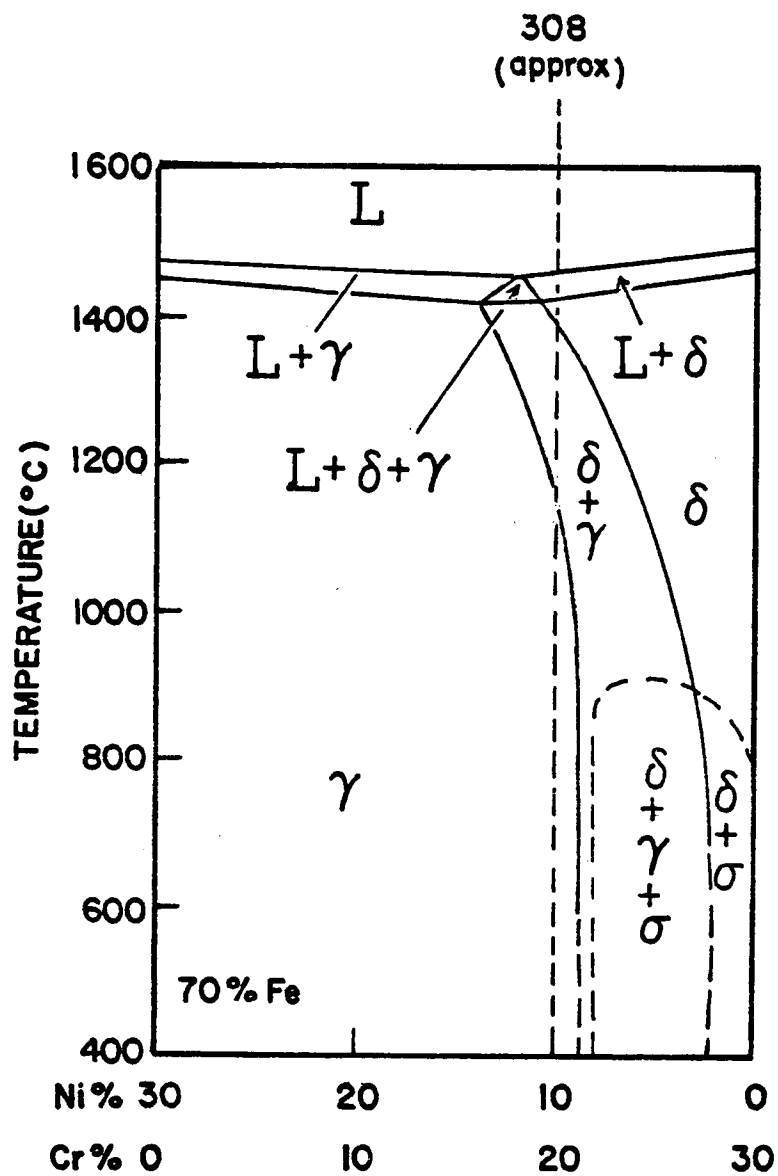
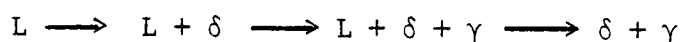


Figure 7. Vertical Section (at 70 wt.% Fe) of Ternary Equilibrium System Fe-Cr-Ni.

right of the three-phase region with the liquid composition being altered with decreasing temperature toward the compositions of the three-phase field. The precise amount of segregation cannot be deduced from the section shown but no doubt exists as to the tendency of the liquid composition to reach the three-phase region. If this occurs, then $\delta + \gamma$ will form simultaneously from the liquid, a fact which many authors (prior to 1979) have decided to overlook or assess with no significance accorded it. Recently, by using the thermal analysis technique and fractographic examination of the solidification cracks induced by the varestraint test, Matsuda et al. (46) concluded that the propagation of solidification cracks in the weld metal containing a small amount of delta-ferrite is arrested by the peritectic/eutectic reaction products. This fact may be the key to the fissuring behavior of the various austenitic stainless steels and may be the reason why composition control is so important.

A series of experiments that combined thermal analysis and interrupted solidification was carried out by David et al. (47) to understand the solidification behavior of type 308 stainless steel weld metal. Results indicate the following sequence of phase separations for the alloy investigated:



Ferrite observed at room temperature was identified as the untransformed primary δ -ferrite that formed during the initial stages of solidification. The mode of freezing and the extent of solid-state diffusion both

control the amount and distribution of ferrite. Microstructural and microprobe analyses indicated both extensive solid-state transformation and solute redistribution during solidification and cooling from the nonequilibrium solidus to room temperature.

4. Mechanism of Hot Cracking in Austenitic Stainless Steel Weld Metal

Although austenitic stainless steels are generally regarded as readily weldable materials, with considerable tolerance for variations in welding conditions without risk of cracking, fully austenitic weld metals may crack both during deposition and in underlying weld runs reheated by subsequent passes. The occurrence of this hot cracking often gives rise to a serious problems for some constructional works.

Many theories have been proposed in regard to the mechanism of hot cracking in welds. Of these theories, the most reasonable one is the one which provided the best understanding of the varied phenomena related to hot cracking and is known as the "Generalized Theory" and is ascribed to Borland (48-50). This theory is essentially a combination of the shrinkage Brittleness Theory of Bochvar and others (51-54) and strain theory of Pellini (55,56). Borland has proposed four solidification stages as shown in Figure 8 (48):

- (1) Primary dendrite formation; in which the solid phase is dispersed and the liquid is continuous, both solid and liquid being capable of relative movement, and therefore no cracking is possible.
- (2) Dendrite interlocking; both solid and liquid are continuous but only the liquid is capable of relative movement and is able to circulate freely between the interlocking dendrites. In this stage if the grains are pulled apart by excessive

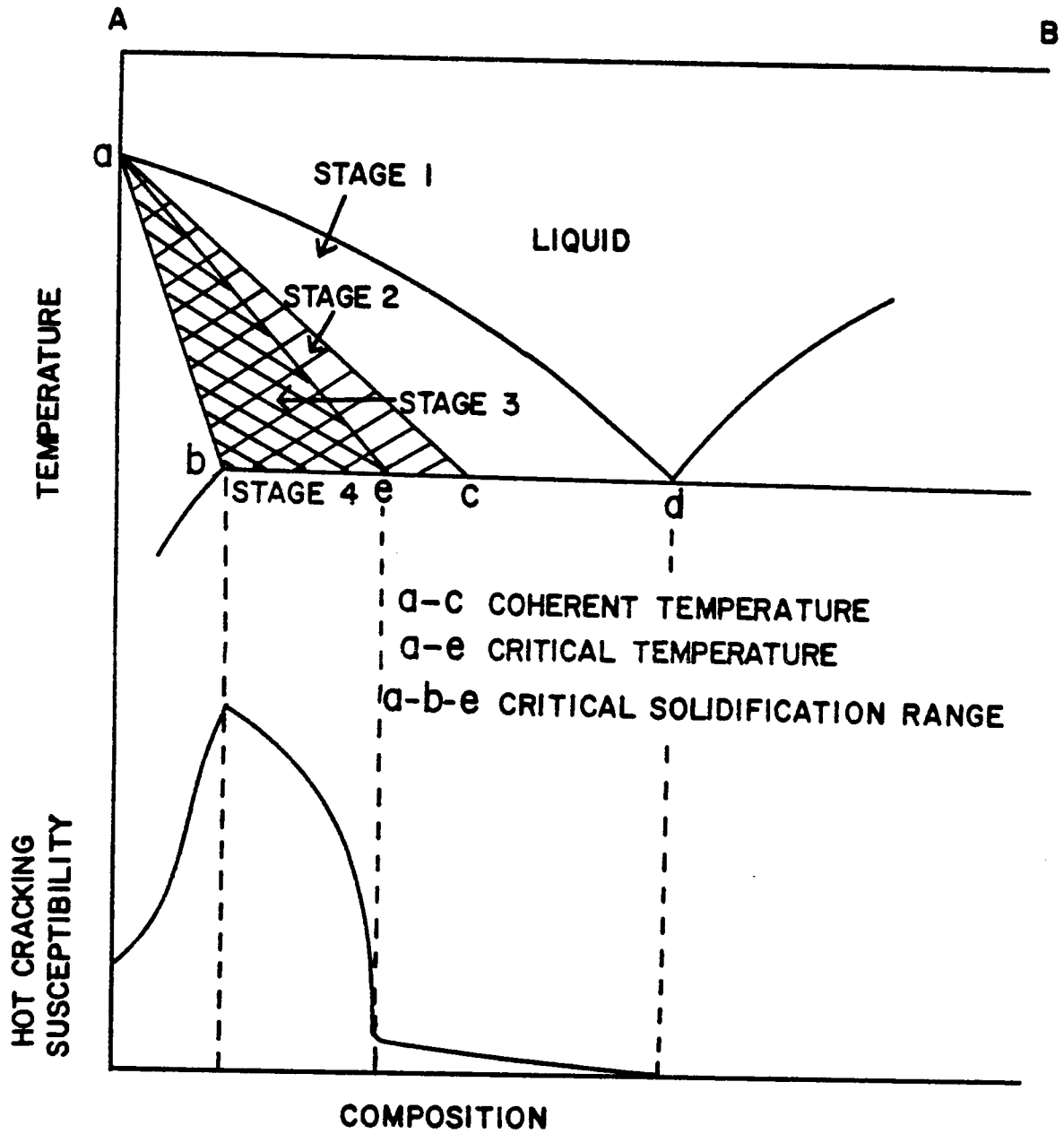


Figure 8. Effect of Constitutional Features on Cracking Susceptibility in Binary Systems.

straining, the cracks are refilled and healed by the inter-connecting liquid.

- (3) Grain boundary development; the solid crystals are in an advanced state of development and a semi-continuous network of solid restricts the free passage of liquid. This is called the "critical solidification range". No healing of cracks is possible if the accommodation strain is exceeded, because the remaining liquid volumes are not interconnected.
- (4) Complete solidification; the remaining liquid has solidified and no cracking occurs.

The amount of cracking experienced is related to the amount of straining in the critical solidification range. The temperature difference in the critical solidification range is considered to be in proportion to that of nominal liquidus and solidus temperature in general. Since welds do not solidify under equilibrium conditions, the nominal solidus will be depressed by coring and lack of time for diffusion. In multicomponent systems with impurities, the effective solidus can be severely depressed. This leads to a wider temperature range during stage 3 for restraint to build up and severe cracking may occur.

Meanwhile, under the same concept that solidification cracking takes place in the final stage of solidification, Prokhorov (57) has represented a low ductility curve in the brittleness temperature range during solidification which corresponds to the stage 3 of Borland. Using the Prokhorov's concept and the Trans-Varestraint Tester (58) which was modified from the original Varestraint Tester (59), Arata et al. (18) rated five types of commercial austenitic stainless steels in

decreasing order of hot crack susceptibility as follows: Type 310, Type 316, Type 347, Type 321 and Type 304.

More recently Masumoto et al. (11) proposed that the determination of weld metal fissuring should be based on composition alone and not on ferrite content. Using the Fe-Cr-Ni phase diagram and weld metal of different compositions, they claim that fissuring depends on whether austenite or delta ferrite is the primary phase solidifying from the liquid. The alloys whose compositions nucleate austenite as the primary phase have a high sensitivity to fissuring. However, for the weld metals that nucleate delta ferrite as the primary phase, the sensitivity to fissuring is low even with high S content (0.15%). Masumoto believes that when the delta ferrite nucleates as the primary phase, it is able to dissolve larger amounts of S and other fissuring-forming elements thus preventing the segregation of these elements to the grain boundaries where they are most influential in causing fissuring.

The question of heat-affected-zone (HAZ) hot cracking has received considerable attention since the early 1950's. Many of the theories of HAZ hot cracking were initially derived from the liquation theories of weld metal cracking and to this date the existence of liquid films is deemed essential in the formation of HAZ hot cracking by most investigators (48,49,55,56,60,61). However, there are a few investigators who continue to advance solid state cracking mechanisms (61,62). These authors also recognize that both liquation and solid state mechanisms may be operable at the same time and within the same temperature range.

The mechanism of formation of liquid films in the HAZ during welding has been related to:

1. Liquation of grain boundaries,
2. Low melting segregates,
3. Absorption of solute from the weld puddle,
4. Constitutional liquation of inclusions and second phase constituents.

The solid state mechanisms postulate that polygonized boundaries, solute segregation at grain boundaries and solute segregation at solidification subboundaries produce strain intolerant microstructural regions and thus sites for hot crack initiation. All theories necessarily involve a strain induced during welding whether of thermal or mechanical origin or combination of both.

At risk of oversimplification, it can be stated that the formation of a crack in a weldment requires (3):

1. That the material comprising the weldment exhibits a phase or a combination of phases which possess a limited capacity to tolerate strain within some critical range of temperature;
2. That the strain imposed upon the weldment by the combined action of thermal and restraint stresses within this critical range of temperature exceed the strain tolerance of the composite microstructure.

Any attack on the problem of hot cracking (fissuring), regardless of the alloy type, must address itself to these two main features before it can become of any consequence.

Although the majority of austenitic stainless steels are considered readily weldable having considerable tolerance for variability in the welding parameter, the fully austenitic weld metals are known to be susceptible to fissuring in the weld metal of single run welds and in the

HAZ of the preceeding weld beads in multipass welds. The liquation theories as outlined above are believed to be relevant to the mechanism of formation of fissures in both cases.

5. Effect of Multiple HAZ Thermal Cycle on Fissuring Tendency

It is recognized that austenitic stainless steel weld metals are susceptible to fissuring in the fusion zone of single-run weld beads (11,58,63,64), in the base metal HAZ (63-70), and in the HAZs of weld metal produced by subsequent beads in multipass welds (12,65,66,70-82). The fissuring in single-run welds is normally combated by reduction of weld metal restraint and by composition control to assure residual ferrite in the room temperature microstructure. The base metal HAZ phenomenon has been largely associated with the stabilized grades of stainless steel such as 347 and in recent years has become a relatively minor problem due to improved residual elemental control in the melting and processing of the wrought materials.

Microfissuring in multipass welds in austenitic stainless steels has become of greater importance in recent years due to the increased utilization of heavy section stainless weldments. There is almost universal agreement among investigators that the fissuring in multipass welds is restricted to the weld metal HAZs produced by the multiple weld passes needed to complete a heavy section weldment (73, 76-82). It has also been established that composition control of the deposited weld metal resulting in a modest amount of ferrite in the room temperature microstructure, will essentially eliminate the microfissuring tendency. This level of ferrite was more firmly established for eight austenitic stainless steel weld metals by Lundin et al. (10).

In some studies (78-80) utilizing actual production weldments of 316 stainless steel, the investigators did not find any instance of base metal HAZ cracking, nor did they find any microfissures in the last or cover passes in the weld metal. All of the fissures found were within the bulk of the weld metal (buried) in the HAZ of subsequently deposited beads. Literally hundreds of microfissures were observed and fissure densities of up to 70 fissures/in were recorded on polished and etched weld sections. Lundin, Delong, and Spond (10) also studied fissuring in a variety of weld metals using the Fissure Bend Test. Their observations confirm those reported for the production weld metals. Haddrill and Baker (73) and Honeycombe and Gooch (12,76,77,81) utilized 310 and 316 weld metals in their work, and found as others had before (72), that fully austenitic weld metals are prone to microfissuring. They confirmed the observations in the multirun welds (using pad specimens) that the fissuring is relegated to the underlying HAZs and virtually never occurs in passes not subjected to subsequent thermal cycles. Unfortunately, the fissuring of austenitic stainless steel weld metals in multipass heavy section weldments has not been studied extensively to date and much work remains to be done.

By utilizing Fissure Bend Test on SMA deposited pads, Lundin and Spond (9) found that the multiple HAZ thermal cycle exposure is a significant factor in the fissuring tendency of 308, 316, and 16-8-2 weld metals at low ferrite levels. In their work they more clearly defined the effect of multiple HAZ thermal cycling has on the fissuring tendency by remelting the surface of SMA deposits of both types 316 and 308 (0.6 FN) using overlapping GTA beads. No fissures were found upon bend

testing along interpass boundaries which were subjected to only a single HAZ experience. Only when a weld bead underwent double or triple HAZ experiences did any fissures appear upon bend testing, thus confirming the results of the directly deposited SMA weld bead studies.

The effect of multiple weld metal HAZ thermal cycles on the fissuring propensity of austenitic stainless steel weld metal is unmistakable in light of the foregoing results. This phenomenon appears to be the most influential element in the gamut of fissuring influences in austenitic stainless steels.

Since the occurrence of a single weld metal HAZ experience does not produce myriads of fissures upon the application of strain, it is clear that the strain tolerance of the grain boundary microstructural regions is significantly reduced by multiple thermal cycling. This degradation of grain boundary ductility can be due to (9):

1. Enhanced segregation due to continued partitioning of harmful trace elements to the grain boundaries followed by liquation and rupture under strain.
2. Additional strengthening of the matrix by thermally-induced precipitate reaction as is the case for some ductility-dip incidences. The increased strength of the matrix may cause strain accumulation in the degraded grain boundary microstructural region thus leading to rupture.
3. Strain-induced precipitation leading to a strengthened matrix and the subsequent accumulation of strain in the degraded microstructural region.
4. The simple accumulation of thermal and restraint strains over

a range of temperatures produced by continued thermal cycling causing the strain tolerance of the grain boundary microstructural region to be exceeded.

Each of the above elements of the phenomenon may be operational at the same time. However, it is also clear that liquation plays a role in the majority of fissuring occurrences and must, therefore, be a part of the mechanism. Thus it appears that enhanced partitioning or segregation during thermal cycling can play a decisive role.

6. Role of Ferrite in Austenitic Stainless Steel Weld Metals

A long-standing problem in the weldability of stainless steels of the austenitic type is "hot cracking" during or immediately after the welding process. It was recognized that the presence of ferrite in the room temperature microstructure indicated a reduced fissuring tendency for austenitic stainless steel weld metals (4). It was shown that weld metal containing three to ten percent ferrite was more resistant to hot cracking than either fully austenitic or fully ferritic alloys. Recently, Spond (3) and Lundin, Delong and Spond (10) have quantitatively evaluated the ferrite requirements needed to eliminate fissuring in a variety of common austenitic stainless steel weld metals. Table 2 (10) shows the ferrite numbers needed for fissure elimination for the weld metals deposited by the SMA process.

Ferrite is a solid solution of one or more elements in body-centered cubic iron. In iron and in mild or low alloy steels it is both the high temperature form, which develops as the alloy freezes, and the room temperature form. Since ferrite is magnetic, the magnetic permeability of stainless steel will increase with ferrite content. For electrical

Table 2. Ferrite Number Needed for Fissure Elimination.

Weld Metal Type	Ferrite Number (FN)
16-8-2	0-2
316L	1.5
308	2.0
316	2.5
308L	3.0
309	4.0
318	5.0 (or less)
347	6.0

applications, this permeability must be kept low. The high Cr content in ferrite enhances the formation of chromium carbides during heat treatments or service in the sensitization range. Sigma formation is also enhanced by the presence of ferrite during aging treatments at temperatures in the range 650° - 900° C (7,36,37). The formation of carbide and sigma causes low temperature embrittlement of austenitic stainless steel weld metal and also appears to reduce creep rupture ductility (83).

The ferrite (in the room temperature microstructure) per se may not be related to the true mechanism of fissure formation but may only reflect the compositional effects directly responsible for fissure formation. Other alloying elements beside Cr and Ni have an effect on the amount of ferrite formed within the weld metal (84). Cr, Mo, Ti, Si, and Cb increase ferrite content and thus are termed "ferrite formers." On the other hand, Ni, C, N, Cu, and Co tend to decrease ferrite content and increase austenite content and are thus termed "austenite stabilizers." The role of Mn is complicated. At low contents, it is generally regarded as an austenite former (85,86), whereas at higher contents it is thought to favor ferrite (13,54,87-89).

Beside chemical composition and thermal effects, there are procedural factors which affect the resulting ferrite content. In summary, where low δ -ferrite levels are desired or where the content is to be restricted to narrow ranges recommended welding procedures involve (90): (a) short arc lengths, (b) preferred use of basic electrodes, (c) use of the smallest practicable electrode size, and (d) use of the lowest practicable amperages (current density).

In 1967 Hull (4) summarized five theories that had been previously presented to explain the beneficial effect of ferrite in reducing hot cracking tendency and then postulated his own theory.

The most common theory for the effect of delta ferrite in reducing cracking is based on the assumption that ferrite has a greater solubility than austenite for certain harmful elements and impurities. Ferrite can thereby reduce the amount of liquid films and the temperature range over which they persist by decreasing impurity segregation during solidification. Borland and Younger have tabulated the solubilities of a number of elements in delta and gamma iron (49). By analogy, they and other authors (71,72) concluded that phosphorus, silicon, sulphur, columbium, and oxygen would be more soluble in delta ferrite than in austenite in stainless steels. They assumed that the function of ferrite is to decrease the concentration of these injurious impurities at the austenite grain boundaries and thus to reduce their damaging effect on hot cracking.

A second effect of ferrite is the generation of a substantial area of interphase boundary between ferrite and austenite in addition to the austenite: austenite boundaries. This extra interfacial area supposedly acts as a sink to decrease the concentration of impurities at the austenite boundaries and reduce hot cracking.

A third effect of ferrite is based on the assumption that in a two-phase alloy the austenite grain size would be refined. A smaller grain size would reduce hot cracking for the same reason mentioned above.

The fourth effect of ferrite is based on the fact that compositions resulting in primary ferrite reduce the solidification range of the weld

metal, and thus there is a reduction in segregation of fissure-forming elements to the grain boundaries. The fallacy in this theory is that the phase diagrams shown only a gradual change in solidification range as a function of composition, whereas there is an abrupt change in hot cracking with only a small amount of ferrite. Furthermore, it can be readily shown that a reduced solidification range does not necessarily mean a reduction in segregation tendency.

The fifth theory of the effect of ferrite is that fissuring is reduced because the bcc ferrite has a smaller coefficient of thermal expansion than the fcc austenite and thus shrinkage stresses are decreased. Hull indicated that this theory lacks substance because the presence of the small percentage of ferrite required to prevent fissuring would have only a minor effect on the total thermal expansion coefficient.

Hull proposed another theory to explain the role of ferrite in preventing fissuring (4). He postulates that the beneficial effect of ferrite results from the fact that the austenite: ferrite boundaries have a lower interfacial energy than the austenite: austenite boundaries. The austenite: austenite boundaries are wet by the last traces of liquid whereas the austenite: ferrite boundaries are not wet. Thus the austenite: ferrite boundaries can sustain the small but increasing stresses imposed by the contraction of the alloy as it freezes under restraint until all the liquid has solidified and the austenite: austenite boundaries can bear their share of the load.

Recently, by applying the fractographic technique to the solidification crack induced with the Varestraint test (46), Matsuda et al.

proposed a new explanation for the role of delta-ferrite. They concluded that propagation of solidification cracks in the weld metal containing a small amount of delta-ferrite is arrested by the peritectic/eutectic reaction products, because the residual liquids are confined as liquid droplets inside austenite grains formed on the occasion of this reaction.

Ferrite can be predicted by utilizing one of two constitutional diagrams. Ferrite content of an undiluted weld deposit is determined from weld metal composition. For gas tungsten-arc and gas metal-arc welding, a reasonable prediction can be made from filler wire composition. Schaeffler devised a diagram (85), relating compositions by empirical formulas to predict the amounts of martensite, ferrite, and austenite in as-deposited weld metals. It utilized the effect of ferrite and austenite forming elements on the metallographically determined ferrite content. Delong (86) provided a similar diagram, but with the addition of nickel equivalent for nitrogen and the change from percent ferrite to Ferrite Number (FN). Ferrite contents were determined magnetically for the Delong diagram. Both diagrams are very sensitive to the accuracy of the chemical analysis. To illustrate, the cumulative effect of errors in C of +0.01%, Ni of +0.2%, and Cr of -0.25% is a shift of about 2.5FN - a rather substantial amount. Poor chemical analysis can increase the spread between the calculated value and that measured magnetically well beyond the ranges claimed for the diagrams.

A third diagram developed by Hull (84), by studying seventy chill-cast stainless steel alloys, was constructed by regression analyses.

Although it is more complex, the Hull diagram did give a better prediction of Ferrite content than both the Schaeffler and Delong diagrams, particularly for those with higher manganese and nitrogen (13,87-89).

7. Factors Controlling the Hot Cracking Tendency of Fully Austenitic Stainless Steel Weld Metal

One of the well known characteristics of fully austenitic stainless steel weld metal is its susceptibility to a type of intergranular microcracking forming in the underlying, reheated weld runs of multipass weldments. The incidence of microcracking in practice will depend on the weld metal composition and on welding conditions. It was found that similar compositional factors will influence both solidification cracking and microcracking, in the same way that ferrite content influences both. So, in reviewing the information available, papers were considered dealing with the effects of various alloying and residual elements in austenitic materials on hot cracking, including both solidification cracking and microcracking.

By means of the Cast Pin Tear Test, the effects of 18 alloying elements on susceptibility to hot cracking of austenitic stainless steel have been evaluated quantitatively by Hull (91). N, Mn, Mo, W, and Cr decrease the susceptibility to hot cracking. On the basis of equal per cent by weight, Co, V, Al, Ni, Ta, Cu, Si, Nb, Ti, C, Hf, Zr, and B cause increasing amounts of cracking, in the order listed. Hull also recommended possible compositions for alloys resistant to hot cracking, emphasizing that Cr should be as high and Ni as low as possible without giving delta-ferrite, that Mn should be higher than normally used, and C and Si low.

The effect of carbon has been found to be dependent on the Si content, increasing carbon being required to compensate for the undesirable influence of Si (19,92,93). A connection between C, Si, and O_2 in the weld pool has been considered important with respect to microcracking (93,94). It has been suggested that, as carbon is increased, grain boundaries are weakened by carbide formation. Such a mechanical effect is, however, likely to be of secondary importance, since most cracks form by liquation. At contents above 0.2% carbon is beneficial since eutectic carbides are formed which can reduce solidification cracking by a healing mechanism (53,93).

It is considered that Si contents above 0.5% in 25/20 alloys cause a significant increase in cracking (18-21,54), not only because the liquidus-solidus interval is widened, but also because intergranular silicate films may form thus embrittling the grain boundaries. To some extent, the effect of Si can be compensated for by increasing the Mn content. This was demonstrated by Gooch (12) and others (13,95), showing that the addition of Mn above 1% reduces markedly the susceptibility of 25% Cr - 20% Ni alloy to microcracking. The optimum effect was observed with Mn contents between 2 and 6%, although Mn above 2% could cause a reduction in corrosion resistance in mildly oxidizing condition. Polgary (93) pointed out that, in a given austenitic alloy type, cracking was always more extensive when Mn was less than 1%. Mn has a higher affinity for S than Cr, Fe, or Ni and these other sulphides form only when the Mn content is too low (14).

Sulphur and phosphorus are very deleterious and from many studies consistent figures of less than 0.015% for each of these elements are

given (11,14-16,53,61,92-98). Phosphorus at less than 0.02 is probably entirely in solution, but even so it has been suggested that above 0.012% there is a large decrease in resistance to all forms of hot cracking. If P was increased above 0.015% it was found necessary to lower the Si content, otherwise cracking was extreme, and, according to a number of workers, the interrelation of C, S, Si, and P is of paramount importance. The presence of 5% delta ferrite greatly increases the tolerance for P and S of austenitic stainless steel weld metals. Titanium when present in amounts greater than 0.08% may reduce resistance to cracking (91), and the presence of Nb in small amounts - around 0.2 - 0.3% - increases microcracking, the presence of liquation by interaction of NbC and the matrix being detected on the crack surface (19,96,99).

Oxygen appears to increase the adverse effect of sulphur, and by oxidizing Al, Ti, Si, Cr, and V and other ferrite formers, it can cause a low ferrite weld metal to become fully austenitic and be prone to microcracking (100). Nitrogen is an austenite stabilizer and a potential strengthener, but it can cause a reduction of cracking in the weld metal (14,46,100). Zirconium, sometimes used as a deoxidant, is a particularly detrimental element (91).

From the above considerations, the following compositional recommendations were made by Gooch (101) to minimize the incidence of microcracking in fully austenitic stainless steel weld metals, which may assist in the design of welding consumables:

- (1) Carbon should be below 0.1%.
- (2) Silicon should be below 0.3%.
- (3) Sulphur and phosphorus should each be below 0.015%.
- (4) Manganese should be between 3 and 6%.
- (5) Molybdenum should be between 2 and 3%.
- (6) Nitrogen should be between 0.1 and 0.2%.

- (7) Niobium should be below about 0.3%.
- (8) Trace impurity elements should be avoided.

There is general agreement among investigators (63,101,102) on the problem of solidification cracking that the arc energy should be low to avoid cracking. The welding current has the greatest effect on microcracking with reduced current being beneficial. However, travel speed and weld pool size appear to have little effect, although it would be generally recommended that heat input should be minimized. Neither joint restraint nor high interpass temperatures appear significant in causing microcracking. Good fitup is preferred to avoid solidification cracking in the root pass, but fitup does not affect microcracking in subsequent runs.

The incidence of microcracking is also dependent on the grain size developed in deposited weld metal. Microcracking is more extensive with coarse grain size (81). Gooch and Honeycombe found that the reduction in microcracking was associated with two distinct microstructural changes, namely the development of a finer, equiaxed grain size, and a reduction in the content of non-metallic matter, giving a very much cleaner weld metal.

8. Evaluation of Hot Cracking and Hot Ductility of Austenitic Stainless Steel Weld Metals

Over thirty different direct welding tests have been developed to investigate various types of cracking in austenitic stainless steels and other alloys. Most of these tests were devised to evaluate one particular type of cracking problem such as base metal cracking, cold cracking, or hot cracking. Since austenitic stainless steels and other alloys that are susceptible to hot cracking or fissuring are often

deposited by the SMAW or GTAW processes as multipass welds, there was a need for developing a test to determine fissuring tendencies in multipass weldments in an economical and reproducible manner. Such a test has been developed and has become known as the Fissure Bend Test (103). The Fissure Bend Test has the favorable features desired in a weldability test in that it is economical, easy to conduct, reproducible, and capable of evaluating fissuring in multipass weldments very similar to those used in current fabrication procedures.

The Fissure Bend Test employs a multirun, double-layer, weld pad to permit the evaluation of fissuring on relatively undiluted weld metal. The weld pad is deposited by the SMAW process on a rigidly clamped base plate as twelve 8" long stringer beads, six beads to each layer. The pad is then ground to produce a smooth surface for examination. The Ferrite Number of the deposit is determined on the ground surface. Examination of the pad for fissuring is conducted both before bending and after bending.

Fissuring evaluation is performed in the center 4 inches of each specimen. This evaluation is a two-step process: (1) fluorescent penetrant testing to detect surface indications, (2) binocular microscopic examination of these indications. Then count and measure the length of those indications verified as fissures. Bend the specimen to an included angle of 120° in a fixture similar to the one shown in Figure 9 (103). Repeat the entire fissure evaluation process on the bent specimen and record the number and the size of the fissures on the bent pad as was done on the unbent pad.

The Fissure Bend Test has capabilities of providing both practical and fundamental information. Since the test involves the actual

deposition of the weld metal to be evaluated, virtually all the welding process variables can be made a part of the testing scheme and their influence on the fissuring tendency can be determined.

Gleeble testing evolved from an original device built at the R.P.I. by Nippes and Savage in 1949 (104). Though the versatility of the Gleeble has led to the adoption of several different testing techniques, the Gleeble is basically a high-speed, resistant-heating control coupled with a high-speed, tensile-testing apparatus. With 0.25 in. diameter specimens (of plain-carbon steel, for example), the Gleeble can reproduce with an accuracy of ± 15 F, a weld HAZ thermal cycle requiring a heating rate less than $3000^{\circ}\text{F}/\text{sec}$ and a cooling rate, at 1000°F , of less than $500^{\circ}\text{F}/\text{sec}$. A block diagram, showing the essential features of the Gleeble, is provided in Figure 10 (105).

The specimen to be heat treated is clamped between two water cooled copper grips that serve as a means for introducing current through the specimen and insure a rapid rate of heat extraction when the current is interrupted. When a preset thermal cycle is initiated, a fine-wire thermocouple, previously percussion welded to the specimen, permits an electronic controller to monitor the specimen temperature, compare the specimen temperature with the instantaneous temperature desired, and to increase, reduce or interrupt the heating current as required. Since the specimen temperature is monitored, and adjusted, each time the 60-cycle current passes through zero, the desired thermal excursion is reproduced with great precision. Under the normal conditions, a uniformly heat-treated microstructure is obtained in the specimen

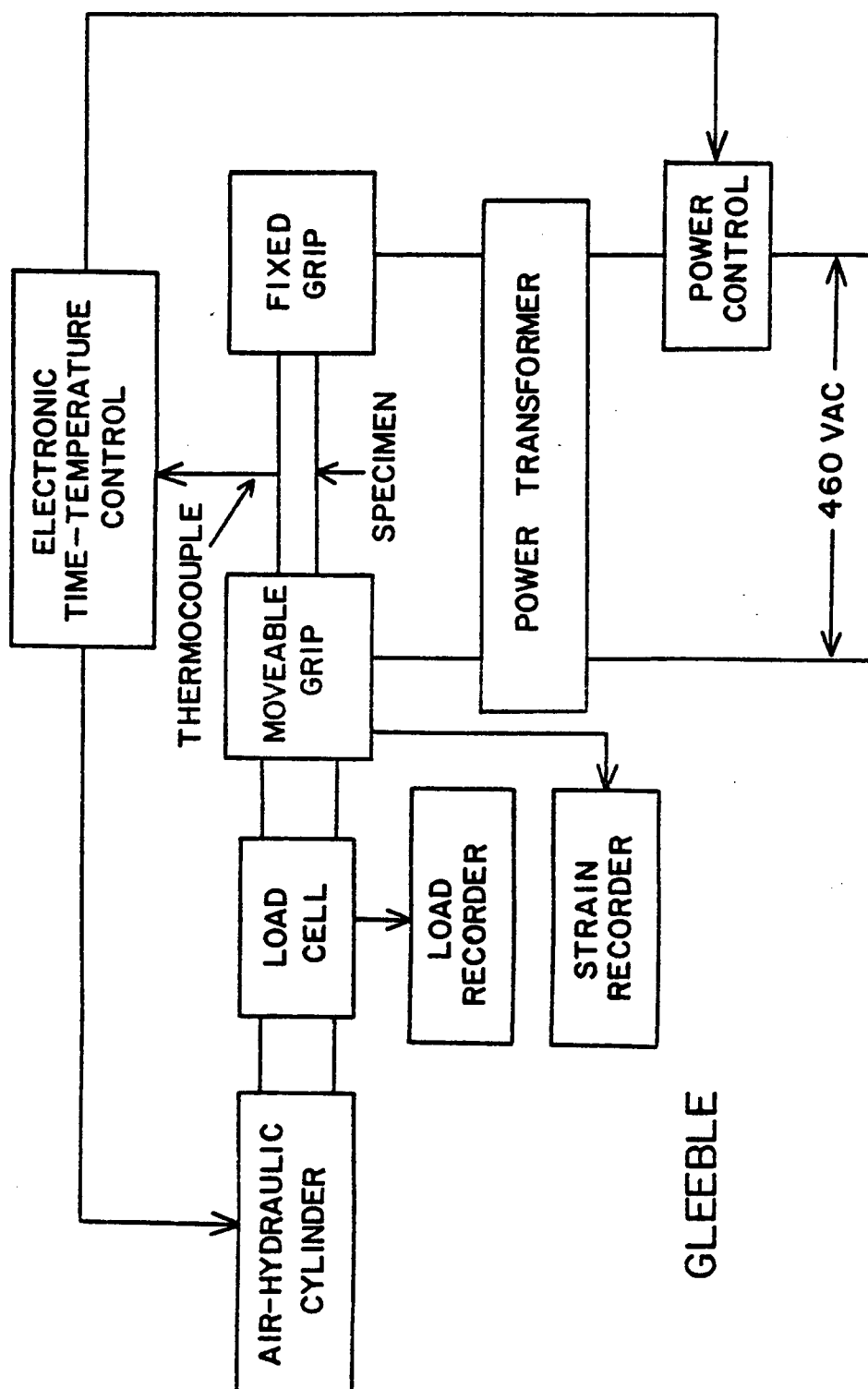


Figure 10. Schematic Diagram of the Gleeble Apparatus.

within the region roughly $1/10$ " on each side of the centrally located thermocouple.

Tensile loading is not employed with all of the Gleeble techniques. If it is required, however, as in hot ductility tests, rapid fracture may be obtained at any preset temperature by a signal from the controller which interrupts the flow of current and actuates the loading mechanism.

The hot ductility test for assessing the elevated temperature strength and ductility of metals is the most commonly employed of the Gleeble techniques. The test employs cylindrical specimens usually $\frac{1}{4}$ " in diameter and 4" long. Generally, twenty specimens are sufficient for a simple materials comparison. Figure 11 (105) provides an illustration of the way hot ductility curves are developed.

With the Gleeble controller set for the desired weld thermal cycle, specimens are heated and quickly broken during the heating portion of the cycle (dark circles, Figure 11, top) at increasing temperatures until a temperature is attained where the specimen fails, without necking, in a brittle manner. This temperature, the "zero-ductility" temperature, is thereafter used as the maximum temperature of exposure for the specimens (open circles) that are broken during the cooling portion of the simulated weld cycle. Comparison of the on-heating and on-cooling ductility curves permits an evaluation of the degree of embrittlement that occurs in the weld HAZ. Materials that exhibit adequate on-heating performance and rapid recovery of on-cooling ductility are expected to accommodate contraction and distortion stresses during welding without cracking. Since the load required to fracture each

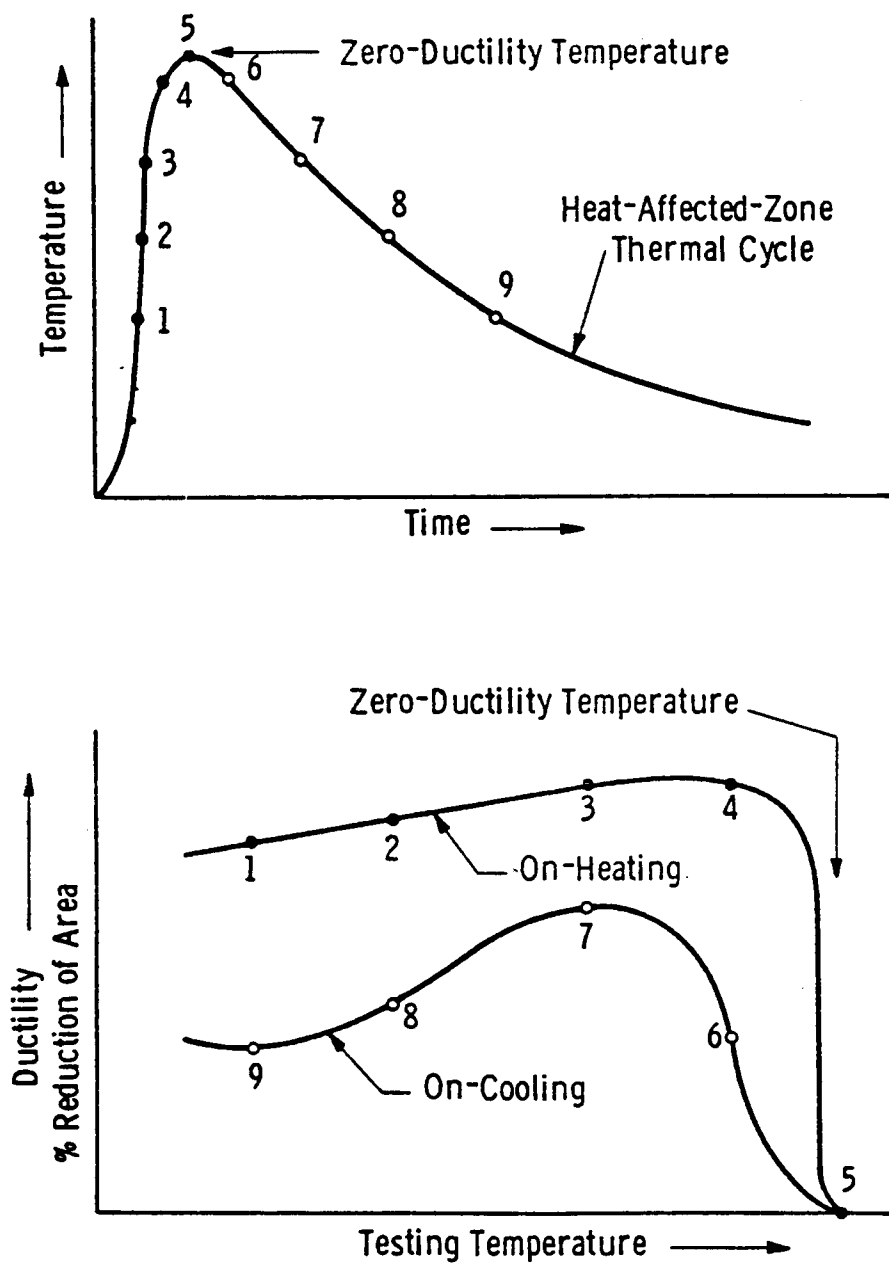


Figure 11. Illustration of the Development of Hot-Ductility-Test Curves.

specimen is also recorded by the Gleeble, similar curves of ultimate strength, may also be plotted. Though exceptions occur, the on-heating and on-cooling strength curves usually show only minor differences and are normally not useful measures of HAZ performance.

After the introduction of the hot ductility test, the method was extensively employed to investigate HAZ fissuring in austenitic stainless steels (106-115). The test not only proved to be invaluable in establishing the relative cracking tendencies of various stainless steels, but demonstrated the ability to differentiate between "good" and "bad" materials of the same nominal composition on a heat-to-heat basis.

After reviewing several criteria currently in use for predicting cracking sensitivity from hot ductility curves, Yeniscavich (116) made a brief analysis of a possible cracking mechanism and how it is related to hot ductility curves. His analysis suggests that the zero ductility range on cooling is the most indicative factor to use in predicting cracking sensitivity. A mid-temperature range dip to zero ductility would also be expected to have a significant on cracking.

In order to eliminate the uncertainties involved in testing the fusion zone and HAZ separately, a testing procedure was developed by Krantz (117) for studying the hot cracking sensitivity of weldments in thin sheets. Briefly, the testing procedure consists of a modified hot ductility test utilizing a composite test specimen containing a longitudinal bead-on-plate GTA weld bead.

In Krantz's test, the transverse cross-section of the hot ductility test specimen is comprised of a full penetration, autogenous

fusion-zone bounded on both sides by actual weld HAZ. Thus, by appropriate selection of the welding conditions used in producing the weld bead, a pattern of microsegregation characteristic of any desired combination of welding variables can be achieved. The specimen is then tested in the Gleeble device by applying predetermined amounts of strain to the composite specimen at any desired temperature during a synthetic weld thermal cycle. A schematic representation of the testing sequence is shown in Figure 12 (117).

In this manner, hot cracking data has been obtained in four heats of Hastelloy X by comparing the effect of controlled strain at elevated temperatures on the their respective cracking behavior. The resulting data indicate a strong correlation between the pattern of microsegregation present in both the weld fusion-zone and the weld HAZ and the location and types of microcracks produced.

For the tests described, the Fissure Bend Test has a demonstrated capability of depositing consumables with various welding conditions and correlates well with fabrication and service conditions. Therefore, the FBT can be used to evaluate the fissuring sensitivity of austenitic consumables with varying compositions and welding processes.

The Gleeble Hot Ductility Test has proved to be promising in evaluating the hot ductility of various materials. It can be easily modified to accomodate a flat sample and can be extended to evaluate the effect of multiple thermal cycles.

The Gleeble Hot Cracking Test developed by Krantz involves a small size of specimen which contains both the weld fusion zone and the weld HAZ. The major advantage of using the Gleeble is an ability to separate the effects of the mechanical and metallurgical factors.

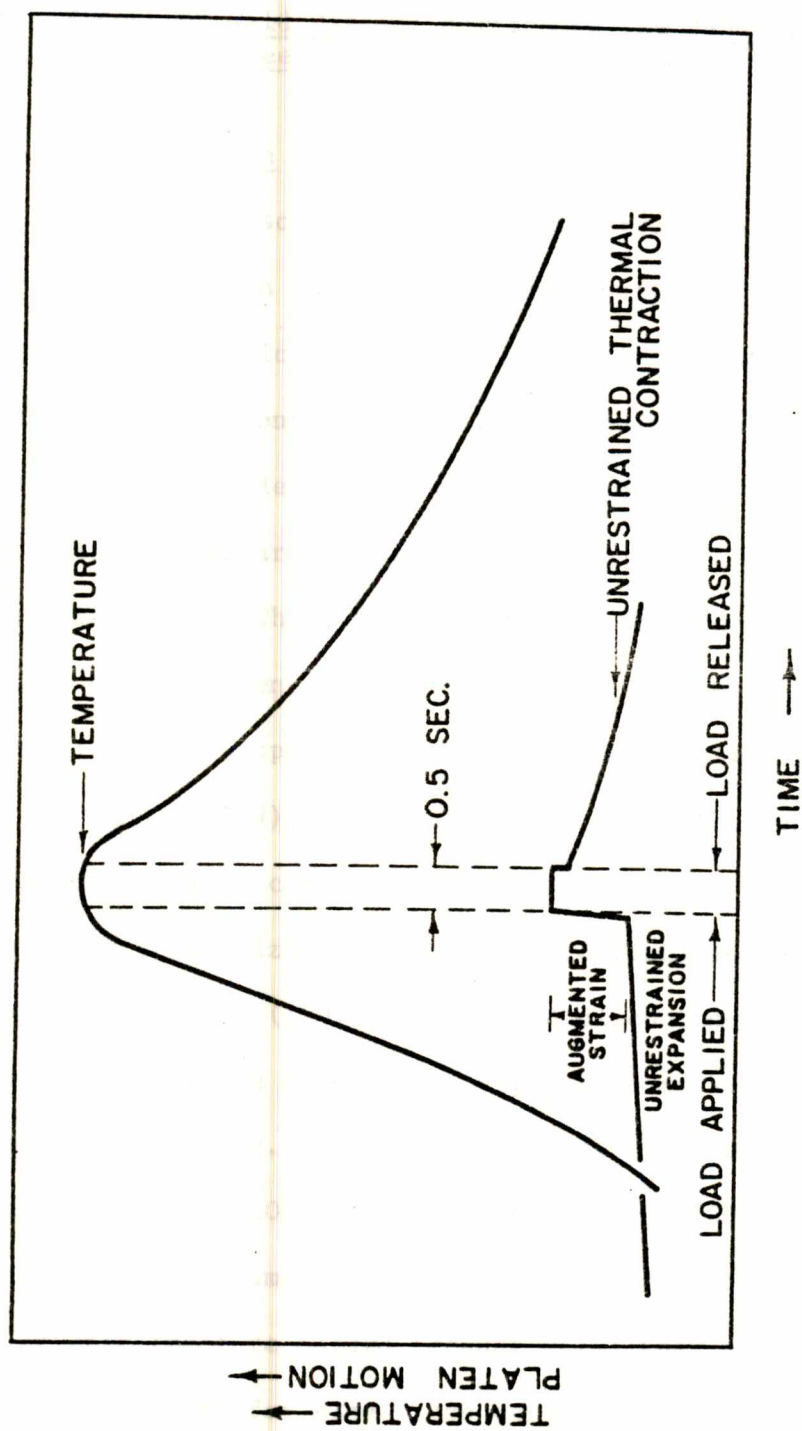


Figure 12. Schematic Representation of Gleeble Hot Cracking Testing Sequence Used by Krantz.

Thus, the relative importance of microstructure, strain, and temperature on hot cracking of weld metals can be explored systematically.

9. Effect of Weld Discontinuities on the Behavior of Austenitic Stainless Steel Weld Metals

9A. Introduction

A weld discontinuity is an interruption of the typical structure of weldment, such as a lack of homogeneity in the mechanical or metallurgical or physical characteristics of the material or weldment. A defect is a discontinuity or discontinuities which by nature or accumulated effect render a part of product unable to meet minimum applicable acceptance standards or specifications (118). This term designates rejectability. Thus, the term defect is used to describe the discontinuity whose presence in the weld metal or the HAZ is known to be detrimental to the proper functioning or service life of the weldment.

Lundin (119), in his review of the literature concerning weld discontinuities, combined the work of various authors to formulate a complete categorization of weld discontinuities as shown in Table 3 (119). Harrison (120) divided weld discontinuities into two major groups: the three dimensional type, such as slag inclusions, porosity, weld profile, etc. and two dimensional type, such as cracks, laminations, lack of fusion (LOF), lack of penetration (LOP), etc. Harrison indicated that the three dimensional discontinuities probably would not create a brittle fracture hazard because-- (a) they are rounded in shape and therefore, size for size, less harmful than cracks; and (b) the through-thickness dimension of such discontinuities is inherently small.

Table 3. Characterization of Weld Discontinuities.

I.	Process and Procedures Related
A.	Geometric
	Misalignment
	Undercut
	Concavity or convexity
	Excessive reinforcement
	Poor reinforcement
	Poor reinforcement Angle
	Overlap
	Burn through
	Backing ring-lack of penetration
	Incert ring-lack of fusion
	Backing-left on
	Incomplete penetration
	Lack of fusion
	Shrinkage
	Surface irregularity-ripples
B.	Other
	Arc strikes
	Slag inclusions
	Tungsten inclusions
	Oxide films
	Weld dressing
	Spatter
	Arc crater
II.	Metallurgical
A.	Cracks or Fissures
	Hot
	Cold or Delayed
	Reheat, stress-relief or strainage
B.	Porosity
	Spherical
	Worm-hole
C.	Heat-affected zone, microstructure alteration
D.	Weld metal and heat-affected zone segregation
E.	Base plate delamination
III.	Design
A.	Changes in section, stress concentration
B.	Weld joint type

While the effects of the three-dimensional weld discontinuities on the behavior of weldments have been emphasized by many workers, a literature survey (for the past twenty years) has shown very few papers that specifically deal with the effects of such discontinuities on the mechanical behavior and properties of austenitic or nickel base alloys. This review shows that attention has been given to the two-dimensional discontinuities and their influence on austenitic steels.

Of all the types of discontinuities in austenitic materials, fissuring has received the most attention. Susceptibility to fissuring can be avoided by using consumables with such compositions that a small portion of the high temperature δ -ferrite phase is retained in deposited weld metal (5). Lundin (10) et al., have quantitatively evaluated the ferrite requirements needed to eliminate fissuring in a variety of common austenitic stainless steel weld metals. The Ferrite Numbers needed for elimination ranged from one to six depending on the composition of the weld metal.

Unfortunately, in certain corrosive environment, the ferrite may lead to preferential attack at the weld metal (6). In high temperature service, a small amount of ferrite (U.S. NRC regulatory Guide 1.31) may promote rapid transformation to the embrittling sigma phase which will reduce the high temperature creep and rupture resistance (83). With the advent of extensive use of austenitic stainless steel in cryogenic service, concern exists that the ferrite in the weld deposit can adversely affect the low temperature toughness (ASME Boiler & Pressure Vessel Code Sec. VIII) because its BCC brittle structure (8).

The influence of discontinuities on the behavior of welds in austenitic stainless steels and many nickel base alloys has stirred much discussion but no large-scale efforts to resolve the doubts in the minds of some. Perhaps the reason for the lack of extensive work is the fact that while fissure-like discontinuities are known to exist in austenitic stainless steel weldments no documented failure data have been compiled concerning their influence. Honeycomb and Gooch (121) in discussion on the effects of microfissures on the mechanical properties of fully austenitic stainless steel weld metals state -- "There must be a considerable amount of operational plants containing cracks, but the authors know of no service failure that can be attributed to micro-cracks."

9B. Effect of Discontinuities on Static Properties

The influence of discontinuities on static properties such as tensile strength and ductility can be considered inconsequential as long as the extent of the area reduction is no greater than five percent of the cross-section under consideration.

Honeycombe and Gooch (121) studied the effect of microfissures on mechanical properties of austenitic stainless steel weld metals. Mechanical tests were conducted on 310 and 316 fully austenitic (fissure containing) welds and the results were compared to those obtained from austenitic-ferritic welds. Results show that microfissures are normally insignificant with respect to the tensile strength of weld joints. Only when microcracks exceed more than 5% of the cross-section area of the weld metal, do they cause a decrease in tensile strength (this would mean that a large number of fissures were present).

Lundin (75) conducted room temperature and elevated temperature (1500°F) tensile tests of Inconel 600 and 310 stainless steel Electron Beam (EB) welds. The results show that the ultimate strength or time-to-rupture at a given stress appeared to be a function of the number and distribution of the discontinuities. In other words, the degrading effect of the discontinuities was only in proportion to the fraction of cross-sectional area they occupied relative to the specimen cross-sectional area. The yield strength appeared to be less affected than was the ultimate tensile strength.

An investigation was conducted by the Southwest Research Institute (122) on the influence microfissures have on type 316 stainless steel weldments. Their results show that the tensile properties are unaffected by the presence of fissures although the weldments investigated had low ferrite content and exhibited a density of from zero to 25 fissures per square inch.

In another study of weld fissuring, Theilsch (123) concluded that microfissures as found in production welds do not significantly reduce the tensile strength. The weld investigated was made to simulate a weld in the cooling system of the Oconee Nuclear Power Station. This weld was designed to have more residual stress than any production weld in the piping system and thus was more susceptible to microfissuring. No code rejectable fissures (Over 1/16 in. long) were found in the weld. Further, the microfissures that were found were narrower than base metal stringer inclusions and were not considered unacceptable.

Theilsch also points out that fully austenitic (indicating presence of fissures) stainless steel welds of type 316 have been used

extensively in fossil fuel power plants operating at high temperatures and pressures. In these cases, not a single weld failure has occurred which, in any way, was related to the presence of microfissures.

Grudzinskii (124) et al. show that the influence of LOF on tensile behavior is a function of temperature and ultimate strength. At normal ambient temperature the LOF discontinuity is relatively innocuous, not reducing static strength at levels up to 10 percent area reduction. However, at -196°C the influence is significant and is related to microstructural instability (transformation of austenite to martensite). The reduction in static tensile strength is in proportion to the ultimate strength. This is particularly evident at low levels of LOF (5%).

Kudryavtsev and Yushchenko (125) have conducted an investigation into the effect of discontinuities (e.g. pores, slag inclusions, and LOP) on the static tensile loading of welded joints of 18-10 and 20-16-6Mn austenitic stainless steels. Figure 13 shows the dependence of the strength of the welded joints on the size of pores and LOF type discontinuities at $+20^{\circ}$ and -196°C . It is seen that in all the steels studied, with the exception of 18-10 at -196°C , the strength decreases in proportion to the reduction in the cross-sectional area of discontinuities. The strength of 18-10 welds rapidly decreases in the presence of a discontinuity with an area of 7%. The mathematical dependences of the strength of the welded joints on discontinuity size makes it possible to calculate the strength of the given welded joints with discontinuities on the basis of the results of non-destructive testing without testing specimens.

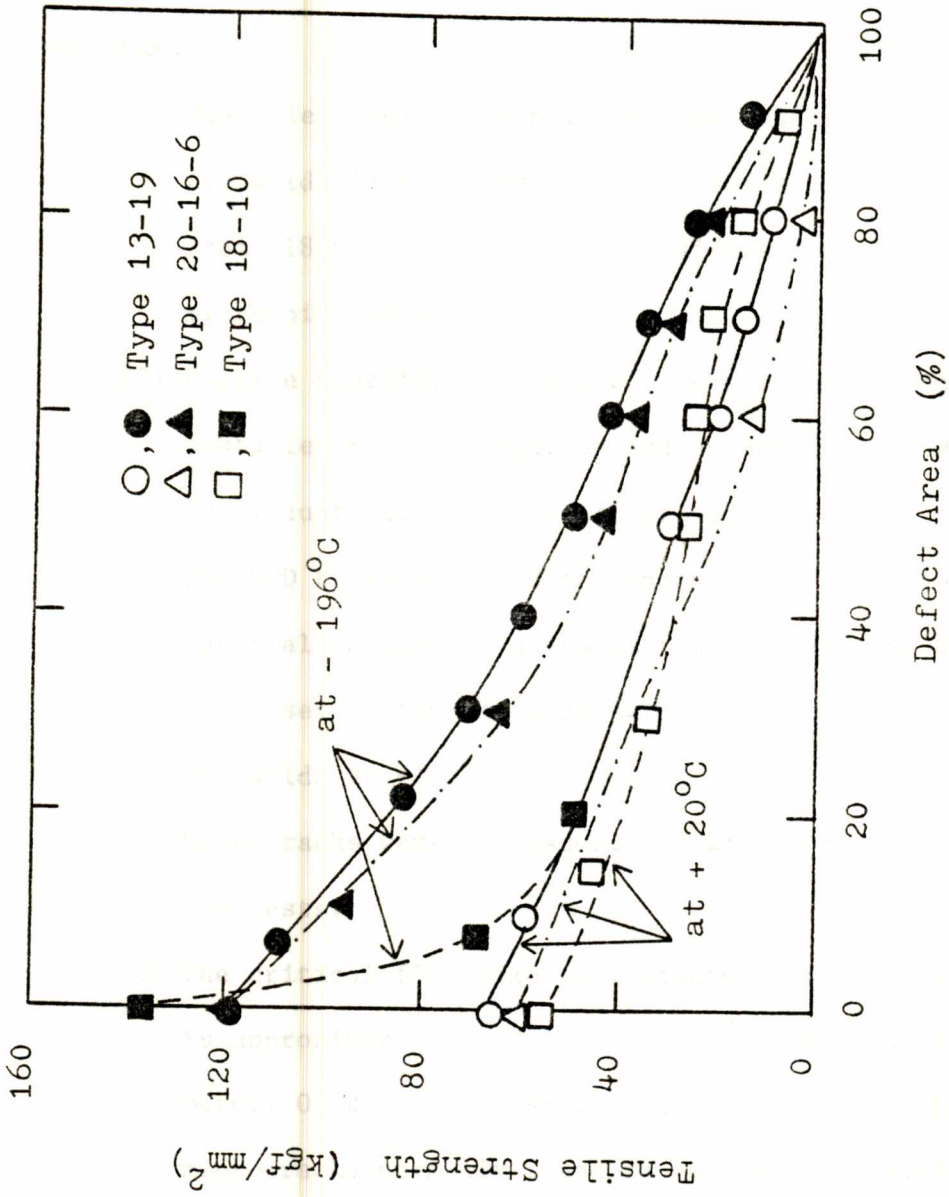


Figure 13. The Effect of Discontinuity Size on the Tensile Strength of Weld Metals.

9C. Effect of Discontinuities on Toughness and Impact Behavior

Honeycombe and Gooch (121) conducted Charpy and Crack-Opening Displacement (COD) tests on 316 and 310 weld metal with microcracks and variable ferrite level as a function of temperature. Their results show that:

- (1) Unstable crack propagation was not observed in fully austenitic welds in the presence of microfissures at temperatures from $+18^{\circ}\text{C}$ to -196°C and the failure mode involved only microvoid coalescence.
- (2) Ferrite containing, fissure free weld metal did exhibit unstable crack propagation and a lower impact energy than fully austenitic weld metal at -196°C .
- (3) The COD tests revealed no tendency for fully austenitic material to exhibit unstable fracture and the COD values for the onset of tearing were twice those of the ferrite containing welds at -196°C .
- (4) Microcracks were not associated with ductile tearing in the COD tests.
- (5) The critical flaw size to initiate tearing at room temperature is approximately 0.3" and at -196°C the flaw size is approximately 0.12". These sizes are far and above the sizes of any microcracks normally found in austenitic stainless steels. These data are for a material thickness of 0.38" and the critical flaw size would be expected to decrease as section size increased. Also the critical size will increase with an increase in temperature above room temperature.

- (6) The authors conclude the microcracks are inconsequential in the behavior of these austenitic materials subject to impact or sudden overload.

The J integral and COD tests have been used by Chipperfield (126) to determine the toughness of wrought 316 stainless steel and relevant 17-8-2 weldments at temperatures up to 850°C. Results have highlighted the fact that the initiation toughness established for a given ductile material can decrease with increasing temperature. It has been shown that the magnitude of this decrease in terms of COD may be associated with the work hardening capacity of the material while for J integral measurements the decrease is additionally dependent on the temperature sensitivity of the yield stress. The absolute values of toughness established for the restrained weldments were found to be broadly insensitive to variations in welding technique, preparation angle and short term stress relieving temperatures below 850°C, but were significantly below the value established for wrought 316 steel at the same temperature. A discontinuity size assessment indicated that, taking residual stresses as additive to applied stresses, tolerable discontinuity sizes were controlled largely by the magnitude of residual stress. Thus, while the initiation toughness of weld metal is not significantly improved by short term stress relief at and below 850°C, the stress relief accompanying such thermal treatments results in an appreciable increase in tolerable discontinuity size.

Kazenov et al. (127) studied the effect of reactor irradiation on the mechanical properties of austenitic 18-10 welds with various amounts of ferrite. It is shown that in cases in which the ferrite content of weld metal does not exceed 7%, the weld metal toughness

properties after irradiation with small doses (up to 10^{21} neutron/cm²) at temperatures up to 500°C, are satisfactory. At higher radiation doses, the toughness and ductility properties of the weld metal rapidly decrease, regardless of the ferrite content.

Austenitic stainless steels are widely used structural materials for large superconducting magnet systems because their good toughness at cryogenic temperatures. In order to reduce hot cracking problems, 5% to 10% delta ferrite is often introduced in austenitic weld metals. Read et al. (31) investigated the effects of delta ferrite on the fracture toughness of AWS E316L and E316 SMA weldments at 295 K, 76 K, and 4 K by conducting Charpy tests and J integral tests.

The results of their work are;

1. Increasing delta ferrite over the ferrite number range 0 to 11 causes a sharp decrease in fracture toughness at 76 and 4 K.
2. The ferrite morphology changes from isolated patches of ferrite at near zero ferrite to a nearly continuous ferrite phase at about Ferrite Number 7.
3. Decreased cooling rate increases the ferrite dendritic spacing and the effective grain size of the austenite matrix and decreases the Charpy impact toughness at 76 K.
4. Increasing the carbon content from 0.03 to 0.06% decreased the fracture toughness at 76 and 4 K.
5. Deliberate sensitization of an AWS E316 weld deposit caused a significant toughness loss at 76 K.
6. Fractography of failed CVN specimens suggests that fracture at 76 K progressed along the ferrite-austenite boundary.
7. The fracture toughness, K_{IC} (J), of E316L welds is approximately half of the K_{IC} (J) for AISI 300-series stainless steel plates.

Same results were found by Szumachowski and Reid (8) on the cryogenic toughness of seven types of austenitic stainless SMA weld metal.

Kudryavtseu and Yushchenko (125) have conducted an investigation into the effect of discontinuities, e.g. pores, slag inclusions, and lack of penetration, on the impact bending of welded joints of 18-10

and 20-16-6 Mn austenitic stainless steels at 20°C and -196°C . Figure 14a shows the dependence of the impact toughness of the weld on the discontinuity size for various cross-sections of the specimens at a temperature of $+20^{\circ}\text{C}$. The data on the variation in the impact toughness of the specimens with different notches, the identical position of the discontinuity in the weld, and a discontinuity size not exceeding 10% of the cross-sectional area of the specimen, are presented in Figure 14b. The test temperature was -196°C . The cross-sectional area of the specimens with the different notches was identical, 80 mm^2 . For comparison, the graph also gives the impact toughness of the identical specimens without discontinuities.

Analysis of the results shows that the characteristics of the dependence of impact toughness on the notch size in the specimens without and with discontinuities (up to 10% of the cross-section of the specimen) are identical.

The presence of weld discontinuities with cross-sections of up to 10% of the cross-sectional area of the specimen reduces the energy necessary of crack initiation, because of the additional stress concentration caused by the discontinuities. The impact toughness of the weld metal with discontinuities decreases by a factor of approximately 1.5 in comparison with the weld without discontinuities at -196°C .

9D. Effect of Discontinuities on High Temperature Creep and Stress Rupture Behavior

There has been concern expressed over the low creep-rupture ductility evidenced by some austenitic stainless steel weld metal and HAZ's. This has been largely related to the microstructural aspects

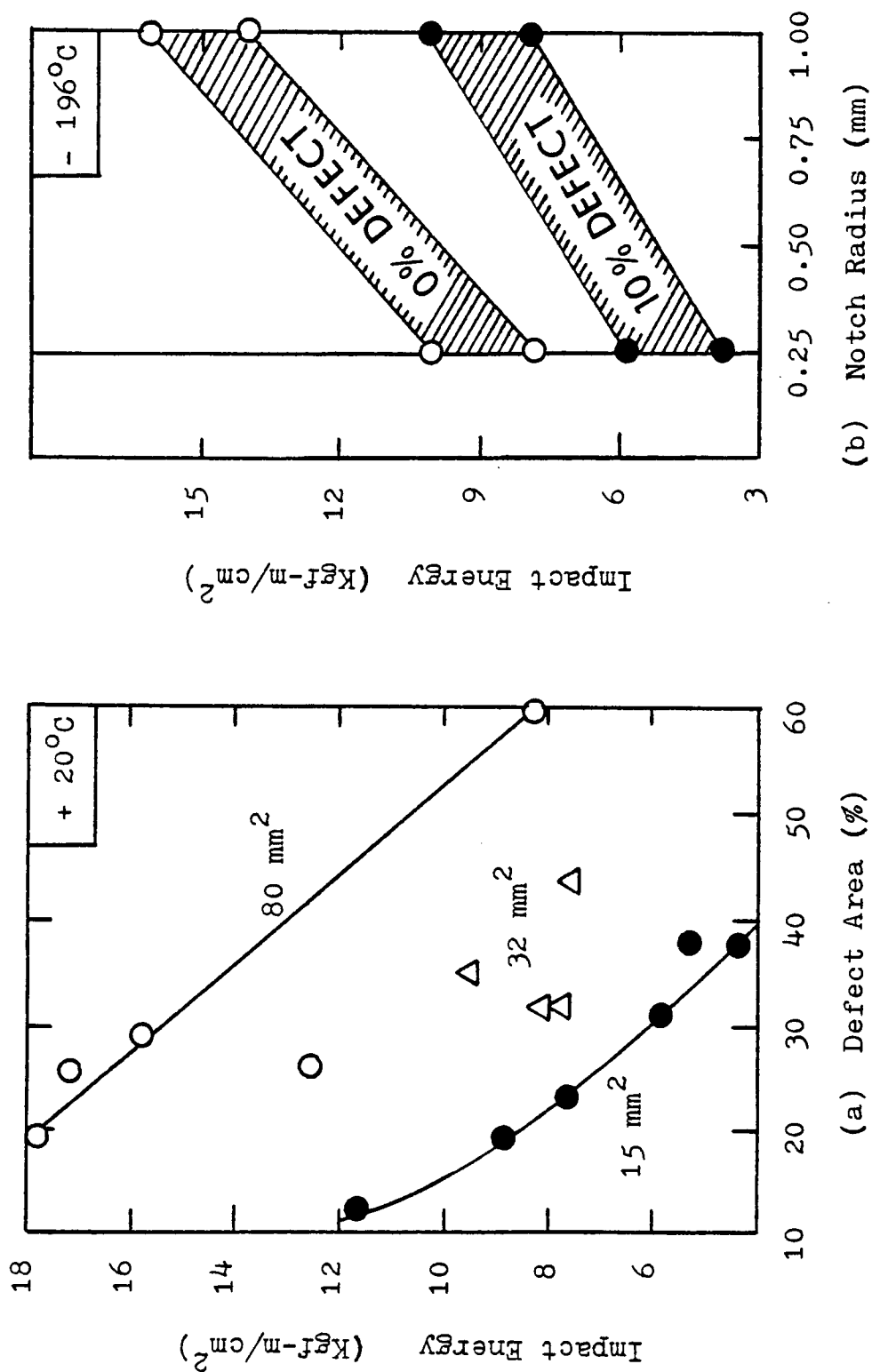


Figure 14. The Dependence of the Impact Toughness of the Weld in Steel 03Kh20N16AG6 on the Discontinuity Area and Notch Radius.
 (a) Discontinuity Area, with Specimens of Different Cross Section Areas.
 (b) Notch Radius, with Specimens of 0 and 10% Discontinuity Area.

rather than the influence of discontinuities. Zemzin and Stanjukovic (128) reviewed this area of concern and showed that the low ductility is due to metallurgical embrittlement during service especially for the stabilized stainless grades. Christoffel (129) reports that weldments of 316 are not notch sensitive at elevated temperature in any weld region.

Goodwin et al. (130) performed some 1200°F creep-rupture tests on as welded materials. They found that 308 weld deposits made with SMAW had higher rupture strengths than welds made with the same nominal composition filler material using the SAW process. It was pointed out that the submerged arc weld deposits had the coarser substructures and ferrite morphologies. Fracture paths were observed to follow the austenitic-ferritic boundaries almost exclusively.

Berggren et al. (83) investigated the structure and elevated temperature properties of Type 308 weld metal with varying ferrite content from 0 to 15%. Their results show that:

- (1) The nominally 0% ferrite weld had the highest tensile and creep strength of the welds investigated.
- (2) The ferrite phase morphology changes from isolated patches paralleling substructural boundaries with increasing ferrite content. Ferrite transformed to sigma phase during creep testing at 1200°F.
- (3) In creep tests at 1200°F, low ductility occurred for high-ferrite welds in relatively long time (approaching 1000 hr) tests.
- (4) Metallographic and fractographic observations are consistent with fracture initiating at austenite-sigma phase boundaries.

In studying the effect of δ ferrite on the creep-rupture properties of austenitic 316 weld metals at 600°C, Thomas (131) found that the optimum combination of ductility and rupture life appears to occur at about 5% δ ferrite. In low ferrite microstructures, fracture is intergranular, crack initiation occurs early and crack propagation and linkage will account for most of the creep life. The presence of about 5% δ ferrite reduces the effective grain size of the weld metal, and creep deformation can now be borne by the austenite grain interiors and also off-loaded onto the δ ferrite which offers less creep resistance.

Using the creep crack growth tests with T-type wedge opening loading specimens of Type 316 stainless steel weld metal, Haigh (132) found that creep cracking can best be avoided by postweld heat treatment, by keeping delta ferrite below about 4% or by depositing small weld beads.

9E. Effect of Discontinuities on Fatigue Behavior

Microfissures or cracks in austenitic materials generate the most alarm when fatigue behavior is being considered. Yeniscavich's work (133) is most often cited in regard to the influence of fissures on fatigue and rightly so, being the first with definitive data. Later Honeycombe and Gooch (121) highlighted the influence with particular reference to stainless steel.

Yeniscavich found that fissures have little effect on the fatigue life of Inconel weld metal if they constitute less than five percent of the load carrying area. Fissure density does not influence fatigue life if within the five percent area reduction. In addition, unfused weld root areas have no measurable effect on fatigue behavior when within

the five percent limit. Low cycle fatigue behavior was superior to high cycle behavior for equivalent fissure areas.

Honeycombe and Gooch (121), in testing 310 and 316 weld metals with reinforcement intact, report no influence of microcracks on fatigue life. All fatigue failures initiated at the toe of the weld and propagated in the HAZ. With the reinforcement removed the microcracks became more influential in the initiation of fatigue cracks but a loss in fatigue strength was not evident.

In an audit study by Battelle (134) actual samples were obtained from piping and strain controlled fatigue tests were conducted on weld metal containing microfissures. The weld metal, generally of low ferrite level (0-1.5 FN), exhibited fissure densities, as measured on polished and etched sections of the welds, averaging approximately 30 fissures/inch² with the average fissure size being approximately 0.015". The fissures were in the HAZ underlying a given weld pass in "ferrite free" areas, with no fissures extending to the outer weld surface. The fatigue specimens were 0.252" diameter reduced section bars. The results of the specimens tested are shown as triangles on the stainless steel design curve in Figure 15. Southwest Research Institute (122) in a study of 316 stainless steel weldments also measured fissure sizes and densities on polished and etched weld sections, and conducted bend tests and fatigue tests. The average ferrite content was less than 2.5 percent. The fissure density ranged upward to 25/inch² and the average size was approximately 0.014". The fissures were again found in ferrite free areas in HAZ's of weld metal. The bend test specimens exhibited a large maximum fissure size than the polished and etched section. The fatigue

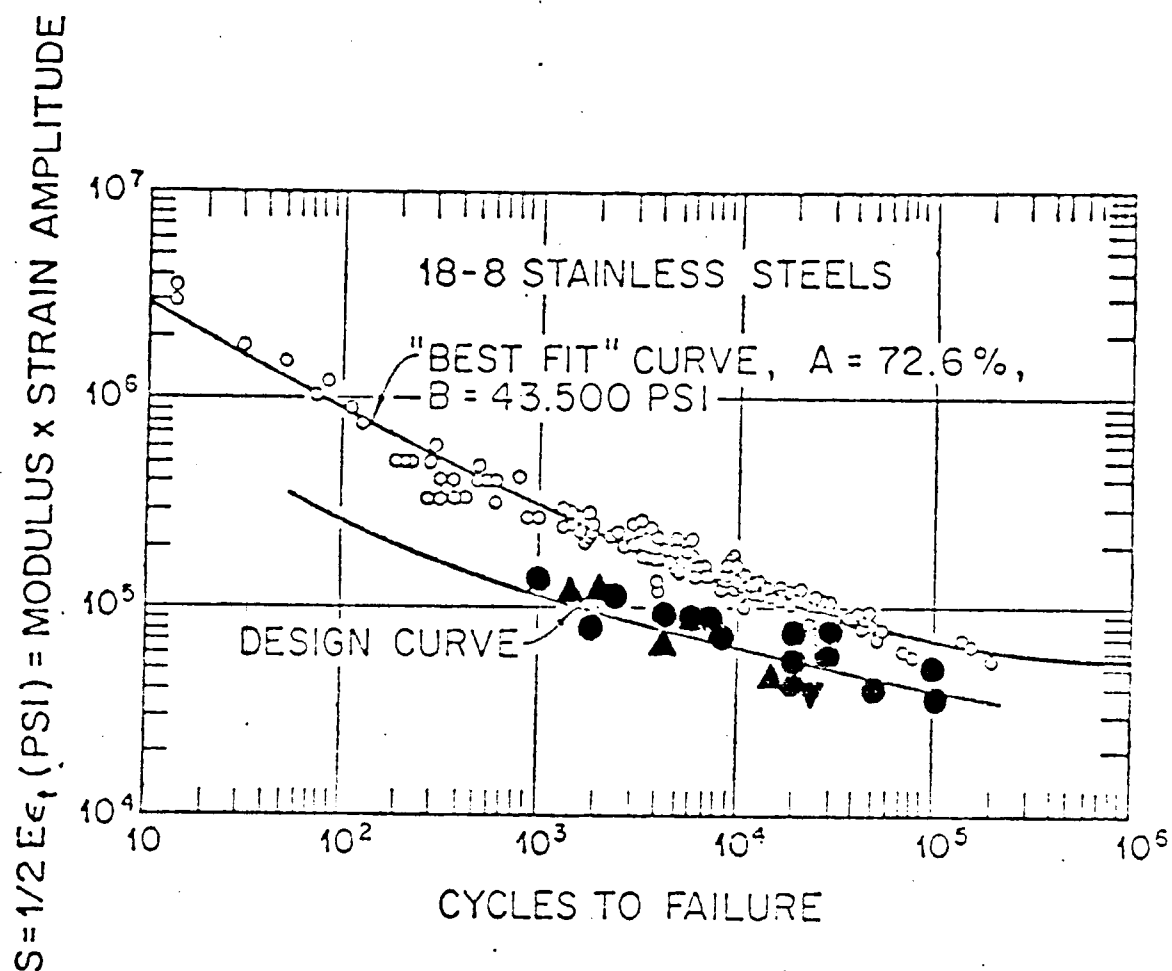


Figure 15. Fatigue Design Curve of Stainless Steel, ASME III, 1974.

specimens were polished and of rectangular cross section (0.825" x 0.375"). The results of the tests are plotted on Figure 15 as the solid circles.

Their report concludes that:

- (1) It was not possible to distinguish between fissure free (high ferrite) and fissure containing welds on the basis of fatigue tests.
- (2) Fatigue crack initiation sites were microfissures, specimen corners, and acceptable weld discontinuities (such as inclusions) with no dominance of one discontinuity over another.

The effect of discontinuities on the fatigue life of the welded joints in austenitic stainless steel 18-10 at 20°C was studied in the alternating bending of plate specimens by Kudryavtseu and Yushchenko (125). Discontinuities such as pores, slag inclusions, and LOP were produced artificially and examined by radiography. The dependence of the number of cycles N to failure on the discontinuity size F_d is shown in Figure 16. Analysis of the test results also shows that the rate of failure is considerably higher if the discontinuity is positioned in the vicinity of the weld surface, or if the discontinuity extends to the surface.

Recently, APTECH Engineering Services (135,136) investigated the significance of LOP discontinuities in SA-312 Type 304 stainless steel piping. Base on flaw inspection data obtained visually, the maximum observed flaw size was 26% of the pipe wall, and the maximum observed flaw eccentricity was 30% of the wall. A fatigue model and assessment procedure was developed to evaluate the extent of subcritical crack

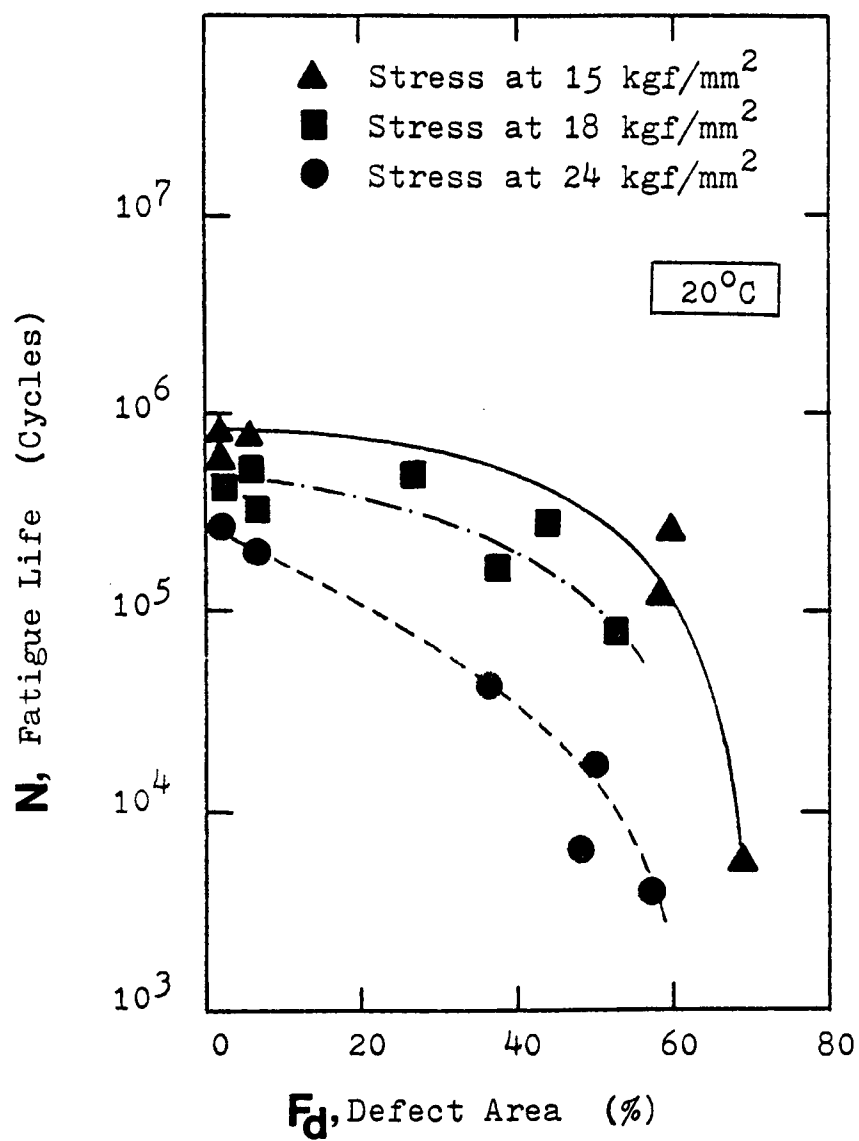


Figure 16. The Fatigue Life of Welded Joints of 12Kh18N10T with Discontinuities.

propagation under service conditions. Although an important factor on the calculated fatigue life was flaw eccentricity, especially when the worst case flaw was assumed for the initial flaw size, the anticipated subcritical crack extension in the design life time is negligibly small for class 2 and 3 piping systems (Nuclear ASME III). The safety factor on burst load based on the load required to fail the pipe by fully plastic collapse is calculated to be greater than four even when the worst case flaw is assumed.

The work of Grudzinskii (124) shows that LOF in stainless steel is detrimental at low temperatures where austenite transforms to martensite during fatigue. Above the austenite instability temperature (M_d) LOF is only detrimental in regard to its cross sectional area reduction. James and Jubb (137) report that the fatigue strength of austenitic stainless fillet welds increases significantly with decreasing temperature below room temperature. All fatigue failures initiated at the toe of the weld as anticipated, again illustrating the influence of weldment geometry and stress concentration.

9F. Effect of Discontinuities on Stress Corrosion Cracking Behavior

Stress Corrosion Cracking (SCC) is a general term describing stressed alloy failures that occur by the propagation of cracks in corrosive environments. SCC has the appearance of brittle fracture, yet it can occur in highly ductile materials. The requirements for SCC to occur are the presence of a tensile stress, either residual, applied, or a combination of both, and the presence of a specific corrosive agent. The cracks form and propagate roughly at right angles to the

direction of the tensile stress at stress levels much lower than those required to fracture the material in the absence of the corrodent.

Stress-corrosion cracking in austenitic stainless steels has been a problem of considerable importance to the nuclear power and chemical industries. Numerous studies have been performed to determine the characteristics of chloride-induced stress corrosion cracking phenomena in both sensitized and unsensitized austenitic stainless steels.

Although SCC in these alloys has been extensively investigated and well documented, the area concerning the SCC properties of ferrite containing weld metal remain poorly understood and highly controversial.

In studying the effect of ferrite content on stress corrosion cracking in duplex stainless steel weld metals, Baeslack et al. (138) made weld metals with ferrite contents ranging from FN=1 to FN=24 by using Type 310 or 312 filler and type 304 base plate. Their results show that the type 304 base metal cracks transgranularly and is less susceptible than the duplex type 304 weld metal, which cracks by the combined occurrence of stress assisted, preferential dissolution of the ferrite and SCC of the austenite. The most susceptible duplex weld metals are those which exhibit a continuous or nearly continuous vermicular ferrite structure.

Baeslack et al. (139) investigated the effect of nitrogen content on the SCC of stainless steel weld metal (18Cr-8Ni). With nitrogen content increasing from 0.04% to 0.24%, the FN drops from six to zero. Their conclusions are:

- (1) Stress corrosion cracking resistance in $MgCl_2$ appears to decrease with increases in nitrogen content and decreases in ferrite level.

- (2) Stress corrosion cracking morphologies observed in weld metals are closely associated with weld metal microstructural characteristics.
- (3) Weld metal which solidifies as primary ferrite exhibits a SCC resistance similar to that of wholly-austenite type 304 base metal. SCC propagation occurs by an interphase/interface mode along the austenite-ferrite boundaries in the weld metal substructure.
- (4) Duplex weld metals which solidify as primary austenite exhibit similar SCC susceptibilities. SCC occurs by a mixed trans-granular-intergranular mode.
- (5) Wholly-austenitic weld metals are highly susceptible to SCC by a predominantly intergranular mode. The large grain size promotes crack propagation.

Gooch (140) has discussed factors determining the risk of SCC attack and the conditions under which practical problems may arise. In principle, the problem may readily be avoided either reducing the tensile stress operative in causing cracking or avoiding the anions involved in initiation and propagation. Although reduction in residual stresses is to be preferred, relief of such stresses by post weld heat treatment does not guarantee avoidance from cracking and is often impracticable. The problem is best controlled by attention to those design and operation features which encourage the concentration of aggressive chemical species.

9G. Conclusions

The assessment of weld discontinuities in regard to code requirements is one of the most difficult areas to be addressed. Under ASME III, a fatigue design curve is developed from smooth specimen tests of base material and a "best fit" to the data obtained. The data for stainless steel are shown in Figure 15 (open circles represent base metal tests). Note that the base metal points fall both above and below the

curve and it is truly a "best fit" curve and not a lower bound. The design curve is obtained by applying a factor of two on stress or 20 on cycles whichever is more conservative. If we treat the present weld fatigue data together without prejudice it will be noted that the data fall around the design curve. Thus, this design curve does not guarantee safety for stainless steel weld metal in actual service.

In summary, from the review of currently available information, the following conclusions are obtained:

- (1) Microfissures, usually fine and short, do not have significant effect on the mechanical behavior during service.
- (2) Delta ferrite, which is required by NRC to avoid hot cracking problems, has an adverse influence on high temperature creep-rupture properties, low temperature toughness, and may influence stress corrosion cracking resistance.
- (3) Small internal discontinuities, such as slag inclusions, LOP/LOF, etc. do not have significant influence on the mechanical behavior as long as the extent of the area reduction is no greater than 5% of the cross section under consideration.
- (4) The need for design data on weld metals and weldments as a whole is certainly emphasized and necessary for evaluation of actual welds.

10. Closure

In light of the above discussions, it is clear that weld metal HAZ cracking is important in regard to the mechanical behavior of the weld. With the increasing use of heavy section weldments (multipass and

repair welding), the understanding of the effect of multiple thermal cycles on weld metal fissuring is necessary. Numerous investigations have been concerned with the compositional effect on the hot cracking from the solidification view point. However, no extensive quantitative evaluation of the effect of multiple thermal cycling on fissuring tendency has been undertaken. Therefore, the investigation on the influence of the alloying elements and multiple thermal cycling on the weld metal HAZ hot cracking resistance is needed.

In actual weldability tests, the inseparability of mechanical and thermal factors and the narrow HAZ induced by deposition provide limitations on testing methods. The Gleeble Hot Cracking Test developed by Krantz has the ability to separate the effects of mechanical and metallurgical variables. In addition, the Gleeble can produce a wide, uniform, hot zone in a small size specimen which is easily prepared and tested.

Additional techniques are needed for rapidly assessing the effect of specific alloying elements on the hot cracking sensitivity of austenitic weld metals. These techniques should have the following features: (1) reproducibility, (2) ease of operation, (3) small sample size, and (4) low cost.

Some of the objects of this research are to; (1) develop a method to produce laboratory-made alloys with variations in composition, (2) evaluate the hot ductility and hot cracking behavior of commercial weld metals in addition to experimental compositions, (3) test separately the influence of strain and thermal history on the hot cracking tendency, and (4) propose a theory on the "Reheat Hot Cracking" of austenitic stainless steel weld metals.

CHAPTER III

EXPERIMENTAL PROCEDURE

1. Materials

In this study seven austenitic stainless steel commercial consumables in the bare wire form were used as filler metal. They are AWS ER308, ER316L, ER310, modified 316L, and modified 310 type wires. The modified consumables are distinguished by higher Mn and/or higher Mo when compared to the standard consumables. The ferrite potential of the deposited weld metals ranged from 0 FN (fully austenitic) to 10 FN. The diameter of the bare wires ranged from 0.030" to 0.063". One of the bare wires was also used as an electrode in the GMAW process.

Five austenitic stainless steel SMA coated electrodes were also used in this research. Four of them are of the fully austenitic type which are AWS E310, modified E310, and modified E316L with AC-DC type coating. The fifth one is E316L-15 with a ferrite potential of 0.5 FN in the deposited weld metal. All the electrodes were 1/8" diameter and 10" long.

Two 1/2" thick austenitic stainless steel plates were used as the base plates in the Fissure Bend Test. They are AISI Types 304 and 304LN with ferrite content of 5.0 FN and 6.3 FN respectively in the autogenously as-welded condition.

The chemical composition and measured FN of the weld metals deposited from the commercial consumables by GTAW or GMAW process are shown in Table 4. W1 to W7 are the weld metals deposited from the bare wires by GTAW process. E1 to E5 are weld metals deposited from the coated electrode by SMAW process. B1 and B2 are the base plates.

Table 4. Chemical Composition and Measured Ferrite Number of Austenitic Stainless Weld Metals Used in Fissure Bend Test Evaluation.

Weld Metal	Chemical Composition, Wt.%								Ferrite	
	C	Si	Mn	P	S	Cr	Ni	Mo	N	Number
W1	0.014	0.15	4.57	0.009	0.009	25.09	21.98	2.12	0.06	0.0
W2	0.017	0.44	2.20	0.009	0.007	18.50	15.20	2.80	0.06	0.0
W3	0.013	0.36	5.26	0.011	0.014	19.03	16.74	2.57	0.13	0.0
W4	0.093	0.37	1.49	0.013	0.009	27.18	20.27	0.09	0.06	0.2
W5	0.110	0.53	1.85	0.015	0.008	26.32	20.81	0.06	0.07	0.0
W6	0.016	0.80	1.70	0.010	0.010	20.00	10.36	0.07	0.06	10.0
W7	0.020	0.40	1.80	0.015	0.015	19.00	11.50	2.20	0.06	9.0
E1	0.032	1.00	5.20	0.019	0.000	18.50	17.40	2.60	0.08	0.0
E2	0.025	0.28	1.94	0.021	0.012	17.80	13.00	2.22	0.08	0.5
E3	0.100	0.26	1.55	0.032	0.016	26.14	20.99	0.40	0.08	0.0
E4	0.030	0.40	4.80	0.015	0.015	25.00	21.80	2.10	0.13	0.0
E5	0.140	0.59	4.15	0.026	0.008	24.80	19.70	0.56	0.06	0.0
B1	0.022	0.56	1.66	0.024	0.004	19.02	8.25	0.35	0.13	5.8
B2	0.041	0.57	1.73	0.017	0.007	18.11	8.19	0.20	0.08	5.2

In order to study the effect of nitrogen in shielding gas on the microstructure of weld metals, eleven austenitic stainless steel sheets in the thickness range of 0.020" to 0.5" were used. The chemical composition, calculated FN, and thickness of the stainless steel materials are shown in Table 5. S1 through S5 are Type 304 austenitic stainless steel with variations in C, Si, and S contents. S6 is a modified 308L strip, S7 and S8 are Types 16-8-2 and 317LM respectively. P1, P2, and P3 are $\frac{1}{2}$ " thick austenitic stainless steel plates of 304, 304LN and 317L respectively.

In order to determine the effect of alloying elements on the hot cracking resistance of weld metal, thirteen austenitic stainless steel materials varied in composition; Mn, Mo, Si, and N were used for the Gleeble Hot Cracking Test. The chemical composition and calculated FN of the weld metals used in this study are shown in Table 6. All the weld metals except B2 and B3 are fully austenitic.

2. Welding Equipment and Fixtures

In this study SMAW, GMAW, and GTAW processes were used to deposit weld metals for the Fissure Bend Test. The welding conditions used in the Fissure Bend Test are shown in Table 7. In SMAW process, a 300 Ampere SMA welder (Hobart TG301 AC-DC GTA/SMA) was used to deposit weld metal from a covered electrode. In GMAW process, a 300 Ampere constant potential GMA welder (linde VI-303A/SMWII GMA) was used to deposit a bare filler wire on a $\frac{1}{2}$ " thick plate. In GTAW process, an Airco SAMPAC precision system with a three phase DC, 750 Ampere function-controlled welding power source was used for the GTA welding. Since the filler metal wires used for evaluation varied in diameter, the

Table 5. Chemical Composition of Austenitic Stainless Steels Used in Studying the Nitrogen Effect.

Material	Thickness	Wt %							FN	
		C	Si	Mn	P	S	Cr	Ni	Mo	N (Calculated)
S1 (304)	0.025"	0.066	0.41	1.68	0.029	0.010	18.00	8.30	0.61	0.083
S2 (304)	0.031"	0.048	0.43	1.69	0.021	0.005	18.36	8.11	-----	-----
S3 (304)	0.035"	0.078	0.60	1.58	0.025	0.004	18.47	8.60	-----	-----
S4 (304)	0.035"	0.067	0.44	1.62	0.023	0.010	18.67	8.17	0.16	0.070
S5 (304)	0.037"	0.057	0.57	1.56	0.019	0.010	18.70	8.17	0.14	0.076
S6 (308L)	0.025"	0.012	0.25	1.54	0.011	0.010	23.40	11.36	0.03	0.017
S7 (16-8-2)	0.020"	0.070	0.43	1.58	0.014	0.015	16.18	8.81	1.85	0.040
S8 (317LM)	0.020"	0.018	0.59	1.02	0.014	0.020	18.05	14.99	4.21	0.045
P1 (304L)	0.500"	0.022	0.41	1.81	0.025	0.008	18.10	9.20	-----	-----
P2 (304LN)	0.500"	0.022	0.56	1.66	0.024	0.004	19.02	8.25	0.35	0.130
P3 (317L)	0.500"	0.019	0.59	1.62	0.020	0.018	18.74	13.95	3.22	-----

Table 6. Chemical Composition and Measured Ferrite Number of Austenitic Stainless Steel Weld Metals Used in the Gleeble Hot Cracking Test Evaluation.

Weld Metal	Wt. %										Measured FN
	Cr	Ni	Mn	Mo	Si	C	P	S	N		
A1	24.34	21.49	4.49	2.16	0.06	0.021	0.007	0.014	0.160	0.0	
A2	20.49	15.95	9.43	2.57	0.22	0.016	0.010	0.017	0.078	0.5	
B1	26.56	21.34	1.85	0.09	0.51	0.090	0.015	0.015	0.064	0.0	
B2	16.18	8.81	1.58	1.85	0.43	0.053	0.014	0.015	0.040	2.2	
B3	18.05	14.99	1.02	4.21	0.59	0.018	0.014	0.020	0.045	2.5	
D1	25.42	21.40	2.46	0.63	0.40	0.079	0.012	0.009	0.086	0.0	
D2	19.49	13.87	2.52	1.65	0.31	0.033	0.010	0.014	0.093	0.0	
D3	20.55	14.10	1.69	0.95	0.46	0.070	0.014	0.015	0.052	0.0	
A1Si	21.90	20.36	4.09	1.97	0.29	0.021	0.009	0.016	0.120	0.0	
A2Si	19.12	15.19	8.06	2.31	0.44	0.022	0.010	0.017	0.074	0.1	
B3Ni	18.78	16.97	1.02	3.55	0.53	0.018	0.019	0.020	0.054	0.0	
B3N	17.52	14.47	1.01	3.57	0.53	0.015	0.019	0.020	0.180	0.0	
B2N	16.16	8.87	1.60	1.84	0.45	0.039	0.015	0.016	0.230	0.0	

Table 7. Welding Parameters for Fissure Bend Test.

Parameter	SMA	GTA
Electrode Diameter	1/8"	0.047" 0.063"
Polarity	DCRP	DCSP
Current	90-100 Amps.	180 Amps.
Voltage	25 V.	10.8 V.
Travel Speed	8 ipm	8 ipm
Interpass Temp.	200°F	250°F
Gas Flow Rate	-----	Ar, 25 cfh
Heat Input	16-17 KJ/in	15 KJ/in

wire-feed rate was adjusted to deposit the same amount of wire per inch of weld. The shielding gas used in the GTA welding are either pure argon or argon plus nitrogen.

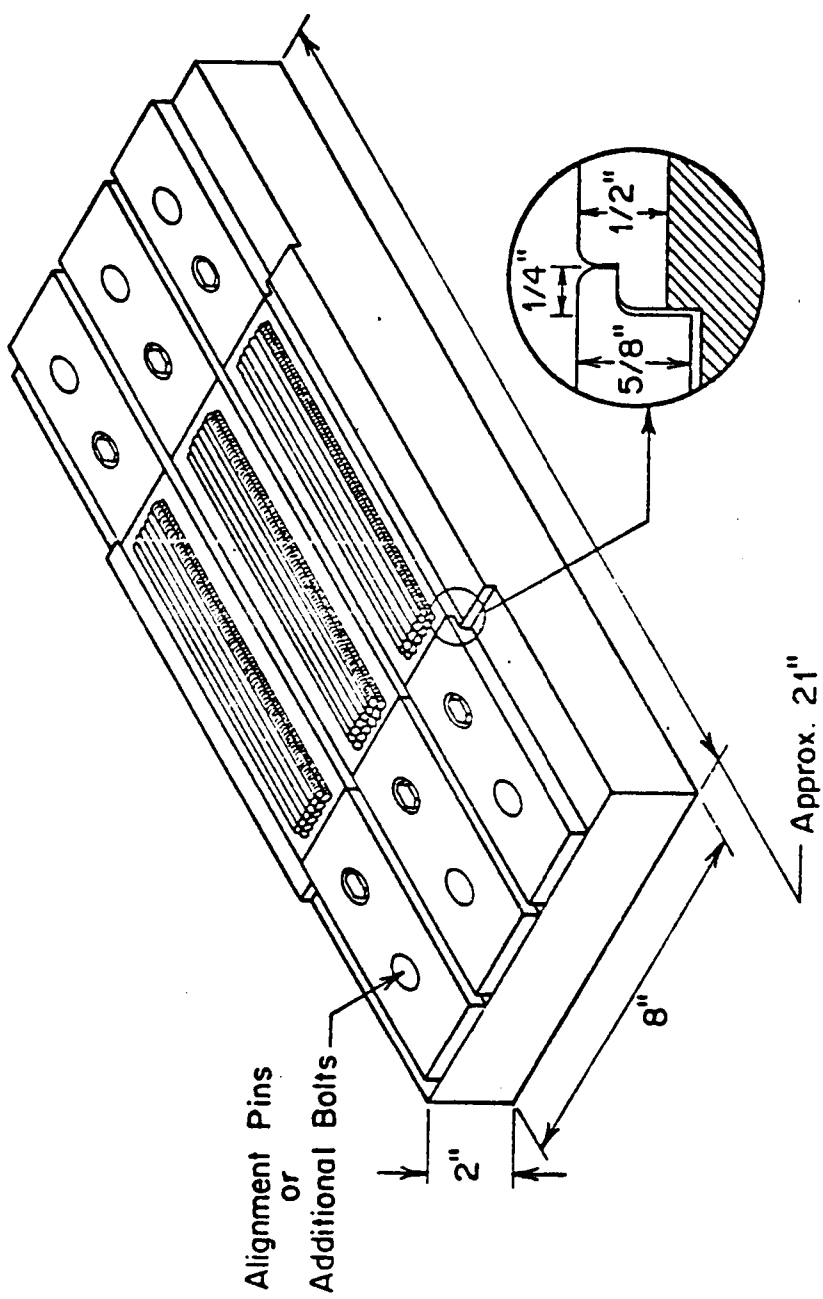
In this investigation, a number of GTA autogenous remelted welds were made. In order to provide restraint and adequate gas protection, two specially designed fixtures were utilized. Figure 17 shows the fixture for clamping the $\frac{1}{2}$ " thick plate. This fixture was also used for depositing weld metal by the SMAW, GMAW, and GTAW process. A second fixture, used for welding thin sheet materials, is shown in Figure 18. This fixture consisted of a one-inch thick copper plate and two aluminum hold down plates. The copper plate had a half-inch wide groove machined so that a backing gas could be used when full penetration welds were made.

3. New Alloy Development

A dual-wire device was designed and developed. This dual-wire system was used when variations in weld metal deposit chemistry were desired and a specific chemistry was obtained by varying either the wire feed speeds or wire compositions. This miniature casting technique for new alloy development is economical and time-saving. Gleeble Hot Cracking Tests were conducted using several alloys developed by this technique. The details of this technique are described later.

4. Ferrite Measurement

Ferrite Number (FN) measurement were made on all weld metal using the Aminco-Brenner Magne Gage. The instrument was calibrated to AWS A4.2 - 74 (141). In order to minimize the error due to weldment



Clamps - 6 required - 5/8" x 2" x 6" Heat treated to prevent deformation
Bolts - 6 required - Allen head cap screw, 1/2" - 13
Backup plate - Carbon steel 2" thick (min.)

Figure 17. Recommended Clamping Fixture for Fissure Bend Test.

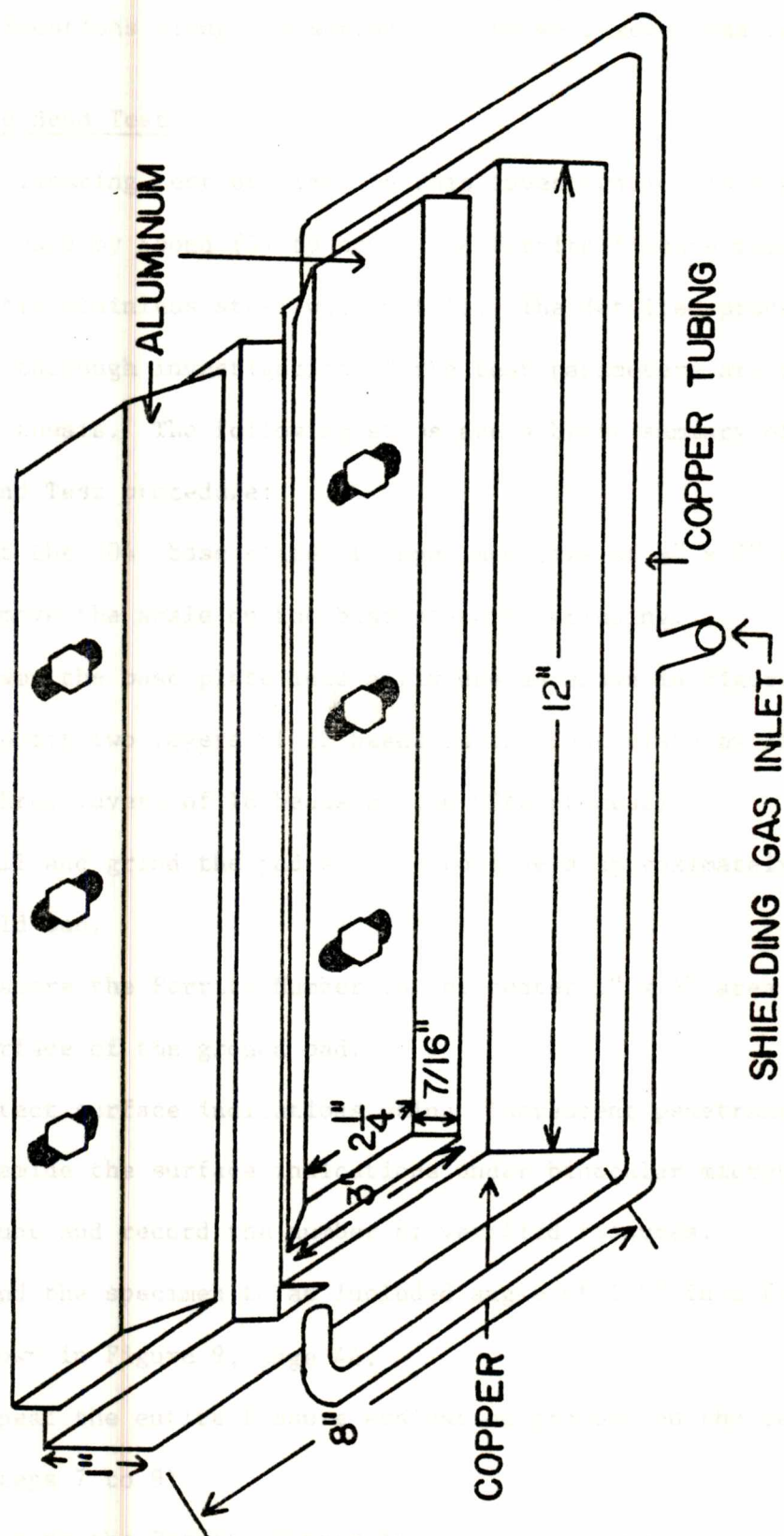


Figure 18. Weld Fixture for Welding Thin Sheet Materials.

inhomogeneity, the average value of several determinations taken from different locations along the surface of the weld metal was recorded.

5. Fissure Bend Test

The fissuring test utilized in this investigation is the "Fissure Bend Test" used by Spond (3) to study the ferrite-fissure relationship in austenitic stainless steel weld metals. The detailed procedures and a more thorough investigation of the test parameters are presented in Spond's thesis. The following steps are a brief summary of the Fissure Bend Test procedure:

1. Cut the 304 base plate to specimen size of $\frac{1}{2}$ " x 2" x 9".
2. Remove the scale on the base plate by grinding.
3. Clamp the base plate into a fixture as shown in Figure 17.
4. Deposit two layers of 12 beads on the base plate by SMAW process (three layers of 18 beads by the GTAW process).
5. Mill and grind the pad surface to give a approximately $\frac{1}{4}$ " x 1" x 8" weld pad.
6. Measure the Ferrite Number in the center 1" x 4" area on the surface of the ground pad.
7. Detect surface indications using fluorescent penetrant testing.
8. Examine the surface indications under binocular microscope.
9. Count and record the number of verified fissures.
10. Bend the specimen to an included angle of 120° in a fixture as shown in Figure 9, page 44.
11. Repeat the entire fissure evaluation process on the bent specimen (steps 7 to 9).
12. Measure the Ferrite Number in the center 4 inches of the bent pad.

6. Gleeble Hot Ductility Test

Hot ductility tests were conducted using the Gleeble. The hot ductility test consists of performing tensile tests on specimens heated in accordance with programmed (simulated) weld thermal cycles. The testing was conducted during both the heating and cooling portions of the cycle. All tests were conducted in an inert atmosphere (argon purged chamber) to avoid high temperature oxidation.

The conditions of heating and cooling were selected to simulate an energy input of 70,000 joules per inch on a 1 ½ inch thick stainless steel plate (standard hot ductility test conditions). A schematic representation of the testing sequence used in this study for both the on-heating and the on-cooling tests is shown in Figure 19.

The geometry of the specimen used in the hot ductility test and the temperature distribution along the sample are shown in Figure 20. Since the temperature is essentially uniform in the center 0.25 inch length, the result of the hot ductility tests obtained with the transverse composite weld specimen was only from the weld itself. A second type of composite weld specimen was also used in this study. Figure 21 shows the dimensions of the longitudinal composite weld specimen and the temperature distribution along the sample. The reason for use of this type of specimen is to correlate the hot ductility test results with hot cracking sensitivities and to conduct simultaneous evaluations of entire weld region which includes fusion zone, HAZ, and base metal.

Since 80% of the specimen deformation occurred within a 0.25 inch length in the center of the specimen gage length, the elongation was calculated as; $\% \text{ elongation} = 0.8 \times \Delta l / 0.25 \times 100$. Because specimens

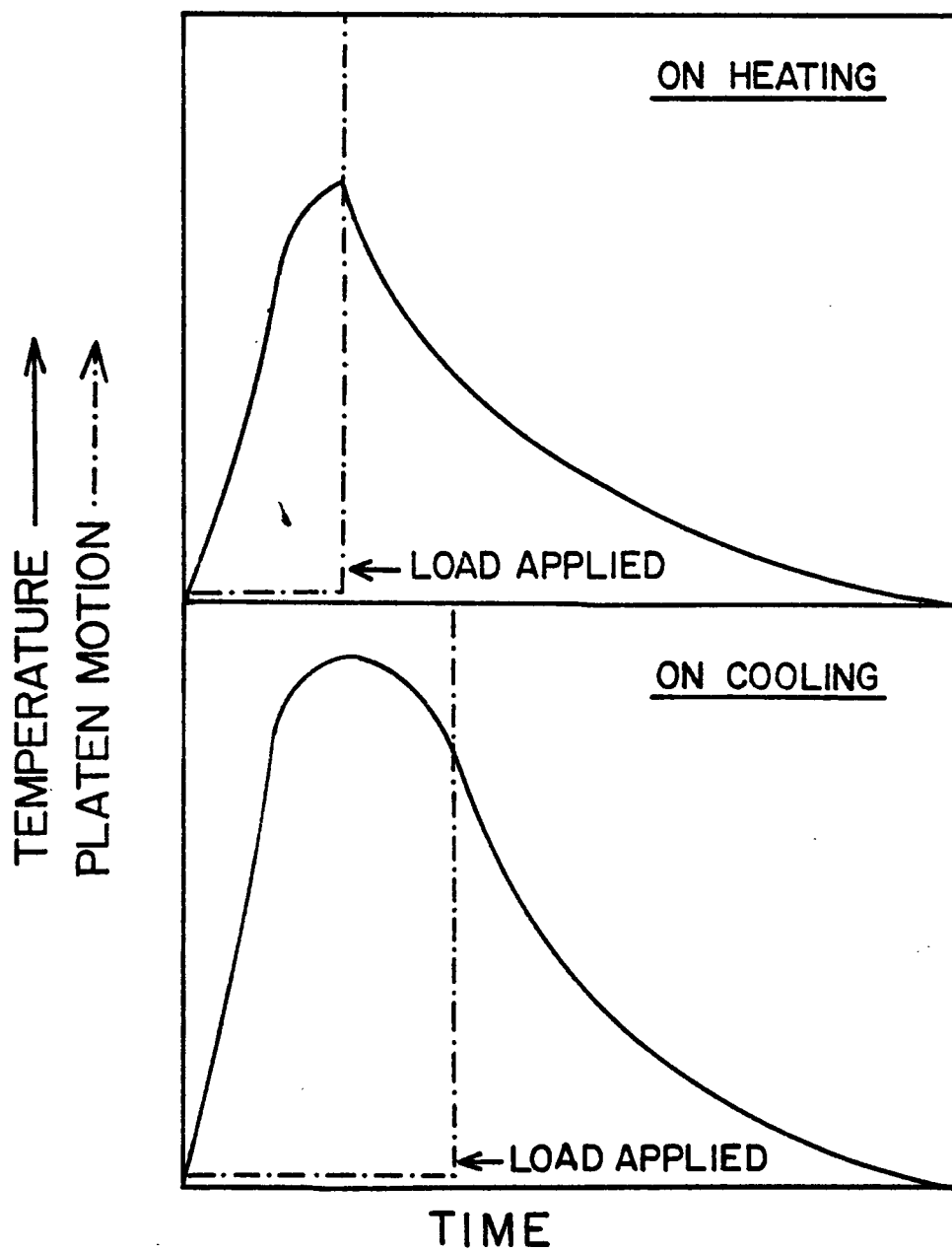


Figure 19. Schematic Representation of Hot Ductility Testing Sequence.

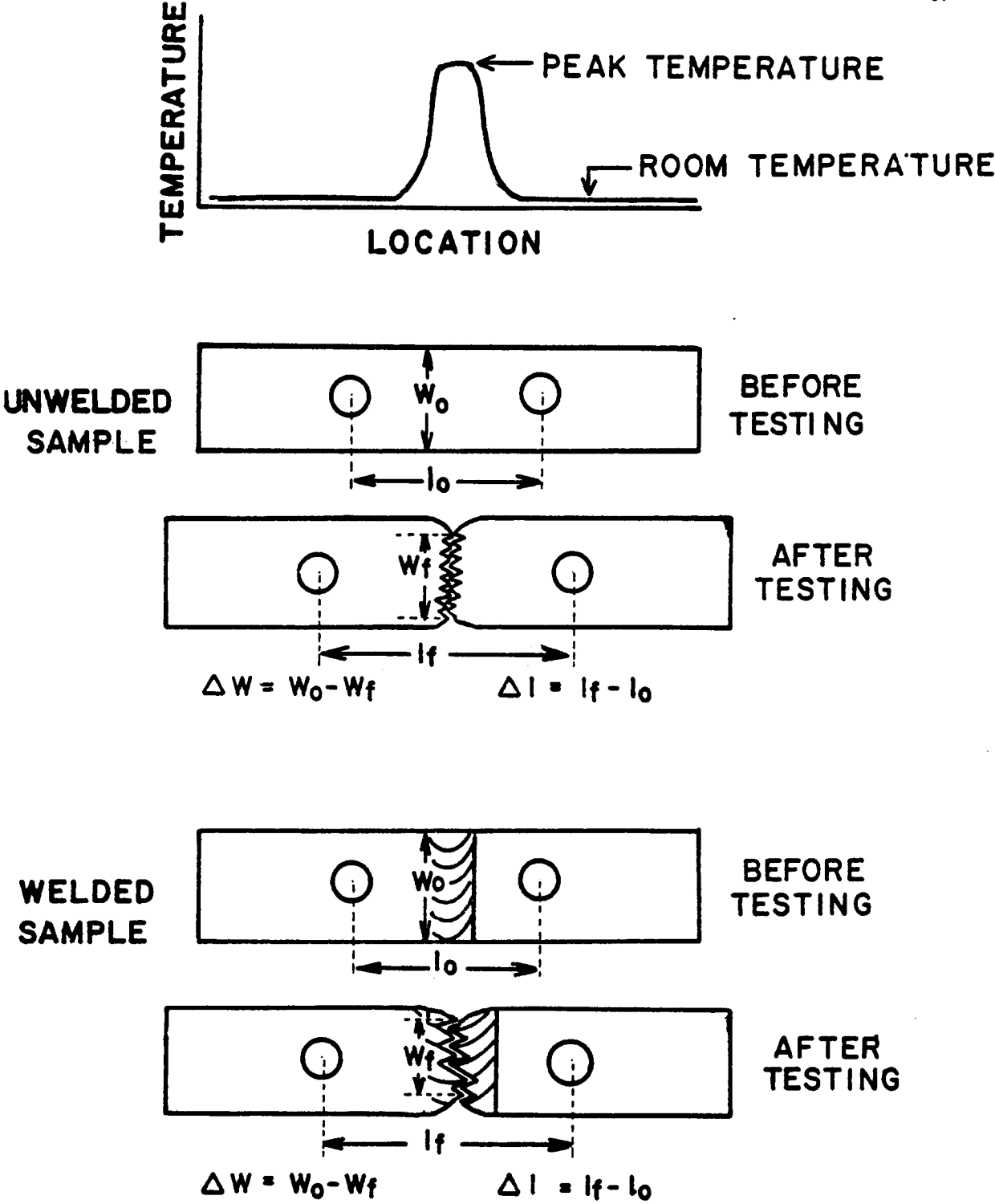
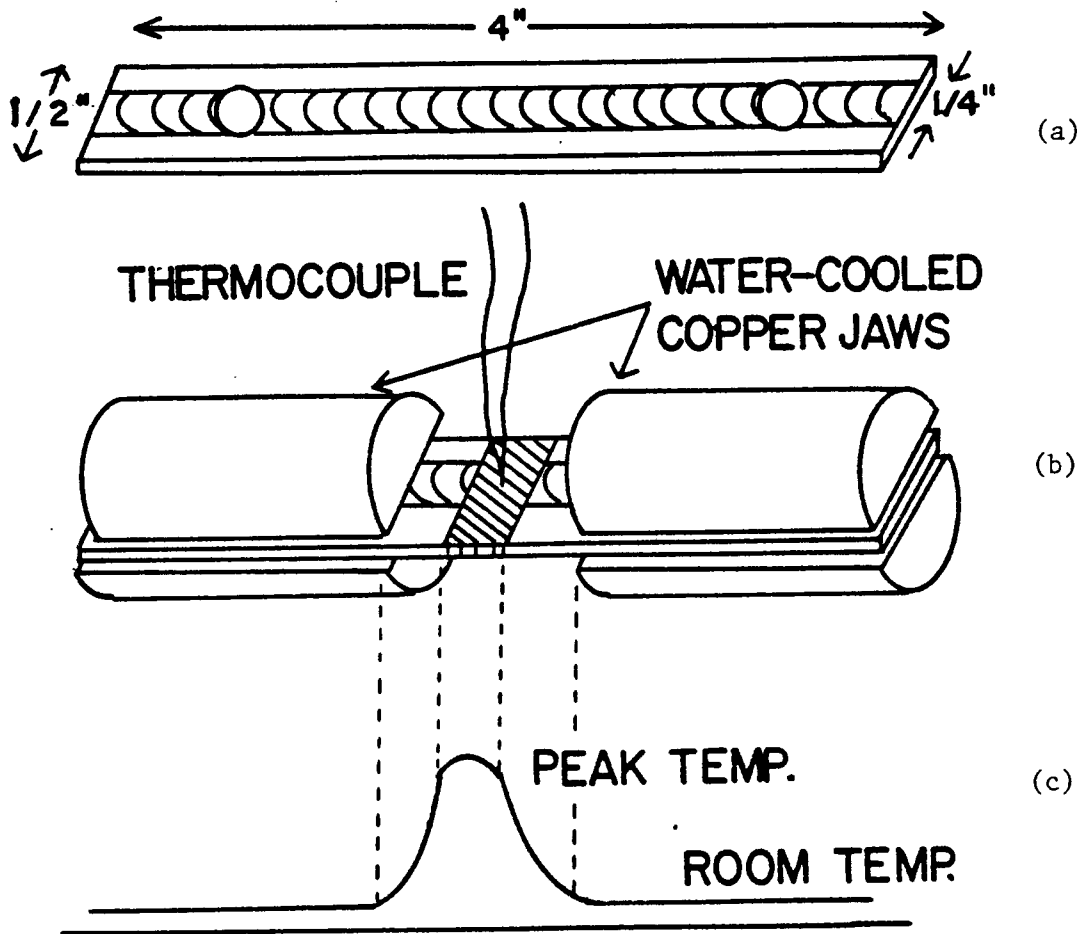


Figure 20. The Geometry and the Temperature Distribution for Unwelded and Transversely Welded Test Samples.



LONGITUDINAL COMPOSITE WELD SPECIMEN USED IN GLEEBLE

Figure 21. Dimensions and Temperature Distribution in the Longitudinal Composite Weld Specimen.

used were thin (0.020"), the reduction of area calculated as the width change; $R.W. = (\Delta w/w_o) \times 100$. The measurement of Δl and Δw are shown in Figure 20.

7. Gleeble Hot Cracking Test

A testing procedure was adopted from Krantz's work (117) and modified for studying the hot cracking sensitivity of weld metals in strip form. Stainless steel sheet was cut into several $\frac{1}{2}$ " wide and 10" long strips. The strip was then cleaned with Scotch scouring pads and degreased with acetone. Autogenous GTA weld bead was made on the strip which was clamped in the fixture shown in Figure 18. The fixturing consists of a one-inch copper backing plate and two aluminum hold-down plates. A trailing shield was used on all welds to protect the weld area from oxidation and contamination. After cooling to room temperature, 4" x 0.5" x 0.02" test specimens were sheared as shown in Figure 21a. To prevent slippage during testing, a 0.180 in. diameter hole was made at both end of each specimen to accomodate 0.180 in. locator pins in the copper jaws of the Gleeble.

A chromel-alumel thermocouple of 0.010 inch diameter was percussion welded to the center of each specimen using the welding equipment provided with the Gleeble. The Gleeble used in this study is model 510 with reference generator model 734 and model 766 programmer. The test specimens were clamped in the Gleeble with a 1.0 inch free span between the water-cooled copper jaws as shown in Figure 21b. The central portion of the specimen was subjected to a thermal cycle which was selected to simulate an energy input of 70,000 joule per inch on a 1.5 inch thick stainless steel plate. A controlled extension was introduced

immediately upon reaching the peak temperature and maintained until cooling to the room temperature. Since 80% of the specimen deformation occurred within a 0.25 in. length in the center of the specimen gage length and the temperature in the present tests was essentially constant in this region, the actual extension strain was calculated as follows:

$$\text{Extension strain} = 0.8 \times (\text{Extension}) / 0.25$$

All tests were conducted in an inert atomosphere (argon purged Chamber) to minimize high temperature oxidation.

After cooling to room temperature, the test specimen was removed and examined under binocular microscope up to 70X. By subjecting a series of identical test specimens to various amounts of strain at each of several testing temperatures, a quantitative evaluation of the hot cracking propensity was obtained.

8. Optical Metallography

Selected samples were cut, cleaned, and mounted using Buehler plastic epoxy. After mounting, the specimens were ground and polished to 0.06 μ . Specimens were then electrolytically etched using a stainless steel cathode at 7-9 volts for approximately 60 seconds. The etching solution consisted of 25 grams CrO_3 , 133 ml glacial acetic acid, and 7 ml distilled water. The specimens were examined and photographed under a Leitz optical microscope using a polaroid 55 film.

9. Scanning Electron Microscopy

Fractographic examination was performed using the AMR 900 Scanning Electron Microscope. Fracture Surfaces were cut from selected specimens

and cleaned with ultrasonic agitation in acetone. No gold coating was used on the metal specimen itself for the SEM work. However, for the metallographic specimen, the epoxy area was coated with a silver paint to provide conduction to the stage of the microscope. An ORTEC detector and energy dispersive multichannel analyzer were used for X-ray fluorescence analysis with the electron beam limited in the desired spot.

CHAPTER IV

RESULTS AND DISCUSSION

1. Introduction

Austenitic stainless steels are generally regarded as readily weldable materials, with considerable tolerance for variations in welding conditions without risk of cracking. However, fully austenitic weld metals may crack during deposition of a single pass and in underlying weld runs reheated by subsequent passes. It is well known that the composition of the consumables as well as the welding procedure has a significant effect on the hot cracking tendency of fully austenitic stainless steel weld metal. Furthermore, it has been long recognized that reheating of the material to temperatures in the vicinity of the solidus increases the fissuring tendency. However, no quantitative evaluation of the effect of multiple thermal cycling on fissuring tendency has been undertaken until recently by Lundin and Spond (9). With the increasing use of heavy section weldments, repair after fabrication and repair in service become more common. Thus, the understanding of the effect of multiple thermal cycles on weld metal fissuring becomes more important and draws more concerted attention.

It is recognized that the presence of ferrite in the room temperature microstructure indicates a reduced fissuring tendency for austenitic stainless steel weld metals. The ferrite level needed to virtually eliminate fissuring is depended upon the composition of weld metal as determined by Lundin et al. (10). More recently Masumoto et al. (11) proposed that the determination of weld metal fissuring should be based on

composition alone and not on ferrite content. Using the Fe-Cr-Ni phase diagram and weld metal of different compositions, they claim that fissuring depends on whether austenite or delta ferrite is the primary phase solidifying from the liquid. The alloys whose compositions nucleate austenite as the primary phase have a high sensitivity to fissuring. However, for the weld metals that nucleate delta ferrite as the primary phase, the sensitivity to fissuring is low even with high sulphur contents (≈ 0.15 percent). They believe that when the delta ferrite nucleates as the primary phase, it is able to dissolve larger amounts of sulphur and other fissuring-inducing elements thus preventing the segregation of these elements to the grain boundaries where they are most influential in causing fissuring.

It has been confirmed that manganese has beneficial effects in reducing hot cracking (12,13) and that the cracking susceptibility increases as the silicon, phosphorous, and sulphur content increases (14,15,16) in the fully austenitic stainless steel weld metals. The role played by manganese and molybdenum in hot cracking is not fully understood because of the lack of fundamental investigations and researches. With the improved technology in the refining process (14), low phosphorous and sulphur content can be obtained and hot cracking resistance improved. It is considered that silicon contents above 0.5% in 25/20 alloys cause a significant increase in cracking (18-21, 54), not only because the liquidus-solidus interval is widened, but also because intergranular silicate films may form thus embrittling the grain boundaries. To some extent, the effect of silicon can be compensated for by increasing the manganese content.

By utilizing Fissure Bend Test on SMA deposited pads, Lundin and Spond (9) found that the multiple HAZ thermal cycle exposure is a significant factor in the fissuring tendency of 308, 316 and 16-8-2 weld metals at low ferrite levels. In their work they more clearly defined the effect that multiple HAZ thermal cycling has on the fissuring tendency by remelting the surface of SMA deposits of both types 316 and 308 (0.6 FN) using overlapping GTA beads. No fissures were found upon bend testing along interpass boundaries which were subjected to only a single HAZ experience. Only when a weld bead underwent double or triple HAZ experiences did any fissures appear upon bend testing, thus confirming the results of the directly deposited SMA weld bead studies.

The fact that multiple weld metal HAZ thermal cycles enhance the fissuring propensity of austenitic stainless steel weld metal is unmistakable in light of the foregoing results. This phenomenon appears to be the most influential factor in the gamut of fissuring influences in austenitic stainless steels.

Since the occurrence of a single weld metal HAZ experience does not produce myriads of fissures upon the application of strain in the Fissure Bend Test, it is clear that the strain tolerance of the grain boundary microstructural regions is significantly reduced by multiple thermal cycling. This degradation of grain boundary ductility can be due to (9);

1. Enhanced segregation due to continued partitioning of harmful trace elements to the grain boundaries followed by liquation and rupture under strain.
2. Additional strengthening of the matrix by thermally-induced precipitate reaction as is the case for some ductility-dip

incidences. The increased strength of the matrix may cause strain accumulation in the degraded grain boundary microstructural region thus leading to rupture.

3. Strain-induced precipitation leading to a strengthened matrix and the subsequent accumulation of strain in the degraded microstructural region.
4. The simple accumulation of thermal and restraint strains over a range of temperatures produced by continued thermal cycling causing the strain tolerance of the grain boundary microstructural region to be exceeded.

Each of the above elements of the phenomenon may be operational at the same time. However, it is also clear that liquation plays a role in the majority of fissuring occurrences and must, therefore, be a part of the mechanism. Thus it appears that enhanced partitioning or segregation during thermal cycling can play a decisive role.

The primary purpose of the present research was to investigate the influence of the alloying elements and multiple thermal cycling on the hot cracking resistance of austenitic stainless steel weld metals. In addition, the effect of composition and multiple thermal cycling on the hot ductility were also determined to correlate with the hot cracking tendency. To more clearly define the variables, the following experimental methods were developed and experiments conducted; (1) method to develop laboratory-made alloys with variations in composition, (2) evaluation of the hot ductility and the hot cracking behavior of commercial weld metals in addition to experimental compositions and (3) testing to separate the influence of strain and thermal history on the hot cracking tendency.

The Fissure Bend Test has the favorable features desired in a weldability test in that it is economical, easy to conduct, reproducible and capable of evaluating fissuring in multipass weldments very similar to those used in current fabrication procedures. This test was utilized to evaluate the hot cracking resistance of several commercial austenitic stainless steel weld metals. In addition, the effect of welding procedures such as GTA remelting, weaving techniques and multiple remelting were also investigated with the Fissure Bend Test.

A Gleeble Hot Cracking Test developed by Krantz (117) was successfully used to evaluate the hot cracking behavior of Hastelloy X. The resulting data indicated a strong correlation between the pattern of microsegregation present in both the weld fusion zone and the weld HAZ and the location and types of microcracks produced. This hot cracking test was chosen for this research because it involves a small size specimen which can be easily prepared and its composition modified for evaluation of elemental effects. The major advantage of using the Gleeble is an ability to separate the effects of the mechanical (strain) and metallurgical (thermal history) factors. Thus, the relative importance of microstructure, strain and temperature on hot cracking of weld metals can be explored systematically.

After the introduction of the RPI hot-ductility test, the method was extensively employed to investigate HAZ fissuring in austenitic stainless steels (107-114). The test not only proved to be invaluable in establishing the relative cracking tendencies of various stainless steels, but demonstrated the ability to differentiate between "good" and "bad" materials of the same nominal composition on a heat-to-heat basis.

A modified specimen (a longitudinal composite weld with strip form) was used in this research to investigate the effect of multiple thermal cycling on the hot ductility behavior.

In order to produce weld deposits of specific chemistry, a dual-wire, dual-control cold wire device was designed and developed. This dual-wire system is used when variations in weld metal deposit chemistry are desired and a specific chemistry can be obtained by varying either the wire feed speeds or wire compositions. This miniature casting technique for new alloy development is economical and time-saving. After deposition in special fixture, the weld build-ups are cut, ground, cold rolled and annealed several times to form a strip of the appropriate thickness. The resultant strip is used to create specimens for the various hot cracking tests.

In addition to composition evaluation with the modified alloys, an additional technique for rapidly assessing the effect of specific alloying element on the hot-cracking sensitivity of weld metals was employed. This method was used by Savage and Lundin in regard to the Varestraint test evaluations (142). A strip of material with the desired alloying element is cut to an appropriate size (to provide the desired composition) and spot welded to a weld test coupon. The coupon is then GTA welded with a fully penetration bead. With knowledge of the chemical composition and dimensions of both the added material and the base material, the desired composition is obtained by controlling the size of the fused weld zone. With this technique several laboratory-made weld metals were produced and used to determine the effect of a specific alloying element on the hot cracking resistance of a fully austenitic stainless steel weld metal.

It is well known that nitrogen is a strong austenite stabilizer and it can also reduce the hot cracking tendency of the weld metal (14,46,100). Nitrogen is a potent element which can be used as substitute for nickel. Nitrogen was chosen as one of the major elements for developing a fully austenitic weld metal. A study was conducted to investigate the effect of nitrogen on microstructure and ferrite content of austenitic steel weld metals.

The work conducted in this research falls into five distinct areas thus the discussions that follow are divided accordingly. The results are discussed in this manner so that the reader can more easily appreciate the salient features of the work.

2. Effect of Nitrogen on the Microstructure and Ferrite Content of Austenitic Stainless Steel Weld Metals

It is well known that nitrogen dissolves interstitially in austenite and is a strong austenite stabilizer. The stabilizing effect of nitrogen is 20-30 times that of nickel (83,86). Nitrogen additions are valuable in increasing the tensile strength at all temperatures (89, 143-145) although this tensile strength increase is sometimes accompanied by reduced ductility and toughness. Nitrogen can also reduce the hot cracking tendency of the weld metal (14,46,100).

Autogenous GTA welds were made on eleven austenitic stainless steels in the thickness range of 0.020" to 0.5". The chemical compositions and the thicknesses of the welded materials are shown in Table 4, page 76. S1 through S5 are Type 304 austenitic stainless steels with variations in C, Si, and S contents. S6 is a modified 308L. S7 and S8 are Types 16-8-2 and 317LM respectively. P1, P2, and P3 are 1/2"

thick austenitic stainless steel plates of 304L, 304LN, and 317L respectively. The welding parameters used for the autogenous GTA welds are presented in Table 8. The Ferrite Number (FN) was measured on the as-welded surfaces with an Aminco Magne Gage. Selected specimens were sectioned, mounted, polished, and electrolytically etched in a CrO_3 -acetic acid solution.

In order to clearly define the influence of nitrogen additions in the shielding gas on the FN and to determine the nitrogen content in the weld metal, two materials were selected. S1 with a calculated ferrite potential of 4 FN and S6 with a calculated ferrite potential of $22\frac{1}{2}$ FN were chosen because of the wide ferrite range. These materials were in strip form and thus permitted full-penetration welds with a single pass avoiding any multipass or unfused base metal effects.

The results of the nitrogen addition study are summarized in Figures 22 to 24. It is to be noted from Figure 22, at 0% N_2 the developed ferrite potential is essentially identical to that calculated. Further, Figure 22 shows a rapid drop in measured FN as the volume of nitrogen in the shield gas is increased. The S6 material is reduced to 4 FN from $22\frac{1}{2}$ FN by the addition of 2% N_2 in the shielding gas and is not influenced further with increasing nitrogen addition. The S1 (4 FN) material experienced a drop to 1 FN with 2% N_2 and remained at this level with further increase in nitrogen in the shielding gas.

The chemical analyses for nitrogen of these test welds, presented graphically in Figure 23, shows that the weld metal becomes essentially saturated with nitrogen after being exposed to a shielding gas containing approximately 2% N_2 and thus no further effect of shield gas nitrogen

Table 8. Welding Parameters for Autogenous GTA Welding.

Material	Voltage (Volt)	Current (Ampere)	Travel Speed (in./minute)	Gas Flow Rate (cfh)	Shielding Gas
S1	10.0	25	1.4	30	Ar and N ₂
S2-S5	11.0	50	3.5	25	Ar and N ₂
S6	11.0	20	1.4	30	Ar and N ₂
S7-S8	7.0	43	6.0	30	Ar and N ₂
P1-P3	10.0	210	8.0	25	Ar and N ₂

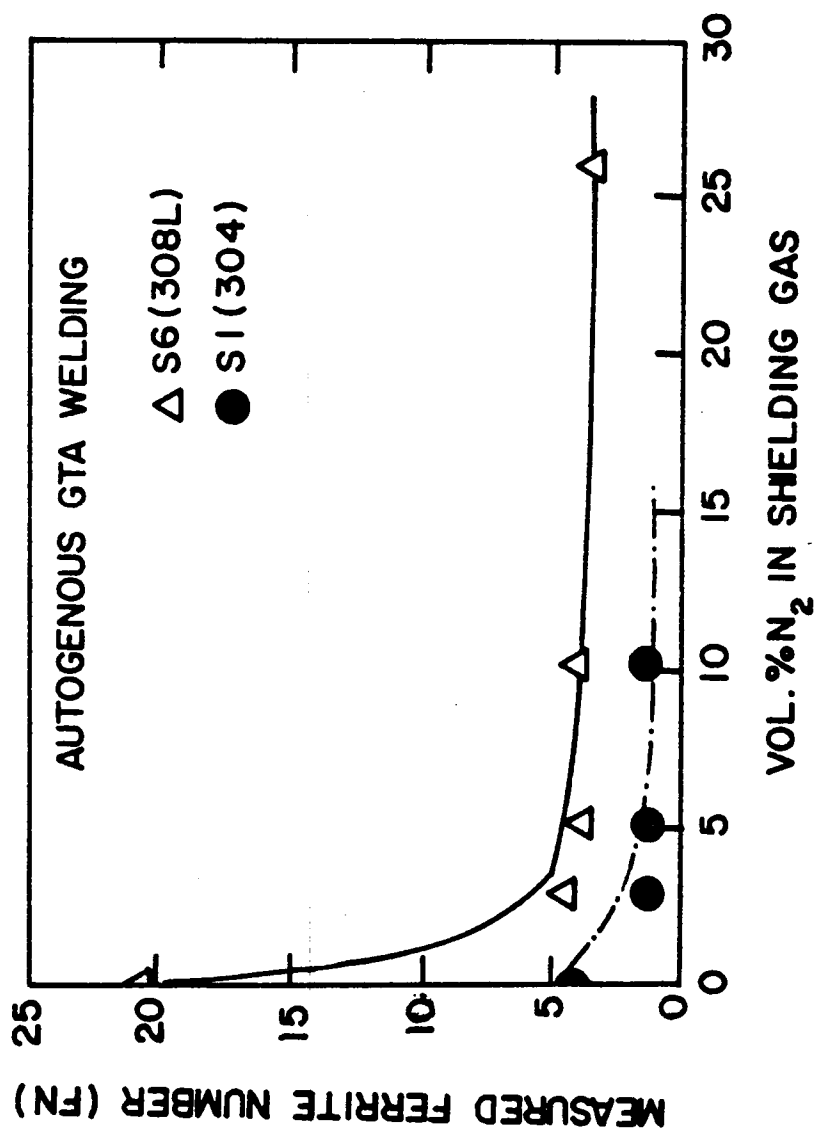


Figure 22. Measured Ferrite Number vs % N₂ in Shielding Gas for Austenitic Stainless Steel Autogenous GTA Welds S1 and S6.

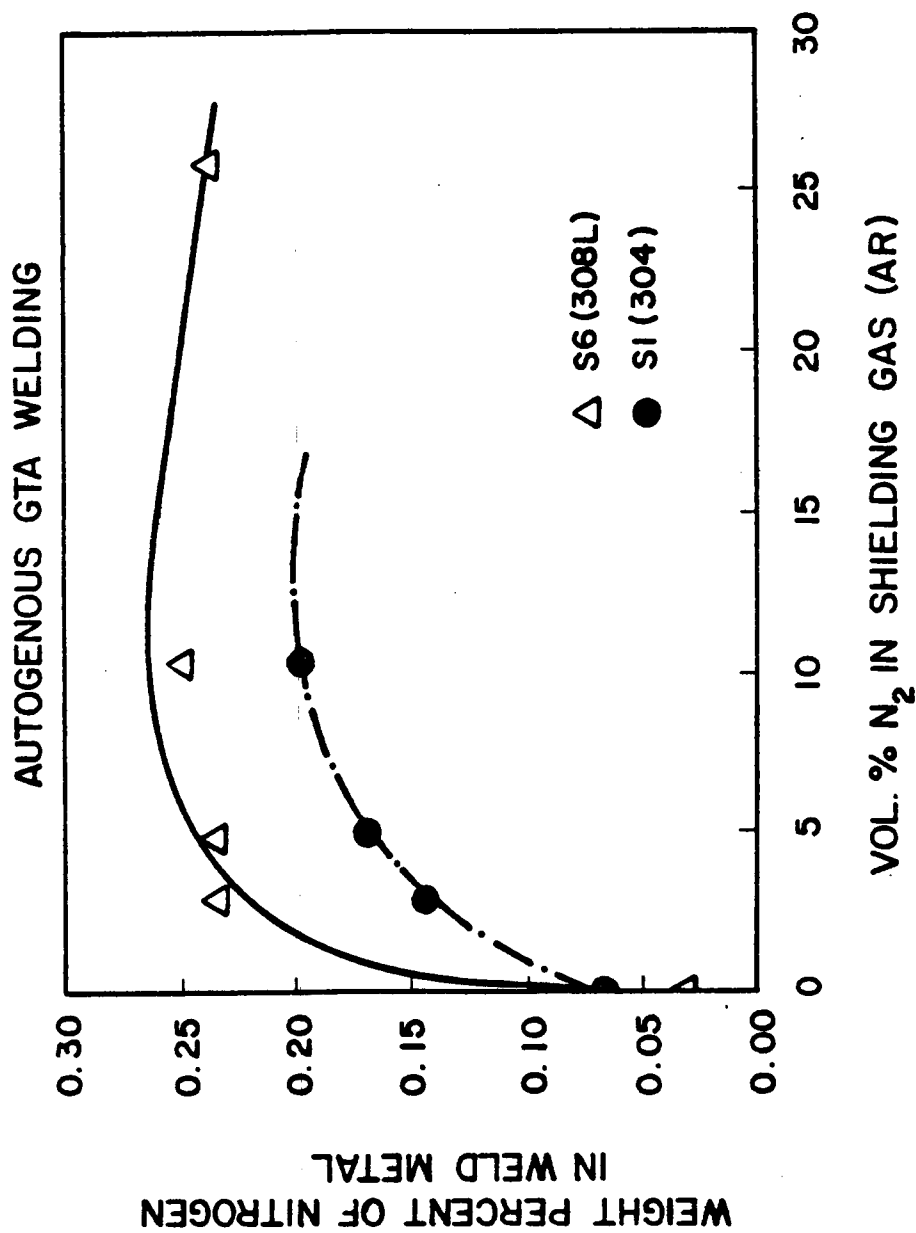


Figure 23. Nitrogen Content in Weld Metal vs $\% N_2$ in Shielding Gas for Austenitic Stainless Autogenous GTA Welds.

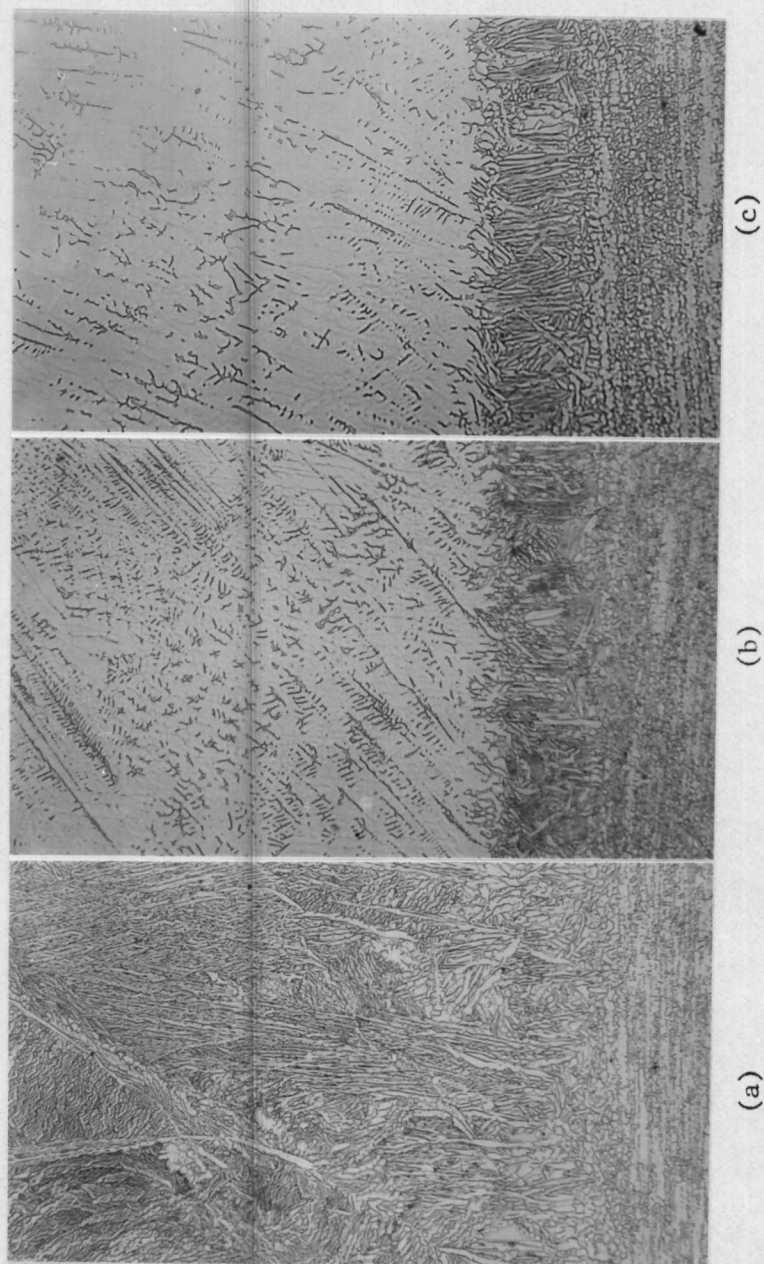


Figure 24. Microstructures of Autogenous GTA Welds of Austenitic Stainless Steel S6 (308L). (OLM, 100X)

- (a) Pure Ar, 21 FN, 0.04% N in Weld.
 (b) Ar + 5% N₂, 4 FN, 0.23% N in Weld.
 (c) Ar + 26% N₂, 3 FN, 0.23% N in Weld.

concentration is noted on FN. This is of importance since the 2% N_2 addition to argon does not influence arc behavior to any great extent whereas greater additions of nitrogen change the arc behavior in a non-advantageous manner by reducing arc stability.

The microstructures of the GTA welds in S6 with different nitrogen levels are shown in Figure 24. It is clearly seen that the ferrite content decreased with increased nitrogen in the shielding gas. There is a good correlation between the measured FN and metallographic observation of ferrite. It is also noted that the ferrite morphology changed from an acicular form to a vermicular form as nitrogen is increased. There is a relatively wide unmixed fused zone located adjacent to the base metal along the edge of the weld. The microstructure in the unmixed zones of welds with different nitrogen contents is identical to the fusion zone of the nitrogen-free weld.

Both the Schaeffler and Delong constitutional diagram were used to predict the ferrite content of the weld metals. Table 9 presents the measured and calculated FN for the S1 and S6 materials with varying nitrogen in the shielding gas. The Delong diagram gives good ferrite predictions with both the nitrogen-free and nitrogen-containing weld metals.

Fully penetrated autogenous GTA welds (with 2% N_2) were also made on sheets S2, S3, S4, S5, S7, and S8. These sheets are austenitic stainless steel Types 304, 16-8-2, and 317LM. Their calculated FN range from 2 to 5. Table 10 presents the calculated and measured FN with pure Ar and Ar plus 2% N_2 in the shielding gas. It is found that with 2% N_2 in the shielding gas, the FN is reduced to near zero. Figure 25 shows

Table 9. Ferrite Number in Austenitic Stainless Steel Weld Metals S1 and S6.

% N ₂ in Shielding Gas	S1 (304)		S6 (308L)	
	Calculated FN		Calculated FN	
	Measured FN	Delong Schaeffler	Measured FN	Delong Schaeffler
0.0	4.2	4.0	6.0	21.1
2.6	1.0	0.0	6.0	4.7
5.0	1.0	0.0	6.0	4.0
10.4	0.8	0.0	6.0	4.0
26.0	---	---	---	3.1
				4.0
				26.0

Table 10. Effect of Nitrogen in Shielding Gas on GTA Autogenous Welds of Austenitic Stainless Steel Sheet.

Weld Metal	Calculated FN (Pure Ar)	Measured Ferrite Number (FN)	
		Pure Ar	Ar + 2%N ₂
S2 (304)	5.0	4.7	0.3
S3 (304)	3.0	4.2	0.3
S4 (304)	5.0	5.5	0.4
S5 (304)	4.5	4.5	0.1
S7 (16-8-2)	2.0	2.5	0.2
S8 (317LM)	4.5	2.5	0.1

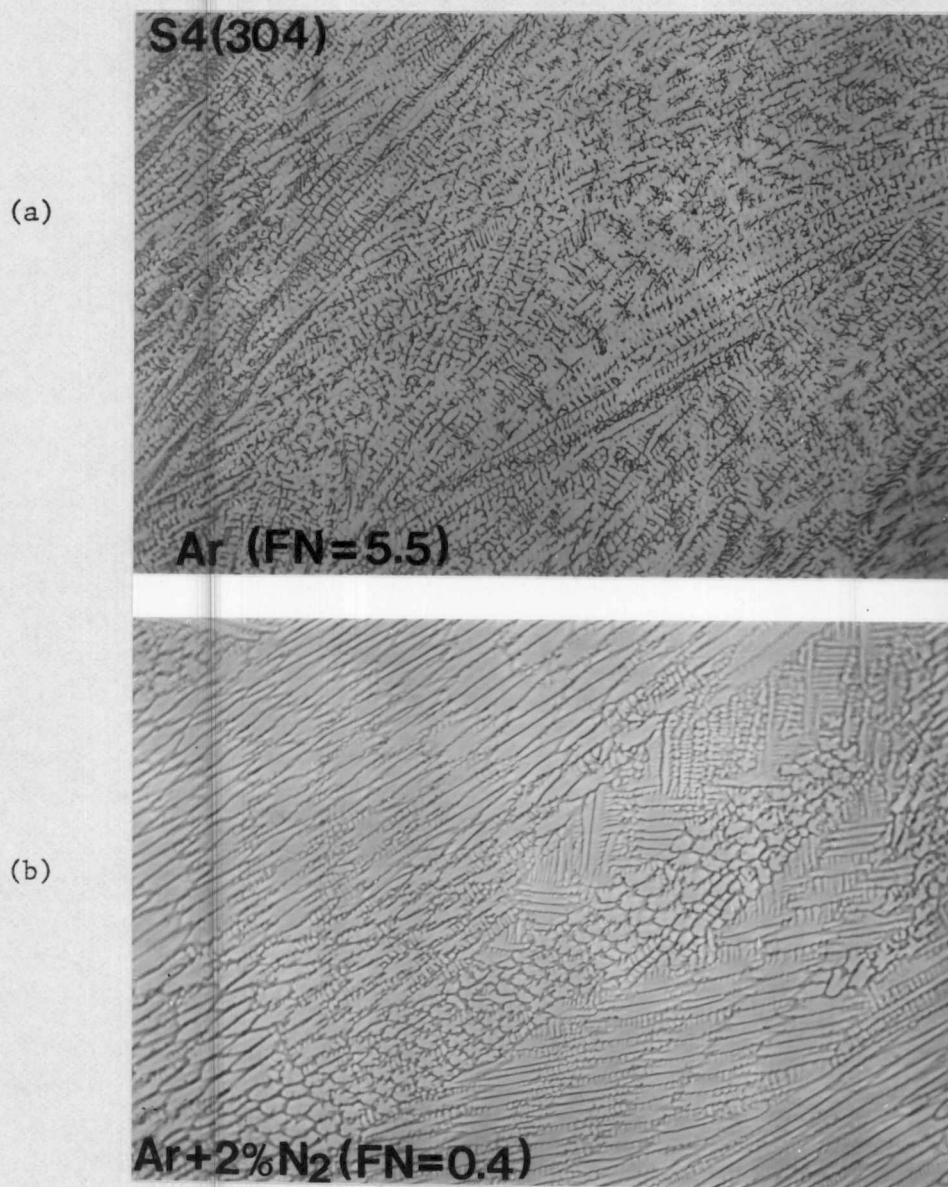


Figure 25. Microstructures of Autogenous GTA Welds of Austenitic Stainless Steel S4 (304). (OLM, 100X)

(a) Pure Ar, FN = 5.5.

(b) Ar + 2% N₂, FN = 0.4.

the microstructural change evidenced by adding N_2 to the argon shielding gas for autogenously melted S4 sheet material. The substructure size does not change significantly as the FN changes from 5.5 with pure argon to 0.4 with the 2% N_2 addition.

Autogenous GTA welds were also made on three thick plates ($\frac{1}{2}$ " thick) of P1, P2, and P3 which are Types 304L, 304LN, 317L respectively. A weld pad deposited with a 316 type electrode yielding 4.8 FN was also used to study the effect of nitrogen in the shielding gas. Evaluations on plate P3 were unsuccessful because of an unusual arc behavior. During GTA welding on plate P3 with Ar plus 2% N_2 as the shielding gas, erratic flashing and some expulsion of molten material from the weld pool occurred. When welding with pure Ar, the arc was very stable with no flashing or expulsion. When the same material was rolled into a sheet and autogenously welded with full penetration, a sound weld could be obtained even with 5% N_2 added in the shielding gas.

Figure 26 presents the results of the measured FN as a function of percentage of nitrogen in the shielding gas for the austenitic stainless steel welds of P1, P2, and SMAW deposited Type 316 weld metal. Again, with 2% N_2 in the shielding gas, the FN is effectively reduced to near zero. This can be clearly seen from Figure 27, which shows the microstructures of autogenously fused 304LN base plate with pure Ar and Ar plus 2% N_2 in the shielding gas. Although the FN is essentially zero in both the bulk weld metal and the base metal, the unmixed zone has the predicted amount of ferrite (calculated FN potential).

In conclusion, adding nitrogen to the shielding gas is an economical and effective way to reduce the ferrite number in the austenitic

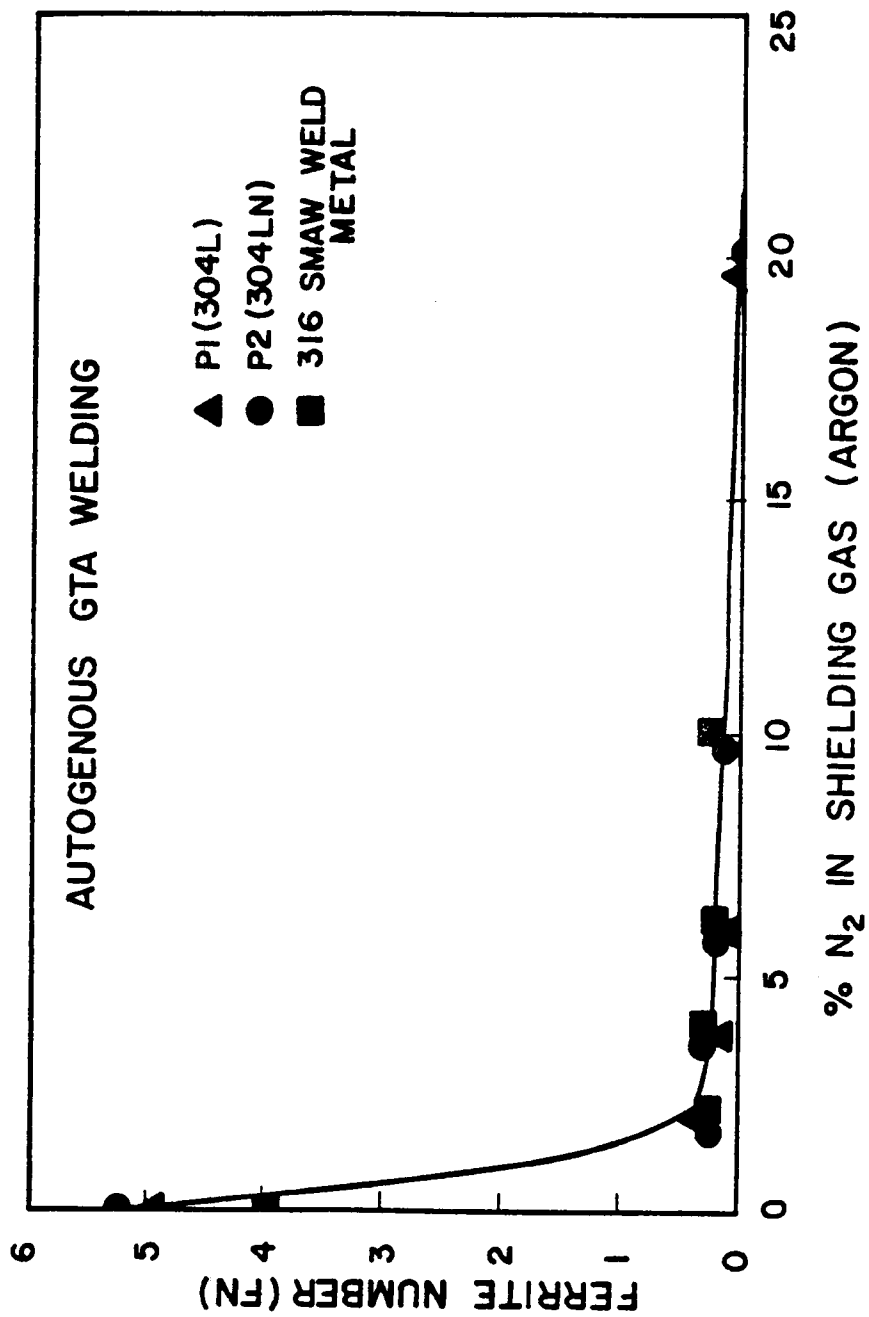


Figure 26. Measured Ferrite Number vs % N₂ in Shielding Gas for Austenitic Stainless Autogenous GTA Welds.

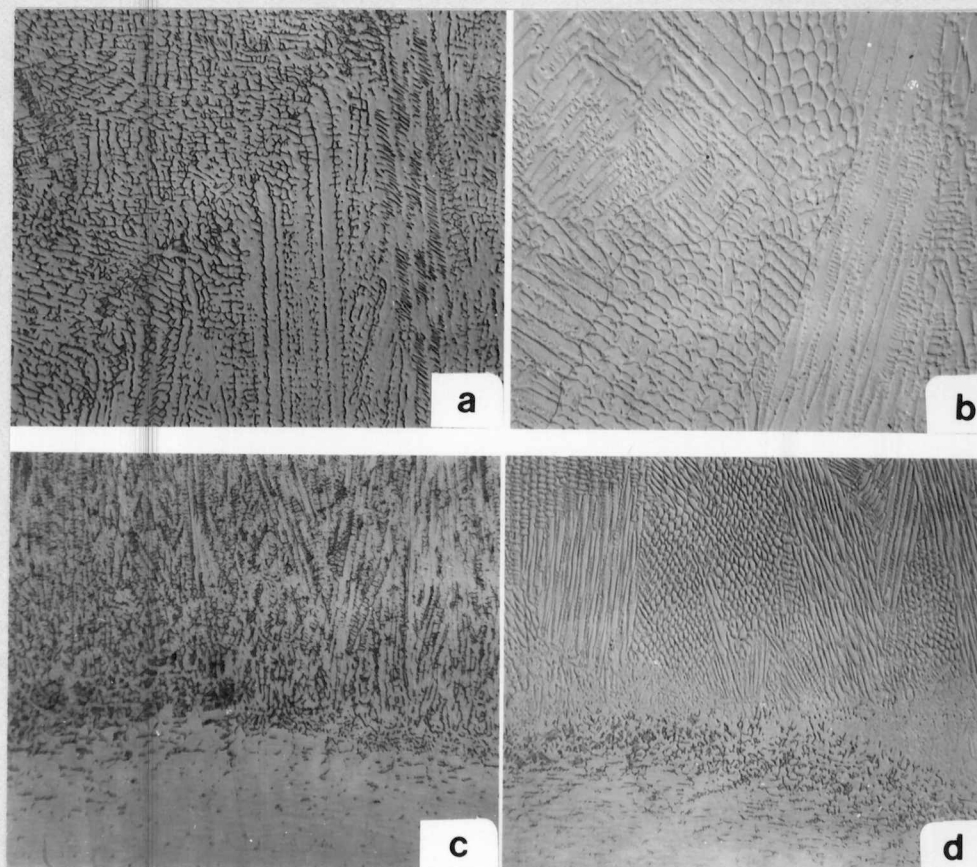


Figure 27. Effect of Nitrogen on the Microstructure of Autogenously Fused Austenitic Stainless Steel P2 (304LN).

- (a) 0% N₂, 5.2 FN. (OLM, 200X)
- (b) 2% N₂, 0.2 FN. (OLM, 200X)
- (c) 0% N₂, 5.2 FN. (OLM, 100X)
- (d) 2% N₂, 0.2 FN. (OLM, 100X)

stainless steel weld metals. Therefore, nitrogen was used as a major element in developing fully austenitic welds which were then evaluated for fissuring tendency with the Gleeble Hot Cracking Test method.

3. Development of Techniques to Produce Modified Alloy Compositions

The effect of specific alloying elements on the hot cracking susceptibility of materials has long been of interest to welding specialists and material producers alike. The effect of specific alloying elements has traditionally been established by preparing small laboratory heats of various compositions and then subjecting these heats to weldability testing. This method is often costly, requiring a considerable expenditure of time and materials.

In this work, three techniques were used to produce chemistry-modified alloys. The dual wire feeder technique was used to produce several weld metal deposits of specific chemistry by varying the wire feed speed or wire composition. The element implant technique was used to modify the alloy chemistry by adding a specific element to an alloy of interest. The nitrogen addition technique was used to stabilize austenite and reduce the ferrite content of austenitic stainless steel welds. The details of each technique and their results are described as follows.

3A. Dual Wire Feeder Technique

In order to produce several alloys of specific chemistry, a dual wire feeder technique was developed in the Welding Research and Engineering Group at the University of Tennessee. This dual-wire system is used when variations in weld metal deposit chemistry are desired and

a specific chemistry can be obtained by varying either the wire feed speeds or wire compositions. This miniature casting technique for new alloy development is economical and time-saving.

A schematic diagram of the dual wire feeder is shown in Figure 28. The dual wire feeder device is attached to a SAMPAC 750 ampere constant current automatic GTA system. Two Inconel guide tubes (0.080" I.D.) were tack welded together to construct a dual wire nozzle. The position of this dual-nozzle unit is controlled by a single control unit. The speed of each wire is controlled independently. In order to calculate the aim composition of new "mixed" weld deposit from two chemistry-known wires, the formula relating the weight fraction of the two wires as shown in Table 11 was used.

The fixture used for making the new alloy coupons is shown in Figure 29. The adjustable, water-cooled copper side blocks can give a 0.25" to 0.5" wide deposited pad. By replacing the $\frac{1}{2}$ " thick base plate with several thin sheets (for example, 0.100" thick), a weld pad with height of 1" can be easily obtained. In order to insure a uniform composition, an autogenous remelting pass is made after each deposition pass. A close-up view of the fixture, the dual-wire nozzle, the GTA torch, and the deposited weld metal is shown in Figure 30.

Figure 31 shows the microstructures of weld metals produced by the dual-wire feeder technique. "B1" and "B2" are weld metals from commercial wires ER310 and ER16-8-2 with a ferrite potential of 0 FN and 2.2 FN respectively. "D3" is a weld deposit with 1:1 ratio of "B1" and "B2" weld metals. The predicted FN values is zero and is consistent with the measured value and metallographic observation. The chemistry predicted

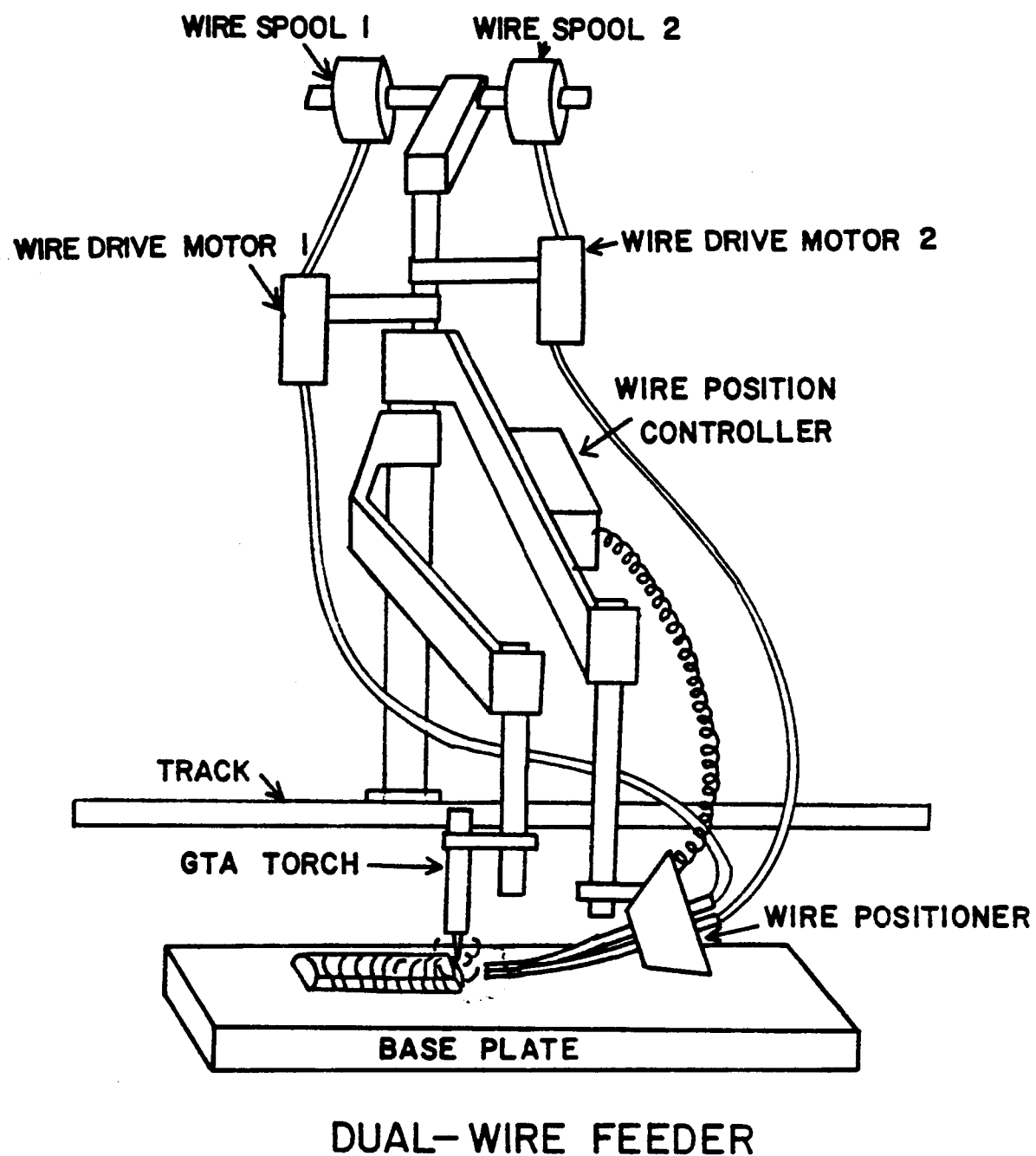


Figure 28. Schematic Diagram of Dual-Wire Feeder.

Table 11. Calculation of Chemistry of Mixed Weld Deposit.

Parameter	Wire 1	Wire 2
Diameter (in.)	D_1	D_2
Feed Rate (ipm)	R_1	R_2
Density (lb/in ³)	d_1	d_2
Weld Metal Deposit Rate	$d_1 \pi (D_1/2)^2 R_1$	$d_2 \pi (D_2/2)^2 R_2$
Weight Fraction	$\frac{d_1 R_1 D_1^2}{d_1 R_1 D_1^2 + d_2 R_2 D_2^2}$	$\frac{d_2 R_2 D_2^2}{d_1 R_1 D_1^2 + d_2 R_2 D_2^2}$
For Same Type Material, $d_1 = d_2$		
Weight Fraction	$\frac{R_1 D_1^2}{R_1 D_1^2 + R_2 D_2^2}$	$\frac{R_2 D_2^2}{R_1 D_1^2 + R_2 D_2^2}$
For Same Diameter, $D_1 = D_2$		
Weight Fraction	$\frac{R_1}{R_1 + R_2}$	$\frac{R_2}{R_1 + R_2}$

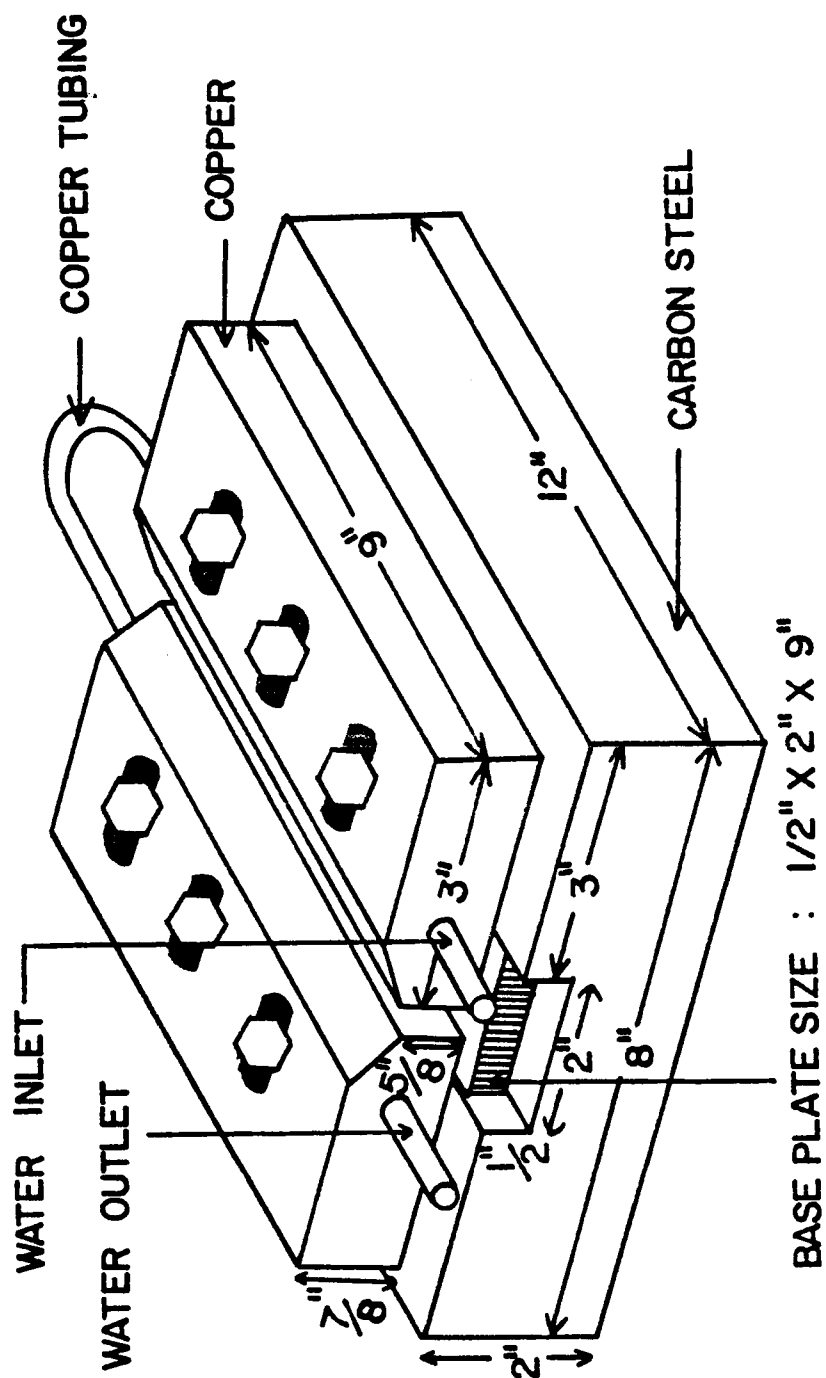


Figure 29. Welding Fixture Used with Dual-Wire Feeder Technique.

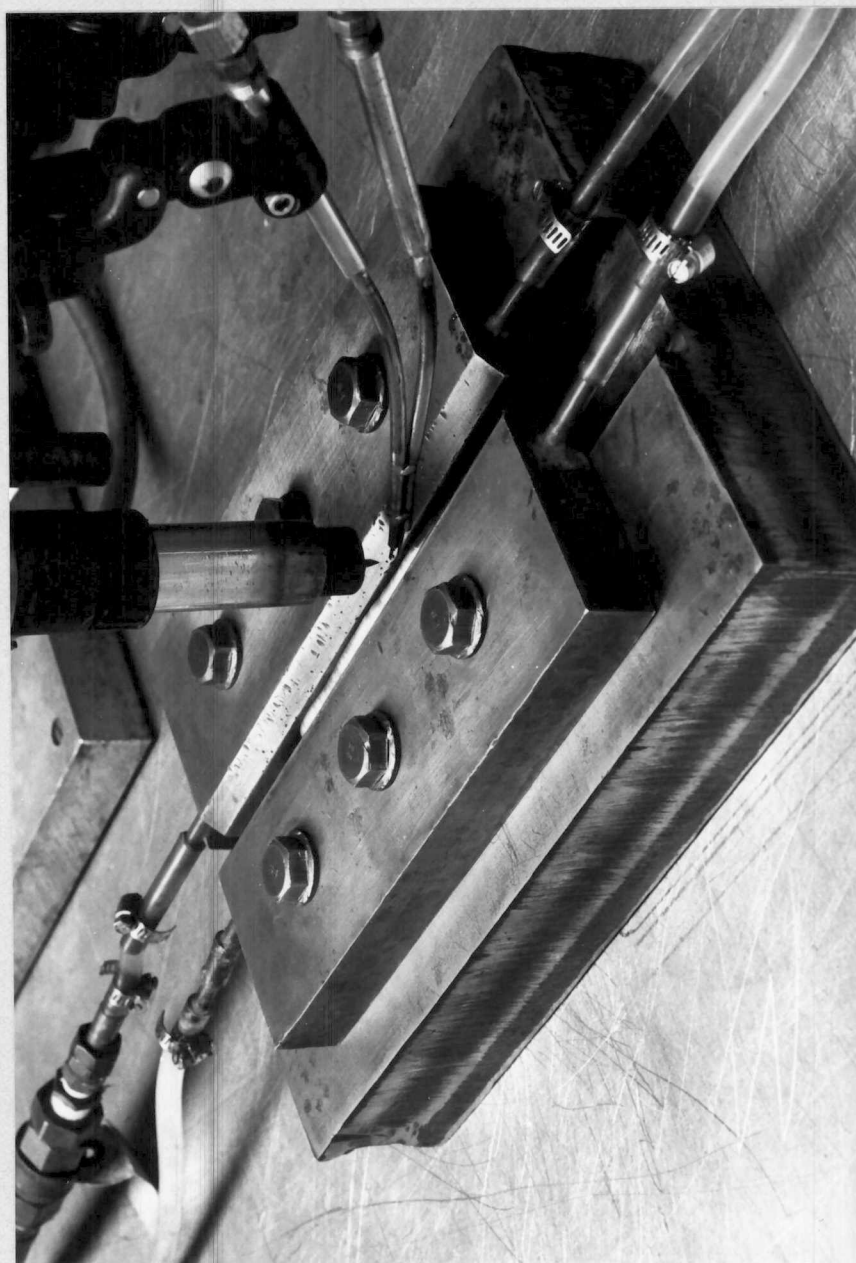


Figure 30. Welding Set Up Showing Torch, Fixture, and Dual-Wire Nozzle.

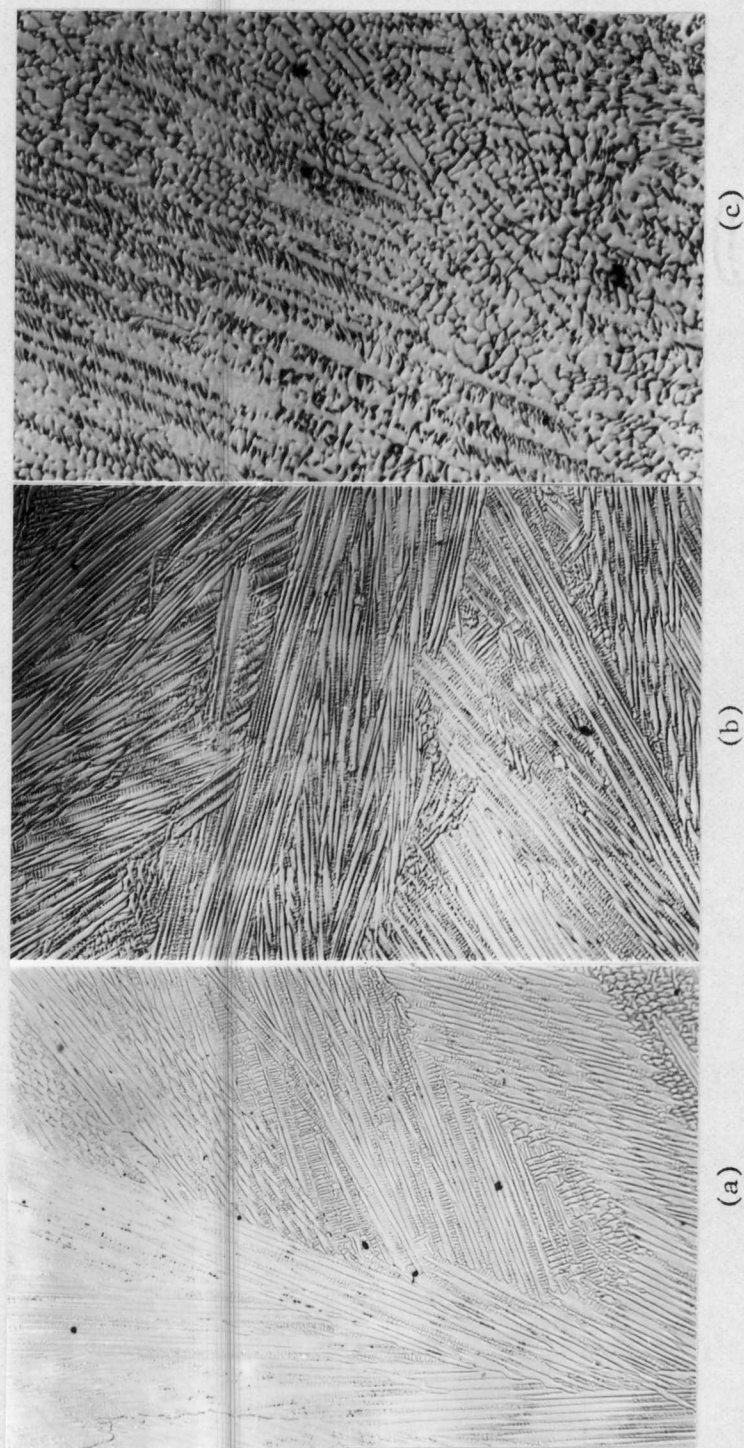


Figure 31. Microstructures of Austenitic Stainless Steel Weld Metals B1, B2, and D3. (OLM, 100X)
 (a) B1, 0 FN. (b) D3, 0 FN. (c) B2, 2 FN.

by calculation using the formula in the Table 11 and the actual chemistry from analysis for the three deposited weld pads are listed in Table 12. A very good agreement is shown between the predicted and measured values.

After deposition, the welded pad is approximately $\frac{1}{4}$ " x 1" x 5". The pad is cut and surface ground into a $\frac{1}{4}$ " x 1" x 4" sample. In order to avoid cracks during rolling operation later, the surface and edge of the sample are dressed carefully by fine grinding. Ferrite Number determinations and hardness measurements are made to check the uniformity of the composition. The sample is then cold rolled and annealed alternatively in order to obtain a strip of 0.020" thickness.. The cold rolling is conducted using a manually fed rolling mill and the annealing treatment is accomplished in an Ar-purged furnace at 1050°C for 10-30 minutes (depending on the thickness of the specimen). Light grinding after annealing is performed on the surface of the weld pad to eliminate the film of oxidized surface. Hardness is measured at each step to control the procedure. RC 30 was usually obtained after 25% CW and RB 80 was obtained after annealing. After the final annealing treatment, the strip was acid pickled in a solution of 1% HF, 15% HNO₃, and 84% H₂O for about ten minutes. Figure 32 illustrates the three steps of the strip preparation procedures.

The microstructures of alloy "D2" resulting from deposition, cold rolling, and annealing are shown in Figure 33. This alloy (D2) was made by deposition from two commercial wires "A1" and "B2" using the dual wire feeder technique. The microstructure in the as-deposited condition is shown in Figure 33a. A typical cellular dendritic structure with segregation pattern is observed. No residual ferrite was found and this is confirmed by FN measurement using the Magne Gage. Figure 33b

Table 12. Predicted and Actual Composition of Weld Metals Produced by Dual-Wire Feeder Technique.

Material	Composition, WT. %									
	Cr	Ni	Mn	Mo	Si	P	S	C	N	FN
D1 Predicted	25.43	21.42	3.17	1.12	0.30	0.011	0.011	0.059	0.118	0
Actual	25.42	21.40	2.46	0.63	0.40	0.012	0.009	0.079	0.086	0
D2 Predicted	20.25	15.15	3.04	2.01	0.25	0.011	0.015	0.037	0.100	0
Actual	19.49	13.87	2.52	1.65	0.31	0.010	0.014	0.033	0.093	0
D3 Predicted	21.37	15.08	1.72	0.97	0.47	0.015	0.015	0.072	0.052	0
Actual	20.35	14.10	1.69	0.95	0.46	0.014	0.015	0.070	0.052	0

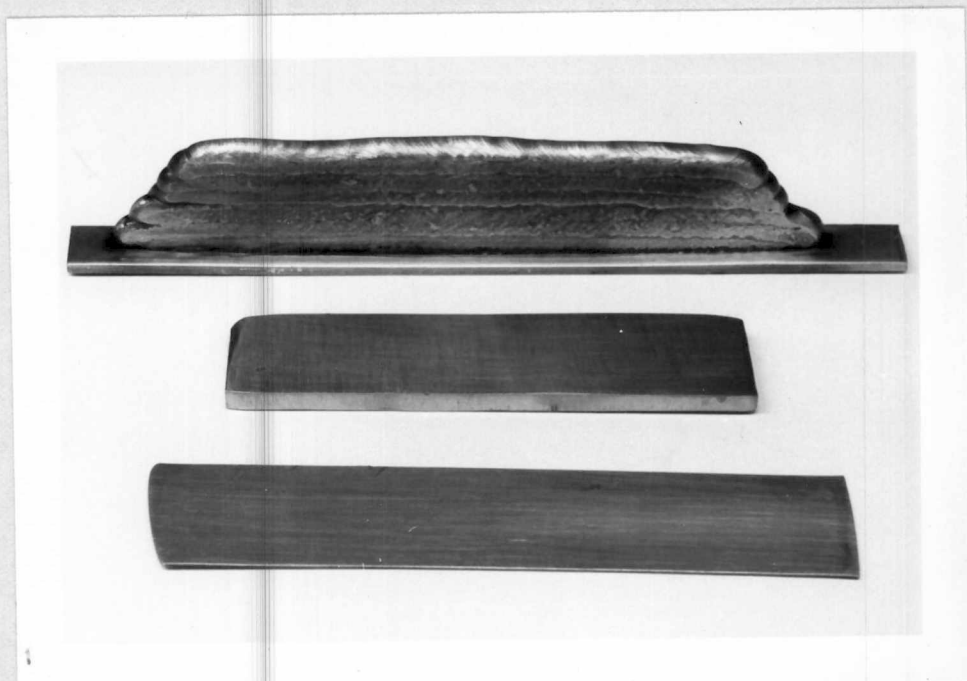


Figure 32. Strip Appearance After Welding, Grinding, and Rolling/Annealing.

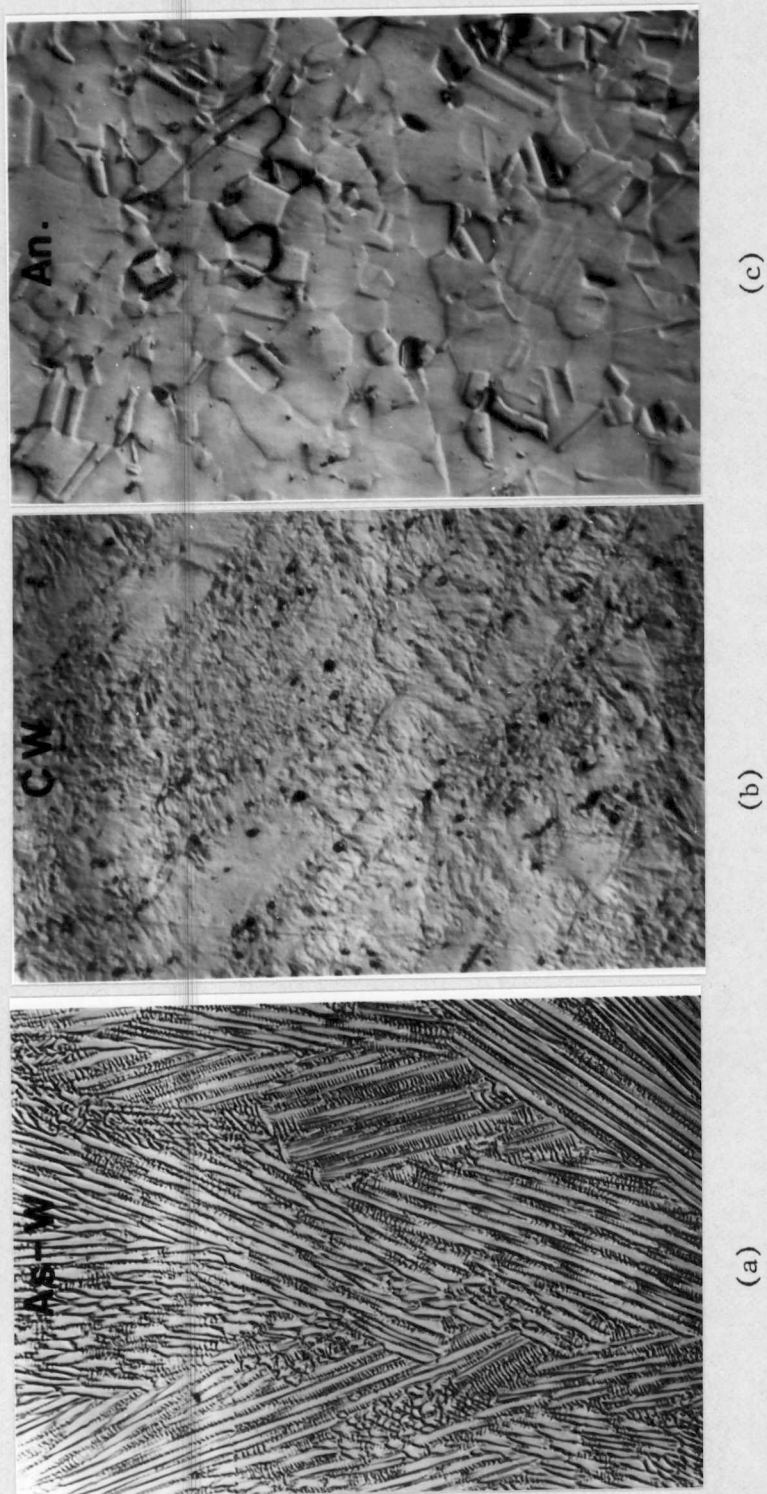


Figure 33. Microstructures of New Alloy D2 Manufactured by the Dual-Wire Feeder Technique.
(OLM, 200X)
(a) As-Welded. (b) Cold Rolled. (c) Annealed.

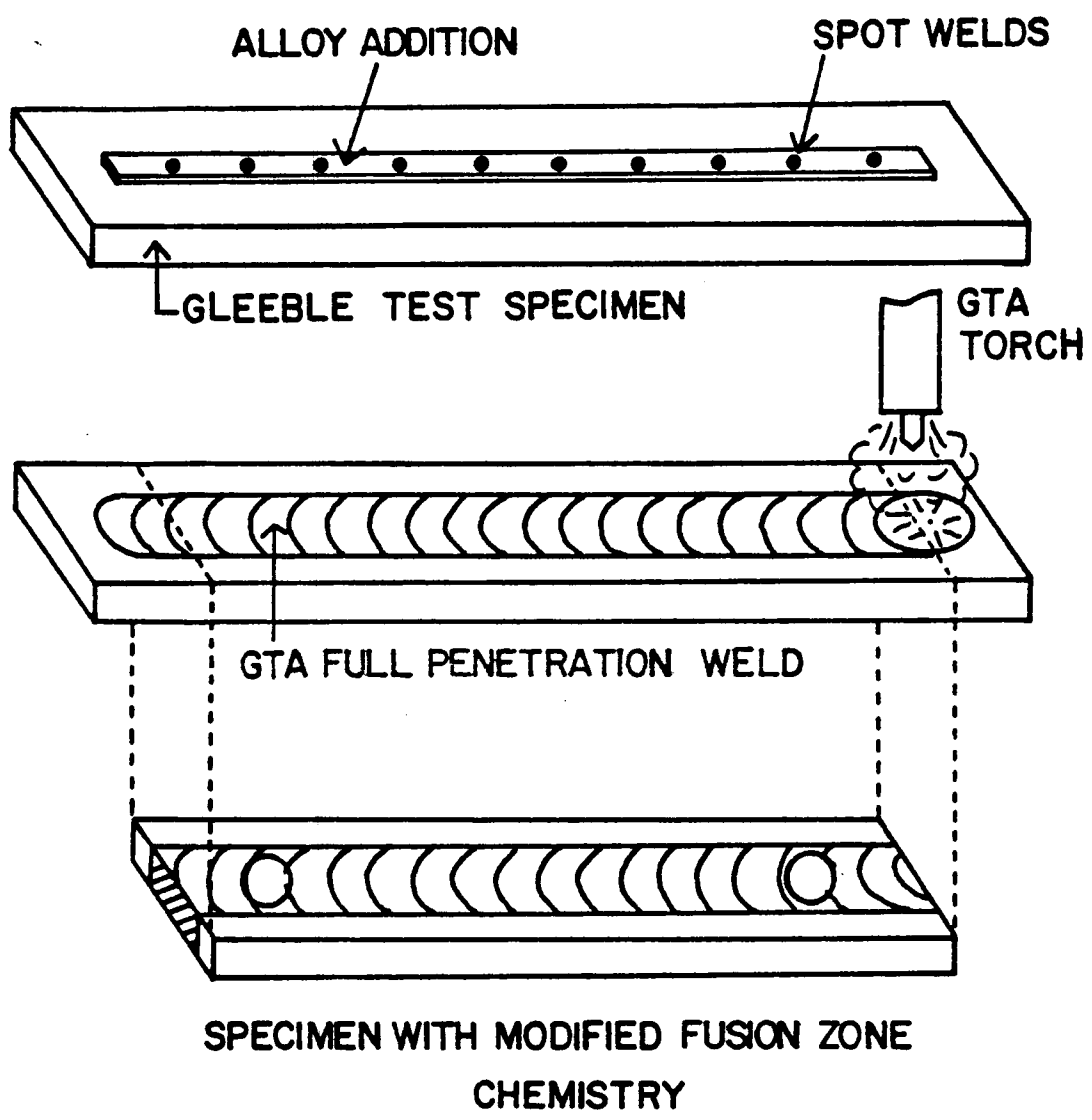
shows the microstructure of the same material after cold rolling with 30% reduction in thickness. After annealing at 1050°C for about twenty minutes, a typical fully austenitic structure with annealing twins is observed in Figure 33c. The final strip form is in the annealed condition and its microstructure is similar to that shown in Figure 33c.

The dual-wire feeder technique shows promise and this technique can be extended to other alloy types for studying alloying element effects. The properties of dissimilar metal welds (transition joints) can be investigated using this method since the precise dilution amount can be controlled. With some modification, a triple-wire feeder can be developed and thus gives more flexibility in the choice of composition.

3B. Element Implant Technique

In order to modify the composition of the strip alloys produced from the dual wire feeder technique, an element implant technique was used. This technique had been devised by Savage and Lundin (142) and was slightly altered for use in this work.

A thin sheet of material with the desired modifying alloying element is cut to an appropriate size (to provide the desired composition) and spot welded to a test strip. The strip is then full penetration GTA welded with argon shielding gas. In order to get adequate mixing, an additional GTA weld pass is made on the opposite surface of the strip. The specimen preparation is shown schematically in Figure 34. With the knowledge of the chemical composition and dimensions of both the added material and the base strip, the desired composition is obtained by controlling the size of the fused weld zone.



STEPS IN PREPARATION OF CHEMISTRY-MODIFIED SPECIMEN (IMPLANT TECHNIQUE)

Figure 34. Steps in Preparation of Chemistry-Modified Specimen With Implant Technique.

3C. Nitrogen Addition Technique

It was found that adding nitrogen to the shielding gas is an economical and effective way to reduce the Ferrite Number in the austenitic welds. It was also found from the Fissure Bend Test that the addition of nitrogen in the shielding gas had little effect on the fissuring tendency of the Type 304 base plate. Therefore the nitrogen was considered as a major element in developing fully austenitic welds.

Since this technique involves only the addition of nitrogen to the argon shielding gas during autogenous GTA remelting of the strip, the basic alloy composition is not changed but the Ferrite Number can be reduced to essentially zero. By employing nitrogen and thus utilizing its austenite stabilizing effect, fully austenitic welds were produced for evaluation using the Gleeble hot cracking testing.

The microstructures of alloy "B3" and its modification employing the element implant technique and the nitrogen addition technique are shown in Figure 35. "B3" is an austenitic stainless steel of Type 317LM with 2.5 FN in the as-welded condition as shown in Figure 35a. In order to obtain a fully austenitic type weld metal within the same composition range of alloy "B3", two techniques were used to make the modification. In the first technique, a small piece of "A nickel" strip (0.006" thick and 1/32" wide) was spot welded on the "B3" strip and then welded with full penetration by GTA using pure Ar as the shielding gas. The microstructure of the new Ni-added weld metal (identified as "B3Ni" as opposed to "B3") is shown in Figure 35b. A fully austenitic microstructure was observed and this was confirmed by Ferrite Number determination. The second technique used to modify the composition of "B3"

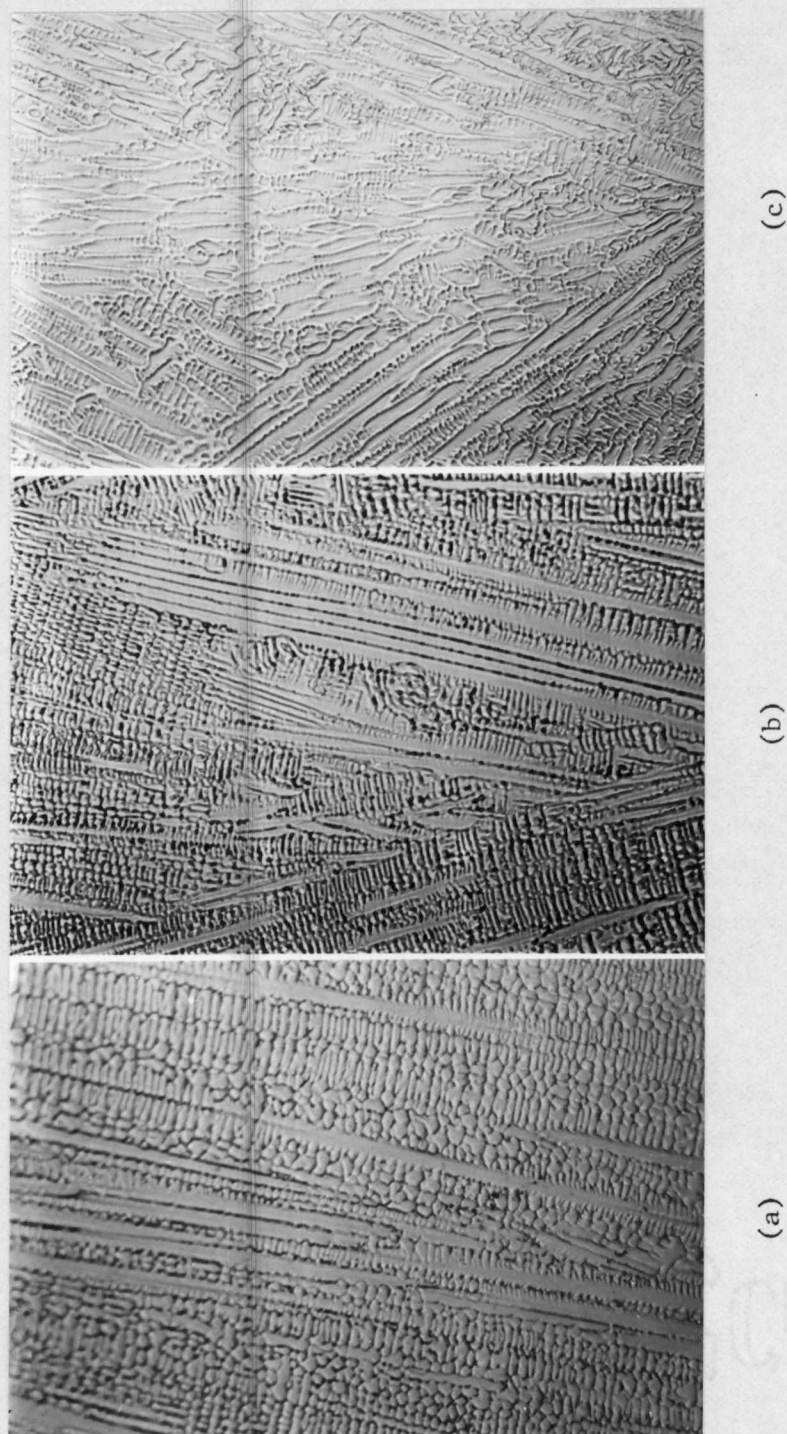


Figure 35. Microstructures of Weld Metal B3 (317LM) and Its Modified Alloys. (OLM, 200X)

(a) B3 (317LM), $FN = 2\frac{1}{2}$.
 (b) B3Ni (317LM-Ni), $FN = 0$.
 (c) B3N (317LM-N₂), $FN = 0$.

involved adding 5% N_2 into the argon shielding gas during welding. The reason for use of 5% N_2 in the shielding gas is to insure the saturation of nitrogen in the weld metal. The microstructure changed from duplex type to a fully austenitic type in the as-welded condition as shown in Figure 35c. As mentioned in the previous section, adding nitrogen to the shielding gas is an effective and economical way to reduce the Ferrite Number in the austenitic stainless steel welds.

The three techniques described above for new alloy development and composition modification are promise, dependable, economical, and time-saving. Using these techniques, a total of eleven weld metals were produced and evaluated in the Gleeble to determine the hot cracking behavior. The results of the Gleeble Hot Cracking test are presented and discussed in a latter section.

4. Hot Cracking Sensitivity Evaluation Using Fissure Bend Test

It is known that the composition of the consumable as well as welding procedure has a significant effect on the hot cracking tendency of fully austenitic stainless steel weld metal. It has been confirmed that Mn and Mo have beneficial effects in reducing hot cracking (12,13) and that the cracking susceptibility increases as Si, P, and S contents increase (14). Low arc energy together with fairly high welding speed is considered effective in reducing microcracking by reducing the weld pool size and the extent to which underlying runs are reheated. The multiple weld metal HAZ thermal cycles enhance the fissuring propensity and this effect requires further study to obtain a full understanding.

The Fissure Bend Test was originally developed and evaluated for determining the cracking resistance of SMAW deposited austenitic weld

metals (9,10,103). In the current work the test was extended to the GTAW and GMAW processes and was used to evaluate the effect of alloying elements, welding procedures, and multiple weld metal HAZ thermal cycles on the fissuring propensity of austenitic weld metals. The results of this study fall into three distinct areas topically thus the discussions that follow are divided accordingly.

4A. Effect of Alloy Elements

In order to study the effect of alloying elements on the hot cracking sensitivity of austenitic stainless steel weld metal, twelve welds utilizing commercial consumables were evaluated. Seven weld metals were deposited by the automatic GTAW process and five weld metals were deposited by SMAW process. Table 4, page 76, shows the chemical composition and calculated Ferrite Number (FN) of the weld metals. The fully austenitic filler metals are basically AISI 310 or 316 composition with a difference in the alloying elements Mn, Mo, Si, and C. Three weld metals (E2, W6, W7) contained ferrite ranging from $\frac{1}{2}$ FN to 10 FN and this permitted the determination of the effect of ferrite content on the hot cracking behavior. The base metal used was $\frac{1}{2}$ " x 2" x 9" Type 304LN or Type 304 plate.

The welding parameters used for depositing weld pads are shown in Table 7, page 80. Since the wires used for evaluation varied in diameter, the wire feed rate was adjusted to deposit the same amount of filler metal per inch of weld. The base plates were characterized by autogenously fusing the surface with six overlapping beads so that ferrite potential could be measured and dilution effects checked.

The Fissure Bend Test method was used to evaluate the welds. Weld pads were ground, bent, penetrant tested, and examined by optical microscope. Three layer pads were used for the GTAW deposits to eliminate base plate dilution effects. The Fissure Bend Test results on the austenitic stainless steel weld metals are shown in Table 13. All weld metals were evaluated in duplicate and the tests showed good reproducibility. The number of cracks shown in Table 13 represents the average of the total number of cracks in the center four inch region of the ground weld pad.

It is obvious that all of the ferrite containing weld metals showed no evidence of hot cracking. The fully austenitic (0 FN) weld metals showed different cracking tendencies. The W1, E1, and E5 weld metals did not crack while the E3, E4, and W5 weld metals exhibited significant cracking. The W2, W3, and W4 weld metals also exhibited cracking although not as extensive as the E3 weld deposit.

Comparing the data in Table 4, page 76, with that in Table 13, it can be seen that the alloying elements Mn and Mo apparently play an important role in hot cracking. With increased amounts of both Mn (of approximately 4-5%) and/or Mo (of 2-2.5%) cracking was eliminated in all weld metals except W3 and E4 which had slightly higher P and S contents. This is consistent with the observations of other investigators (12,13).

The effect of nitrogen on the fissuring tendency of the austenitic stainless steel base material Type 304 was determined by employing the Fissure Bend Test. The plate surface was autogenously fused with six overlapping beads by the GTAW process. These welds were made at 220 Amps, 10.8 ipm, 200°F interpass temperature. Welds were shielded with both

Table 13. Fissure Bend Test Results and Measured Ferrite Number of Austenitic Stainless Steel Weld Metals.

Weld Metal	As-Welded		After Bending	
	FN	Number of Cracks	FN	Number of Cracks
<u>GTA Welds</u>				
W1	0.1	0	0.3	0
W2	0.2	0	0.4	1
W3	0.0	0	0.0	3
W4	0.0	0	0.1	4
W5	0.0	4	0.0	11
W6	11.4	0	12.8	0
W7	10.8	0	12.5	0
<u>SMA Welds</u>				
E1	0.1	0	0.1	0
E2	0.6	0	0.9	1
E3	0.0	10	0.0	55
E4	0.0	0	0.0	13
E5	0.0	0	0.0	0
<u>Autogenous GTA Welds</u>				
B1	6.3	0	7.3	0
B2	5.2	0	6.4	0

pure Ar and Ar plus 2% N₂ in the shielding gas. The pure Ar autogenous welds exhibited a Ferrite Number of 5.2 and the Ar plus 2% N₂ weld had 0.2 FN.

No fissures were found in either the pure Ar or the Ar plus 2% N₂ welds with the Fissure Bend Test, even though 2% N₂ added weld had only 0.2 FN. This indicates that the addition of nitrogen in the shielding gas had little effect on the fissuring tendency of the base plate derived weld metal and that low ferrite per se may not be the reason for fissuring. It is recognized that this effect may be composition dependent and is not to be considered as general behavior. The effect of nitrogen addition on the hot cracking resistance of austenitic weld metal was extensively studied using Gleeble and is discussed later.

The microstructures typical of unfissured fully austenitic weld metals are shown in Figure 36a. The cellular dendritic solidification mode and segregation pattern is clearly seen in this three dimensional view. Microfissures were found in the unbent fully austenitic E3 weld metal deposit. All cracking was along the grain or subgrain boundaries as shown in Figure 36b. The segregation pattern is similar to that of the unfissured weld metals and optical microscopy does not reveal the cause of difference in behavior (cracked vs uncracked).

Figure 37 shows the microstructures of two of the austenitic stainless steel weld deposits near the fusion line. It is clearly evident that there is a significant amount of ferrite (FN = 6) in the unmixed zone which located between the HAZ of the base metal and composite fusion zone (basically filler metal). Although the unfused base metal and fusion zone show essentially 0 FN, the unmixed zone has about 6 FN.

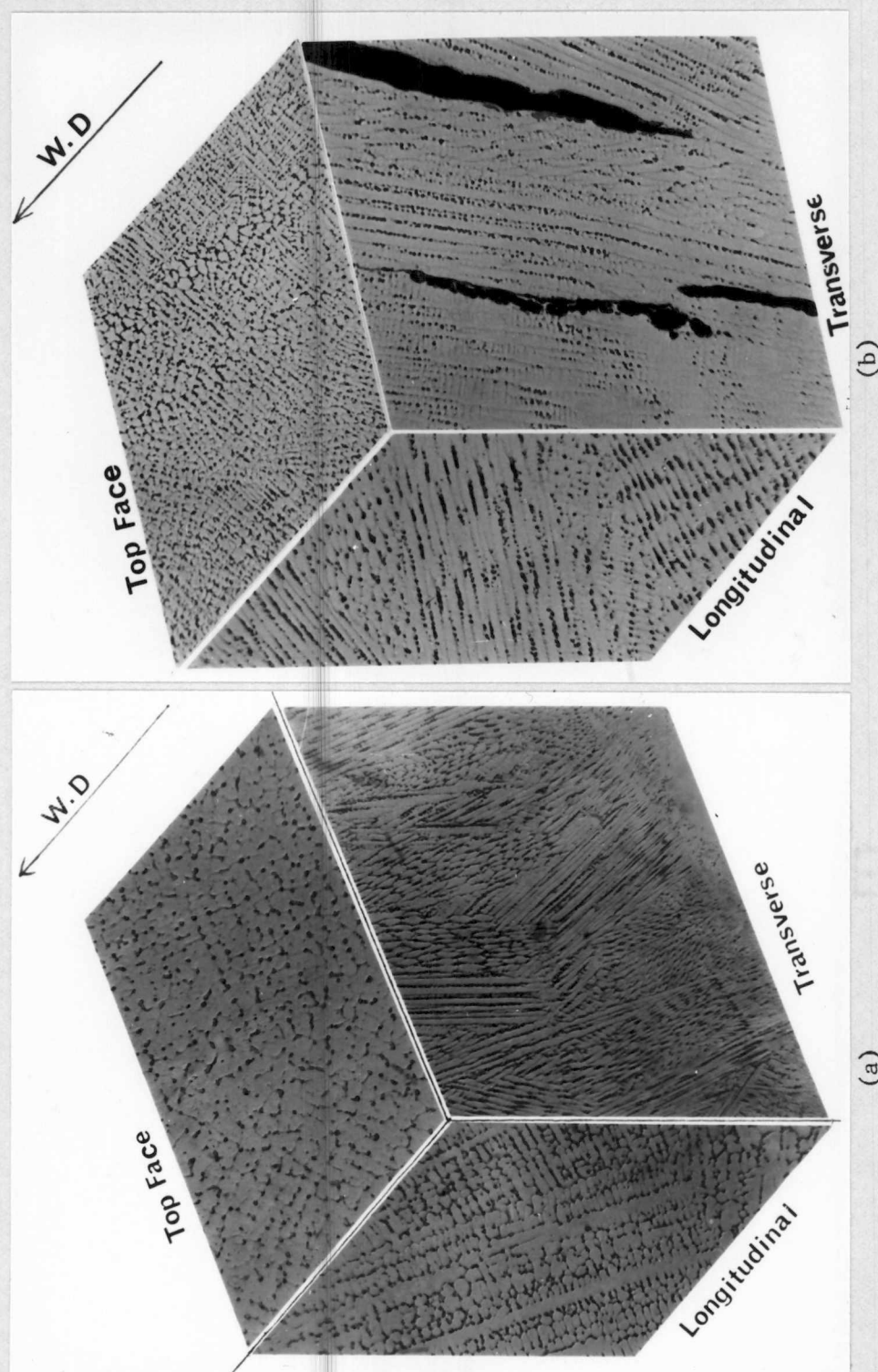
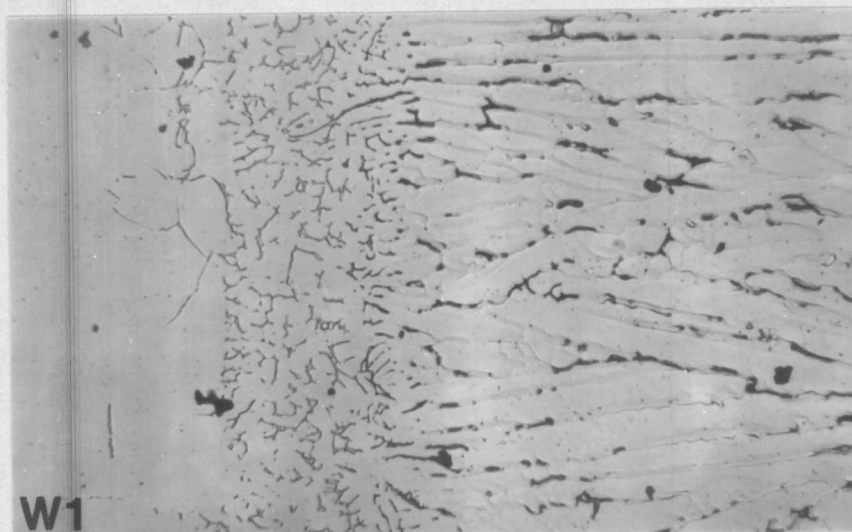


Figure 36. Microstructures of Fully Austenitic Stainless Steel Weld Metal. (OLM, 200X)
(a) E1, Unbent. (b) E3, Unbent.

(a)



(b)



Figure 37. Microstructures of As-Deposited Fully Austenitic Stainless Steel Welds. (OLM, 200X)

(a) GTA W1 Weld Metal, Longitudinal Direction.

(b) SMA E3 Weld Metal, Transverse Direction.

(Note Fusion Boundary at Left in Micrographs)

In the case where low permeability (low FN) is desired one should be aware of the fact that the outer periphery of a deposited weld will have increased permeability due to the "unmixed zone" characterized by the ferrite potential of the base plate. Some dilution is also to be expected adjacent to the fusion line and in the multilayer GTAW pads the first layers averaged 0.5 to 1 FN. The second layers were essentially ferrite free. Thus, in this testing (with two or three layer pads), the undiluted weld metal was evaluated.

In order to evaluate a relationship between hot cracking tendency, ferrite potential, and Cr and Ni equivalents, the DeLong diagram, as shown in Figure 38, was used. The weld metals used are shown on the diagram. It is apparent that a small amount of ferrite in the weld metal can very effectively increase the cracking resistance. No consistent trends were found for the fully austenitic welds. This lack of correlation with Cr and Ni equivalents alone indicates that both the welding procedures and the consumable chemistry are important in hot cracking behavior. The effect of specific alloying elements on the hot cracking resistance was investigated using Gleeble device and is discussed in the next section.

4B. Effect of Welding Procedure/Process

It is known that variations in welding procedures influence the incidence of hot cracking of austenitic stainless steel weld metals. The Fissure Bend Test was utilized to determine the effect of welding process (GMAW vs GTAW), GTA remelting and energy input on the hot cracking sensitivity of austenitic welds.

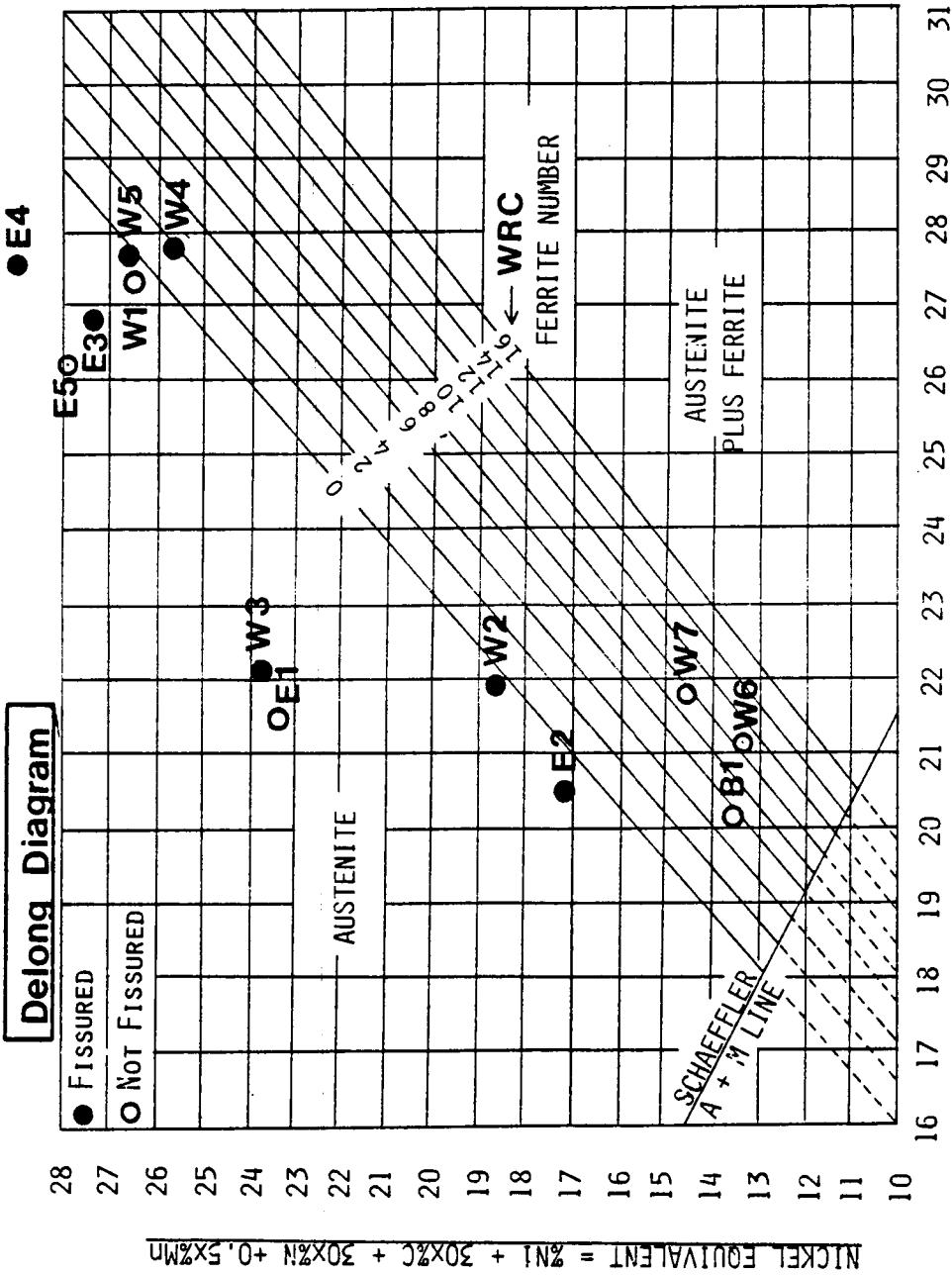


Figure 38. Calculated Weld Metal Compositions on Delong Diagram.

The filler metal W1, when deposited by the cold wire GTAW process, had no sensitivity toward cracking. However, it is to be noted that this same filler metal had a definite tendency toward fissuring when deposited by GMAW. This tendency was first observed in full scale heavy section procedure qualifications utilizing the same heat of W1 wire deposited by GTAW (cold wire) and GMAW processes (146). The dichotomy in cracking behavior was duplicated with the Fissure Bend Test procedure in the laboratory. Pads of GMAW deposition, GTAW deposition, and auto-genous GTA remelting after GMAW deposition were evaluated. The welding conditions for these pads are shown in Table 14.

The Fissure Bend Test results are presented in Table 15. GTAW (cold wire) deposited W1 weld metal showed no tendency toward fissuring. However, the W1 weld metal deposited by GMAW with both pure Ar and Ar plus 2% O₂ shield gas showed extensive fissuring and this directly reflects the full scale procedure qualification test results.

It was initially considered that due to the process differences some modification of the deposited weld chemistry might account for the significant difference in cracking behavior. Chemical analysis of the wire and the weld metals deposited by both techniques, as shown in Table 16, did not support this contention. There are some minor differences but, in reality, nothing which could account for significant cracking sensitivity change was found. To put to rest the chemistry question, weld metal deposited by GMAW was autogeneously re-fused by GTAW over the weld pad surface and tested by the normal Fissure Bend Test methods. This remelting of the GMAW deposit virtually eliminated cracking as shown in Table 15. Thus, it appears that the deposition technique,

Table 14. Welding Conditions of Fully Austenitic Stainless Steel Weld (Consumable W1).

Welding Process	Voltage (Volts)	Current (Amps)	Travel Speed (ipm)	Wire Speed (ipm)	Energy Input (KJ/in.)	Gas Flow (cfh)	Interpass (°F)
GMA, Pure Ar	30.0	305	17.0	220	32.3	30	250
GMA, Ar + 2% O ₂	29.0	300	16.0	250	32.6	40	250
GMA, Pure Ar Plus	27.0	270	16.5	220	26.5	35	250
GTA Remelting	10.0	200	7.6	---	15.8	35	250
GTA, Pure Ar	10.8	180	8.0	26	14.6	25	250

Table 15. Results of the Fissure Bend Test on Fully Austenitic Stainless Steel Weld Metal W1.

Welding Process	Shielding Gas	Number of Hot Cracks	
		As-Welded	After Bending
GMA	Ar	8	66
GMA	Ar + O ₂	10	95
GMA Plus GTA Remelting	Ar	0	8
GTA	Ar	0	0

Table 16. Chemical Analysis on the W1 Bare Wire, GTAW Deposited Weld Metal, and GMAW Deposited Weld Metal.

Consumable Conditions	Wt. %						
	C	Si	Mn	P	S	Cr	Ni Mo N
Bare Wire	0.019	0.130	4.40	0.010	0.010	24.67	22.49 2.23 0.120
GTAW Weld Metal	0.023	0.190	3.99	0.007	0.013	24.35	21.06 2.03 0.100
GMAW Weld Metal	0.029	0.210	4.08	0.010	0.011	23.26	20.20 1.91 0.100

probably related to both solidification mechanics and the HAZ thermal cycle severity, is a primary reason for the behavior.

The microstructure of the welds deposited by GMAW and GTAW remelting are shown in Figure 39. There is no significant difference in the grain or substructure size. The macroscopic solidification rate was approximately the same for these welds and thus this similarity of microstructure might be an anticipated result. However, the grain boundary conditions are apparently altered (cracking tendency indicates this) by the deposition mode.

In order to define the effect of GTA remelting, two weld metals deposited by the SMAW process were evaluated in the Fissure Bend Test. E2 is a SMAW electrode with ferrite potential of $\frac{1}{2}$ FN and E4 is a fully austenitic SMAW electrode. After deposition, the pad was surface ground and GTA remelted with six overlapped stringer beads. The Fissure Bend Test results are shown in Table 17. It is clearly seen that GTA remelting of a previously SMAW deposited weld metal reduces the number of microcracks for fully austenitic type weld metals. The microstructures of both the SMAW deposits and SMAW plus GTA remelting deposit are shown in Figure 40. There is no significant difference in the grain or substructure size. The heat input and macroscopic solidification rate were approximately the same for these welds. However, remelting did appear to cause a reduction in inclusion content, especially at grain boundaries.

In order to determine the effect of GTA remelting on reducing inclusion content in previously SMA deposited weld metal, two samples ($\frac{3}{4}$ " x 1" x 2") were cut from the weld pad deposited with electrodes E2 and E4. After surface grinding, a GTA remelt spot weld was made at 200 Amp, 10.5 V, 40 cfh Ar with $\frac{3}{32}$ " electrode. After five seconds

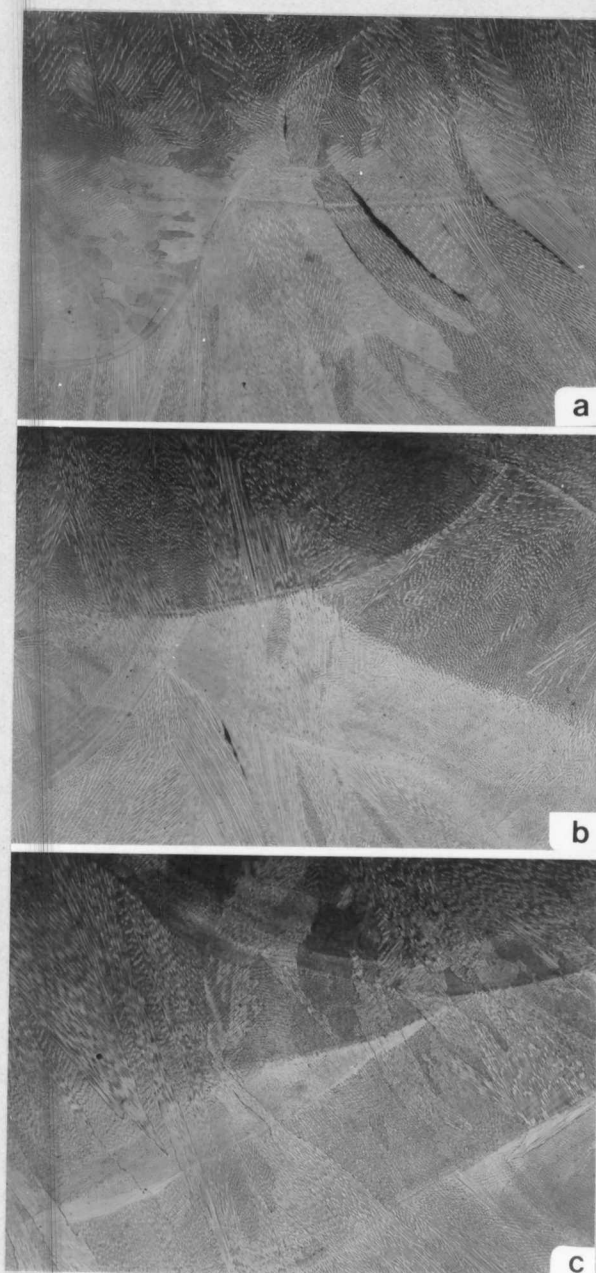


Figure 39. Microstructures of Weld Metal W1, Transverse Section.
(OLM, 50X)
(a) GMA Deposit (Ar).
(b) GMA Deposit (Ar + 2% O₂).
(c) GMA Deposit Plus GTA Remelting (Ar).

Table 17. Fissure Bend Test Results of GTA Remelting of Previously SMA Deposited Weld Metals.

Weld Metal	Welding Process	Measured FN	Number of Cracks
E2	SMA Deposit	0.5	1
E2	SMA Deposit Plus GTA Remelting	0.5	0
E4	SMA Deposit	0.0	13
E4	SMA Deposit Plus GTA Remelting	0.0	0

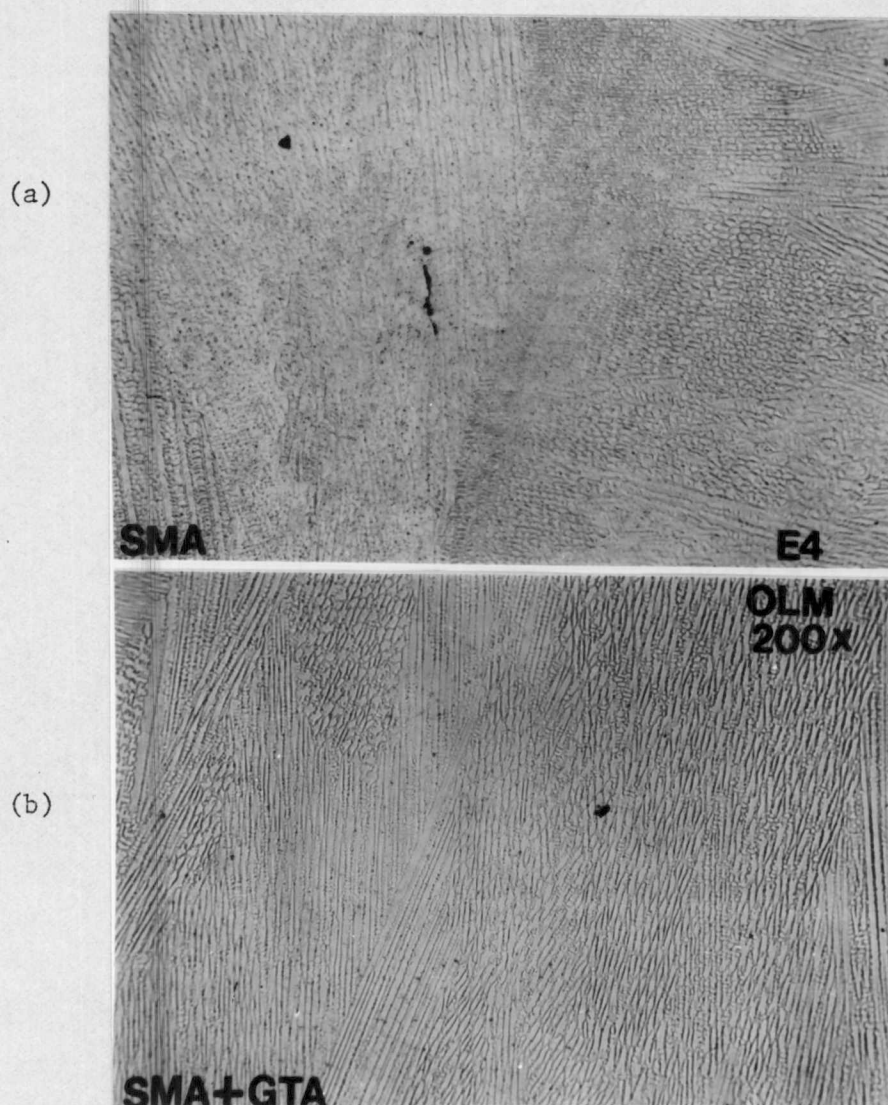


Figure 40. Microstructures of Fully Austenitic Stainless Steel Weld Metal (E4). (OLM, 200X)
(a) SMA Deposit. (b) SMA Deposit Plus GTA Remelting.

of arc on time the weld was terminated and a 0.4" diameter spot weld resulted. The appearances of the weld pads before and after spot welding are shown in Figure 41. It can be clearly seen that slag appear around the edge and in the center of the spot weld. The E2 weld is significantly cleaner than E4 weld. SEM examination shows a typical cellular dendritic structure on the as-welded surface. A rather brittle appearance is seen in the slag on the surface. EDAX results (Figure 42) showed that there is higher Al, Si, Ti, V, and Mn in the slag than in the bulk of weld metal E4. This indicates that the inclusion content in the weld metal can be reduced by a GTA remelting process and thus the possibility of hot cracking is also probably reduced.

It is known that heat input should be minimized to avoid cracking in welding austenitic stainless steels. In order to determine the effect of energy input on the hot cracking sensitivity of weld metal W1 for both GTAW and GMAW processes. Fissure Bend Tests were conducted on pads deposited (or remelted on previously deposited pad) using different travel speeds. The effect of energy input on the hot cracking sensitivity for both the GTAW and GMAW processes is shown in Figure 43. The energy input is calculated using the formula: $E.I. = V \times I \times 60 / v$ (where V = arc voltage in volts, I = arc current in amperes, and v = arc travel speed in inches per minute). It was found that the number of hot cracks increases significantly as the energy input increases for both the GMAW and GTAW processes. The high energy inputs probably enhance the segregation of harmful elements to the grain boundaries in the HAZ of the previously deposited weld metal.

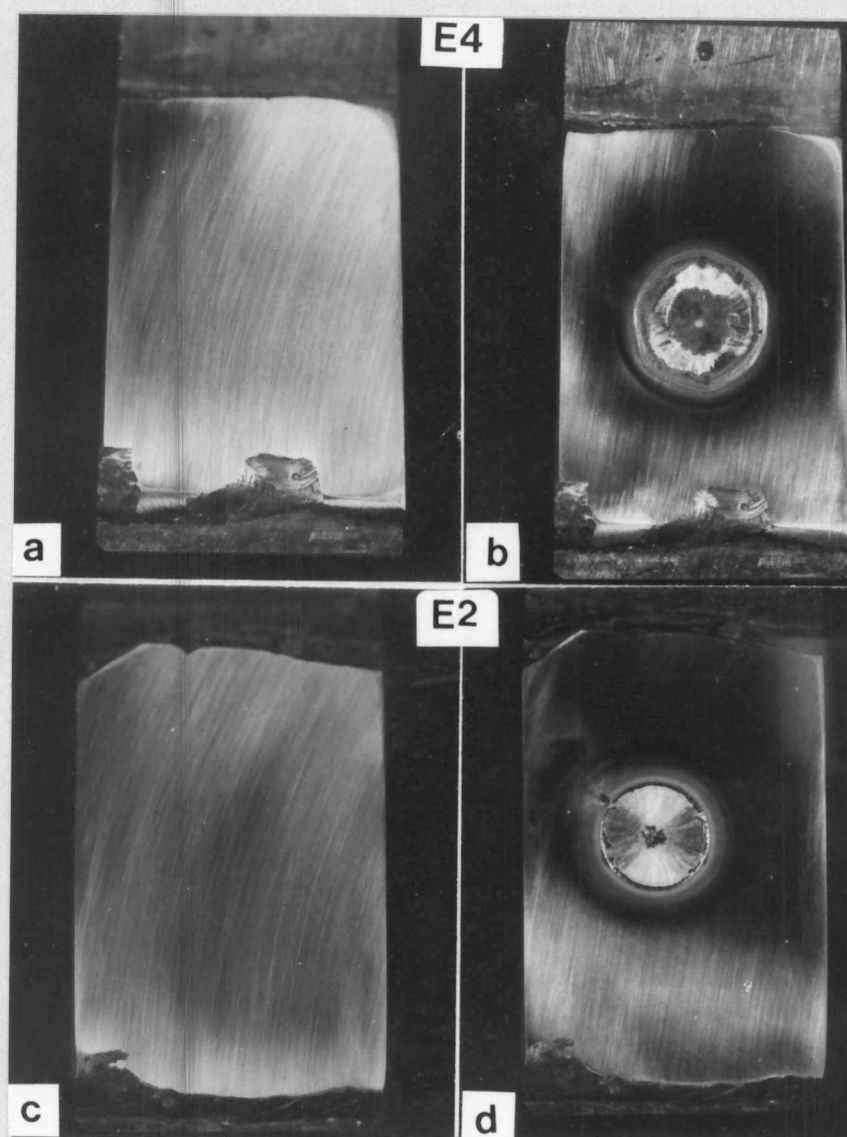


Figure 41. Surface of SMA Deposited Weld Metal. (3X)

- (a) E4 Weld, Before GTA Spot Welding.
- (b) E4 Weld, After GTA Spot Welding.
- (c) E2 Weld, Before GTA Spot Welding.
- (d) E2 Weld, After GTA Spot Welding.

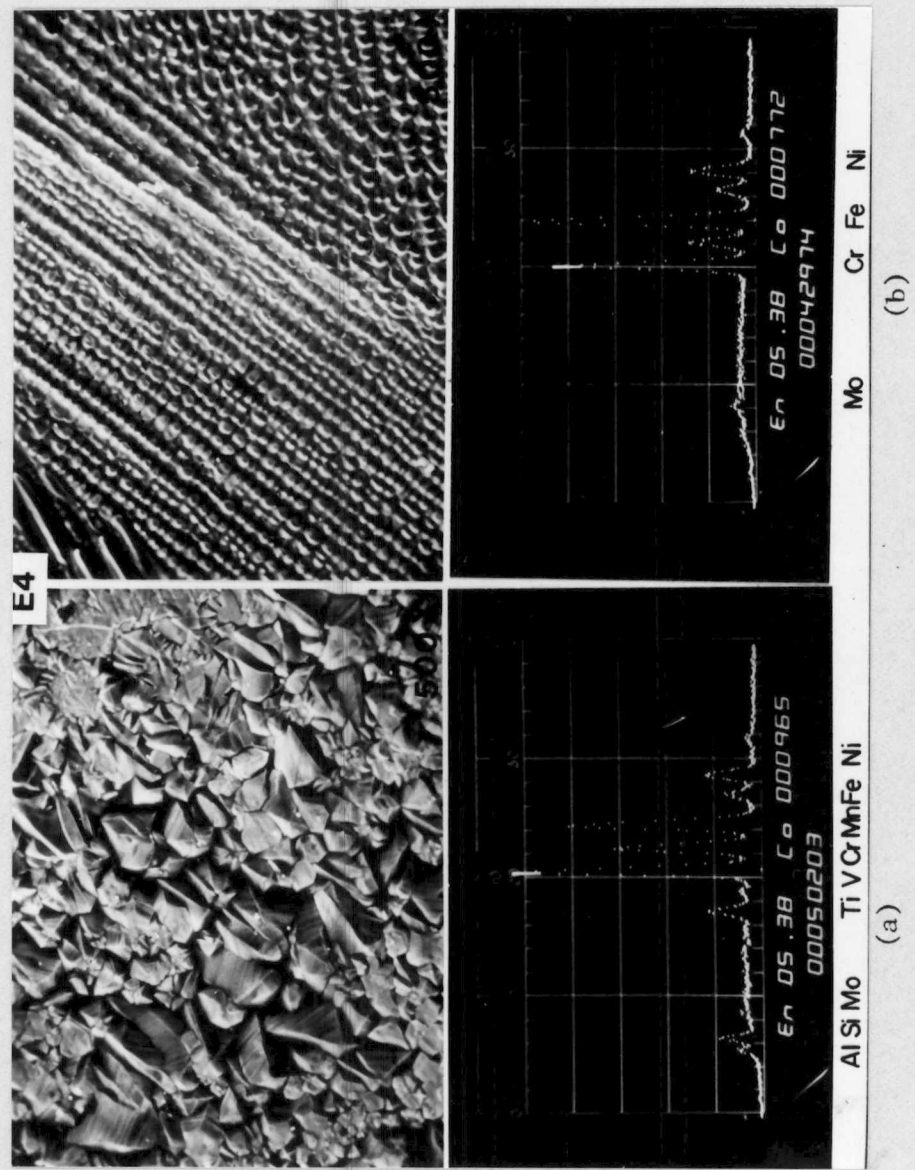


Figure 42. EDAX of Spot Weld of Fully Austenitic Stainless Steel Weld Metal E4. (SEM, 500X)
(a) Slag. (b) Remelted Weld Metal.

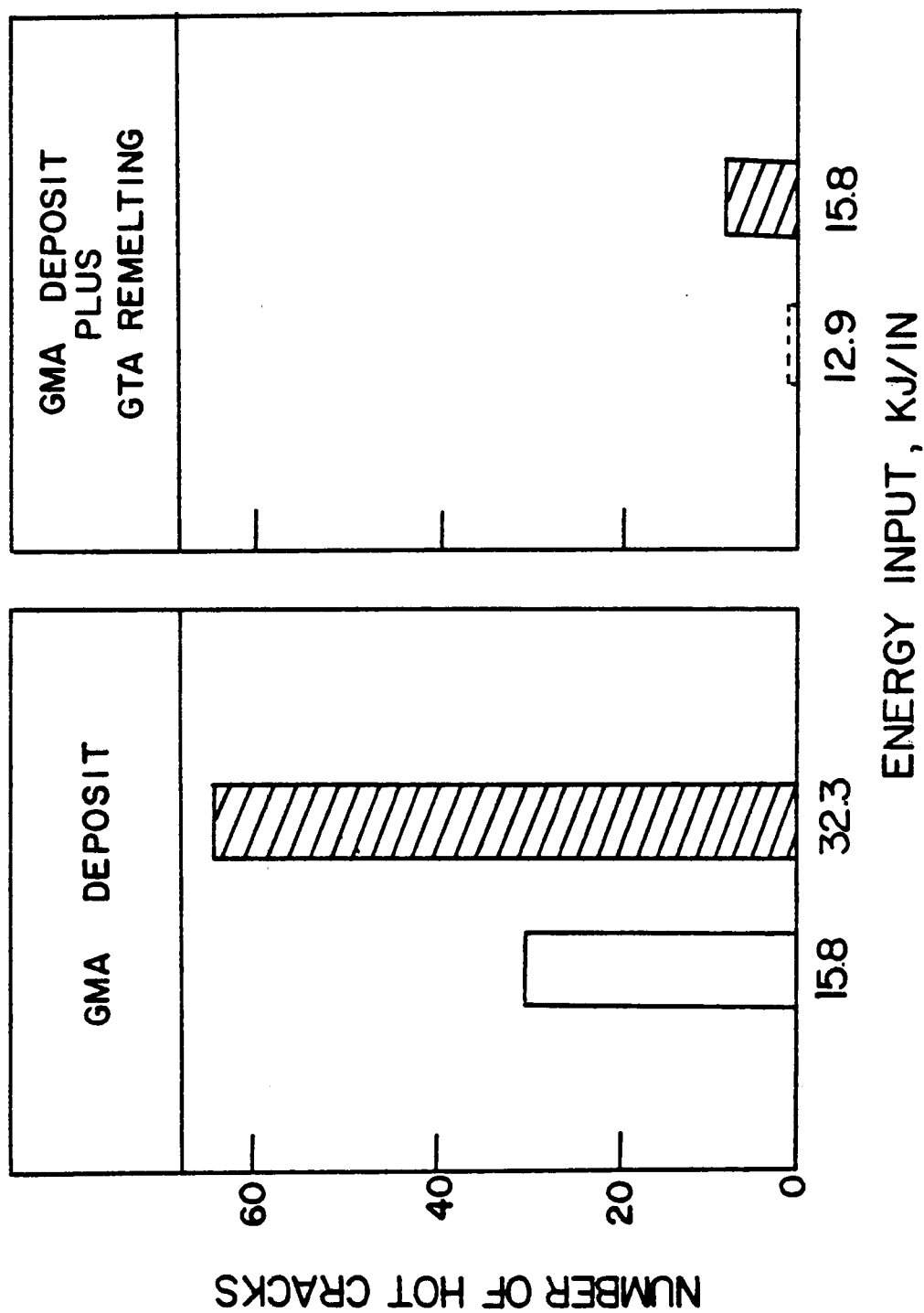


Figure 43. Number of Hot Cracks as a Function of Energy Input for GTAW and GMAW Processes (Weld Metal WI).

4C. Effect of Multiple Weld Metal HAZ Thermal Cycles

With the increasing use of heavy section weldments, repair after fabrication and repair in service become more common. The understanding of the effect of multiple thermal cycles on weld metal fissuring becomes important due to increased utilization of fully austenitic stainless steel structures and increased section thickness. The Fissure Bend Test was utilized to determine the effect of multiple weld metal HAZ thermal cycles on the fissuring tendency of austenitic stainless steel weld metal. Two commercial consumables; a fully austenitic GTAW deposited wire (W1) and a $\frac{1}{2}$ FN austenitic stainless steel SMAW electrode (E2) were evaluated together with the Type 304 base plate (ferrite potential of 5 FN).

Twelve beads in two layers of E2 were deposited on the 304 stainless steel base plate by SMAW process (standard Fissure Bend Test conditions), while eighteen beads in three layers of W1 wire were GTAW deposited (in order to give the same pad height). Then, sixteen autogenous GTA remelting beads were made as shown schematically in Figure 44.

The welding conditions for depositing weld pads and GTA remelting are presented in Table 18. By sequencing the autogenous GTAW remelt beads as shown in Figure 44, the HAZ of the previous bead along the pad surface experience multiple exposures of an identical nature. For example the first remelt bead (left side of pad) experience only one HAZ cycle from the second bead (immediately to the right) and then the second bead receives two HAZ experiences from the third bead which is a double remelt pass and so forth until the last bead (which is remelted five times) exposes the preceding bead to five HAZ excursions. The welding conditions were carefully controlled to insure a precise remelting procedure so that the HAZ excursions were identical.

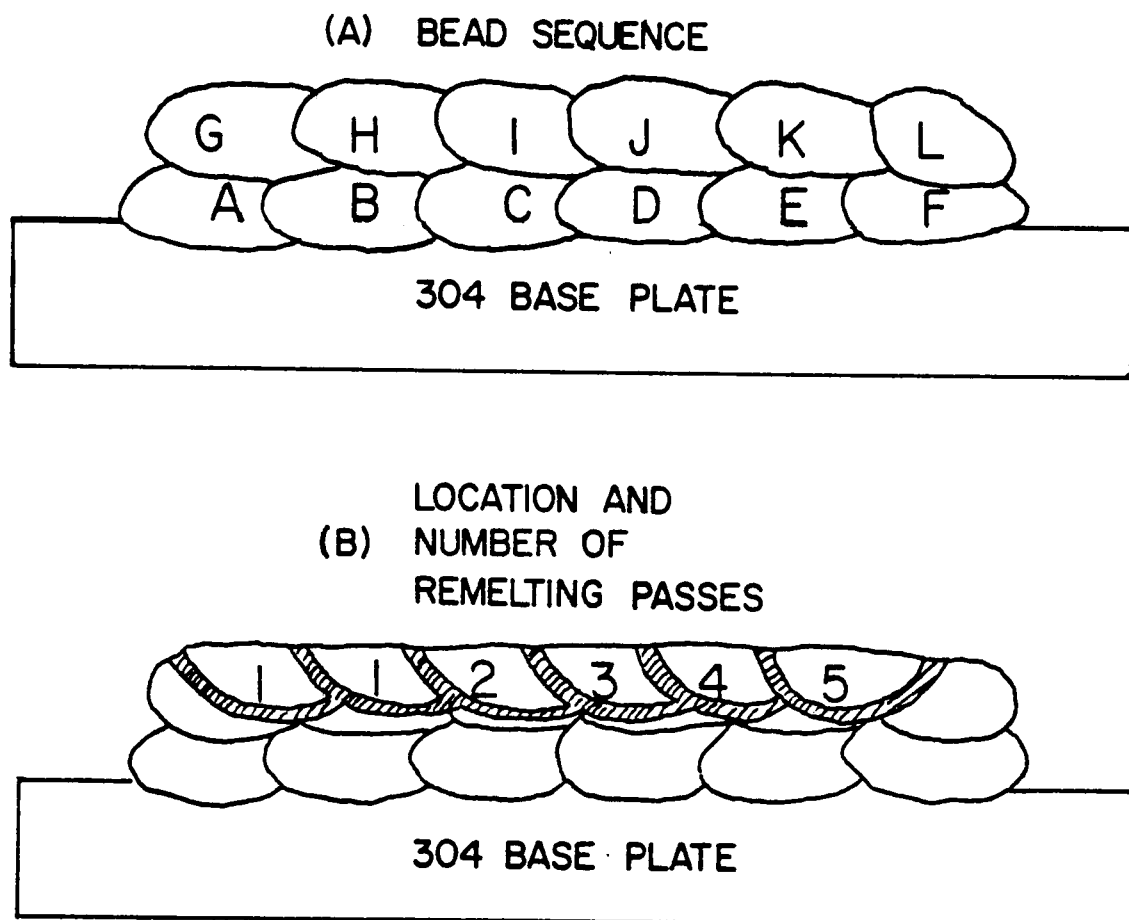


Figure 44. Typical Weld Pad Cross-Sections Showing Bead Sequence for Multipl HAZ Study.
(a) GTA or SMA Deposit Bead Sequence.
(b) GTA Remelting Passes.

Table 18. Welding Parameters for As-Deposited Weld Metal and GTA Remelting.

Weld Metal	Process	Voltage	Current	Travel Speed	Interpass Temp.
W1	GTAW deposit	10.8 V	195 Amp	7.6 ipm	250°F
	GTAW remelting	9.5 V	180 Amp	7.7 ipm	90°F
E2	SMAW deposit	25.0 V	100 Amp	7.8 ipm	200°F
	GTAW remelting	10.5 V	180 Amp	8.8 ipm	90°F
B1	GTAW remelting	10.5 V	180 Amp	8.8 ipm	90°F

The remelted pads were then ground, polished, electrolytically etched, and bent in the Fissure Bend Test device. Figure 45 shows the results of the tests. It was found that the number of cracks in the HAZ increased as the number of multiple thermal cycles experienced by the HAZ increases. With the fully austenitic stainless steel weld metal 12 cracks were found in a double HAZ and 16 cracks were found in the triple HAZ. In the four cycle exposed HAZ 30 cracks were found. Finally, 70 fissures were found in the five cycle exposed HAZ. In the $\frac{1}{2}$ FN weld metal (which was deposited by SMAW) no cracks were found in the double HAZ and only two cracks were found in the triple HAZ but 10 to 20 fissures were found in the four and five cycle exposed HAZ's. No cracks were found in the high ferrite containing weld metal (B1) even after five HAZ exposures.

The microstructures of the tested welds are shown in Figure 46. The measured FN is consistent with microstructural observations. Transverse sections of the bent weld pads are shown in Figure 47. The fissures occur primarily along grain boundaries or subgrain boundaries in the HAZ and terminate at the fusion boundaries. A partial explanation would revolve around liquation of the segregated grain boundary region produced upon solidification and/or those boundaries with enhanced segregation produced during thermal cycling. In addition, the continued HAZ straining by subsequent weld passes no doubt plays an important part.

In order to study the fracture morphology of the fissure surfaces, SEM fractography was performed. Bent and fissured weld pads were cut into a $\frac{1}{4}$ " x $\frac{1}{4}$ " x 2" samples. The samples were broken after immersion in liquid nitrogen to expose the fissure surfaces. Two types of fissures fracture morphology were observed. The first fracture morphology is

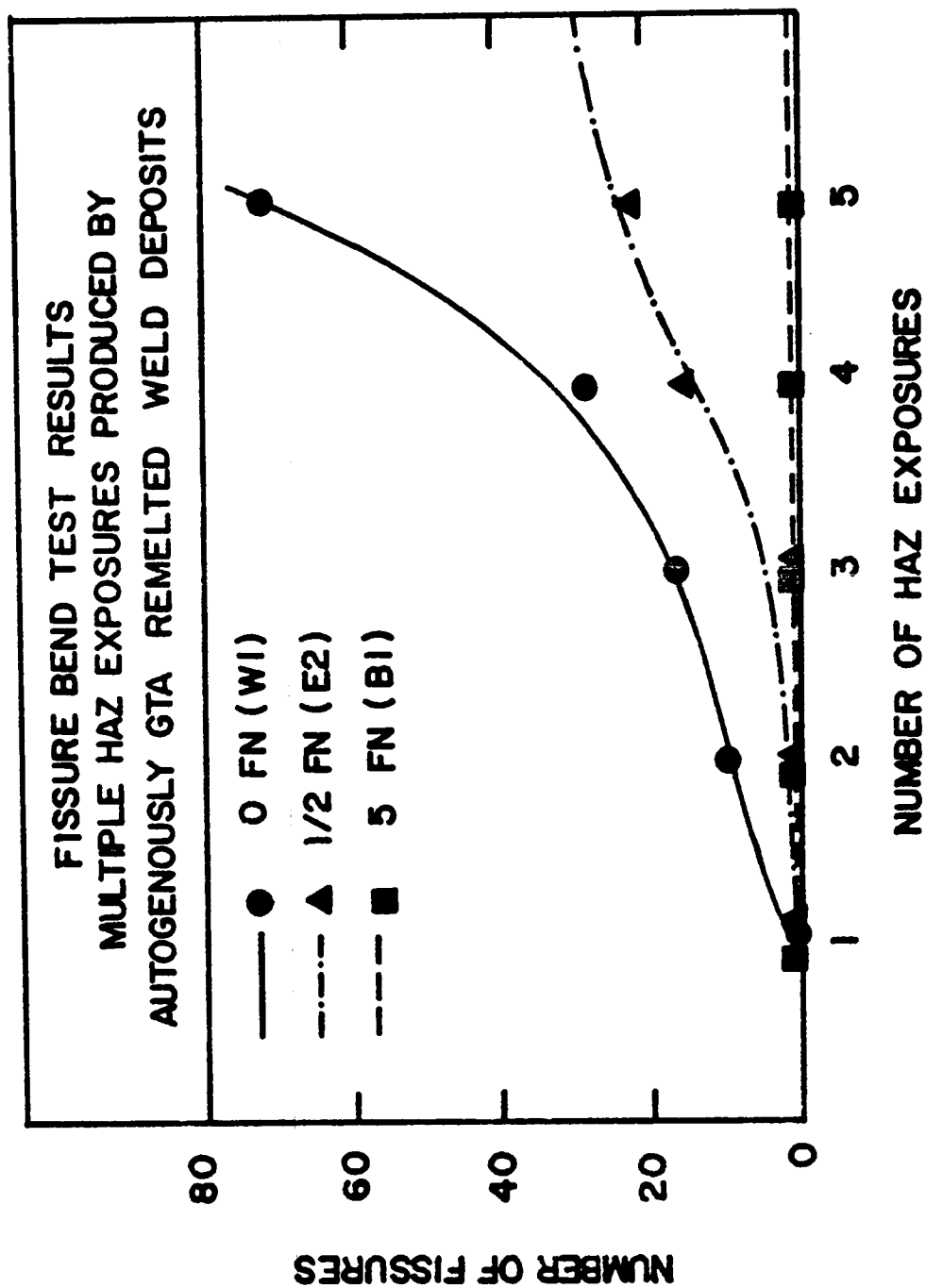


Figure 45. Fissure Count as a Function of Multiple HAZ Thermal Cycles.

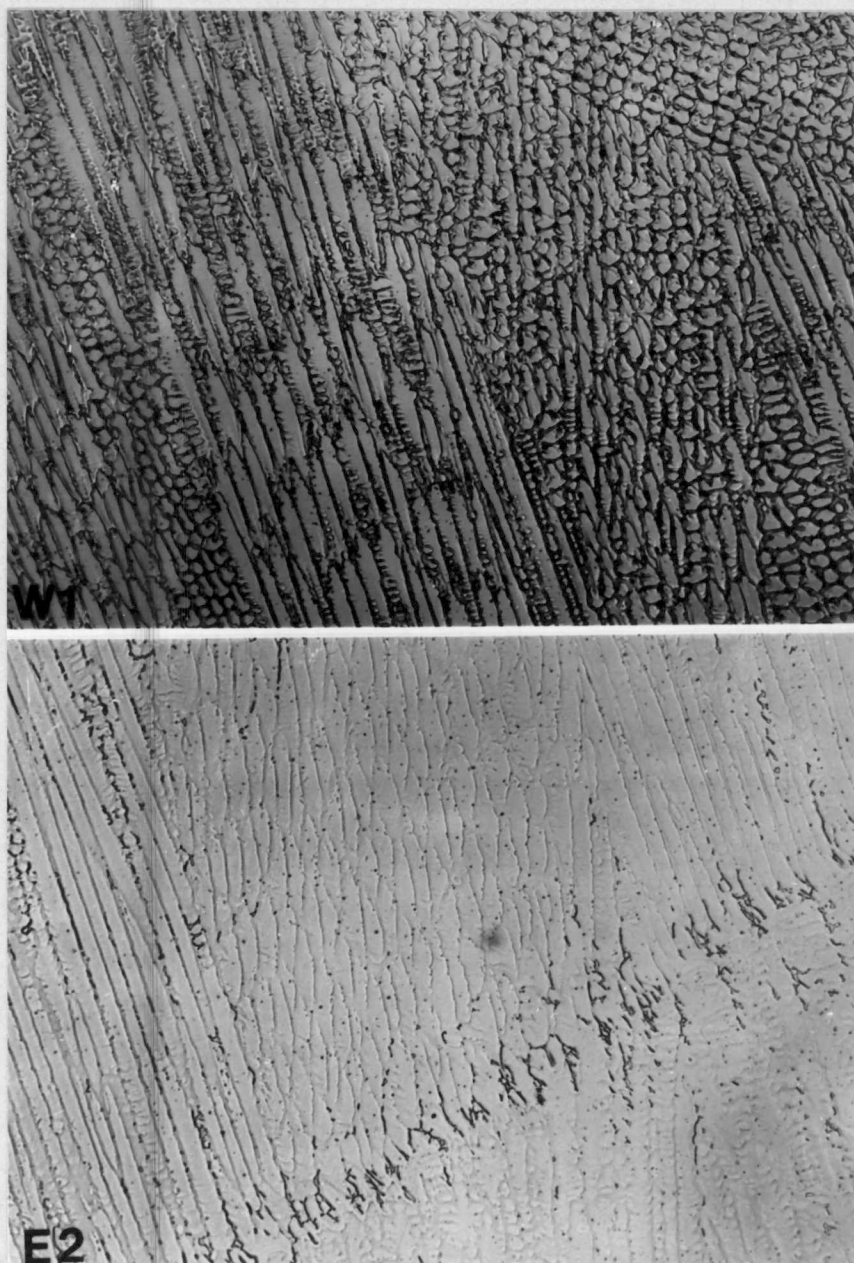


Figure 46. Microstructures of GTA Remelted Fully Austenitic Stainless Steel Weld Metal. (OLM, 200X)
(a) W1, FN = 0.0.
(b) E2, FN = 0.5.

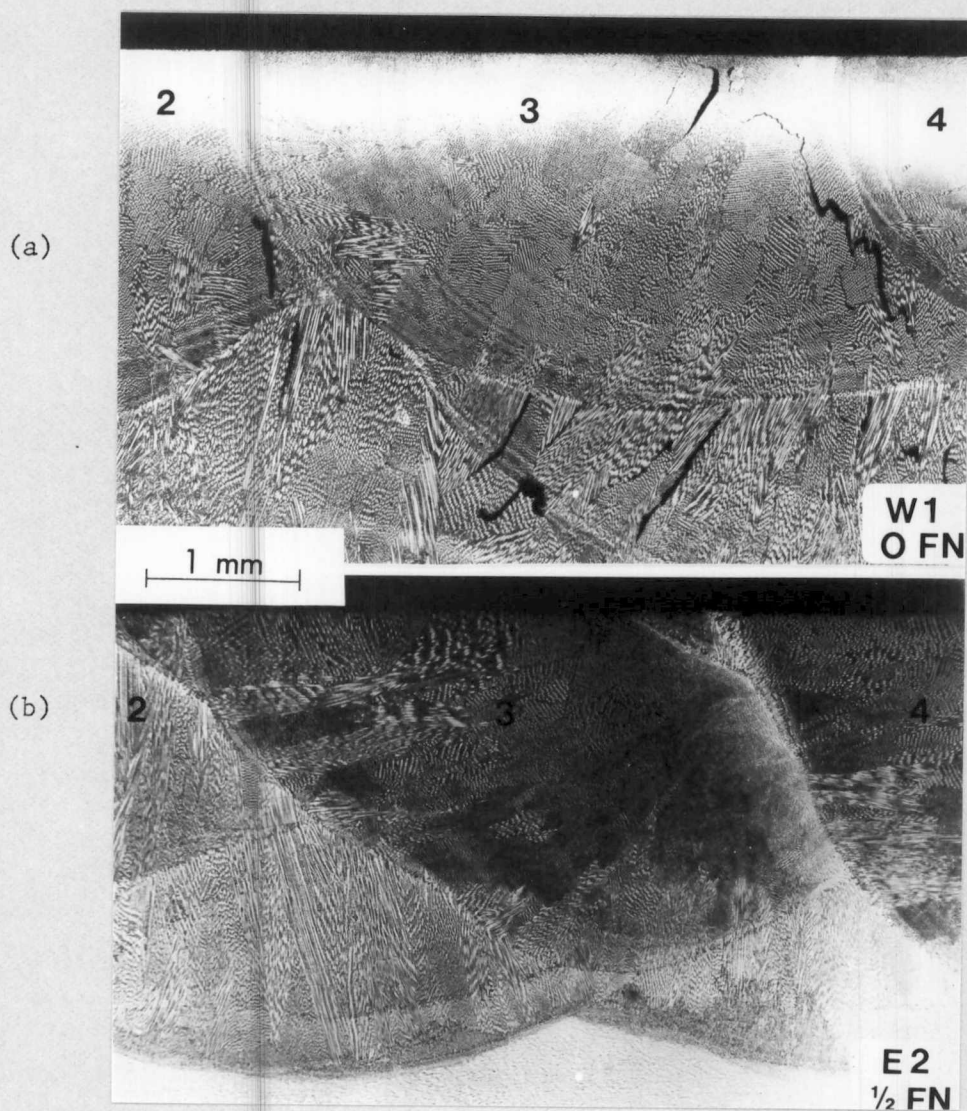


Figure 47. Effect of Multiple Thermal Cycles on Austenitic Stainless Steel Weld Metal. (OLM, 50X)
(a) W1 Weld Metal. (b) E2 Weld Metal.
(Number on Fusion Zone Indicates Multiplicity of Remelting Passes.)

shown in Figure 48. The schematic location of a broken surface is shown in Figure 48a and an SEM fractograph of the fractured surface is shown in Figure 48b. Area A is in the last remelted bead and area B is in the weld metal HAZ which was exposed to five thermal cycles. The fusion line is indicated by a broken line in Figure 48b. A high magnification fractograph of area A is shown in Figure 49. A typical ductile overload region with uniform dimple size is noted. The cellular dendritic type subgrain boundaries and grain boundaries are clearly defined by the dimple ridges. Note that there are appreciable amounts of inclusions/second phase particles contained in the dimples. Area B in Figure 48b is a typical type of fissure surfaces. The fissures of this type appear along grain boundaries and have a smooth or liquated appearance. This appearance is clearly shown in the high magnification fractograph in Figure 50.

The second type of fissure surface morphology is shown in Figure 51. The fissures in Figure 51 are in the weld metal HAZ exposed to two thermal cycles during deposition of the original GTAW cold wire pad and GTA remelting of a previously GMAW deposited pad. There is no difference in appearance between these fissure surfaces. No evidence of a liquid film appeared on these fissure surfaces. These fissures may have formed in response to the thermal strains in the ductility-dip temperature range.

The observation of the two types of fissure morphology from the Fissure Bend Test suggests that cracking may arise as a result of two mechanism. One involves liquation and takes place at temperatures close to the bulk solidus, while the other occurs at lower temperatures.

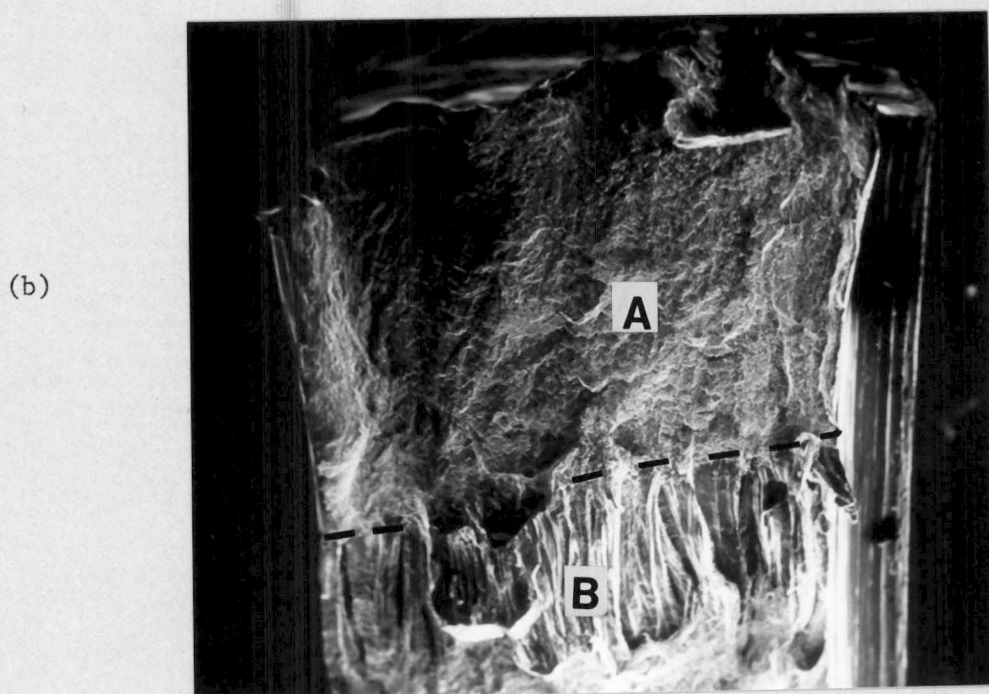
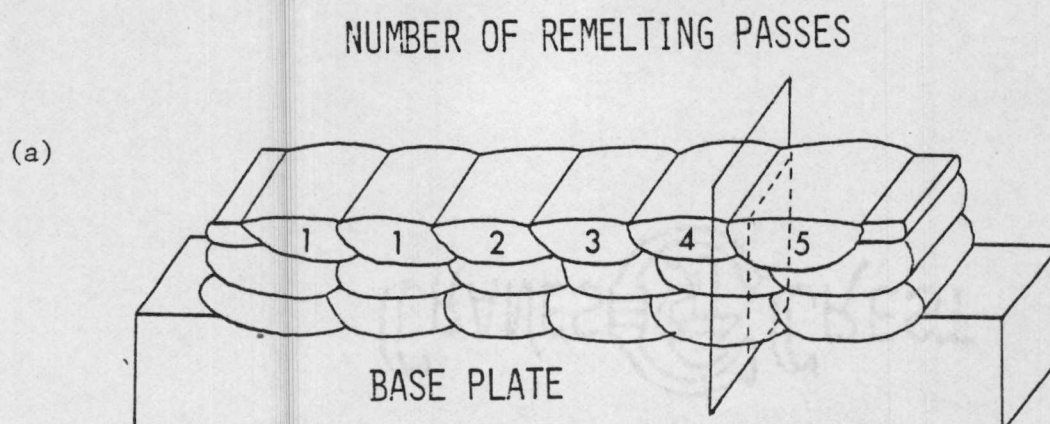


Figure 48. Fracture Surface of W1 Weldment.
(a) Location of the Fracture Plane.
(b) SEM Fractograph of Fractured Surface. (SEM, 20X)
(A: Deposit Region)
(B: Multiple HAZ Exposure Region)

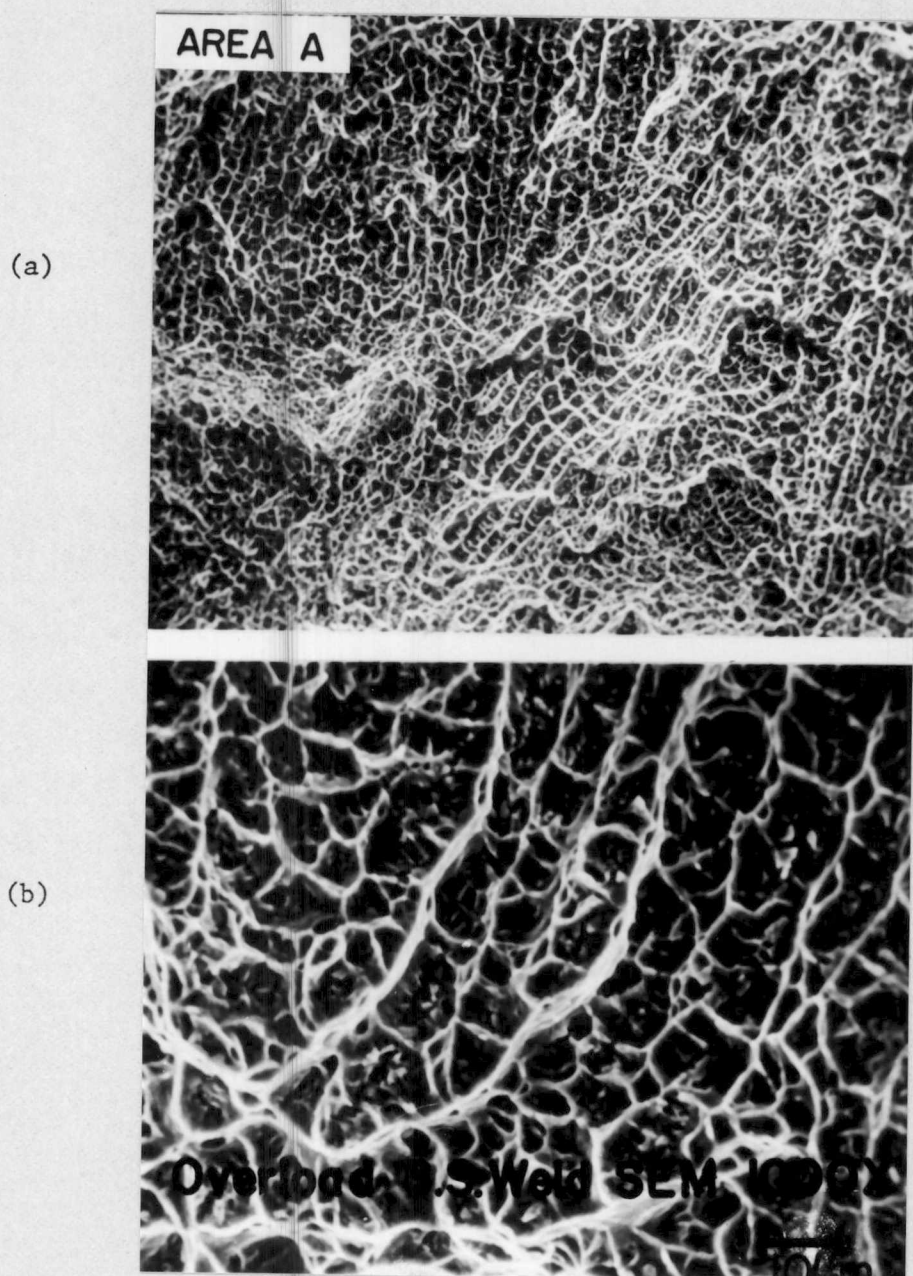


Figure 49. Fracture Surface of Deposit Region from W1 Weldment.
(Area A in Figure 48)
(a) SEM, 100X. (b) SEM, 1,000X.

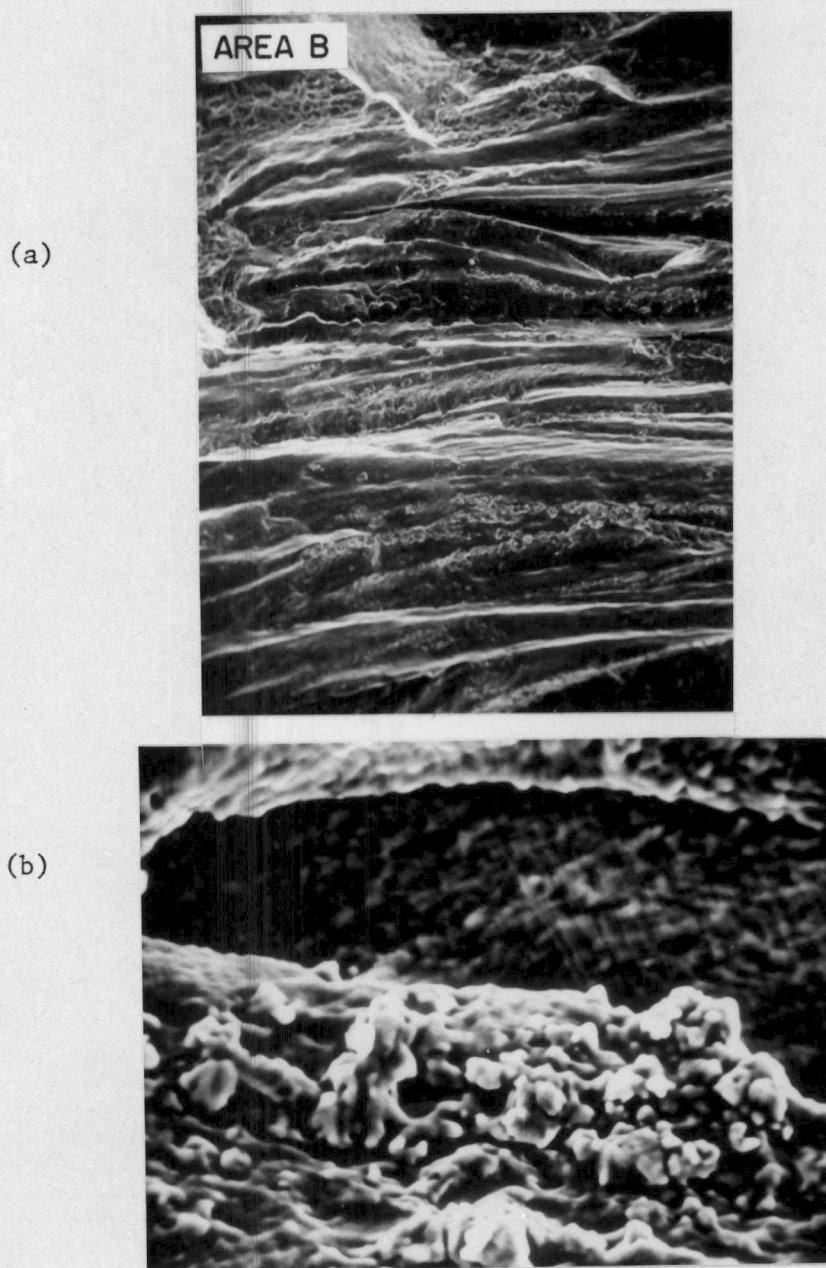
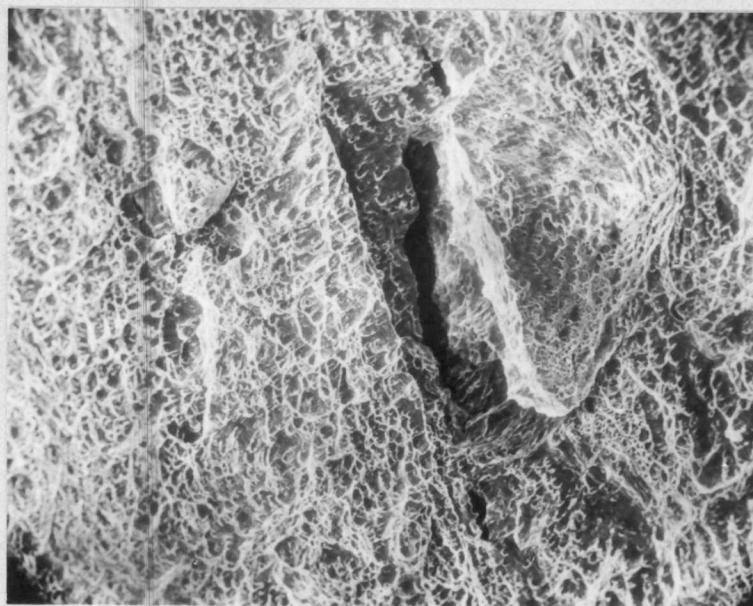


Figure 50. Fracture Surface of Multiple HAZ Exposure Region from W1 Weldment. (Area B in Figure 48)
(a) SEM, 100X. (b) SEM, 1,000X.

(a)



(b)

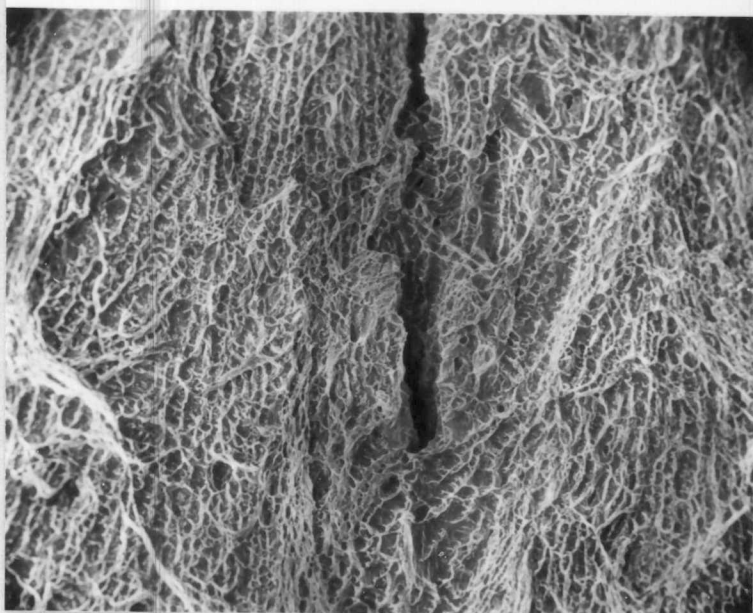


Figure 51. Fracture Surface of Austenitic Stainless Steel Weld Metal (W1) HAZ. (SEM, 200X)
(a) GMAW Deposit Plus GTAW Remelting.
(b) GTAW Deposit.

In order to study the segregation of minor elements on the fissure surface, energy dispersive X-ray analysis was performed on the broken surface of weld metal HAZ as shown in Figure 52. "A" represents a liquated fissure surface. X-ray analysis revealed the presence of Si on the fissure surface but this indication was absent in the overload region B. In addition, Mn was slightly higher when the Si was present. No other significant difference in elemental distribution was noted. The detailed study on the mechanism of hot cracking and the effect of alloying elements on the hot cracking resistance of fully austenitic stainless steel weld metal is discussed in section 6 titled "Hot Cracking Sensitivity Evaluation Using Gleeble Test".

The avoidance of weaving is called to the users attention in manufacturer's electrode handbook (147). There are also cases reported where hot cracking occurred when the welding position changed from flat to vertical (with vertical welding, weaving is required to accomplish good deposition). In order to study the influence of the weaving technique on the hot cracking tendency, a torch oscillator was used in evaluation of three austenitic stainless steel weld metals W1, E2, and B1. W1 and E2 weld pads were deposited by GTAW process and SMAW process respectively. The pads were then surface ground and remelted by GTAW process using a weaving pattern and the welding parameters as shown in Figure 53. Stringer beads 1 and 2 were made first. Weave beads 3 and 4 were then made with a dwell at the locations (dots) shown in Figure 53. The same weaving pattern and welding conditions were used for remelting the B1 base plate. A second type of weaving pattern as shown in Figure 54, was used for the SMAW deposited pad of consumable E2. Two stringer

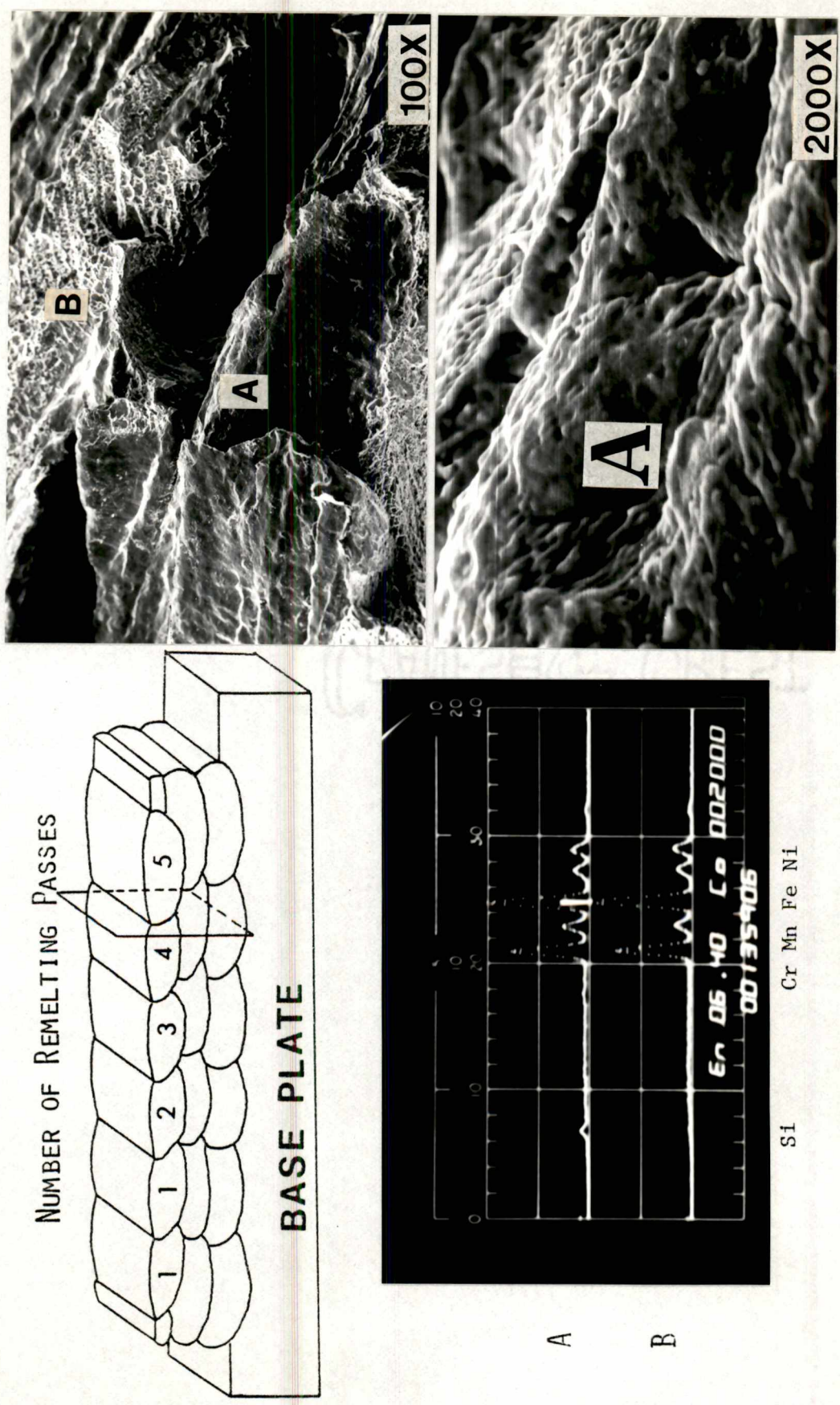
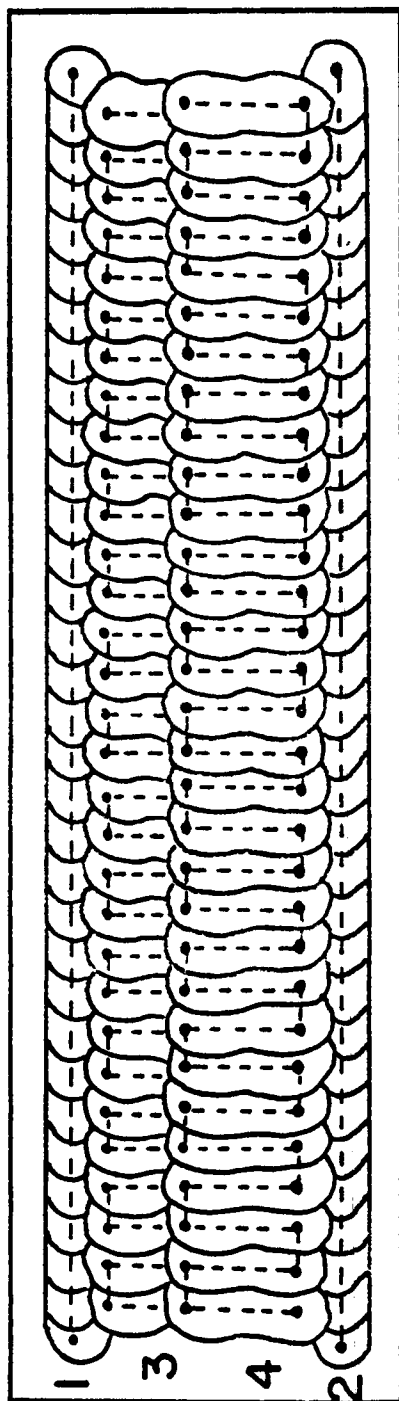


Figure 52. EDAX of Fissure of W1 Weld Metal HAZ.
(A: Fissure Surface; B: Overload Region)



WELDING PARAMETERS

**BEADS 1 AND 2 : 180 AMP, 10.0V, 8 IPM, 40 CFH (ARGON)
(STRINGER BEADS)**

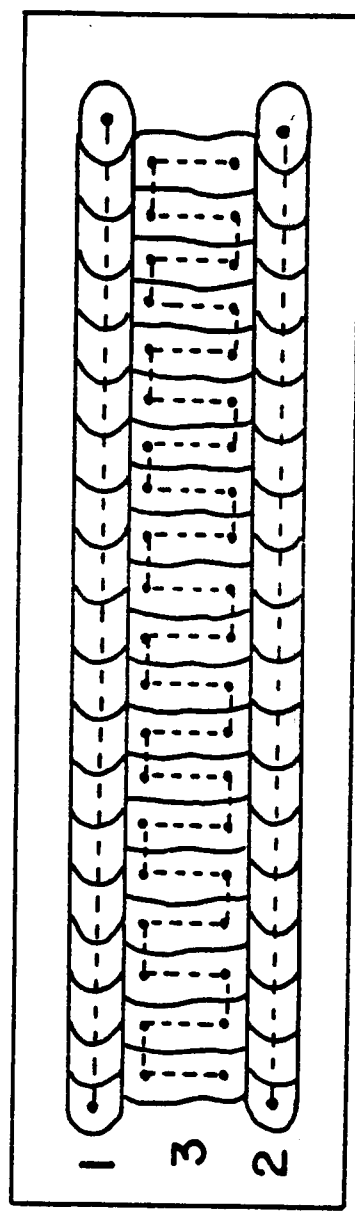
BEADS 3 AND 4 : 230 AMP, 11.0V, 5 IPM, 40 CFH (ARGON)

(WEAVE BEADS) OSCILLATORY FREQUENCY : 70 CYCLES/MIN.

OSCILLATORY AMPLITUDE : 0.25 IN.

DWELL TIME : 0.25 SEC.

Figure 53. Weaving Pattern and Welding Parameters Used in GTA Weave Bead.



WELDING PARAMETERS

**BEADS 1 AND 2 : 100 AMP, 22 V, 8 IPM, 200° F INTERPASS TEMP.
(-STRINGER BEADS)**

**BEAD 3 : 100 AMP, 22 V, 3.5 IPM, 200° F INTERPASS TEMP.
(WEAVE BEAD)**

Figure 54. Weaving Pattern and Welding Parameters Used in SMA Weave Bead.

beads were made first with 0.5" apart. The third bead was then made manually with weaving to fill up the groove. Two layers were deposited with the same pattern.

The Fissure Bend Test results and measured FN are presented in Table 19. The FN is the same for both the stringer bead and weave bead technique. With a low ferrite level ($\frac{1}{2}$ FN and 5 FN) in the weld metal, hot cracking was not evident for the GTA remelting with weaving technique. In the fully austenitic stainless steel weld metal, weaving degraded the weld metal by inducing hot cracks. In the SMAW deposit of the E2 consumable (FN = 0.5), weaving enhanced the hot cracking tendency significantly. The cracks always occurred in weld metal which was the HAZ of the subsequent weave bead. Figure 55 shows the hot cracks as they appear on the top surface of the weave bead HAZ of the autogeneously remelted GTAW W1 consumable. Figure 55b shows a sketch noting the relationship of the fusion lines and 1630°F isotherms (most cracks were found in this HAZ region). The 1630°F isotherms are about 0.050" from the fusion line (determined by Lundin utilizing the thermocouple implantation technique). The two hot cracks are in the region which was reheated three times by the subsequent weave beads. This is consistent with previous work which showed that hot cracking is enhanced by multiple cycling in the weld metal HAZ.

In summary, the hot cracking sensitivity of SMAW, GTAW, and GMAW deposits can be effectively evaluated using the Fissure Bend Test and the results closely correlate with fabrication experience. From the evaluation of several fully austenitic stainless steel weld metals using Fissure Bend Test, it was found that Mn and Mo are favored as alloying elements for preventing hot cracking. Silicon level should be kept low

Table 19. Effect of Weaving on Fissure Bend Test Results.

Material	Welding Procedure	FN	Number of Cracks
B1	GTA Remelting	5	0
B1	GTA Weaving Remelt	5	0
E2	SMA Deposit + GTA Remelting	0.5	0
E2	SMA Deposit + GTA Weaving Remelt	0.5	0
W1	GTA Deposit	0	0
W1	GTA Deposit + GTA Weaving Remelt	0	2
E2	SMA Deposit	0.5	1
E2	SMA Deposit (Weave Technique)	0.5	7

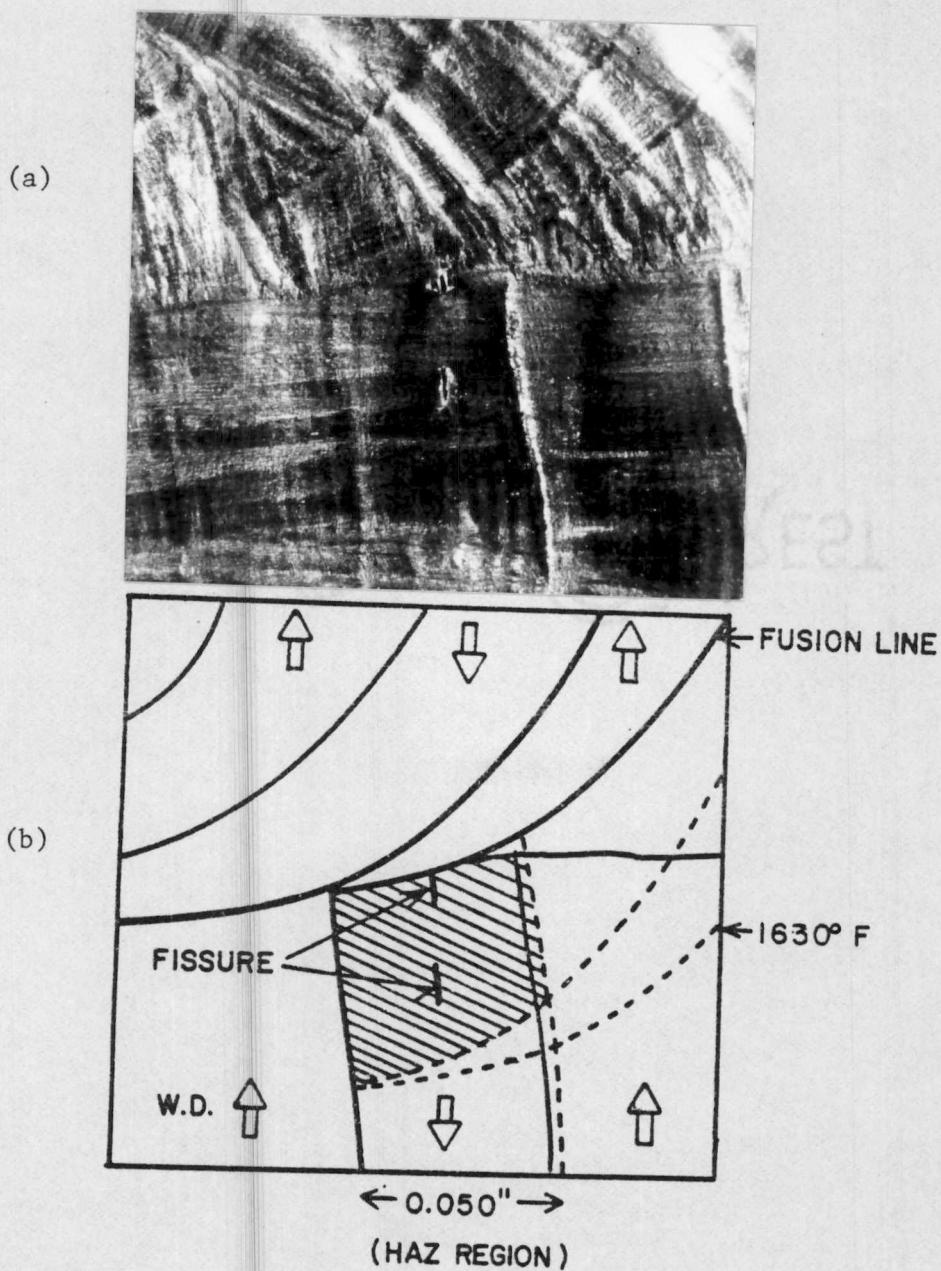


Figure 55. Fissures Resulting From Weaving Techniques of GTAW Remelted W1 Consumable.
 (a) Fissures on the Top Surface of Weave Bead. (20X)
 (b) Sketch of the Top Surface of Weave Bead.

to reduce hot cracking tendency. The effect of these and other alloying elements on hot cracking resistance was extensively studied using Gleeble technique and is discussed in a latter section.

The influence of changes in welding technique such as shielding gas, energy input, GTA remelting, weave or stringer bead on the hot cracking sensitivity was determined. It was found that the addition of nitrogen in the shielding gas had little effect on the fissuring tendency of the Type 304 base plate even though the FN dropped from 5.0 to 0.2. Fissuring tendency is enhanced as the energy input increases for both the GMAW and GTAW processes. Autogenous GTA remelting technique can effectively reduce the hot cracking tendency. Weave techniques used during welding enhance the hot cracking tendency significantly because some areas adjacent to the fusion line are reheated to high temperatures two or three times by the subsequent weave beads.

The most important result obtained in this study is that multiple thermal cycles imposed on weld metal HAZ can significantly increase the hot cracking tendency of austenitic stainless steel weld metals. This part of research is investigated more thoroughly using Gleeble technique and a new theory about the "Reheat Hot Cracking" is developed in subsequent sections.

5. Hot Ductility Behavior and Effect of Multiple Thermal Cycles on Ferrite Content

The material adjacent to the weld is subjected to temperatures varying from the melting temperature to the initial temperature of the plate. While one portion of the weld zone is being heated, another part is cooling from the melting temperature. Thus, it can be realized that the ductility of a material will control cracking behavior as the

thermal strains must be accommodated both elastically and plastically. Austenitic stainless steel has a relatively low heat conductivity and a relatively high thermal expansion coefficient. Thus, ductility plays an important role in the fabrication and service performance of austenitic welds in heavy-sections. Many investigators found a good correlation between fissuring sensitivity and the hot ductility measured under welding conditions (109, 113-115). The basic methodology utilizes the synthetic specimen technique (Gleeble apparatus). In this work, the Gleeble technique was used to study the hot ductility of the austenitic stainless steel weld metals. The effect of composition, peak temperature, and number of thermal cycles on the hot ductility and the microstructural changes that occurred was investigated and correlated with weld hot cracking behavior.

Hot ductility testing consists of performing a tensile test on specimen heated and cooled in accordance with a programmed (simulated) weld thermal cycle. Testing is conducted during either the heating or cooling portions of the thermal cycle. In this work the specimens tested were in the unwelded, welded, and longitudinal composite welded conditions. The details of the testing procedures were described in Chapter three. All tests were conducted in an inert atmosphere (argon purged chamber) to minimize high temperature oxidation.

Three austenitic stainless steel filler metals in strip form and one stainless steel wrought sheet were used in this portion of the study. Autogenously welded samples were prepared by GTA welding on the strip or sheet at 50 Amps, 7.5 V, and 60 ipm, using a 1/16" electrode and 40 cfh argon as the shielding gas. The welds were full penetration and controlled in width (0.25"). The chemical composition and measured Ferrite Number

of the autogenously welded samples is shown in Table 20. The thickness of the samples "A1" and "A2" is 0.20" and the thickness of the samples "L" and "H" is 0.025" and 0.031" respectively.

5A. Hot Ductility Tests on the Unwelded and Welded Materials

The results of the hot ductility tests on unwelded "A1" are shown in Figure 56. The on-heating ductility increased with temperature, reached a maximum at 2100°F and then rapidly dropped to zero at 2400°F (zero ductility temperature, ZDT). When specimens were tested on-cooling from the peak temperature of 2400°F (results compared with those obtained on-heating), it was found that heating to the peak temperature of 2400°F significantly reduced the hot ductility at all testing temperatures on-cooling. Similar results were obtained for the unwelded "A2" as shown in Figure 57. Unwelded "A1" showed better hot ductility behavior than unwelded "A2".

Figure 58 shows the macroscopic appearance of tested specimens from unwelded "A2". Specimen A2-H2100 was tested at 2100°F on-heating and specimen A2-C2100 was tested at 2100°F on-cooling from a peak temperature of 2400°F. It can be clearly seen that the difference in ductility between these specimens is significant. In addition to the lower ductility, specimen A2-C2100 also showed a shiny, rough appearance in the center hot zone. The examination with SEM shows that this region exhibits liquated grain boundaries.

The hot ductility test was also conducted on a GTA welded "A1" (all weld metal gage section). This testing was achieved by using a transverse composite weld specimen as shown in Figure 20, page 87. Figure 59 shows the results of the hot ductility tests on welded "A1". The on-heating ductility increased slightly with temperature, reached a maximum

Table 20. Chemical Composition and Measured Ferrite Number of Austenitic Stainless Steel Welds Used for Gleeble Hot Ductility Testing.

Material	Composition, Wt. %										FN
	Cr	Ni	Mn	Mo	Si	P	S	C	N		
A1	24.34	21.49	4.49	2.16	0.06	0.007	0.014	0.021	0.160	0	
A2	20.30	16.00	9.35	2.57	0.24	0.005	0.011	0.017	0.073	0.5	
L	18.36	8.11	1.69	N.A.	0.43	0.021	0.005	0.048	0.060	5.0	
H	23.40	11.36	1.54	N.A.	0.25	0.011	0.010	0.012	0.060	25.0	

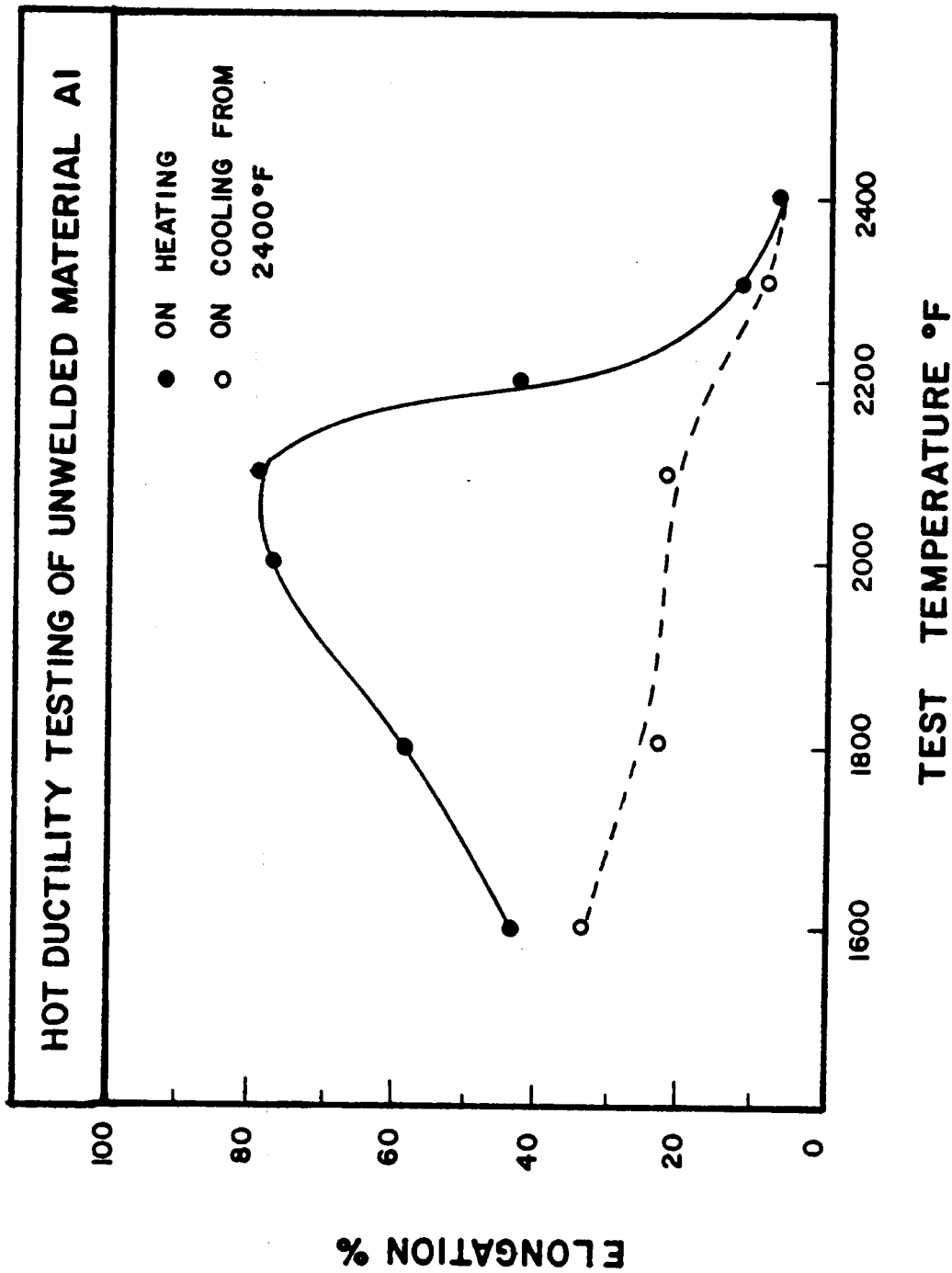


Figure 56. Hot Ductility Test Results for Unwelded Material "AI".

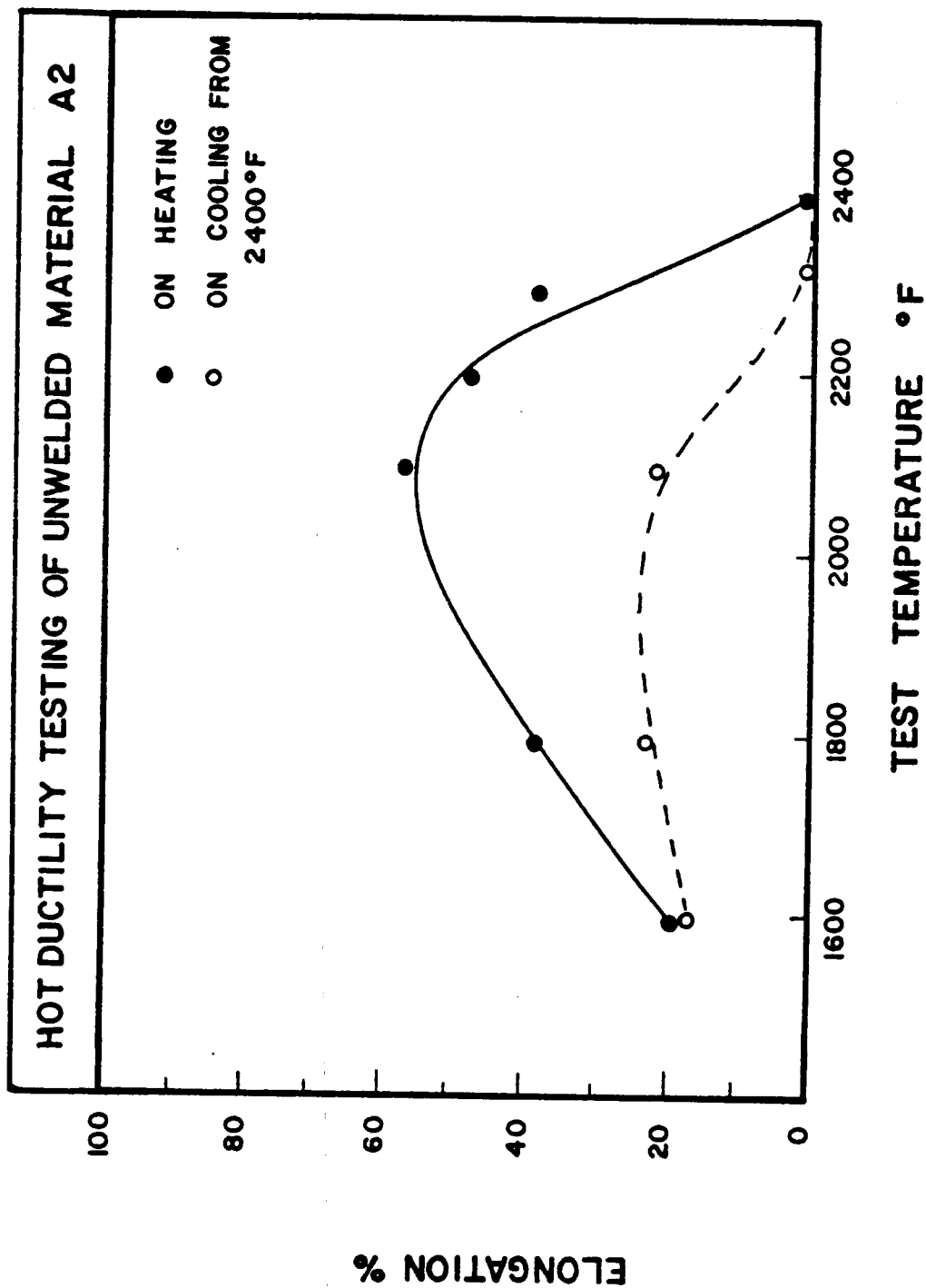
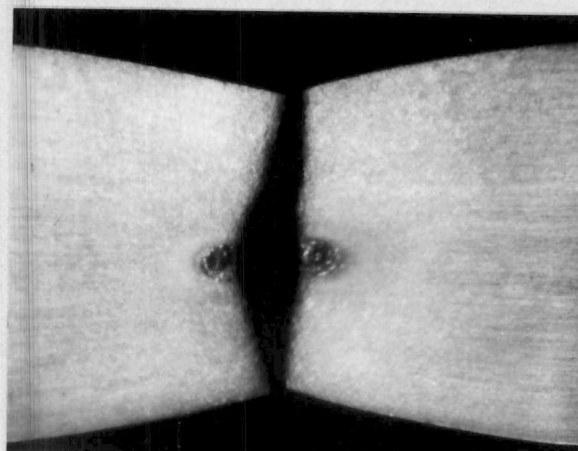


Figure 57. Hot Ductility Test Results for Unwelded Material "A2".

(a)



(b)

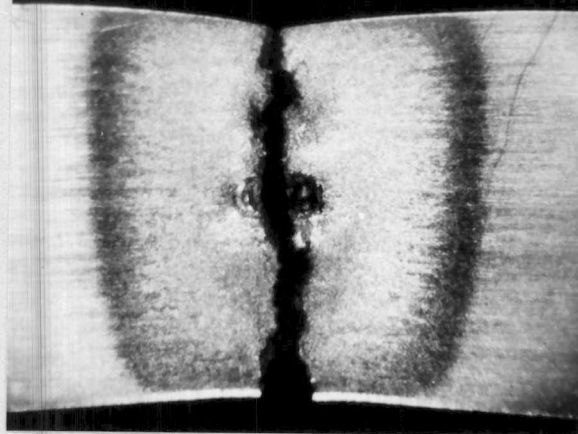


Figure 58. Hot Ductility Tested Specimens of Unwelded Material "A2".
(a) Tested On-Heating to 2100°F.
(b) Tested On-Cooling to 2100°F After Heating to 2400°F.
(4X)

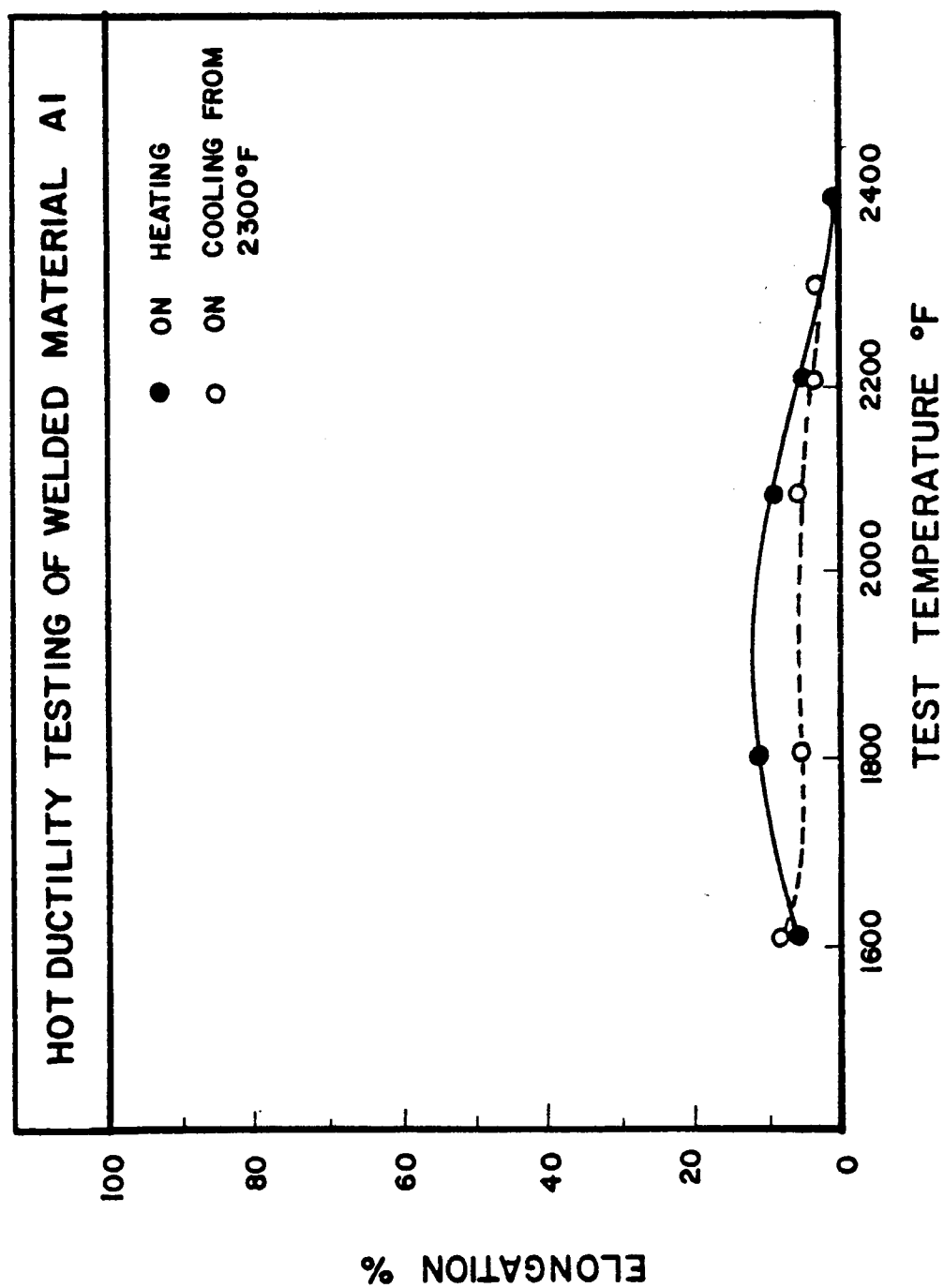


Figure 59. Hot Ductility Test Results for Welded Material "A1" (Transverse Composite Sample).

at about 1900°F and then decreased to zero at 2400°F . When specimens were tested on-cooling from the peak temperature of 2300°F and the results compared with those obtained on-heating, it was found that heating to the peak temperature of 2300°F reduced the hot ductility in the temperature range 1800°F to 2200°F on cooling. It is clearly seen that the hot ductility of the welded material is lower than the unwelded material (see Figures 56 and 59). This is most likely related to the fact that the weld metal experiences segregation during solidification and large columnar grains are produced. The reduced grain boundary area tends to produce a greater concentration of low melting constituents segregated during solidification and these may become even more highly segregated during subsequent thermal cycling.

The results of the hot ductility tests for welded "A2" are shown in Figure 60. It is clearly seen that the hot ductility of the fully austenitic stainless welded "A1" is lower at all temperatures tested than that of the 0.5 FN containing welded "A2". The higher hot ductility behavior of welded "A2" leads to a greater fissuring resistance as will be discussed in a later section.

The lowering of the ductility at high temperatures has been discussed by Yeniscavich (112); (1) Liquation begins at about 2100°F by formation of isolated areas of very thin films that occupy only a small fraction of the grain boundary area. (2) As the temperature increases, the extent of the film increases and thus, the fraction of grain boundary area covered by the liquid film increases also. (3) When a critical fraction of the grain boundary area has liquated the ductility of the specimen will be reduced to zero because of the notching effect caused by the distributed liquid films. The result of heating to a high temperature (2400°F)

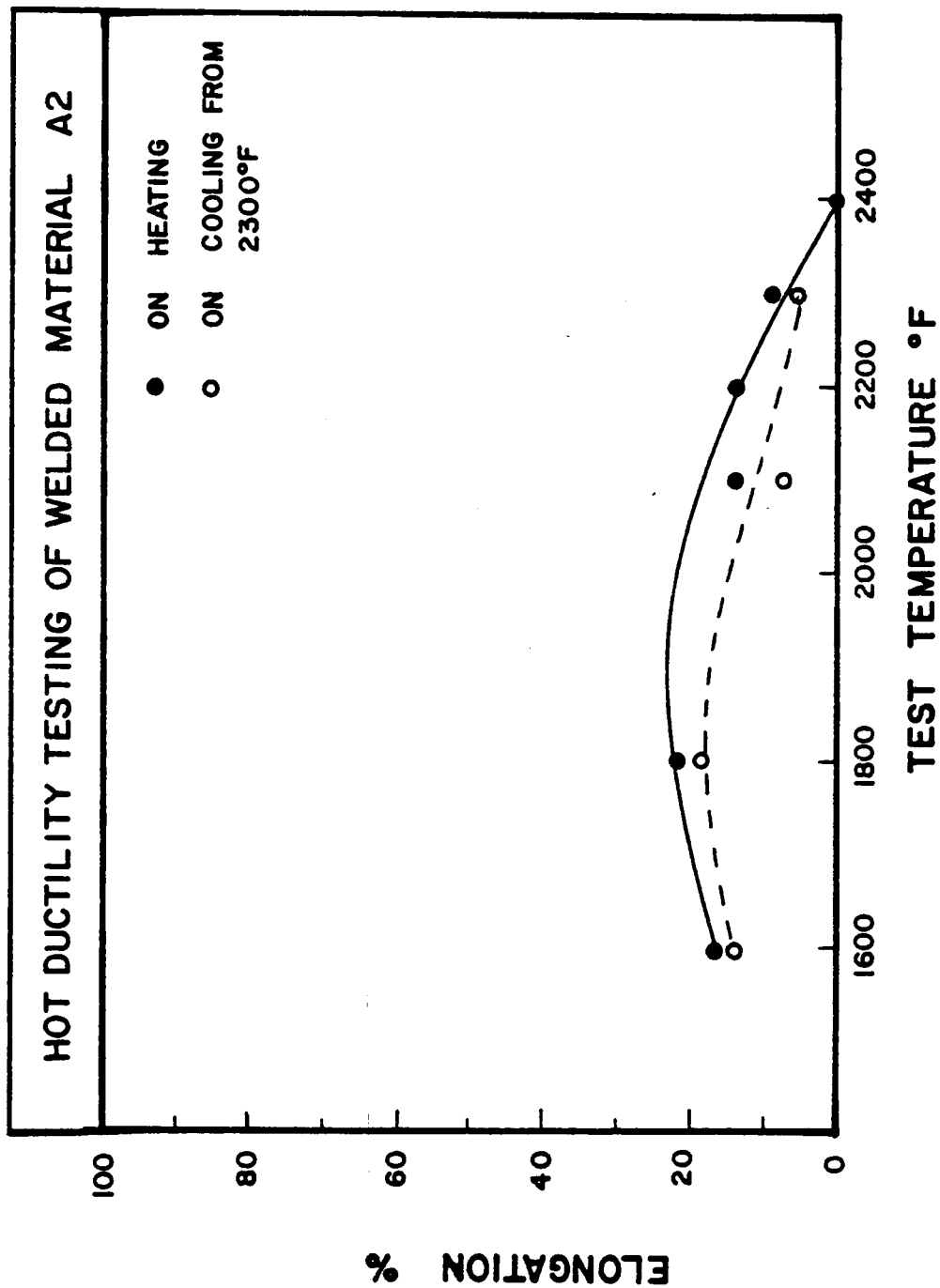


Figure 60. Hot Ductility Test Results for Welded Material "A2" (Transverse Composite Sample).

is the growth of the liquated areas in the grain boundaries. On subsequent cooling, these liquated areas may be supercooled and cause a displacement of the maximum ductility to a lower temperature.

The results of the SEM fractography studies in the current work show no evidence of a liquid film on the surface of specimens tested on-cooling below 2100°F. Thus an alternative explanation to that proposed by Yeniscavich, of the continued reduction in ductility, is necessary. The probable reason for the reduced ductility of specimens tested on-cooling to lower temperatures is the embrittlement of the grain boundaries by the solidified liquid films and/or partitioning of elements by liquid film formation resulting in lower ductility compositions.

5B. Effect of Multiple Thermal Cycles on the Hot Ductility

The effect of multiple thermal cycles on the hot ductility was determined with the Gleeble technique using the longitudinal composite weld specimen as shown in Figure 21, page 88. The reason for using this type specimen tests is to simultaneously evaluate the ductility of all regions - weld metal, HAZ and base metal and to correlate the results with those from the hot cracking tests. Three austenitic stainless steels ("A1", "L", and "H") were used in this study. "A1" is fully austenitic. "L" is Type 304 with ferrite potential of 5 FN and "H" is modified Type 308L with ferrite potential of 25 FN. Specimens were strained to fracture at different peak temperatures (1800°F, 2100°F, and 2300°F) after being exposed zero, two, and five prior thermal cycles to the same peak temperature. The results are graphically presented in Figures 61 to 63. Figure 61 shows the effect of peak temperature and number of thermal cycles on the ductility of the fully austenitic stainless steel "A1". It is clearly

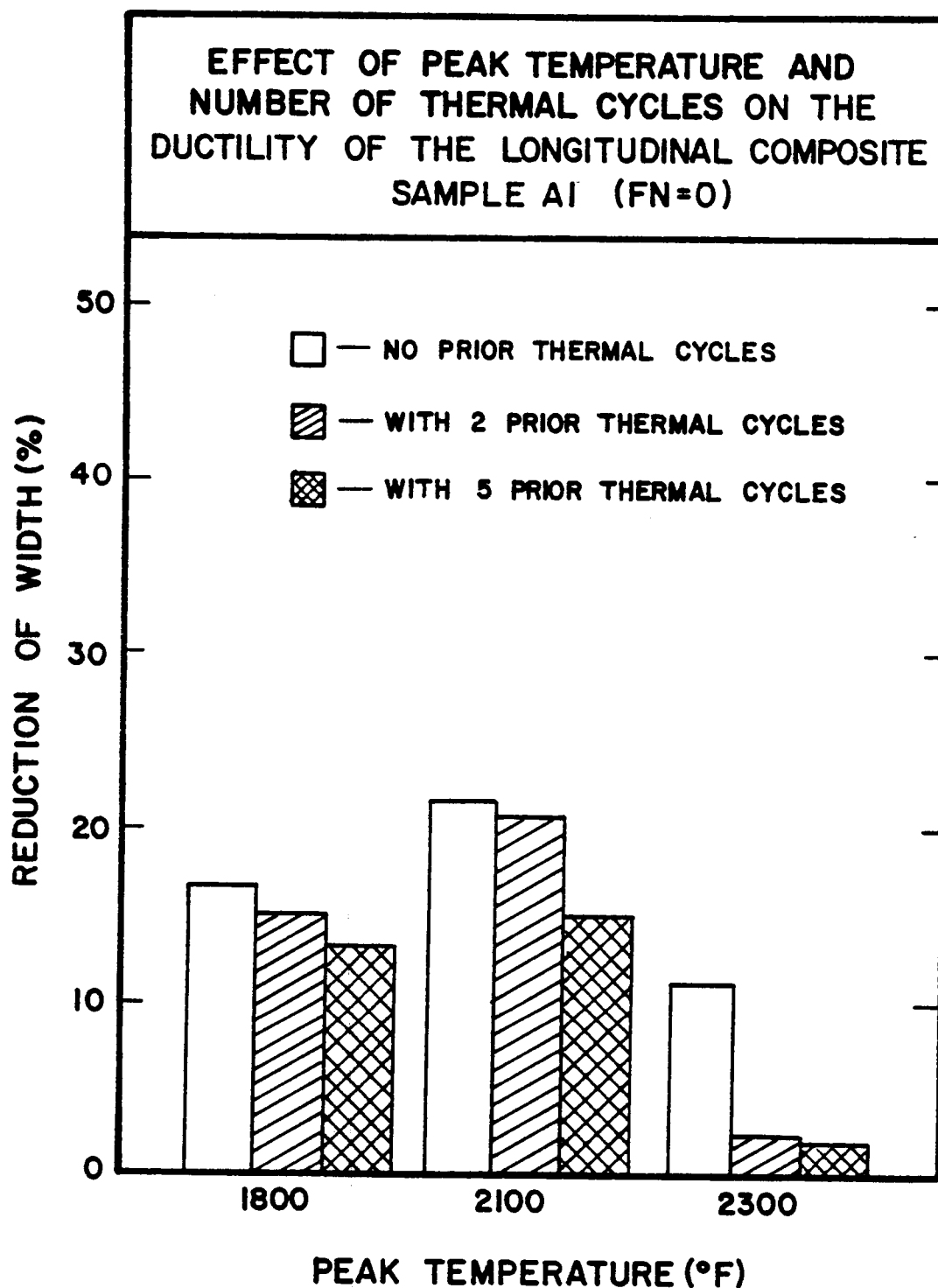


Figure 61. Effect of Peak Temperature and Number of Thermal Cycles on the Hot Ductility of Sample "A1".

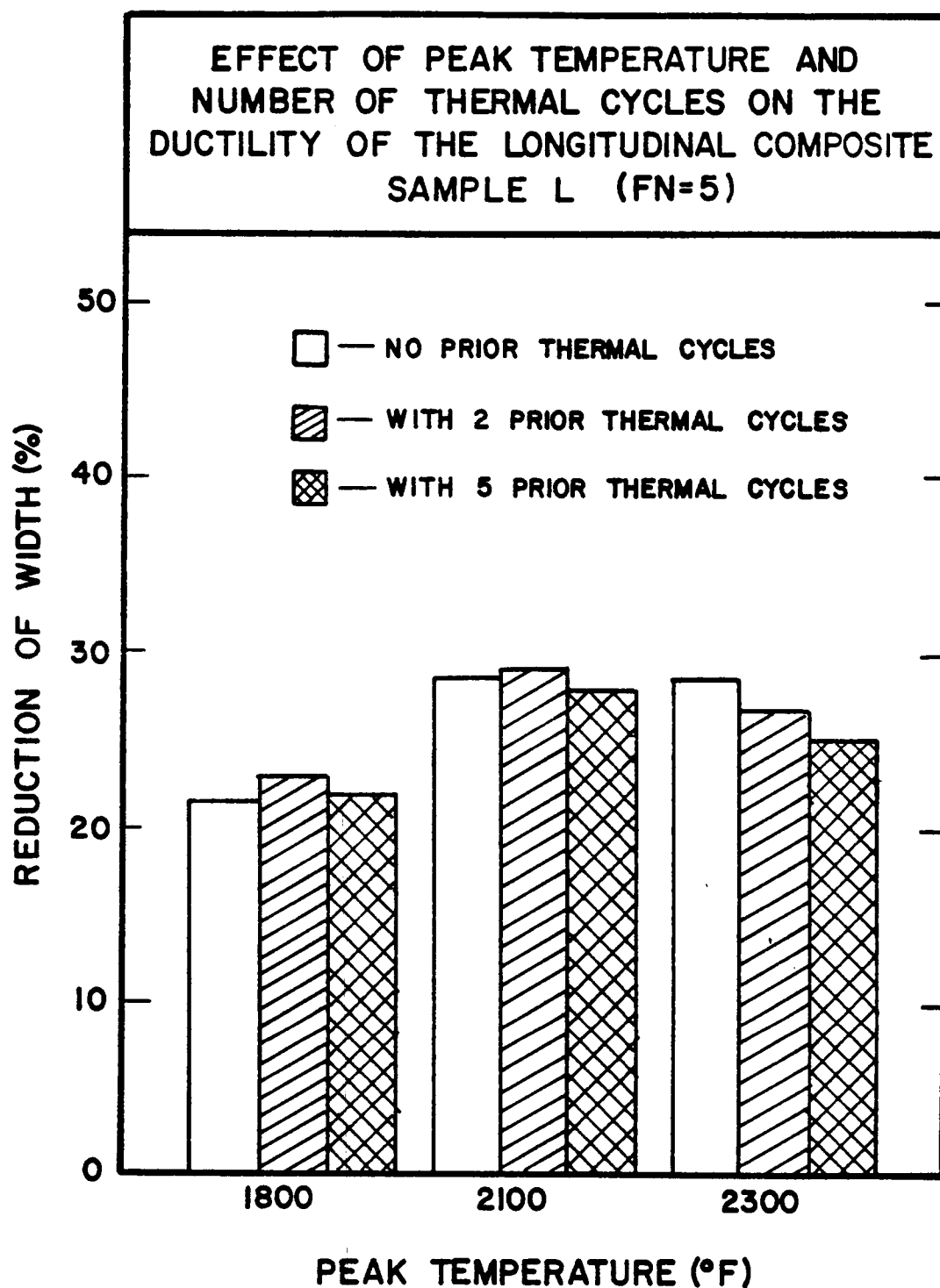


Figure 62. Effect of Peak Temperature and Number of Thermal Cycles on the Hot Ductility of Sample "L".

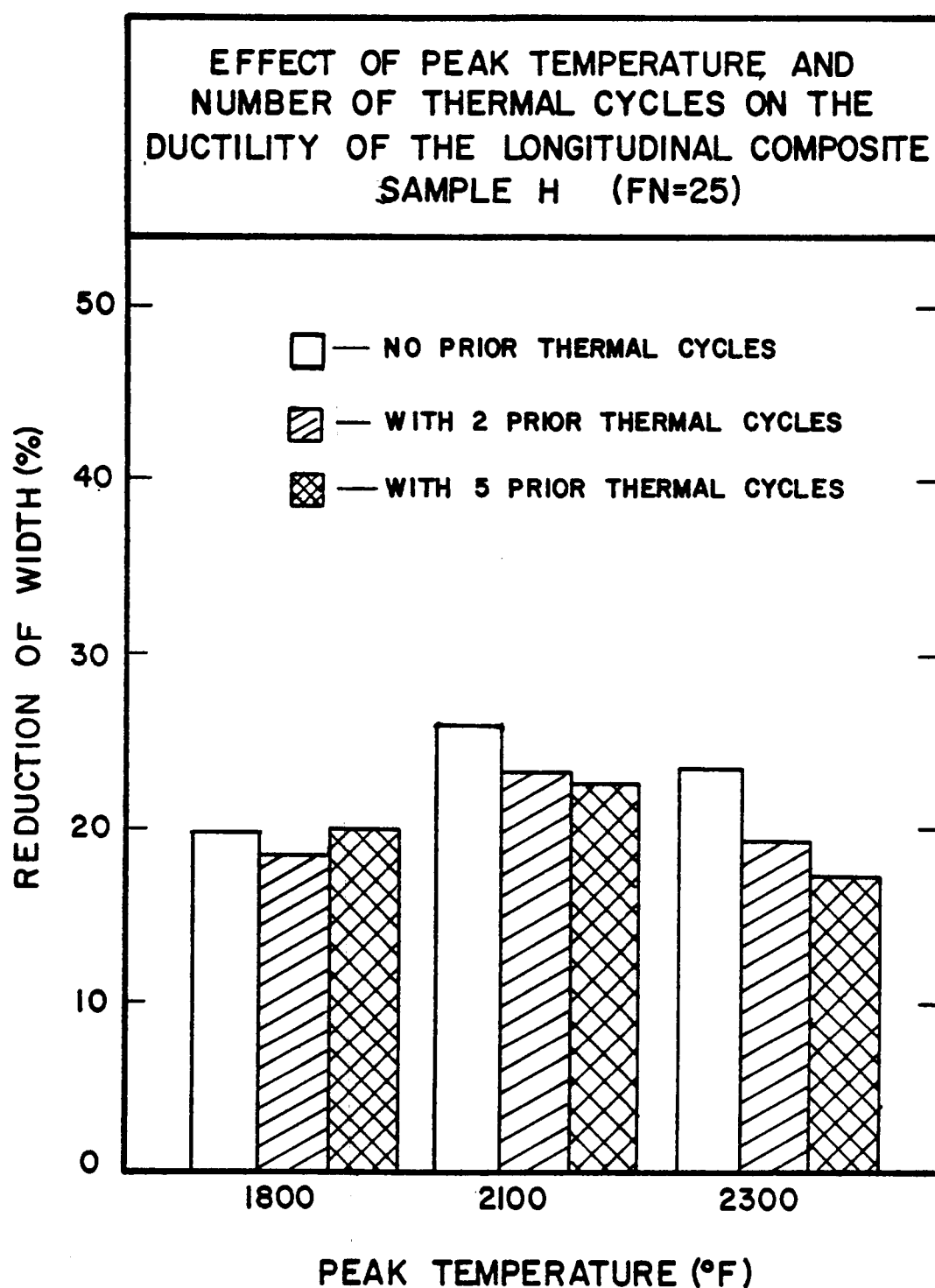


Figure 63. Effect of Peak Temperature and Number of Thermal Cycles on the Hot Ductility of Sample "H".

seen that peak temperature and number of cycles have a significant effect on the ductility at temperatures higher than 2100°F. At a peak temperature of 2300°F, the hot ductility is significantly lower than that at a peak temperature of 2100°F. The hot ductility is lower at a peak temperature of 1800°F as compared to the hot ductility at 2100°F. The multiple thermal cycling reduces the hot ductility at all testing temperatures especially at 2300°F. The cause of the low ductility is liquid film formation.

For the ferrite-containing materials ("L" and "H"), the effect of peak temperature and number of cycles on ductility is not significant as shown in Figures 62 and 63. There is no difference in hot ductility at peak temperatures of 2100°F and 2300°F. The hot ductility slightly decreased as the number of thermal cycles was increased. It appears that the hot ductility behavior at high temperatures is probably related to the amount and morphology of ferrite in the welds.

Figure 64 shows the macroscopic appearances and percent reduction of width obtained under the three different thermal cycling conditions for "Al". Specimens Al-2100-6 and Al-2300-6 were thermal cycled five times to peak temperatures of 2100°F and 2300°F respectively before hot ductility testing. For comparison specimen Al-2300-1 was tested on-heating to 2300°F. With five thermal cycles and subsequent testing at 2300°F peak temperature, the ductility is reduced (R.W. = 2%) and is lower than a similar sample tested with a 2100°F peak temperature (R.W. = 15%). Thermal cycling five times with a peak temperature of 2300°F reduces the hot ductility from 9% to 2%. These results indicate that exposure to both multiple thermal cycles and high peak temperatures has adverse effect on the hot ductility of the welded fully austenitic stainless steel "Al".

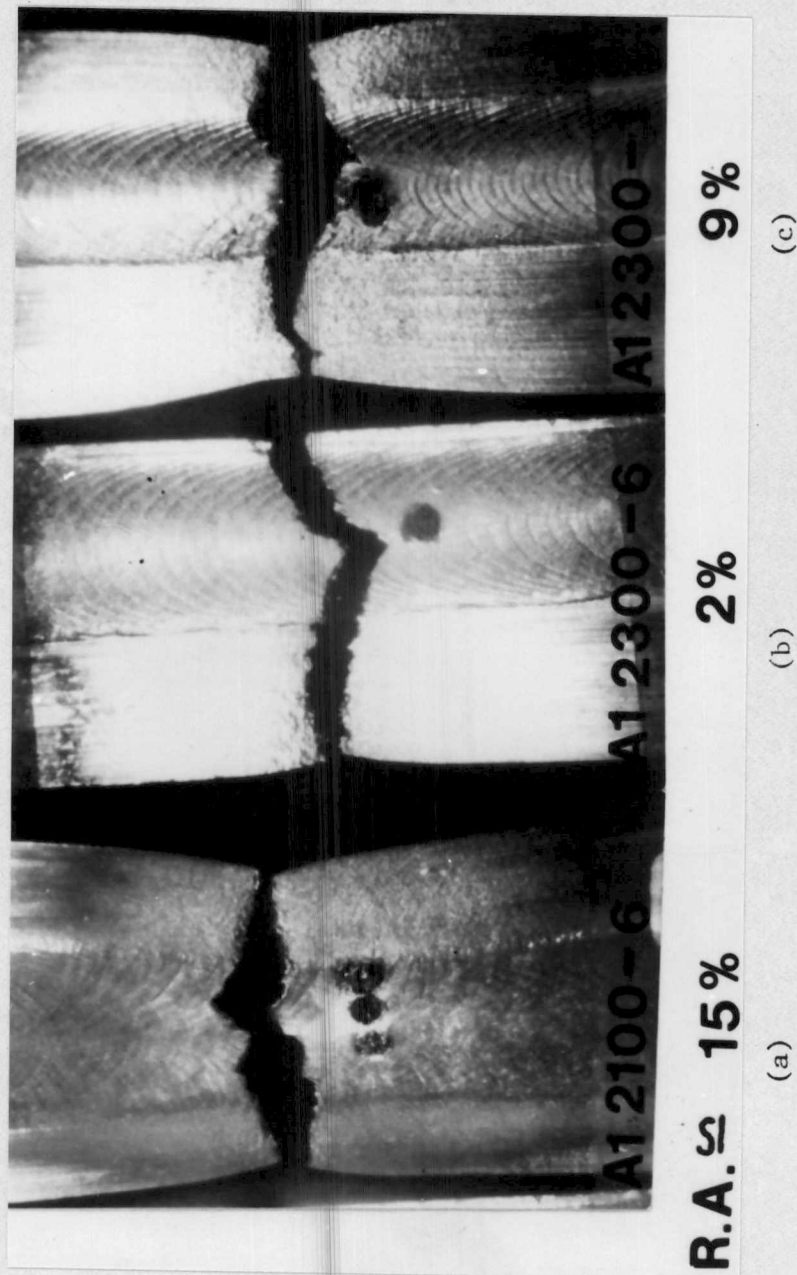


Figure 64. Hot Ductility Test Specimens of Longitudinal Composite Welded Material "Al". (3X)
 (a) Tested On-Heating to 2100°F with Five Cycles to Peak Temperature of 2100°F.
 (b) Tested On-Heating to 2300°F with Five Cycles to Peak Temperature of 2300°F.
 (c) Tested On-Heating to 2300°F.

5C. SEM Fractography and Energy Dispersive X-ray Analysis of the Hot Ductility Samples

Selected hot ductility samples were cleaned ultrasonically in acetone and examined with the SEM. The macroscopic appearance of the outer surface of a hot ductility specimen is shown in Figure 65. This specimen is from unwelded "A1" and was ruptured on-heating at 2300°F. Secondary cracks along the grain boundaries are observed in the hot zone (2300°F). The appearance of the fracture surface of the tested specimen is shown in Figure 66. Liquid film evidence is observed on the grain boundaries indicating that local melting had occurred along the grain boundaries during the heating to the peak temperature of 2300°F.

The outer surface of unwelded "A1" tested on-cooling at 2100°F after heating through the peak temperature of 2400°F is shown in Figure 67. The region E is near the fracture where the peak temperature was 2400°F. There are no secondary cracks on the outer surface in this tested condition. The groove-like appearance of the grain boundaries in the hot zone indicates that the grain boundaries were liquated and had solidified upon cooling to 2100°F (the testing temperature). Figure 68 shows the results of EDAX on the outer surface at region E. A greater level of Mo, Cr, and Mn was found in the grain boundaries than in the grain interior. Since the EDAX analysis technique is not capable of detecting the elements C, N, and O, the constituent identification can not be determined. However, by examination of binary and ternary phase diagrams of Fe, Cr, Ni, Mo, Mn, and C, a lower solidus temperature and in some cases a eutectic reaction would be expected with the increase of these alloying elements in the iron matrix. The appearance of the fracture surface of the same specimen is shown in Figure 69. This fracture displays a combination of inter-

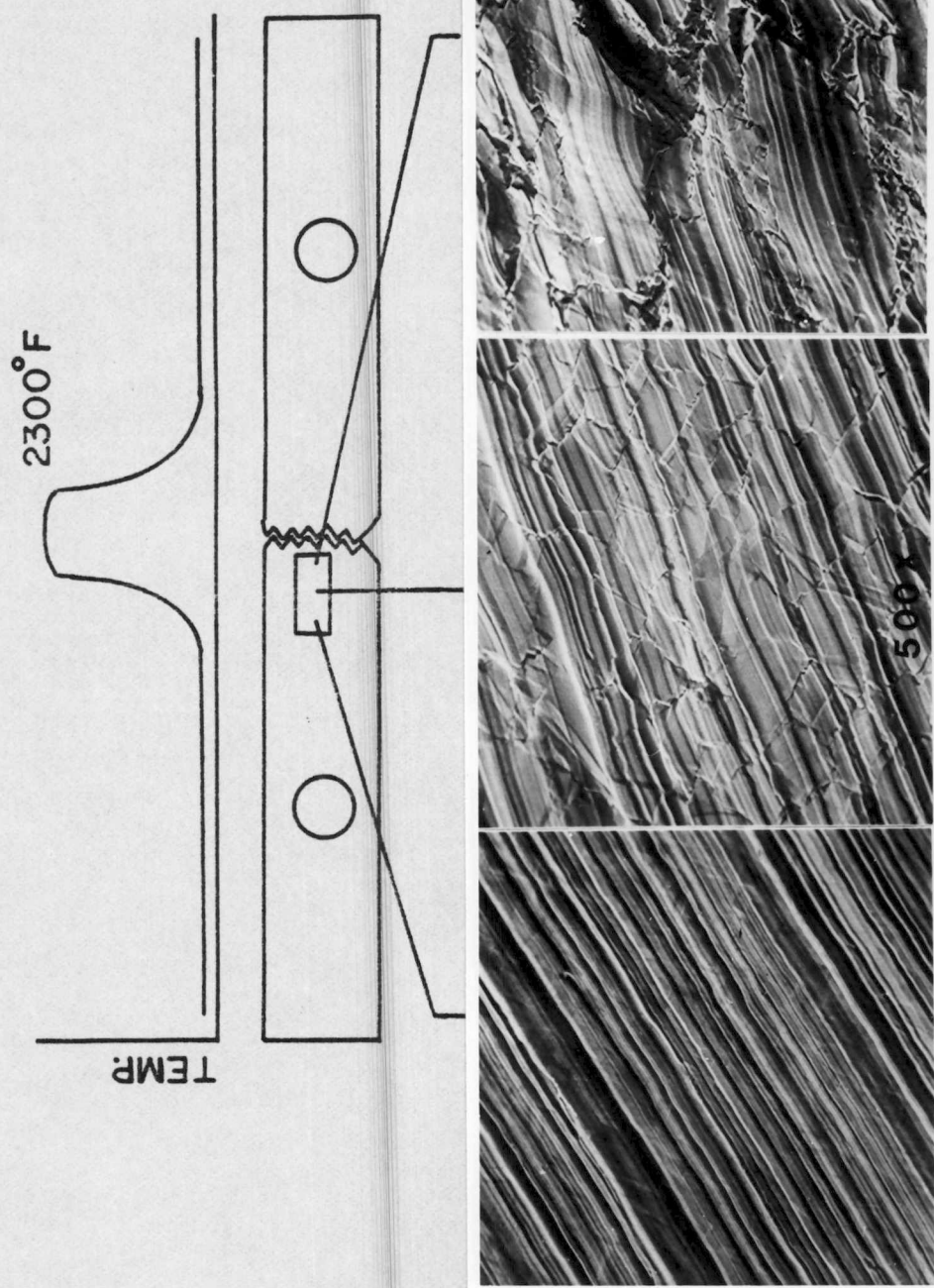
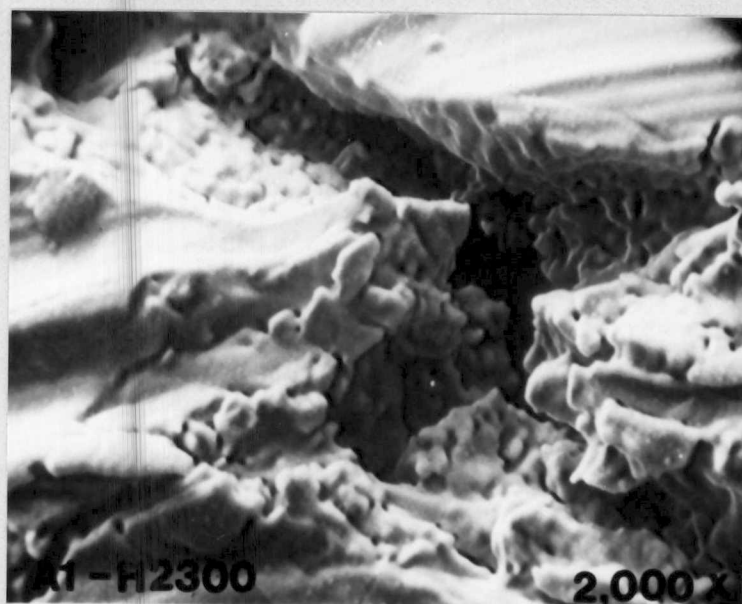


Figure 65. The Outer Surface of Hot Ductility Tested Specimen of Unwelded "Al" Tested On-Heating to 2300°F. (SEM, 500X)

(a)



(b)

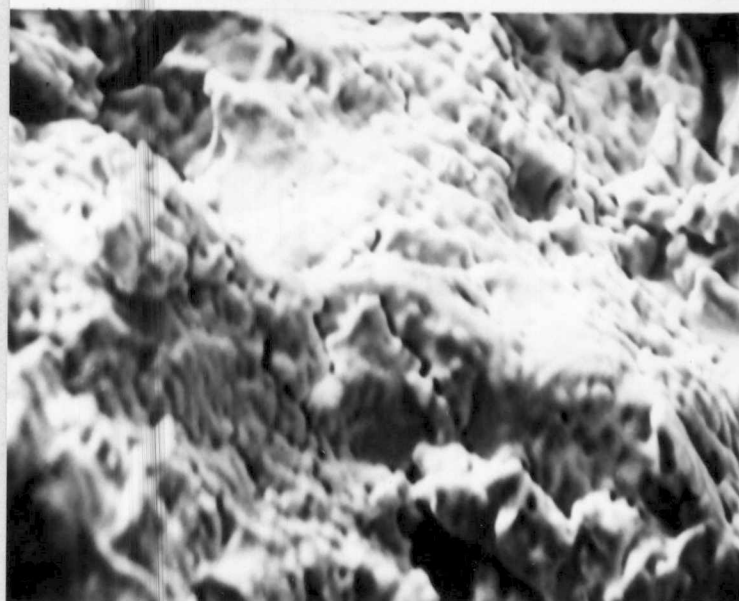


Figure 66. Hot Ductility Tested Specimen of Unwelded "A1" Tested On-Heating to 2300°F. (SEM, 2000X)
(a) Outer Surface. (b) Fracture Surface.

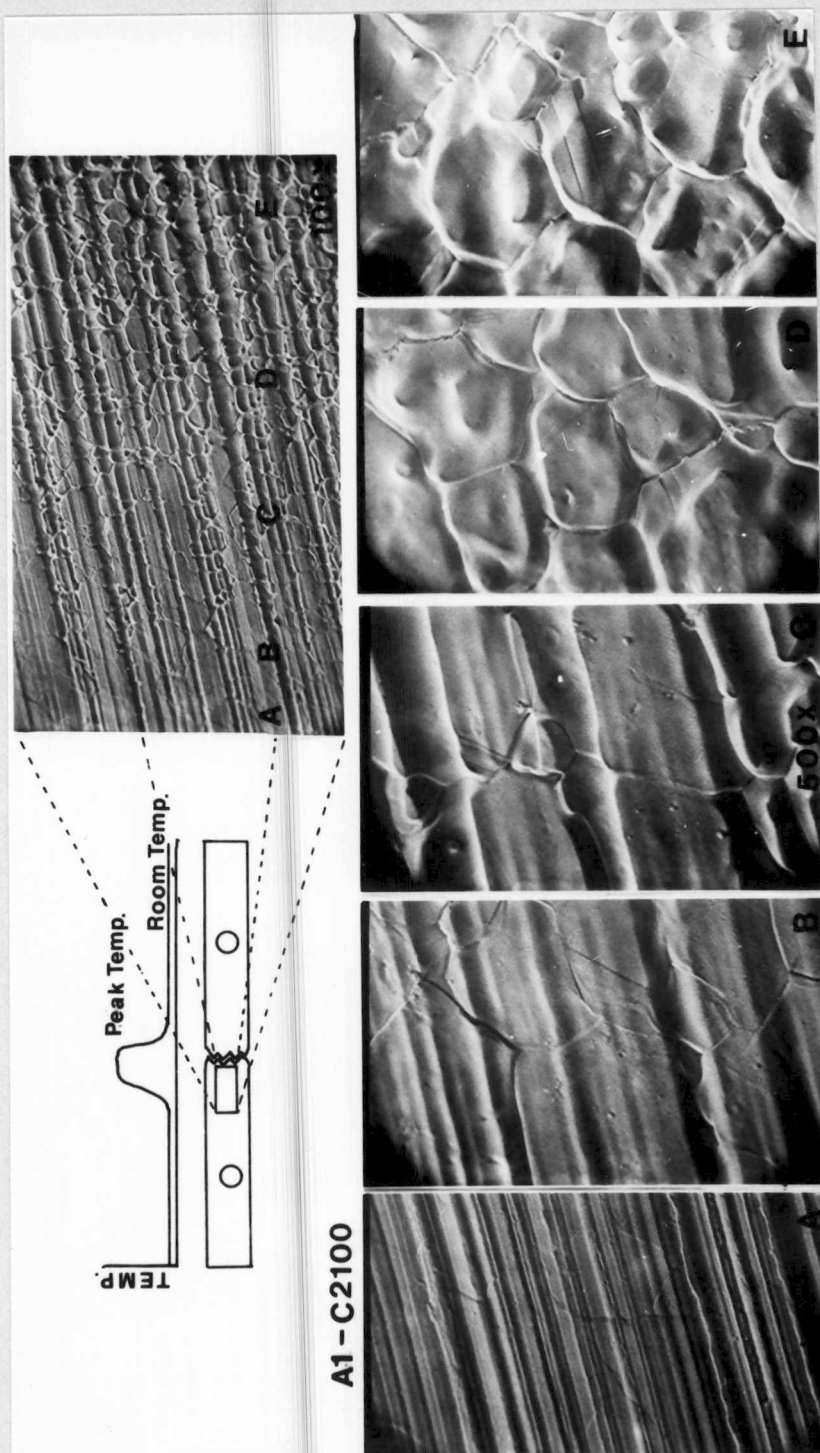


Figure 67. The Outer Surface of Unwelded "A1" Tested On-Cooling to 2100°F After Heating to 2400°F. (SEM, 500X)

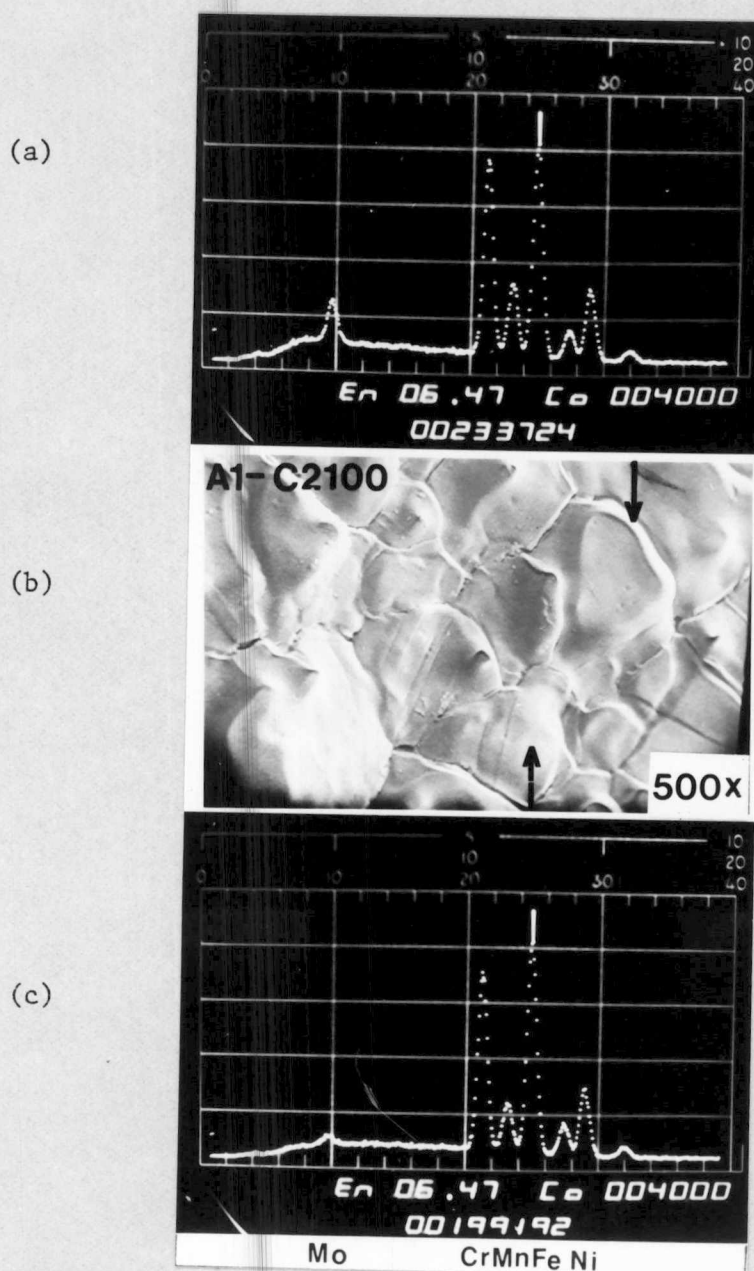


Figure 68. Outer Surface of Unwelded "A1" Tested On-Cooling to 2100°F After Heating to 2400°F.
 (a) EDAX of the Grain Boundary.
 (b) The Outer Surface. (SEM, 500X)
 (c) EDAX of the Grain Interior.

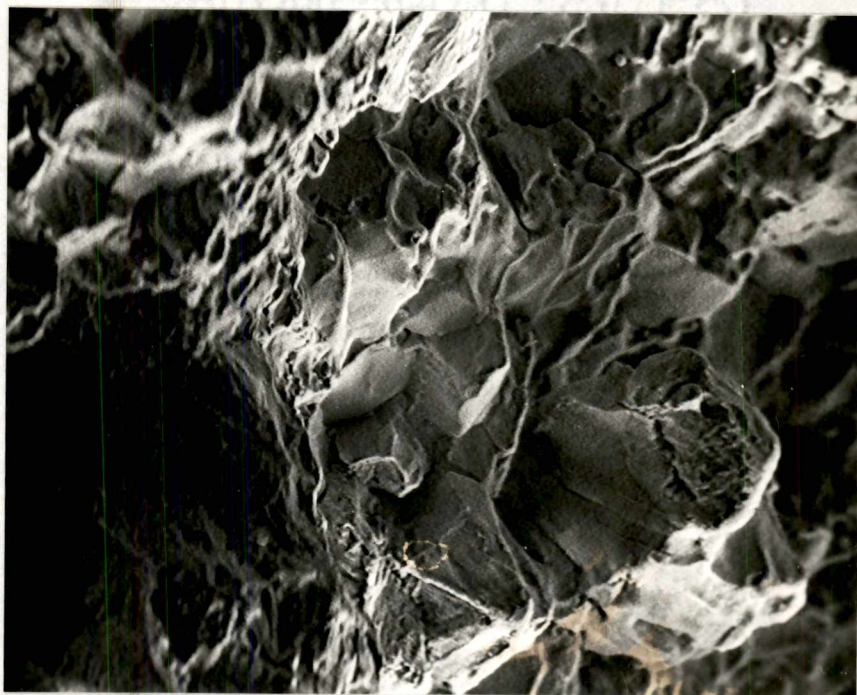


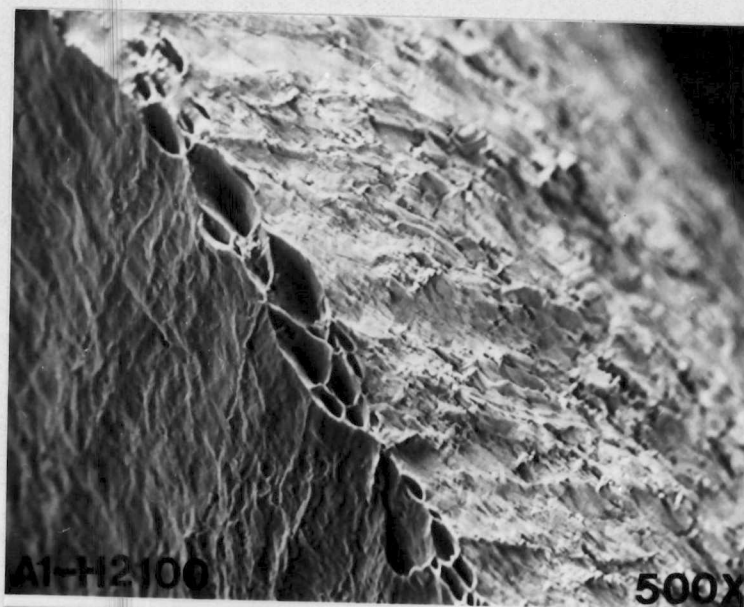
Figure 69. The Fracture Surface of Unwelded "A1" Tested On-Cooling to 2100°F After Heating to 2400°F. (SEM, 500X)
(See Figure 68 for Outer Surface Evaluation)

granular facets and dimples. This indicates that some of the previously liquated grain boundaries were embrittled at the time of rupture.

Figure 70 shows the appearances of the fracture surface of unwelded "Al" specimens tested on-heating to peak temperatures of 2100°F and 2300°F with a ductility of 80% R.W. and 13% R.W. respectively. The shear lips exhibited on the fracture surface of the specimen ruptured on heating to 2100°F give evidence of the plastic flow which occurred during rupture. At 2300°F the ductility is drastically reduced and liquid film evidence is noted on the fracture surface. By comparing Figure 70a with Figure 69, it is clearly seen that heating to 2400°F reduces the ductility measured on-cooling to 2100°F (80% R.W. on-heating at 2100°F and 21% R.W. after the 2400°F exposure). The 2400°F peak temperature exposure has a strong influence on the properties of the grain boundaries (liquated at about 2300°F).

Detailed SEM fractography and energy dispersive x-ray analysis was performed to study the cause of low ductility fractures induced by multiple thermal cycles. The longitudinal composite weld specimen of "Al" tested on heating to 2100°F peak temperature after being thermal cycled five times to a peak temperature of 2100°F is shown in Figure 71. Several secondary cracks are adjacent to the fracture. Liquation evidence was observed along the secondary crack surfaces. Evidence of liquid film formation was also present on the fracture surface in the weld metal regions, while in the base metal, ductile dimples were observed indicating that unwelded material shows greater ductility and a reduced liquation tendency than the weld metal under the same multiple thermal cycling treatment (see Figure 72). The EDAX results show that the Cr and Si content is higher on the fracture surface than that found in the bulk metal (see

(a)



(b)

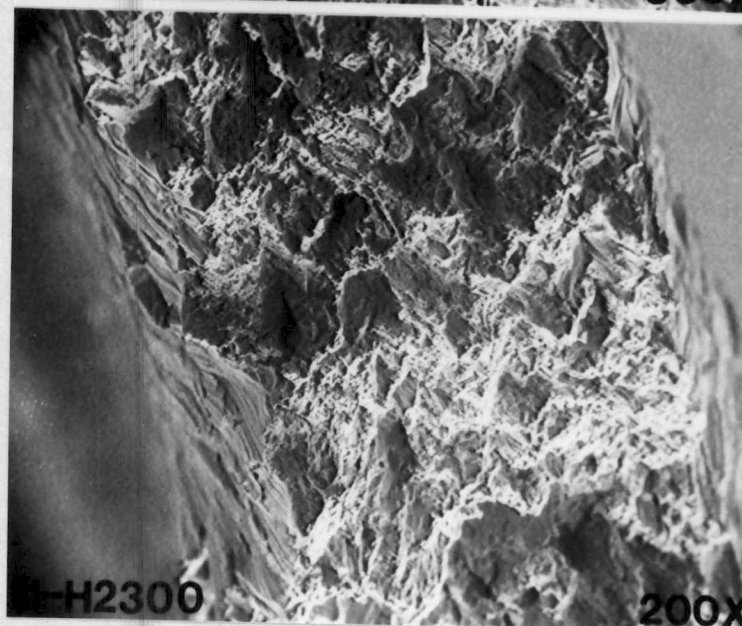
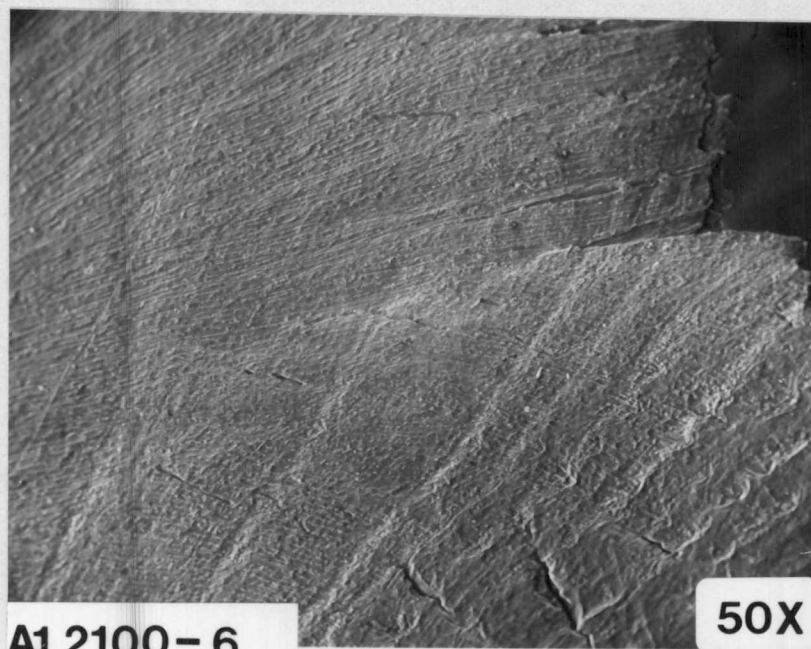


Figure 70. The Fracture Surface of Unwelded "Al" Tested On-Heating.
(a) To 2100°F. (SEM, 500X)
(b) To 2300°F. (SEM, 200X)

(a)



(b)

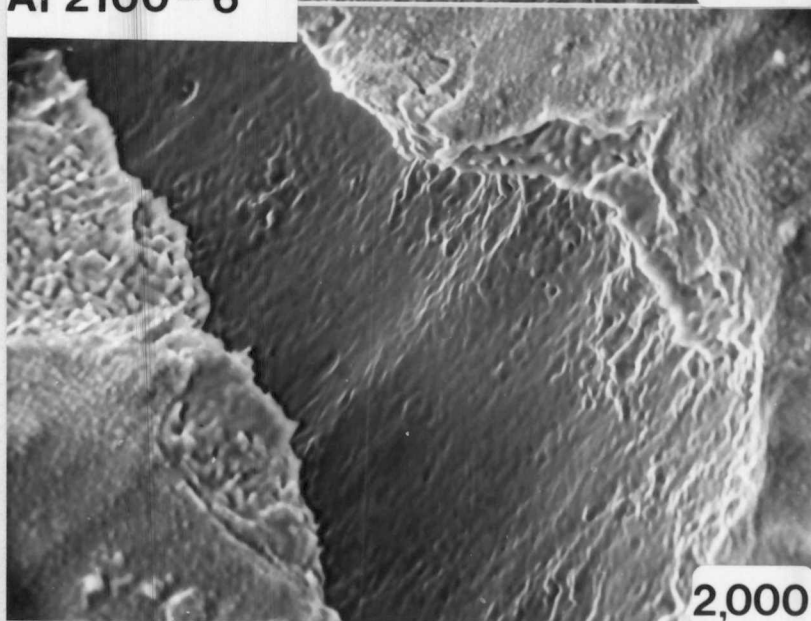


Figure 71. The Outer Surface of Weld Metal of a Longitudinal Composite Welded "A1" Tested On-Heating to 2100°F After Five Thermal Cycles with Peak Temperature of 2100°F .

(a) General Appearance. (SEM, 50X)

(b) High Magnification of a Secondary Crack. (SEM, 2000X)

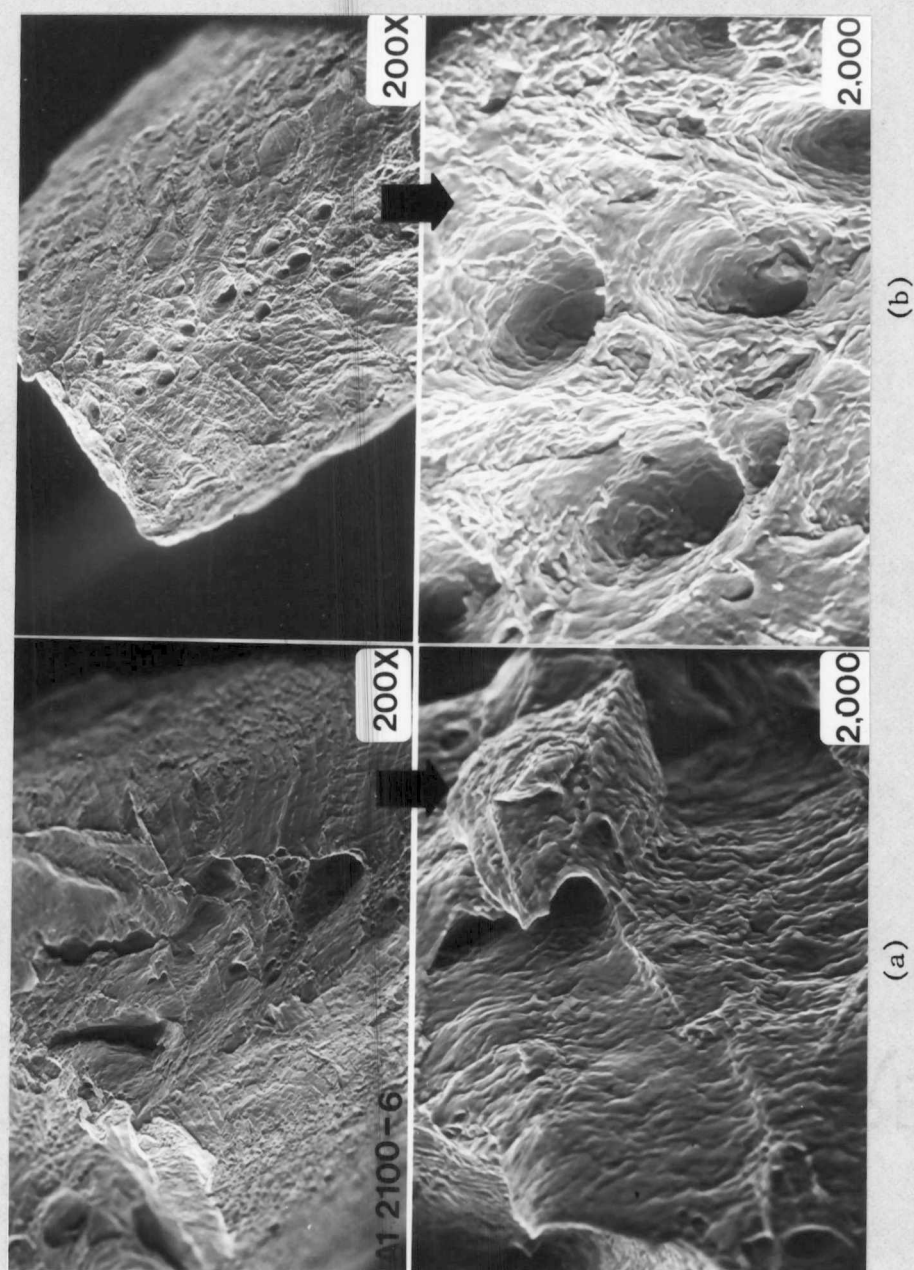


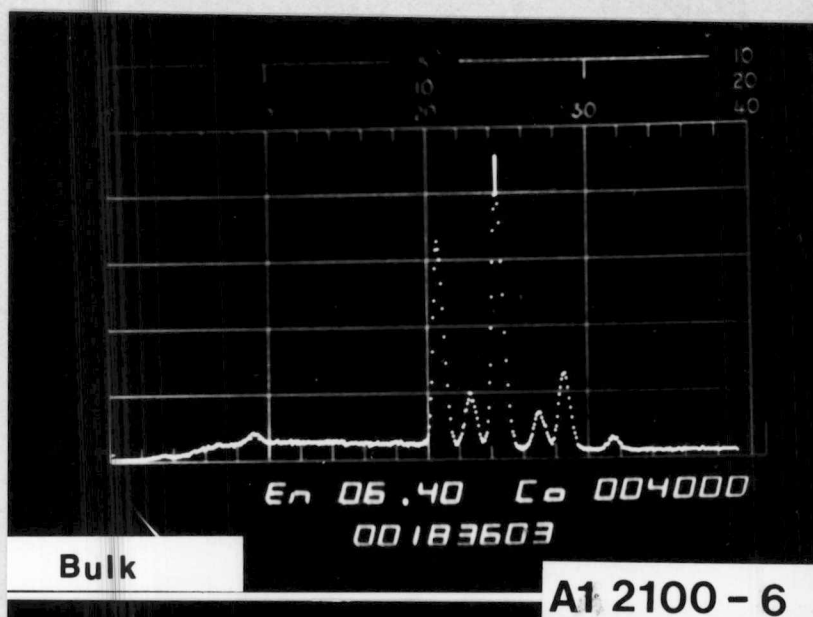
Figure 72. The Fracture Surfaces of a Longitudinal Composite Welded "Al" Tested On-Heating to 2100°F After Five Thermal Cycles with Peak Temperature of 2100°F.
 (a) Weld Metal. (SEM, 200X and 2000X)
 (b) Base Metal HAZ. (SEM, 200X and 2000X)

Figure 73). It is to be noted that using a longitudinal composite weld specimen, information on both the weld metal and the base metal can be obtained simultaneously.

The appearance of the outer surface of the longitudinal composite weld specimen of "Al" tested on-heating to 2300°F is shown in Figure 74. It is apparent that the grains in the fusion zone grow directly from the unmelted base metal. Note that the secondary cracks in the weld metal propagated along the grain boundaries and terminated at the fusion boundary. No cracks were observed in the base metal HAZ. This indicates that the base metal HAZ is more ductile than the weld metal. The cause of the low ductility in the weld metal can be explained by the liquid film formation evident on the grain boundaries (see Figure 75). Higher Cr and Si were found in the liquid film regions as shown in Figure 75. The fracture appearance of the specimen is shown in Figure 76. Secondary cracks along the grain boundaries are clearly seen. Fine dimples on some grain faces indicate that some ductile rupturing occurred during the testing.

With five thermal cycles exposures to a peak temperature of 2300°F, the hot ductility of the longitudinal composite type sample of "Al" was reduced. The appearance of the outer surface and the fracture surface of the specimen is shown in Figure 77. Brittle intergranular fracture features can be seen in both the weld and the base metal HAZ. The EDAX of the fracture surface in the weld metal region again shows that a greater amount of Cr and Si is evident on the fracture surface as compared to the bulk material. The amount (peak height response) of these segregated elements was about the same as that found on the fracture surface of specimens tested without multiple thermal cycles.

(a)



(b)

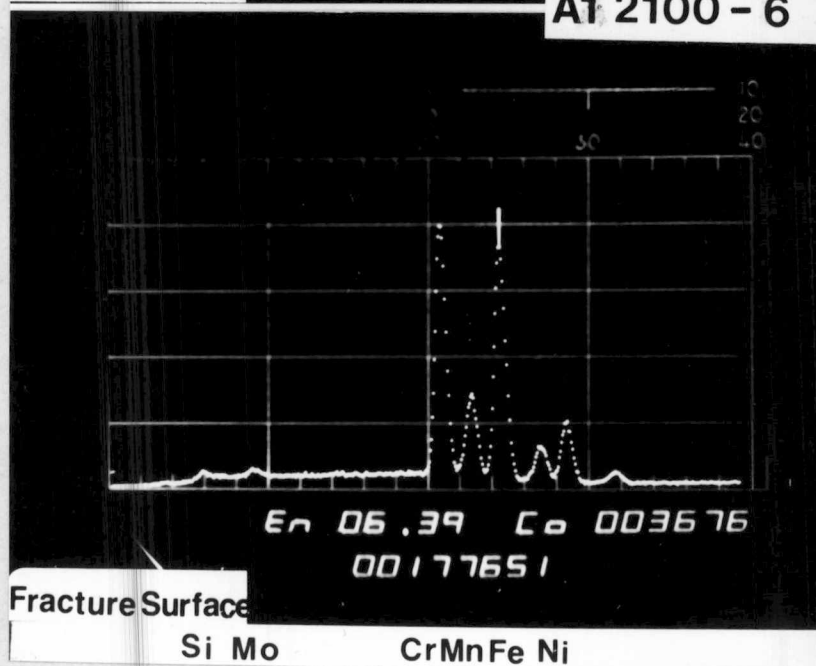


Figure 73. EDAX of a Longitudinal Composit Welded "Al" Tested On-Heating to 2100°F After Five Thermal Cycles with Peak Temperature of 2100°F.
 (a) EDAX of the Bulk of the Specimen.
 (b) EDAX of the Fracture Surface.

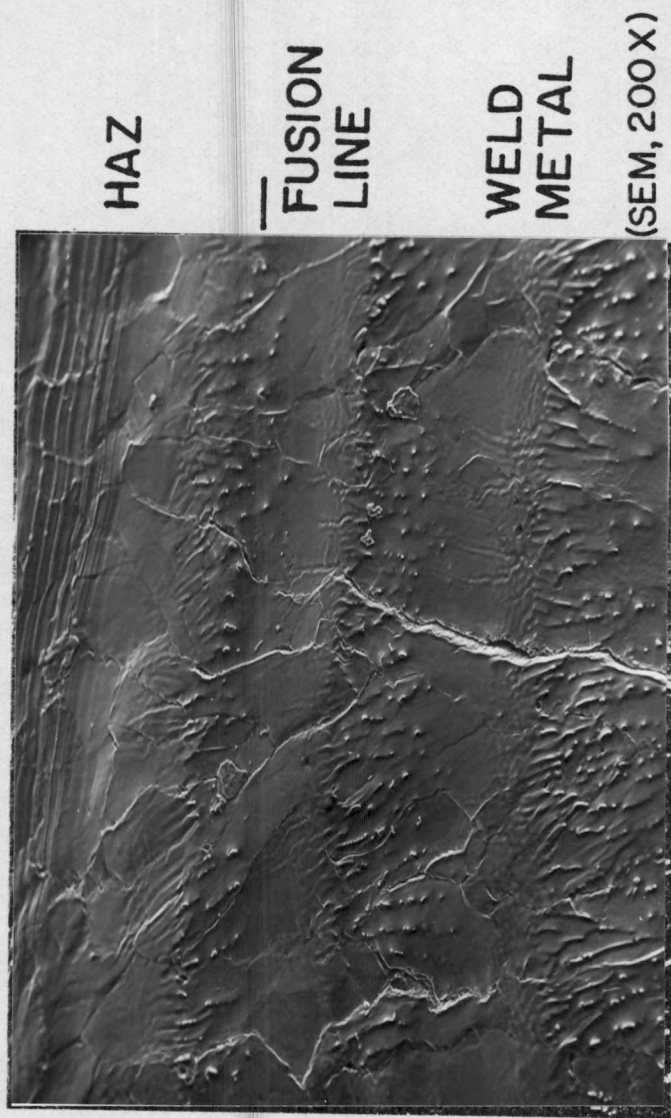


Figure 74. The Outer Surface of the Longitudinal Composite Weld "A1" Tested On-Heating to 2300°F.
(SEM, 2000X)

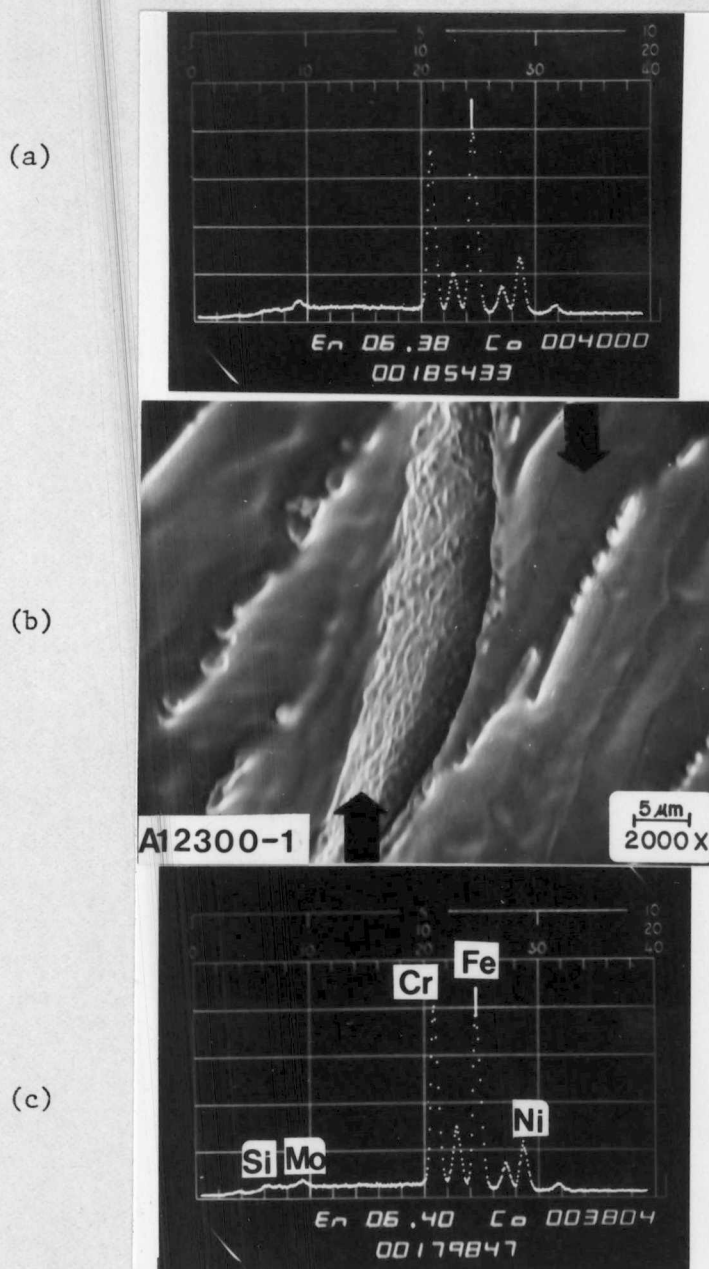


Figure 75. Longitudinal Welded Composite "A1" Tested On-Heating to 2300°F.

- (a) EDAX of Grain Interior.
- (b) Secondary Crack. (SEM, 2000X)
- (c) EDAX of an Secondary Crack surface.

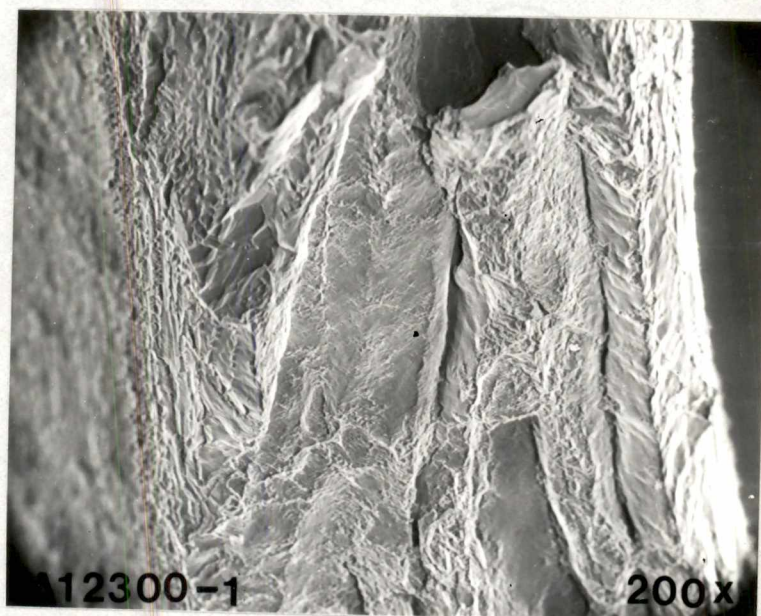


Figure 76. The Fracture Surface of a Longitudinal Composite Welded Specimen "A1" Tested On-Heating to 2300°F. (SEM, 200X)

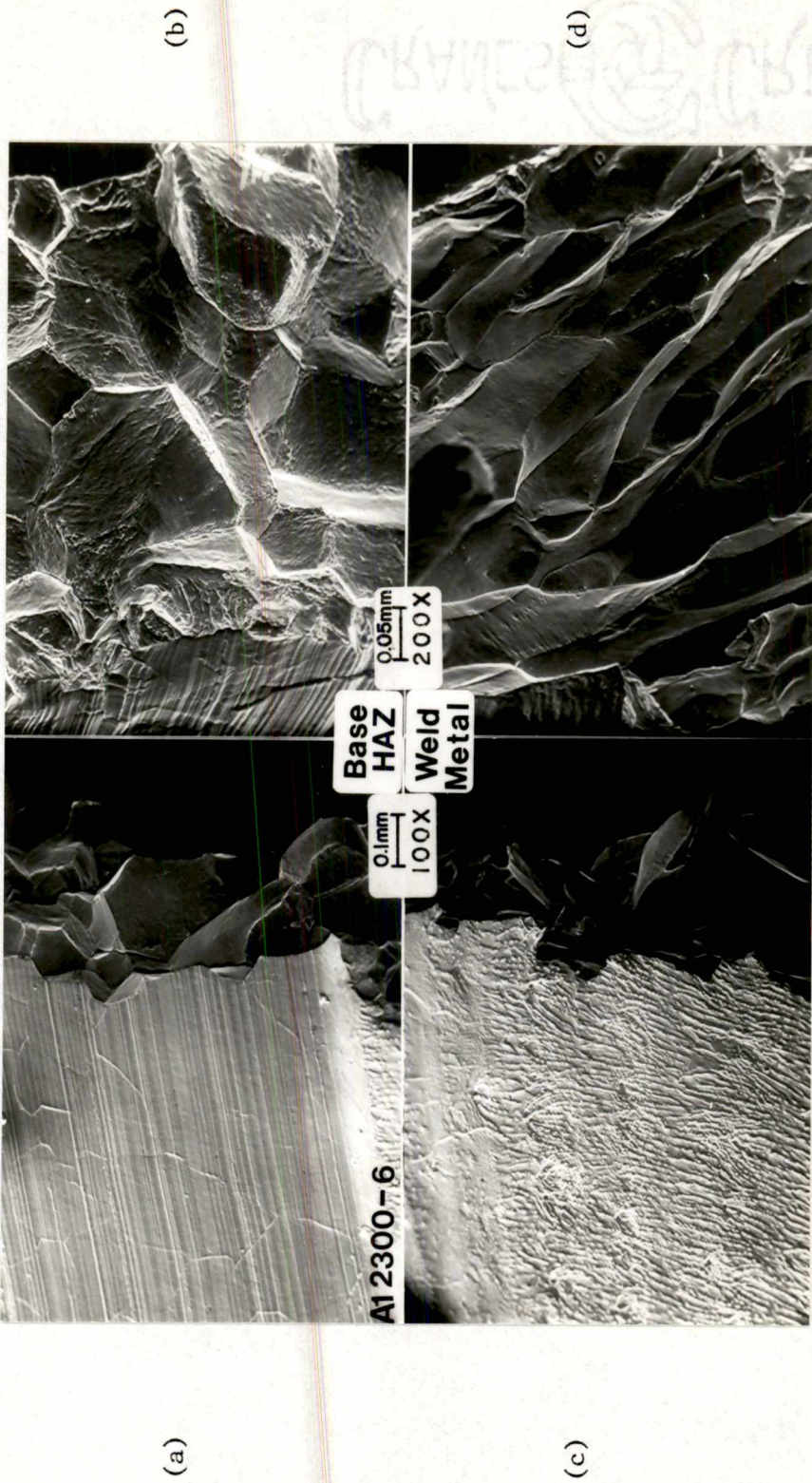


Figure 77. A Longitudinal Composite Welded Specimen "Al" Tested On-Heating to 2300°F After Five Thermal Cycles with Peak Temperature of 2300°F.
(a) The Outer Surface of Base Metal HAZ. (SEM, 100X)
(b) The Fracture Surface of Base Metal HAZ. (SEM, 200X)
(c) The Outer Surface of Weld Metal. (SEM, 100X)
(d) The Fracture Surface of Weld Metal. (SEM, 200X)

Comparing Figure 72 and Figure 77, it was found that the ductility decreased when the specimen was tested at higher peak temperatures after multiple thermal cycling. The intergranular fracture appearance shows brittle behavior. With multiple thermal cycling to a peak temperature greater than 2100°F , the hot ductility was significantly reduced. This can be seen by comparing Figure 76 and Figure 77. In the previous sections, it was reported that multiple thermal cycling reduced the hot cracking resistance of fully austenitic stainless weld metal. The data show a very good correlation between the hot ductility behavior and the hot cracking sensitivity. The number of hot cracks in the weld metal which experienced multiple thermal cycles increased because the embrittlement of the grain boundaries was enhanced by the multiple thermal cycles.

The fracture surface of a specimen of material "H" fractured at a peak temperature of 2300°F after five prior thermal cycles is shown in Figure 78. A ductile surface morphology was observed. No liquid film evidence was found on the fracture surface. The retention of ductility after multiple thermal cycling to a high temperature is probable related to a fact that a greater amount of ferrite still remained at high temperatures. The ferrite can absorb more of the harmful elements (Si, P, and S) than does austenite. Therefore, the formation of liquid films and grain boundary degradation was decreased. The effect of peak temperature and the number of thermal cycles on the ferrite content in the austenitic stainless steel weld metals is discussed in the next section.

5D. Effect of Multiple Thermal Cycles on Ferrite Content

It was found that the hot ductility is higher in the ferrite-containing weld metals than in the fully austenitic weld metals.

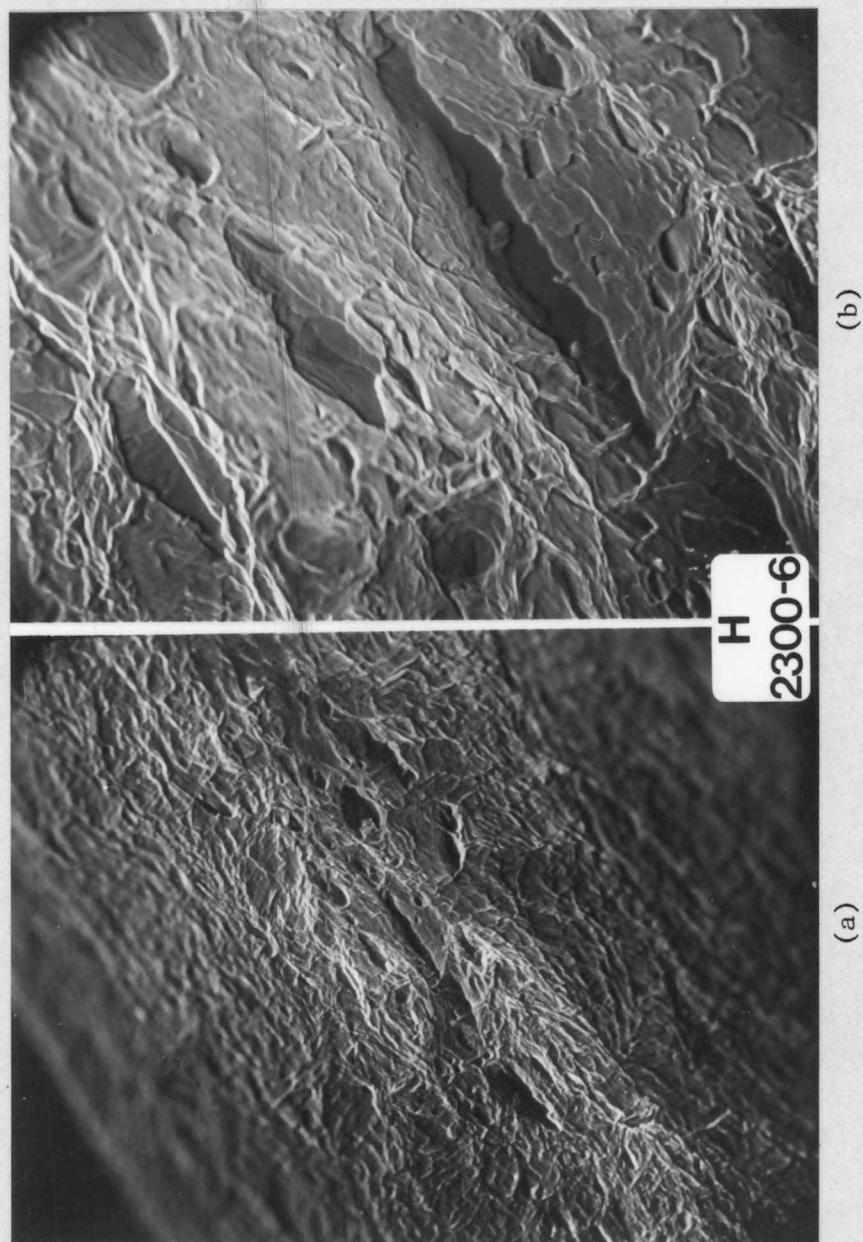


Figure 78. The Fracture Surface of Material "H" Fractured at a Peak Temperature of 2300°F After Five Thermal Cycles to 2300°F. (SEM)
(a) At 200X. (b) At 1000X.

It was also found that in the ferrite-containing welds there is no difference in hot ductility at peak temperatures of 2100°F and 2300°F . The hot ductility decreased slightly as the number of thermal cycle exposures increased. It appears that the hot ductility behavior at high temperatures is related to the amount and morphology of ferrite in the welds. An easy, fast, and reproducible method to determine the effect of temperature-time cycle on the ferrite content of the weld metals was developed. With the Gleeble device, the effect of peak temperature, cooling rate and the number of thermal cycles on the ferrite content was determined. The longitudinal composite weld specimen was used. The test specimens were subjected to pre-programmed weld thermal cycles (multiple cycles at different peak temperatures). After cooling to room temperature, the test specimens were removed and Ferrite Number determined.

Two ferrite-containing austenitic stainless steels "L" and "H" were used in this portion of the study. The Ferrite Numbers of "L" and "H" measured in the as-welded condition are 5 and 25 respectively.

The FN distribution along the weld in sample "H" (FN = 25) which was thermal cycled five times to a peak temperature 2100°F is shown in Figure 79. The FN decreased to 12 from the initial value of 25 in the central portion. There is a ferrite gradient along the longitudinal direction of the specimen because a temperature gradient exists. With a peak temperature of 2400°F , an unusual FN distribution curve was observed as shown in Figure 80. The FN in the central portion remains, while at about $1/8$ " from the center an appreciable decrease in FN was observed. The trough in the FN distribution curve corresponds to the location which was exposed to the temperature of about 2000°F . This

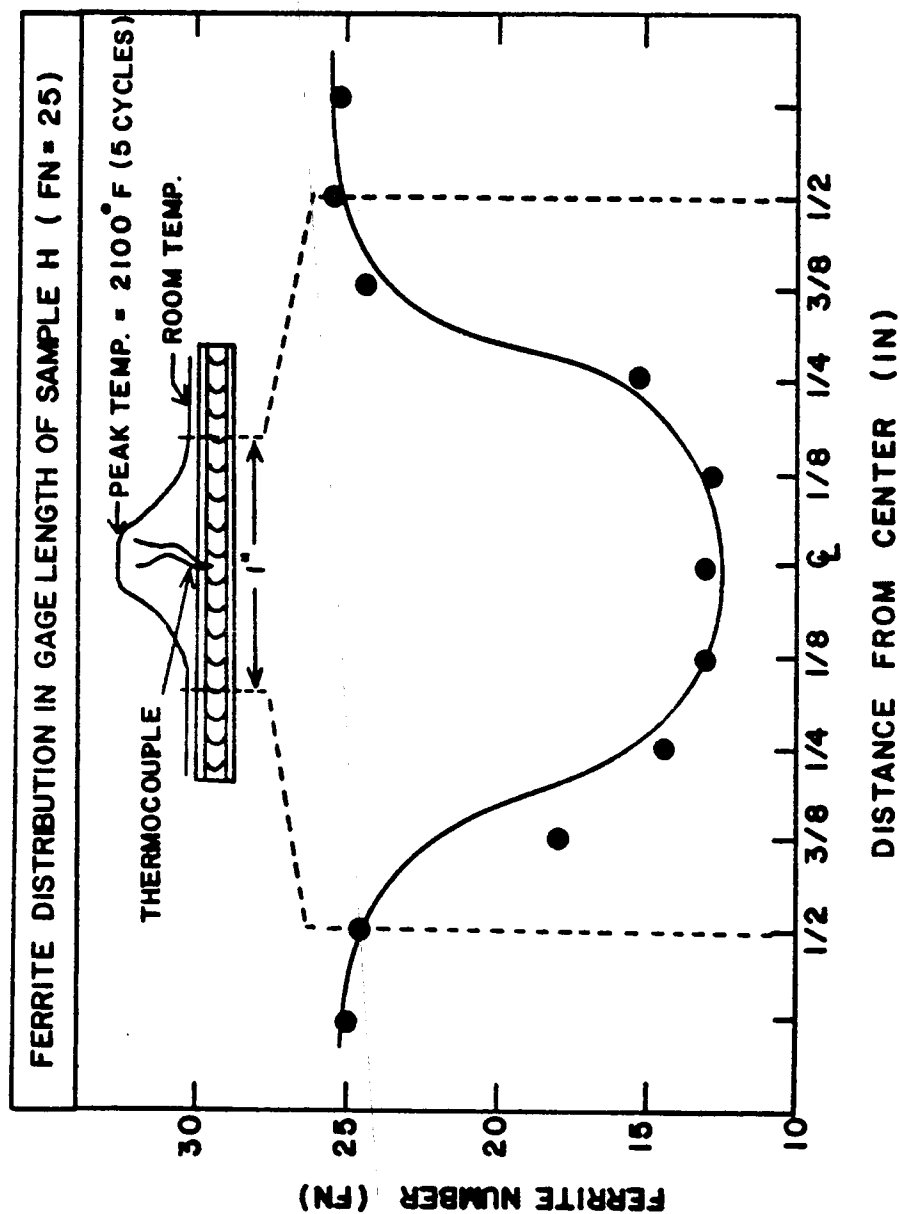


Figure 79. Ferrite Number Distribution Along the Gage Length of Sample "H" After Five Thermal Cycles to a Peak Temperature of 2100°F.

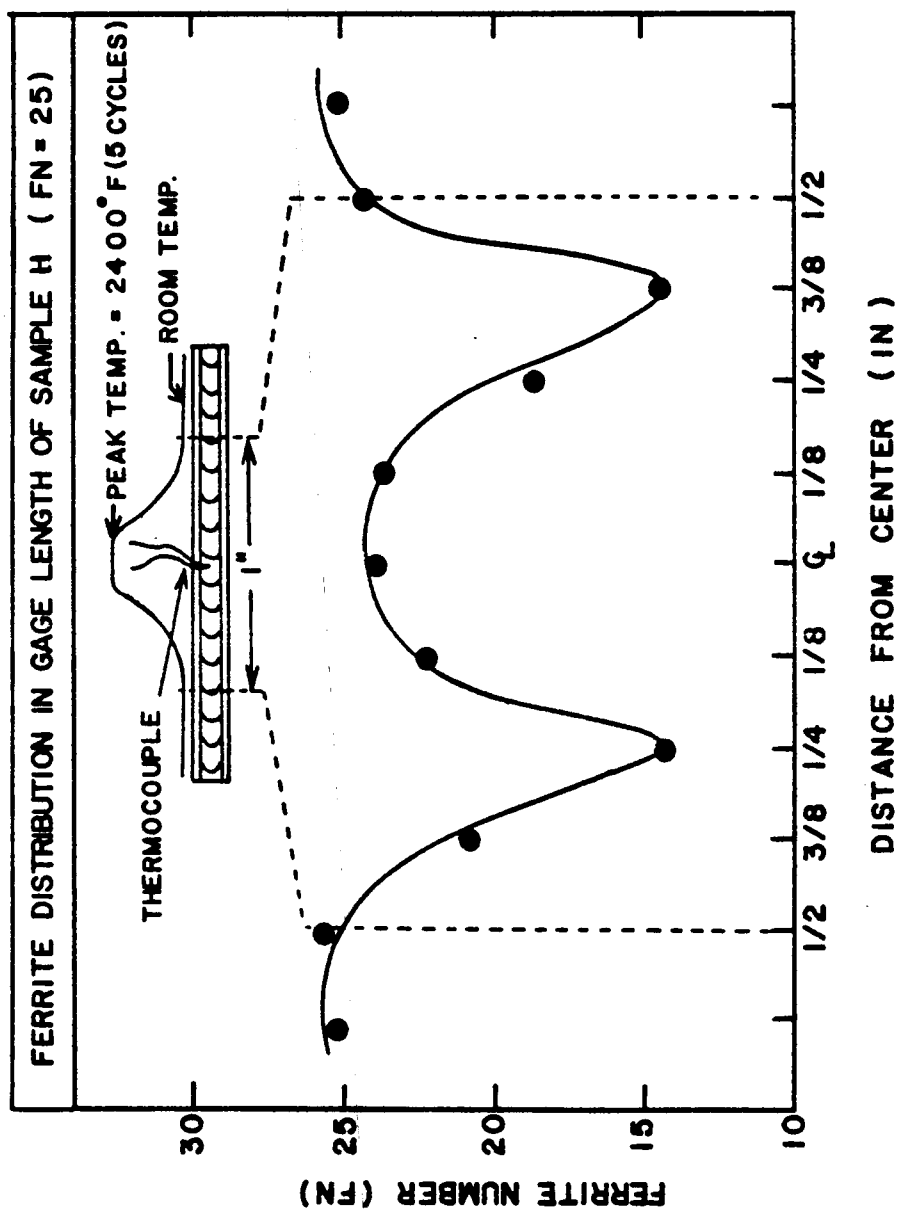


Figure 80. Ferrite Number Distribution Along the Gage Length of Sample "H" After Five Thermal Cycles to a Peak Temperature of 2400 F.

phenomenon is more clearly revealed if the FN vs. peak temperature is plotted as shown in Figure 81. The temperature at which the maximum decrement of FN occurs is approximately 2000°F. The Ferrite Number decrement becomes smaller as the peak temperature increases above 2000°F. The same behavior is also observed for the sample "L" (FN = 5) as shown in Figure 82. The temperature at which the maximum decrement of FN occurs is about 2200°F. The microstructural evidence is clearly seen in Figure 83. Specimens of material "L" were thermal cycled five times to different peak temperatures. The ferrite content after 1800°F and 2400°F exposure is higher than that after 2100°F exposure.

The above observations can be partially explained from the Iron-Chromium-Nickel pseudo binary phase diagrams (70% Iron and 65% Iron) as shown in Figure 84 (11). It is to be noted however that the presence of other alloying elements, and rapid cooling of the weld metal, the solidification behavior and sequences of phase separation in austenitic stainless steel weld metal depart considerably from that predicted by the phase diagram. The compositions of materials "L" and "H" are indicated on the phase diagrams in Figure 84. From these diagrams, an approximate solution treating temperature can be estimated. For example, at 2200°F the weld metal "L" is in the single phase region while "H" is in the two phase region ($\gamma + \delta$). Thus, 2200°F is selected as the solution treating temperature for weld metal "L". The optimum solution treating temperature for "H" is predicted to be approximately 2000°F.

The effect of number of thermal cycles on the FN of welded sample of material "H" thermal cycled to peak temperature of 2300°F is shown in Figure 85. The FN vs. the number of thermal cycles is plotted for two

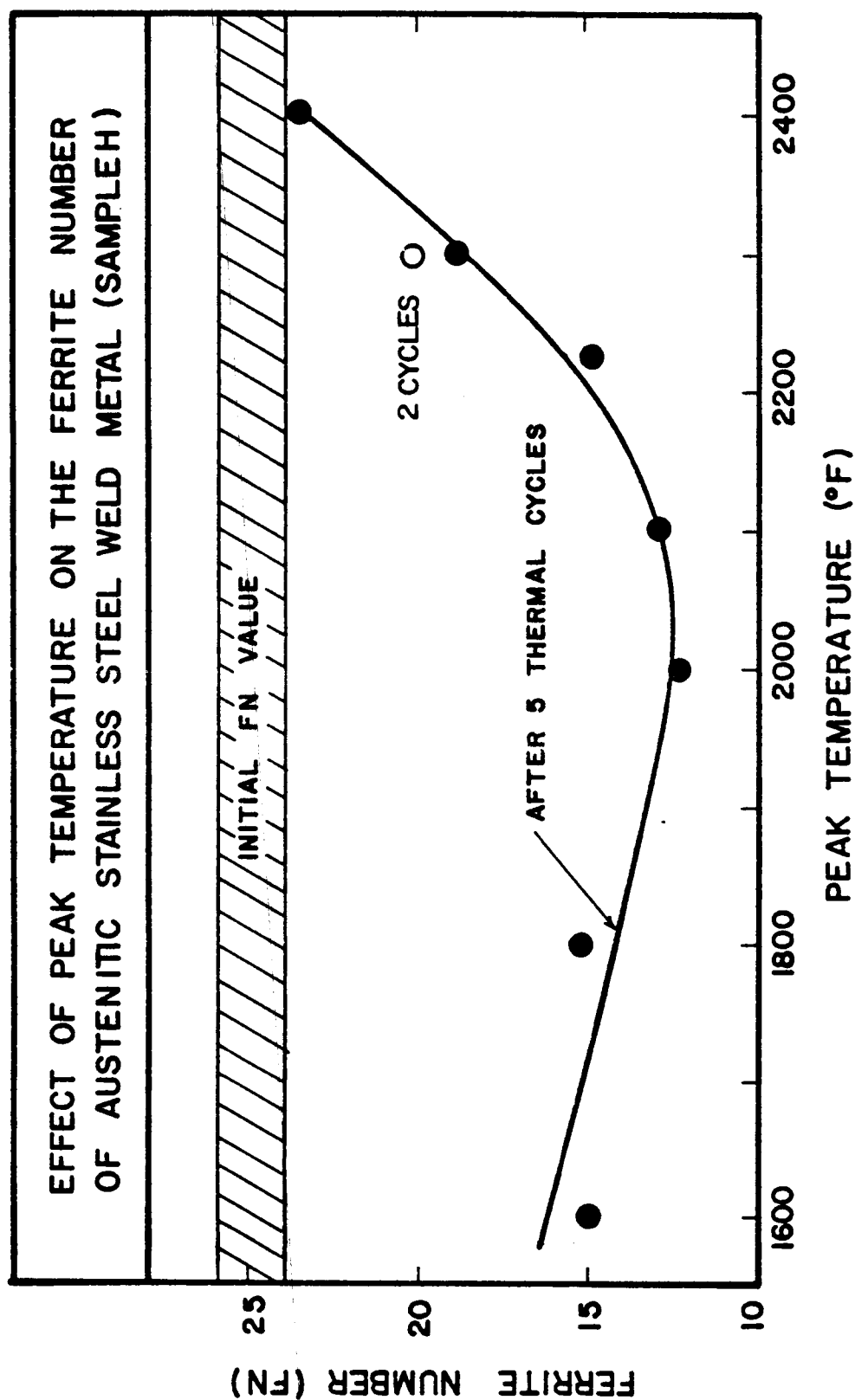


Figure 81. Effect of Peak Temperature on the Ferrite Number of Austenitic Stainless Steel Weld Metal "H".

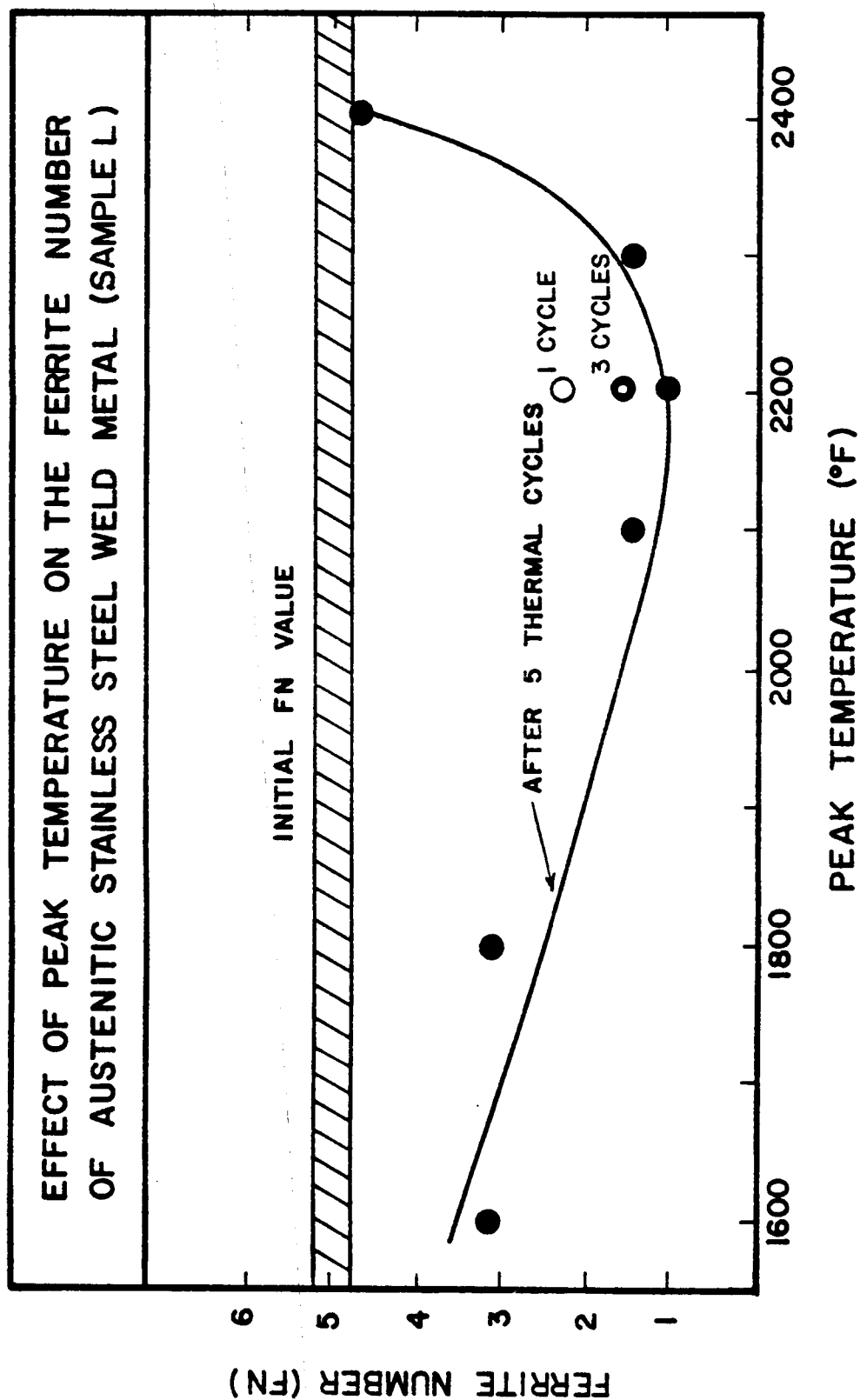


Figure 82. Effect of Peak Temperature on the Ferrite Number of Austenitic Stainless Steel Weld Metal "L".

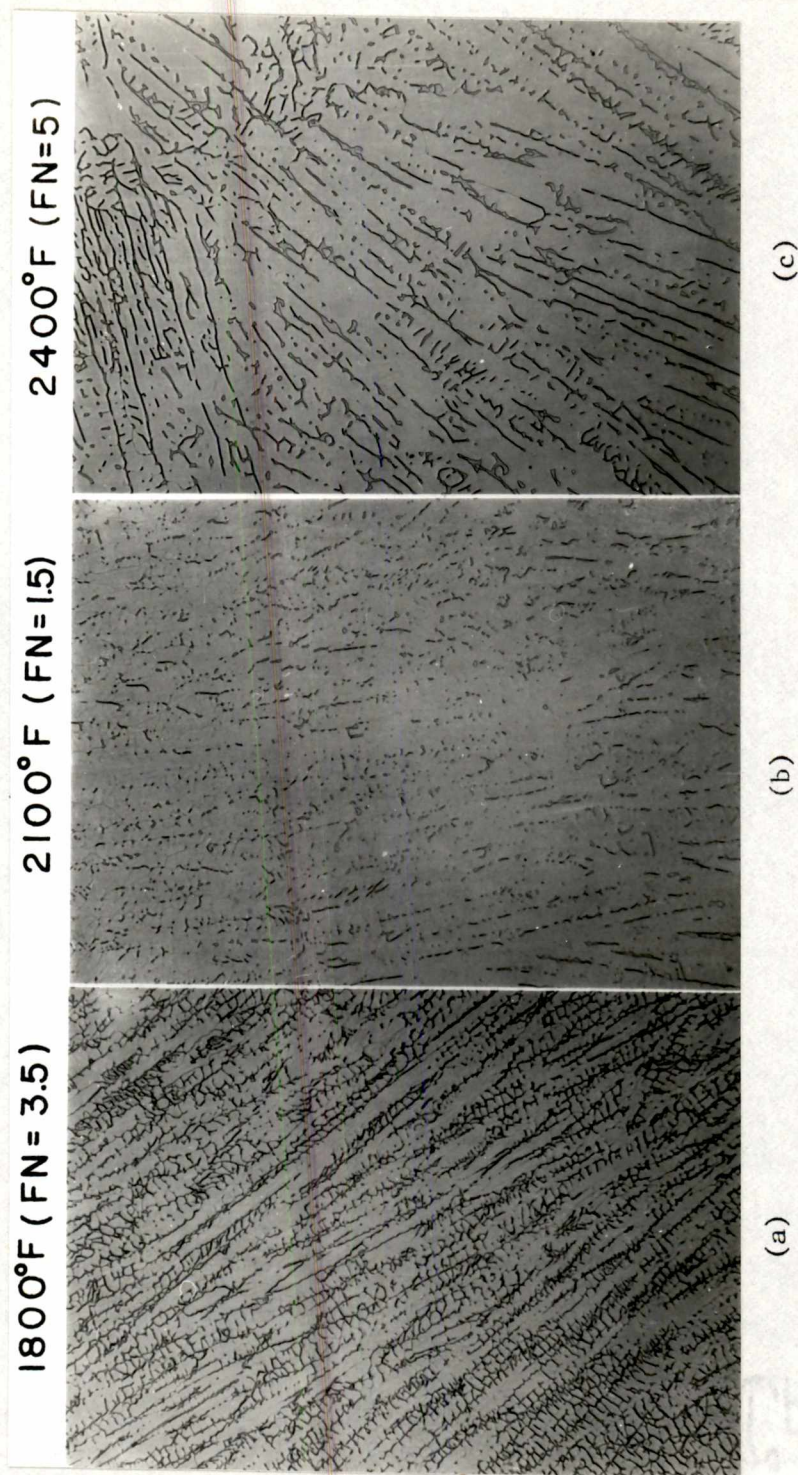


Figure 83. Microstructures of Specimen "L" Thermal Cycled Five Times to a Peak Temperature of; (a) 1800°F, (b) 2100°F, and (c) 2400°F. (OLM, 200X)

FE-CR-NI PSEUDO BINARY PHASE DIAGRAMS

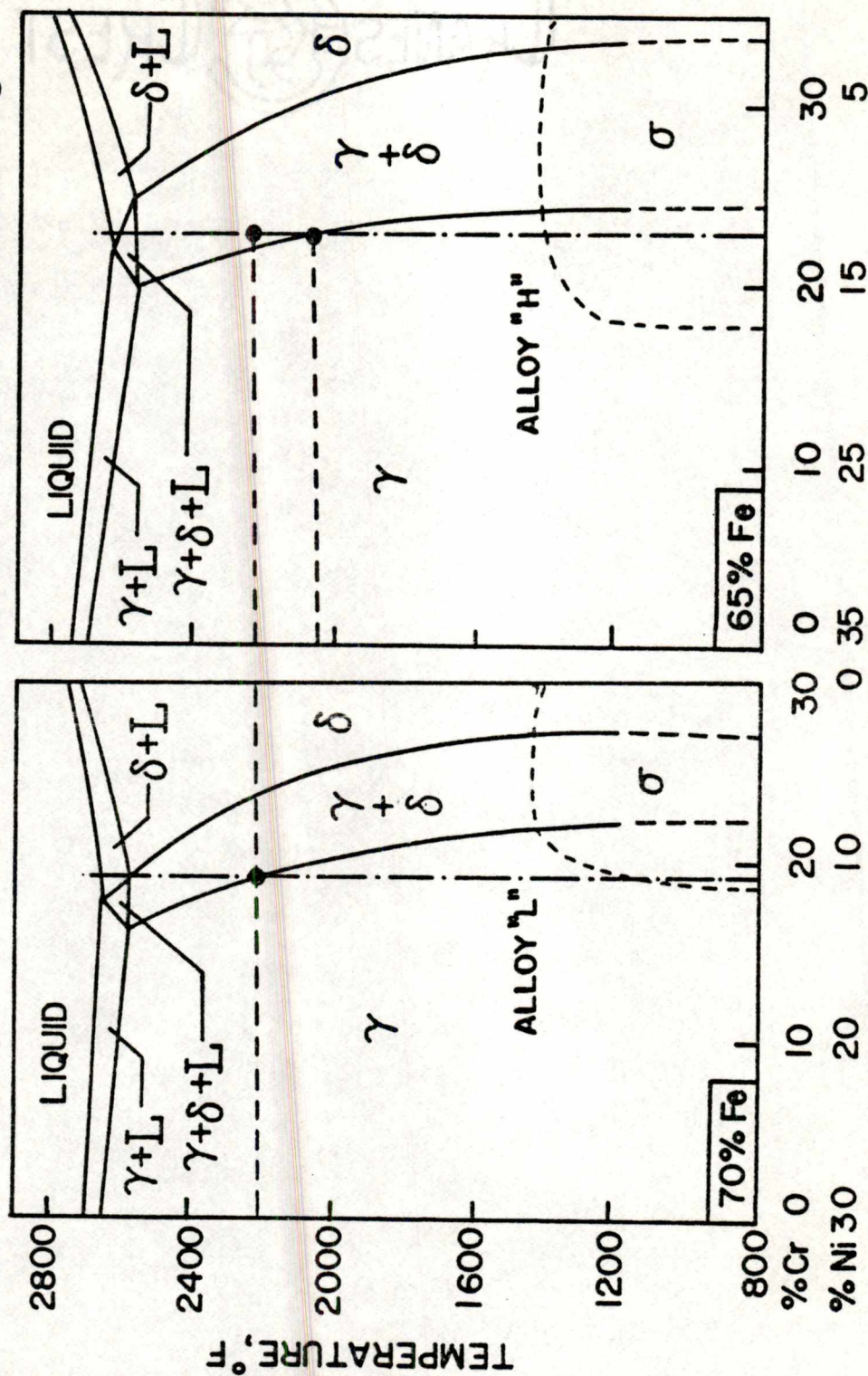


Figure 84. Fe-Cr-Ni Pseudo-Binary Phase Diagrams for 70% Fe and 65% Fe.

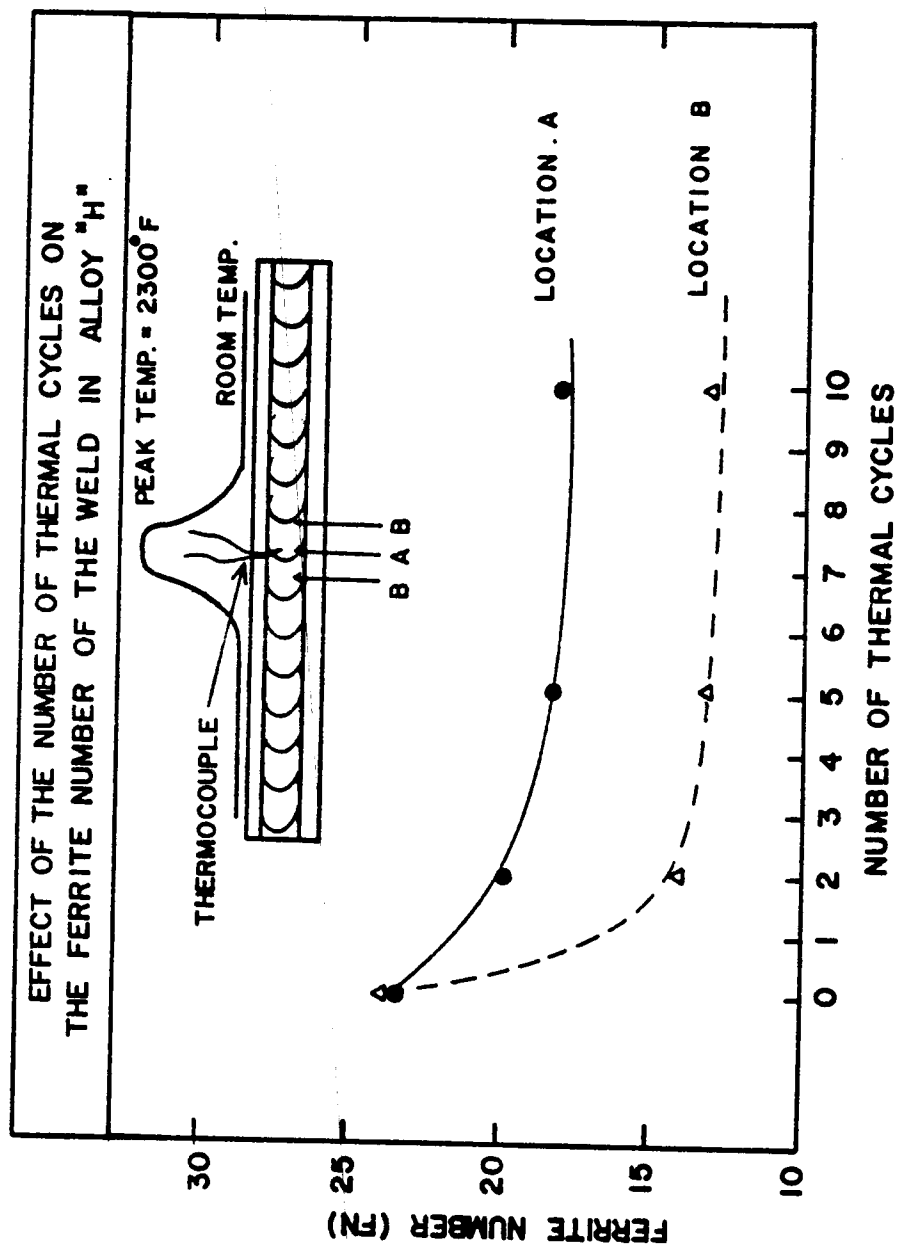


Figure 85. Effect of the Number of Thermal Cycles on the Ferrite Number of the Weld in Alloy "H".

regions in the specimen. Region A is in the center hot zone which was exposed to the peak temperature of 2300°F . Region B is in the area adjacent to the hot zone and was exposed to a lower temperature (approximate 2000°F). It is clearly seen that the FN decreased rapidly after two cycle exposures and no appreciable change in the FN occurs as the number of thermal cycles is increased above five. It is also observed that the FN decreased to a substantially low value in region B while the FN was slightly decreased in region A after the thermal cycle exposure. The microstructural evidence can be clearly seen in Figure 86. Welded material "H" was thermal cycled five times to 2300°F . In the center hot zone (region A in Figure 85), the FN decreased to 18 (from 25 FN in the as-welded condition) after exposure five cycles to 2300°F . In the region adjacent to the hot zone (region B in Figure 85), the FN decreased to 14 after exposure five times to approximate 2100°F .

An interesting result was obtained from the multiple thermal cycle history of welded material "L". Figure 87 shows the measured FN as a function of the number of thermal cycles. In the center hot zone, the FN decreased from 5 to 1 after exposure of ten cycles to 2200°F but increased again to 4 FN after an additional five cycles to 2400°F . In region B, which corresponds to an area exposed to a lower temperature (not above 2200°F), the FN decreased to near 1 FN after exposure to fifteen thermal cycles. Microstructural evidence can be clearly seen in Figure 88. The FN of the as-deposited weld is 5. After exposure to 10 cycles to 2100°F and five cycles to 2000°F , a large fraction of the ferrite is transformed to austenite as shown in Figure 88b. When five additional cycles to a peak temperature of 2400°F were superimposed on the center



Figure 86. Microstructures of Alloy "H" Thermal Cycled Five Times to a Peak Temperature of 2300°F . (OLM, 200X)
(a) As-Welded. (b) Low Temperature Region (2100°F). (c) Central Region (2300°F).

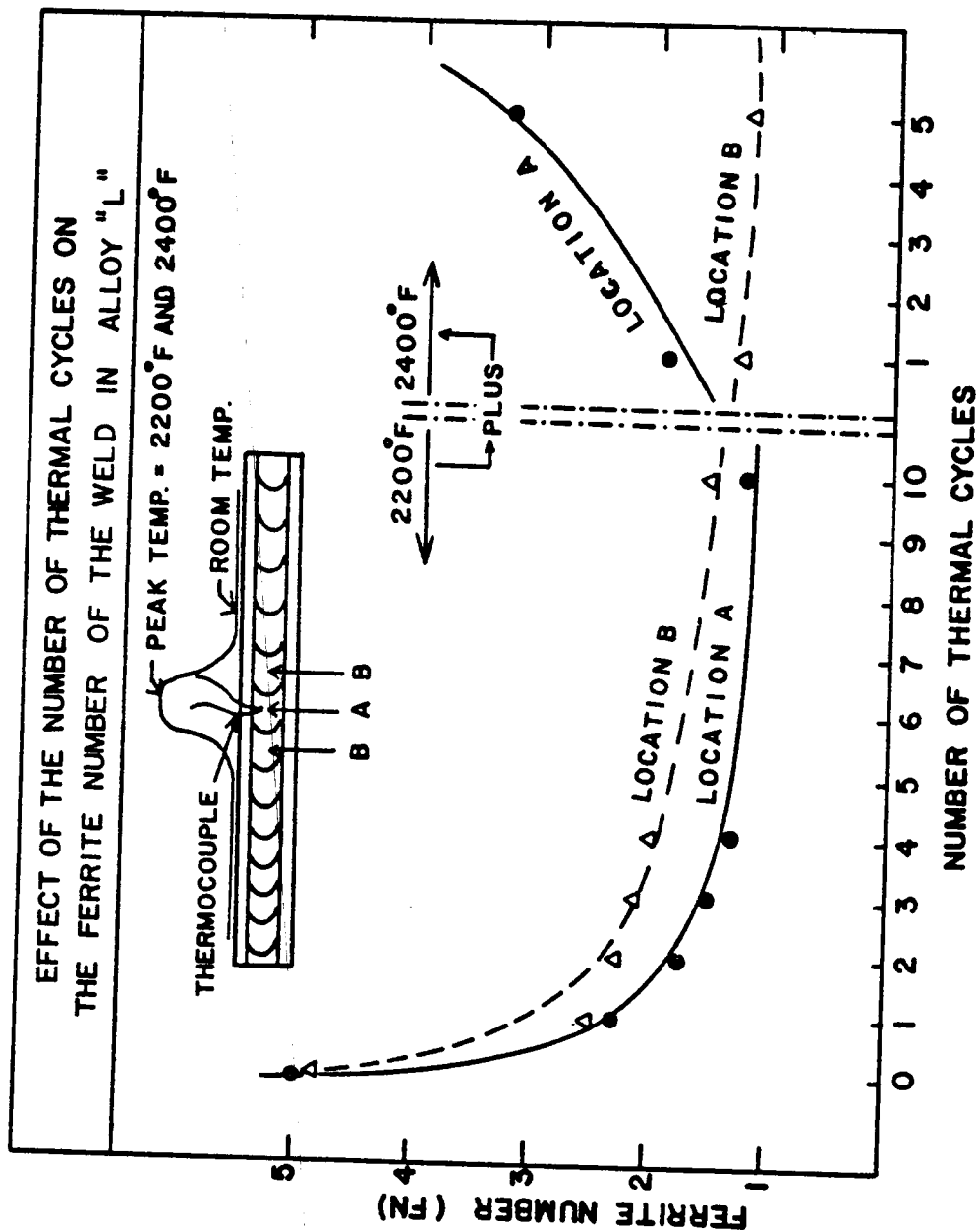


Figure 87. Effect of the Number of Thermal Cycles on the Ferrite Number of the Weld in Alloy "L".

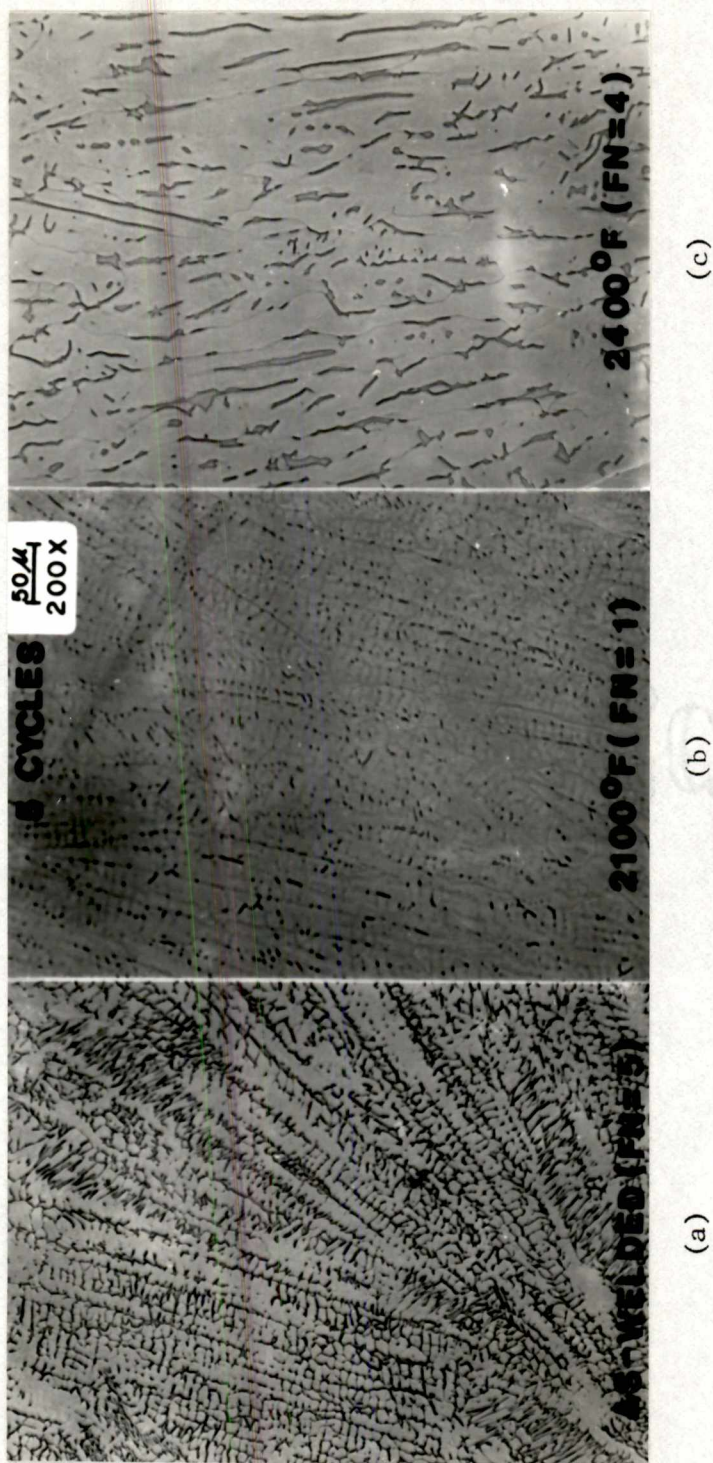


Figure 88. Microstructures of Alloy "L" Thermal Cycled Ten Times to a Peak Temperature of 2200°F and Five Times to a Peak Temperature of 2400°F. (OLM, 200X)
 (a) As-Welded. (b) Low Temperature Region (2100°F). (c) Central Region (2400°F).

hot zone (which was previously exposed to ten cycles to 2200°F), the ferrite reprecipitated along the austenite grain boundaries as shown in Figure 88c. This indicates that the ferrite was nucleated in the austenite at the grain boundaries during the high temperature exposure (2400°F). The ferrite was identified by SEM. Figure 89 shows the EDAX results on a metallographically prepared specimen of welded material "L" which was thermal cycled five times to 2400°F. Higher Cr, Mo plus S, and Si were noted in the ferrite, while higher Ni and Mn were found in the austenite matrix.

In order to determine the ferrite content of weld metal at high temperature, a quenching technique was developed and tests conducted using the Gleeble. This quenching system is shown schematically in Figure 90. When the specimen reaches the desired peak temperature, the cycle is interrupted and helium gas is blasted onto the hot zone of the specimen from both sides. A cooling rate of 2000°F/sec is easily obtained and the microstructure of the weld metal at the peak temperature is "frozen-in".

Using the quenching technique, a series of tests were conducted on materials "L" and "H" thermal cycled to different peak temperatures. All specimens were thermal cycled four times using the normal welding thermal cycle and rapidly quenched from the peak temperature during the fifth cycle. The results are graphically presented in Figure 91. The measured Ferrite Number is plotted as a function of peak temperature. The open circles represent the data from the quenching technique. Data from the normal cooling technique (refer to Figure 82, page 205) are presented as solid circles for comparison. It is clear that for above

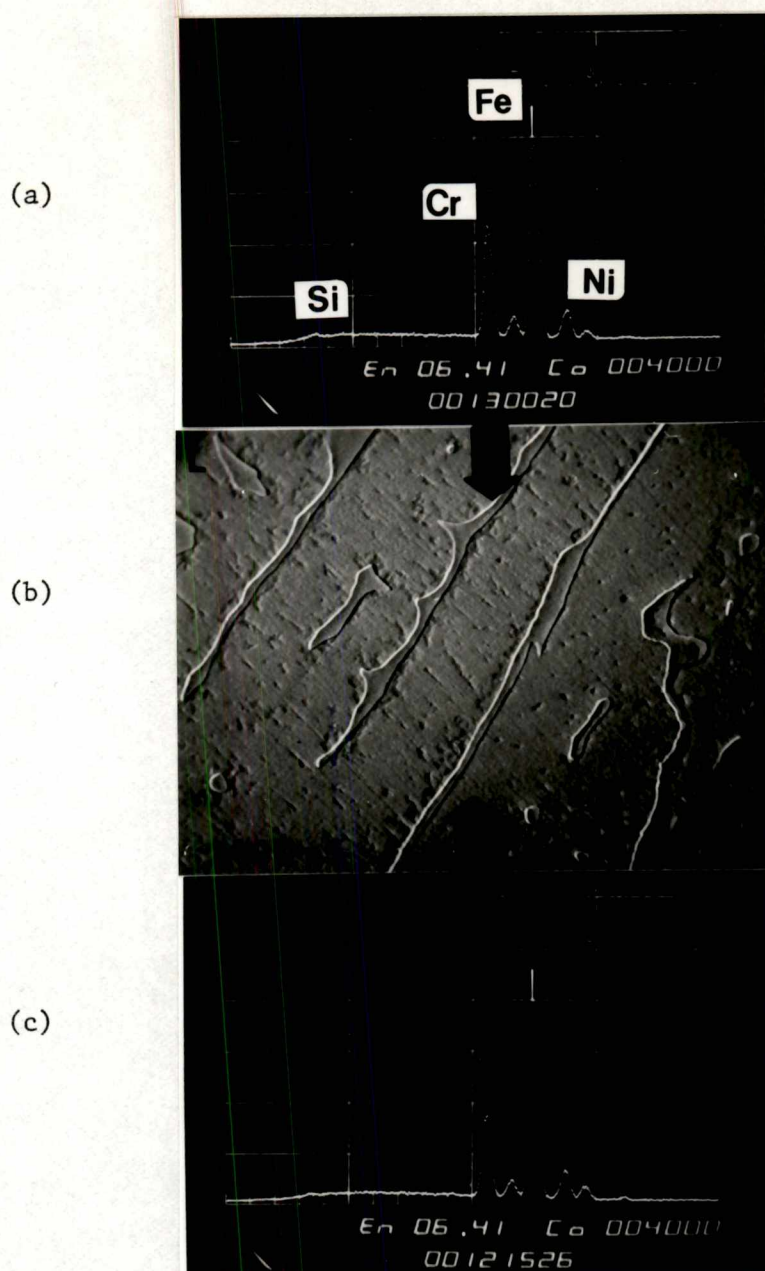


Figure 89. EDAX of Ferrite and Austenite in Specimen "L" Thermal Cycled Five Times to a Peak Temperature of 2400°F. (SEM, 2000X)

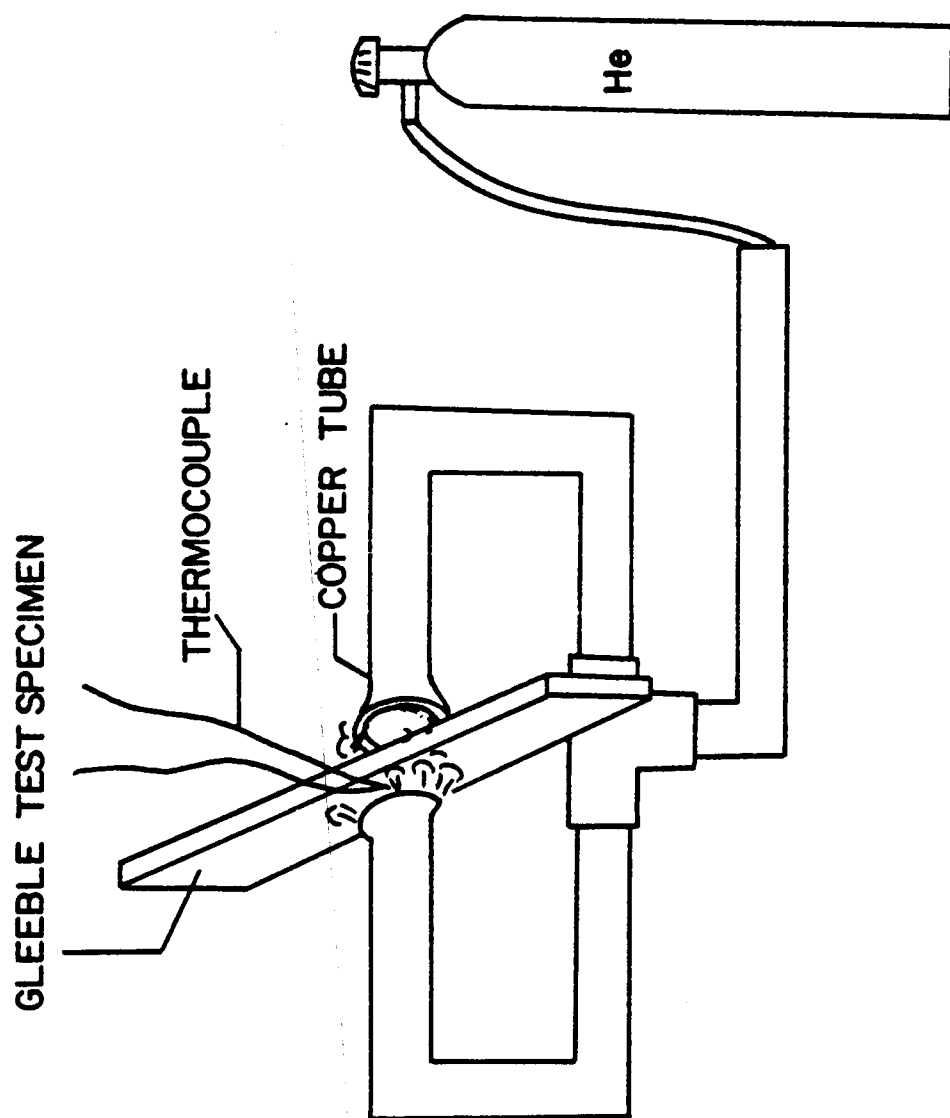


Figure 90. Schematic Diagram Showing the Helium-Blast Cooling Technique.

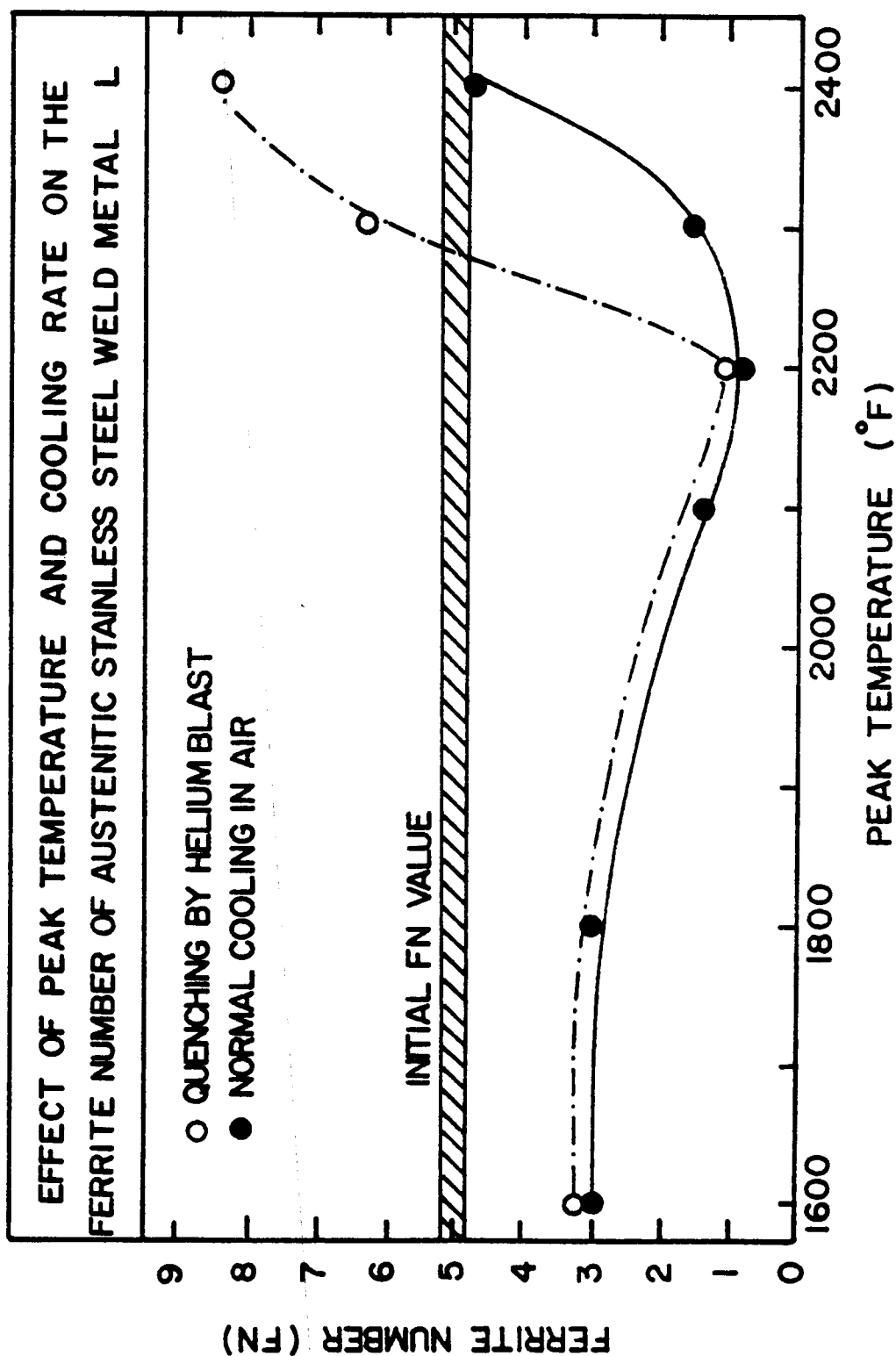


Figure 91. Effect of Peak Temperature and Cooling Rate on The Ferrite Number of Austenitic Stainless Steel Weld Metal "L".

2200°F, the Ferrite Number measured in the weld metal after rapid cooling is higher than that measured in the weld after normal weld cooling. There is no difference in FN measurement between the two cooling methods in the weld metals heated to temperatures up to 2200°F. Similar behavior is also observed for the high ferrite-containing weld sample "H" as shown in Figure 92. Note that the temperature at which a minimum FN was obtained is shifted to about 2000°F.

The above phenomenon can be explained using the Fe-Cr-Ni pseudo-binary phase diagram at 70% iron as shown in Figure 84, page 207. Starting with a weld metal containing approximately 5% residual ferrite (alloy "L"), it can be seen that as the weld metal is reheated to temperatures below the γ -solvus temperature ($\approx 2200^\circ\text{F}$), the residual ferrite should transform to γ by a solid state diffusion-controlled reaction. The higher the temperature the faster the dissolution of ferrite should become because the diffusion rate increases rapidly with temperature. As the temperature is increased above the γ -solvus temperature, "new" δ -ferrite should form. The δ -ferrite content increases as the temperature is increased above the γ -solvus temperature.

The "new" precipitated δ -ferrite can be retained if the temperature is decreased rapidly by quenching (as with a helium blast) after heating. However, the precipitated δ -ferrite may transform back to austenite upon cooling, especially as the cooling rate decreases. Normal weld cycle cooling rates permit some transformation on cooling.

The above observations and explanations are presented schematically in Figure 93. The δ -ferrite decreases from its initial level to a low value with the increase of temperature up to the γ -solvus temperature.

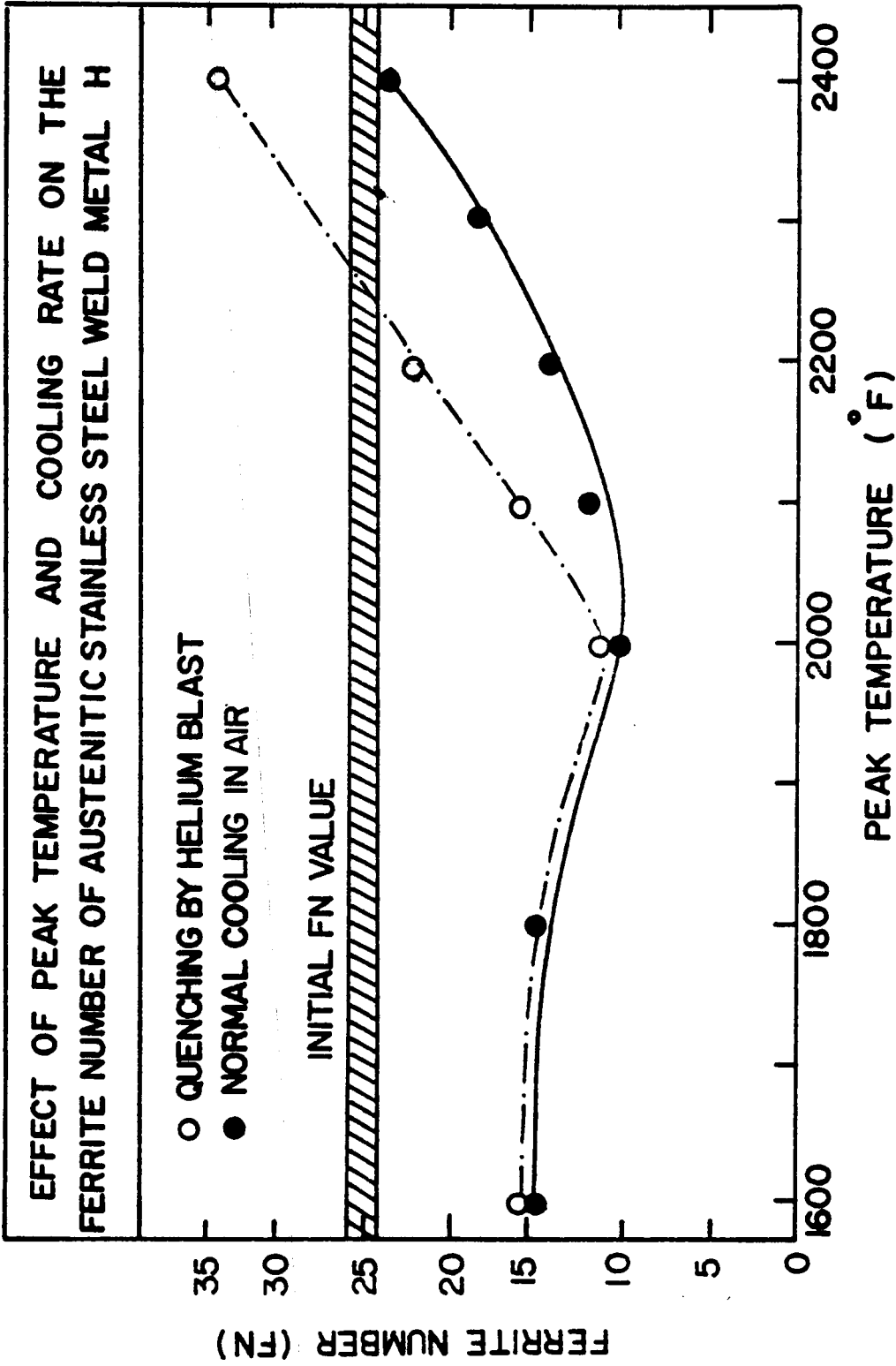


Figure 92. Effect of Peak Temperature and Cooling Rate on the Ferrite Number of Austenitic Stainless Steel Weld Metal "H".

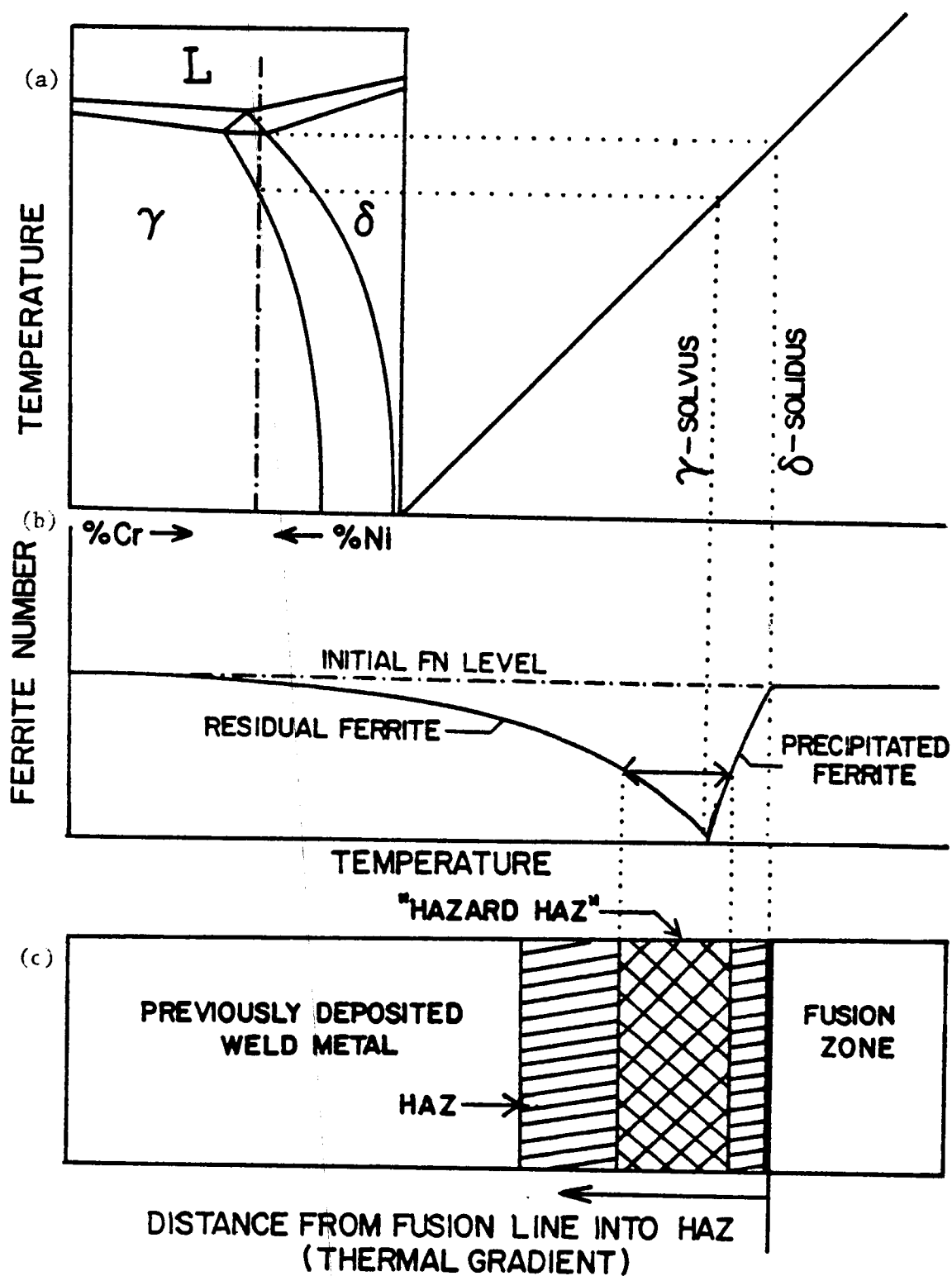


Figure 93. Schematic Diagram Showing the "Hazard HAZ" Region.

"New" precipitated δ -ferrite forms at the γ -solvus temperature and increases to some finite value as the temperature increases toward the solidus temperature. The predicted ferrite content as a function of reheat temperature is plotted in Figure 93b. This prediction is quite consistent with the experimental results. A schematic representation of a weld pass (Figure 93c) shows the weld fusion line and the HAZ region produced during deposition. Note that in the HAZ region there is an area in which the previously as-deposited FN has decreased substantially and this region has been named the "Hazard HAZ" by the author. In this "Hazard HAZ" region, the ductility is relatively low and fissures may form as the strain imposed upon the weld exceeds the strain tolerance of the microstructure.

The cause of the low ductility in the "Hazard HAZ" region is the embrittlement of austenite grain boundaries. It is known that delta ferrite has a greater solubility than austenite for certain "harmful" elements, such as silicon, phosphorous, and sulphur which are believed to cause fissuring. Ferrite reduces the extent of liquid film formation by decreasing the tendency of these "harmful" elements to segregate to the austenite grain boundaries during welding. As the weld metal is reheated to high temperature by subsequent passes, the residual ferrite transforms into austenite and the "harmful" elements are released and redistributed in the austenite grains and at the grain boundaries. As the temperature reaches the γ -solvus temperature, "new" delta ferrite nucleates at the austenite grain boundaries and again absorbs the "harmful" elements. Upon cooling when the temperature drops below the solvus temperature, most of the delta ferrite transforms back to austenite

leaving behind the "harmful" elements trapped at the austenite grain boundaries. It is in these degraded austenite grain boundaries that the ductility is decreased when compared to the adjacent grain interiors. The region which experiences this phenomenon and exhibits lower ductility is called the "Hazard HAZ" region. It is noted that this "Hazard HAZ" region is separated somewhat from the fusion zone of the subsequent bead. In the near-fusion-line region, which is reheated to temperatures well above the γ -solvus temperature, an appreciable amount of ferrite is formed during welding. Therefore, a greater amount of ferrite is retained during cooling and less of the "harmful" elements are released at the grain boundaries (as compared to the "Hazard HAZ" region). Thus, a higher ductility is retained in the high-temperature HAZ region which is adjacent to the fusion zone. However, if this high-temperature HAZ region is reheated again to a temperature just below the γ -solvus temperature, it becomes an active "Hazard HAZ" region. This is because that most of the ferrite is situated, and thus dissolves, at the austenite grain boundaries and the "harmful" elements are released and embrittle grain boundaries.

For a high-ferrite potential weld, the "Hazard HAZ" region is narrower as compared to a low-ferrite potential weld. Figure 94 shows a pseudo-binary diagram in schematic form with the approximate relative locations of low-ferrite and high-ferrite welds indicated by the vertical lines. In order to obtain a high-ferrite potential weld, the Cr/Ni ratio is increased with constant iron content. Therefore, the vertical line representing the weld metal composition is shifted to the right and the intersection with the solvus line decreases in temperature.

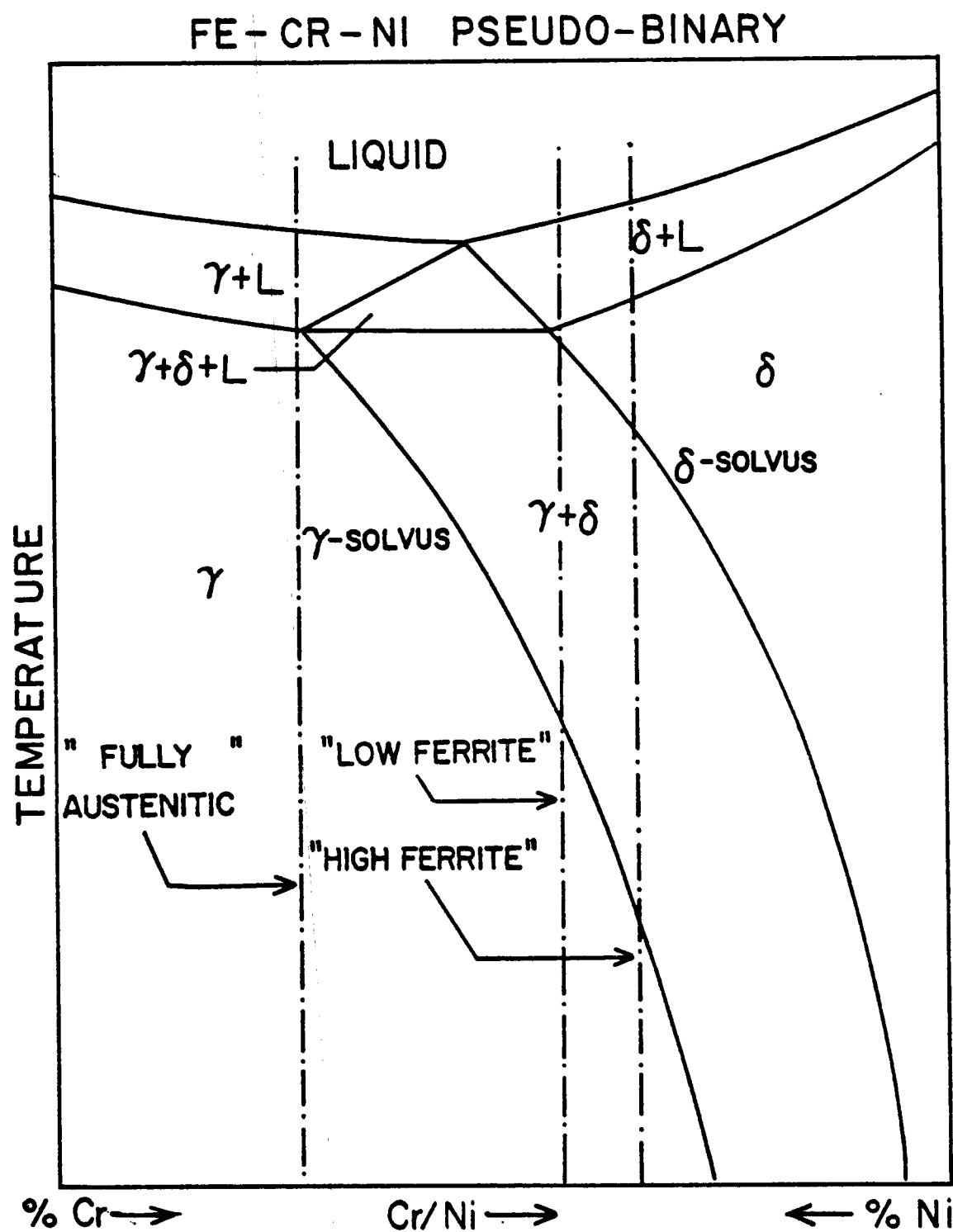


Figure 94. Generalized Pseudo-Binary Diagram Showing Relative Positions of Low Ferrite Alloy and High Ferrite Alloy.

Since the γ -solvus temperature is lower for the high-ferrite weld than the low-ferrite weld, a narrower "Hazard HAZ" region is predictable. The fraction of ferrite remaining as a function of reheat temperature is plotted for both weld metals "L" and "H" as shown in Figure 95. An arbitrary horizontal line indicating 50% remaining ferrite is also illustrated in Figure 95. The HAZ area corresponding to the temperature region in which the ferrite remaining after reheating is less than 50% of the initial value is the extent of the "Hazard HAZ". The temperature range corresponding to the "Hazard HAZ" region in the low-ferrite weld "L" is from approximate 1900°F to 2350°F, while the temperature range in the high-ferrite weld "H" is from approximate 1950°F to 2150°F. Therefore, the "Hazard HAZ" region in the high-ferrite weld is narrower than the "Hazard HAZ" region in the low-ferrite weld.

In actual multipass welds, the low-ferrite "Hazard HAZ" region is too narrow ($\approx 0.030''$) to be detected with any conventional magnetic instrument. This also may be the reason why many investigators have overlooked this phenomenon. Using the Gleeble technique, a wider, uniform, hot zone is obtained. The "Hazard HAZ" was detected by Magne Gage and confirmed by metallographic examination. Lundin and Spond (9) reported that the majority of the fissures form at temperatures significantly below the bulk solidus in a range of 1950°F-2400°F. They also reported the occurrence of fissures in ferrite-free areas while the ferrite level was recorded to 4.0 FN when measured with a Magne Gage. The "Hazard HAZ" region discovered in this study shows a good correlation with their results on cracking location in austenitic welds.

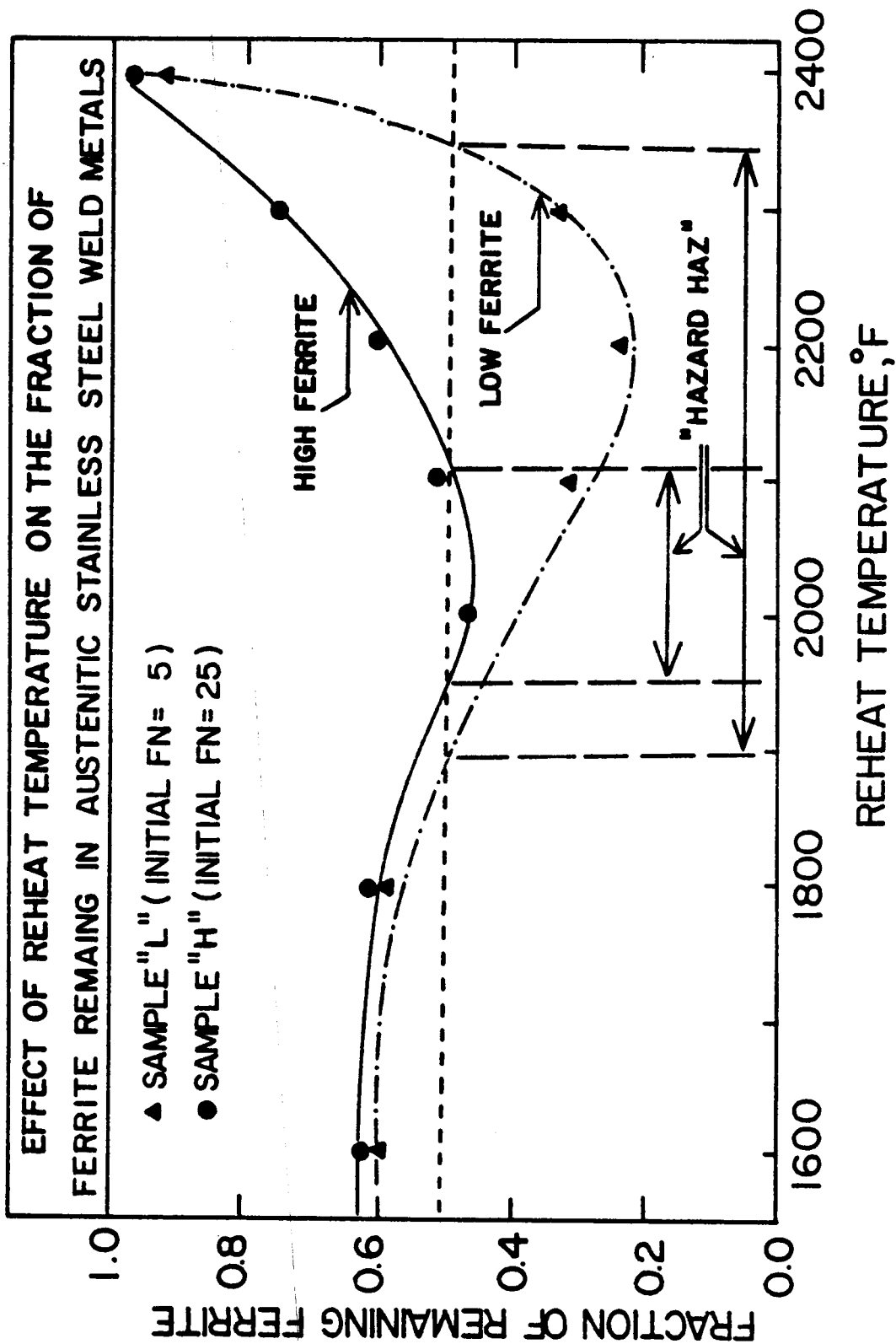


Figure 95. Effect of Reheat Temperature on the Fraction of Ferrite Remaining in Austenitic Stainless Steel Weld Metals.

In the fully austenitic stainless steels, the Cr/Ni ratio is decreased and the vertical line representing the weld metal composition is shifted to the left as shown in Figure 95. As the temperature increases the diffusion rate increases and the segregation of "harmful" elements to austenite grain boundaries may be enhanced. With no restrictions to grain growth due to the absence of ferrite, the austenite grains grow appreciably upon heating. Thus, the concentrations of "harmful" elements per unit surface area is greater. (The "moving" grain boundaries also enhance partitioning as they sweep through the structure.) A grain surface enriched with Si, S, and P may result in embrittlement and the grain boundaries may even be liquated if a sufficiently high temperature is attained during reheating. The temperature range in which the ductility is at a low value is approximately from 1800°F to the solidus temperature. This "Hazard HAZ" region in fully austenitic welds is more extensive than the "Hazard HAZ" region in the ferrite-containing welds. Thus, possibility of fissuring in the weld metal HAZ is higher in the fully austenitic welds than in the ferrite-containing welds. The mechanism of hot cracking in the "Hazard HAZ", the effect of alloying elements, the magnitude of strain, the peak temperature, and the number of thermal cycles on the hot cracking sensitivity was investigated using the Gleeble and are discussed in the next section.

6. Hot Cracking Sensitivity Evaluation Using the Gleeble Technique

It has been found that fissures occur primarily along the grain boundaries of reheated weld metal (HAZ produced by subsequent passes) and that multiple weld metal HAZ thermal cycles enhance the fissuring

propensity in austenitic weld metals. In order to study the cause and mechanism of fissuring and the effect of the magnitude of strain, peak temperature, alloying elements, and the number of thermal cycles on the fissuring tendency, the Gleeble Hot Cracking Test was utilized. This test was originally developed by Krantz for studying the hot cracking in Hastalloy X (117).

The test was selected after considering that microfissuring is more commonly found in multipass welds than in single pass welds. Furthermore, the most prevalent location has been reported to be in those portions of adjacent weld beads which constitute the HAZ of a subsequent pass. Thus, in multipass welds, the thermal cycles and shrinkage strains imposed by each consecutive pass upon the adjacent weld beads appear to contribute to the formation of microcracks.

The major advantage of using the Gleeble lies in the ability to separate the effects of mechanical variable (strain) and metallurgical variable (thermal history). Thus, the relative importance of microstructure, strain, and temperature on hot cracking of weld metals can be explored systematically.

In the Gleeble Hot Cracking Test, a longitudinal composite weld specimen is used. The specimen is clamped in the Gleeble with a 1.0" free span between the water-cooled copper jaws as shown in Figure 21, page 88. The central portion of the specimen is subjected to a thermal cycle, which duplicates the thermal conditions at a given point in the HAZ of an actual weld. A schematic representation of the hot cracking testing sequence is shown in Figure 96. A controlled extension strain is introduced when the sample attains the peak temperature and the sample length is maintained fixed during cooling. In this manner, the total

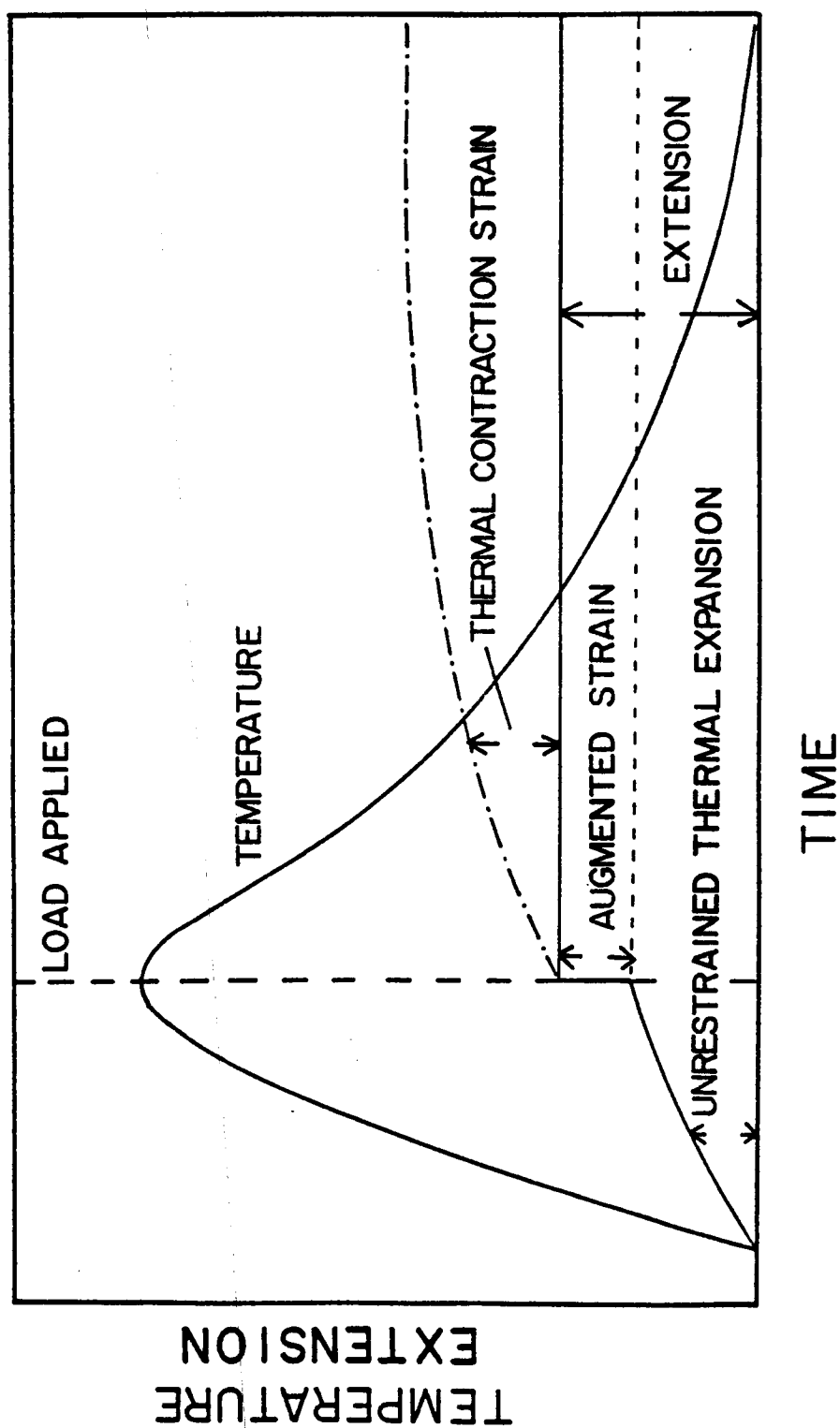


Figure 96. Schematic Representation of Testing Sequence Used in Gleeble Hot Cracking Test.

extension strain (augmented strain plus strain induced by natural contraction) gradually increases as the temperature decreases to room temperature (similar to a weld restraint). All tests are conducted in an inert atmosphere (argon purged chamber) to minimize high temperature oxidation.

Thirteen weld metals were used in this testing. The chemical compositions and calculated FN of the weld metals are shown in Table 6, page 79. "A1" and "A2" are commercial filler metals. "B1", "B2", "B3" are laboratory-made deposits from single commercial bare wires. "D1", "D2", and "D3" are laboratory-made alloys from two wires using the Dual-Wire Feeder technique. "A1Si", "A2Si", "B2N", "B3N", and "B3Ni" are laboratory-made alloys with modified fusion zone chemistry. The measured FN is quite close to the calculated FN for all materials tested except "A2" and "B3". This is because higher Mn content in "A2" and the Mo in "B3" are out of the valid composition range for calculating the chromium equivalent and nickel equivalent in accordance with the Delong Diagram.

6A. Effect of Magnitude of Strain and Peak Temperature on the Hot Cracking Sensitivity

The major parameters involved in the Gleeble Hot Cracking Test are the peak temperature of the thermal cycle and the magnitude of the strain applied during testing. In order to accurately control the magnitude of the applied strain, a differential-screw device was used.

The effect of the magnitude of the extension strain on the hot cracking propensity of fully austenitic stainless steel weld metal "A1" is shown in Figure 97. A series of specimens were thermal cycled with a different magnitude of strain applied at peak temperatures of 2200°F and 2300°F. The strain rate used was 2.5 (in/in)/sec. It was found that the

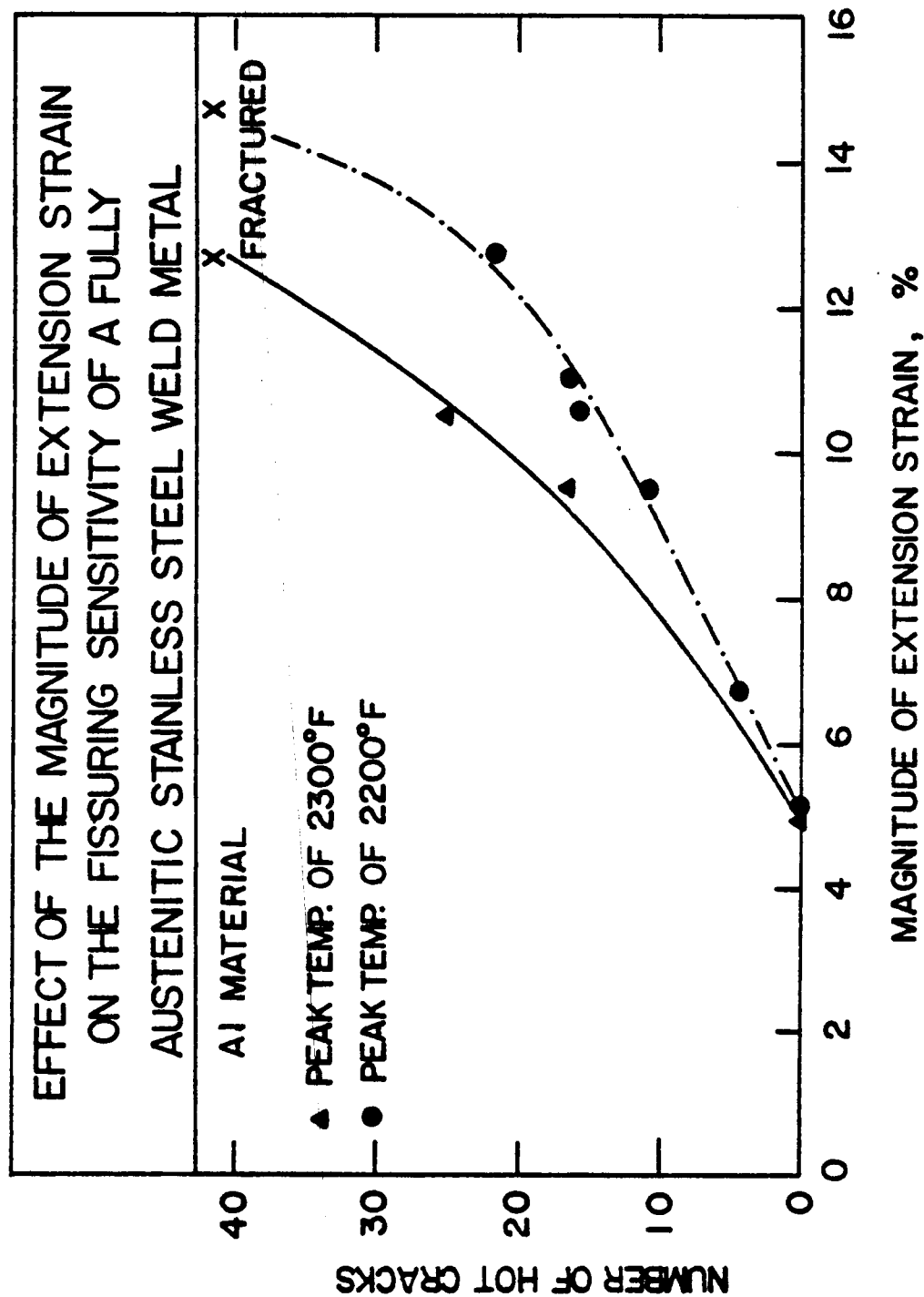


Figure 97. Effect of the Magnitude of Extension Strain on the Fissuring Sensitivity of Weld Metal "Al".

number of fissures increased with increase in the magnitude of extension strain. At a peak temperature of 2200°F , the specimens fractured with greater than 14.7% total extension strain. And the specimens fractured with greater than 12.7% total extension strain when tested at a peak temperature of 2300°F . With this information, the extension strain selected for the remainder of the study was 9.6% to 10.7% depending on the fissuring sensitivity of the weld metal.

The optical macroscopic appearance of fissures is shown in Figure 98. The arrows in the figure indicate the direction of the tensile load applied during the test. Samples were cycled to a peak temperature of 2300°F and extension strains of 10.7% and 16.0% were applied at peak temperature and held during cooling. It is obvious that all the fissures are along the solidification grain boundaries which are perpendicular to the ripple marks.

The peak temperature of a thermal cycle has a strong effect on the fissuring tendency. This can be seen in Figure 99. A series of tests were conducted on samples of Al which were thermal cycled to different peak temperatures. During thermal cycling, an extension strain was applied at peak and the sample length was maintained fixed during cooling. For a total strain of 10.7%, the number of fissures increased as the peak temperature increased. The threshold temperature was approximately 1800°F and fracture occurred at a 2400°F peak temperature. (It should be noted that to maintain the same total extension strain, the augmented strain applied was decreased as the peak temperature of the thermal cycle was increased). The data presented in Figure 99 should be treated with caution especially at the high peak temperatures. The hot cracks found in the tested specimen formed either at the peak temperature immediately

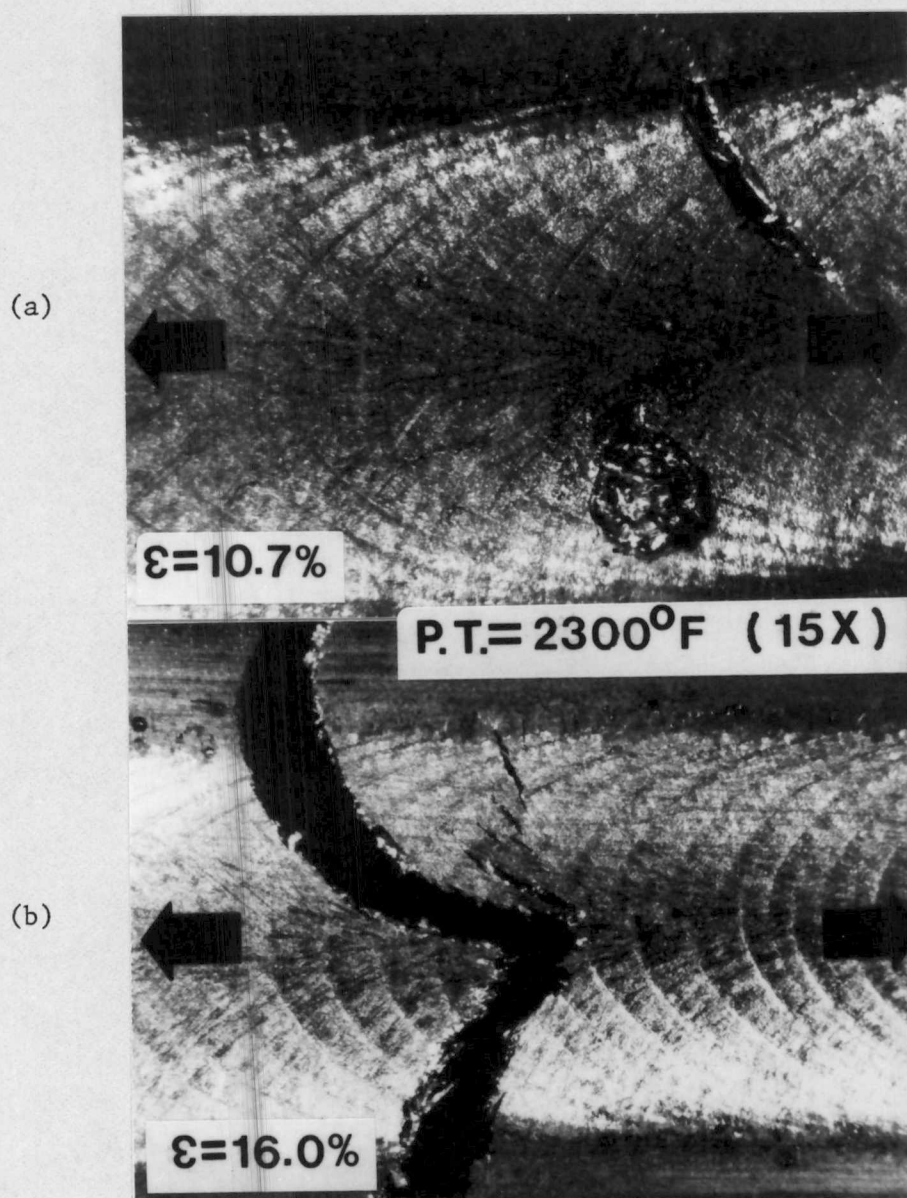


Figure 98. Specimens of "Al" Thermal Cycled to a Peak Temperature of 2300°F with Total Extension Strain of; (a) 10.7% and (b) 16.0%. (15X)

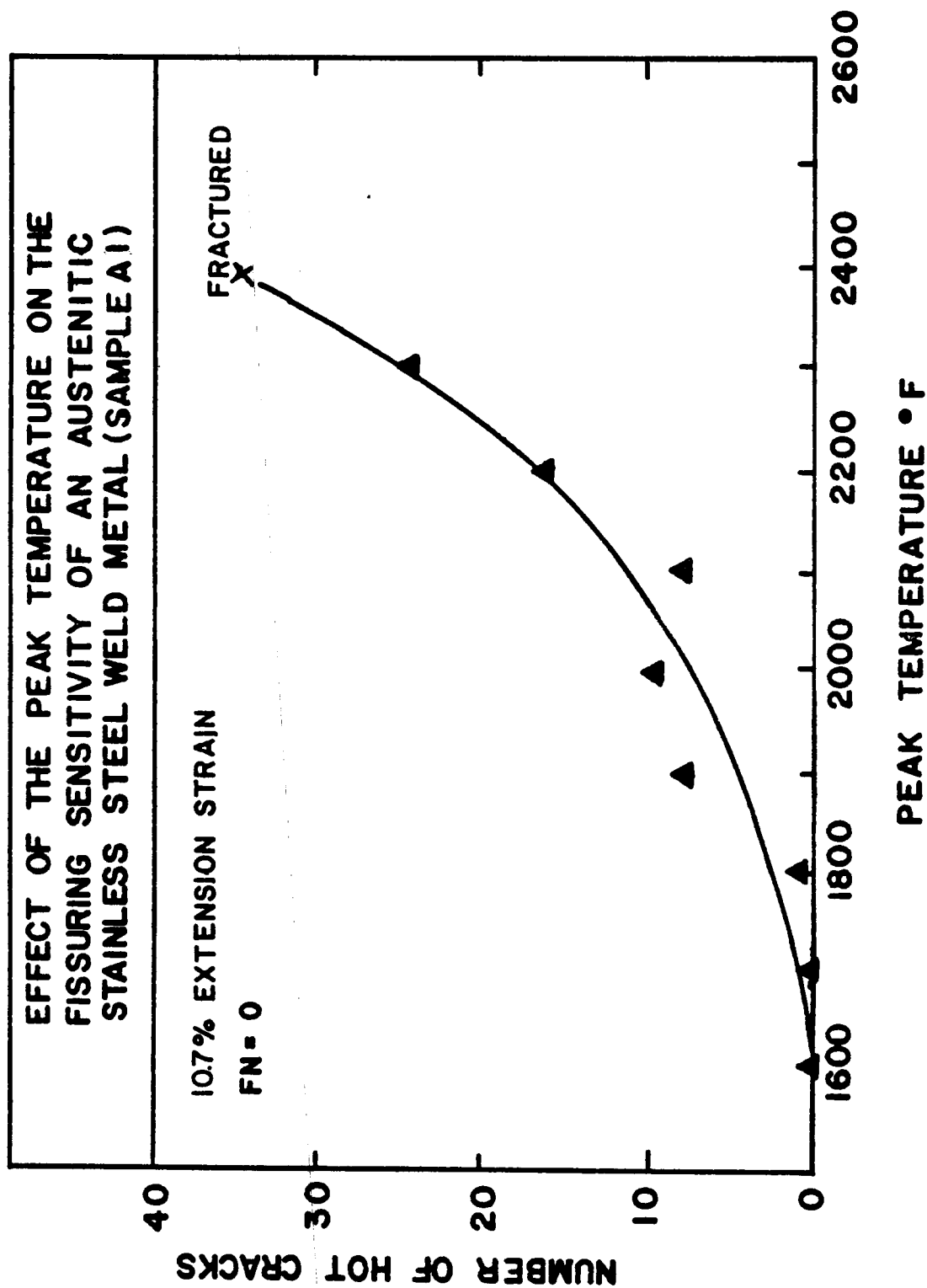


Figure 99. Effect of the Peak Temperature on the Fissuring Sensitivity of Weld Metal "A1".

upon application of the augmented strain or during cooling since the total extension strain (augmented strain plus strain induced by natural contraction of the specimen) increases as the temperature decreases.

In order to determine the temperature range in which the cracks formed, two additional test schemes were employed. The three methods are schematically presented in Figure 100. The standard Gleeble Hot Cracking test is also presented as method 3 in Figure 100. In the first method, the differential screw was adjusted to permit a platen travel equal to the natural thermal expansion of the specimen up to the peak temperature. The platen was fixed at peak temperature inducing only the contractual strains during cooling. In the second method, augmented strain was applied at the peak temperature for 0.5 seconds and then the load was released allowing free contraction of the specimen. In this manner, no contractual strain was imposed during cooling. In the third method (which is the standard test used in this research program), a small augmented strain is applied at the peak temperature and the platen is fixed. Thus, the total extension strain increased during cooling due to the contractual strains. This method is considered to be similar to actual welding since, as the fusion zone cools to the room temperature, the strain in the HAZ gradually increases because of the external restraint offered by the surrounding material.

The three methods were used for peak temperature of 2300°F. The number of cracks found after each evaluation is indicated at the right corner of each graph in Figure 100. Six cracks were found when the contractual-strain-only method was performed using a peak temperature of 2300°F. Hence, all six cracks were formed during cooling. When an augmented strain of 3.5% was applied at the peak temperature for only 0.5

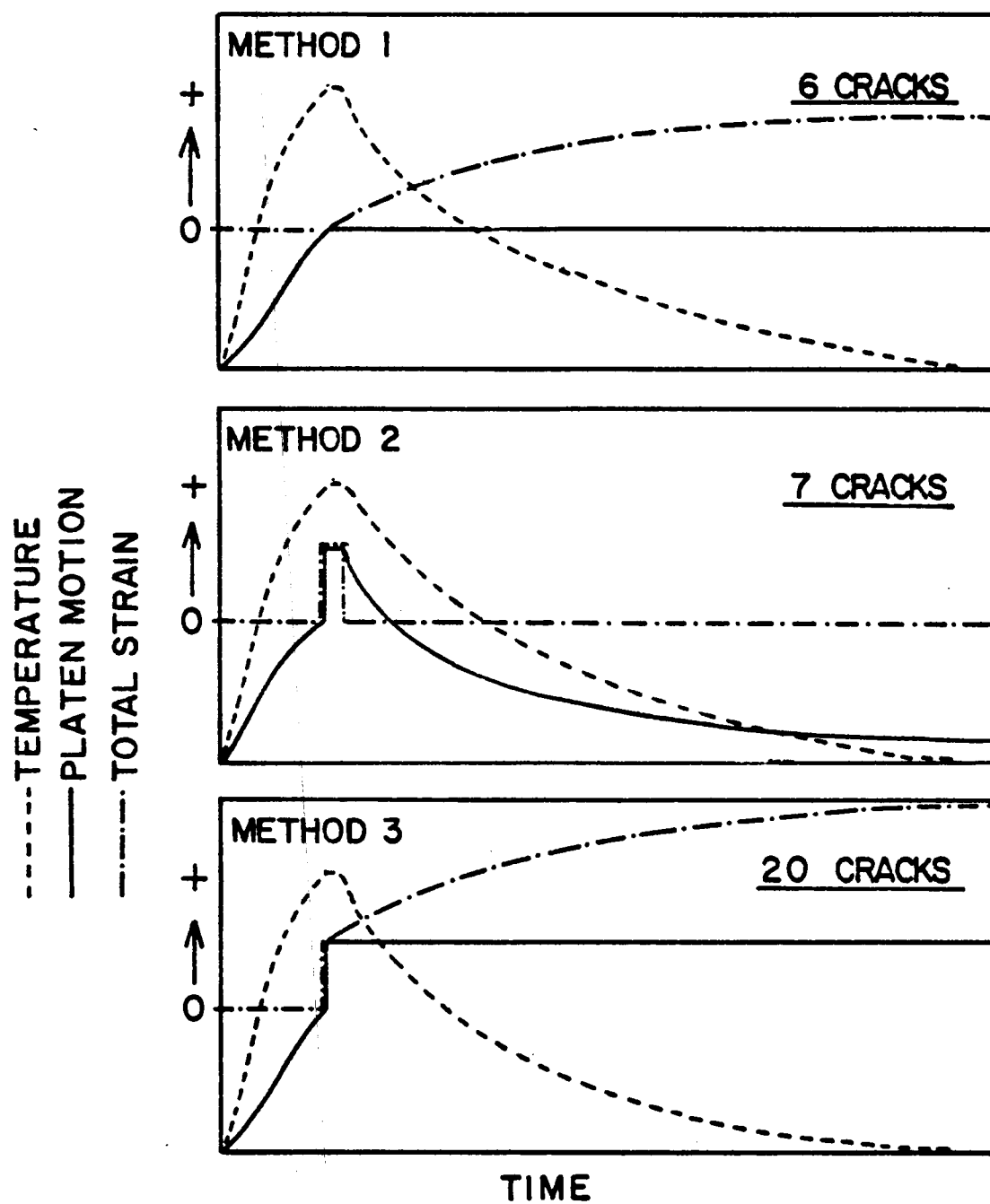


Figure 100. Interrelation of Strain and Temperature for the Three Methods Used in the Gleeble Hot Cracking Test.

seconds and no strain imposed during cooling (method 2), seven cracks were found. These cracks formed at the peak temperature when the augmented strain was applied. In the third method, 3.5% augmented strain was applied at the peak temperature and the total extension strain was allowed to increase as the temperature decreased. Twenty cracks were found in the third method as compared to the seven cracks which were found for the second test method. This indicates that many cracks formed during cooling. Similar results were also observed for the tests performed with a peak temperature of 2100°F.

The above observations are predictable and they can be partially explained with the consideration of elemental segregation and redistribution by diffusion. A typical thermal cycle used in this study is shown in Figure 101. This thermal cycle simulates an energy input of 70,000 joules per inch on a 1½" thick stainless steel plate. Superimposed on the thermal cycle is a temperature band in which diffusion should be considered. The low temperature bound corresponds to a temperature approximately 1100°F (which is about ½ of the absolute melting temperature of stainless steels). The time " t_1 " is the time required for a sample to reach 1100°F, " t_2 ", the time required to attain the peak temperature, and " t_3 ", the time required to fall to 1100°F during cooling from the peak temperature. It is apparent that the time interval ($t_3 - t_1$) is about four times the time interval ($t_2 - t_1$). The average diffusion distance is approximated by $x = 2 \sqrt{D \Delta t}$, where "D" is the average diffusion coefficient over the temperature range, and " Δt " is the corresponding time interval. A solute atom can travel to a longer distance in the time interval ($t_3 - t_1$) as compared to the time interval ($t_2 - t_1$).

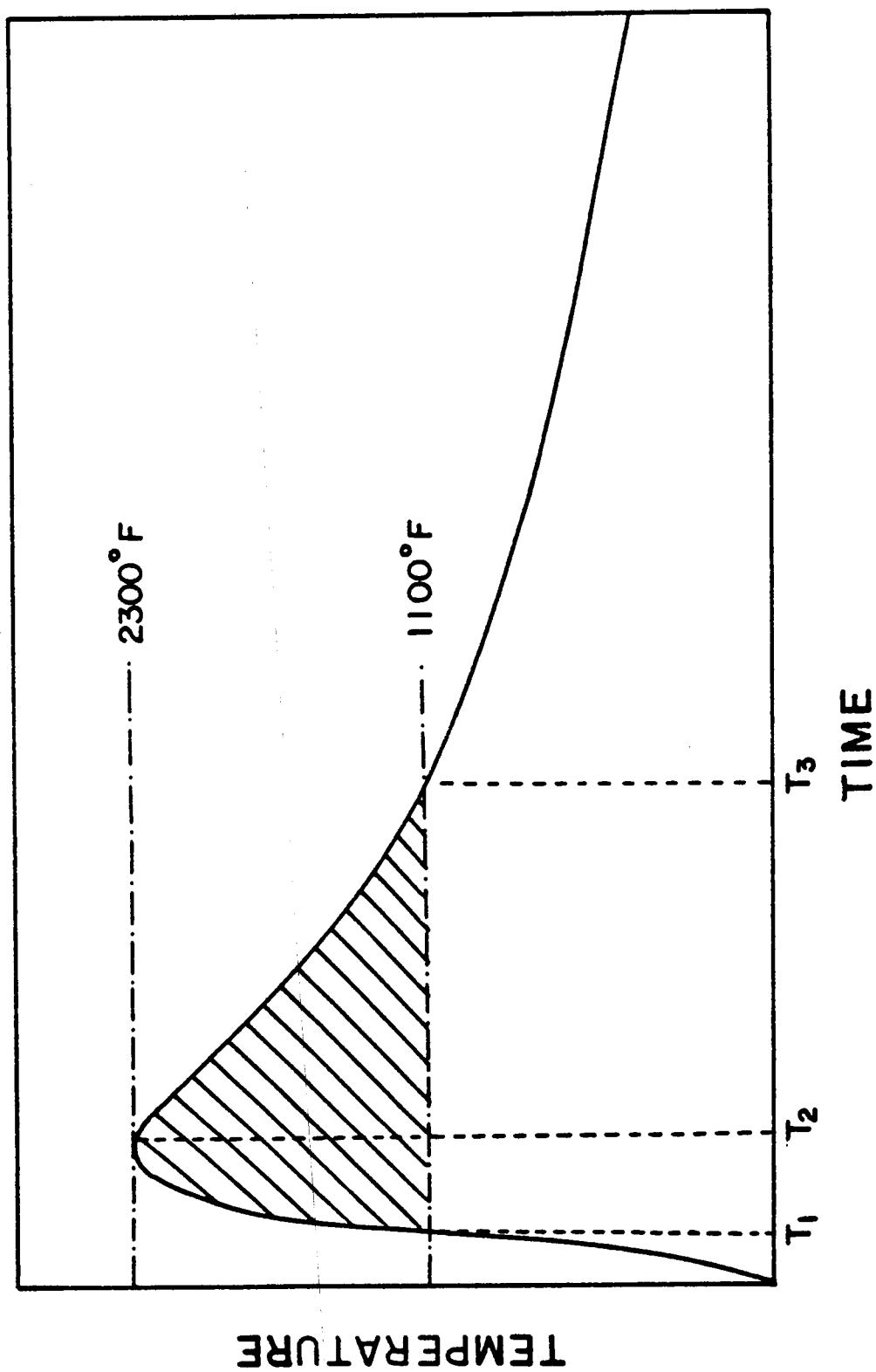


Figure 101. Schematic Thermal Cycle Used in the Gleeble Hot Cracking Test.

Thus, elemental migration and grain boundary movement are expected to be more extensive during cooling than heating and the probability of fissuring is increased when strain is being experienced by the material during cooling. The elemental redistribution and grain growth become more extensive if multiple thermal cycles are experienced by the material prior to strain application. The multiple thermal cycle effect on the hot cracking sensitivity is considered later in this section.

The effect of peak temperature on the hot cracking sensitivity of weld "A2" is shown in Figure 102. Fissuring increases rapidly above 2200°F. At 2400°F peak temperature fracture occurred with 10.7% total extension strain. Comparing these data with the results from the fully austenitic weld "A1", it is found that the "A2" material has a higher fissuring resistance than the fully austenitic "A1". From the results of the Gleeble Hot Ductility Test on the welded materials "A1" and "A2", a very good agreement is obtained between the hot ductility behavior (as presented and discussed in the previous section) and fissuring sensitivity of austenitic stainless steel weld metals. The Ferrite Number in the as-welded "A2" is 0.5 FN and thus it is probable that the "harmful" elements, such as Si, P, and S, are more effectively dissolved in the ferrite. Thus, the segregation of these "harmful" elements to the grain boundaries is greatly reduced. The ferrite was identified by SEM. Figure 103 shows the EDAX results on the as-welded "A2" which was metallographically prepared. Higher Cr, Mo plus S, and Si were noted in the ferrite while higher Ni and Mn were found in the austenite matrix.

Optical photomicrographs of samples of "A1" which experienced the various thermal cycles are shown in Figure 104. No fissures were observed

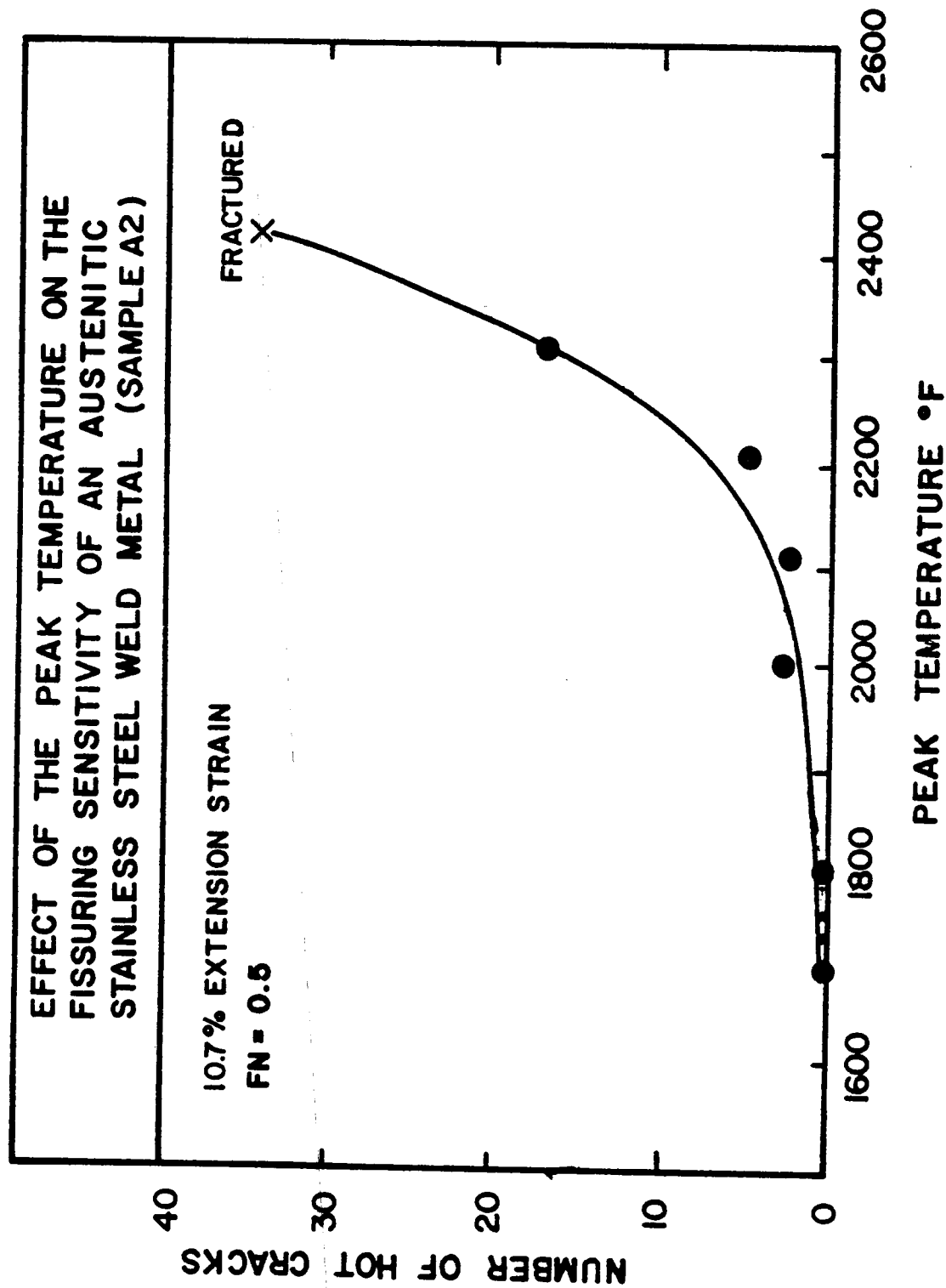


Figure 102. Effect of the Peak Temperature on the Fissuring Sensitivity of Weld Metal "A2".

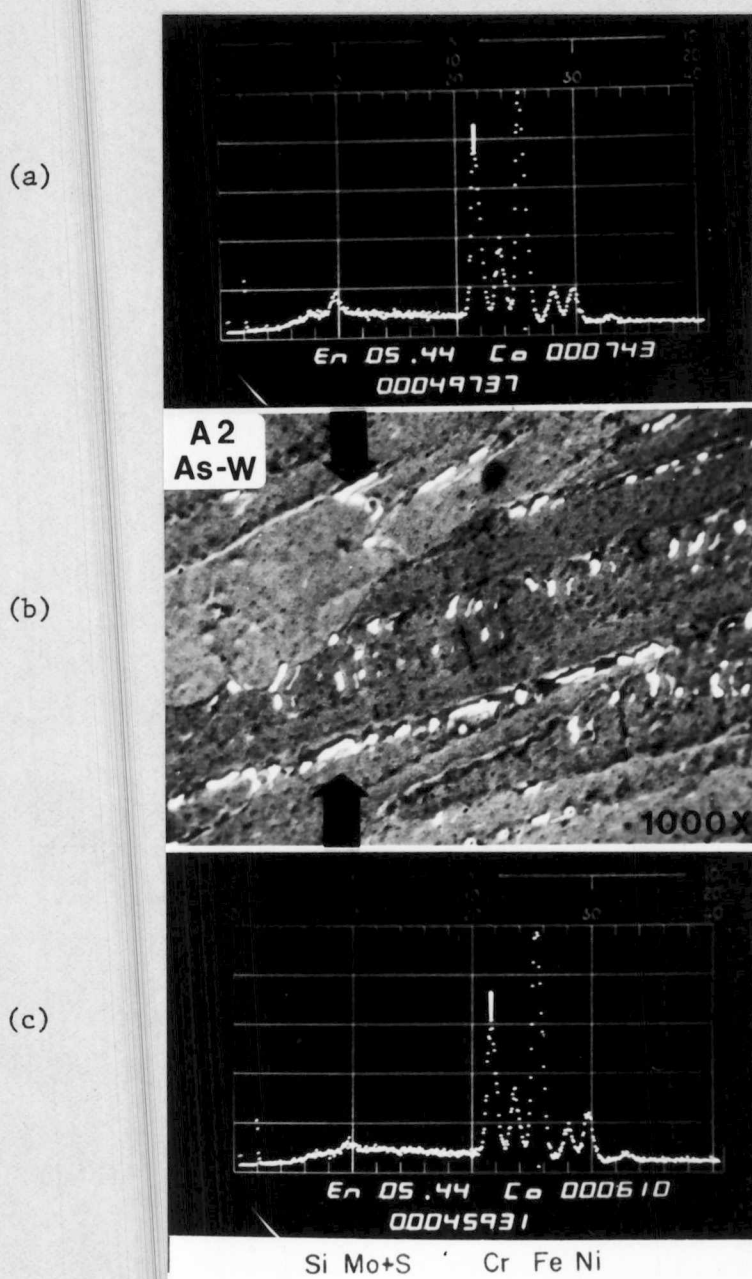


Figure 103. Microstructure of As-Welded "A2".
 (a) EDAX of the Ferrite Phase.
 (b) Microstructure of the Weld Metal. (SEM, 1000X)
 (c) EDAX of the Austenite Matrix.

(a)



(b)



Figure 104. Microstructures of Fully Austenitic Stainless Steel Weld Metal "A1". (OLM, 100X)
(a) Thermal Cycled Twice to 2300°F Without Strain.
(b) Thermal Cycled to 2300°F with 10.7% Strain.

in the weld metal which experienced two thermal cycles with peak temperature of 2300°F without strain applied. With a single thermal cycle to a peak temperature of 2300°F with an extension strain of 10.7%, many fissures were formed. All the fissures were along the grain boundaries.

In order to more clearly define the mechanism and the cause of hot cracking in fully austenitic stainless steel weld metals, the surfaces of fissures were examined utilizing the SEM. In addition, EDAX was performed on the fissure surfaces in an attempt to document segregation of certain elemental species. The sample shown in Figure 104b was notched and fractured under impact tensile load at room temperature. The fissure surface of the fractured sample is shown in Figure 105. A liquid film morphology is seen on the surface as shown at higher magnification in Figure 105b. A high Si EDAX response was found on the fractured surface.

The fissure surface of a sample thermal cycled to 2400°F peak temperature with 10.7% extension strain applied at peak temperature and hold during cooling is shown in Figure 106. Two types of fracture morphologies are observed. Area A shows typical cellular dendrites which formed during initial weld solidification. Area B has a smooth appearance with a number of protuberances extending upward from the surface. The presence of these protuberances suggests that a liquid film existed in this region at the time separation of the material occurred. The protuberances extending upward from the surface result from the liquid metal necks that developed during separation with the liquid film present. The necked regions subsequently ruptured as the fissures opened.

EDAX of the remnant protuberances of the liquid film is shown in Figure 107a. For comparison, the EDAX of base metal is presented in

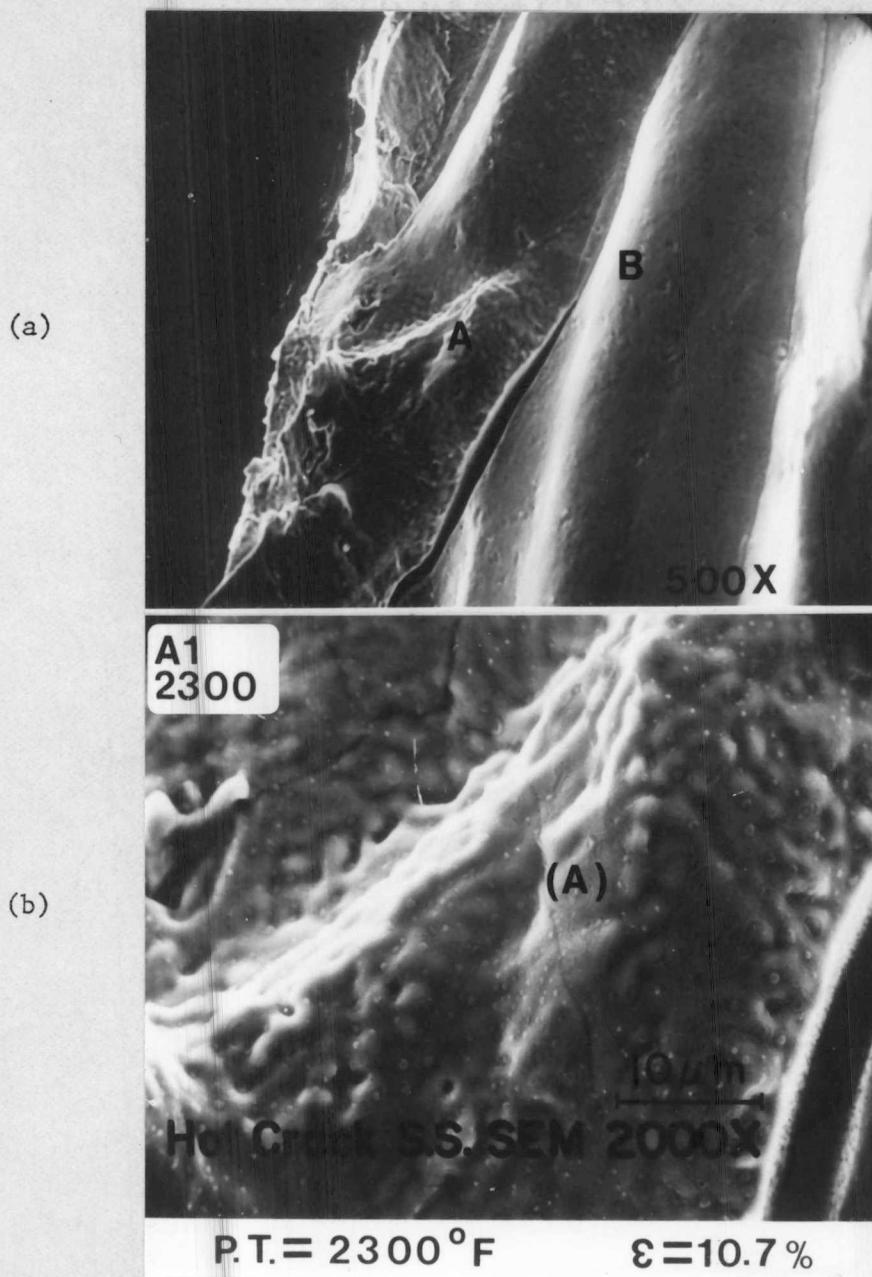
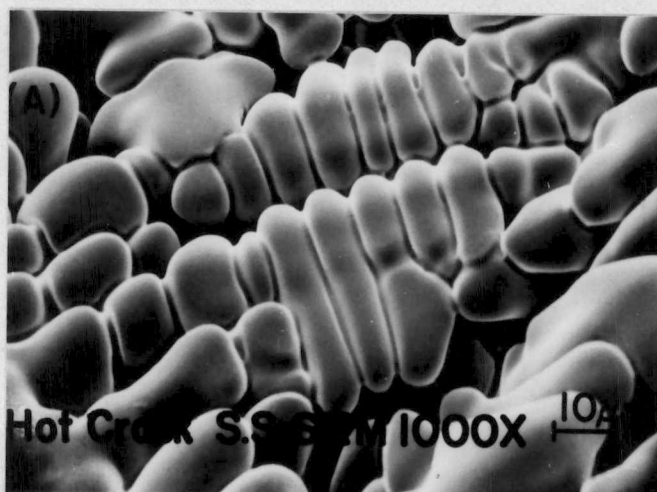
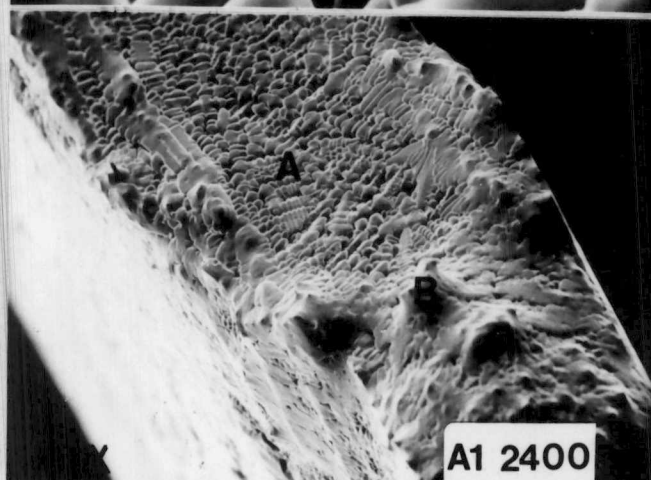


Figure 105. Fissure Surface of "A1" Thermal Cycled to 2300°F with 10.7% Total Extension Strain Applied.
 (a) Fissure Surface. (SEM, 500X)
 (b) High Magnification of Area A. (SEM, 2000X)

(a)



(b)



(c)

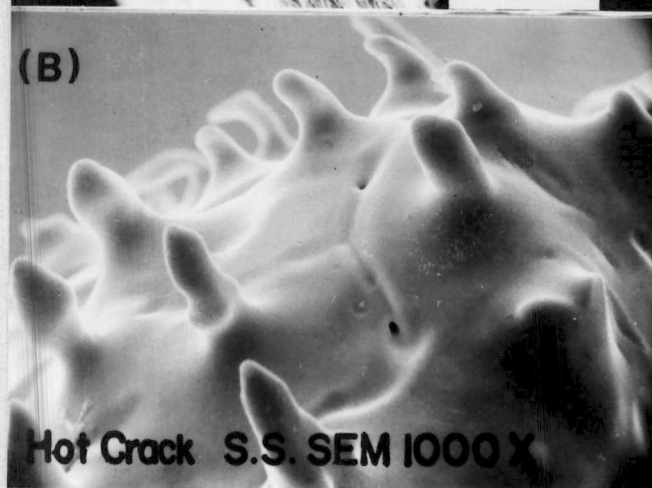


Figure 106. Fissure Surface of "A1" Thermal Cycled to 2400°F with 10.7% Total Extension Strain Applied.

(a) High Magnification of Area A. (SEM, 1000X)

(b) Fissure Surface. (SEM, 50X)

(c) High Magnification of Area B. (SEM, 1000X)

(a)

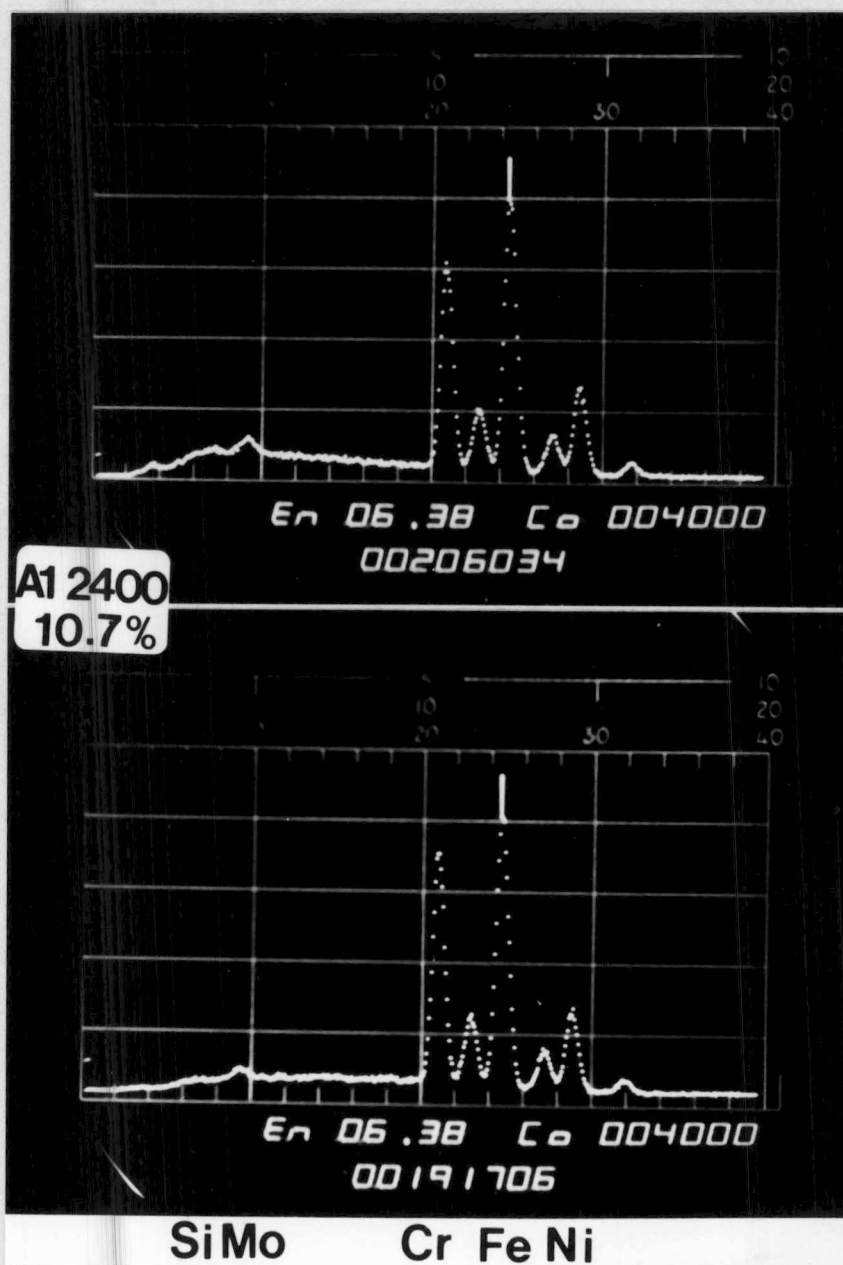


Figure 107. EDAX Results of Fissure Surface of "A1" Thermal Cycled to 2400°F with 10.7% Total Extension Strain Applied.

(a) EDAX of Liquid Film Remnants (Necks).
 (b) EDAX of Base Metal.

Figure 107b. It is obvious that Si content is higher in the liquid film region than in the base metal. This indicates that low melting silicates may form by segregation during weld metal solidification. This is consistent with the results from Fissure Bend Tests.

6B. Effect of Multiple Thermal Cycling on the Hot Cracking Sensitivity

It was found that multiple thermal cycling enhanced the hot cracking tendencies of fully austenitic stainless steel weld metal in the Fissure Bend Test. In order to evaluate the multiple thermal cycle phenomenon more definitively, the multiple HAZ exposures were reproduced in Gleeble samples with and without added extension strain. In these multiple cycle studies, extension strain was applied at peak temperature and maintained during cooling to provide 5% total strain. After each cycle the samples were removed from the test device, examined under the binocular microscope, remounted in the test fixture, and thermal cycled again as appropriate. The peak temperatures used ranged from 1600°F to 2300°F. The results are presented in Figure 108. The number of fissures increased significantly as the number of thermal cycle exposures increased. This was more evident for samples thermal cycled to the higher peak temperatures. For example, there were only twelve fissures found after four cycles to 2300°F peak temperature. The use of 5% extension strain required no augmented strain application for the samples thermal cycled at higher peak temperatures (2100°F to 2300°F) since the thermal expansion equaled approximately 5% (method 1). The results indicate that the accumulated thermal contraction strain plus multiple high temperature thermal cycle exposure has an adverse effect on the hot cracking resistance.

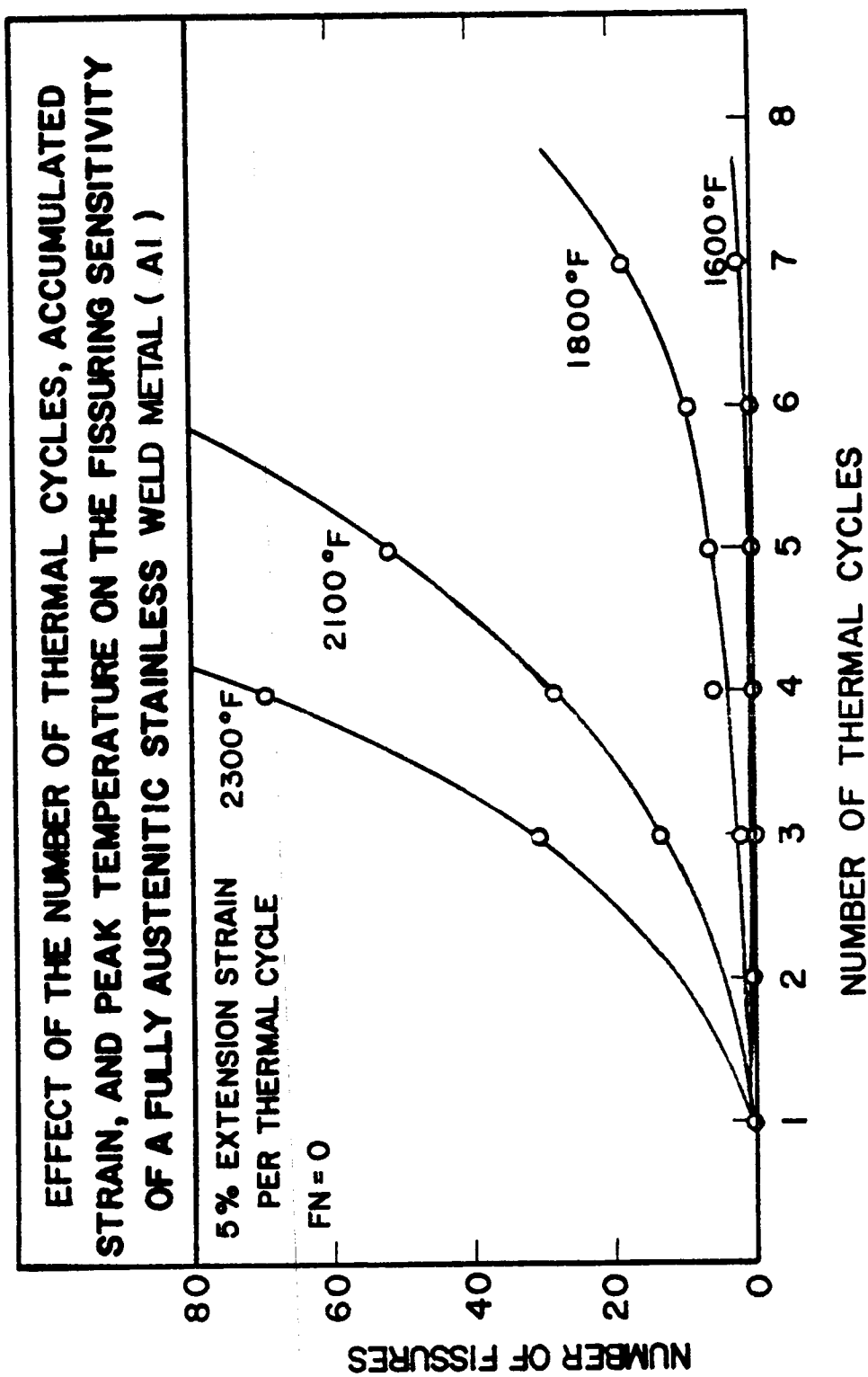


Figure 108. Effect of the Number of Cycles, Accumulated Strain, and Peak Temperature on the Fissuring Sensitivity of Weld Metal "AI".

In order to separate the influence of the thermal history and strain on the fissuring sensitivity, a modified Gleeble test was developed and employed. A specimen is thermal cycled several times without the application strain (specimen allowed to expand and contract freely). During the last thermal cycle, an extension strain is applied at the peak temperature and maintained during cooling to room temperature. Figure 109 shows the results of prior thermal cycling influence on the fissuring sensitivity of weld metal "A1". In Figure 109, the open triangles represent the data obtained from samples thermal cycled once with 10.7% extension strain. The solid triangles represent those obtained from samples which experienced two prior thermal cycles without strain before being cycled a third time with 10.7% total extension strain applied. The number of fissures in the weld metal increased with the prior thermal cycle exposure. Fracture occurred with sample treated this way at a 2300°F peak temperature. This work clearly indicates that high temperature thermal cycling enhances the hot cracking tendency of fully austenitic stainless steel weld metal.

The Ferrite Number was measured for the base metal and weld metal before and after Gleeble testing. All the samples were fully austenitic (FN = 0). Selected specimens were cut, mounted, polished, etched, and examined with optical microscope. The microstructures of fully austenitic stainless steel base metal and weld metal (A1) are shown in Figure 110. A typical fully austenitic structure with annealing twins is observed in the base metal and a cellular dendritic structure with segregation pattern is shown in the weld metal. The microstructures of the composite weld from different locations in a Gleeble tested specimen are shown in .

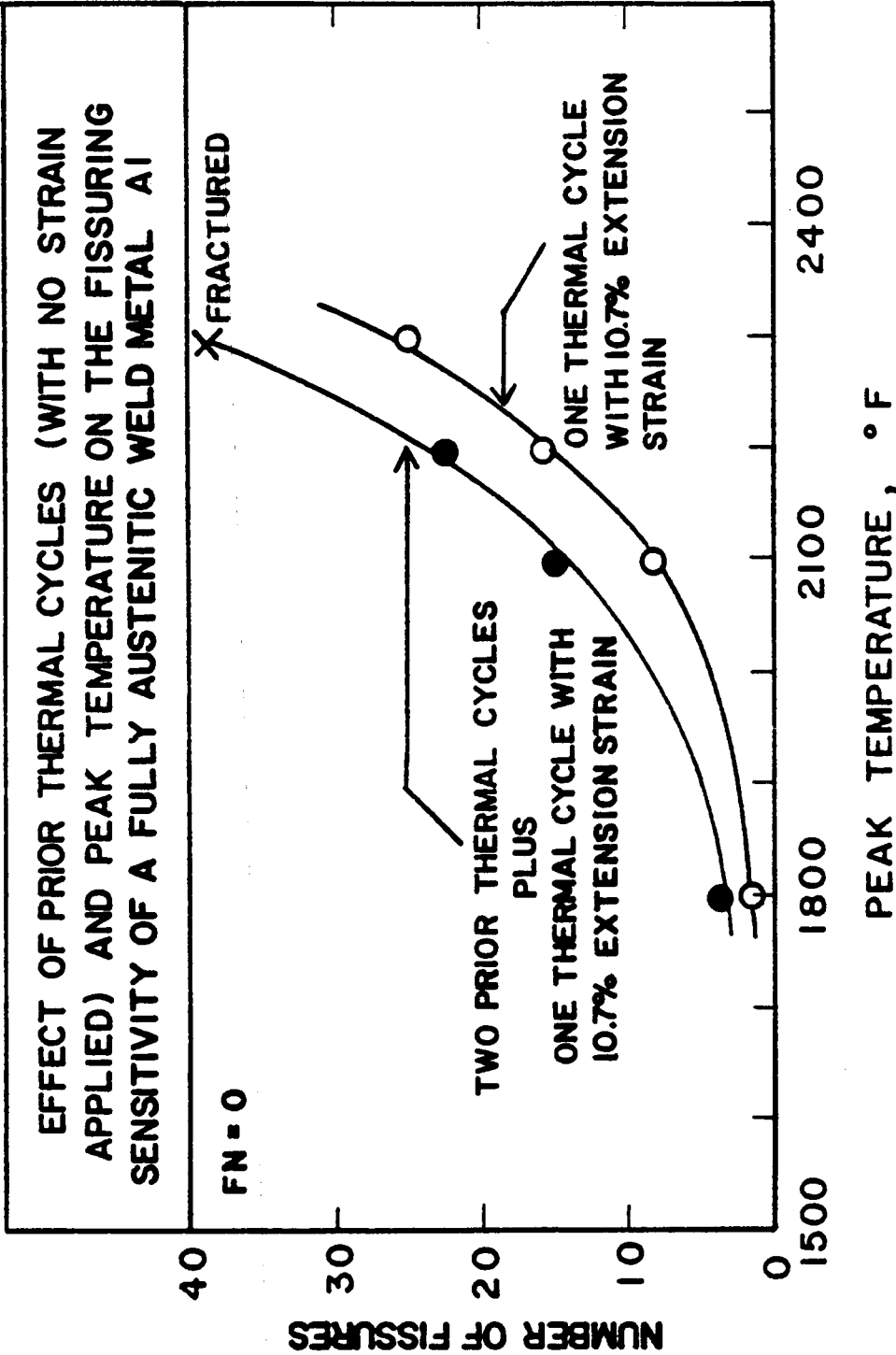


Figure 109. Effect of Prior Thermal Cycles on The Fissuring Sensitivity of "AI".

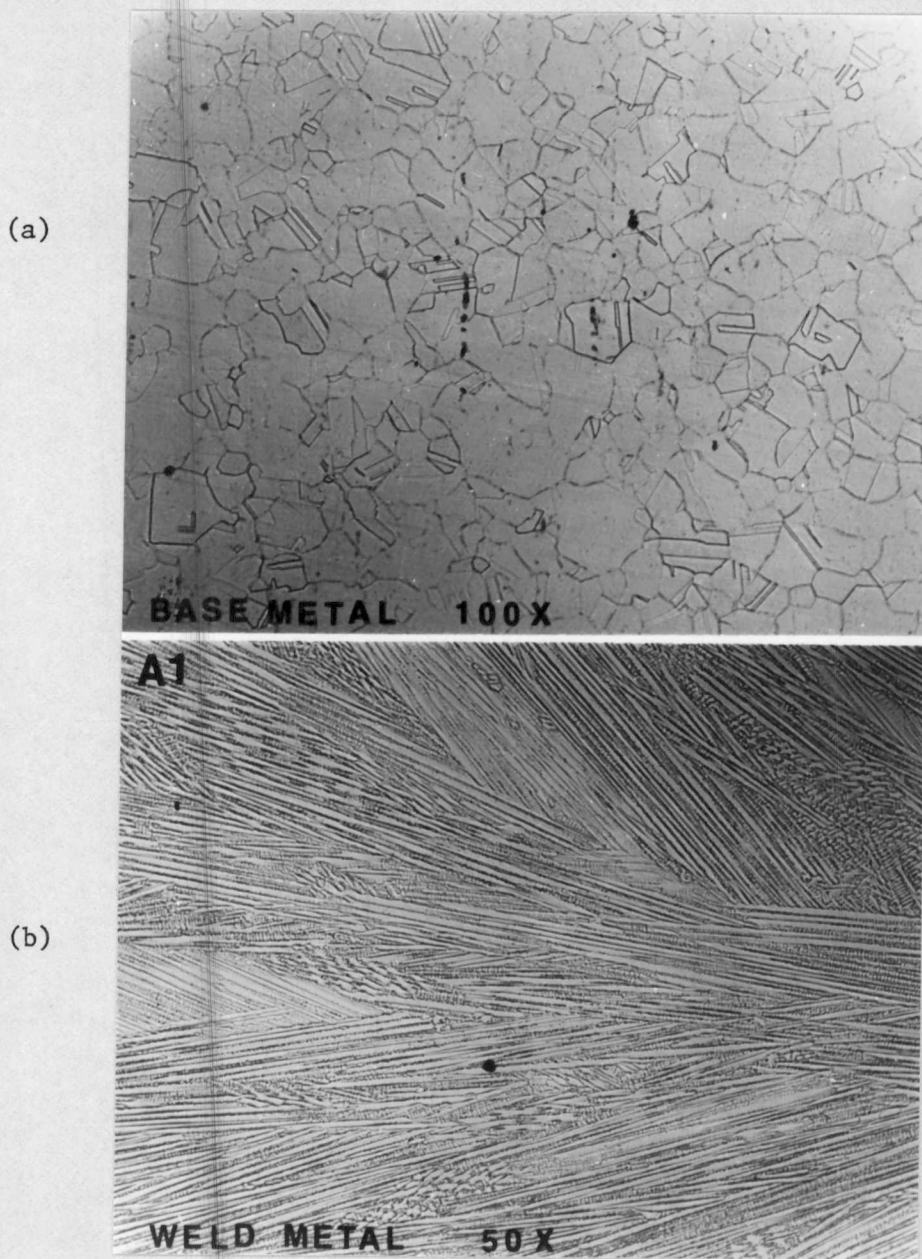


Figure 110. Microstructures of Fully Austenitic Stainless "A1".
(a) Base Metal. (OLM, 100X)
(b) Weld Metal. (OLM, 50X)

Figure 111. The specimen was thermal cycled twice to a peak temperature 2100°F. It was found that grains in the coarsened HAZ region became larger and the degree of segregation in the weld metal was reduced in the central part of the specimen which experienced the highest temperature. Figure 112 Shows the effect of the peak temperature of a thermal cycle on the microstructure of weld metal. A typical cellular dendritic solidification mode with segregation pattern is clearly seen in Figure 112a. Here, Cr-rich regions (interdendritic subgrain boundaries) appear dark; Cr-depleted regions (subgrain cores) appear light. The segregation of Cr at subgrain boundaries was greatly reduced and coarser grain size was observed in the weld metal after thermal cycling to a high peak temperatures as shown in Figures 112b and 112c. This is to be expected since the higher the temperature the higher the diffusion rate of Cr and the more extensive the homogenization.

EDAX was conducted on the as-welded samples and Gleeble tested samples which were thermal cycled six times to a peak temperature of 2100°F. Only the major alloying elements (Fe, Cr, Ni, Mn, Mo) were detected. It was found that Cr content was higher in the subgrain boundaries than in the subgrain cores of the as-welded metal (see Figure 113). This is to be expected if austenite is the primary phase formed during solidification. The "harmful" elements such as Si, P, and S are also expected to be higher in the subgrain boundaries than in the subgrain cores. The lack of detection of these "harmful" elements with EDAX is probably due to the low concentration per unit area (because the number of subgrain boundaries is large). After multiple thermal cycles, chromium segregation was reduced as shown in Figure 114. Since time and temperature are major factors in

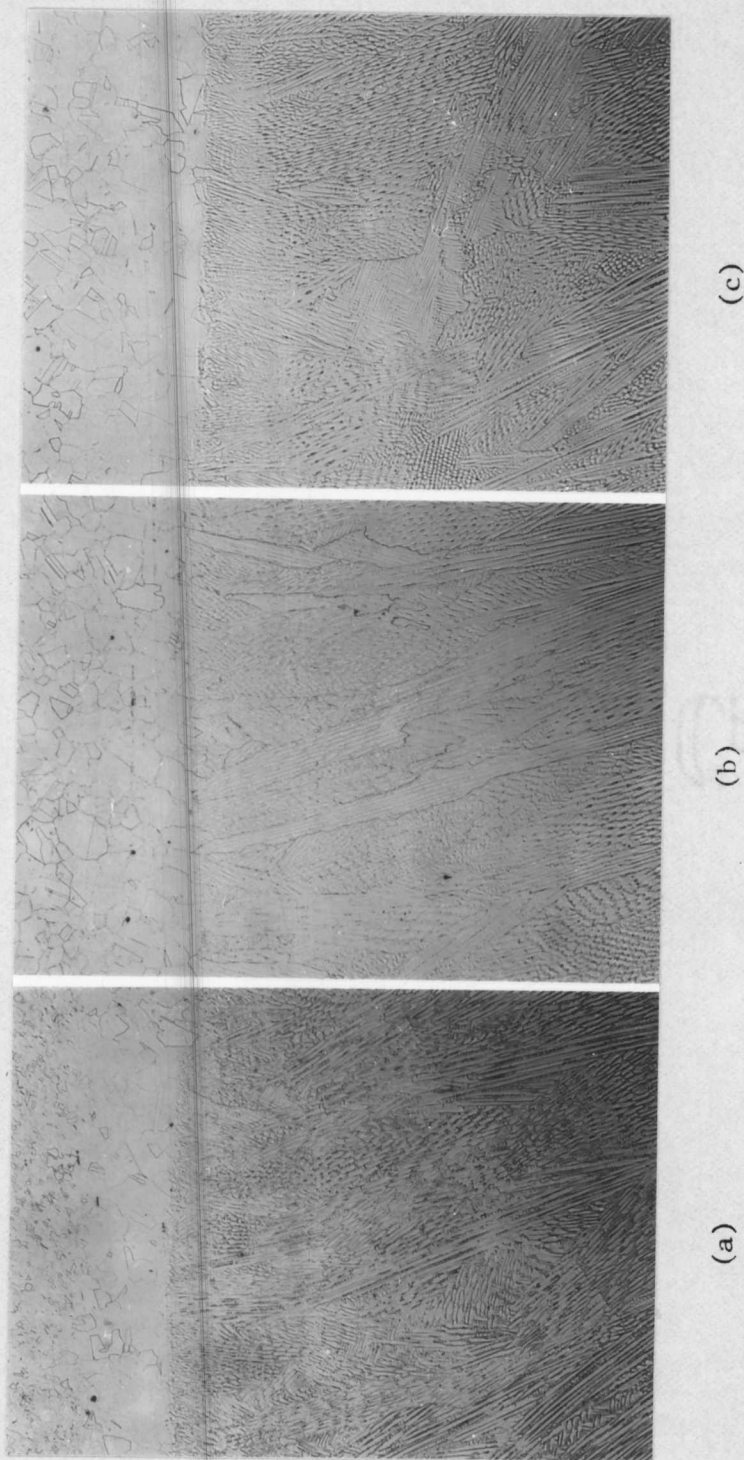


Figure 111. Microstructures of Weld Metal "Al" Thermal Cycled Twice to 2100°F Peak Temperature. (OLM, 50X)
 (a) As-Welded. (b) Low Temperature Region (1800°F). (c) Central Region (2100°F).

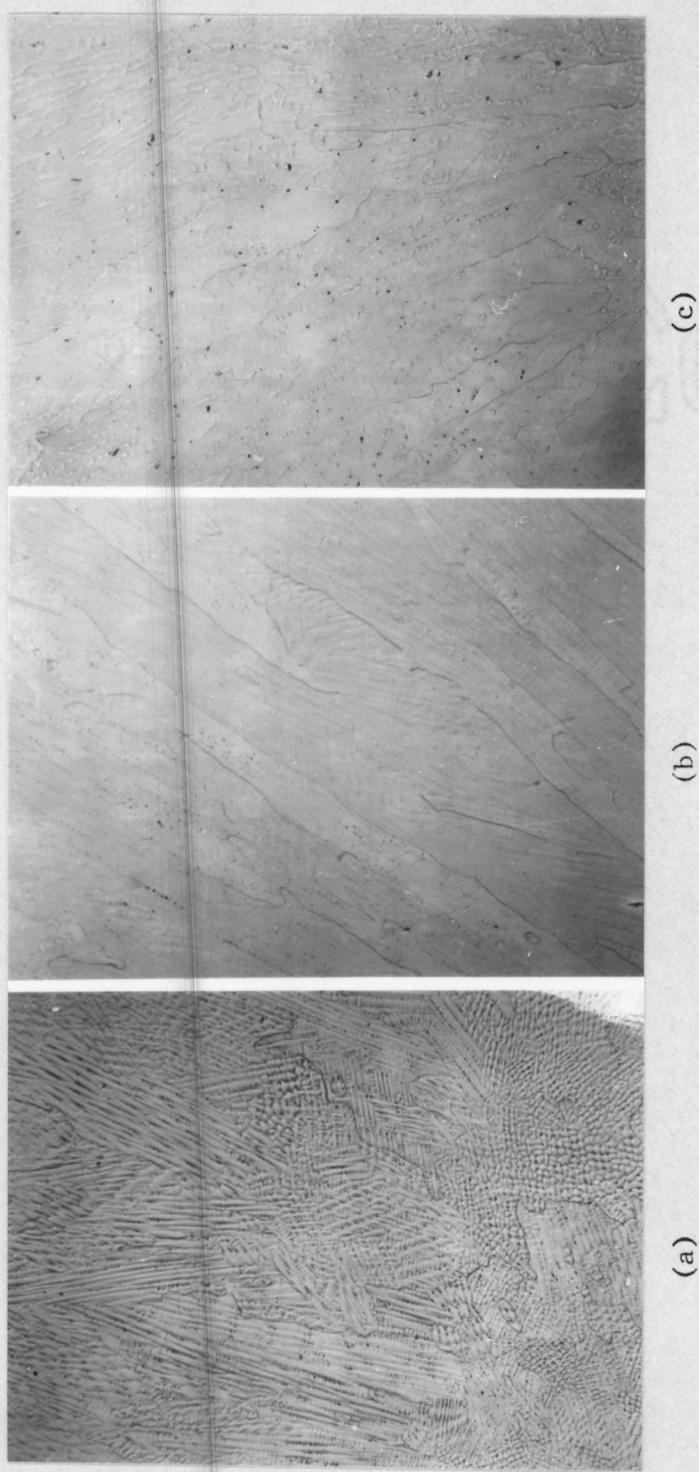


Figure 112. Microstructures of Weld Metal "Al" Thermal Cycled Twice to Peak Temperatures of;
(a) 2200°F, (b) 2300°F, and (c) 2400°F. (OLM, 100X)

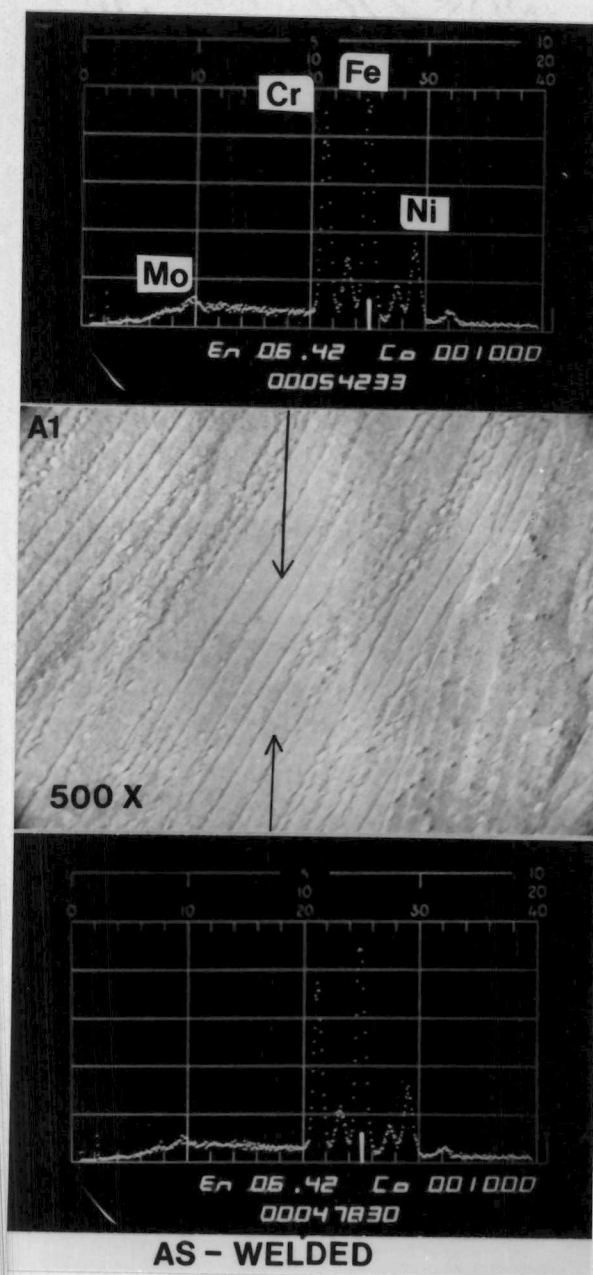


Figure 113. Energy Dispersive X-Ray Analysis of an As-Welded Specimen of "A1". (SEM, 500X)

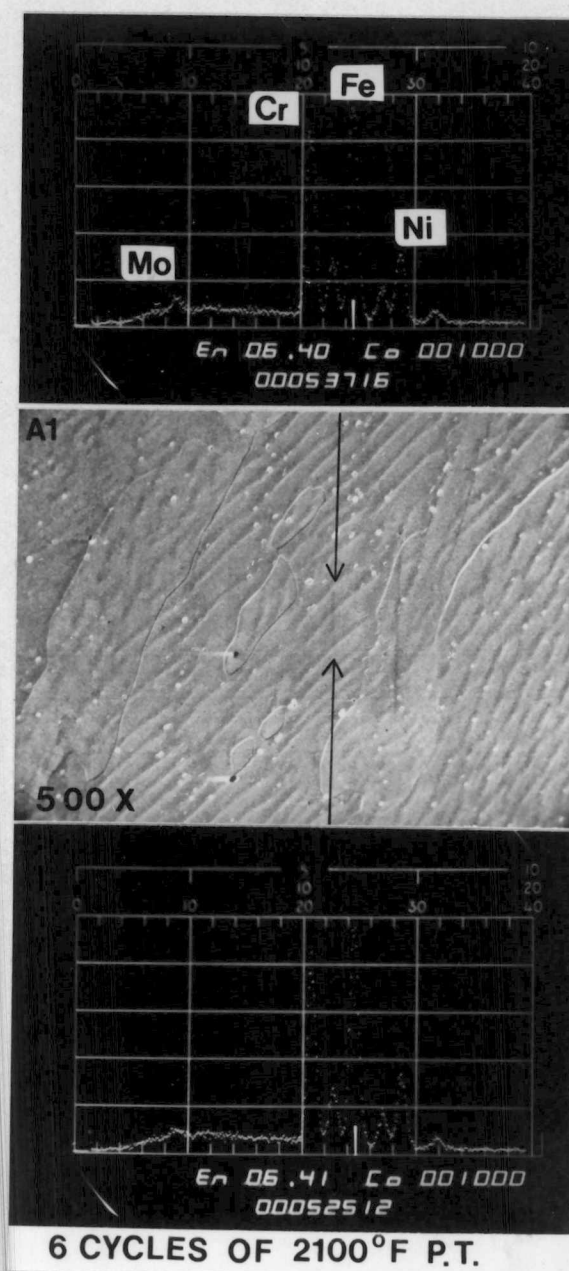


Figure 114. Energy Dispersive X-Ray Analysis of a Specimen of "A1" Thermal Cycled Six Times to 2100°F. (SEM, 500X)

the diffusion mechanism, high peak temperature and more thermal cycles are expected to reduce the segregation of Cr at the subgrain boundaries. As the chromium content decreases in the subgrain boundaries, the "ferrite potential" is also decreased. The "harmful" elements, such as Si, P, and S, are then redistributed in the austenite grains. As the grain growth occurs, solute atoms are "swept-up" by the moving grain boundaries, thus progressively increasing the solute concentration in the grain boundaries. Therefore, the fissuring tendency is increased.

6C. Effect of Alloying Elements on Hot Cracking Sensitivity

The effect of specific alloying elements on the hot-cracking susceptibility of materials has long been of interest to welding specialists and material producers alike. The effect of specific alloying elements has traditionally been established by preparing small laboratory heats various compositions and then subjecting these heats to weldability testing. This method is often costly, requiring a considerable expenditure of time and materials.

A technique for rapidly assessing the effect of a specific alloying element on the hot-cracking sensitivity of strip materials was employed. This technique involved three major steps: (1) making weld pads of desired composition by utilizing the dual-wire feeder device, (2) making strips from the deposited weld pad by rolling, annealing, and cleaning processes, (3) modifying the composition of the fusion zone of the welded strips by adding alloying elements or changing the composition of shielding gas. The details of these steps have been described in section 3.

With the dual-wire feeder technique and the new methods for adding a specific alloying element, a total of 13 weld metals were made. These

13 alloys can be grouped into three categories according to their Cr and Ni contents. Table 21 lists the materials in three groups and gives the chemical analysis of each material. Weld metals in Group I are basically in the range of AISI Type 310. Weld metals in Group II are close to Type 316. Weld metals in Group III are Type 16-8-2 and its modifications.

From the earlier results, it is found that multiple thermal cycling has an adverse effect on the hot cracking resistance. And the fissuring tendency is greater the higher the peak temperature. In order to evaluate the effect of composition and the effect multiple thermal cycling more systematically, a set of testing parameters were selected as shown in Table 22. The thermal cycle, peak temperature, magnitude of strain, and applied load were all carefully calibrated in order to assure the same conditions for all tests. Two specimens were tested at each test condition, thus providing a replication of the entire experiment.

Figure 115 shows the Gleeble Hot Cracking Test results for weld metals in Group I. The number of hot cracks is plotted as a function of chemical composition. Only the weight percentage of selected elements are shown. The major difference in specific elements in the two grades under comparison is underlined. For example, the major difference between weld metal "Al" and "AlSi" is the Si content (0.06% for "Al" and 0.29% for "AlSi") although there is a slight difference in the other elements (Cr, Ni, Mn, and Mo). It can be clearly seen from Figure 115 that multiple thermal cycling has a detrimental effect on the hot cracking resistance for all the fully austenitic weld metals tested at both 2100°F and 2300°F peak temperatures. Thermal cycling to

Table 21. Three Groups of Weld Metals Used for Evaluating the Effect of Specific Alloying Elements With the Gleeble Hot Cracking Test.

Group	Weld Metal	Wt. %							Cr/Ni	
		Cr	Ni	Mn	Mo	Si	C	N		
I (25-20)	A1	24.34	21.49	4.49	2.16	0.06	0.021	0.160	0.0	1.13
	B1	26.56	21.34	1.85	0.09	0.51	0.090	0.064	0.0	1.24
	D1	25.42	21.40	2.46	0.63	0.40	0.079	0.086	0.0	1.19
	A1S1	21.90	20.36	4.09	1.97	0.29	0.021	0.120	0.0	1.08
	II (20-15)	A2	20.49	15.95	9.43	2.57	0.22	0.016	0.078	0.3
B3		18.05	14.99	1.02	4.21	0.59	0.018	0.045	2.5	1.20
D2		19.49	13.87	2.52	1.65	0.31	0.033	0.093	0.0	1.41
D3		20.55	14.10	1.69	0.95	0.46	0.070	0.052	0.0	1.46
A2S1		19.12	15.19	8.06	2.31	0.44	0.022	0.074	0.1	1.26
III (15-10)	B3Ni	18.78	16.97	1.02	3.55	0.53	0.018	0.054	0.0	1.11
	B3N	17.52	14.47	1.01	3.57	0.53	0.015	0.180	0.0	1.21
	B2	16.18	8.81	1.58	1.85	0.43	0.053	0.040	2.2	1.84
	B2N	16.16	8.87	1.60	1.84	0.45	0.039	0.230	0.0	1.82

Table 22. Gleeble Hot Cracking Test Parameters for Evaluation of the Effect of Specific Alloying Elements on Hot Cracking Sensitivity.

Test Code	Peak Temperature	Extension Strain	Prior Thermal Cycles (No Strain Applied)
2100-1	2100°F	9.6%	0
2100-6	2100°F	9.6%	5
2300-1	2300°F	9.6%	0
2300-3	2300°F	9.6%	2

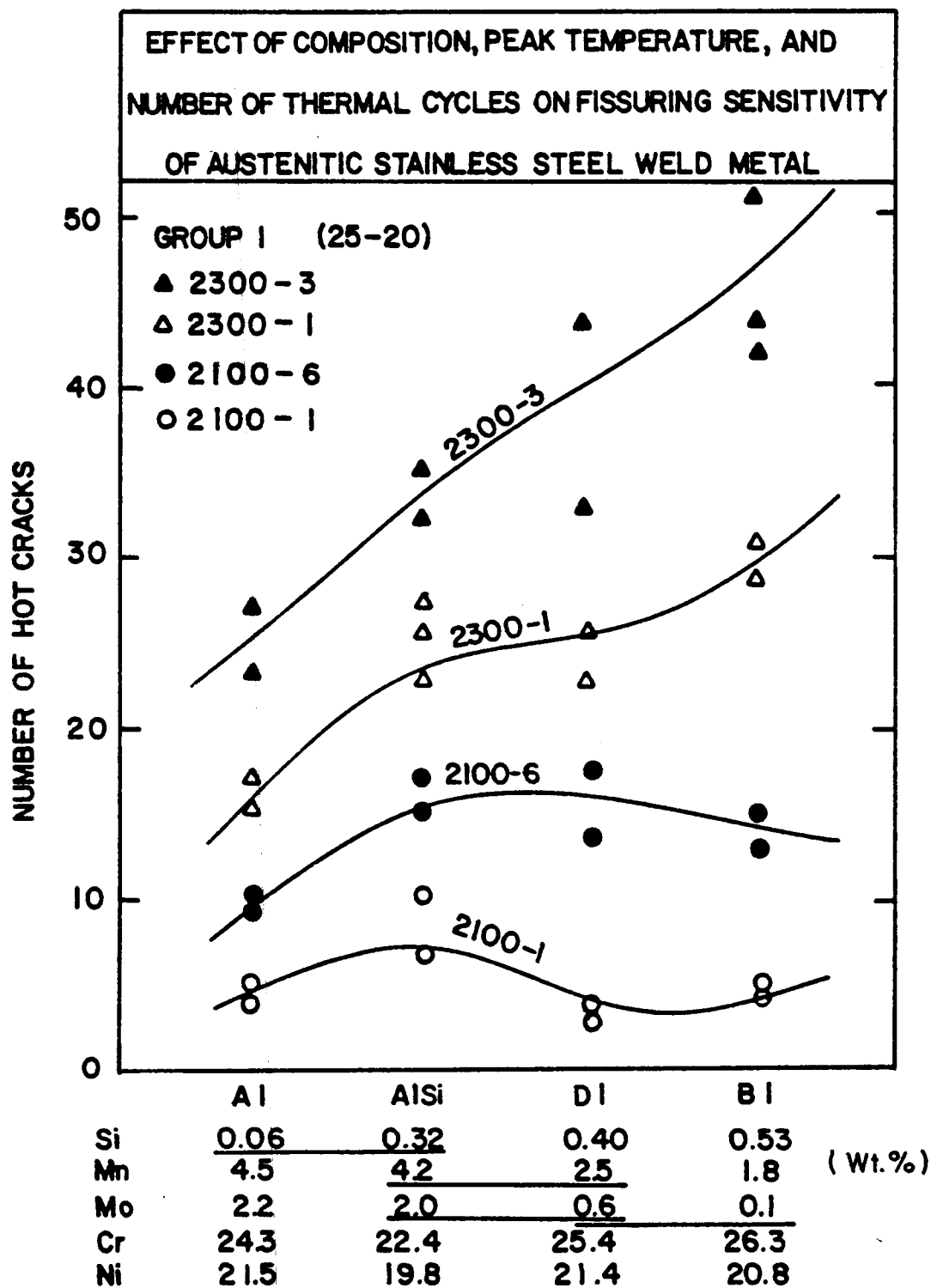


Figure 115. Gleeble Hot Cracking Test Results of Weld Metals in Group I.

higher temperature (2300°F) was more detrimental than the lower temperature (2100°F) thermal cycling.

The "A1" weld metal has the best cracking resistance in this group. The number of cracks increased as the silicon content increased from 0.06% to 0.29% (compare "A1" and "A1Si"). As Mn and Mo contents were reduced along with a slight increase in Si content, the number of hot cracks increased (compare "A1Si", "D1", and "B1"). This shows that Si is detrimental to hot cracking resistance and that Mn plus Mo increases the hot cracking resistance of fully austenitic stainless weld metals.

Figure 116 shows the Gleeble Hot Cracking Test results for weld metals in Group II. Again, multiple thermal cycling increased the number of hot cracks significantly especially at the higher peak temperature (2300°F). The most fissuring resistant weld metal in this group is "A2" which has the highest Mn content. With the increase of Si from 0.22% to 0.45% (compare weld metals "A2" and "A2Si") the number of hot cracks increased considerably. As Mn and Mo contents decreased along with an increase in Si content (compare "A2", "D2", and "D3"), the number of hot cracks increased to a significant extent. In order to separate the individual effect of Mn and Mo on the hot cracking resistance, two "high Mo content" weld metals "B3N" and "B3Ni" were made and tested. Although the Si content was increased slightly from 0.46% to 0.53% and Mn content was reduced from 1.7% to 1.0%, with the increase of Mo from 1.0% to 3.61% (compare "B3Ni" and "D3") the number of hot cracks was reduced. This indicates that Mo alone can effectively reduce the hot cracking tendency. There was no difference in the number of hot cracks as the Mo content was reduced from 3.6% to 2.3% while Mn content was raised from 1.0% to 8.1% (compare "B3Ni" and "A2Si"). Therefore, Mn also can increase the hot

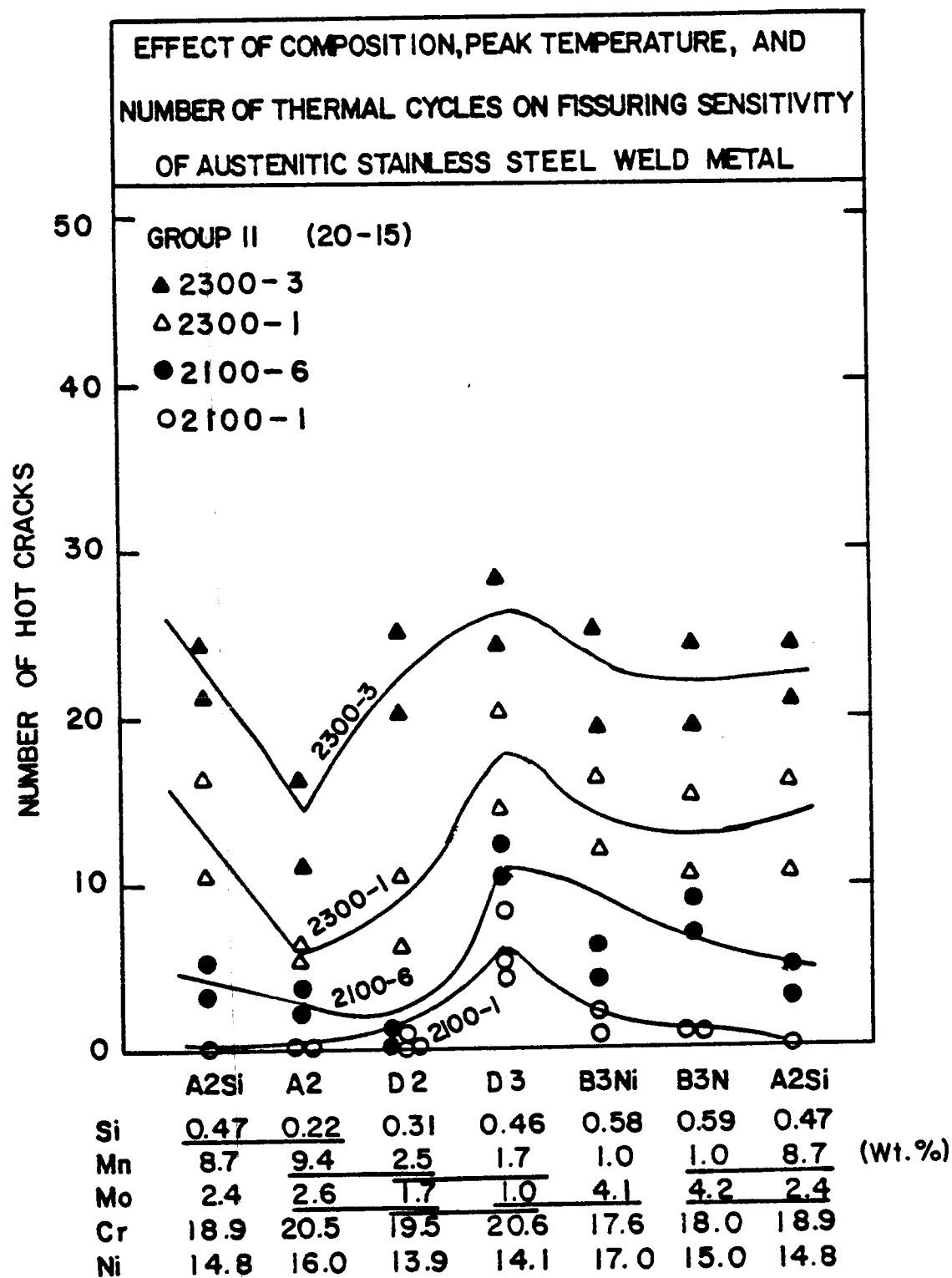


Figure 116. Gleeble Hot Cracking Test Results of Weld Metals in Group II.

cracking resistance although not as effectively as Mo.

Figure 117 shows the Gleeble Hot Cracking Test results for weld metals in Group III. There were no cracks observed for weld metal "B2" even after exposure to two prior thermal cycles to 2300°F peak temperature. Weld metal "B2" is a duplex weld metal with ferrite content of 2.5 FN. This is consistent with the results of other investigators (10). In order to obtain a fully austenitic weld, 5% nitrogen was added in the shielding gas during autogenous GTA welding of additional samples of "B2". The new weld metal was essentially fully austenitic (0.2 FN measured by Magne Gage) and is termed "B2N" weld metal. As the FN was reduced from 2.5 to 0.2, the number of hot cracks increased slightly. By comparing these results with those of the "A1" weld metal in group I and "A2" in group II, it is clearly seen that "B2N" (16-8-2) gives the optimum response of all the fully austenitic weld metals tested.

Comparing Figures 115, 116, and 117, it is found that the weld metals in Group III have the best cracking resistance and the weld metals in Group I exhibit the lowest. This can be graphically presented on the Delong Diagram as shown in Figure 118. The symbols represent four ranges in the number of hot cracks counted for specimens thermal cycled three times to a peak temperature of 2300°F with strain applied during the third cycle. The Cr/Ni ratio is listed in Table 21, page 257, and it is found that as Cr/Ni increases the cracking resistance increases. This is also clearly seen in Figure 118. As Cr/Ni increases and Cr plus Ni decreases the number of hot cracks decreases. This is predictable if the Fe-Cr-Ni ternary phase diagram shown in Figure 119 is examined. Alloy

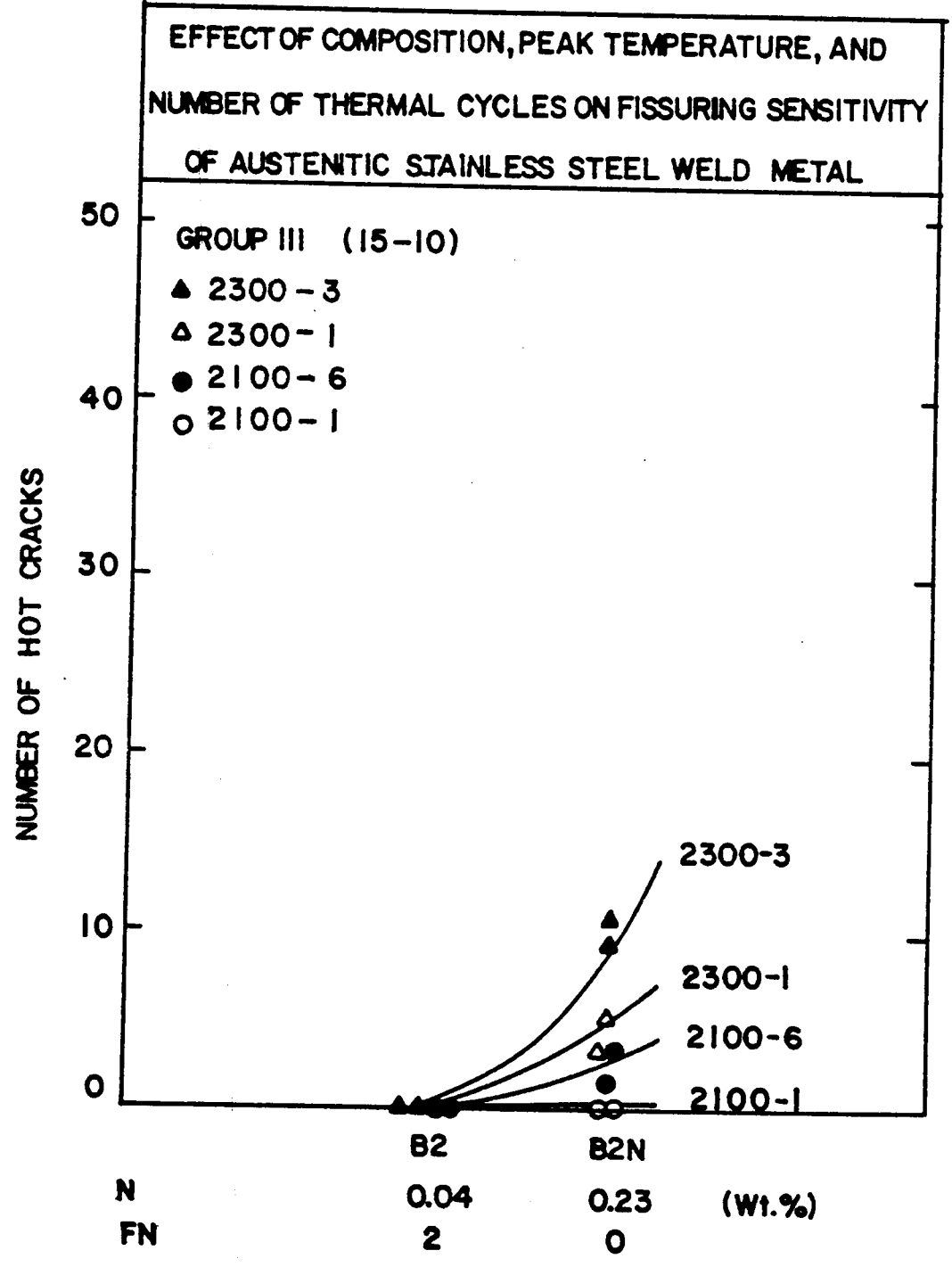


Figure 117. Gleeble Hot Cracking Test Results of Weld Metals in Group III.

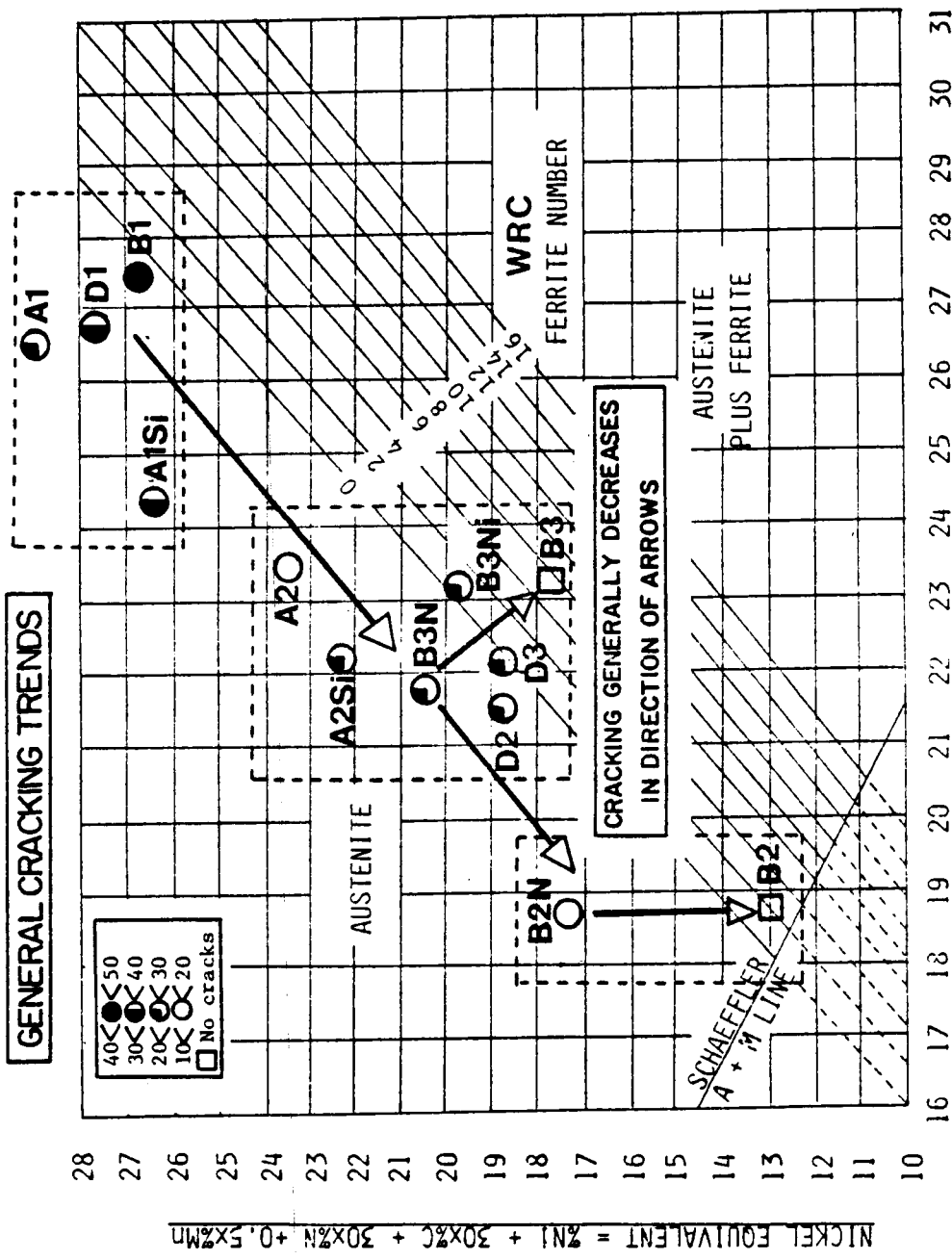


Figure 118. General Cracking Trends of Austenitic Stainless Steel Weld Metals.

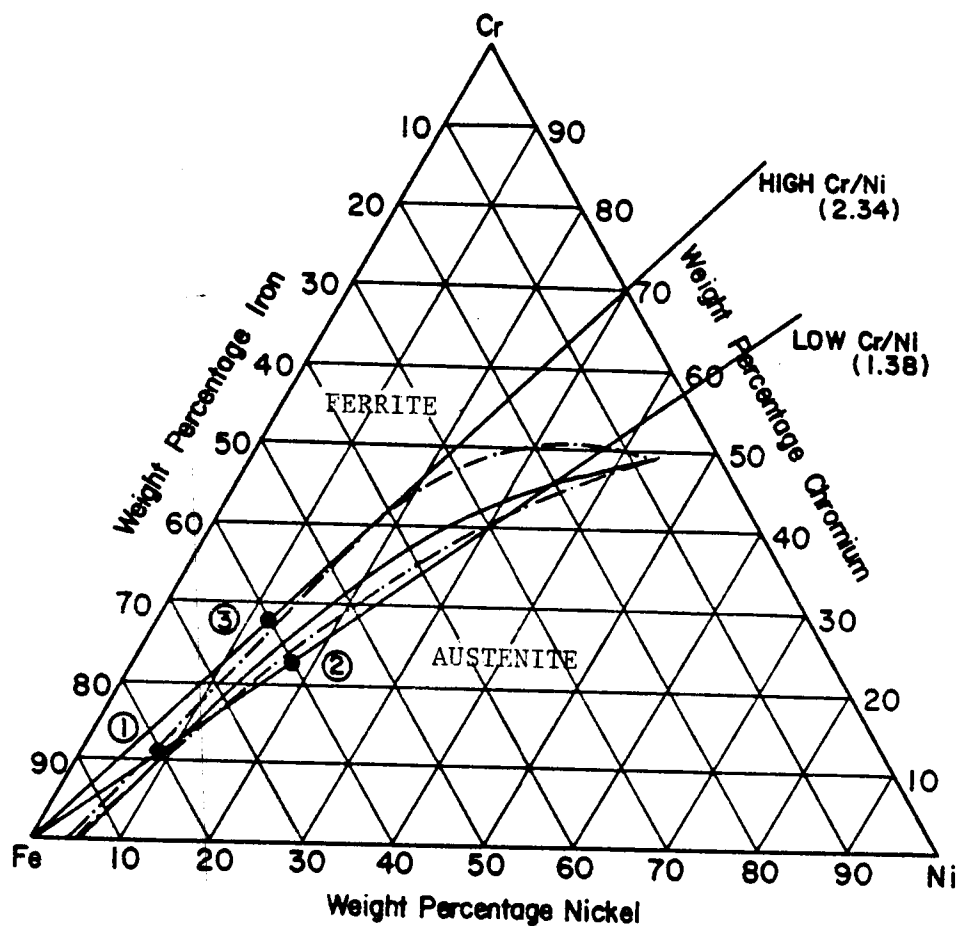


Figure 119. Fe-Cr-Ni Ternary Diagram with Hypothetical Alloys - Showing the Effect of Cr/Ni Ratio and Total Cr + Ni Content.

compositions lying on the Ni-rich side of the liquidus curve solidify as primary austenite and alloy compositions on the Cr-rich side solidify as primary delta ferrite. Masumoto et al. (11) reported that welds solidifying as primary austenite were extremely susceptible to hot cracking whereas welds solidifying as primary delta ferrite did not crack. It is clearly seen from Figure 119 that at the same iron content, the initial phase formed upon solidification changes from austenite to delta ferrite as the Cr/Ni increases. For example, the iron content for alloy 2 and alloy 3 is the same (60%), alloy 3 ($\text{Cr/Ni} = 2.34$) solidifies as primary delta ferrite while alloy 2 ($\text{Cr/Ni} = 1.38$) solidifies as primary austenite. With the same Cr/Ni ratio, welds with low total alloy content (sum of Cr and Ni) are favored because ferrite is the initial phase solidified (compare alloy 1 and alloy 2). When the weld metal that solidified as primary ferrite is reheated to a temperature above the γ -solvus temperature new ferrite precipitates at the grain boundaries and absorbs the "harmful" elements such as Si, P, and S (ferrite has greater solubility) and thus the fissuring sensitivity is lower. When the weld metal that solidified as primary austenite is reheated, elemental migration and grain growth become more extensive due the absence of ferrite. Thus, the fissuring tendency is higher.

The summary of the effect of individual alloying element on the hot cracking resistance is shown in Figure 120. It is clear that ferrite-containing austenitic weld metals are less prone to hot cracking than fully austenitic welds. In the fully austenitic welds, Mn and Mo are found to be favored as alloying elements for preventing hot cracking. Silicon decreases the hot cracking resistance significantly. Nitrogen can reduce the FN and increases the hot cracking slightly.

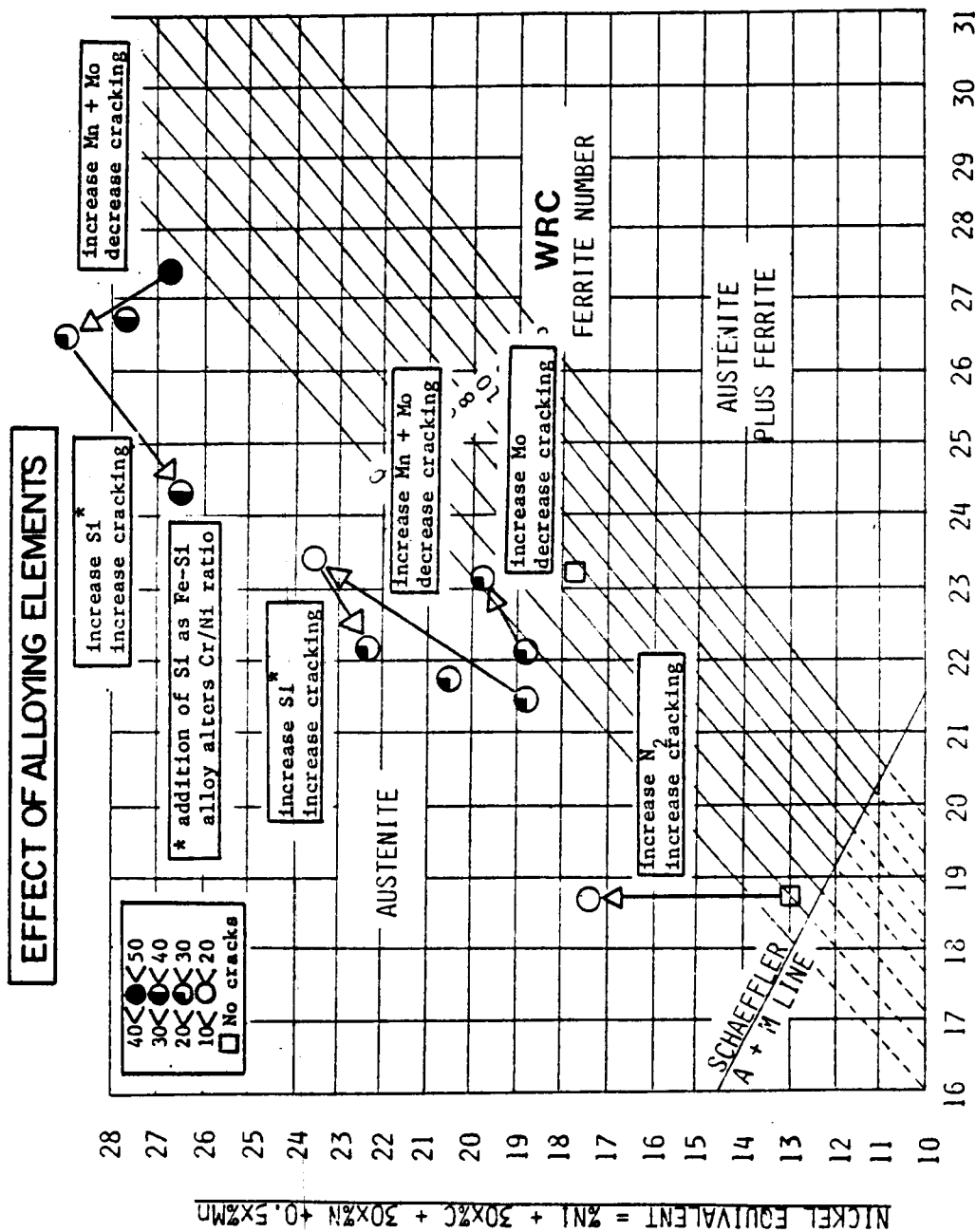


Figure 120. Effect of Alloying Elements on the Hot Cracking Behavior of Austenitic Stainless Steel Weld Metals.

In order to more clearly define the cause of hot cracking and investigate the effect of individual alloying element on the hot cracking resistance, an intensive SEM, EDAX, and optical metallographic study was performed. Figure 121, a light micrograph showing typical fissures in weld metal "D3" (which was thermal cycled three times and strained during the third cycle with a peak temperature of 2300°F), illustrates that fissures are primarily located along grain boundaries. Adding nickel or nitrogen to a duplex weld metal changes the microstructure (at room temperature) from duplex type to a fully austenitic type and hot cracking resistance is significantly reduced. This is illustrated in Figure 122. Both specimens were tested with the same conditions; thermal cycled three times to a peak temperature of 2300°F with strain applied during the third cycle. No fissures were observed in the weld metal "B3" (Figure 122a) which had ferrite content of 2.5 in the as-welded condition. By adding Ni, the alloy becomes fully austenitic (new alloy "B3Ni" as shown in Figure 122b). The number of fissures observed in the metallographic sample increased rapidly.

Examination of other metallographic specimens revealed that there is no difference in the microstructure of the fully austenitic weld metals tested under the same conditions. No second phase (carbide, for example) or inclusions (such as sulphide, phosphide, silicate, etc.) were noted in the microstructures (refer to Figures 121 and 122). However, the number of fissures changed dramatically as the composition varied. This is probably because the alloying elements control the degree of potential segregation as indicated above especially the effects of Mn and Mo.

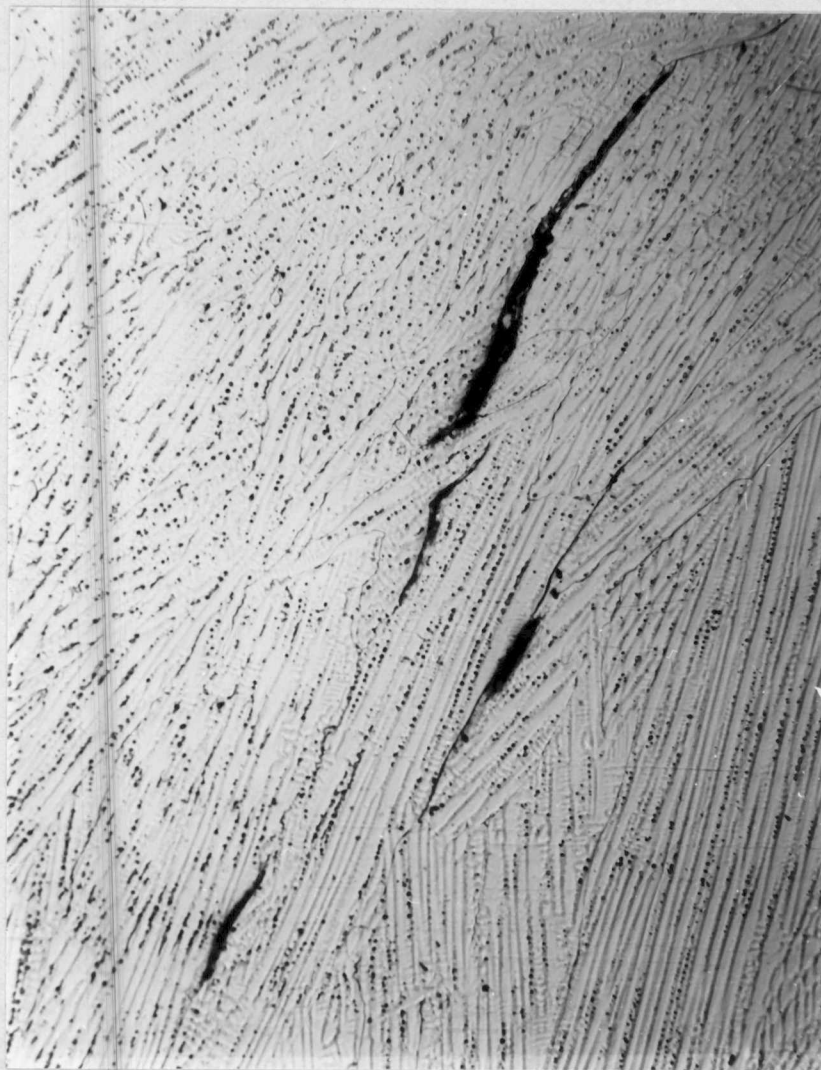


Figure 121. Microstructure of Weld Metal "D3" Tested to 2300°F with Three Thermal Cycles. (OLM, 100X)

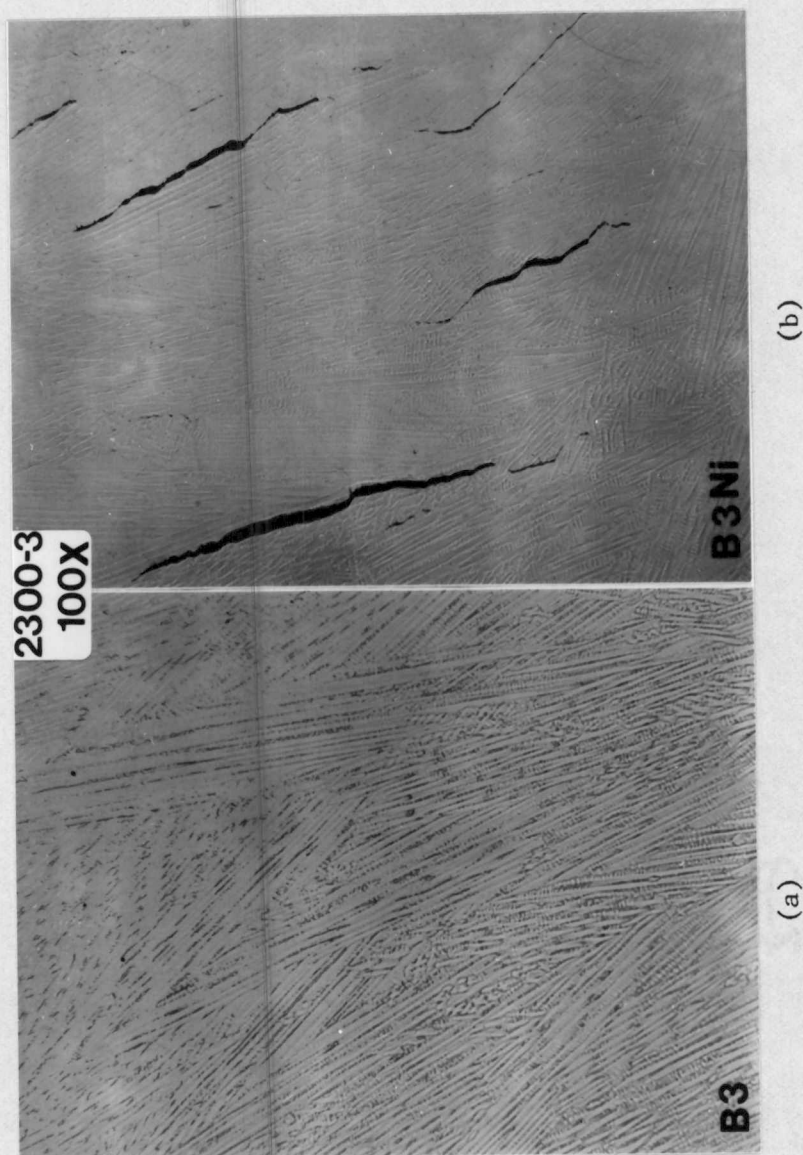


Figure 122. Microstructures of Weld Metals "B3" and "B3Ni" Tested to 2300°F with Three Thermal Cycles. (OLM, 100X)
 (a) Weld Metal "B3". (b) Weld Metal "B3Ni".

In order to study the fracture morphology and elemental distribution on the fissure surfaces, specimens containing fissures were fractured at room temperature. The schematic method of specimen preparation for the fractographic studies is shown in Figure 123. The specimen was notched on both sides of a fissure and an impact tensile load was applied at room temperature. The specimens were cleaned ultrasonically and examined in the SEM. Figure 124 shows the fractured surface of a specimen of weld metal "B1". Area A is a fissure surface. Area B is a ductile overload region, and area C is an as-welded surface. These areas are shown in Figure 124 at higher magnification. Liquid film evidence is clearly seen on the fissure surface as shown in Figure 124b. Figure 124d shows fine dimples and indicates ductile behavior in the tensile overload region. A typical as-welded surface appearance is shown in Figure 124c, with a cellular dendritic type of solidification structure clearly seen. EDAX was employed on all three surfaces to detect any elemental differences. Figure 125 shows the EDAX results on the fissure surface and the overload region. A higher Si content was found on the fissure surface than on the overload region, indicating that a silicate film may have formed during thermal cycling influencing fissuring as sufficient strain was induced. Similar results are also obtained from the fractured surface of weld metal "AlSi" as shown in Figure 126. Again, a greater amount of Si was evident on the fissure surface as compared to the overload region. This implies that segregation of Si may occur in the solidifying weld metal during the initial deposition and that the segregation of Si at the grain boundary may be enhanced upon reheating. In addition, as the grain growth occurs due to reheating, Si atoms may be "swept-up" by the moving grain boundaries, thus progressively increasing the Si concentration in

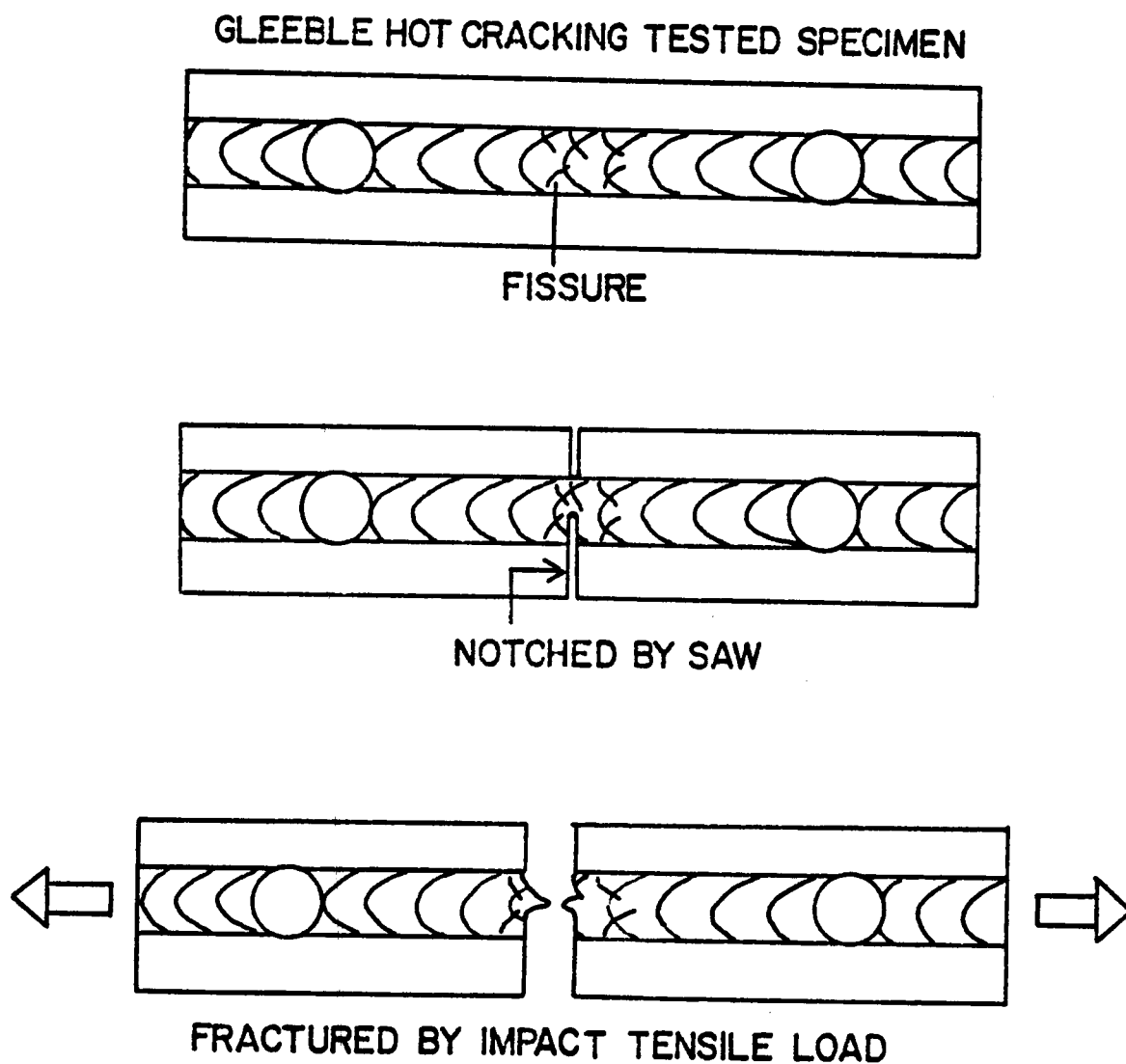


Figure 123. Schematic Representation of Specimen Preparation for Fractography.

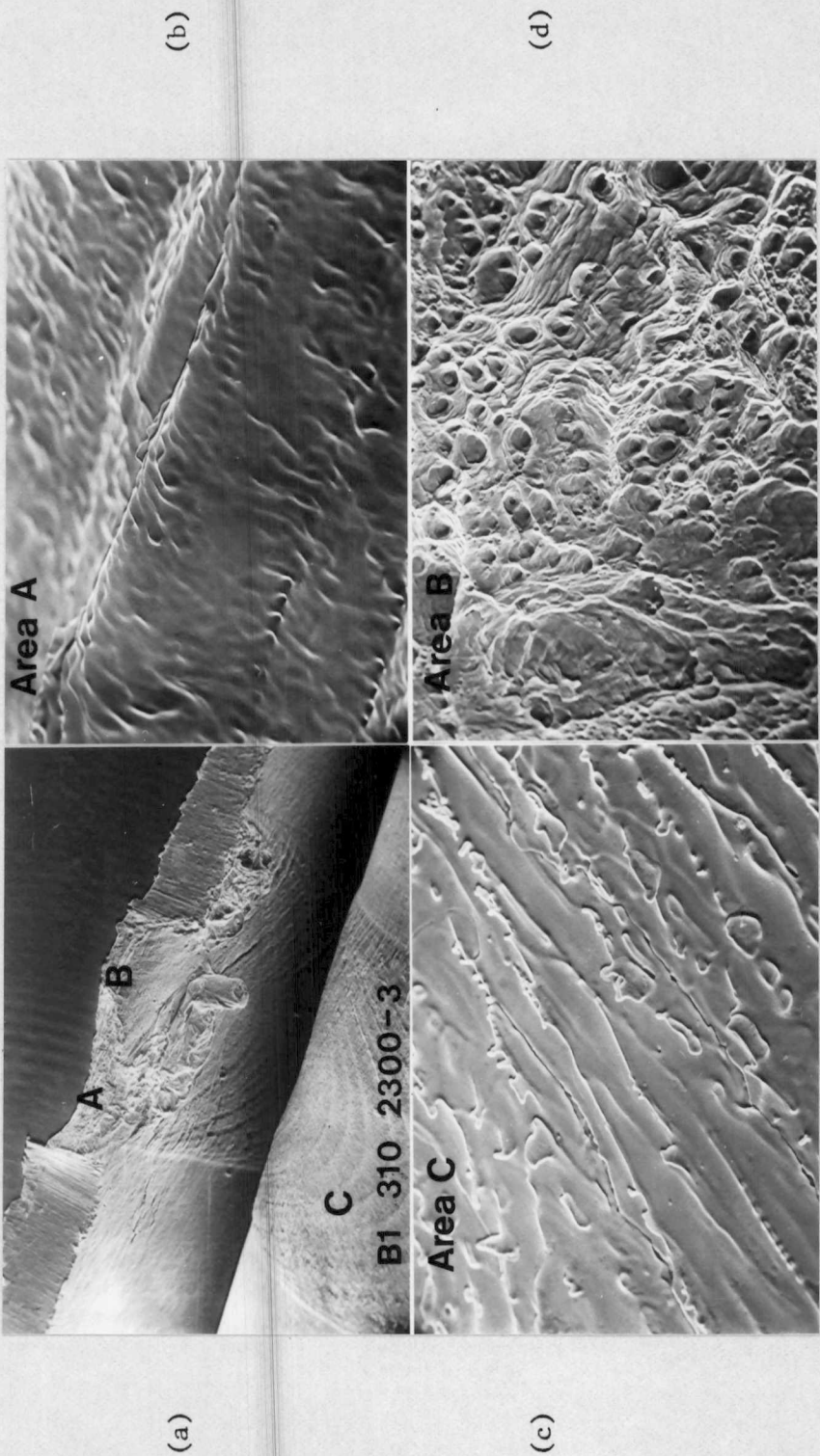


Figure 124. Fracture Surface of Weld Metal "B1" Tested to 2300°F with Three Thermal Cycles.
(a) General View of Fractured Surface. (SEM, 20X)
(b) High Magnification of Area A. (SEM, 2000X)
(c) High Magnification of Area C. (SEM, 1000X)
(d) High Magnification of Area B. (SEM, 2000X)

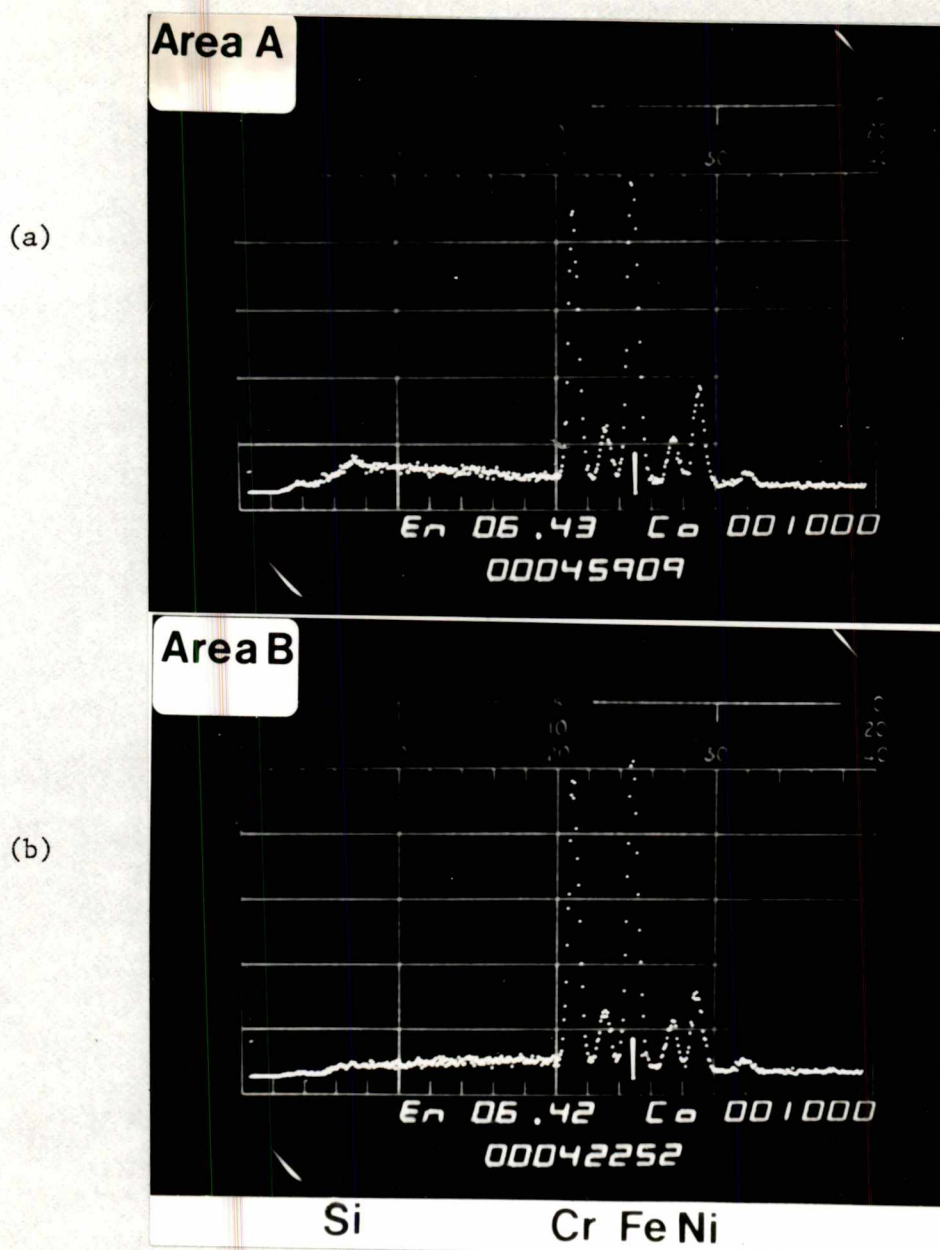


Figure 125. Energy Dispersive X-Ray Analysis Results of The Fractured Surface of "B1".
(a) Area A, Fissure Surface.
(b) Area B, Overload Region.

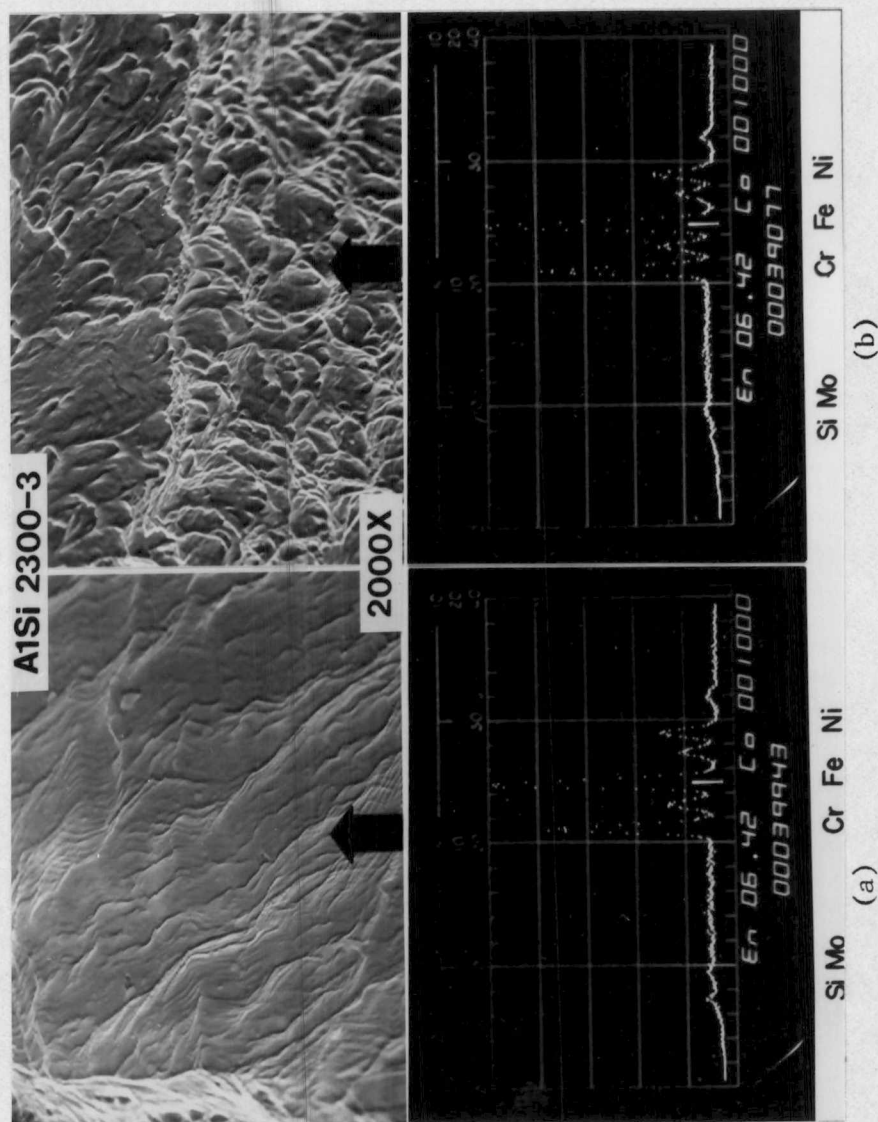


Figure 126. Fractured Surface of Weld Metal "AlSi" Tested to 2300°F with Three Thermal Cycles and Their EDAX Results.
 (a) Fissure Surface. (SEM, 2000X)
 (b) Overload Region. (SEM, 2000X)

the boundaries. The exact composition of the liquid film can not be determined because the SEM fluoresces the material to a considerable depth thus precluding elemental characterization of the remnants of the liquid film alone. Auger electron spectroscopy is perhaps an alternative approach to the problem.

EDAX was also employed on the fractured surface of the high Mo containing weld metal "B3Ni". Figure 127 shows the fractured surface of weld metal "B3Ni" thermal cycled three times to a peak temperature of 2300°F with strain applied during the third cycle. High magnification micrographs of areas A, B, and C are shown in Figure 128. Liquid film evidence on the fissure surface (area A) can be seen in Figure 128a. Fine dimples are shown in the overload region (B). Liquation evidence is observed along the crack surface in area C. The EDAX results again shown that the Si content on the fissure surface is higher than that found in the overload region (see Figure 129).

The beneficial effect of adding Mn and Mo in preventing hot crack-ing in fully austenitic stainless steel weld metal can be expected in regard to their stablization of ferrite at high temperature. Many investigators (87-89) have found that Mn is a ferrite stablizer when the Mn content is over about 4%. The two phase $\gamma + \delta$ region (in Fe-Cr-Ni system) becomes wider as the Mn and/or Mo content increases. Also, the γ -solvus temperature decreases for the alloys with higher Mn and/or Mo content. As weld metal of higher Mn and/or Mo content ($> 4\%$) is reheated to a temperature above γ -solvus, ferrite forms from austenite and, since the solubility of Si is higher in the ferrite than in the austenite, the degree of segregation of Si at the austenite grain boundaries is

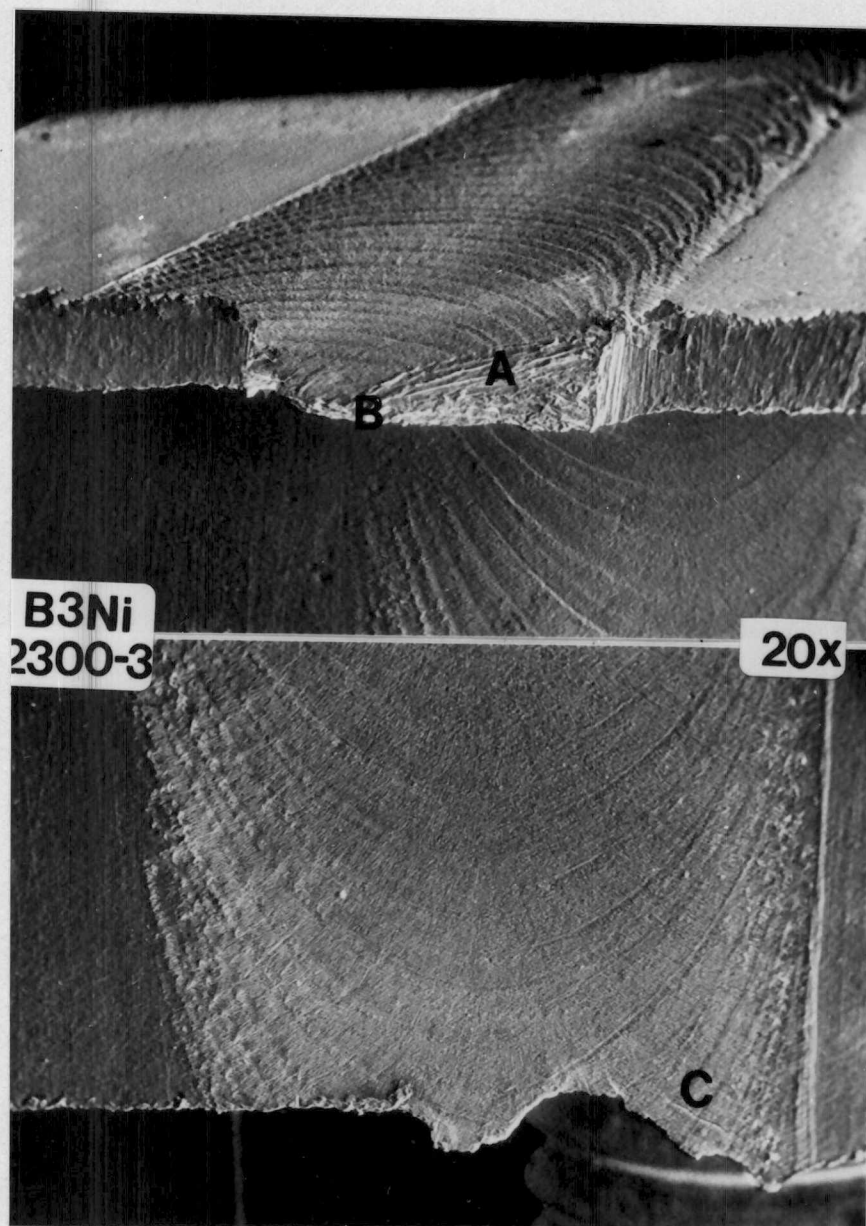


Figure 127. General View of Fractured Surface of Weld Metal "B3Ni"
Tested to 2300°F with Three Thermal Cycles. (SEM, 20X)

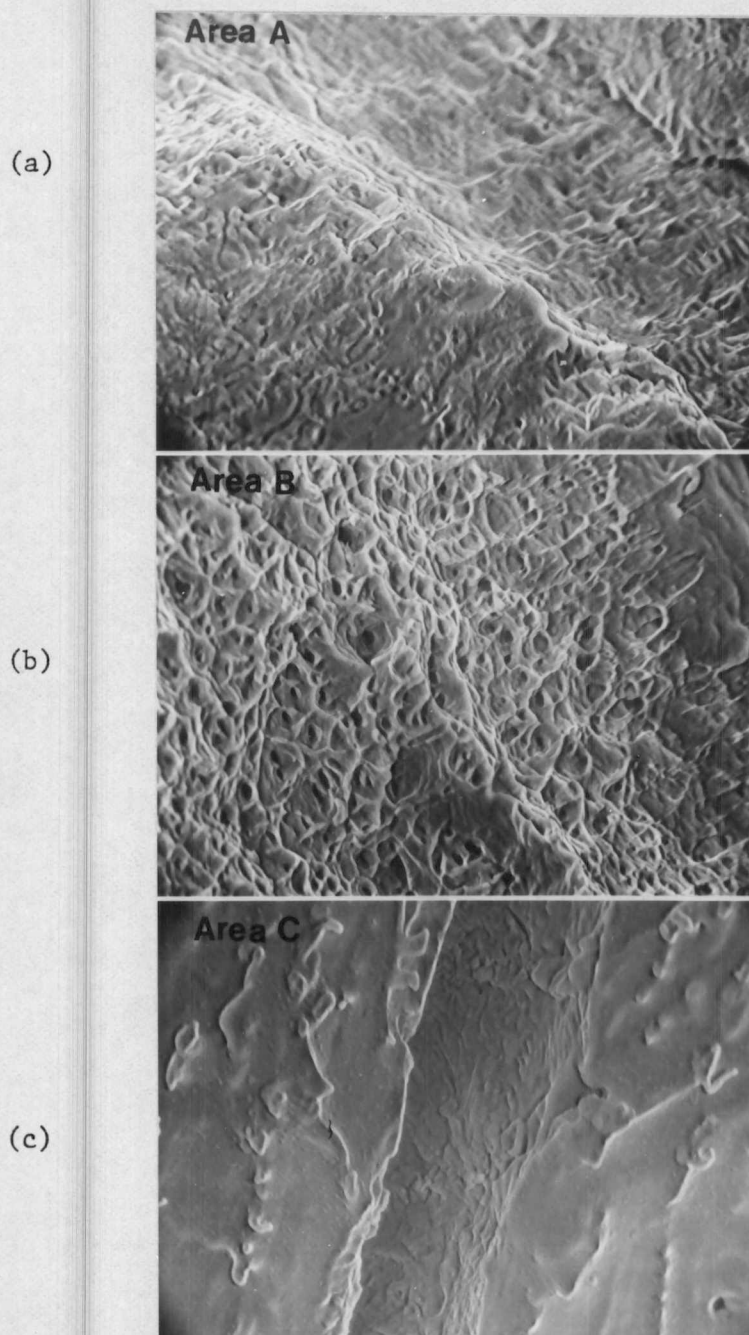


Figure 128. High Magnification of Areas A, B, and C in Figure 127.
(SEM, 2000X)

- (a) Area A, Fissure Surface.
- (b) Area B, Overload Region.
- (c) Area C, Outer Surface.

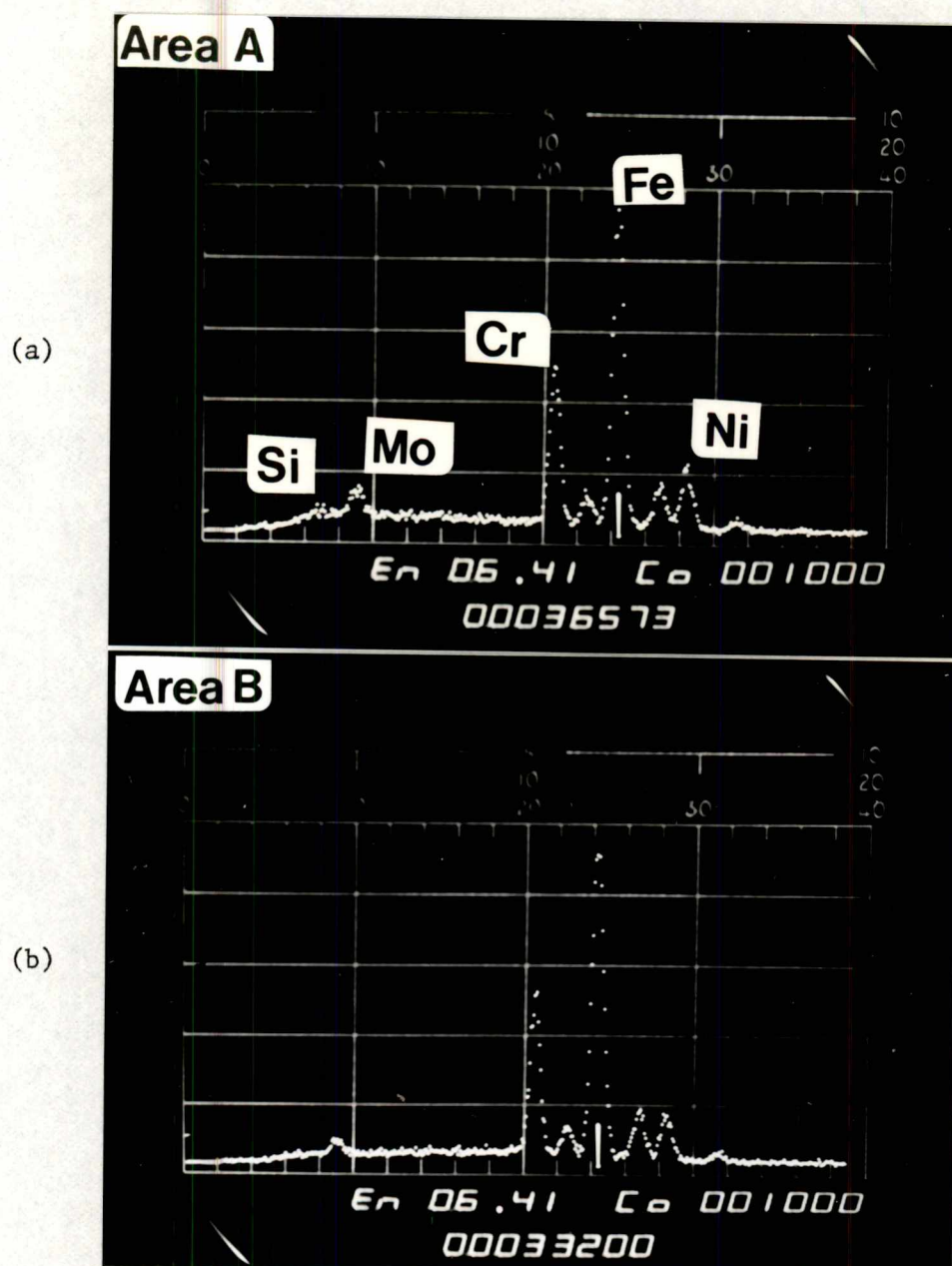


Figure 129. EDAX Results of Fractured Surface of Weld Metal "B3Ni" Tested to 2300°F with Three Thermal Cycles.
(a) Area A, Fissure Surface.
(b) Area B, Overload Region.

reduced and thus chance of hot cracking is also reduced. For example, alloy "A1" (4.5% Mn, 2.2% Mo) has better hot cracking resistance than alloy "B1" (2.0% Mn, 0.1% Mo). Alloy "A2" (9.4% Mn, 2.6% Mo) shows better hot cracking resistance than the alloy "A1". Alloy "A2" was predicted (using Delong Diagram) to be fully austenitic. However, with a rather high Mn content, the microstructure becomes duplex at room temperature (0.5 FN). As this weld metal (A2) is reheated above the γ -solvus temperature, more delta ferrite (> 0.5 FN) is expected to form in contrast to low Mn welds. The ferrite formed at high temperature aids in reducing the level of "harmful" elements (S, P, and Si) at the austenite grain boundaries. Therefore, the fissuring tendency is decreased.

The estimated FN of the as-welded "A2", actually using both Delong diagram and Schaeffler diagram, is zero (about -8 FN on Delong's scale). The 0.5 value obtained from Ferrite Number measurement and the metallographic observation of ferrite in the weld metal was totally contrary to the predicted value. From Table 6, page 79, it is noted that Mn content in "A2" was higher than those commonly found in welding consumables. A similar problem was also encountered in regard to ferrite calculation by other investigators on high-Mn "fully" austenitic stainless steel weld metals (81-84). It was noted and later confirmed that Hull's modification of the diagrams more accurately predicts the residual ferrite in high-Mn alloys. Calculations using Hull's criteria give values of delta ferrite which more closely agree with the measured values for weld metal "A2" (about -1 FN on Hull's scale).

From the results obtained in the Gleeble Hot Cracking tests, it was found that ferrite-containing austenitic stainless steel weld metals are less prone to hot cracking than fully austenitic welds. The major

reason for the beneficial effect on reducing hot cracking is that delta ferrite has a greater solubility than austenite for certain elements, such as P, S, and Si. In the fully austenitic welds, Mn and Mo are found to be favored as alloying elements for preventing hot cracking. Silicon decreases the hot cracking resistance significantly and enhanced Si levels were always found to be associated with grain boundary cracking and liquated fissure surfaces of fully austenitic weld metals.

Multiple thermal cycles imposed on weld metal HAZ can significantly increase the hot cracking tendency of fully austenitic stainless steel weld metals. This is consistent with the results from the hot ductility test as discussed in the previous section. The cause of hot cracking is the enhanced segregation of Si at the grain boundaries. Since the diffusion is temperature and time dependent, more Si can diffuse into the grain boundaries under multiple thermal cycle conditions. With no restrictions to growth due to the absence of ferrite, the austenite grain size also becomes appreciable. Thus, the concentration of Si and other segregation elements per unit surface area is greater. When the Si content at the grain boundaries (which act as diffusion sinks) reaches a certain level, a low-melting silicate forms. The silicon level appears to be about the same on the fractured surface regardless of number of thermal cycles experienced. Since the number of hot cracks increases significantly with multiple cycling, the degraded grain boundary area apparently increases with an increase in the number of thermal cycles.

Using the above information and the results on the effect of multiple cycles on the ferrite content, a new theory on "Reheat Hot Cracking" and a "Hazard HAZ" region is proposed and discussed in the next section.

7. Theory of Reheat Hot Cracking

It is known that austenitic stainless steel weld metals are susceptible to fissuring in the fusion zone of single run beads and in the HAZ of both the base metal and weld metal produced by subsequent beads in multipass welds. Therefore, a complete understanding of the thermal effects on weld metal fissuring is important to the assessment of cracking in multipass or repair welds.

The hot cracking susceptibility of austenitic stainless steels has been the subject of considerable research over the past twenty years. Initial investigations determined that there was a direct correlation between the ferrite content of the as-deposited weld and the hot cracking susceptibility of the alloy. Since that time, numerous investigations concerned with the compositional effect on the solidification hot cracking were conducted using a variety of methods including the Varestraint test (59), the Transvarestraint test (58), and Tigmajig test (148). The overall conclusion was that weld deposits which contained at least five volume percent delta ferrite were relatively immune to hot cracking. Furthermore, alloys which exhibited either a fully austenitic or low ferrite as-welded microstructure were found to be more susceptible to hot cracking.

The "Reheat Hot Cracking" theory proposed in this study concerns the hot cracking which occurs in the weld metal HAZ (induced by subsequent passes during the multipass welding and/or repair welding). This theory

is proposed in two parts: (a) "Reheat Hot Cracking" of ferrite-containing weld metal and (b) "Reheat Hot Cracking" of fully austenitic weld metal.

7A. "Reheat Hot Cracking" of Ferrite-Containing Weld Metal

Based on the experimental results in this investigation, a theory of "Reheat Hot Cracking" is proposed. There is a HAZ region which is separated somewhat from the fusion line of a given bead in which the ferrite content is lower when compared with immediately adjacent HAZ areas. This region exhibits a relatively low ductility and fissures occur if the strain imposed upon the weld exceeds the strain tolerance in this local microstructural region. This region has been termed the "Hazard HAZ".

The cause of the low ductility in this region is the embrittlement of austenite grain boundaries. This embrittlement can be partially explained by utilizing a schematic Fe-Cr-Ni pseudo-phase diagram as shown in Figure 94, page 222. The approximate relative locations of "low ferrite" and "high ferrite" composition are indicated by the vertical lines. For a low ferrite potential material, it can be seen that as the material is reheated to temperatures below the γ -solvus temperature, any residual ferrite should transform to austenite by a solid state diffusion controlled reaction. The higher the temperature the faster the dissolution of ferrite should become because the diffusion rate increases rapidly with the temperature. Ferrite has a greater solubility than austenite for certain elements, such as Si, P, and S which are believed to cause fissuring. As the residual ferrite transforms into austenite during reheating, these "harmful" elements are released and redistributed at

the grain boundaries. As the temperature reaches the γ -solvus temperature, the new delta ferrite which nucleates at the austenite grain boundaries "reabsorbs" the "harmful" elements. Upon cooling, when the temperature drops below the γ -solvus temperature, the ferrite tends to transform back to austenite leaving behind the "harmful" elements trapped at the austenite grain boundaries. (The ferrite/austenite interfaces become austenite grain boundaries upon transformation.) It is in these degraded austenite grain boundaries that the ductility is decreased when compared to the adjacent grain interiors. The region which experiences this phenomenon and exhibits lower ductility is called the "Hazard HAZ" region.

For a high-ferrite potential material, the "Hazard HAZ" region is narrower when compared to a low-ferrite potential material. In order to obtain a high-ferrite potential material, the Cr/Ni ratio is increased at constant iron content. Therefore, the vertical line representing the high ferrite potential composition is shifted to the right and the intersection with the solvus line decreases in temperature. Since the γ -solvus temperature is lower for the "high ferrite" material than the "low-ferrite" material, a narrower "Hazard HAZ" region is predictable for "high-ferrite" material. Thus, the HAZ fissuring tendency for a "high-ferrite" weld is reduced compared to that for the "low-ferrite" weld.

In summary, a narrow weld metal "Hazard HAZ" region separated from the fusion line of a given bead exists in ferrite-containing weld metal. The "Hazard HAZ" region exhibits a relatively low ductility and fissures may occur in this region upon welding under

conditions of sufficient restraint. The extent (width) of the region and the ferrite level therein depend upon the ferrite potential of the welding consumable used and the thermal conditions induced by welding.

7B. "Reheat Hot Cracking" of Fully Austenitic Weld Metal

Since no ferrite exists in fully austenitic stainless steel weld metal, an alternative explanation on the "Reheat Hot Cracking" in fully austenitic materials is proposed.

For fully austenitic stainless steel weld metals, the fissuring propensity becomes significant as the weld metal is reheated to a high temperature. Due to the absence of ferrite in the austenite matrix, the grain boundary segregation of certain "harmful" elements and the grain growth during reheating are the major factors influencing the fissuring propensity.

After welding, in the as-solidified condition, the "harmful" elements, such as Si, P, and S are segregated in interdendritic regions. When the weld is reheated to a high temperature (by a subsequent pass), grain growth occurs. The grain boundaries moving through the material during grain growth have a tendency to "sweep up" and enhance the segregation of the "harmful" elemental species and to trap them in the grain boundary. Since the diffusion is temperature and time dependent, diffusion enhanced by reheating and the occurrence of multiple thermal cycles exacerbates segregation. The segregation of certain elements (P, S, and Si in particular) to grain boundaries is responsible for lower ductility. The grain boundary segregation of the "harmful" elements lowers the "cohesion" between the austenite grains. Therefore, the hot ductility drops significantly and intergranular fissures can occur.

The temperature range in which the ductility is at the lowest value ranges from approximately 1800°F to the solidus temperature. This "Hazard HAZ" region in the fully austenitic welds is generally more extensive than the "Hazard HAZ" region in the ferrite-containing welds. Thus, the probability of fissuring in the weld metal HAZ is higher in the fully austenitic welds than in the ferrite-containing welds.

Fully austenitic stainless steel welds with higher Mn (> 4%) and/or Mo (> 2%) content may have a narrower "Hazard HAZ" region than welds with low Mn and Mo content. This is because welds with higher Mn and Mo content have a lower γ -solvus temperature than welds with low Mn and Mo content. Therefore, ferrite may form in the region near the fusion line (where the reheating temperature is above the γ -solvus temperature) and absorb the "harmful" elements. The ductility is retained at the temperature (above γ -solvus) and the fissuring tendency is decreased. For the fully austenitic welds with low Mn and/or Mo content, the γ -solvus is near the solidus and therefore a wider "Hazard HAZ" region is expected.

In summary, a theory of "Reheat Hot Cracking" is proposed for both ferrite-containing and fully austenitic welds. A region, called "Hazard HAZ", exists near the fusion zone of subsequent bead upon multipass or repair welding. This "Hazard HAZ" region exhibits a relatively low ductility and fissures occur in this region upon welding under conditions of sufficient restraint. The extent of the "Hazard HAZ" region depends primarily on the composition of weld metal as with respect to the Cr/Ni ratio combined with the chromium content and the Mn + Mo levels as well as the P, S, and Si content.

CHAPTER V

CONCLUSIONS

Based on the experimental results in this investigation, a theory of "Reheat Hot Cracking" is proposed for both ferrite-containing and fully austenitic stainless steel welds. A "Hazard HAZ" region was discovered in the weld metal of a previously deposited bead, near but adjacent to the fusion zone. This "Hazard HAZ" region exhibits a relatively low ductility and fissures occur if the strain imposed upon the weld exceeds the strain tolerance of this local microstructural region. The cause of the low ductility in the "Hazard HAZ" region is the embrittlement of austenite grain boundaries. This embrittlement can be partially explained by utilizing a Fe-Cr-Ni pseudo-phase diagram. In the ferrite-containing weld, the ferrite dissolves below γ -solvus and forms above γ -solvus upon heating. Ferrite has a greater solubility than austenite for certain "harmful" elements, such as Si, P, and S which are believed to cause fissuring. As the ferrite transforms into austenite, these "harmful" elements are released and redistributed at the grain boundaries. It is in these degraded austenite grain boundaries that the ductility is decreased when compared to the adjacent grain interiors. For the fully austenitic stainless steel weld metals, the fissuring propensity becomes significant as the weld metal is reheated to a high temperature. Due to the absence of ferrite in the austenite matrix, the grain boundary segregation of certain "harmful" elements and the grain growth during reheating are the major factors influencing the fissuring propensity.

The extent of the "Hazard HAZ" region depends primarily on the composition of weld metals as with respect to the Cr/Ni ratio and Cr+Ni content as well as the Mn+Mo level.

The additional conclusions derived from the experimental work are as follows:

A. Nitrogen Effect on Microstructure:

1. Adding nitrogen to the shielding gas is an economical and effective way to reduce the Ferrite Number in the austenitic welds.
2. By using an autogenous GTA welding technique, it is found that with 2% N₂ in shielding gas the Ferrite Number is reduced to almost zero from 4-5 FN with pure argon as a shielding gas.
3. A high ferrite unmixed zone usually occurs at the filler metal-base metal interface if the base metal has a positive ferrite potential.

B. Techniques for Producing Chemistry-Modified Alloys:

1. The dual-wire feeder technique can be used to produce small-size laboratory-made alloys by the mixing of two chemistry-known wires.
2. The dual-wire feeder technique is reproducible, easy to conduct, inexpensive, and time-saving.
3. With a little modification, a triple-wire feeder can be developed and thus give more flexibility in the choice of composition.
4. The element implant technique used for alloy modification is promising, dependable, and easy to conduct.

5. The nitrogen-addition technique can be used to modify the microstructure of stainless steel weld metal from a duplex type to a fully austenitic type.

C. Fissure Bend Test:

1. The Fissure Bend Test can be used to evaluate SMAW, GTAW, and GMAW deposits and the results closely correlate with fabrication experience.
2. Fissures found in the austenitic weld metals occur primarily along the grain boundaries of reheated weld metal (HAZ produced by subsequent passes).
3. Fissure Bend Test can be effectively used to study the effect of multiple thermal cycles on the HAZ of austenitic stainless steel weld metal.
4. Multiple weld metal HAZ thermal cycles enhance the fissuring propensity in austenitic weld metals.
5. Weaving has an adverse effect on the hot cracking resistance of fully austenitic stainless steel weld metal.
6. The energy input should be low to avoid hot cracking in fully austenitic weld metals.
7. GTA remelting can reduce fissuring sensitivity of previously GMA or SMA deposited weld metal.
8. Mn and Mo are to be strongly favored as alloying elements for preventing hot cracking in the austenitic weld metals.
9. Liquid film evidence was seen in the fissure surfaces which had higher silicon content than the bulk weld metal.

D. Gleeble Hot Ductility Test:

1. The hot Ductility Tests of composite weld showed a good correlation with the fissuring sensitivity of fully austenitic stainless steel weld metal.
2. The hot ductility of the welded material is lower than that of the unwelded material with the same composition.
3. The hot ductility of both welded and unwelded materials tested on-cooling from the zero ductility temperature is lower than that tested on-heating.
4. The cause of hot ductility drop on-heating at high temperature (above 2100°F) is due to the liquation at the grain boundaries.
5. The cause of low hot ductility on-cooling from the zero ductility temperature ($\approx 2400^{\circ}\text{F}$) is due to the liquid film which when solidified is more brittle than the matrix.
6. The Gleeble can be used to study the effect of HAZ thermal cycles on the hot ductility of deposited weld metal.
7. In fully austenitic welds, multiple thermal cycles reduce the hot ductility and enhance the hot cracking tendency.
8. In ferrite-contained welds, the effect of multiple thermal cycling on hot cracking resistance and hot ductility is relatively insignificant.
9. The peak temperature of a thermal cycle, the cooling rate, and the number of cycles has a significant effect on the FN measured at room temperature.

E. Gleeble Hot Cracking Test:

1. The strip specimen used in the Gleeble can effectively evaluate the hot cracking resistance and showed a good correlation with the Fissure Bend Test results.
2. The Gleeble can be used to study the effect of HAZ thermal cycles on hot cracking tendencies and microstructural changes.
3. Multiple HAZ thermal cycles reduce the hot cracking resistance of fully austenitic welds.
4. Strain and peak temperature play major roles in the hot cracking mechanism.
5. General interdendritic segregation of major elements is reduced by thermal cycling to high peak temperatures (segregation to grain boundaries may be enhanced however).
6. Liquid film formation that occurs during the heating portion of a thermal cycle is of prime importance in the hot cracking mechanism.
7. Mn and/or Mo are strongly favored in preventing hot cracking in the austenitic stainless steels.
8. Enhanced Si levels were found to be associated with grain-boundary cracking and liquated fracture of fully austenitic stainless steel weld metal.

LIST OF REFERENCES

LIST OF REFERENCES

1. Spruiell, J. E., "Studies on Microstructure-Property Relationships in Austenitic Stainless Steels," A proposal to ERDA, University of Tennessee (1975).
2. Honeycombe, J. and Gooch, T. G., "Microcracking in Fully Austenitic Stainless Steel Weld Metal," Metal Const. and British Welding J., 2(9), 375-380 (1970).
3. Spond, D. F., "A Study of the Ferrite-Fissuring Relationship in Austenitic Stainless Steel Weld Metals," M.S. Thesis, University of Tennessee (1975).
4. Hull, F. C., "Effect of Delta Ferrite on the Hot Cracking of Stainless Steel," Welding J., 46(9), 399s-409s (1967).
5. DeLong, W. T., "Ferrite in Austenitic Stainless Steel Weld Metal," Welding J., 53(7), 273s-286s (1974).
6. Gooch, T. G., "A Question of Ferrite", The Welding Institute Research Bulletin, 15(7), 183-188 (1974).
7. Willingham, D. C. and Gooch, T. G., "Sigma Formation in Austenitic Stainless Steel Weld Metal," The Welding Institute Research Bulletin, 12(8) (1971).
8. Szumachowski, E. R. and Reid, H. F., "Cryogenic Toughness of SMA Austenitic Stainless Steel Weld Metals: Part 1 - Role of Ferrite," Welding J., 57(11), 325s-333s (1978).
9. Lundin, C. D. and Spond, D. F., "The Nature and Morphology of Fissures in Austenitic Stainless Steel Weld Metal," Welding J., 55(11), 356s-367s (1976).
10. Lundin, C. D., DeLong, W. T., and Spond, D. F., "Ferrite-Fissuring Relationship in Austenitic Stainless Steel Weld Metals," Welding J., 54(8), 241s-246s (1975).
11. Masumoto, I., Tamaki, K., and Katsuna, M., "Hot Cracking of Austenitic Stainless Steel Weld Metal," Journal of JWS, 41(11), 1306-1314 (1972).
12. Honeycombe, J. and Gooch, T. G., "Effect of Manganese on Cracking and Corrosion Behavior of Fully Austenitic Stainless Steel Weld Metals," Metal Const. and British Welding J., 4(12) (1972).

13. Backman, A. and Lundqvist, B., "Properties of a Fully Austenitic Stainless Steel Weld Metal for Severe Corrosion Environments," Welding J., 56(1), 23s-28s (1977).
14. Ogawa, T. and Tsunetomi, E., "Hot Cracking Susceptibility of Austenitic Stainless Steels," Welding J., 61(3), 82s-93s (1982).
15. Arata, Y., Matsuda, F. and Katayama, S., "Solidification Crack Susceptibility in Weld Metals of Fully Austenitic Stainless Steels (Report IV)," Trans. of JWRI, 7(2), 169-172 (1978).
16. Brooks, J. A. and Lambert, F. J., "The Effect of Phosphorus, Sulfur, and Ferrite Content on Weld Cracking of Type 309 Stainless Steel," Welding J., 57(5), 139s-143s (1978).
17. Medovar, B. I., "Effect of Solubility of Alloying Elements Upon Weld Hot Cracking," Avtom. Svarka., 8(2), 79-80 (1955). (Brutcher Trans. No. 3554)
18. Arata, Y., Matsuda, F., and Katayama, S., "Solidification Crack Susceptibility in Weld Metals of Fully Austenitic Stainless Steels (Report II)," Trans. of JWRI, 6(1) (1977).
19. Medovar, B. I. and Latash, Yu. V., "All Austenite Welds Not Suscetible to Hot Cracking," Avtom. Svarka., 10(2), 32-45 (1957).
20. Rapatz, F., "Weldability of Cr-Ni Steels Resistant to Chemical Agents," Berg und Huttenmannische Monatshefte, 95(12), 368-370 (1950)
21. Jordan, D. E., "Some Aspects of Welding Cast Austenitic Steels," 173-179.
22. Castro, R. and De Castenet, J. J., Welding Metallurgy of Stainless and Heat-Resisting Steels, Cambridge Univ. Press (1974).
23. "Wrought Stainless Steel," ASM Metal Handbook, 9th edition, Vol.3, 3-40 (1980).
24. Brickner, K. G., "Stainless Steels for Room and Cryogenic Temperature," Metals Engr. Quarterly, May (1968).
25. Design Guidlines for the Selection and Use of Stainless Steel, Committee of Stainless Steel Producers, AISI (1977).
26. Parker, T. D., "Strength of Stainless Steel at Elevated Temperature," Metals Engr. Quarterly, May (1968).
27. Sedriks, A. J., Corrosion of Stainless Steels, John Wiley & Sons, Inc. (1979).

28. Weiss, B. and Stickler, R., "Phase Instabilities During High Temperature Exposure of 316 Austenitic Stainless Steels," Met. Trans., 3, 851-866 (1972).
29. Read, D. T., and Reed, R. P., "Toughness, Fatigue Crack Growth, and Tensile Properties of Three Nitrogen-Strengthened Stainless Steels at Cryogenic Temperature," NBSIR 78-884, 93-154 (1978).
30. Espy, R. H., "Weldability of 21-6-9 Stainless Steel," NBSIR 78-884, 169-177 (1978).
31. Read, D. T., McHenry, H. I., Steinmeyer, P. A., and Thomas, R. D., "Metallurgical Factors Affecting the Toughness of 316L SMA Weldments at Cryogenic Temperatures," Welding J., 59(4), 104s-113s (1980).
32. Szumachowski, E. R. and Reid, H. F., "Cryogenic Toughness of SMA Austenitic Stainless Steel Weld Metals: Part 2 - Role of Nitrogen," Welding J., 58(2), 34s-44s (1979).
33. Linnert, G. E., "Welding Characteristics of Stainless Steels," Metals Engr. Quarterly, Nov., 1-15 (1967)
34. Hemsworth, B., Boniszewski, T., and Eaton, N. F., "Classification and Definition of High Temperature Welding Cracks in Alloys," Metal Const. and British Welding J., 1(2), 5-16 (1969).
35. "Arc Welding of Stainless Steel," ASM Metals Handbook, 8th edition, Vol. 6, 245-276 (1971).
36. Malone, M. O., "Sigma and 885^oF Embrittlement of Chromium-Nickel Stainless Steel weld Metals," Welding J., 46(6), 241s-253s (1967)
37. Poole, L. K., "Sigma - An Unwanted Constituent in Stainless Weld Metal," Metal Progress, June, 108-112 (1954).
38. Medovar, B. I. and Pinchuk, N. I., "Prevention of Embrittlement in Austenitic Type 25-20 Welds," Avtom. Svarka., 10(2), 46-49 (1962).
39. Curran, R. M. and Rankin, A. W., "Welding Type 347 Stainless Steel for 1100^oF Turbine Operation," Welding J., 34(3), 205s-213s (1955).
40. Henry, O. H., Cordovi, M. A., and Fischer, G. J. "Sigma Phase in Austenitic Stainless Steel Weldments," Welding J., 34(2), 75s-81s (1955).
41. Savage, W. F. and Aronson, A. H., "Preferred Orientation in the Fusion Zone," Welding J., 45(2), 85s-89s (1966).
42. Savage, W. F., Lundin, C. D., and Chase, T. F., "Solidification of Fusion Welds in Face-Centered Cubic Metals," Welding J., 47(11), 522s-526s (1968).

43. Savage, W. F., Lundin, C. D., and Aronson, A. H., "Weld metal Solidification Mechanism," Welding J., 44(4), 175s-181s (1965).
44. Savage, W. F., Nippes, E. F., and Erickson, J. S., "Solidification Mechanisms in Fusion Welds," Welding J., 55(8), 213s-221s (1976).
45. Davies, G. J. and Garland, J. G., "Solidification Structures and Properties of Fusion Welds," International Metallurgical Review, Vol. 20, 83-106 (1975).
46. Matsuda, F., Nakagawa, H., Uehara, T., Katayama, S. and Arata, Y., "A New Explanation for Role of Delta-Ferrite Improving Weld Solidification Crack Susceptibility in Austenitic Stainless Steel," Trans. of JWRI, 8(1) (1977).
47. David, S. A., Goodwin, G. M., and Braski, D. N., "Solidification Behavior of Austenitic Stainless Steel Filler Metals," Welding J., 58(11), 330s-336s (1979).
48. Borland, J. C., "Generalized Theory of Super-Solidus Cracking in Welds (and Casting)," British Welding J., 7(8), 508-512 (1960).
49. Borland, J. C. and Younger, R. N., "Some Aspects of Cracking in Welded Cr-Ni Austenitic Steels," British Welding J., 7(1) (1960).
50. Borland, J. C., "Suggested Explanation of Hot Cracking in Mild and Low Alloy Steel Welds," British Welding J., 8(11) (1961).
51. Bochvar, A. A. and Siderskaya, Z. A., "Failure of Castings Caused by Shrinkage Stresses During Crystallization, as Related to Composition," Izvest. Akad., Nauk SSSR, Otd. Tekhn. Nauk, No. 3 (1947) (Brutcher Trans. No. 3806).
52. Pumphrey, W. I. and Jennings, P. H., "A Consideration of the Nature of Brittleness at Temperature Above the Solidus in Castings and Welds in Aluminum Alloys," J. Inst. Met., 75, 235-256 (1948).
53. Medovar, B. I., "On the Nature of Weld Hot Cracking," Avtom. Svarka, 7(4) (1954). (Brutcher Trans. No. 3400)
54. Medovar, B. I., "Effect of Solubility of Alloying Elements Upon Weld Hot Cracking," Avtom. Svarka, 8(2), 79-80 (1955). (Brutcher Trans., No. 3554)
55. Pellini, W. S., "Strain Theory of Hot Tearing," Foundry, 80(11), 125-133 (1952).
56. Apblett, W. R. and Pellini, W. S., "Factors Which Influence Weld Hot Cracking," Welding J., 3(2), 83s-90s (1954).

57. Prokhorov, N. N., "The Technological Strength of Metals While Crystallising During Welding," Svar. Proiz., No. 4, 1-5 (1962).
58. Senda, T., Matsuda, F., Takano, G., Watanabe, K., Kobayashi, T., and Matsuzaka, T., "Fundamental Investigations on Solidification Crack Susceptibility for Weld Metals with Trans-Varestraint Test," Trans. of JWS, 2(2) (1971).
59. Savage, W. F. and Lundin, C. D., "The Varestraint Test," Welding J., 44(10), 433s-442s (1965).
60. Movchan, B. A., "Development, Structure and Some Properties of the Crystallize Boundaries in Cast Metals and Alloys," Izvest. Akad. Nauk SSSR, OTN, Met i Toplivo, 5 (1959). (Brutcher Trans. No. 5663)
61. Savage, W. F. and Krantz, B. M., "An Investigation of Hot Cracking in Hastelloy X," Welding J., 45(1) (1966).
62. Shorshorov, N. K. and Sokolov, Y. V., "The Temperature Range in Which Hot Cracks Develop When Single Phase Nickel Alloys are Fusion Welded," Svar. Proiz., No. 4, 9-11 (1962).
63. King, B. L., "Weld Metal and Heat Affected Zone Hot Cracking in Wrought Austenitic Stainless Steels," BWRA Report C 139/A/2/65, February 1966.
64. Arata, Y., Matsuda, F., and Saruwatari, S., "Varestraint Test for Solidification Crack Susceptibility in Weld Metal of Austenitic Stainless Steels," Trans. of JWRI, 3(1) (1974).
65. Poole, L. K., "The Incidence of Cracking in Weld Type 347 Steels," Welding J., 32(8), 403s-412s (1953).
66. Puzak, P. P., Apblett, W. R., and Pellini, W. S., "Hot Cracking of Stainless Steel Weldments," Welding J., 35(1) 9s-17s (1956).
67. Puzak, P. P. and Rischall, H., "Further Studies on Stainless Steel Hot Cracking," Welding J., 36(2), 57s-61s (1957).
68. Shackleton, D., "A Metallographic Study of Heat Affected Zone Cracks Formed in the Type 304 Stainless Steel Forgings," BWRA Report M/16/67 (1969).
69. Younger, R. N. and Baker, R. G., "Heat Affected Zone Cracking in Welded High-Temperature Austenitic Steels," J. Iron and Steel Institute, 196(10), 188-194 (1960).
70. Tamura, H. and Watanabe, T., "Mechanism of Liquation Cracking in Weld Heat Affected Zone of Austenitic Stainless Steels," Trans. of JWS, 4(1) (1973).

71. Rollason, E. C. and Bystram, M. C., "Hot Cracking of Austenitic Welds with Specific Reference to 18/3/1 Cr-Ni-Nb Alloy," JISI, 169(12) (1951).
72. Cordea, J. N., Evans, R. M., and Martin, D. C., "Investigation to Determine Causes of Fissuring in Stainless Steel and Nickel-Base Alloy Weld Metals," Technical Documentary Report No. ASD-TDR-62-317, Battelle Memorial Institute, June (1962).
73. Haddrill, D. M. and Baker, R. G., "Microcracking in Austenitic Weld Metal," Brit. Welding J., 13(1) (1966).
74. Fredriks, H. and Van Der Toorn, L. J., "Hot Cracking in Austenitic Stainless Steel Weld Deposits," Brit. Welding J., 15(4), 178-182 (1968).
75. Lundin, C. D., "Welding Defects, Their Origin and Effect," Second Interamerican Conference on Materials Technology, Mexico City, Mexico, August (1970).
76. Honeycombe, J. and Gooch, T. G., "Microcracking in Fully Austenitic Stainless Steel Weld Metal," Metal Const. and British Welding J., 2(9), 375-380 (1970).
77. Honeycombe, J. and Gooch, T. G., "Microcracking in Fully Austenitic Stainless Steel Weld Metal," Metal Const. and British Welding J., 7(3), 146-148 (1975).
78. Burghard, H. C., Norris, E. B., Pickett, A. G., and Watson, P. D., "An Investigation of the Properties of Type 316 Stainless Steel Weldments Containing Austenitic Microfissures," SWRI Topical Report No. 1, Project 17-3078, August 15, (1971).
79. Buchhoit, R. D., Jaske, C. E., and Bennet, R. W., "Laboratory Studies of Type 316 Stainless Steel Weld Metal From Maine Yankee Reactor Coolant System," Battelle Final Report, Sept. 17, (1971).
80. Thielsch, H., "Examination of Type 316 Stainless Steel Weld Deposit for Microfissuring and Delta Ferrite Contents for the Oconee Nuclear Power Station, Duke Power Company," Report No. 1057, December 23, (1972).
81. Gooch, T. G. and Honeycombe J., "Welding Variables and Microfissuring in Austenitic Stainless Steel Weld Metal," Welding J., 59(8), 233s-241s (1980).
82. Witherell, C. E., "Welding Stainless Steels for Structures Operating at Liquid Helium Temperature," Welding J., 59(11), 326s-342s (1980).

83. Berggren, R., Cole, N. C., Goodwin, G. M., Stiegler, J. O., Slaughter, G. M., Gray, R. T., and King, R. T., "Structure and Elevated Temperature Properties of Type 308 Stainless Steel Weld Metal with Varying Ferrite Contents," Welding J., 57(6), 167s-174s (1978).
84. Hull, F. C., "Delta Ferrite and Martensite Formation in Stainless Steels," Welding J., 52(5), 193s-203s (1973).
85. Schaeffler, A. L., "Constitution Diagram for Stainless Steel Weld Metal," Metal Progress, 56(11), 680 and 680B (1949).
86. Delong, W. T., Ostrom G. A., and Szumachowsik, E. R., "Measurement and Calculation of Ferrite in Stainless-Steel Weld Metal," Welding J., 35(11), 521s-528s (1956).
87. Brooks, J. A., "Weldability of High N, High Mn Austenitic Stainless Steel," Welding J., 54(6), 189s-195s (1975).
88. Suutala, N., "Effect of Manganese and Nitrogen on the Solidification Mode in Austenitic Stainless Steel Welds," Met. Trans., 13A(12), 2121-2130 (1982).
89. Espy, R. H., "Weldability of Nitrogen-Strengthened Stainless Steels," Welding J., 61(5), 149s-156s (1982).
90. Miller, T. M., "Ferrite-Free Consumables for Welding Austenitic Stainless Steels-AWRA Project Panel Weld Metal," Australian Welding Journal, Jan/Feb., 22-23 (1978).
91. Hull, F. C., "Effect of Alloying Additions on Hot Cracking of Austenitic Chromium-Nickel Stainless Steels," Trans. ASTM, Vol. 60, 667-690 (1960).
92. Dahl, W., Duren, C., and Musch, H., "Effect of Various Alloying Elements on the Hot Cracking Susceptibility of Austenitic Steel During Welding," Stahl Und Eisen, 93(18), 813-822 (1973).
93. Polgary, S., "The Influence of Silicon Content on Cracking in Austenitic Stainless Steel Weld Metal With Particular Reference to 18Cr-8Ni Steel," Svetsaren Vol. 6, 1-2, p8-13 (1970).
94. Nagaoka, T. and Shone, S., "Study on Hot Cracking of Austenitic Stainless Steel in Overlay Welding," Mitsubishi Heavy Industries Technical Review, Sept. (1966).
95. Rabensteiner, G., Tosch, J., and Schabereiter, H., "Hot Cracking Problems in Different Fully Austenitic Weld Metals," Welding J., 62(1), 21s-27s (1983).
96. Astrom, H., Loberg, B., Bengtsson, B., and Easterling, K. E., "Hot Cracking and Micro-Segregation in 18-10 Stainless Steel Welds," Material Science, July, 225-234 (1976).

97. Musch, H., "Causes of Hot Cracking of Welds on Austenitic Steels Containing 16% Cr and 16% Ni," Schweissen und Schneiden, 25(6), p225 (1973).
98. Lippold, J. C. and Savage, W. F., "Solidification of Austenitic Steel Weldments: Part III - The Effect of Solidification Behavior on Hot Cracking Susceptibility," Welding J., 61(12), 388s-396s (1982).
99. Kammer, P. A., Masubuchi, K., and Monroe, R. E., "Cracking in High-Strength Steel Weldments - A Critical Review," DMIC Report 197, Feb. (1964).
100. Linnert, G. E., "Weldability of Austenitic Steels as Affected by Residual Elements," ASTM STP 418, 105-119 (1967).
101. Honeycombe I. and Gooch, T. G., "The Effect of Compositional and Process Variables on Microcracking in Fully Austenitic Stainless Steel Weld Metal," The Welding Institute Research Report M/74/73.
102. Dem'yanenko, G. P. and Kakhovskii, Yu. N., "Effect of the Welding Conditions and the Temperature of the Metal Welded on the Resistance to Hot Cracking of Single-Phase Austenitic Welds," Avt. Svarka, No. 5, 14-17 (1972).
103. Lundin, C. D., Delong, W. T., and Spond, D. F., "The Fissure Bend Test," Welding J., 55(6), 145s-151s (1976).
104. Nippes, E. F., and Savage, W. F., "Development of Specimen Simulating Weld Heat-Affected Zones," Welding J., 28(11) 534s-546s (1949).
105. Stout, R. D. and Doty, W. D., "Weldability of Steels," Welding Research Council, (1978).
106. Nippes, E. F., Savage, W. F., Bastian, B. T., Mason, H. F., and Curran, R. M., "An Investigation of the Hot Ductility of High Temperature Alloys," Welding J., 34(4), 183s-196s (1955).
107. Weiss, B., Grotke, G. E., and Stickler, R., "Physical Metallurgy of Hot Ductility Testing," Welding J., 49(10), 471s-487s (1970).
108. Brooks, J. A., "Effect of Alloy Modifications on HAZ Cracking of A-286 Stainless Steel," Welding J., 53(11) 517s-523s (1974).
109. Duvall, D. S. and Owczarske, W. A., "Further Heat-Affected Zone Studies in Heat-Resistant Nickel Alloys," Welding J., 46(9), 423s-432s (1967).
110. Owczarske, W. A., Duvall, D. S., and Sullivan, C. P., "A Model for Heat Affected Zone Cracking in Ni-Base Superalloys," Welding J., 45(4) 145s-155s (1966).

111. Owczarski, W. A., Duvall, D. S., and Sullivan, C. P., "A Model for Heat Affected Zone Cracking in Ni-Base Superalloys," Welding J., 45(4), 145s-155s (1966).
112. Yeniscavich, W., "A Correlation of Ni-Cr-Fe Alloy Weld Metal Fissuring With Hot Ductility Behavior," Welding J., 45(8), 344s-356s (1966).
113. Solda, H. M. and Mayne, C. R., "Ductility Related to Service Performance of Heavy-Wall Austenitic Pipe Welds," Welding J., 36(3), 141s-150s (1957).
114. Heuschkel, J., "Initial Characteristics of Chromium-Nickel Steel Weld Metals," Welding J., 34(10), 484s-504s (1955).
115. Wellnitz, G. and Schmidtman E., "The Effect of Sulphides on Hot Cracking and Nutch Toughness in the HAZ in the Welding of High Strength Steels," DVS Berichte, No. 32, 177-183 (1974).
116. Yeniscavich, W., "Correlation of Hot Ductility Curves With Cracking During Welding," Proceedings of the Symposium on "Methods of High-Alloy Weldability Evaluation," by WRC (1970).
117. Krantz, B. M., "An Investigation of Hot Cracking in Hastelloy X," Ph.D. Dissertation, R.P.I. (1966).
118. Welding Terms and Definitions, AWS, A30-77 (1977).
119. Lundin, C. D., "The Significance of Weld Discontinuities - A Review of Current Literature," WRC Bulletin 222, Dec. (1976).
120. Harrison, J. D., "Defect Standards in Relation to Design," International Conference on Welding Research. P. 1-9, Sept. 17-21, (1972).
121. Honeycombe, J. and Gooch, T. G., "Effects of Microcracks on Mechanical Properties of Austenitic Stainless Steel Weld Metals," Welding Research Abroad, Oct., 46-53 (1973).
122. "An Investigation of the Properties of Type 316 Stainless Steel Weldments Containing Austenitic Microfissures," Topical Report #1, SWRI Project, 17-3078, Aug. 15 (1971).
123. Thielsch, H., "Examination of Type 316 Stainless Steel Weld Deposits for Microfissuring and Delta Ferrite Contents for the Oconee Nuclear Power Station," Duke Power Company, Dec. 23, (1972).
124. Grudzinskii, B. V., "Influence of Lack of Fusion on the Strength of Welded Joints in Steel Kh18N10T at Cryogenic Temperature," Svar. Proiz. (12), 31-33 (1973).

125. Kudryavtsev, O. A. and Yushchenko, K. A., "The Mechanical Properties of High-Alloy Welds With Defects at Cryogenic Temperatures," Avt. Svarka, No. 12, 7-12 (1978).
126. Chipperfield, C. C., "A Toughness and Defect Size Assessment of Welded Steel Components," I Mech E Conference Publications 1978-10, London, p125-138 (1980).
127. Kazennov, Yu, I., "The Effect of Reactor Irradiation on the Mechanical Properties of Austenitic Welded Joints with Various Amounts of Ferrite," Welding Production, Vol. 25, No. 3 (1978).
128. Zemzin, I. V. N. and Stanjukovic, A., "Evaluation of the Brittle Failure Susceptibility of Austenitic Steel Welds at High Temperatures," Welding Research Abroad, 10(9), Nov., p66-71, (1964).
129. Christoffel, R. T., "Notch Sensitivity of the HAZ in Typr 316 Material," Welding J., 42(1) (1963).
130. Goodwin, G. M., Cole, N. C., and Slaughter, G. M., "A Study of the Ferrite Morphology in Austenitic Stainless Steels," Welding J., 51(9), 425s-429s (1972).
131. Thomas, R. G., "The Effect of Delta Ferrite on the Creep Rupture Properties of Austenitic Weld Metals," Welding J., 58(3), 81s-86s (1978).
132. Haigh, J. R., "Creep Crack Growth in a Type 316 Stainless Steel Weld Metal," Welding J., 57(5), 149s-153s (1977).
133. Yeniscavich, W., "The Effect of Fissures on the Fatigue Strength of Ni-Cr-Fe Alloys," Welding J., 45(3), 111s-123s (1966).
134. Buchheit, R. D., Jaske, C. E., and Bennet, R. W., "Final Report on Laboratory Studies of Type 316 Stainless Weld Metal from Maine Yankee Reactor Coolant System to Maine Yankee Atomic Power Company," Battlle Report, Sept., (1971).
135. McNaughton, W. P., Egan, G. R., and Cipolla, R. C., "Significance of Centerline Lack-of-Penetration Defects in Double-Seam Welded SA-312 Stainless Steel Pipe," APTECH Engineering Service Report, May, (1980).
136. Cipolla, R. C., Grover, J. L., Beaupre, G. S., and McNaughton, W. P., "Signification of Centerline Lack-of-Penetration Defects in Double Seam Welded SA-312 Stainless Steel Pipe," APTECH Engineering Service Report, May (1980).
137. James, E. A. and Jubb, J. E., "The Fatigue Strength of Fillet Welded Joints in 18-8 Type Austenitic Stainless Steels at Ambient and Low Temperatures," Aust. Weld J., 16(7), (1982).

138. Baeslack III, W. A. Duquette, D. T., and Savage, W. F., "The Effect of Ferrite Content on Stress Corrosion Cracking in Duplex Stainless Steel Weld Metals at Room Temperature," Corrosion, 35(2), (1979).
139. Baeslack III, W. A., Savage, W. F., and Duquette, D. T., "Effect of Nitrogen on the Microstructure and Stress Corrosion Cracking of Stainless Steel Weld Metals," Welding J., 59(3), 83s-90s (1979)
140. Gooch, T. G., "Welding and Corrosion Resistance of Austenitic Stainless Steels," IIW Annual Assembly, Public Session, Subdivision No.1. (1975).
141. AWS A4.2-74, "Standard Procedure for Calibrating Magnetic Instruments to Measure the Delta Ferrite Content of Austenitic Stainless Steel Weld Metal," April 19, (1974).
142. Savage, W. F. and Lundin, C. D., "Application of the Vareststraint Technique to the Study of Weldability," Welding J., 45(11), 497s-503s (1966).
143. Enjo, T., Kikuchi, Y., Kobayashi, T., and Kuwana, T., "Low Temperature Mechanical Properties of High-Nitrogen Type 316 Austenitic Stainless Steel Weld Metals," Trans. of JWRI, 10(1), 55-62 (1981).
144. Enjo, T., Kikuchi, Y., Kobayashi, T., and Kuwana, T., "Gas Shielded Arc Welding of High-Nitrogen Austenitic Stainless Steel (Report I)," Trans. of JWRI, 9(2), 173-180 (1980).
145. Irvine, K. J., Gladman, T., and Pickering, F. B., "The Strength of Austenitic Stainless Steels," JISI, July, 1017-1028 (1969).
146. Seth, C. W., C.B.I., Houston, Private Communication, May (1979).
147. Thermanit Welding Consumables, Catalogue 1151/3E, Thyssen Edelstahl Werke AG, July (1982)
148. Savage, W. F., Nippes, E. F., and Goodwin, G. M., "Effect of Minor Elements on Hot Cracking Tendencies of Inconel 600," Welding J., 56(8), 245s-253s (1977).

VITA

The author was born in Sen-Tung Province of Mainland China on June 5, 1947. He moved with his family to Taiwan in 1948. He earned his Bachelor of Science degree in the field of Physics at the Fu Jen University in 1969. He entered the Metallurgical Engineering Department of The University of Tennessee, Knoxville, in the Fall of 1975. He received a M.S. degree in Metallurgical Engineering in December, 1977 and continued his graduate work toward a Ph.D. degree at The University of Tennessee in the field of Metallurgical Engineering. He is a member of the American Society for Metals, the American Welding Society, and Sigma Xi.