

Naval Surface Warfare Center Carderock Division

West Bethesda, MD 20817-5700

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June 2020

Platform Integrity Department
Technical Report

Temperature-Dependent Material Property Databases for Marine Steels – Part 2: HSLA-65

by

Jennifer K. Semple

Daniel H. Bechetti

Wei Zhang

Charles R. Fisher



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CONTENTS

	<i>Page</i>
FIGURES.....	iv
TABLES	x
ADMINISTRATIVE INFORMATION	xii
ACKNOWLEDGEMENTS.....	xii
EXECUTIVE SUMMARY	14
BACKGROUND	14
APPROACH	15
Material Testing Program	15
<i>Chemical Composition.....</i>	<i>16</i>
<i>Heat Affected Zone Phase Transformation Analysis</i>	<i>16</i>
<i>Thermo-Physical Property Analysis</i>	<i>18</i>
<i>Thermo-Mechanical Property Analysis</i>	<i>19</i>
<i>Fabrication of Welded Specimens</i>	<i>21</i>
<i>Microstructural Analysis</i>	<i>22</i>
RESULTS AND DISCUSSION	22
Chemical Composition.....	22
Solidus and Liquidus Analysis	22
Base Metal Microstructure.....	23
Heat Affected Zone Phase Transformation Analysis.....	23
<i>Dilation Curve Analysis Method.....</i>	<i>23</i>
<i>Austenite Transformation Temperature Measurement</i>	<i>25</i>
<i>Continuous Cooling Transformation Diagrams</i>	<i>26</i>
Weldment Microstructures.....	37
Thermo-Physical Property Analysis	42
Thermo-Mechanical Property Analysis	45
SUMMARY	51
APPENDICES	52
Appendix A: Plate Conformance Certificate and Chemical Composition for HSLA-65 Steel.....	52
Appendix B: Continuous Cooling Transformation Curves	54
Appendix C: Representative Microstructures of HSLA-65 CCT Specimens.....	68

Appendix D: Microhardness Measurements.....	75
Appendix E: Thermo-Physical Property Datasets for HSLA-65.....	79
Appendix F: Thermo-Mechanical Properties.....	81
REFERENCES	87

FIGURES

	<i>Page</i>
Figure 1.	Photographs of the Gleeble-based dilatometry experimental set-up 17
Figure 2.	Specimen dimensions for Gleeble-based mechanical testing of HSLA-65 steel: a) single-reduced gauge section, b) double-reduced gauge section (dimensions in mm). 19
Figure 3.	Photographs of the Gleeble-based mechanical testing experimental set-up. 20
Figure 4.	Representative a) butt and b) tee joint weldments investigated for plate, weld, and HAZ microstructures. 21
Figure 5.	Representative micrographs of HSLA-65 steel plate, showing a) base metal microstructure and b) a binary representation of that micrograph produced using <i>ImageJ</i> used to calculate phase fraction. 23
Figure 6.	Representative on-heating dilatometry curve for HSLA-65 steel, showing the linear extrapolation and derivative curve methods for evaluating the ferrite + pearlite to austenite reaction. 24
Figure 7.	Variation in austenite transformation start and finish temperatures with heating rate for HSLA-65 material. 25
Figure 8.	Comparison of on-heating dilation curves measured at 10 °C/s (18 °F/s) heating and calculated under equilibrium conditions using Thermo-Calc 2020a 26
Figure 9.	CCT diagram for HSLA-65 heated to peak temperature of 825 °C (1517 °F). 27
Figure 10.	CCT diagram for HSLA-65 heated to peak temperature of 1000 °C (1832 °F). 27
Figure 11.	CCT diagram for HSLA-65 heated to peak temperature of 1150 °C (2102 °F). 28
Figure 12.	CCT diagram for HSLA-65 heated to peak temperature of 1350 °C (2462 °F). 28
Figure 13.	Variation of start temperature for the ferrite transformation shown in Figures 9-12. 29
Figure 14.	Variation of start temperature for the pearlite transformation shown in Figures 9-12. 30
Figure 15.	Variation of start temperature for the martensite transformation shown in Figures 9-12. 30

Figure 16.	Measured Vickers microhardness for HSLA-65 dilatometry specimens as a function of peak temperature and cooling rate.....	32
Figure 17.	Representative light optical micrographs of the second pass of the HSLA-65 fillet weld a) fusion line, with fusion zone above the dotted line and CGHAZ below, b) CGHAZ, c) FGHAZ, and d) ICHAZ.	38
Figure 18.	Representative light optical micrographs of the second pass of the HSLA-65 butt joint a) fusion line, with fusion zone to the right of the dotted line and CGHAZ to the left, b) CGHAZ, c) FGHAZ, and d) ICHAZ.....	38
Figure 19.	Composite image of light optical micrographs of the HSLA-65 fillet weld showing placement of microhardness indents.	39
Figure 20.	Average Vickers microhardness across the two lines of indents made for each fillet weld pass, as shown in Figure 19	39
Figure 21.	Composite image of light optical micrographs of the HSLA-65 butt joint weldment showing placement of microhardness indents.....	40
Figure 22.	Vickers microhardness traverses corresponding to the indents shown in Figure 21	41
Figure 23.	Vickers microhardness traverses corresponding to the indents shown in Figure 21	41
Figure 24.	Representative on-heating dilatometry curve for HSLA-65, showing typical CTE analysis and results.	42
Figure 25.	Density of HSLA-65 steel as a function of temperature.....	42
Figure 26.	Measured specific heat for HSLA-65 steel at various temperatures.....	43
Figure 27.	Measured thermal diffusivity for HSLA-65 steel at various temperatures.....	44
Figure 28.	Calculated thermal conductivity for HSLA-65 steel at various temperatures.....	44
Figure 29.	Assumed elevated temperature elastic modulus for HSLA-65 steel as adapted from references [40] [41].....	45
Figure 30.	Measured elevated temperature 0.2% offset yield strength for HSLA-65 steel.	46
Figure 31.	Measured elevated temperature ultimate tensile strength for HSLA-65 steel.	46

Figure 32.	Yield strength of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at different rates as compared with base material yield strength.....	47
Figure 33.	On-heating flow stress behavior for HSLA-65 steel at various temperatures.....	48
Figure 34.	On-heating flow stress behavior for HSLA-65 steel at testing temperatures above the austenitic phase transformation.	48
Figure 35.	Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 1 °C/s (1.8 °F/s) from a peak temperature of 1350 °C (2462 °F).	49
Figure 36.	Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 10 °C/s (18 °F/s) from a peak temperature of 1350 °C (2462 °F).	49
Figure 37.	Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 25 °C/s (45 °F/s) from a peak temperature of 1350 °C (2462 °F).	50
Figure 38.	Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 100 °C/s (180 °F/s) from a peak temperature of 1350 °C (2462 °F).	50
Figure 39.	Plate conformance certificate provided by the manufacturer for HSLA-65 material used in this report.....	52
Figure 40.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 1 °C/s.....	56
Figure 41.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 5 °C/s.....	56
Figure 42.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 10 °C/s.....	57
Figure 43.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 25 °C/s.....	57
Figure 44.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 100 °C/s.....	58
Figure 45.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 200 °C/s.....	58
Figure 46.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 1 °C/s.....	59

Figure 47.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 5 °C/s.....	59
Figure 48.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 10 °C/s.....	60
Figure 49.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 25 °C/s.....	60
Figure 50.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 100 °C/s.....	61
Figure 51.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 200 °C/s.....	61
Figure 52.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 1 °C/s.....	62
Figure 53.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 5 °C/s.....	62
Figure 54.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 10 °C/s.....	63
Figure 55.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 25 °C/s.....	63
Figure 56.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 100 °C/s.....	64
Figure 57.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 200 °C/s.....	64
Figure 58.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 1 °C/s.....	65
Figure 59.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 5 °C/s.....	65
Figure 60.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 10 °C/s.....	66
Figure 61.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 25 °C/s.....	66
Figure 62.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 100 °C/s.....	67

Figure 63.	Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 200 °C/s.....	67
Figure 64.	Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 825 °C (1517 °F) and continuously cooled at various rates. (a-f) : 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).	68
Figure 65.	Representative microstructure from the HSLA-65 dilatometry specimen heated to a peak temperature of 825 °C (1517 °F) and continuously cooled at 1 °C/s (1.8 °F/s).....	69
Figure 66.	SEM images of the microstructure of a HSLA-65 dilatometry specimen heated to a peak temperature of 825 °C (1517 °F) and cooled at 25 °C/s (45 °F/s).	69
Figure 67.	SEM images of the microstructure of a HSLA-65 dilatometry specimen heated to a peak temperature of 825 °C (1517 °F) and cooled at 100 (top row) and 200 °C/s (bottom row) (180 and 360 °F/s).	70
Figure 68.	Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 1000 °C (1832 °F) and continuously cooled at various rates. (a-f) : 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).	71
Figure 69.	SEM images of the microstructure of a HSLA-65 dilatometry specimen heated to a peak temperature of 1000 °C (1832 °F) and cooled at 100 and 200 °C/s (180 and 360 °F/s) showing islands of martensite interspersed with bainite.....	72
Figure 70.	Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 1150 °C (2102 °F) and continuously cooled at various rates. (a-f) : 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).	73
Figure 71.	Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 1350 °C (2462 °F) and continuously cooled at various rates. (a-f) : 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).	74
Figure 72.	Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 1 °C/s (1.8 °F/s), then reheated to the test temperature.	85
Figure 73.	Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 10 °C/s (18 °F/s), then reheated to the test temperature.	85

Figure 74.	Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 100 °C/s (180 °F/s), then reheated to the test temperature.	86
Figure 75.	Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 100 °C/s (180 °F/s), then reheated to the test temperature.	86

TABLES*Page*

Table 1.	Test matrix for HSLA-65 HAZ CCT diagram development. A_{c1} and A_{c3} are the on-heating austenite transformation start and finish temperatures, respectively.	17
Table 2.	Nominal welding parameters used to fabricate HSLA-65 butt and tee joints.....	21
Table 3.	Chemical composition of HSLA-65 plate material (wt%)	22
Table 4.	Solidus and liquidus temperatures for HSLA-65 steel, measured by Concurrent Technologies Corporation [14] [15]	23
Table 5.	Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 825 °C (1517 °F).....	36
Table 6.	Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 1000 °C (1832 °F).....	36
Table 7.	Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 1150 °C (2102 °F).....	37
Table 8.	Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 1350 °C (2462 °F).....	37
Table 9.	Chemical Composition of HSLA-65 Base Plate as Measured by OSU. All values in wt. %.	53
Table 10.	Austenite transformation temperatures illustrated in Figure 7	54
Table 11.	Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 825 °C (1517 °F)..	54
Table 12.	Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 1000 °C (1832 °F).	54
Table 13.	Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 1150 °C (2102 °F).	55
Table 14.	Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 1350 °C (2462°F).	55
Table 15.	Vickers microhardness of HSLA-65 as a function of peak temperature and cooling rate.....	75
Table 16.	Vickers microhardness measurements across the experimental tee joint weldment illustrated in Figure 20	76

Table 17.	Vickers microhardness measurements across the experimental butt joint weldment illustrated in Figure 21	77
Table 18.	Experimentally measured specific heat of HSLA-65.	79
Table 19.	Experimentally measured thermal properties of HSLA-65.	80
Table 20.	Assumed elevated temperature elastic modulus for HSLA-65 base material, as shown in Figure 29	81
Table 21.	Elevated temperature mechanical properties of HSLA-65 base material, as shown in Figures 30-31	81
Table 22.	Yield strength of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at different rates, as illustrated in Figure 32	82
Table 23.	On-heating flow behavior of HSLA-65 base material, as shown in Figures 33-34	82
Table 24.	Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 1 °C/s (1.8 °F/s), as illustrated in Figure 35 . Terminal values are at the UTS.	83
Table 25.	Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 10 °C/s (18 °F/s), as illustrated in Figure 36	83
Table 26.	Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 25 °C/s (45 °F/s), as illustrated in Figure 37	84
Table 27.	Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 100 °C/s (180 °F/s), as illustrated in Figure 38	84
Table 28.	Engineering fracture strain for the specimens shown in Figures 35-37	84

ADMINISTRATIVE INFORMATION

The work described in this report was performed by the Welding, Processing, and Nondestructive Evaluation Branch (Code 611) of the Platform Integrity Department at the Naval Surface Warfare Center, Carderock Division (NSWCCD) and by the Department of Materials Science and Engineering at the Ohio State University. The work was funded in FY17 by the Office of Naval Research (ONR), in support of the Lightweight Innovations for Tomorrow (LIFT) Institute's program entitled *Robust Distortion Control Methods and Implementation for Construction of Lightweight Metallic Structures*.

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EXECUTIVE SUMMARY

The Lightweight Innovations for Tomorrow (LIFT) project entitled Robust Distortion Control Methods and Implementation for Construction of Lightweight Metallic Structures sought to address distortion issues with thin-plate steel fabrication of U.S. Navy ships. Integrated Computational Materials Engineering (ICME)-based tools and techniques were identified as the best path forward for distortion mitigation through computational simulation of the welding process. ICME tools require temperature-dependent material properties to achieve accurate computational results for distortion and residual stress. Properties of note include specific heat, thermal conductivity, coefficient of thermal expansion, elastic modulus, yield strength, and flow stress of alloys from room temperature to near melting. In addition, the temperatures associated with on-heating and on-cooling phase transformations and their variation with heating rate, cooling rate, and peak temperature are also important for the prediction of residual stress and distortion evolution.

The integrated project team (IPT) made generating pedigreed, temperature-dependent material property databases of Navy-relevant steels a key task within the LIFT project. The testing plan included some of the most common marine steels used in the construction of U.S. Naval vessels, namely DH36, HSLA-65, HSLA-80, HSLA-100, HY-80, and HY-100. Material testing for each of the six steel grades was performed jointly by the Welding Engineering Program within the Department of Materials Science and Engineering at the Ohio State University (OSU) and the Welding, Processing, and Nondestructive Evaluation Branch at the Naval Surface Warfare Center, Carderock Division (NSWCCD). The temperature-dependent material property data was then sent to ESI for adaptation for use in SYSWELD, their welding-based finite-element analysis (FEA) software.

This report is the second part of a seven-part series detailing the pedigreed steel data. The first six reports will report and discuss the material properties for each of the individual steel grades, whereas the final report will compare and contrast the measured steel properties across all six steels and compare them to the available literature data. This report will focus specifically on the data associated with HSLA-65 steel.

BACKGROUND

The Lightweight Innovations for Tomorrow (LIFT) project entitled *Robust Distortion Control Methods and Implementation for Construction of Lightweight Metallic Structures* is a collaborative research project aimed at addressing the complex fabrication problems that arise from the use of thin, high-strength steel panels for ship construction. Lightweighting has gained attention in the commercial and military shipbuilding sectors in recent decades as a way to increase performance and capability while also reducing total ownership cost. Over the past decade, the production ratio of thin plate (10 mm [3/8-in.] or less) to thicker plate structures for steel construction of United States Coast Guard National Security Cutters (NSCs) at Huntington Ingalls Industries – Ingalls Shipbuilding has risen to over 4:1. Modern naval vessel designs also make greater use of complex panels with inserts and cutouts, further increasing the fabrication complexity to achieve weight savings while meeting structural requirements.

European shipbuilding research suggests that 30% of hull fabrication costs can be attributed to re-work and fit-up issues due to distortion [1]. ICME tools can be used to predict distortions associated with

the fabrication process of complex stiffened panels or other lightweight structures. These tools, once validated on selected product forms, could be used to establish recommended fitting, fixturing, welding, and assembly sequencing for optimized distortion control in thin-plate steel construction.

These ICME tools require detailed, reliable databases of temperature-dependent material properties in order to increase the accuracy of calculated distortion and residual stresses in welded metallic structures. The thermo-physical and thermo-mechanical properties of the material(s) being joined are of highest importance to the fidelity of such models. The properties of note include specific heat, thermal conductivity, coefficient of thermal expansion, elastic modulus, yield strength, and flow stress of the alloy, from room temperature to near melting. The temperatures associated with on-heating and on-cooling phase transformations and their variation with heating rate, cooling rate, and peak temperature are also important for the prediction of stress and distortion evolution.

While some of these data exist for the most common Navy steels, they are generally not suitable for use in high-fidelity welding-focused computational models for several reasons. First, the methods of data generation are inconsistent because the testing was completed over many decades by separate researchers and programs. Also, significant gaps exist in the data, especially at temperatures above approximately 400 °C (~750 °F). Additionally, some available data are restricted against public distribution. Finally, the data are, in general, subject to reliability issues because of changes in material fabrication and testing practices that have occurred in the decades since the data were generated. To combat this lack of comprehensive and consistent data, the integrated project team (IPT) made generating pedigreed, temperature-dependent material property databases of Navy-relevant steels a key task within the LIFT project. A major portion of the tasking involved physical simulation and analysis of heat-affected zone (HAZ) microstructures, because the identity and start/finish temperatures of phase transformations during welding significantly impact residual stresses. HAZ microstructures of Navy-relevant steels also do not have well-established mechanical property data, especially as a function of temperature and heating or cooling rate.

The testing plan included some of the most common marine steels used in the construction of U.S. Naval vessels, namely DH36, HSLA-65, HSLA-80, HSLA-100, HY-80, and HY-100. HSLA is short for “high-strength low-alloy,” whereas HY stands for “high yield.” The number designation behind HSLA and HY is the minimum yield strength of the plate material in units of ksi. Current (as of this writing) specifications for these alloys can be found in references [2] [3] [4]. Material testing for each of the six steel grades was performed jointly by the Welding Engineering Program within the Department of Materials Science and Engineering at the Ohio State University (OSU) and the Welding, Processing, and Nondestructive Evaluation Branch at the Naval Surface Warfare Center, Carderock Division (NSWCCD). The temperature-dependent material property data was then sent to ESI for adaptation for use in their welding-based, finite-element analysis (FEA) software colloquially known as *SYSWELD*.

This report is the second part of a seven-part series describing the pedigreed steel data. The first six reports (the first of which is given in [5]) will focus on establishing the material properties for each of the individual steel grades, whereas the final report will compare and contrast the measured steel properties across all six steels, while also comparing them to the available literature data. Follow-on reports detailing use of these temperature-dependent material databases within the LIFT-funded program are also expected.

APPROACH

Material Testing Program

The material property assessment program was developed through inputs from welding engineers at NSWCCD, researchers at OSU, and modeling experts at ESI. The program focused on generating the

types of data required to develop Navy-relevant material databases for use by ESI's commercial software, *SYSWELD*. The raw data could also be adapted for use by other FEA tools, such as MSC's *Simufact Welding* or the welding module for *Abaqus*. Specific data included thermo-physical and mechanical properties of the alloys of interest, from room temperature up to near-melting. Density, heat capacity, thermal conductivity, and coefficient of thermal expansion (CTE) were identified as the most important thermo-physical properties. Mechanical properties of interest included the elastic modulus, yield strength, and flow stress.

Material from a single HSLA-65 steel plate procured to ASTM A945 (Grade 65) [3] was analyzed for this project. The plate was acquired by Huntington Ingalls Industries – Ingalls Shipbuilding (HII – Ingalls) and was nominally 4.76 mm (3/16-in.) thick. The room temperature yield strength, ultimate tensile strength, and elongation requirements for the alloy are 450 MPa (65 ksi), 540 to 690 MPa (78 to 100 ksi), and 18% in 200 mm (8-in.) or 22% in 50 mm (2-in.), respectively [3]. The plate conformance certification sheet for the material investigated in this study is given in **Appendix A**. A portion of this plate was cut into 610 mm by 305 mm (24-in. by 12-in.) pieces and shipped to both NSWCCD and OSU for determination of temperature-dependent material properties.

Chemical Composition

The chemical composition of the base material was measured at OSU using optical emission spectroscopy in accordance with ASTM E415 [6].

Heat Affected Zone Phase Transformation Analysis

Phase transformations in HSLA-65 as a function of austenitization temperature and cooling rate were assessed via dilatometry. First, on-heating tests using a range of heating rates were performed in order to establish the austenite transformation temperature range and its dependence on heating rate. Then, continuous cooling transformation (CCT) diagrams were developed for four regions of the HAZ that are subject to microstructural transformation during welding: the intercritical region (ICHAZ); the low-temperature, fine-grained region (FGHAZ1); the high-temperature, fine-grained region (FGHAZ2); and the coarse-grained region (CGHAZ).

The experimental heating and cooling rates were defined by combining typical low alloy steel production heat inputs (0.4 to 2.6 kJ/mm [10 to 65 kJ/in.], as suggested by HII-Ingalls engineers), with analytical heat flow simulations using Sandia National Laboratories' *SmartWeld* program [7]. From these calculations, it was determined that heating rates between 10 and 2000 °C/s and cooling rates between 1 and 200 °C/s (1.8 and 360 °F/s) would provide information covering a sufficiently broad range of phase transformation behavior relevant to both slowly-cooled welds (*e.g.*, high heat input welds on thin plate) and rapidly-cooled welds (*e.g.*, low heat input welds on thick plate). Peak temperatures for three of the four HAZ regions were selected based on previous work with DH36 [5]. Selection of the ICHAZ peak temperature required knowledge of the on-heating austenite transformation temperature range and was therefore performed after the completion initial heating rate tests. **Table 1** shows the full test matrix utilized for HAZ CCT diagram development.

Execution of this test matrix was performed at NSWCCD by employing a Gleeble 3500 thermo-mechanical simulation apparatus to apply the prescribed heating and cooling rates. The Gleeble operates on the principles of resistive heating, conductive cooling through water-cooled fixturing, and rapid (50 kHz) control system feedback to apply precise thermal cycles even at rapid heating and cooling rates that cannot be easily achieved by other apparatus. The Gleeble 3500 system, in combination with the apparatus's standard "Pocket Jaw" setup, used full contact copper grips and cylindrical specimens 70 mm (2.76-in.) in length and 3 mm (0.118-in.) in diameter. A linear variable differential transformer (LVDT)-type contact extensometer was used to measure diametric dilation of the specimens during heating and

cooling. All dilatometry specimens were extracted perpendicular to the rolling direction of the plate, and all dilatometry tests were performed in an argon atmosphere. The typical experimental setup for the dilatometry testing is shown in **Figure 1**.

Table 1. Test matrix for HSLA-65 HAZ CCT diagram development. A_{c1} and A_{c3} are the on-heating austenite transformation start and finish temperatures, respectively.

Test Type	Region	Peak Temperature, °C (°F)	Heating Rate, °C/s (°F/s)	Soak Time, s	Cooling Rate, °C/s (°F/s)
A_{c1}/A_{c3} Determination	N/A	1200 (2192)	10 (18) 100 (180) 200 (360) 500 (900) 1000 (1800) 2000 (3600)	N/A	N/A
CCT Diagram Development	ICHAZ	825 (1517)	200 (360)	2	1 (1.8)
	FGHAZ1	1000 (1832)			5 (9)
	FGHAZ2	1150 (2102)			10 (18)
	CGHAZ	1350 (2462)			25 (45) 100 (180) 200 (360)

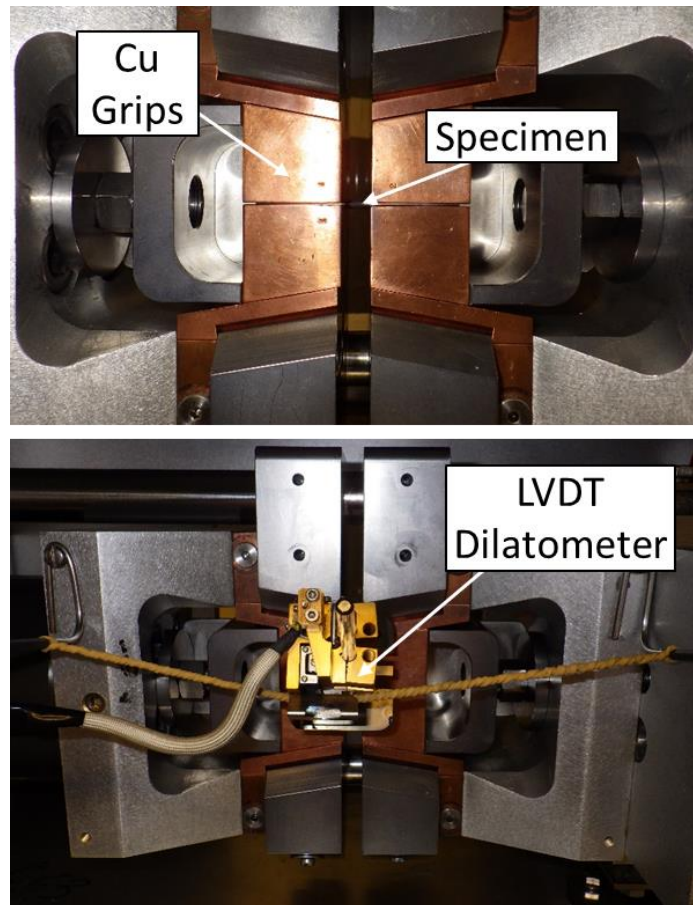


Figure 1. Photographs of the Gleeble-based dilatometry experimental set-up.

Thermo-Physical Property Analysis

Specimens were sent to the Thermophysical Properties Research Laboratory, Inc. (TPRL) in West Lafayette, IN for analysis of thermal diffusivity (α) and specific heat (C_p). Thermal diffusivity was measured using the laser flash method according to ASTM E1461 [8]. Specific heat was measured using differential scanning calorimetry (DSC) methods according to ASTM E1269 [9]. Coefficient of thermal expansion (CTE) was measured from the on-heating Gleeble dilatometry data in two regimes: 150 to 650 °C (302 to 1202 °F) for the base material microstructure, and 1000 to 1300 °C (1832 to 2192 °F) for austenite. No attempt was made to determine CTE during the on-heating transformation. CTE calculations were either performed by linear fitting or by discrete analysis using the dilation data endpoints from each temperature range in conjunction with **Equation 1**. The material was assumed to be isotropic.

$$\alpha_T = \left(\frac{\Delta d}{d_0} \right) \left(\frac{1}{\Delta T} \right) \quad (1)$$

where $\alpha_T \equiv$ Linear coefficient of thermal expansion [$^{\circ}\text{C}^{-1}$]
 $\Delta d \equiv$ Change in specimen diameter over the chosen temperature range [cm]
 $d_0 \equiv$ Initial specimen diameter [cm]
 $\Delta T \equiv$ Temperature range [$^{\circ}\text{C}$]

The temperature-dependent density (ρ) was calculated using a theoretical cubic volume element under the assumption of isotropic thermal expansion:

$$\rho(T) = \frac{\rho_0 V_0}{V(T)} \quad (2)$$

where $\rho(T) \equiv$ Density at temperature T [g/cm^3]
 $\rho_0 \equiv$ Room temperature density provided by TPRL [g/cm^3]
 $V_0 \equiv$ Room temperature volume of the theoretical cubic element [cm^3]
 $V(T) \equiv$ Volume of the theoretical cubic element at temperature T [cm^3]

V_0 was calculated assuming a side length equal to the room temperature width of OSU's dilatometry specimens. This value was chosen because, as shown in **Figure 1**, it corresponds to the starting gauge length for the dilatometer used to measure thermal expansion. A theoretical cubic volume element was chosen instead of the actual specimen dimensions to subvert potential complications with measuring dimensions that were not directly measured by the dilatometer (*e.g.*, the effect of thermal gradients on measurement of the specimen's longitudinal expansion). V_T was simply calculated by:

$$V(T) = [w(T)]^3 \quad (3)$$

where $w(T) \equiv$ Dilatometer-measured specimen width at temperature T [cm]

This method allowed calculation of the density through the austenite transformation. Finally, thermal conductivity (λ) as a function of temperature was calculated using **Equation 4**:

$$\lambda = \rho \cdot c_p \cdot \alpha \quad (4)$$

where $\lambda \equiv$ Thermal conductivity [$\text{W}/\text{cm}\cdot^{\circ}\text{C}$]
 $c_p \equiv$ Specific heat capacity [$\text{J}/\text{g}\cdot^{\circ}\text{C}$]
 $\rho_0 \equiv$ Room temperature density [g/cm^3]
 $\alpha \equiv$ Thermal diffusivity [cm^2/s]

Thermo-Mechanical Property Analysis

Tension testing of the HSLA-65 base material at temperatures between 22 and 1100 °C (72 and 2012 °F) was subcontracted to IMR Test Labs in Portland, OR. Room temperature testing was performed in accordance with ASTM A370 [10], and elevated temperature testing was performed in accordance with ASTM E21 [11]. Specimens were of rectangular cross-section with 12.7 mm (0.50 in.) gauge width and approximately 50.8 mm (2.0 in.) gauge length. Testing was performed in strain control mode at target rates of 0.005 min⁻¹ prior to yield and 0.05 min⁻¹ after yield.

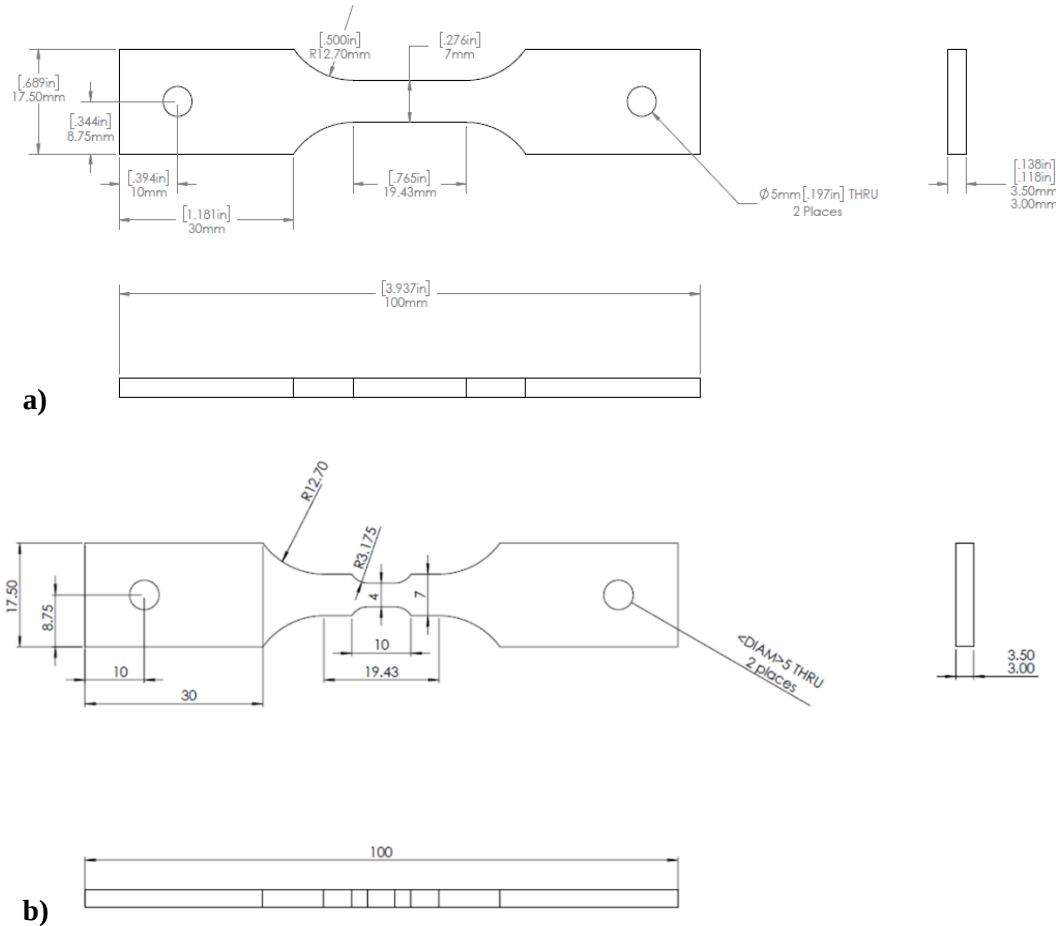


Figure 2. Specimen dimensions for Gleeble-based mechanical testing of HSLA-65 steel: **a)** single-reduced gauge section, **b)** double-reduced gauge section (dimensions in mm).

HAZ mechanical properties were measured using the Gleeble 3500 at NSWCCD. Prior to mechanical testing, specimens of the geometry shown in **Figure 2a** were thermally cycled to generate different initial microstructures based on the results of the CCT diagram development described above. For HSLA-65, microstructural analysis indicated that substantially different microstructures formed when the material was cooled at 1, 10, 25, and 100 °C/s (1.8, 18, 45, and 180 °F/s). Because of programmatic time constraints, only a peak temperature of 1350 °C (2462 °F) was applied to the tensile specimens. All specimens were heated in an argon atmosphere, and forced air was used to meet cooling rates of 10 °C/s (18 °F/s) and above. All specimens were cooled to room temperature prior to affixing an extensometer. The specimens were then reheated to the test temperature at a rate of 10 °C/s (18 °F/s) and soaked for 10 sec. before being pulled to failure in crosshead control mode at strain rates targeting those prescribed in ASTM E21 [11]. The crosshead displacement rate used was 0.30 mm/min (0.012 in/min). It is

acknowledged that this is not the most representative method of testing for welding-related applications because the stresses that develop during welding do so rapidly and while the material is cooling and undergoing phase transformations. However, given the programmatic intent of establishing the mechanical behavior of specific microstructures and the need to avoid exposing the extensometer device to damagingly high temperatures, the method of cooling to room temperature and testing after reheating was determined to be an acceptable compromise. In cases where the induced microstructure was stronger than the base material, a second reduced gauge section was machined in the specimen prior to testing to ensure specimen fracture in the region of interest. This revised specimen geometry is shown in **Figure 2b**.

Thermal cycling and subsequent tensile testing were performed using the apparatus's "Pocket Jaw" setup with minimal contact stainless steel grips used to fixture the specimens. These grips were selected to minimize the longitudinal thermal gradient in the gauge length of the specimens during testing. Graphite foil was inserted between the specimen and the grips to further minimize the thermal gradient. The initial gauge length for the tests varied between 8 and 10 mm (0.31 and 0.39 in.) depending on the specimen geometry (**Figure 2**). Elevated temperature tests were performed in an argon atmosphere. Preliminary testing as part of the DH36 test program [5] indicated that for this setup, the temperature variation across the initial gauge length was 10 to 30 °C (18 to 54 °F). The typical experimental setup for the tensile testing is shown in **Figure 3**.

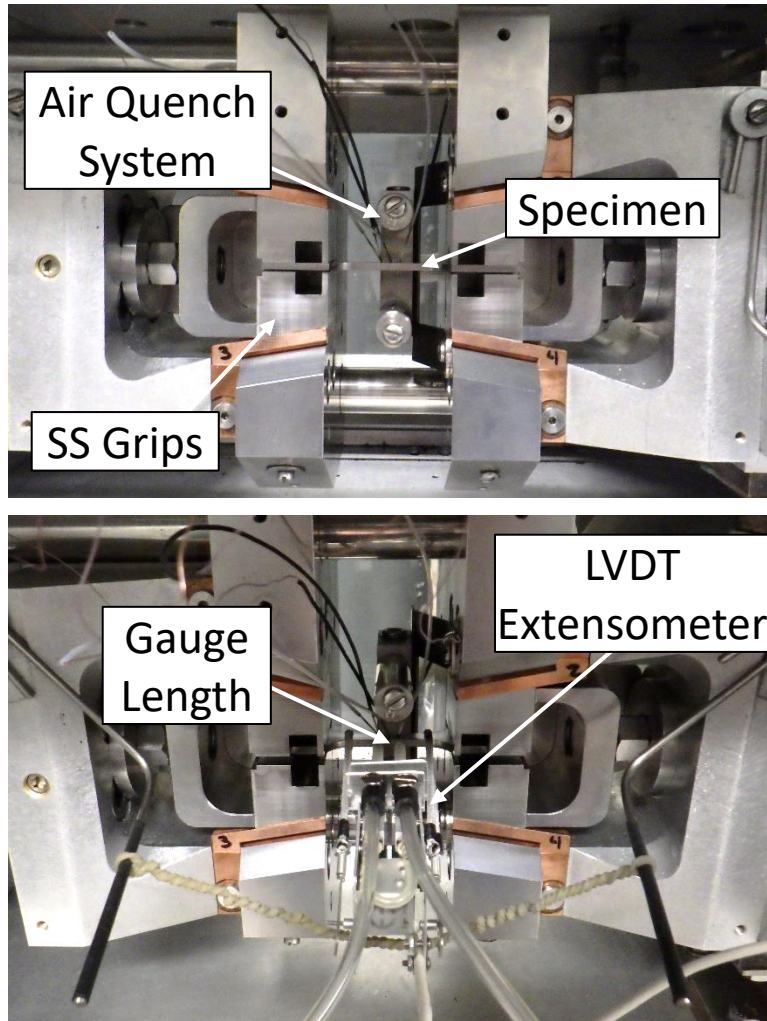


Figure 3. Photographs of the Gleeble-based mechanical testing experimental set-up.

Fabrication of Welded Specimens

Two sets of HSLA-65 plates were welded at HII-Ingalls and sent to OSU for analysis of the base plate, weld metal, and heat-affected zone (HAZ) microstructures. One set of plates was welded using the submerged arc welding (SAW) process in a two-sided butt joint configuration, as shown in **Figure 4a**. The other set was welded using flux cored arc welding (FCAW) in a two-sided tee-joint configuration, as shown in **Figure 4b**. Welding parameters for both welds are given in **Table 2**. These weldments enabled correlation of the HAZ microstructures generated through thermal simulation with those produced by arc welding. The two joint configurations are representative of frequently used thin plate structural welds: seaming butt joints and stiffener fillet joints. These joint types were also the focus of larger fabrication efforts in other tasks within the greater LIFT project as described in [12].

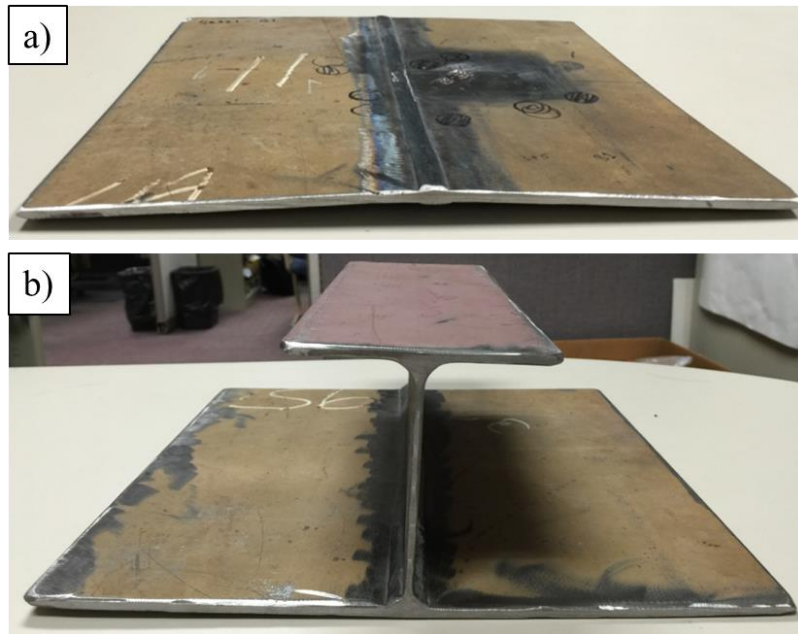


Figure 4. Representative **a)** butt and **b)** tee joint weldments investigated for plate, weld, and HAZ microstructures. Photos show weldments of DH36 material with the same joint configurations that were used for the HSLA-65 specimens.

Table 2. Nominal welding parameters used to fabricate HSLA-65 butt and tee joints

	Butt Joint	Tee Joint
Welding Process	Submerged arc (SAW)	Flux cored arc (FCAW)
Current (A)	300	200
Voltage (V)	30	27
Travel Speed (mm/s [in/min])	14.8 (35)	6.8-7.6 (16-18)
Heat Input (kJ/mm [kJ/in])	0.61 (15.4)	0.71-0.79 (18-20.3)
Consumable Designation(s)	AWS A5.23 EM12K	AWS A5.20 E71T-1
Filler Wire Diameter (mm [in])	3.2 (0.125)	1.3 (0.052)
Interpass Temperature, Max. (°C [°F])	149 (300)	149 (300)

Microstructural Analysis

To quantify and characterize the microstructures observed in HSLA-65, thermally-cycled specimens from the HAZ CCT diagram development were metallographically prepared to a final polish of 0.05 μm using standard techniques and etched by immersion in 4% Picral (4 g picric acid dissolved in 100 mL ethanol) for approximately 10 seconds followed by immersion in 2% Nital (2 mL nitric acid in 100 mL of ethanol) for 3 to 5 seconds. Identification and phase fraction measurement of the microstructural constituents in each specimen were evaluated using image analysis through light optical microscopy (LOM) and scanning electron microscopy (SEM). Vickers microhardness testing was also used to determine the hardness of the constituent(s) in each CCT specimen. This analysis facilitated proper CCT analysis, provided phase fraction input for the HSLA-65 material property database, and helped determine which thermal cycles were the most appropriate for application in the mechanical testing portion of the program described above. Additionally, a cross-section of each of the weldments represented in **Figure 4** was analyzed. Specimens were metallographically prepared in the manner described above and characterized using LOM and SEM. Vickers microhardness values at 200 g (7.05 oz.) load were taken from the fusion zone through the HAZ to the base material for both weldments.

RESULTS AND DISCUSSION

Chemical Composition

Table 3 shows the averaged results of five base material chemistry measurements for the HSLA-65 plate, plus manufacturer's plate certification values (shown in **Appendix A**) and the ASTM A945 Grade 65 specifications [3]. As expected, the measured chemical composition values generally agreed with the manufacturer's plate certification and were well within the ASTM specifications. The carbon equivalency number (CEN) was calculated according to [13]. The individual measurements made by OSU to obtain the average composition values are located in **Appendix A**.

Table 3. Chemical composition of HSLA-65 plate material (wt%)

Type	C	Mn	P	S	Si	Cu	Ni	Cr	Mo	V
Measured Avg.	0.086	1.48	0.009	<0.0005	0.27	0.006	0.014	0.026	0.006	0.079
Plate Certificate	0.08	1.49	0.007	0.001	0.281	0.010	0.01	0.02	0.007	0.075
ASTM A945 Grade 65	<0.10	1.1-1.65	<0.025	<0.010	0.1-0.4	<0.35	<0.40	<0.20	<0.08	<0.10

Type	Ti	Al	B	Nb	N	Sn	Fe	CEN [13]
Measured Avg.	0.012	0.029	0.001	0.030	0.006	0.0006	Bal.	0.261
Plate Certificate	0.013	0.032	0.0003	0.029	0.005	0.001	Bal.	---
ASTM A945 Grade 65	0.007-0.020	<0.08	---	<0.05	<0.012	---	Bal.	---

Solidus and Liquidus Analysis

Solidus and liquidus temperatures were not experimentally measured for HSLA-65 base material. Data reproduced from two studies conducted as part of a Navy Manufacturing Technology (ManTech) program with the Naval Metalworking Center from the mid-2000s [14] [15] are included in **Table 4**. No data for the liquidus of the 12.7 mm (0.5-in.) thick plate was reported [15].

Table 4. Solidus and liquidus temperatures for HSLA-65 steel, measured by Concurrent Technologies Corporation [14] [15]

Nominal Plate Thickness (mm [in])	Solidus (°C)	Liquidus (°C)
12.7 (0.5)	1517	---
31.75 (1.25)	1522	1538
63.5 (2.5)	1521	1533

Base Metal Microstructure

The microstructure of the HSLA-65 base metal is shown in the light optical micrograph of **Figure 5a**. The white-etched structure is equiaxed ferrite ($89\% \pm 1\%$), and the dark-etched constituent is pearlite ($11\% \pm 1\%$). The amount of each constituent was quantified by standard image threshold techniques applied to three fields in *ImageJ* software [16], as shown in **Figure 5b**. Two of these fields were taken at 500x magnification and one at 1000x magnification. The average Vickers microhardness of the base metal was 204 HV₂₀₀.

Observing the HSLA-65 base metal in both the longitudinal and transverse planes revealed equiaxed grains, but banding of microstructural features provided evidence of cross-rolling. The grain size was measured using the “intercept method” detailed in ASTM E1382 [17]. Using this technique, a circle is drawn at random locations within a micrograph, and the number of times a grain boundary intercepts that circle is used to calculate the average grain size. Using a circle instead of a line ensured that the presence of anisotropy (*e.g.*, rolling texture) did not bias results. Measurements were taken twice on a sample of base metal with the average ASTM grain size calculated to be $G = 12$.

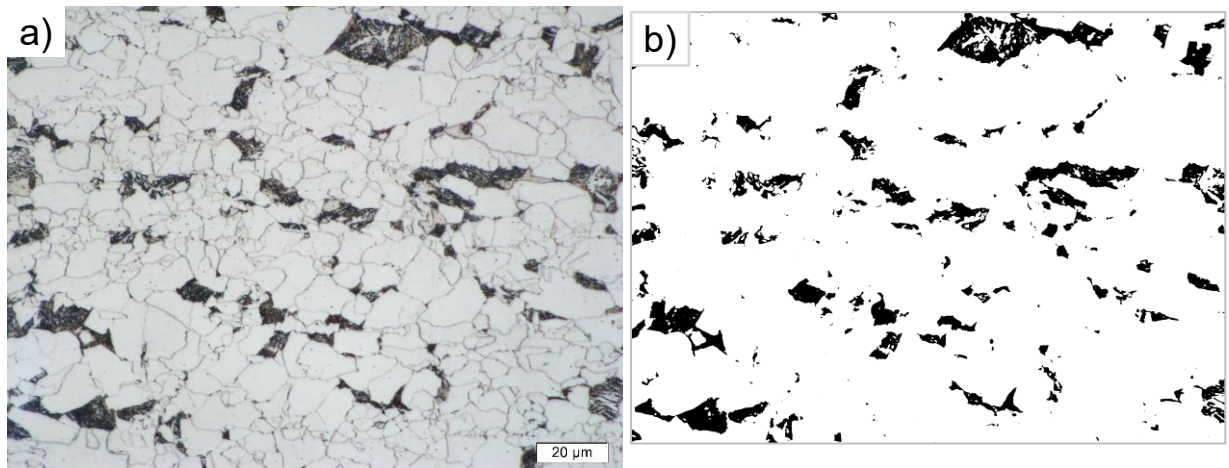


Figure 5. Representative micrographs of HSLA-65 steel plate, showing **a)** base metal microstructure and **b)** a binary representation of that micrograph produced using *ImageJ* used to calculate phase fraction.

Heat Affected Zone Phase Transformation Analysis

Dilation Curve Analysis Method

Figure 6 shows a representative on-heating portion of a Gleeble-generated dilatometry curve. The black curve (left axis) represents the change in sample diameter with temperature, and non-linearity

in the curve is indicative of a phase transformation's occurrence. Non-linearity is shown in the red derivative curve (right axis). Linear extrapolations of the black curve are shown by the blue lines. In this case, the transformation of interest is that of the HSLA-65 base metal microstructure to austenite.

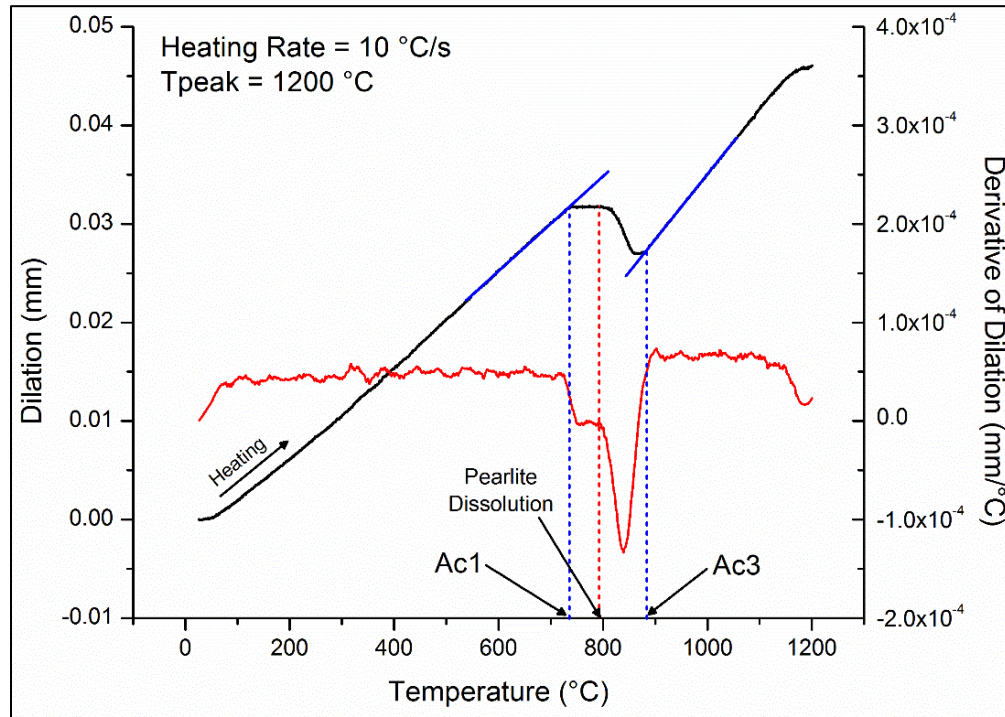


Figure 6. Representative on-heating dilatometry curve for HSLA-65 steel, showing the linear extrapolation and derivative curve methods for evaluating the ferrite + pearlite to austenite reaction. Black curve, left axis: dilation data. Red curve, right axis: derivative of dilation data. Blue dotted lines: extrapolation lines for determining transformation start (A_{c1}) and finish (A_{c3}) temperatures. Red dotted line: extrapolation line for determining temperature at which pearlite begins to dissolve.

Two methods for extracting start and finish temperatures for phase transformations from Gleeble dilatometry data are commonly used. The first method involves superimposing a straight line of matching slope along the low temperature portion of the dilation curve and extrapolating it past the onset of transformation. The point at which the experimental data begins to deviate from the straight line is then taken as the transformation start temperature. The same technique is then used to find the transformation finish temperature, with the straight line being extrapolated backward from the high-temperature portion of the curve after the transformation. The second method involves calculating and plotting the derivative of the dilation curve. Then the points at which the derivative begins to deviate from a constant value are taken as the transformation start and finish temperatures. Both methods for determining phase transformation temperatures are inherently somewhat subjective, and both also have advantages and disadvantages. The linear extrapolation method is faster and easier to interpret but is difficult to use when multiple transformations occur. The derivative method is more quantitative and can more easily distinguish multiple transformations but can be difficult to interpret when the dilation data is noisy. For this work, the linear extrapolation method was chosen as the primary method of data analysis, and the derivative method was used as a complementary technique for distinguishing concurrent phase transformations. Both methods are shown in **Figure 6**, where the derivative curve is displayed in red and the linear extrapolations are displayed in blue.

Austenite Transformation Temperature Measurement

On-cooling, solid-state phase transformation temperatures and products in steels can vary substantially with peak temperature due to differences in austenite grain size at different peak temperatures [18] [19] [20] [21]. It was therefore expected that each of the investigated HAZ regions would produce slightly different CCT diagrams. The residual stress (and subsequent distortion) that develops during weld cooling is greatly influenced by: 1) the microstructure that forms on cooling and 2) the temperature difference between the end of the on-cooling transformations and the final temperature of the component. Thus, full analysis must include as much HAZ phase transformation behavior as possible in order to provide the most complete and accurate material database for welding simulations.

Figure 7 presents the variation in austenite transformation behavior as a function of heating rate for HSLA-65 base material. As shown, the austenite transformation start temperature (A_{c1}) varies significantly over the range of experimental heating rates. This is a consequence of thermal diffusion during heating outpacing the mass diffusion required for the transformation to take place. The austenite transformation finish temperature (A_{c3}) remains relatively stable across the experimental heat rate range. The reason for the difference in heating rate response of the A_{c1} and A_{c3} temperatures in this material is not readily determined without further investigation. Researchers who have investigated similar behavior in other ferrous materials have associated heating rate dependencies of the austenite transformation with those of the controlling diffusion mechanism (i.e., volume carbon diffusion or grain boundary substitutional element diffusion), the ferrite recrystallization process, and/or austenite nucleation and growth rates [22] [23] [24]. These data are important for the refinement of welding simulation results because different regions of a weldment and its HAZ are subject to significantly different heating rates during the welding process. Thus, the on-heating transformation behavior must be incorporated in order to accurately predict which areas around the weld will transform to austenite, and will therefore be subject to re-transformation (and associated changes in mechanical properties and residual stresses) during cooling. The raw data points for **Figure 7** can be found in **Table 10** within **Appendix B**.

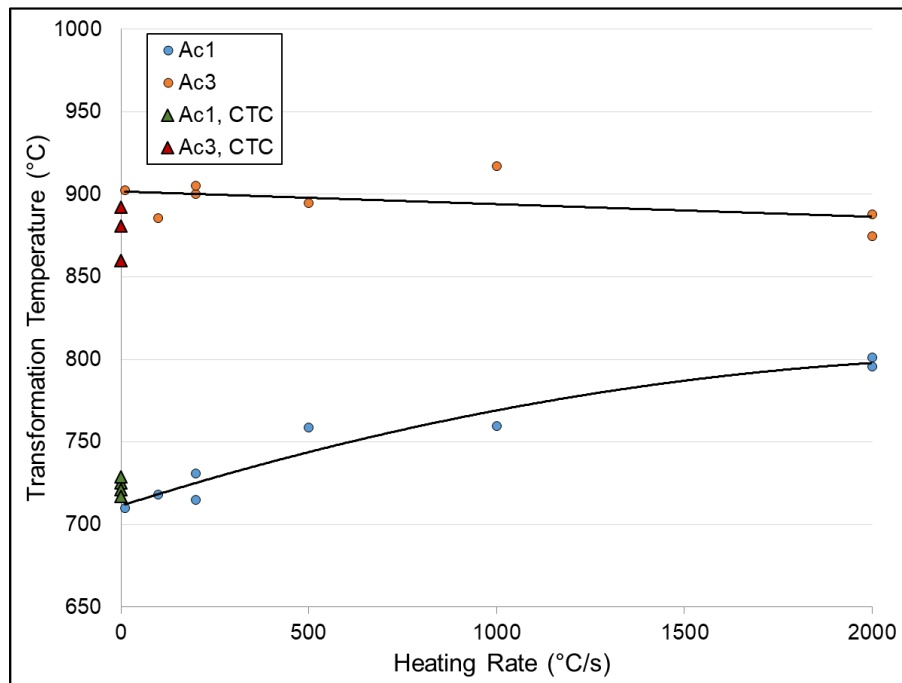


Figure 7. Variation in austenite transformation start and finish temperatures with heating rate for HSLA-65 material. Lines illustrate general trends in the data generated in this study.

As a comparison, data from the ManTech studies noted previously are also included in **Figure 7** [14] [15]. This data is labeled “CTC” for Concurrent Technologies Corporation, where the testing was conducted. In order to assess the ability of thermodynamic models to accurately predict austenite transformation temperatures, a dilation curve was calculated under equilibrium conditions using Thermo-Calc 2020a and the TCFE9 thermodynamic database [25]. As no experimental curve was produced under equilibrium conditions, the sample that was most slowly heated (at 10 °C/s [18 °F/s]) was used as a point of comparison. Dilation behavior matched well, with experimental transformation temperatures approximately 40 °C (104 °F) higher than calculated. This discrepancy likely occurs because of the difference between the thermal and mass diffusion during the experimental trial, as noted above.

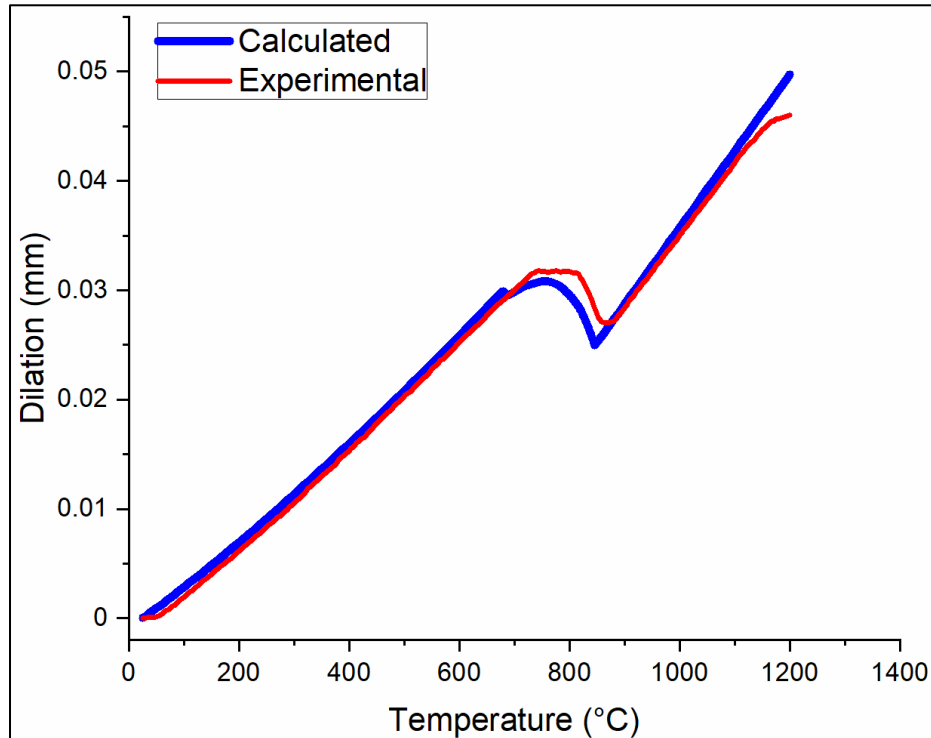


Figure 8. Comparison of on-heating dilation curves measured at 10 °C/s (18 °F/s) heating and calculated under equilibrium conditions using Thermo-Calc 2020a

Continuous Cooling Transformation Diagrams

CCT diagrams, assembled from dilatometry of HSLA-65 thermally cycled to peak temperatures representative of the four selected HAZ regions, are given in **Figures 9-12**. The austenite transformation temperatures described above led to the selection of 825 °C (1517 °F) as the ICHAZ peak temperature for HSLA-65. This temperature was 50 °C (90 °F) lower than the temperature used for DH36 [5]. The A_{c1} and A_{c3} temperatures labeled on the CCT diagrams are averages calculated from the individual A_{c1} and A_{c3} values for all of the specimens. The black curves are the actual specimen cooling curves. The target cooling rates for all specimens except those cooled at 100 and 200 °C/s (180 and 360 °F/s) were maintained through the on-cooling phase transformations. The latent heat released during low temperature transformation slowed the cooling rates of both the 100 °C/s (180 °F/s) and the 200 °C/s (360 °F/s) tests with peak temperatures of 1150 and 1350 °C (2102 and 2462 °F). As a result the target cooling rate was maintained until the start of the transformation, but was slowed by about 33% and 71% as the transformations occurred for 100 and 200 °C/s (180 and 360 °F/s) tests, respectively. The raw data for

these figures are given in **Tables 11-14 of Appendix B**, and the individual dilation curves are given in **Figures 40-63 of Appendix B**.

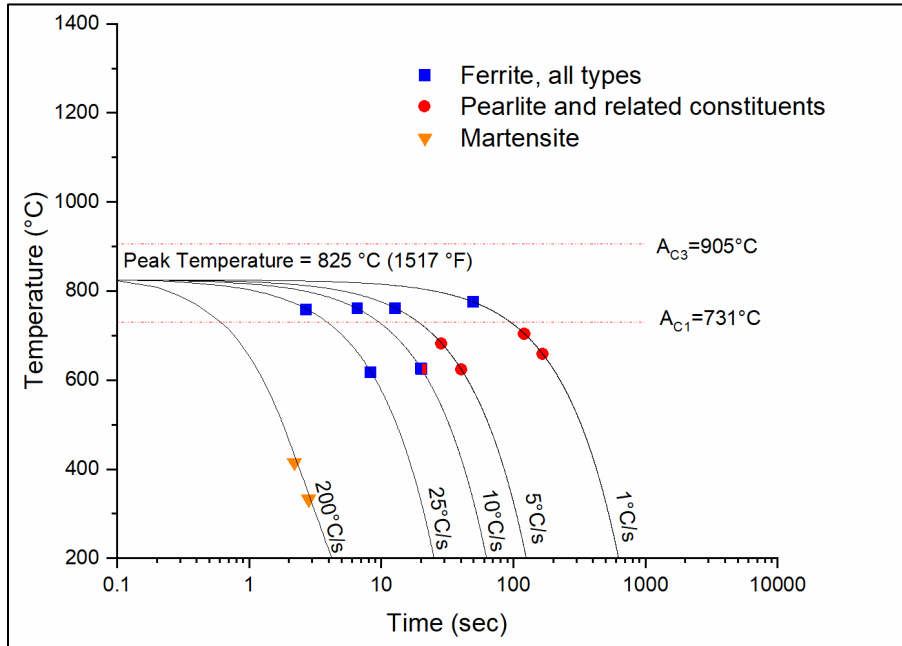


Figure 9. CCT diagram for HSLA-65 heated to peak temperature of 825 °C (1517 °F). Note that while martensite was observed in the sample cooled at 100 °C/s (180 °F/s), the transformation temperature could not be resolved from the dilation curve and is not represented in the diagram.

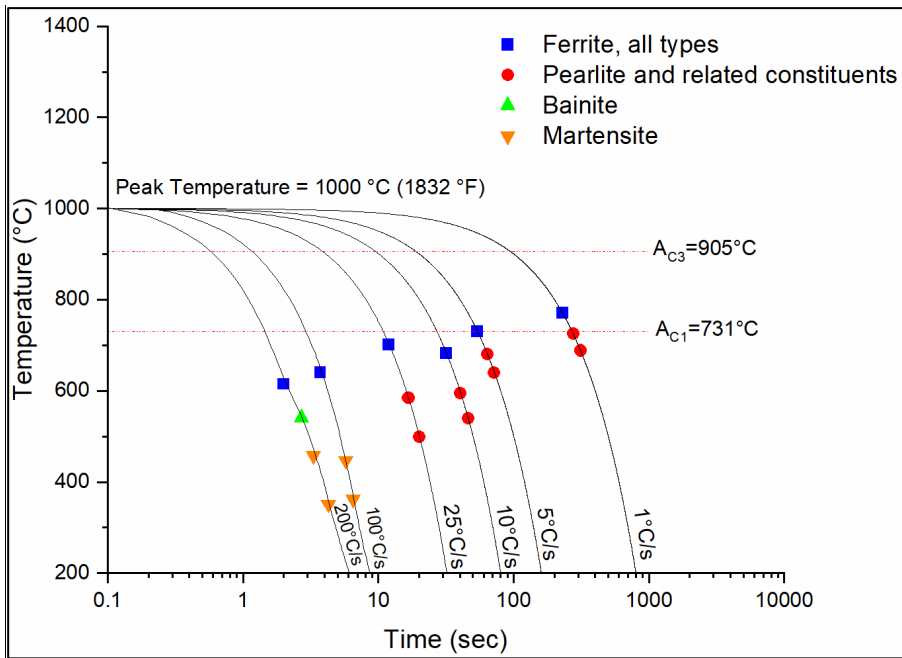


Figure 10. CCT diagram for HSLA-65 heated to peak temperature of 1000 °C (1832 °F). Note that while bainite was observed in the sample cooled at 100 °C/s (180 °F/s), the transformation temperature could not be resolved and is not represented in the diagram.

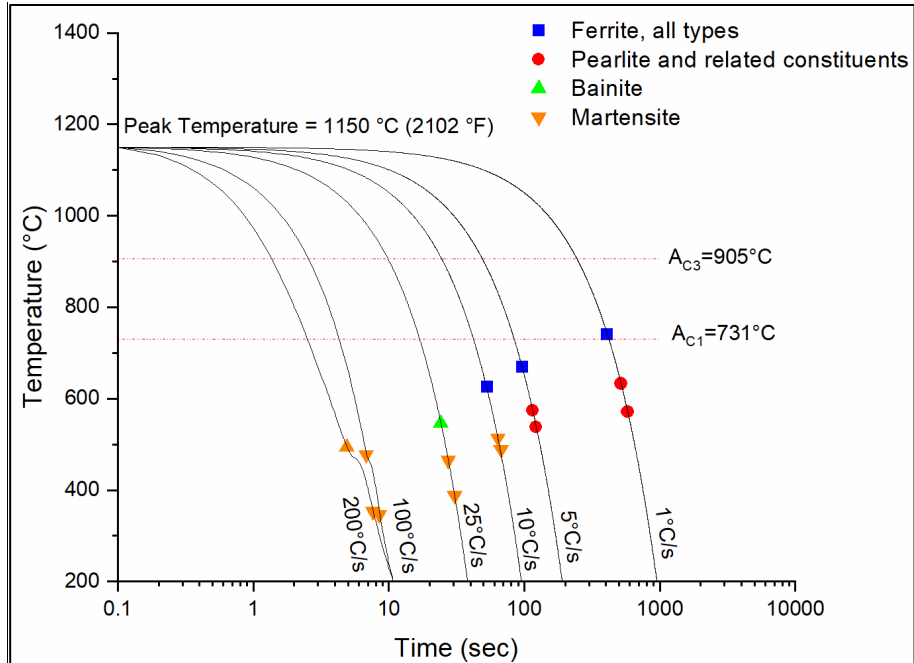


Figure 11. CCT diagram for HSLA-65 heated to peak temperature of 1150 °C (2102 °F). Note that while bainite was observed in the samples cooled at 100 and 200 °C/s (180 and 360 °F/s), the transformation temperatures could not be resolved and are not represented in the diagram.

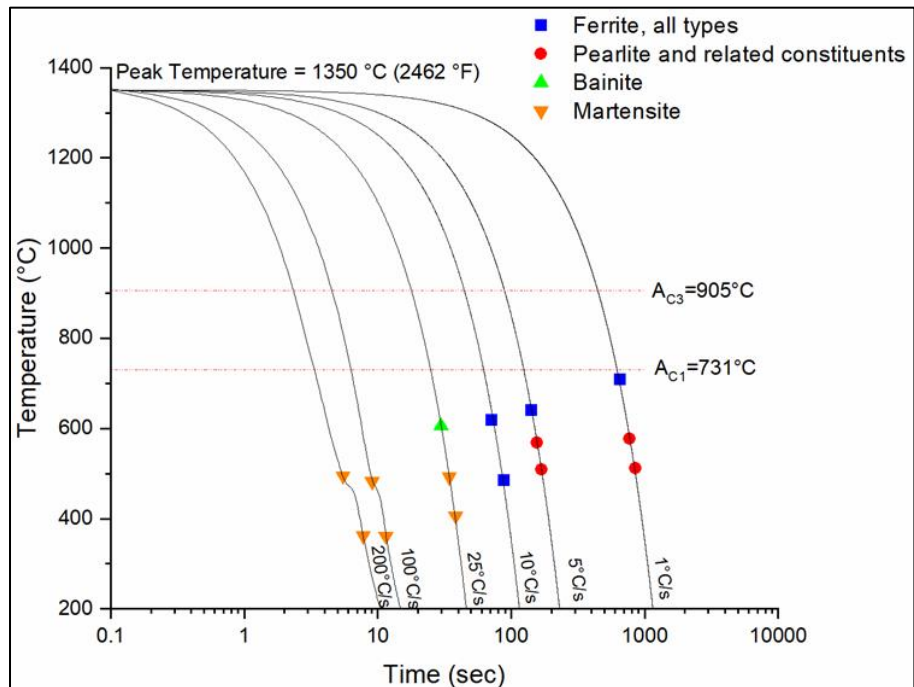


Figure 12. CCT diagram for HSLA-65 heated to peak temperature of 1350 °C (2462 °F). Note that while bainite was observed in the samples cooled at 100 and 200 °C/s (180 and 360 °F/s), the transformation temperatures could not be resolved and are not represented in the diagram.

The colored symbols in **Figures 9-12** correspond to the start and finish temperatures for the various on-cooling phase transformations. Each simulated HAZ region was subject to three regimes of on-cooling transformation start temperatures across the cooling rate range assessed in this work. The first occurred at high temperatures (619 to 775 °C [1146 to 1427 °F]) and is denoted in **Figures 9-12** by blue square symbols. The start of this transformation generally occurred near A_{c1} , though it decreased slightly with increasing peak temperature and increasing cooling rate. The second type of transformation identified is indicated by red circle symbols. This transformation generally started about 40 to 110 °C (104 to 230 °F) below the start of the higher temperature transformation and about 35 to 90 °C (95 to 194 °F) prior to the completion of austenite decomposition. It appeared to follow the same trends as the high temperature transformation with respect to peak temperature and cooling rate, but the transformation was not detected in the dilation curve for cooling rates above 5 °C/s (9 °F/s) in any specimen. The third transformation regime identified in the CCT diagrams, indicated as green and orange triangle symbols, generally started at temperatures of 414 to 500 °C (779 to 932 °F) and was only observed at cooling rates of 25 °C/s (45 °F/s) and higher. There is a slight trend of increasing transformation temperature as peak temperature increased. Identification of these transformations is discussed in more detail below, but for interpretation of the diagrams presented in **Figures 9-12**, the blue, red, green, and orange symbols reflect ferrite, pearlite, bainite, and martensite transformations, respectively.

The effects described above are summarized in an alternate way in **Figures 13-15**. **Figure 13** gives the variation in transformation start temperature for the ferrite transformation as a function of peak temperature and cooling rate. As shown, for a given cooling rate, increasing the peak temperature generally suppresses the onset of the transformation, resulting in lower start temperatures. This most likely occurs because PAG size increases with increasing peak temperature, resulting in less grain boundary area and therefore fewer nucleation sites for transformation. The onset of transformation is additionally suppressed by undercooling as cooling rate is increased. There was little change in ferrite transformation for samples heated to 825 °C (1517 °F) regardless of cooling rate, as no set of testing conditions provided significant time for austenite growth or carbon diffusion. At all other peak temperatures, an increase in cooling rate from 1 to 10 °C/s (1.8 to 18 °F/s) was associated with an approximately drop in ferrite start temperature. The ferrite start temperature decreased with increase peak temperature at constant cooling rate, likely because of the reduced austenite grain boundary area for ferrite nucleation.

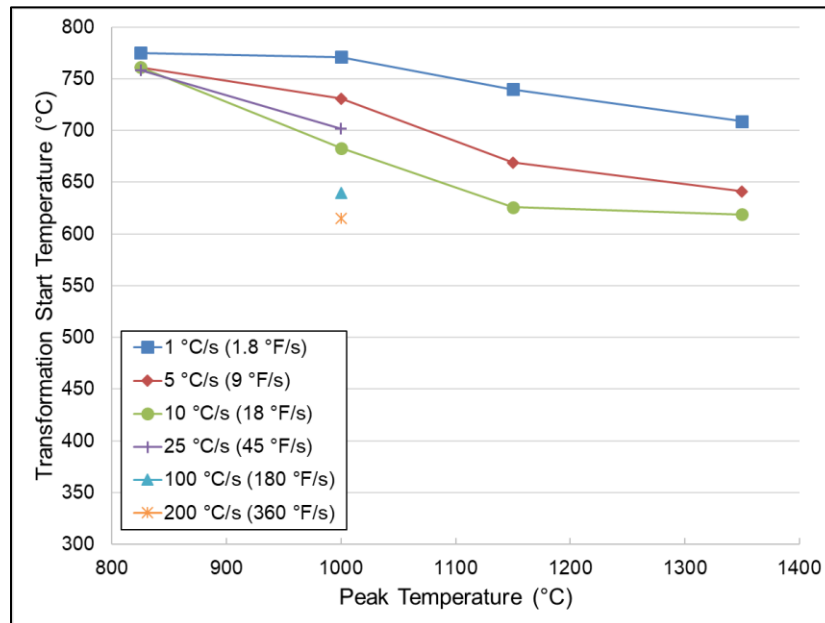


Figure 13. Variation of start temperature for the ferrite transformation shown in **Figures 9-12**.

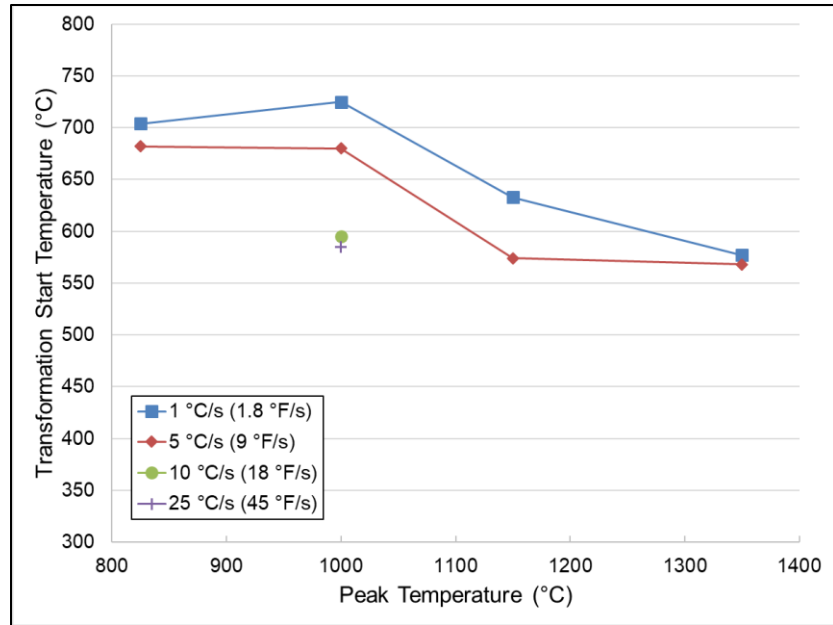


Figure 14. Variation of start temperature for the pearlite transformation shown in **Figures 9-12**.

Figure 14 shows that the pearlite transformation behaved in a similar manner to that of ferrite, with decreasing transformation start temperatures measured for both increasing cooling rates and increasing peak temperatures. The exception to this trend was the change in transformation temperature as peak temperature increased from 825 to 1000 °C – instead of decreasing, the transformation temperature showed negligible change for the 5 °C/s (9 °F/s) cooling rates and rose by 20 °C (68 F) for the 1 °C/s (1.8 °F/s) cooling rates.

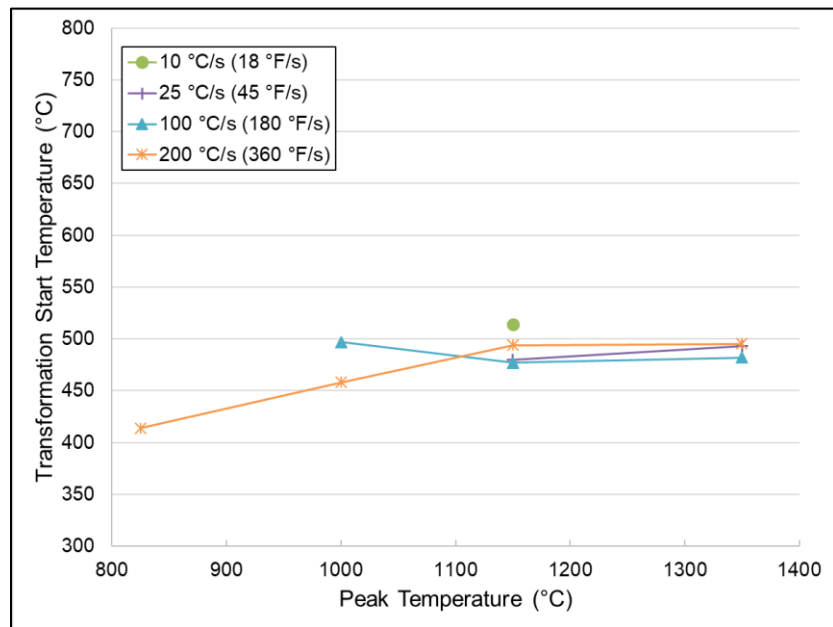


Figure 15. Variation of start temperature for the martensite transformation shown in **Figures 9-12**.

Trends in bainite transformation temperatures are not evaluated here as very few instances were detectable in dilation curves. Martensite was the final transformation product examined, with measured temperatures recorded in **Figure 15**. The increase in transformation temperature with decreasing peak temperature indicated that these transformation products formed more readily following precipitate dissolution and coarsening of prior austenite grains. When both were present in a single specimen, it was frequently unclear where bainite transformation completed and martensitic transformation began, with dilation curves appearing to show only one transformation despite micrographs reflecting two distinct microstructures. All specimens cooled at 100 to 200 °C/s (180 to 360 °F/s) appeared to contain both bainite and martensite, with the exception of the specimens with a peak temperature of 825 °C (1517 °F). This behavior is further discussed below in the context of the phases that were present for each combination of peak temperature and cooling rate.

In order to confidently establish the identities of the on-cooling phase transformations, LOM was performed on all dilatometry specimens, and select specimens were analyzed via SEM. Micrographs collected from the dilatometry specimens are given in **Figures 64-71** of **Appendix C**. Additionally, the measured microhardness for each specimen is given in **Figure 16** below and **Table 15** of **Appendix D**. For this discussion, the following definitions for the morphology of microstructural constituents are adopted [26] [27] [28] [29] [30]. The typical definitions for crystal structure and composition of the constituents are assumed.

- Primary ferrite (F_P)* – Carbide-free grain boundary or intragranular allotriomorphic or idiomorphic ferrite
- Widmanstätten ferrite with second phase (WF_{SP})* – High aspect ratio ferrite with either aligned or unaligned carbide
- Acicular ferrite (AF)* – Fine, interlocking structure formed by impingement of multiple Widmanstätten plates growing from intragranular inclusions
- Pearlite (P)* – Lamellar aggregate of ferrite and cementite
- Martensite-austenite constituent (M/A)* – Structure represented by a combination of martensite and residual austenite
- Bainite (B)* – General term for fine aggregates of ferrite laths or plates and cementite particles
- Lath martensite (M)* – Low-carbon (typically) martensite consisting of high aspect ratio platelets with highly dislocated substructures that are grouped into larger packets

Representative micrographs of all ICHAZ CCT specimens heated to 825 °C (1517 °F) are shown in **Figure 64** of **Appendix C**. This temperature falls between A_{c1} and A_{c3} (731 and 905 °C [1348 and 1661 °F], respectively) in order to achieve partial transformation to austenite prior to cooling. This partial transformation is expected to occur via austenite nucleation at the pearlite colonies as the cementite solvus temperature is exceeded and austenite-stabilizing carbon is released into solution [28].

Samples that were heated to intercritical temperatures and then slowly cooled partially retained microstructures from the original base material. The samples subjected to 1, 5, and 10 °C/s (1.8, 9, and 18 °F/s) cooling rates (shown in **Figure 64a-c**) all exhibited a mix of equiaxed ferrite, pearlite, and what appeared to be mixed martensite-austenite (M/A) constituent. Ferrite transformation temperatures fell slightly from 775 to 761 °C (1427 to 1402 °F) as cooling rate increased. Pearlite transformation temperatures followed the same trend, dropping from 704 to 682 °C (1299 to 1260 °F) as cooling rate increased from 1 to 5 °C/s (1.8 to 9 °C/s). The fraction of pearlite in more rapidly cooled samples was too low to detect a clear transformation in the dilation curves. Strong pearlite banding was present throughout the 1 °C/s (1.8 °F/s) specimen, with a band spacing of approximately 100 microns shown at low magnification in **Figure 65**. Pearlite banding is a common occurrence in low alloy steels and is attributable to remnant compositional segregation from the material's primary processing steps [31]. When examined with LOM, pearlite occasionally appeared as a black and white lamellar structure, but more often the individual lamellae were unresolvable and the entire colony appeared as a solid black

color. M/A constituent appeared as varying tan or brown shades, with color intensity dependent upon etching time. The distinction between the M/A constituent and pearlite morphologies was more pronounced when examined via SEM, as illustrated in **Figure 66**. High magnifications revealed pearlite to have a classic lamellar structure. M/A constituent appeared as more homogeneous regions containing very fine lath-like features. The proportion of martensite and austenite in these local regions could not be determined through optical or electron microscopy.

As cooling rate increased from 1 to 25 °C/s (1.8 to 45 °F/s), the proportion of M/A constituent to pearlite also increased, with no pearlite observed in the 25 °C/s (45 °F/s) specimen, as shown in **Figure 64d**. Microhardness values followed a similar trend, rising from an average of 179 to 208 HV₂₀₀ as cooling rate increased from 1 to 25 °C/s (1.8 to 45 °F/s), as illustrated in **Figure 16**. Higher cooling rates are expected to result in lower amounts of pearlite because its formation relies upon short-range carbon diffusion and is thus time-dependent. When cooling rates were increased to above 25 °C/s (45 °F/s), the islands of austenite that formed upon heating were cooled rapidly enough to form martensite, as seen in **Figure 64e-f**. The short time spent above A_{c1} allowed for dissolution of the carbides found within pearlite, but there was not adequate time for significant carbon diffusion. High carbon content is associated with hardenability in steels (*i.e.*, it encourages martensite formation), thus the locally carbon-rich regions that contained pearlite in the base material are expected to correspond with the location of martensite in the thermally cycled specimens. The absence of a high-temperature transformation indicated that the light-etched equiaxed ferrite in these specimens is untransformed base material. The presence of this ferrite kept the average microhardness for these specimens at a relatively low 225 and 238 HV₂₀₀ despite the presence of martensite. Microhardness of locally martensitic regions was not evaluated, but high standard deviations of 21 and 28 HV₂₀₀ for 100 and 200 °C/s (180 and 360 °F/s), respectively, are indicative of two regions with extremely different hardness values.

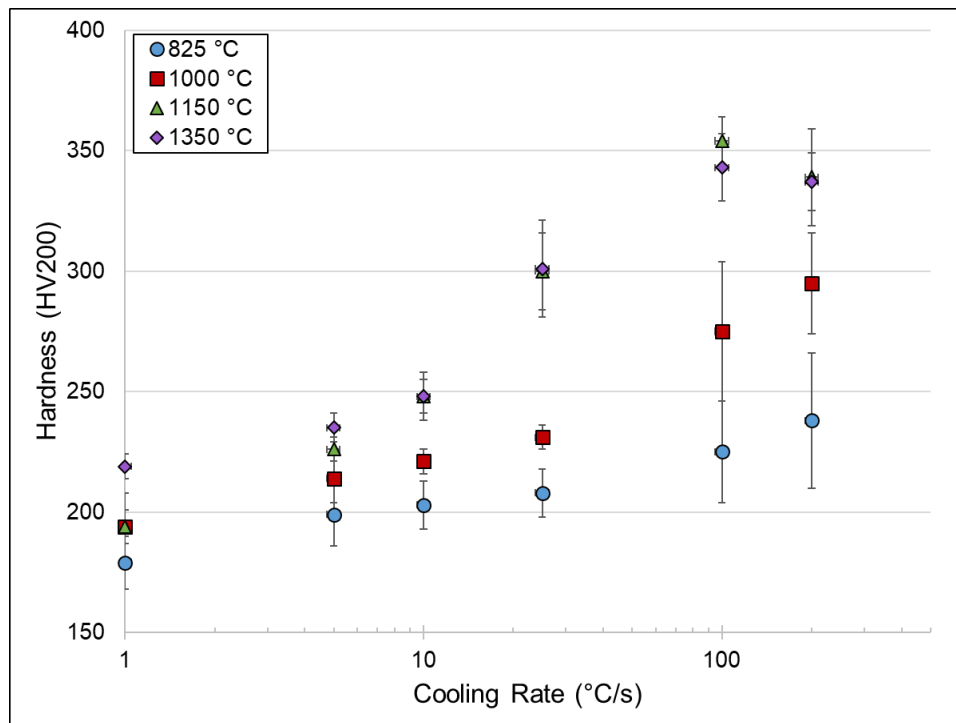


Figure 16. Measured Vickers microhardness for HSLA-65 dilatometry specimens as a function of peak temperature and cooling rate. Error bars are one standard deviation. The number of indents per specimen ranged from 10 to 90, as given in Table 15.

The martensitic islands were investigated via SEM in **Figure 67** and shown to have a classic lath morphology, though the small prior austenite grain size resulted in small and fragmented packets. The specimen cooled at 100 °C/s (180 °F/s) did not show a distinct on-cooling transformation in the dilatometry curve in either of two trials. While this absence has no obvious explanation, the microhardness values and microstructure morphology suggest that the dark-etched phase is indeed martensite. The morphology in the 100 °C/s (180 °F/s) specimen appeared to be the same as that seen in the 200 °C/s (360 °F/s) specimen, which experienced a phase transformation starting at 414 °C (777 °F). While bainite is frequently seen in HSLA-65 steels, the transformation temperature in these samples falls nearly 200 °C (392 °F) below calculated bainite start temperatures and agree with calculated martensite temperatures, as discussed in the following paragraph.

Various constitutive equations for predicting the start temperatures for bainite and martensite transformations based on alloy chemistry are available in the literature [32] [33]. Such models are generally oversimplifications and often are only accurate for alloys similar to those used to create them, but they offer the ability to gather an initial understanding of whether a given transformation was likely to be bainite or martensite. As such, the models developed by Capdevilla *et al.* [34] and Kirkaldy [35] were applied to the averaged HSLA-65 composition given in **Table 3**:

$$M_s = 491.05 - 302.6w_C - 30.6w_{Mn} - 16.6w_{Ni} - 8.9w_{Cr} + 2.4w_{Mo} - 11.3w_{Cu} + 8.58w_{Co} + 7.4w_W - 14.5w_{Si} \quad (5)$$

$$B_s = 656 - 57.7w_C - 75w_{Si} - 35w_{Mn} - 15.3w_{Ni} - 32w_{Cr} - 41.2w_{Mo} \quad (6)$$

where $M_s \equiv$ martensite start temperature (°C)
 $B_s \equiv$ bainite start temperature (°C)
 $w_i \equiv$ concentration of element i (wt%)

From **Equations 5** and **6**, the M_s and B_s temperatures for HSLA-65 are predicted to be 415 and 578 °C (780 and 1072 °F), respectively. In samples where a bainitic transformation was detected in the dilation curve, B_s was measured between 541 and 606 °C (1006 and 1123 °F), indicating acceptable applicability of this particular model to HSLA-65. Across all experimental conditions, M_s temperatures were measured between 414 and 514 °C (777 and 957 °F). Samples brought to lower peak temperatures generally had lower M_s temperatures than those heated to high temperatures for reasons discussed below.

Martensite transformation temperature was also calculated using Thermo-Calc 2020a software using the TCFE9 thermodynamic database [25]. The model used was based on work by Stormvinter, *et al.*, and updated to include work by Gulapura Hanumantharaju [36] [37]. In this approach, material data including chemical composition are used to model the transformation barrier with fitted analytical equations. Only lath martensite was considered as a viable transformation product, as other types of martensite typically require carbon levels of at least 0.5 wt% to form [27]. The model performs calculations under the assumption of equilibrium conditions, though the experimental trials likely conformed more closely to paraequilibrium conditions, which assume full redistribution of interstitial elements (*i.e.*, carbon) and no redistribution of substitutional elements. Because HSLA-65 has extremely low alloying content and carbon is the primary driver of the phase transformation of interest, approximating the behavior with an equilibrium model and the nominal alloy composition likely had minimal impact on calculation results. Use of the model with the nominal composition shown in **Table 3** predicted an M_s temperature of 434 °C (813 °F).

None of the experimental conditions produced fully martensitic microstructures. As such, phase transformations occurring prior to martensite formation would have created a heterogeneous carbon distribution within the bulk sample – that is, any ferrite formation would have rejected carbon into the remaining austenite due to differences in carbon solubility. The resulting carbon enrichment of the

austenite would have changed its composition relative to that of the base material and therefore changed the martensite start temperature relative to that of the nominal composition. To interrogate the effects of these local chemistry changes on the M_s temperature, Thermo-Calc was used to predict the temperature for two austenite chemistries. To do this, the expected chemical composition of ferrite near the measured M_s temperatures under equilibrium conditions was first calculated. Equilibrium was assumed to be representative because of rapid carbon diffusion (as described above), the low substitutional alloy content of HSLA-65, and the sufficient solubility for the substitutional alloying elements in ferrite. Next, two austenite compositions near the measured M_s temperatures were calculated using the nominal composition, ferrite volume fractions of either 0.40 or 0.60, and the calculated ferrite compositions at the corresponding ferrite volume fraction. The ferrite fractions correspond roughly to those measured in specimens containing both ferrite and martensite. The use of volume fractions without conversion to weight fraction prior to performing the required mass balance calculation was expected to have little effect on the end results because of the small difference in densities between the involved phases. For the cases of 40% and 60% ferrite, M_s temperatures calculated in this manner were 403 and 371 °C (757 and 700 °F), respectively. From this, a trend emerges showing a correlation between increased ferrite content – and subsequently increased carbon content in the available austenite – with decreased M_s temperature. This relationship is well-documented in literature, and is observable in the experimental results for ferrite fractions between 0.35 and 0.53 (**Tables 5-6, 11-12**). Experimental M_s temperatures were consistently higher than the predicted temperatures, but the reason for this is unknown.

Micrographs of HSLA-65 specimens heated to 1000 °C (1832 °F) are shown in **Figure 68**. These were intended to reproduce FGHAZ microstructures and thus were fully austenitized prior to cooling. When cooled at 1 °C/s (1.8 °F/s), a high temperature transformation was detected at 771 °C (1420 °F) and followed by a secondary transformation at 725 °C (1337 °F). This along with a low microhardness of 194 ± 7 HV₂₀₀ suggested the formation of ferrite followed by pearlite. Examination of a LOM micrograph in **Figure 68a** revealed a microstructure consisting of equiaxed ferrite, pearlite, and M/A constituent. Higher cooling rates limited pearlite formation and lowered the ferrite and pearlite transformation temperatures. The 5 and 10 °C/s (9 and 18 °F/s) specimens contained primary ferrite along with M/A constituent and minimal pearlite, and the 25 °C/s (45 °F/s) specimen contained almost entirely primary ferrite and M/A constituent. This increase in M/A proportion was accompanied by a corresponding increase in microhardness, to 231 ± 7 HV₂₀₀ in the sample cooled at 25 °C/s (45 °F/s). This change in both phase fraction and ferrite morphology can be followed visually in **Figure 68b-d**. As SYSWELD does not distinguish between ferrite morphologies, the distinction has been noted but not quantified. The low transformation finish temperature of 499 °C (930 °F) measure in the sample cooled at 25 °C/s (45 °F/s) coincides with the approximate M_s temperature of martensitic specimens in the test matrix, suggesting that a small fraction of the microstructure may consist of martensite. The low hardness of 231 ± 5 HV₂₀₀ indicates that the amount of martensite present, if any, is negligible, and the sample has been classified as a mixture of primary ferrite and M/A constituent.

Samples cooled at 100 and 200 °C/s (180 and 360 °F/s) were a complex mixture of several phases. The first of these was marked by a high-temperature transformation that began at 640 and 615 °C (1184 and 1139 °F) for the two cooling rates, respectively. The product of this transformation was primary ferrite, which appeared as a featureless light-etched phase in **Figure 68e-f**. A bainitic transformation was detected at 541 °C (1006 °F) in the specimen cooled at 200 °C/s (360 °F/s). Micrographs indicated that both samples contained bainite, appearing as a light-etched phase with aligned laths, but the transformation in the dilation curve of the sample cooled at 100 °C/s (180 °F/s) could not be resolved. A martensitic transformation began near 447 and 458 °C (837 and 856 °F) for the two specimens, with both the microhardness and the proportion of martensite increasing with cooling rate. Despite similar phase fraction of martensite to that measured in the ICHAZ sample, the M_s temperature was significantly higher. This was influenced by the presence of bainite. While the large amount of ferrite in the ICHAZ samples had low carbon solubility, bainite contains carbide precipitates and would have decreased the carbon content associated with the austenite prior to its transformation to martensite. As

was the case with the rapidly cooled ICHAZ specimens, the large standard deviation in microhardness was indicative of the multi-phase microstructure.

Microstructures representative of a higher peak temperature FGHAZ (1150 °C [2102 °F]) are shown **Figure 70**. They are similar to those observed in the 1000 °C (1832 °F) specimens, with larger prior austenite grain size driving minor differences. The specimen cooled at 1 °C/s (1.8 °F/s) had a microstructure of equiaxed ferrite and pearlite, with very small amounts of M/A constituent discernible via LOM. Compared to pearlite colonies in the corresponding 1000 °C (1832 °F) peak temperature sample, pearlite colonies were coarser, bordering on degenerate. This was likely due to the increased time spent in a high temperature regime. The sample cooled at 5 °C/s (9 °F/s) was comprised of a mixture of polygonal and Widmanstätten ferrite, M/A constituent, and trace amounts of pearlite. Increasing the cooling rate to 10 °C/s (18 °F/s) increased the degree of undercooling and resulted in a shift to a Widmanstätten ferrite morphology. This constituent was characterized by packets of coarse, aligned laths. It is acknowledged that this constituent is morphologically similar to bainite, but the small packet size and low aspect ratio relative to bainite seen in more rapidly cooled specimens suggests that the primary transformation product here is Widmanstätten ferrite. Small islands of martensite were visible at the LOM level, with M_s and M_f temperatures of 514 and 489 °C (957 and 912 °F), respectively.

The sample cooled at 25 °C/s (45 °F/s) experienced a significant decrease in transformation temperatures, with the initial transformation from austenite occurring at 546 °C (1015 °F) and the secondary transformation product beginning transformation at 480 °C (896 °F). The initial transformation product appeared to be bainite, with needle-like morphology reminiscent of Widmanstätten ferrite but a higher aspect ratio and forming 80 °C (176 °F) lower than the ferrite seen in the sample cooled at 10 °C/s (18 °F/s). The relatively steep transformation in the dilation curve suggests a rapid displacive transformation, and while Widmanstätten ferrite also forms via displacive mechanisms, its transformation rate is limited by carbon diffusion. The lower temperature transformation was martensite, as evidenced by a low transformation finish temperature of 388 °C (730 °F) and a high hardness of 300 HV₂₀₀. The phase fraction of martensite was similar to those of samples heated to lower peak temperatures, but the M_s temperature was higher due to the absence of a ferritic transformation. The presence of bainite did not seem to have a significant effect on M_s temperature in any samples with a dual-phase bainite and martensite microstructure. At cooling rates of 100 and 200 °C/s (180 and 360 °F/s), the microstructures were predominantly transformed to lath martensite, as evidenced by low transformation temperatures of 477 and 494 °C (891 and 921 °F), respectively, and with microhardness values above 330 HV₂₀₀. A small fraction of bainite was identified via LOM, though it could not be resolved in the dilation curves for either sample. Representative micrographs in **Figure 71e-f** show typical lath martensite morphologies.

The final peak temperature tested was 1350 °C (2462°F), representing temperatures reached by the CGHAZ. Representative microstructures are shown in **Figure 71**. HSLA-65 is typically strengthened with microalloying additions including niobium, vanadium, and titanium, which can form nitrides or carbides that pin austenite grain boundaries and maintain a fine grain structure [38]. At high temperatures these precipitates dissolve, allowing for unhindered austenite grain growth. This is demonstrated in the specimen cooled at 1 °C/s (1.8 °F/s), where enough grain boundary ferrite formed to clearly delineate prior austenite grain boundaries. These boundaries act as nucleation sites for ferrite, and with the larger grains—and overall lower total boundary length—there are fewer opportunities for polygonal ferrite nucleation. As such, there was more austenite available at lower temperatures to transform into primary Widmanstätten ferrite, visible in **Figure 71a** as aligned ferrite plates and pearlite, growing from grain boundary ferrite. Intragranular acicular ferrite was also observed. This morphology requires high amounts of undercooling to form and would have similarly benefited from a high fraction of austenite in a low temperature regime.

Despite the large fraction of soft grain boundary ferrite, the microhardness of the 1 °C/s (1.8 °F/s) specimen was significantly higher than that seen in the slow-cooled samples associated with lower peak

temperatures. The samples cooled at 1 °C/s (1.8 °F/s) with peak temperatures of 1000 and 1150 °C (1832 and 2102 °F) demonstrated average microhardness values of 194 HV₂₀₀, while the sample with a 1350 °C (2462 °F) peak temperature had average microhardness of 219 HV₂₀₀. This is likely a Hall-Petch effect resulting from the relatively fine acicular ferrite. The specimens cooled at 5 and 10 °C/s (9 and 18 °F/s) seemed to be comprised of a mix of acicular and Widmanstätten ferrite, though with longer laths of Widmanstätten ferrite than seen in the samples heated to 1150 °F (2102 °C). This may be explained by the increased time in the high temperature regime, allowing prior austenite grains to coarsen. The smaller prior austenite grain size of the 1150 °C (2102 °F) samples would limit the growth of laths. In addition, the dissolution of intragranular precipitates at high temperatures would discourage nucleation of ferrite at locations other than austenite grain boundaries or allotriomorphic grain boundary ferrite, thus reducing competitive growth and allowing unhindered growth of long plates.

The sample cooled at 25 °C/s (45 °F/s) was a mix of bainite and martensite. The B_s temperature was measured at 606 °C (1123 °F), with martensite beginning to form at 493 °C (919 °F) and finishing at 406 °C (763 °F). Both phases are present in **Figure 71d**, which shows their two distinct morphologies. The intermediate microhardness of 301 HV₂₀₀ falls below that of pure martensite but above typical values recorded for ferrite or bainite alone. Specimens cooled at 100 and 200 °C/s (180 and 360 °F/s) were both primarily martensitic, with low transformation start temperatures of 482 and 495 °C (900 and 923 °F), respectively, and high average microhardness values of 343 and 347 HV₂₀₀, respectively. As with the 1150 °C (2102 °F) peak temperature specimens, a small amount of bainite was identified via LOM, though it was not apparent in the dilation curves. **Figure 71e-f** show that the representative microstructures for these two specimens resemble bainite along with typical lath martensite. **Table 5** gives the measured fractions of each microstructural constituent for these specimens. As noted previously, ferrite and pearlite are not distinguished in SYSWELD software, so for the purposes of inclusion into a database for that software, all types of ferrite, pearlite, and M/A constituent were grouped together.

Table 5. Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 825 °C (1517 °F)

Cooling Rate		Peak Temperature = 825 °C (1517 °F)		
°C/s	°F/s	F + P	B	M
1	1.8	1	0	0
5	9	1	0	0
10	18	1	0	0
25	45	1	0	0
100	180	0.53	0	0.47
200	360	0.50	0	0.50

Table 6. Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 1000 °C (1832 °F)

Cooling Rate		Peak Temperature = 1000 °C (1832 °F)		
°C/s	°F/s	F + P	B	M
1	1.8	1	0	0
5	9	1	0	0
10	18	1	0	0
25	45	1	0	0
100	180	0.43	0.17	0.40
200	360	0.35	0.06	0.59

Table 7. Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 1150 °C (2102 °F)

Cooling Rate		Peak Temperature = 1150 °C (2102 °F)		
°C/s	°F/s	F + P	B	M
1	1.8	1	0	0
5	9	1	0	0
10	18	1	0	0
25	45	0	0.46	0.54
100	180	0	0.09	0.91
200	360	0	0.08	0.92

Table 8. Measured phase fraction of HSLA-65 CCT specimens thermally cycled to a peak temperature of 1350 °C (2462 °F)

Cooling Rate		Peak Temperature = 1350 °C (2462 °F)		
°C/s	°F/s	F + P	B	M
1	1.8	1	0	0
5	9	1	0	0
10	18	1	0	0
25	45	0	0.67	0.33
100	180	0	0.34	0.66
200	360	0	0.26	0.74

The phase transformation information developed here is critically important for ensuring that weld simulation software can draw on thermo-physical and thermo-mechanical property information from the appropriate phases at the appropriate times during calculations. The results in **Figures 9-15** highlight the importance of developing multiple HAZ-related CCT diagrams.

Weldment Microstructures

Metallographic specimens were removed from both weldments and analyzed in a manner similar to the Gleeble specimens. **Figure 17** presents representative micrographs from the tee joint. **Figure 17a** shows that the fusion zone was a mixture of coarse primary ferrite, acicular ferrite, and unidentified carbides. **Figure 17b** shows that the CGHAZ was composed of coarse WF_{SP} and small islands of pearlite and M/A constituent. **Figure 17c** shows that the FGHAZ was composed of the same microstructures, albeit with a finer lath size for WF_{SP} . **Figure 17d** shows that the tee joint ICHAZ had a similar microstructure to other HAZ regions but with a shift in ferrite morphology from Widmanstätten to quasi-polygonal ferrite. As in the CGHAZ and FGHAZ, ferrite was interspersed with both pearlite colonies and M/A constituent. The microstructures found in the butt joint were generally the same. Representative micrographs are shown in Figure 18. The dilatometry specimen microstructures that most closely correspond with those observed in the fillet weld CGHAZ, FGHAZ, and ICHAZ are those associated with 5 °C/s (9 °F/s) cooling. Inspection of optical micrographs reveals similar microstructural morphologies in both sets. Corresponding microhardness values also match well, with measurements across the relevant microstructures for both weldments and thermally cycled specimens falling within the same range of 170-240 HV₂₀₀.

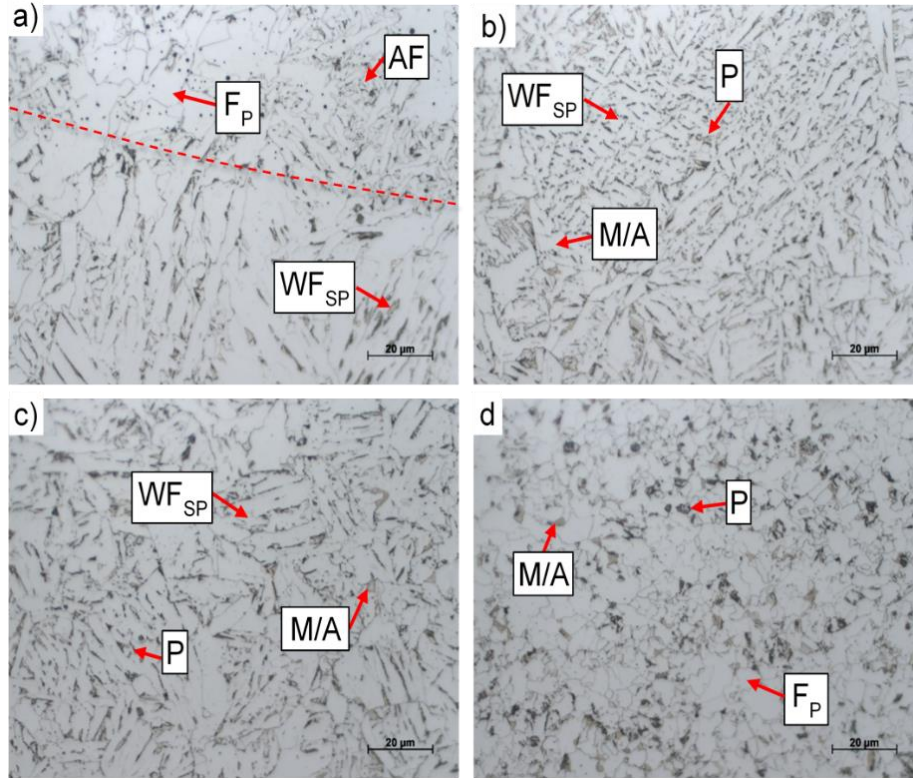


Figure 17. Representative light optical micrographs of the second pass of the HSLA-65 fillet weld a) fusion line, with fusion zone above the dotted line and CGHAZ below, b) CGHAZ, c) FGHAZ, and d) ICHAZ.

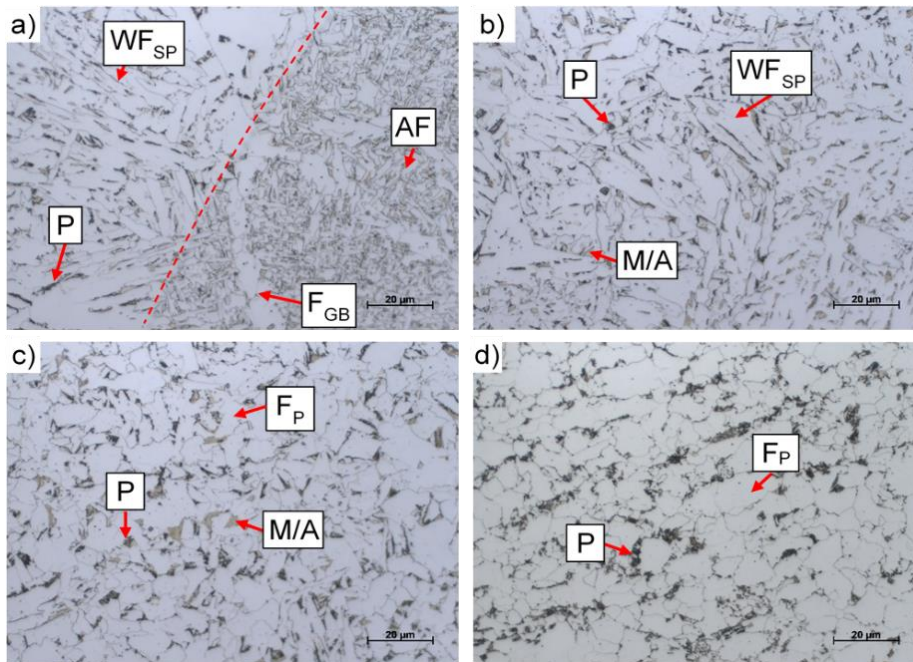


Figure 18. Representative light optical micrographs of the second pass of the HSLA-65 butt joint a) fusion line, with fusion zone to the right of the dotted line and CGHAZ to the left, b) CGHAZ, c) FGHAZ, and d) ICHAZ.



Figure 19. Composite image of light optical micrographs of the HSLA-65 fillet weld showing placement of microhardness indents. Indents appear as parallel lines of black dots on the left and right sides of the image.

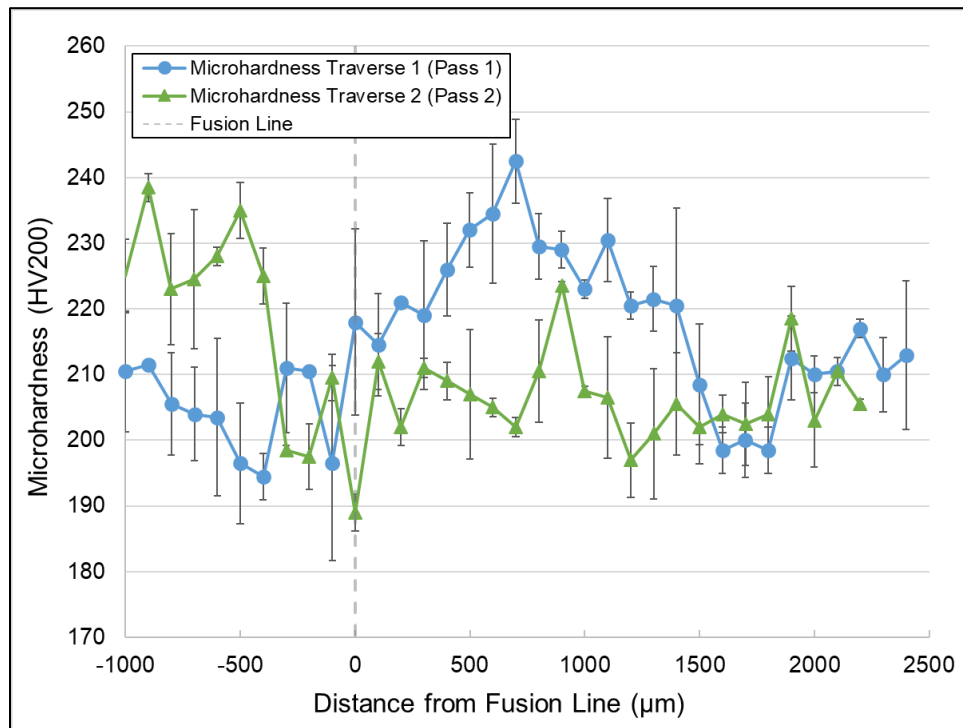


Figure 20. Average Vickers microhardness across the two lines of indents made for each fillet weld pass, as shown in **Figure 19**. Error bars are one standard deviation. Dotted line demarcates fusion line, as reported in **Table 16** of **Appendix D**.

The light optical micrographs in **Figure 19** demonstrate placement of microhardness indents within the microstructural gradient of the tee joint's fillet weld heat affected zones. **Figure 20** shows the averaged measurements from the two linear traverses associated with each weld bead. Microhardness data is presented as distance from the fusion line, though the width of the heat affected zone varies depending

on location within the weldment (*i.e.*, near the surface or the center of the plate). As such, care should be taken when comparing values taken at the same distance from the fusion line, as they do not necessarily correspond with the same region of the heat-affected zone. Inconsistencies also exist between microhardness values at the same linear distance from the fusion line due to challenges associated with collecting hardness traverses perpendicular to the non-semicircular HAZ boundaries in thin plates. In the case of the fillet weld, microhardness trends were different for the HAZ of the two beads. The first pass showed an increase in microhardness from the CGHAZ to the FGHAZ, a drop to the minimum hardness value in the ICHAZ, and a slight increase back to the base metal microhardness. The second pass had microhardness values between 190-220 HV₂₀₀ but with no clear pattern throughout the heat affected zone.

Similar testing was performed on the butt joint, with a composite light optical micrograph shown in **Figure 21** that demonstrates placement of microhardness traverses. Microhardness was assessed near each weld toe in the butt joint, with values presented in **Figure 22** and **Figure 23** for the first and second pass, respectively. Microhardness values in the heat affected zones of the first pass followed a pattern of high hardness in the CGHAZ, followed by a drop throughout the FGHAZ and ICHAZ before rising slightly to the base metal microhardness. The maximum hardness in the CGHAZ was significantly higher for the indents in traverse 1 than those in traverse 2. This may be due to placement of the weld bead: the two passes were slightly offset, resulting in a greater degree of reheating undergone by traverse 2. This relative softness may have resulted from dissolution of precipitates and consequent grain coarsening. Microhardness in the HAZ of the second weld bead fell from the fusion zone to the ICHAZ. Traverse 3 showed a distinct increase back to base material hardness values, whereas traverse 4 leveled out at an average microhardness of 184 HV₂₀₀. This is likely an effect of measurement location within the weldment, as traverse 3 was positioned closer to the center of the plate, where slower cooling rates resulted in a wide heat affected zone, thus the traverse did not extend far enough to reach base material. The raw data points for **Figures 20, 22, and 23** can be found in **Appendix D**.

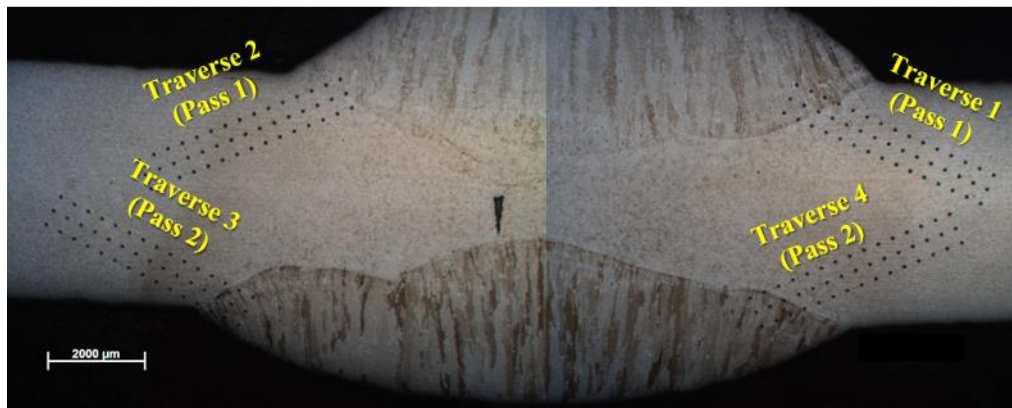


Figure 21. Composite image of light optical micrographs of the HSLA-65 butt joint weldment showing placement of microhardness indents.

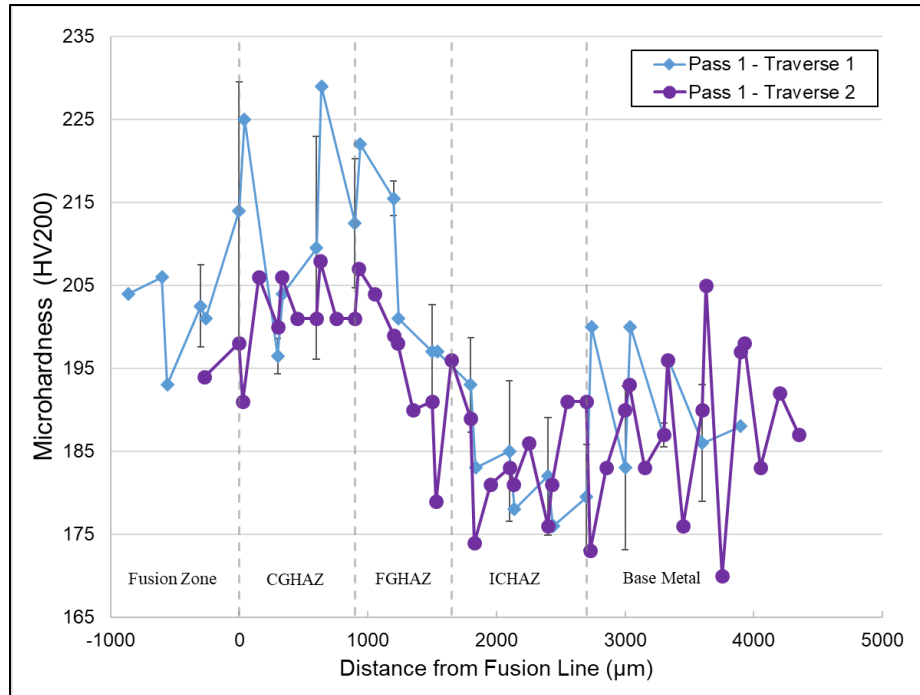


Figure 22. Vickers microhardness traverses corresponding to the indents shown in **Figure 21**. Error bars of one standard deviation are shown where two measurements were taken at the same distance from the fusion line. Dotted lines demarcate HAZ regions, as reported in **Table 17** of **Appendix D**.

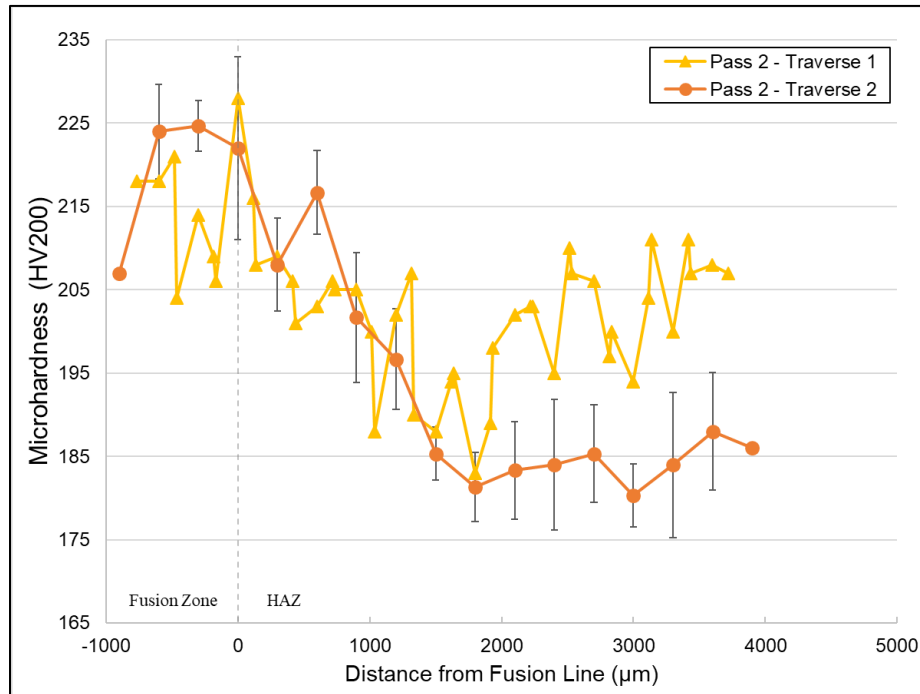


Figure 23. Vickers microhardness traverses corresponding to the indents shown in **Figure 21**. Error bars of one standard deviation are shown where two measurements were taken at the same distance from the fusion line. Dotted line demarcates the fusion line, as reported in **Table 17** of **Appendix D**.

Thermo-Physical Property Analysis

The CTE values measured from the dilation curves (n=3) were $1.5 \times 10^{-5} \pm 5.9 \times 10^{-8} \text{ }^{\circ}\text{C}^{-1}$ ($8.6 \times 10^{-6} \pm 3.3 \times 10^{-8} \text{ }^{\circ}\text{F}^{-1}$) for the untransformed base metal below 650 °C (1202 °F) and $2.0 \times 10^{-5} \pm 8.9 \times 10^{-7} \text{ }^{\circ}\text{C}^{-1}$ ($1.1 \times 10^{-5} \pm 4.9 \times 10^{-7} \text{ }^{\circ}\text{F}^{-1}$) for austenite above 1000 °C (1832 °F). A representative graph showing the CTE measurement is given in **Figure 24**.

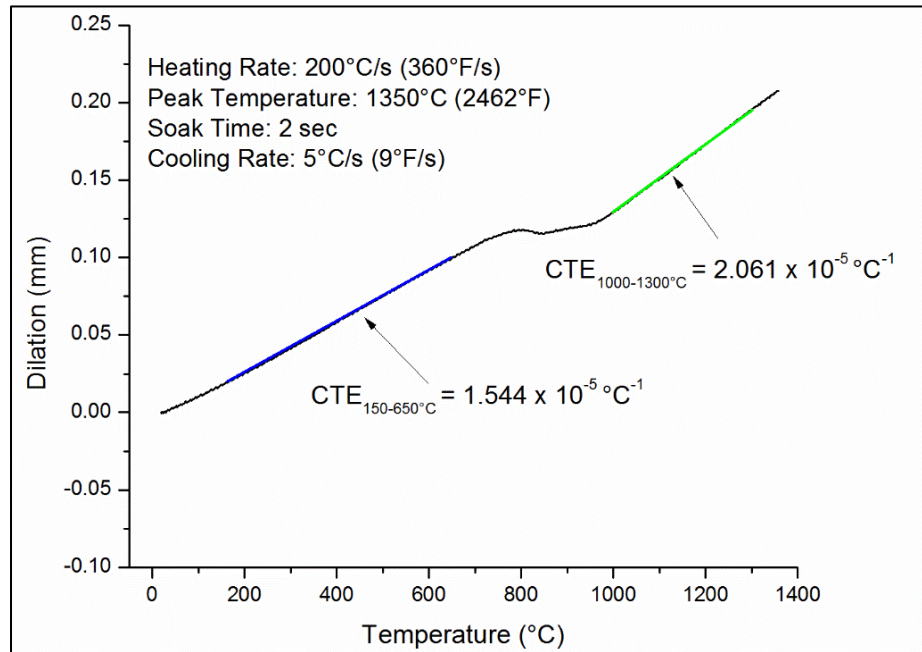


Figure 24. Representative on-heating dilatometry curve for HSLA-65, showing typical CTE analysis and results.

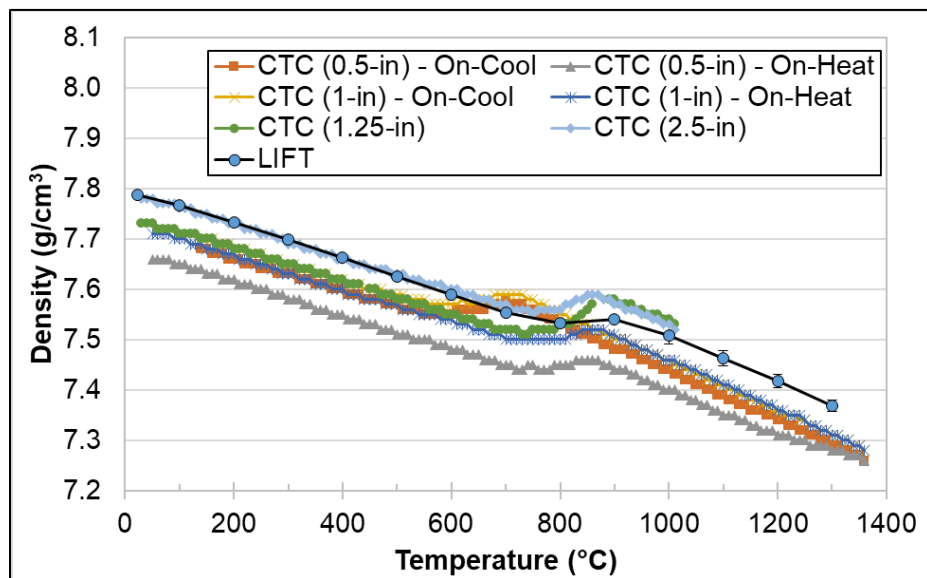


Figure 25. Density of HSLA-65 steel as a function of temperature. LIFT data measured using analysis of Gleeble-based dilation curves in accordance with **Equations 2 and 3**. CTC data is adapted from references [14] [15]. Error bars on LIFT data are one standard deviation from the mean of three tests.

Temperature-dependent density values generated in this work and reproduced from the ManTech program described previously [14] [15] are shown in **Figure 25**. The ManTech-generated data is labeled “CTC” and includes the nominal thickness of the material tested (in inches) and whether the measurements were taken on heating or on cooling. Because the slopes of the lines (*i.e.*, the CTE) are nearly identical, differences between the data can be primarily attributed to a difference in the measured room temperature densities between the three studies.

The measured specific heat and thermal diffusivity data for HSLA-65 can be found in **Figures 26-27**. The peaks or cusps in the data correspond to the effects of thermal energy absorption during phase transformations, particularly during the austenitic transformation between approximately 700 to 900 °C (1292 to 1652 °F) and the δ ferrite and melting transformations beginning around 1400 °C (2552 °F). The peaks in **Figure 26** for the two on-cooling datasets match poorly with specific heat data generated in this study, as the on-heating datasets were influenced by different transformation behavior. The δ ferrite transformation is only visible in **Figure 26** (beginning at approximately 1300 °C [2372 °F]) and is missed by **Figure 27** because of the larger temperature increment between thermal diffusivity measurements. Thermo-Calc equilibrium calculations using the TCFe9 database [25] predict a δ -ferrite stability range of 1452 to 1520 °C (2646 to 2768 °F). As shown, the data from this program compare very well with the data generated in the Navy ManTech studies.

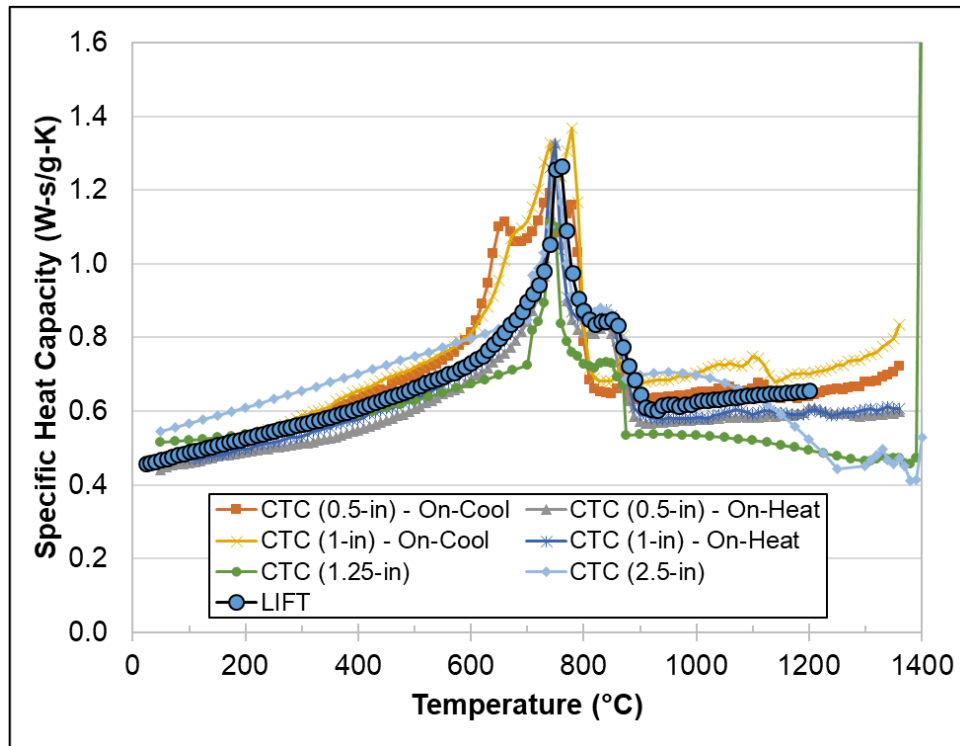


Figure 26. Measured specific heat for HSLA-65 steel at various temperatures. CTC data adapted from references [14] [15].

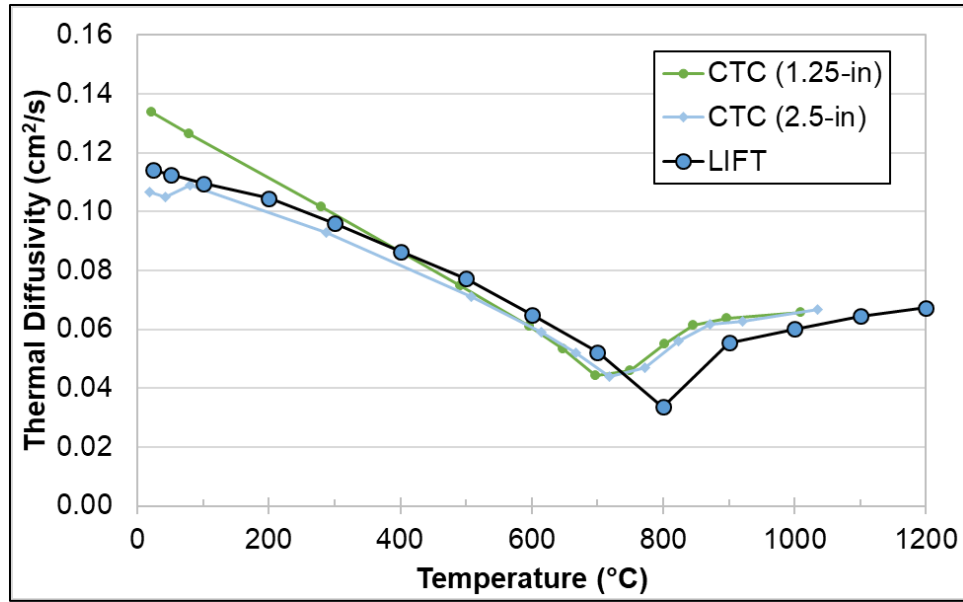


Figure 27. Measured thermal diffusivity for HSLA-65 steel at various temperatures. CTC data adapted from reference [14].

Finally, the data from **Figures 25-27** were used in conjunction with **Equation 4** to calculate the temperature-dependent thermal conductivity as shown in **Figure 28**. Once again, the data coincides well with the previous reporting from CTC. The raw data points for all of the thermo-physical property graphs can be found in **Appendix E**.

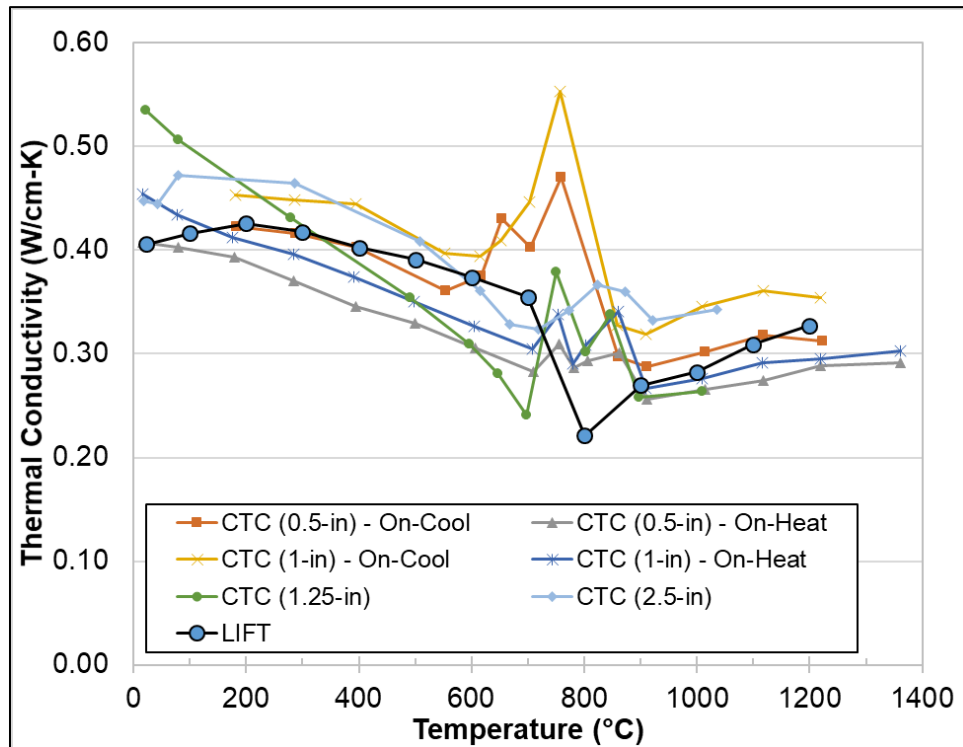


Figure 28. Calculated thermal conductivity for HSLA-65 steel at various temperatures. CTC data adapted from references [14] [15].

Thermo-Mechanical Property Analysis

Figure 29 shows the values for elastic modulus as a function of temperature that were assumed in this work. The elevated temperature mechanical testing described above was performed in accordance with ASTM E21 [11] rather than the stricter ASTM E111 [39], which is prescribed when elastic modulus measurements are to be made. As such, it was determined that the experimentally measured elevated temperature elastic moduli were likely inaccurate. In order to provide elevated temperature modulus values for a HSLA-65 property database, the data in **Figure 29** were adapted from the European standard for fire design of steel structures [40] [41]. Of the potential sources for this information, the European standard contained the most complete data covering temperatures relevant to this study. A room temperature elastic modulus of 210 GPa (30.5 Msi) was assumed [40], as it had been previously for DH36 [5]. It was also assumed that rigorously measured elastic moduli for HSLA-65 steel would be very similar to data presented in **Figure 29** because elastic properties tend to be consistent within a given material class (*e.g.*, steel). **Figure 30** shows the experimentally measured temperature-dependent 0.2% offset yield strength for HSLA-65 base material, and **Figure 31** shows the measured temperature-dependent ultimate tensile strength (UTS). Comparison data for the measured yield and tensile strengths were found in one of the CTC reports, albeit for slightly thicker material than was used in this study [15]. As expected, increased temperatures result in significant strength loss. At temperatures above 800 °C (1472 °F), the base metal microstructure is completely transformed to the weaker austenite phase, resulting in only minor differences in strength with increasing temperature.

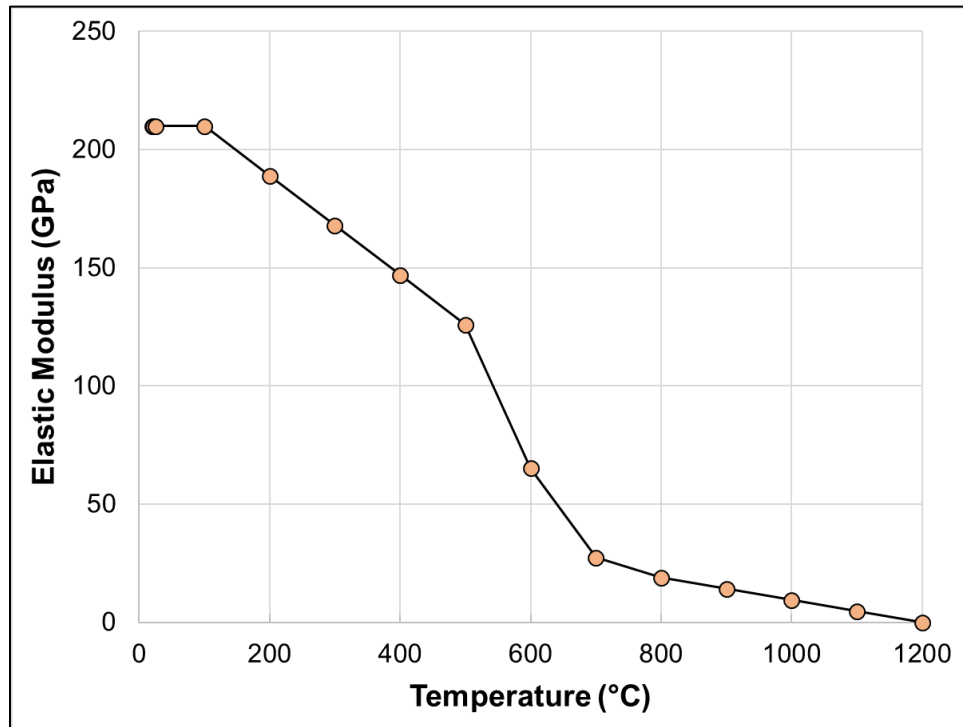


Figure 29. Assumed elevated temperature elastic modulus for HSLA-65 steel as adapted from references [40] [41].

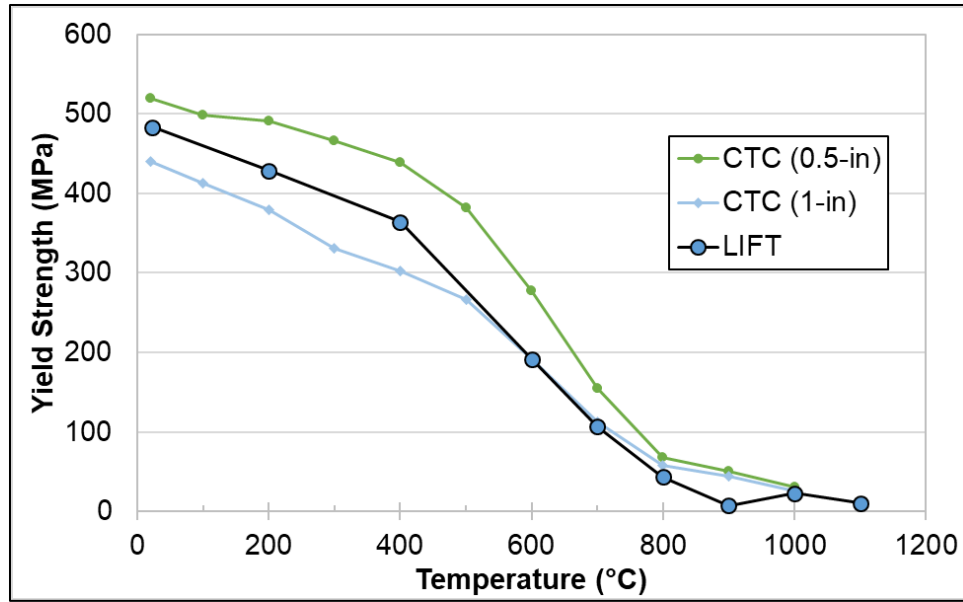


Figure 30. Measured elevated temperature 0.2% offset yield strength for HSLA-65 steel. CTC data adapted from reference [15].

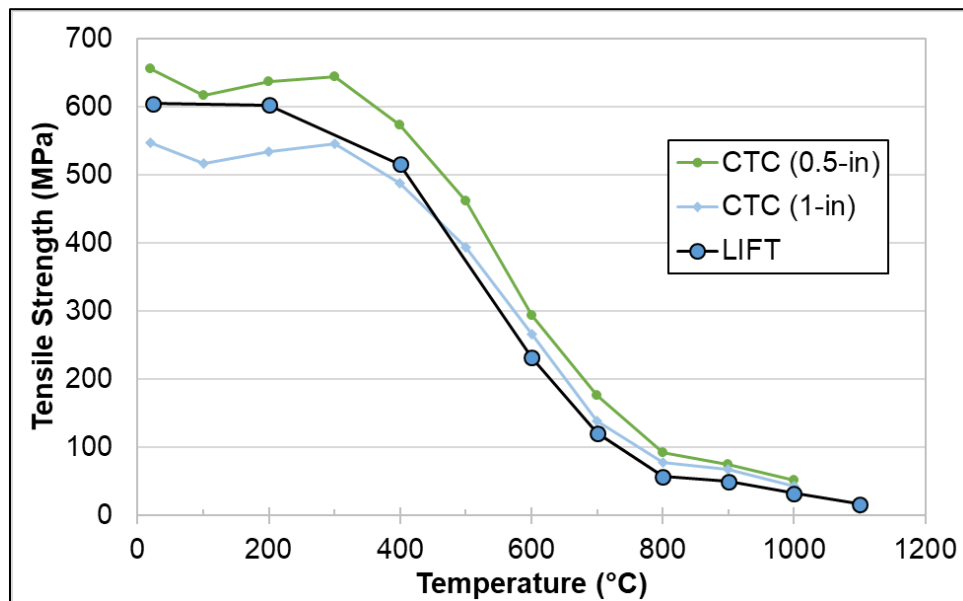


Figure 31. Measured elevated temperature ultimate tensile strength for HSLA-65 steel. CTC data adapted from reference [15].

Figure 32 shows the measured yield strengths from specimens cycled to a peak temperature of 1350 °C (2462 °F) along with that of the base material yield strength for comparison. The stress-strain curves associated with the data are given in **Figures 72-75** of **Appendix F**. At room temperature, the yield strength of the 1 °C/s (1.8 °F/s) specimen was nominally the same as the base material. This is consistent with their similar microstructures. The 100 °C/s (180 °F/s) specimen was over 50% stronger than the 1 °C/s (1.8 °F/s) specimen, with yield strength of 10 and 25 °C/s (18 and 45 °F/s) specimens falling between the two extremes. These observations are consistent with the change in transformation products from various ferrite and pearlite morphologies to increasing fractions of martensite and bainite. Typically, increases in cooling rate are associated with higher yield strengths in steels. This trend was

generally observed except at room temperature, where the specimen cooled at 10 °C/s (18 °F/s) had a yield strength that was almost 17% higher than that of the 25 °C/s (45 °F/s) specimen. At 200 °C (392 °F), the 1, 10, and 100 °C/s (1.8, 18, and 180 °F/s) specimens experienced virtually no change in yield strength from room temperature, but the 25 °C/s (45 °F/s) specimen saw a 17% increase. It is unclear why this occurred. The 25 °C/s (45 °F/s) specimen starting microstructure may have experienced some additional carbide precipitation during reheating to 200 °C (392 °F), but confirmation would require additional microstructural analysis. In all other cases, an increase in testing temperature was associated with a decrease in yield strength for a given cooling rate.

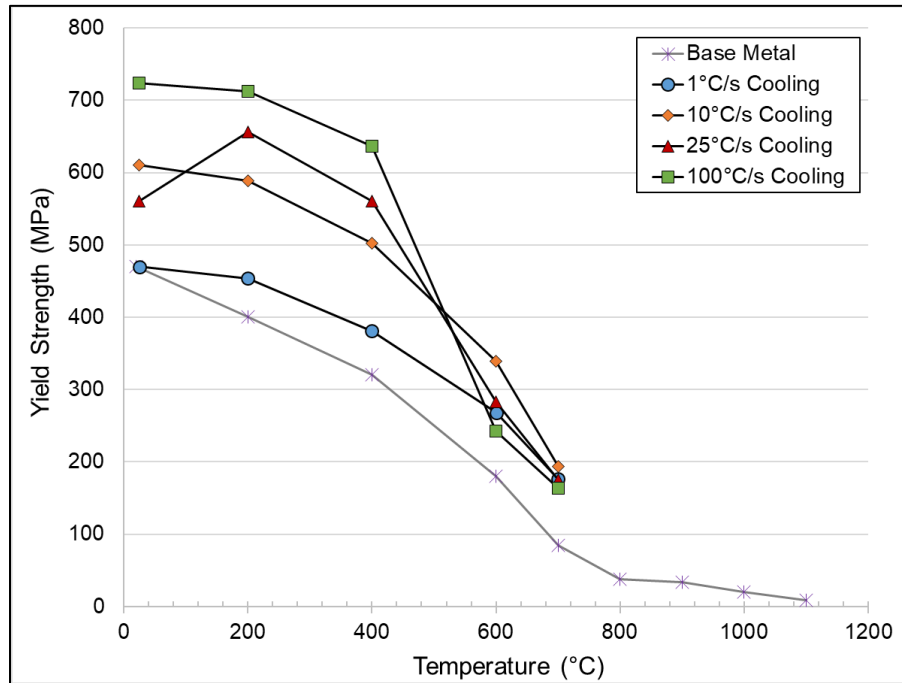


Figure 32. Yield strength of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at different rates as compared with base material yield strength.

As indicated in **Figure 32**, there was a monotonic decrease in yield strength with increasing temperature for all specimens above 200 °C (392 °F). The effect of a stronger starting microstructure was wiped out for the 100 °C/s (180 °F/s) specimens above 400 °C (752 °F) as their strength converged with those of the 10 °C/s (18 °F/s) samples. The yield strengths of all specimens then converged to that of the 1 °C/s (1.8 °F/s) specimen at 700 °C (1292 °F), though they remained slightly stronger than the base material.

Elevated temperature flow stress measurements for the HSLA-65 base material are shown in **Figure 33** for all temperatures, and **Figure 34** highlights the flow behavior for tests at temperatures above 700 °C (1292 °F). The raw data points for these graphs can be found in **Appendix F**. The data are true stresses and strains calculated by applying the traditional conversion equations to engineering stress-strain data provided by the test lab. Extensometers were used through yield, but for the protection of laboratory equipment, extensometers were removed from each specimen between 0.7 and 3.6% strain at the discretion of the testing personnel. For all cases except where the test temperature was 800 °C (1472 °F), this removal occurred well before UTS was reached. As a result, the flow stress behavior is not fully characterized for those samples. As expected, increasing the test temperature tends to flatten the flow stress curves, since work hardening is made more difficult by dynamic recovery and/or recrystallization effects.

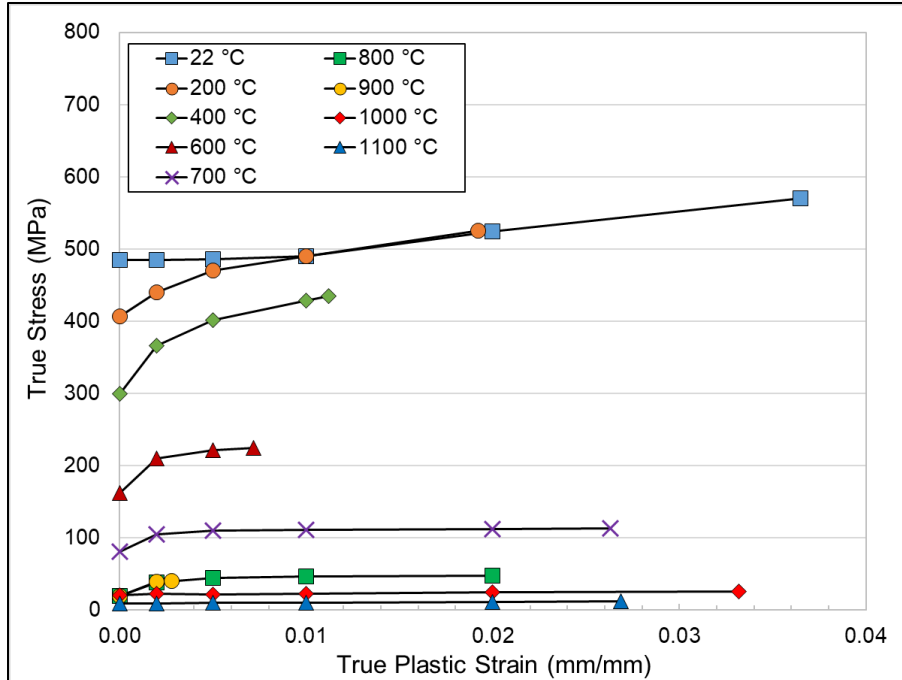


Figure 33. On-heating flow stress behavior for HSLA-65 steel at various temperatures.

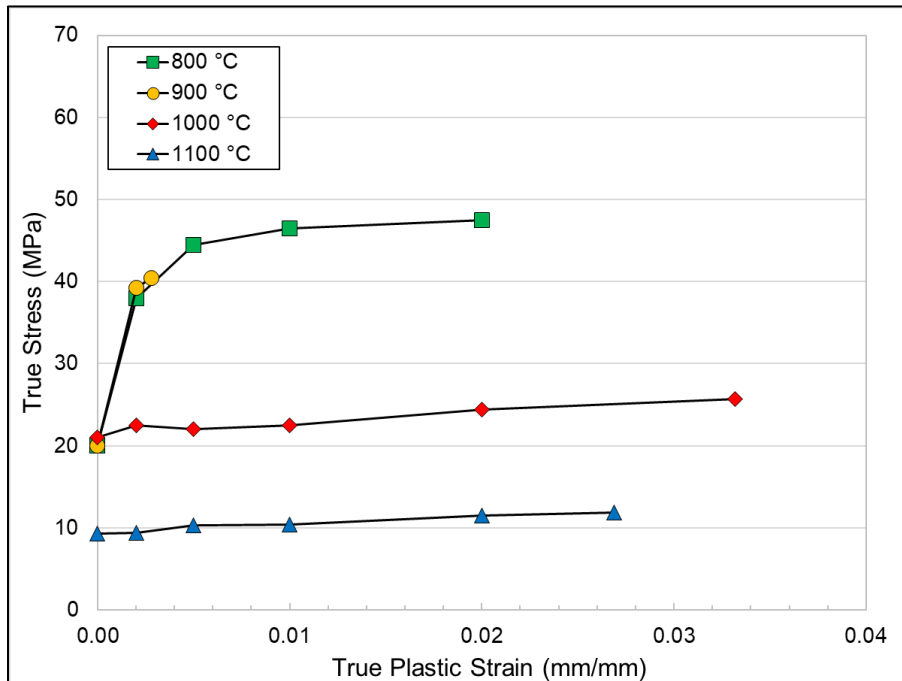


Figure 34. On-heating flow stress behavior for HSLA-65 steel at testing temperatures above the austenitic phase transformation.

Flow stress curves for the thermally cycled HSLA-65 are given in **Figures 35-38**. Since these tests were performed under inert atmosphere, the extensometer was not removed during the tests and more complete flow stress information is available. Once again, the data are true stresses and strains calculated by conversion from engineering stress-strain data. Such conversions are invalid past the onset

of necking, so the terminal data points in **Figures 35-38** are at the UTS. All raw data points for these figures, along with the engineering fracture strains for the specimens, are found in **Appendix F**.

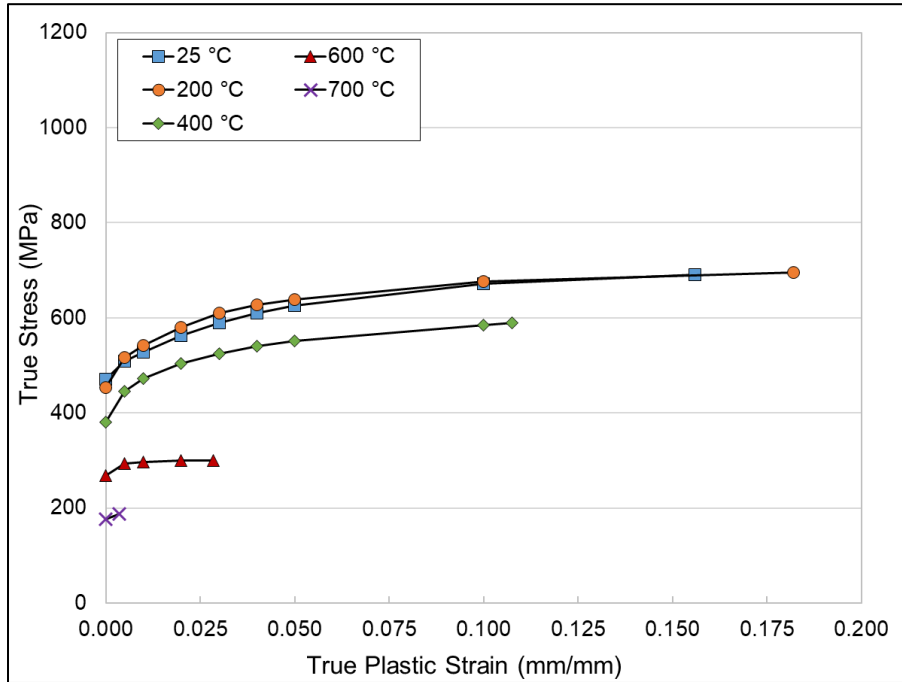


Figure 35. Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 1 °C/s (1.8 °F/s) from a peak temperature of 1350 °C (2462 °F). Terminal data points are at the UTS.

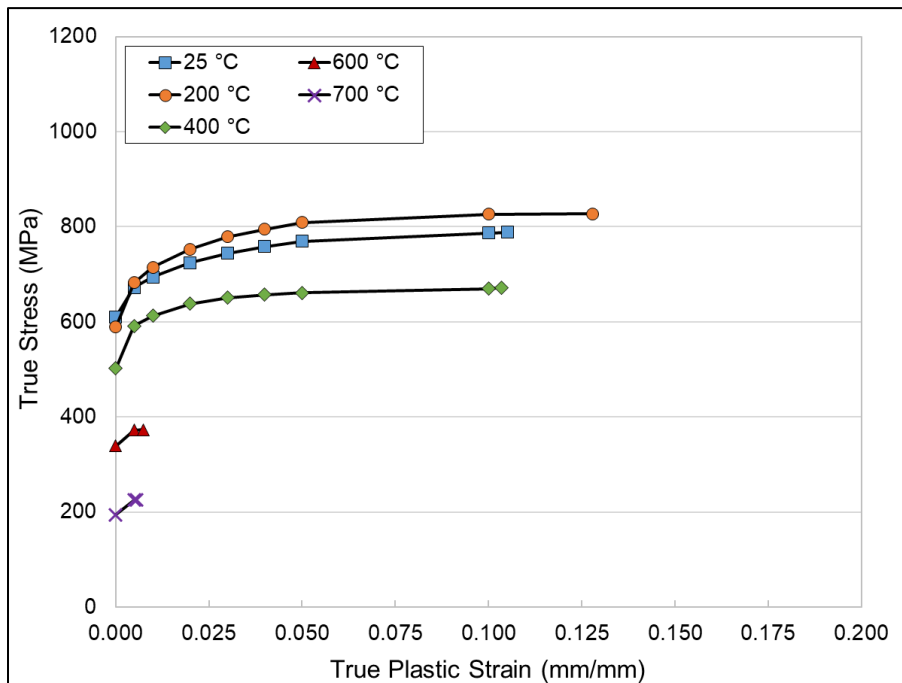


Figure 36. Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 10 °C/s (18 °F/s) from a peak temperature of 1350 °C (2462 °F). Terminal data points are at the UTS.

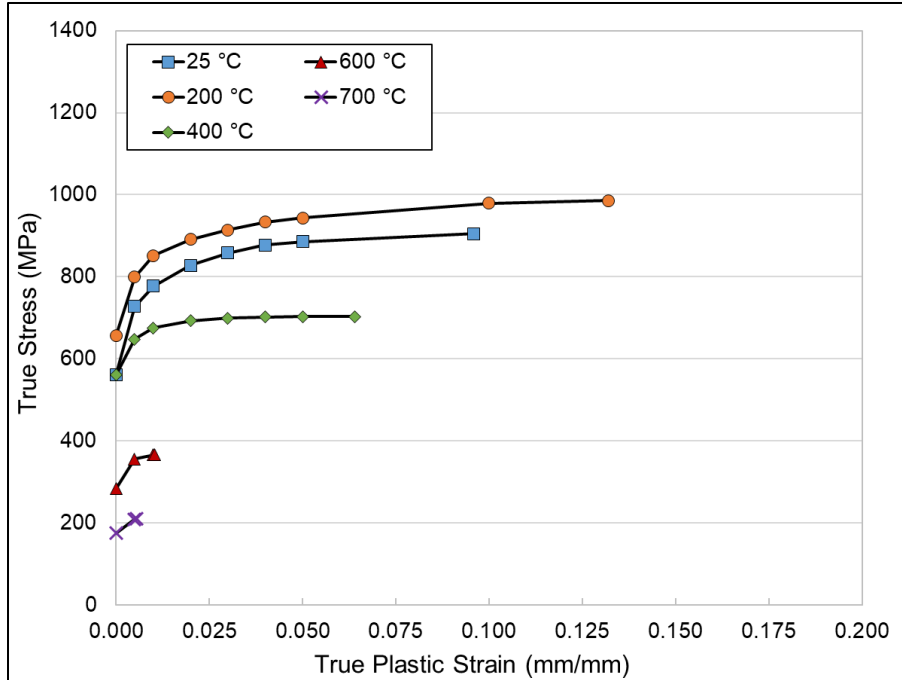


Figure 37. Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 25 °C/s (45 °F/s) from a peak temperature of 1350 °C (2462 °F). Terminal data points are at the UTS.

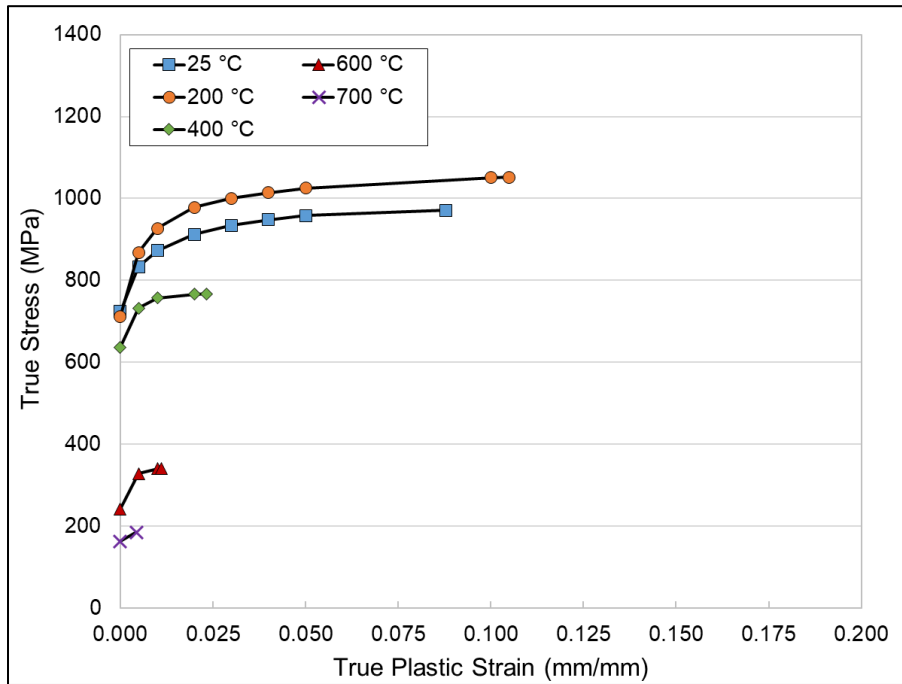


Figure 38. Flow stress behavior at various temperatures for HSLA-65 steel after cooling at 100 °C/s (180 °F/s) from a peak temperature of 1350 °C (2462 °F). Terminal data points are at the UTS.

SUMMARY

Temperature-dependent material property data of a pedigreed plate of HSLA-65 steel from room temperature up to nearly the steel's melting point were determined. The thermo-physical properties investigated include specific heat, thermal diffusivity, thermal conductivity, coefficient of thermal expansion, and density. Thermo-mechanical properties including elastic modulus, yield strength, and flow stress were also measured. The temperatures associated with on-heating and on-cooling phase transformations and their variation with heating rate, cooling rate, and peak temperature were determined and used to develop welding-focused continuous cooling transformation (CCT) diagrams. Investigation of heat-affected zone microstructures from the CCT specimens and arc welds concluded the analysis. This effort is essential for increasing the fidelity of finite element models used to predict welding-induced distortion and residual stress in marine structures. The data generated in this program have been provided to ESI for immediate incorporation into their *SYSWELD* software. A machine-readable version of the collected data will be uploaded to the University of Michigan's *Materials Commons* data repository <<https://materialscommons.org>> at a later date.

APPENDICES

Appendix A: Plate Conformance Certificate & Chemical Composition for HSLA-65 Steel

ArcelorMittal Burns Harbor Plate																
SHIPMENT NO.					DATE SHIPPED					CAR OR VEHICLE NO.						
803-07049					06-03-16					TRLR 2						
HUNTINGTON INGALLS INDUSTRIES INC					HUNTINGTON INGALLS INDUSTRIES INC					PAGE 1						
INGALLS SHIPBUILDING DIV					INGALLS SHIPBUILDING DIV											
PO BOX 149					C/O METALS USA PLATES & SHAPES SE I											
PASCAGOULA MS 39568-0149					1 FOUNDRY RD											
					WAGGAMAN LA 70094-2349											
S E R I A L N O	P A T N O	H E A T N O	N O. P C S	SIZE AND QUANTITY				Y I E L D P O I N T	T E N S I L E S T R E N G T H	E L O N G.	R E D.					
				T H I C K N E S S	W I D T H O R D I A	L E N G T H	W E I G H T									
				INCHES	INCHES	INCHES	POUNDS	PSI	PSI	IN	%					
QUALITY STEEL MELTED & MANUFACTURED IN THE U. S. A.																
PLATES - ASTM A945-06E1 GR 65 KLD FINE GRAIN																
PRAC, CH-V A673 FREQ (P) T 70 FTLB																
AT -40F --- PLT CONTROL ROLL																
NO WELD REPAIR WAS PERFORMED ON BELOW PLATE(S)																
MFST - MFST MILL SERIAL# MFST MILL SERIAL# & PATTERN#																
MFST PPI 0045046- 0001 LIFT MAX 30 TON UNLDG																
OH-MAGNET																
CO# 903853 GH 823-0364																
OUTSIDE INSPECTION BY AMERICAN BUREAU OF SHIPPING																
B259514	151P16170	1	.1875	96	300	1531	74300	90500	2	40						
(M67)MFST CUST ITEM NO:001																
B259212	152P33580	1	.1875	96	300	1531	76400	87900	2	40						
(M67)MFST CUST ITEM NO:001																
Q-QUENCH TEMPERATURE																
T-TEMPER TEMPERATURE																
N-NORMALIZE TEMPERATURE																
WE HEREBY CERTIFY THAT THE MATERIAL DESCRIBED HEREIN HAS BEEN MADE BY THE																
ARCELORMITTAL BH PLNT TO THE APPLICABLE SPECIFICATION BY AN APPROVED PROCESS																
AND HAS BEEN TESTED IN ACCORDANCE WITH THE REQUIREMENTS OF THE AMERICAN																
BUREAU OF SHIPPING RULES FOR THE INSPECTION AND TESTING OF MATERIALS TO THE																
SATISFACTION OF THE SURVEYORS.																
S E R I A L N O	P A T N O	H E A T N O	H A R D B H N	B E N D	T H I C K N E S S INCHES	T Y P E	S I Z E	D I R	T E S T T E M P F	C H A R P Y I M P A C T			S H E A R(%)	L A T. E X P	M I L S	
										1	2	3				
B259514	151P16170				.187	V	1/3	T	-40	34	32	31				
B259212	152P33580				.187	V	1/3	T	-40	37	40	38				
C H E M I C A L A N A L Y S I S																
H E A T N O	C	Mn	P	S	Si	Cu	Ni	Cr	Mo	V	Ti	Al	B	Cb	N	Sn
151P16170	.08	1.49	.007	.001	.281	.010	.01	.02	.007	.075	.013	.032	.0003	.029	.005	.001
152P33580	.08	1.49	.010	.001	.277	.007	.01	.03	.006	.071	.012	.030	.0002	.025	.005	.001
I certify that the above results are a true and correct copy of actual results contained in records maintained by ArcelorMittal Burns Harbor and are in full compliance with the requirements of the specification cited above. This test report cannot be altered and must be transmitted intact with any subsequent third party test reports, if required.																
BHLTRPT.TF																
SUPV. QUALITY ASSURANCE																
R. SPANGLER II LSS																

Figure 39. Plate conformance certificate provided by the manufacturer for HSLA-65 material used in this report.

Table 9. Chemical Composition of HSLA-65 Base Plate as Measured by OSU. All values in wt. %.

Element	Test 1	Test 2	Test 3	Test 4	Test 5	Average
C	0.082	0.089	0.087	0.084	0.086	0.086
Si	0.27	0.27	0.27	0.27	0.270	0.27
Mn	1.47	1.48	1.49	1.49	1.480	1.48
P	0.009	0.009	0.008	0.009	0.009	0.009
S	<0.0005	<0.0005	<0.0005	<0.0005	<0.0005	---
Cr	0.026	0.026	0.026	0.026	0.026	0.026
Mo	0.006	0.006	0.006	0.005	0.005	0.006
Ni	0.014	0.014	0.013	0.013	0.014	0.014
Al	0.028	0.028	0.029	0.029	0.029	0.029
Co	0.003	0.003	0.003	0.003	0.003	0.003
Cu	0.006	0.006	0.006	0.006	0.006	0.006
Nb	0.030	0.031	0.030	0.030	0.031	0.030
Ti	0.012	0.012	0.012	0.012	0.012	0.012
V	0.079	0.080	0.079	0.079	0.079	0.079
W	<0.005	<0.005	<0.005	<0.005	<0.005	---
Pb	0.003	0.003	0.003	0.003	0.003	0.003
Sn	0.0006	0.0006	0.0006	0.0006	0.0006	0.0006
As	0.004	0.005	0.004	0.004	0.004	0.0042
Zr	0.001	0.001	0.001	0.001	0.001	0.001
Bi	0.001	<.001	<.001	<.001	0.001	0.001
Ca	0.003	0.003	0.003	0.002	0.002	0.003
Ce	<0.001	<0.001	<0.001	<0.001	<0.001	---
Sb	0.021	0.021	0.021	0.021	0.021	0.021
Se	0.004	0.003	0.004	0.004	0.004	0.004
Te	0.026	0.025	0.025	0.025	0.025	0.025
Ta	0.041	0.05	0.048	0.042	0.045	0.045
B	0.001	0.001	0.001	0.001	0.001	0.001
Zn	0.002	0.002	0.002	0.002	0.002	0.002
La	<0.0003	<0.0003	<0.0003	<0.0003	<0.0003	---
N	0.006	0.006	0.006	0.006	0.006	0.006
Fe	97.8	97.8	97.8	97.8	97.8	97.8

Appendix B: Continuous Cooling Transformation Curves

Table 10. Austenite transformation temperatures illustrated in **Figure 7**.

Heating Rate (°C/s)	A _{c1} (°C)	A _{c3} (°C)
10	710	903
100	718	886
200	723	903
500	759	895
1000	760	917
2000	799	881
CTC: Nominal Plate Thickness (mm [in])	A _{c1} (°C)	A _{c3} (°C)
12.7 (0.5)	725	860
25.4 (1.0)	721	860
31.75 (1.25)	717	892
63.5 (2.5)	729	881

Table 11. Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 825 °C (1517 °F). Colors correspond to the transformation products discussed in **Figures 9-15**.

Cooling Rate (°C/s) [°F/s]	1 [1.8]	5 [9]	10 [18]	25 [45]	100 [180]	200 [360]
Transformation Temperature (°C) [°F]	775 [1427]	761 [1402]	761 [1402]	758 [1396]	Not Detected	414 [777]
	704 [1299]	682 [1260]	626 [1159]	617 [1143]		333 [631]
	659 [1218]	623 [1153]				

Table 12. Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 1000 °C (1832 °F). Colors correspond to the transformation products discussed in **Figures 9-15**.

Cooling Rate (°C/s) [°F/s]	1 [1.8]	5 [9]	10 [18]	25 [45]	100 [180]	200 [360]
Transformation Temperature (°C) [°F]	771 [1420]	731 [1348]	683 [1261]	702 [1296]	640 [1184]	615 [1139]
	725 [1337]	680 [1256]	595 [1103]	585 [1085]	447 [837]	541 [1006]
	688 [1270]	639 [1182]	540 [1004]	499 [930]	362 [684]	458 [856]
						351 [664]

Table 13. Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 1150 °C (2102 °F). Colors correspond to the transformation products discussed in **Figures 9-15**.

Cooling Rate (°C/s) [°F/s]	1 [1.8]	5 [9]	10 [18]	25 [45]	100 [180]	200 [360]
Transformation Temperature (°C) [°F]	740 [1364]	669 [1236]	626 [1159]	546 [1015]	477 [891]	476 [889]
	633 [1171]	574 [1065]	514 [957]	480 [896]	346 [655]	353 [667]
	571 [1060]	538 [1000]	489 [912]	388 [730]		

Table 14. Experimentally measured on-cooling transformation temperatures for HSLA-65 dilatometry specimens cooled from 1350 °C (2462°F). Colors correspond to the transformation products discussed in **Figures 9-15**.

Cooling Rate (°C/s) [°F/s]	1 [1.8]	5 [9]	10 [18]	25 [45]	100 [180]	200 [360]
Transformation Temperature (°C) [°F]	709 [1308]	641 [1186]	619 [1146]	606 [1123]	482 [900]	495 [923]
	577 [1071]	568 [1054]	485 [905]	493 [919]	361 [682]	362 [684]
	511 [952]	509 [948]		406 [763]		

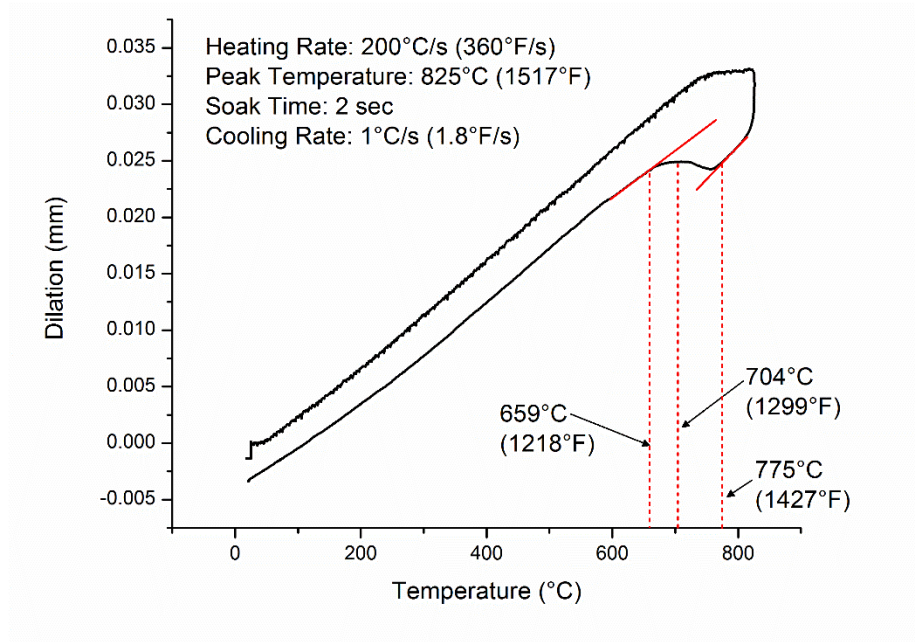


Figure 40. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 1 °C/s.

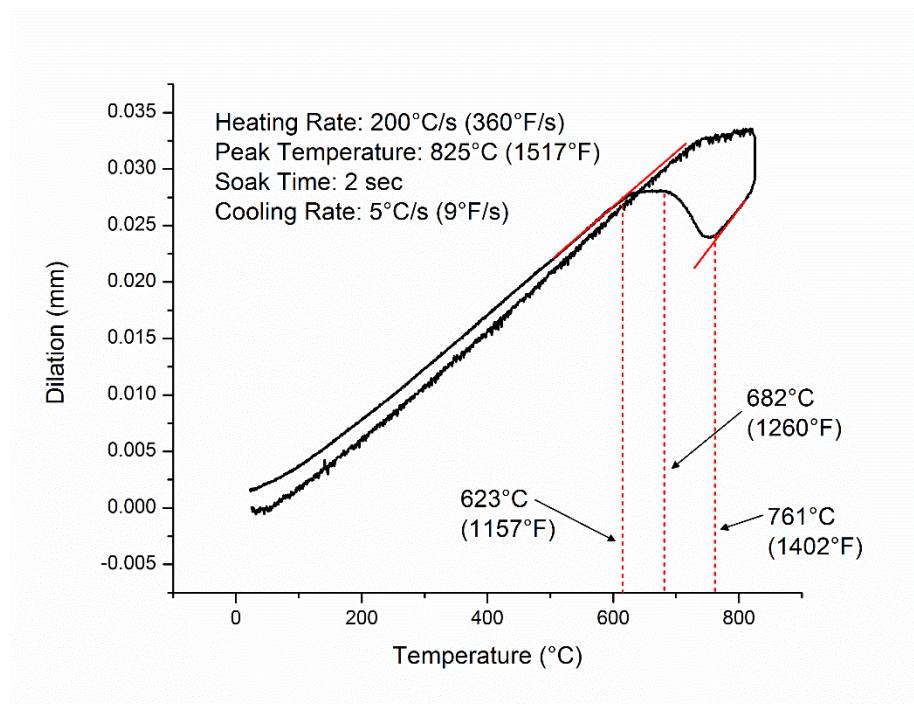


Figure 41. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 5 °C/s.

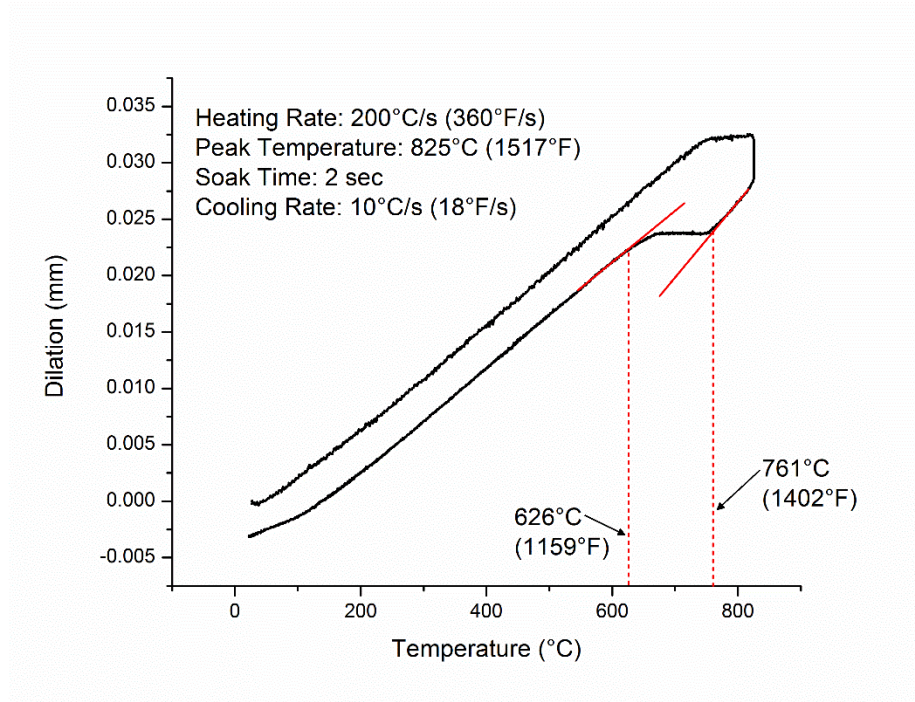


Figure 42. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 10 °C/s.

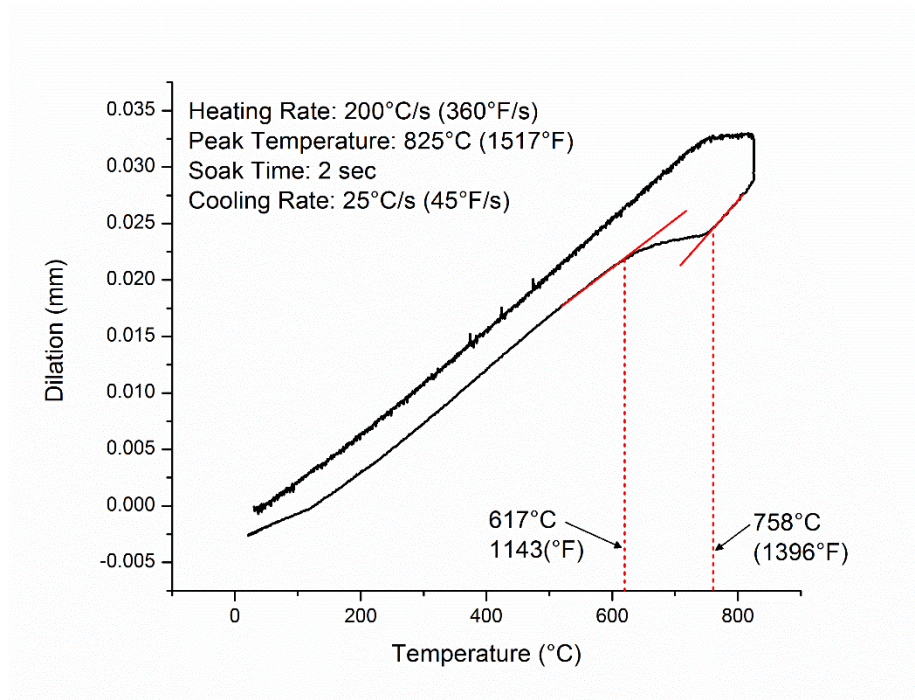


Figure 43. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 25 °C/s.

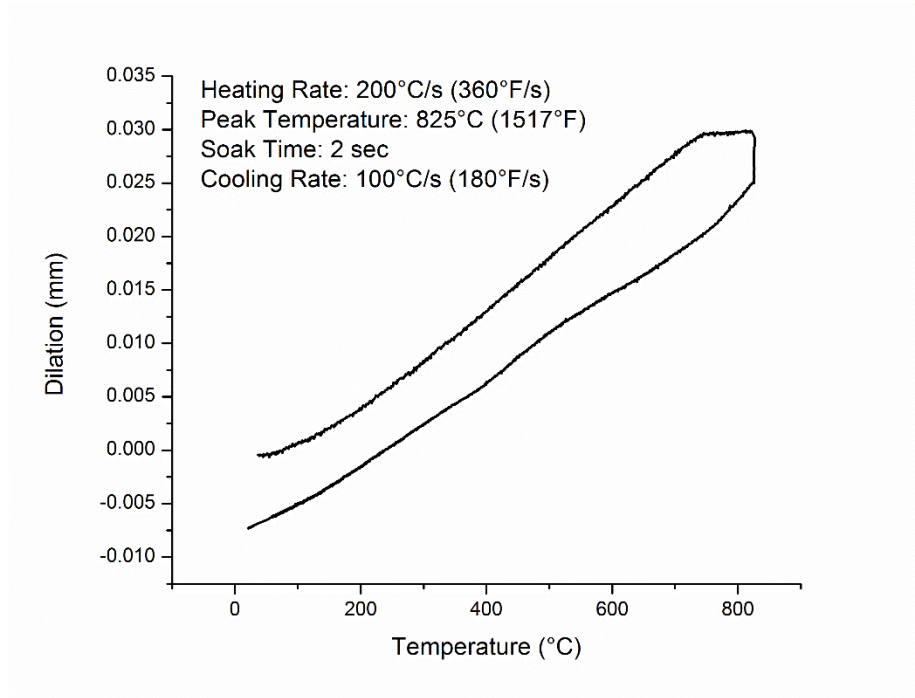


Figure 44. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 100 °C/s.

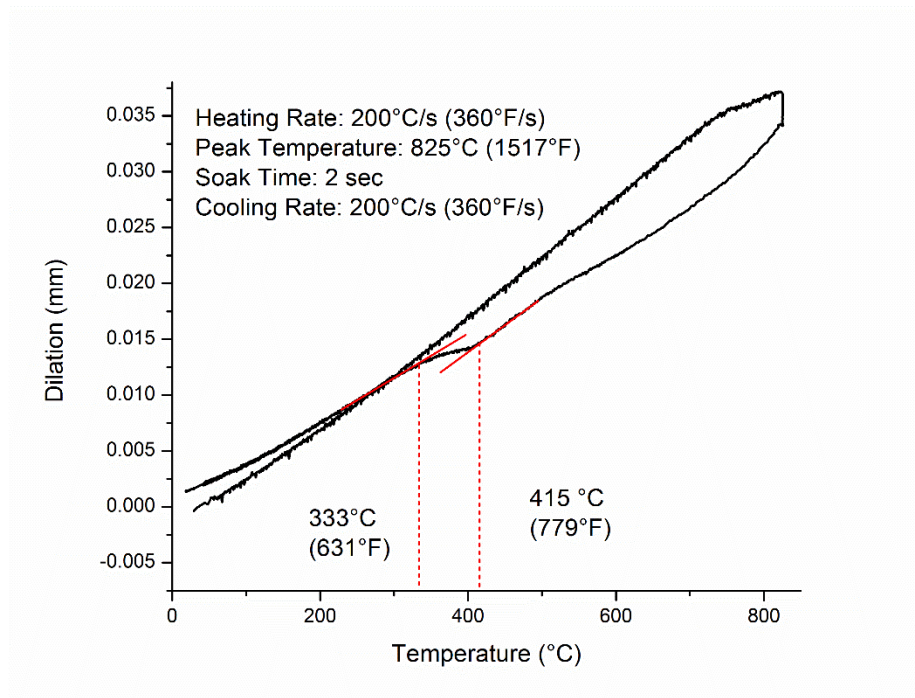


Figure 45. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 825 °C and cooled at 200 °C/s.

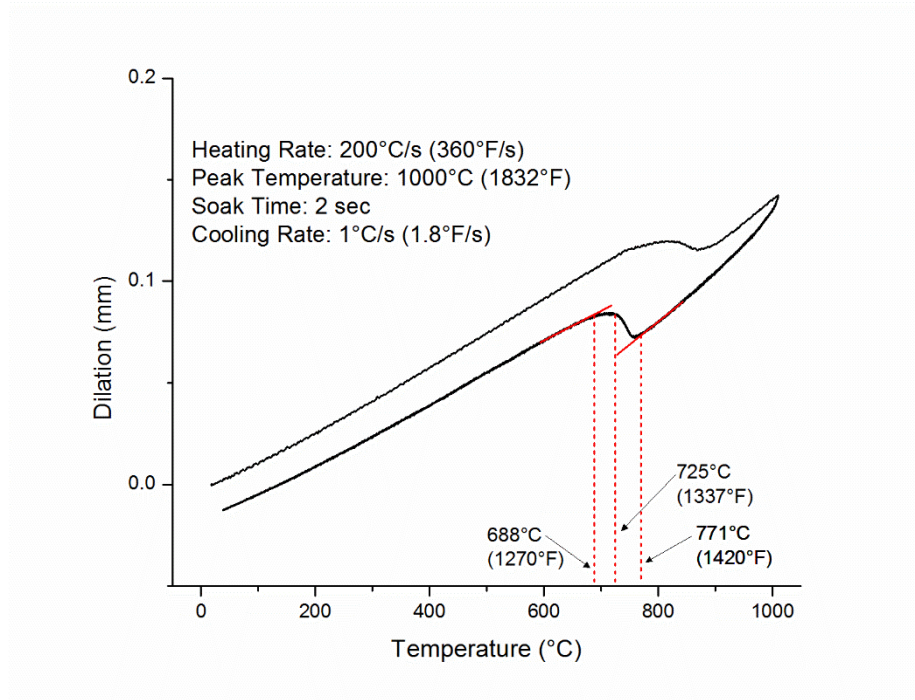


Figure 46. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 1 °C/s.

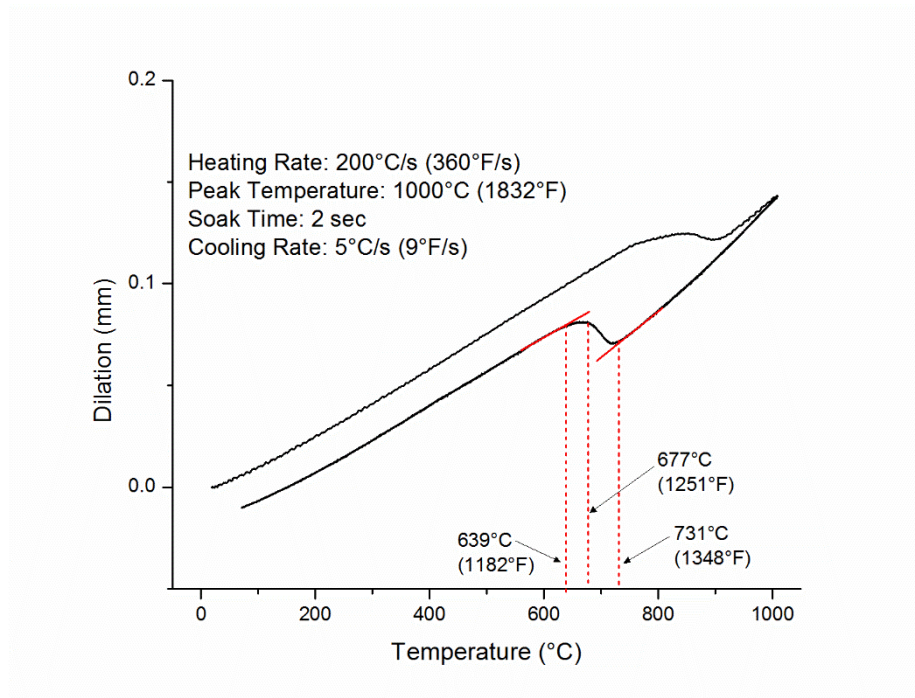


Figure 47. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 5 °C/s.

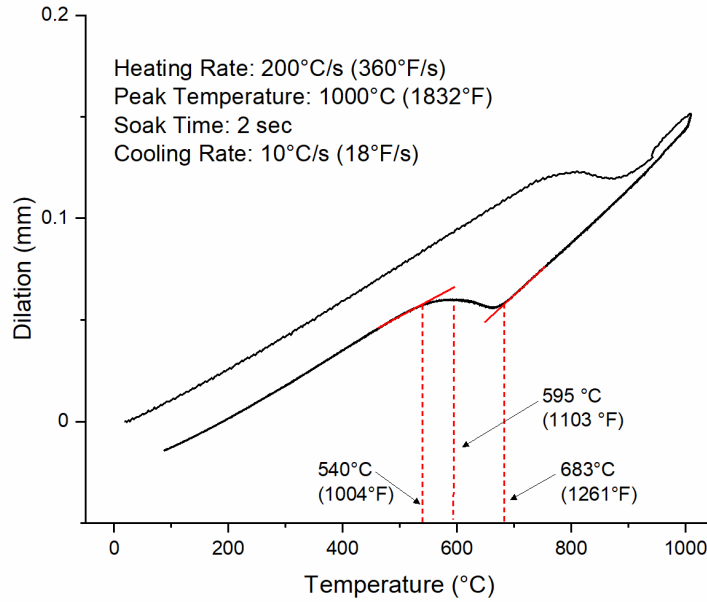


Figure 48. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 10 °C/s.

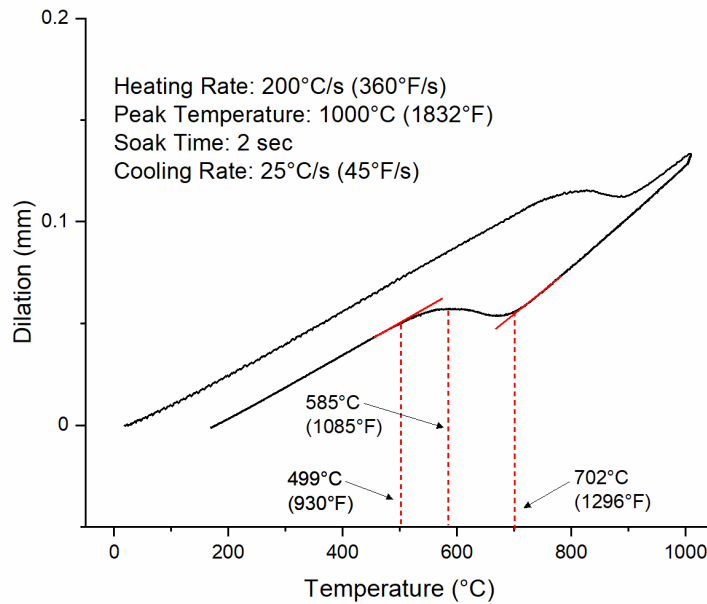


Figure 49. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 25 °C/s.

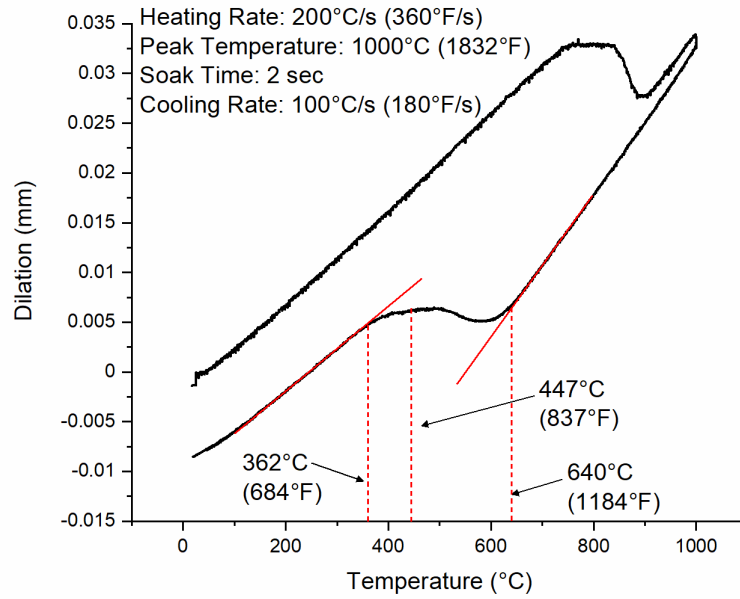


Figure 50. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 100 °C/s.

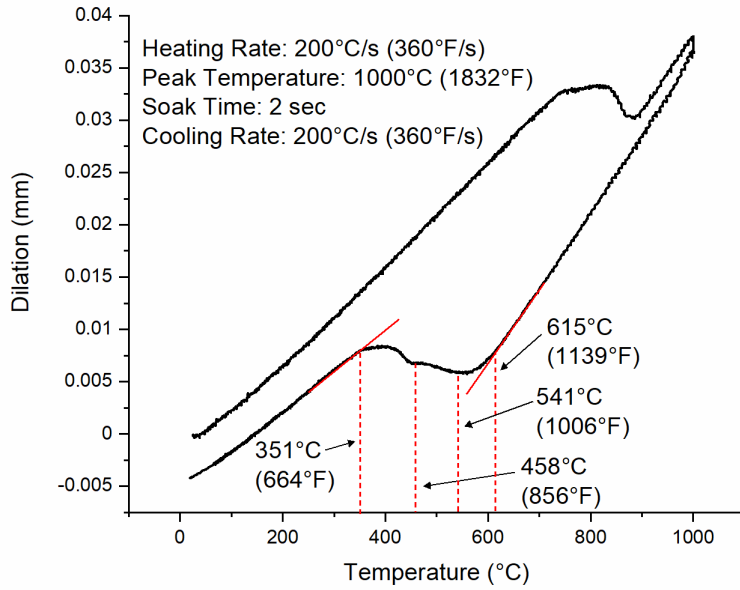


Figure 51. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1000 °C and cooled at 200 °C/s.

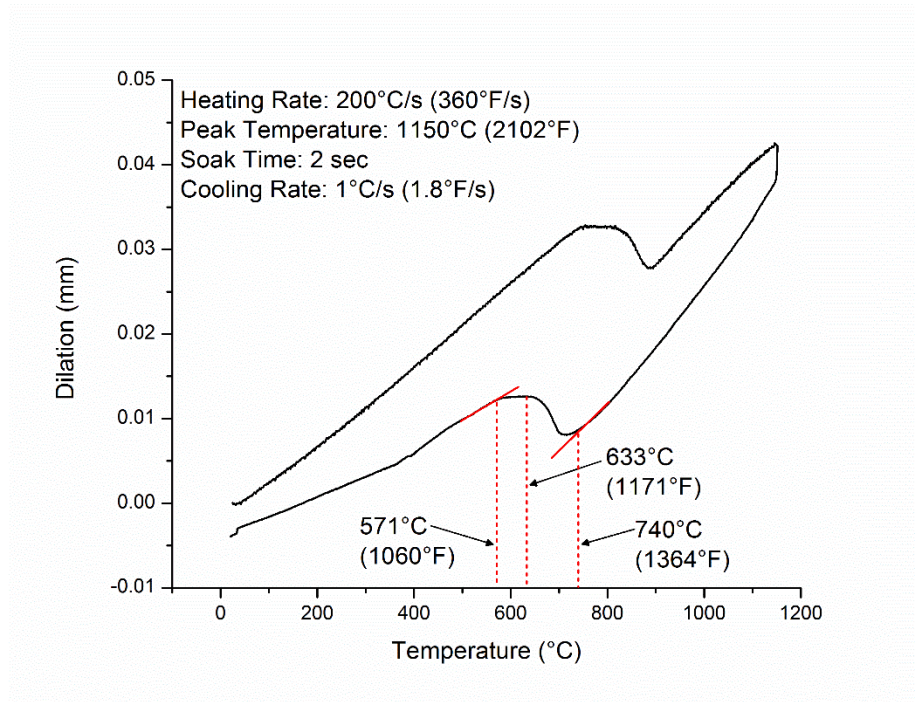


Figure 52. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 1 °C/s.

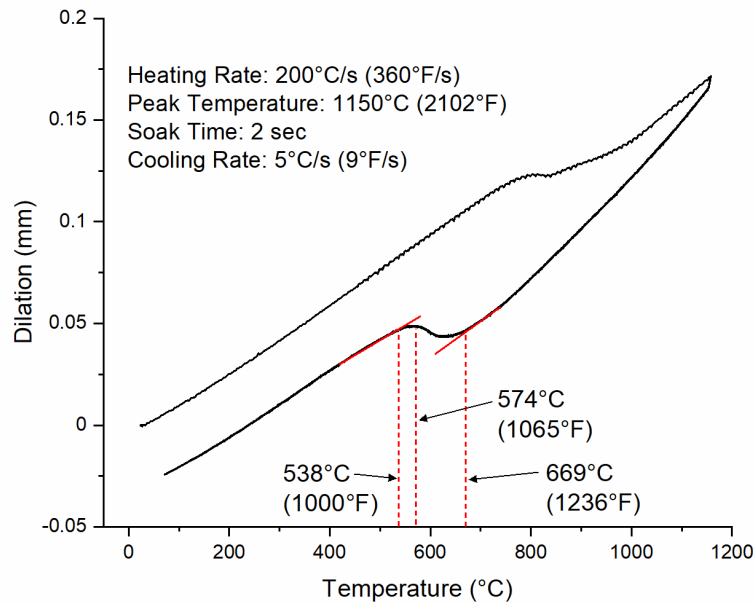


Figure 53. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 5 °C/s.

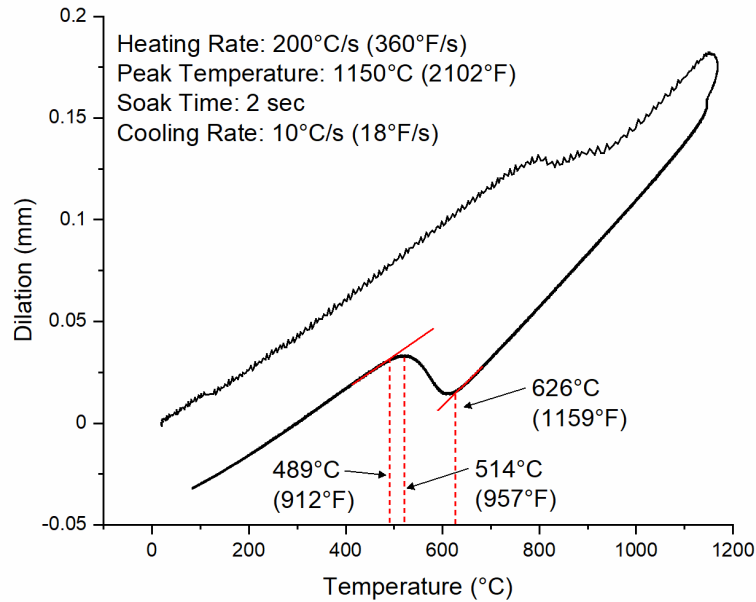


Figure 54. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 10 °C/s.

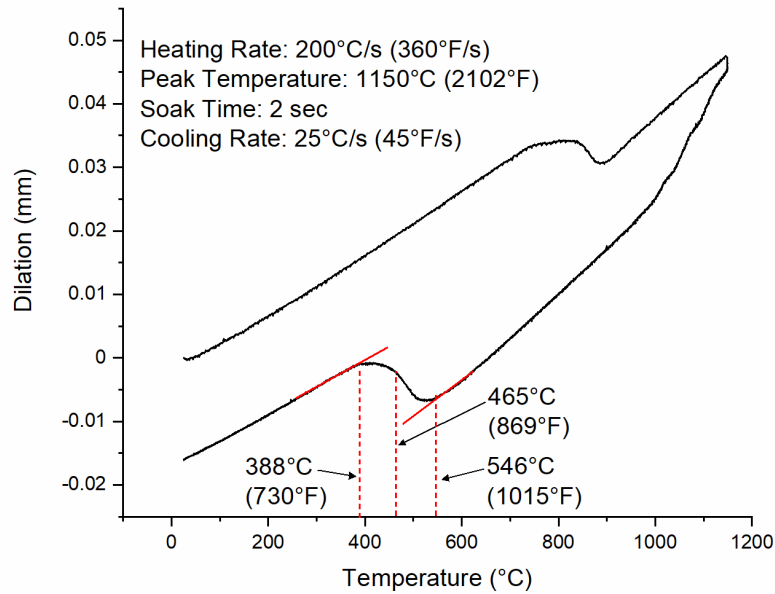


Figure 55. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 25 °C/s.

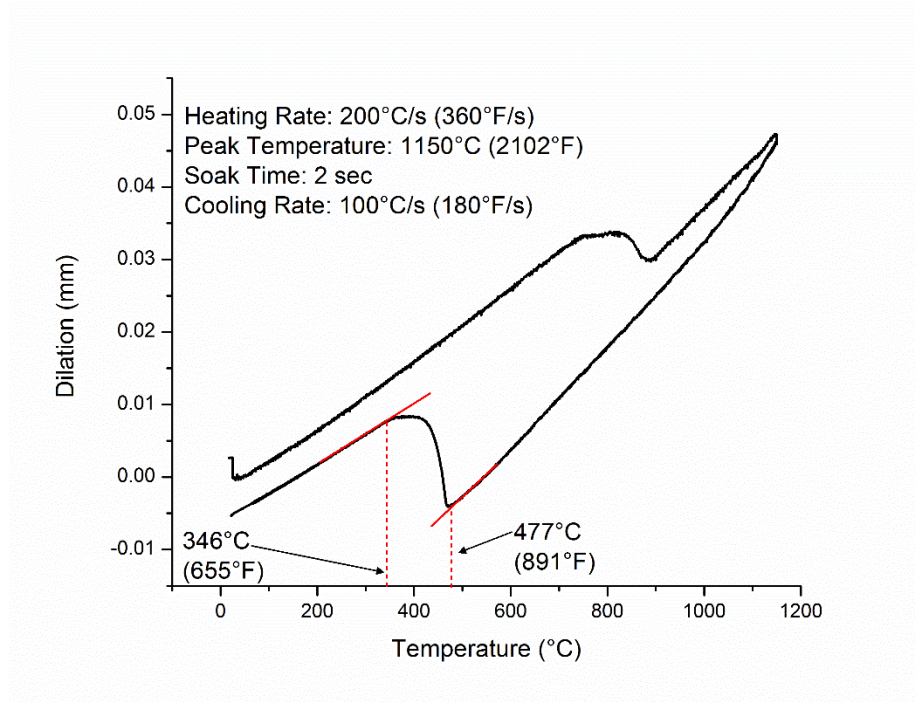


Figure 56. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 100 °C/s.

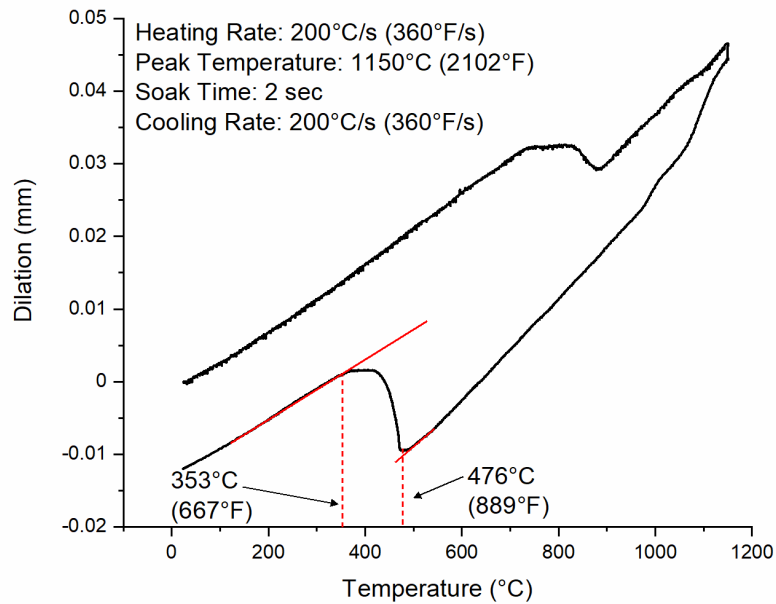


Figure 57. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1150 °C and cooled at 200 °C/s.

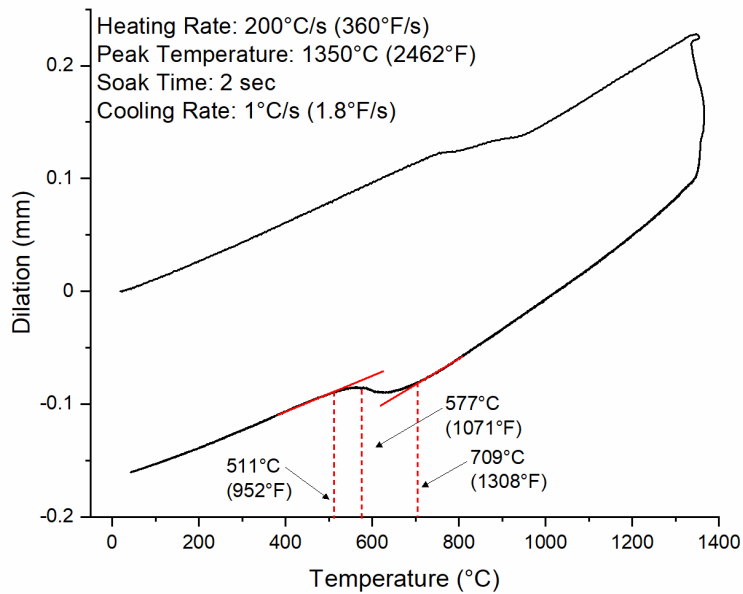


Figure 58. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 1 °C/s.

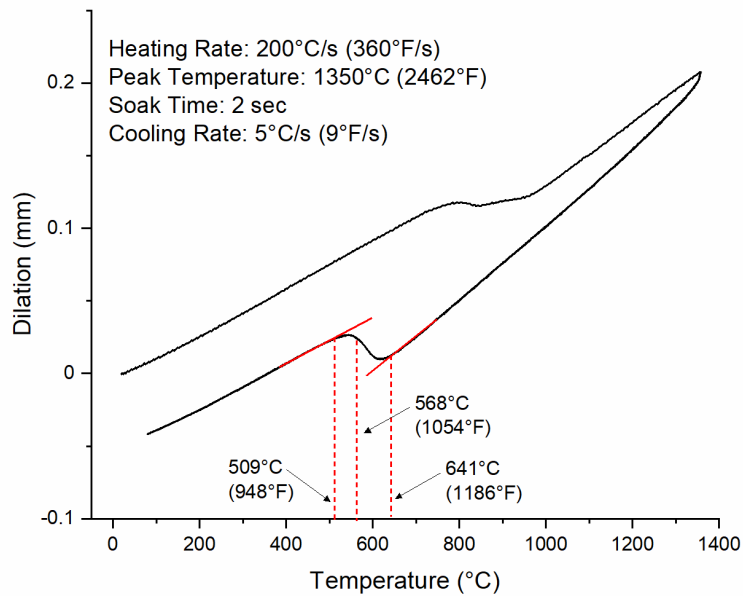


Figure 59. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 5 °C/s.

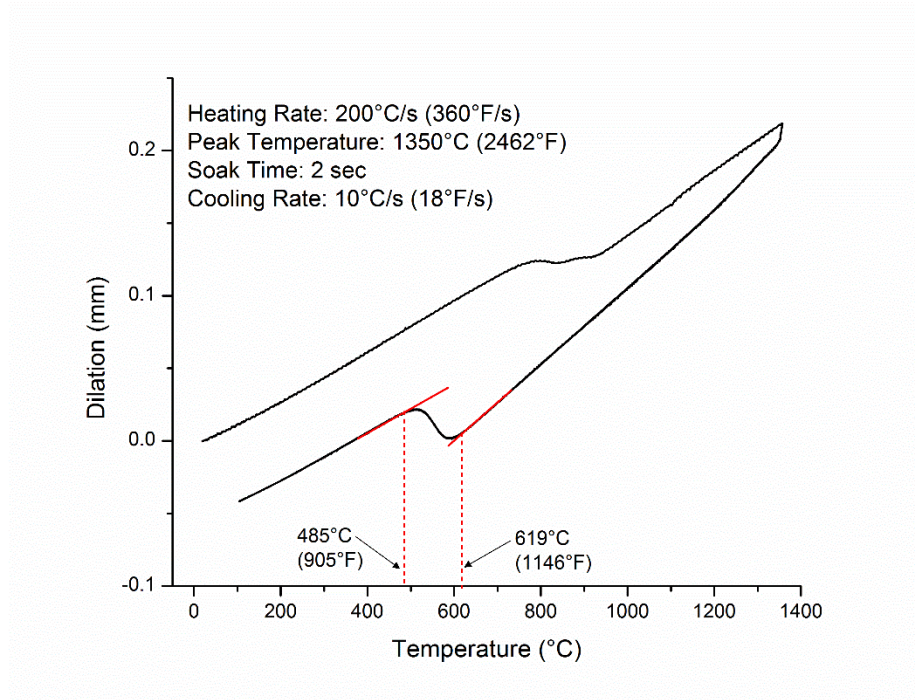


Figure 60. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 10 °C/s.

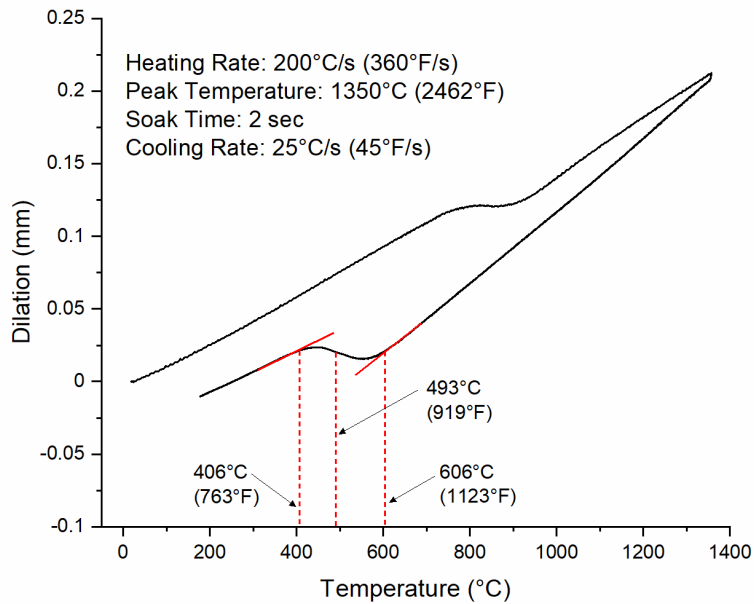


Figure 61. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 25 °C/s.

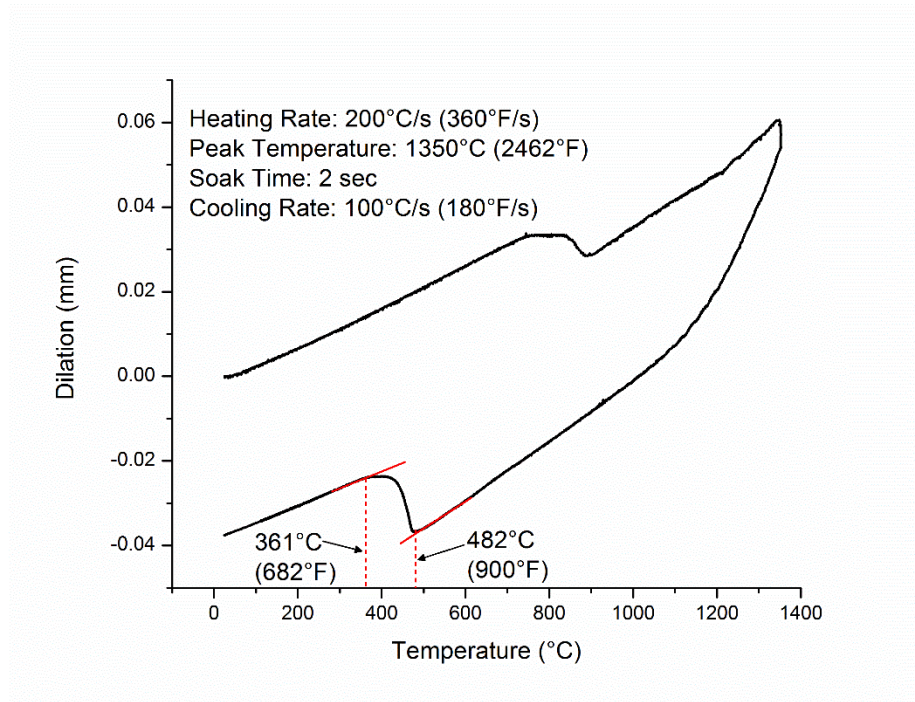


Figure 62. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 100 °C/s.

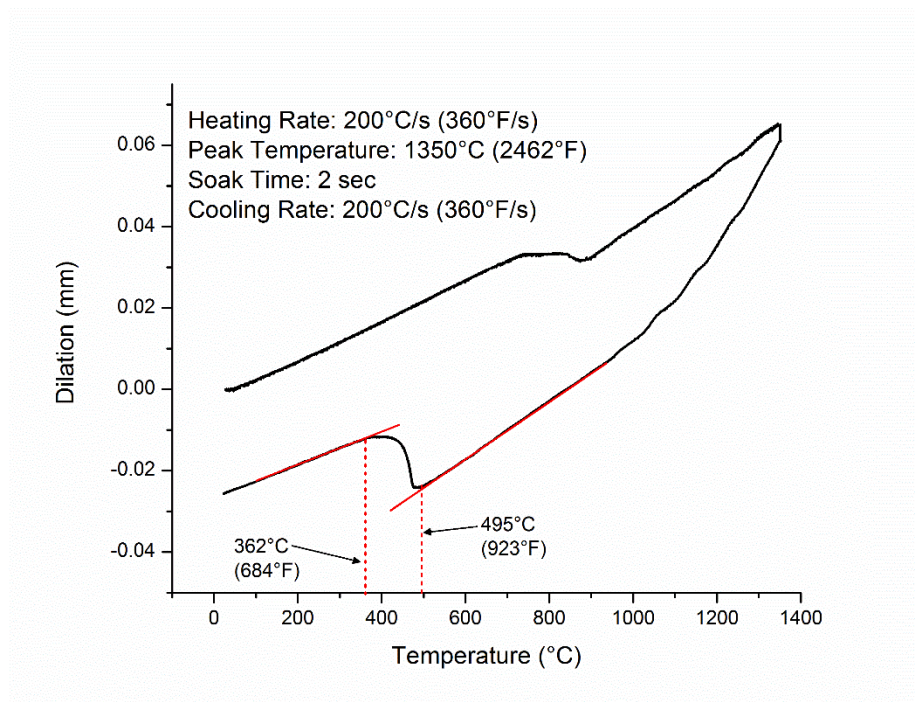


Figure 63. Dilation curve from a HSLA-65 Gleeble sample heated to a peak temperature of 1350 °C and cooled at 200 °C/s.

Appendix C: Representative Microstructures of HSLA-65 CCT Specimens

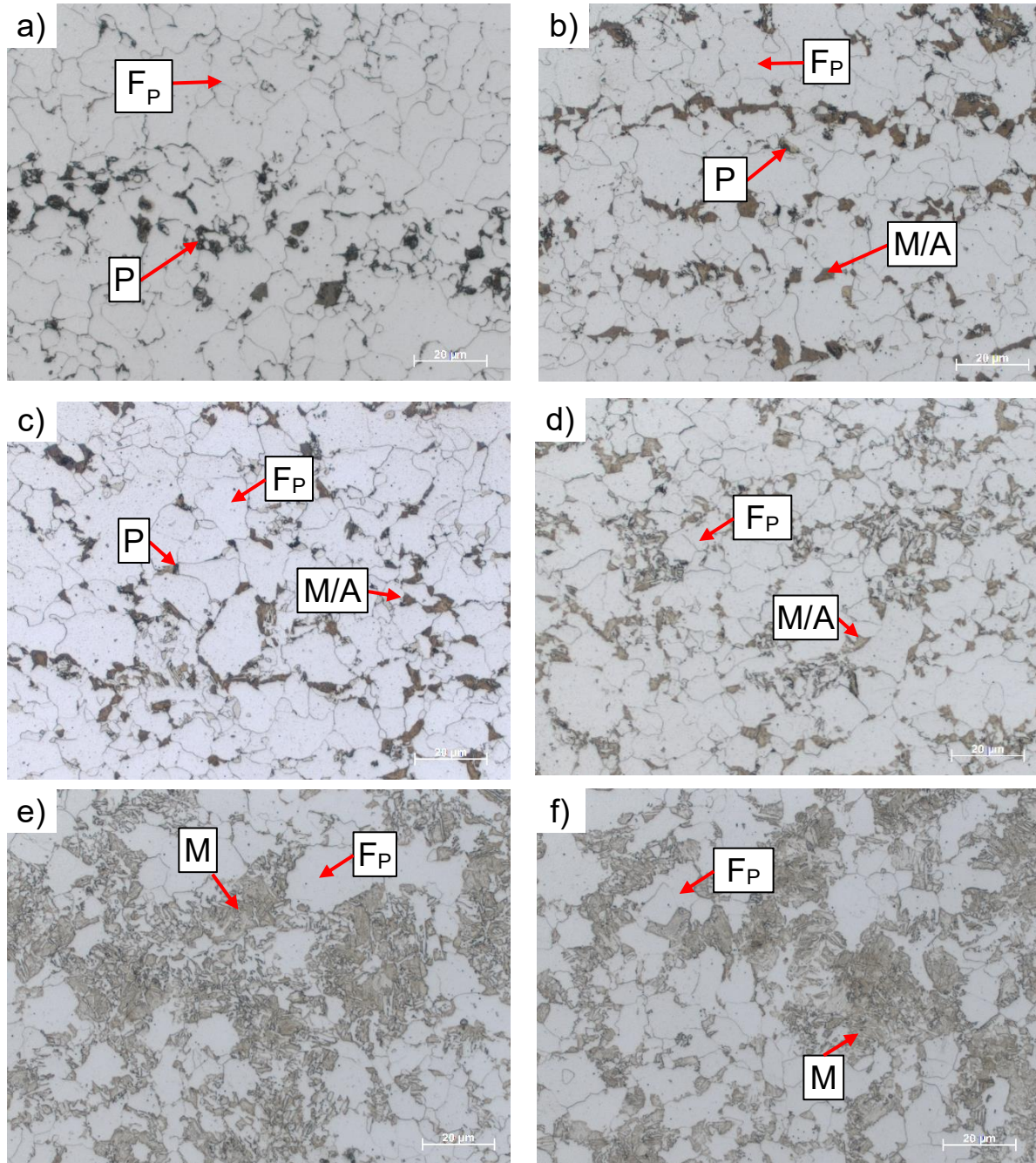


Figure 64. Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 825 °C (1517 °F) and continuously cooled at various rates. **(a-f):** 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).



Figure 65. Representative microstructure from the HSLA-65 dilatometry specimen heated to a peak temperature of 825 °C (1517 °F) and continuously cooled at 1 °C/s (1.8 °F/s). Banding of pearlite (dark etched constituent) is apparent at low magnifications.

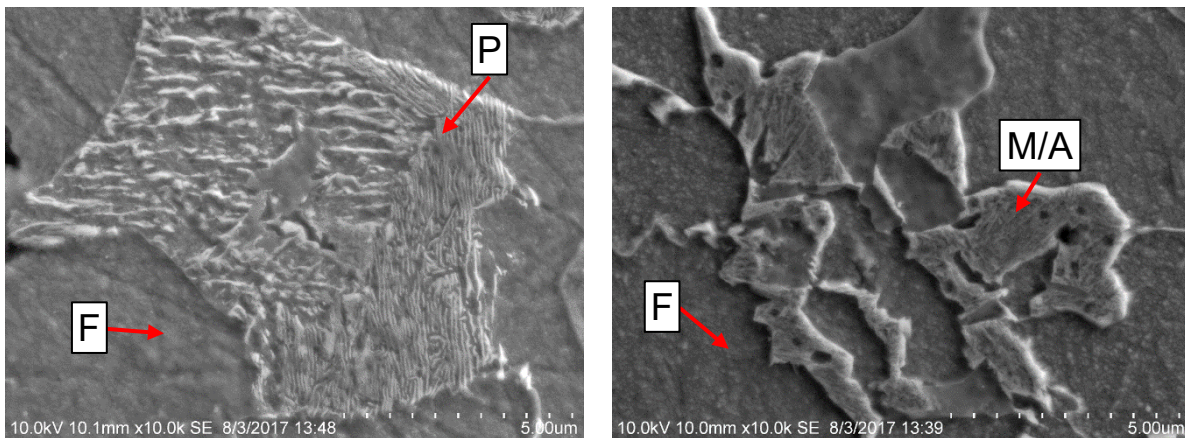


Figure 66. SEM images of the microstructure of a HSLA-65 dilatometry specimen heated to a peak temperature of 825 °C (1517 °F) and cooled at 25 °C/s (45 °F/s). Image on the left shows a typical pearlite colony found within a ferrite matrix in the base material, while the image on the right shows M/A constituent in the thermally cycled material.

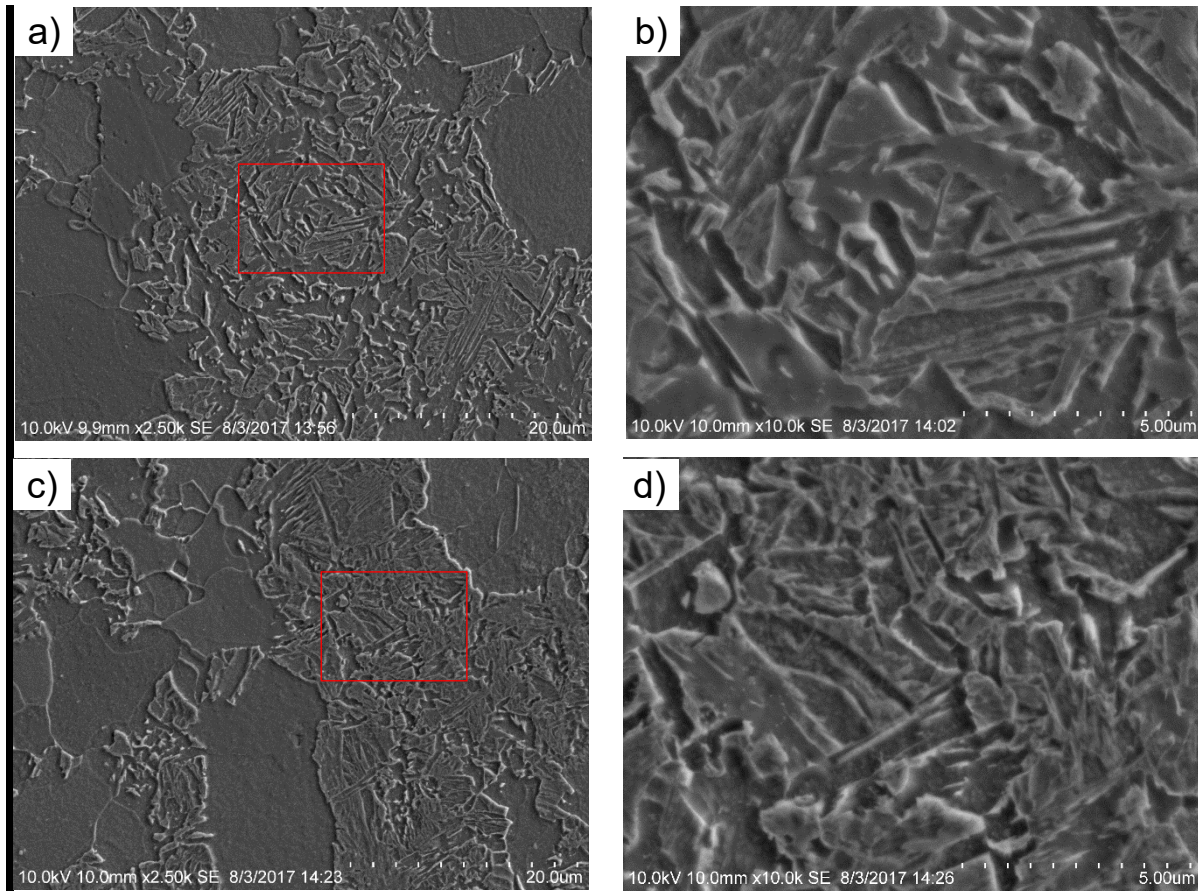


Figure 67. SEM images of the microstructure of a HSLA-65 dilatometry specimen heated to a peak temperature of 825 °C (1517 °F) and cooled at 100 (top row) and 200 °C/s (bottom row) (180 and 360 °F/s). Red boxes outline regions shown in images on the right.

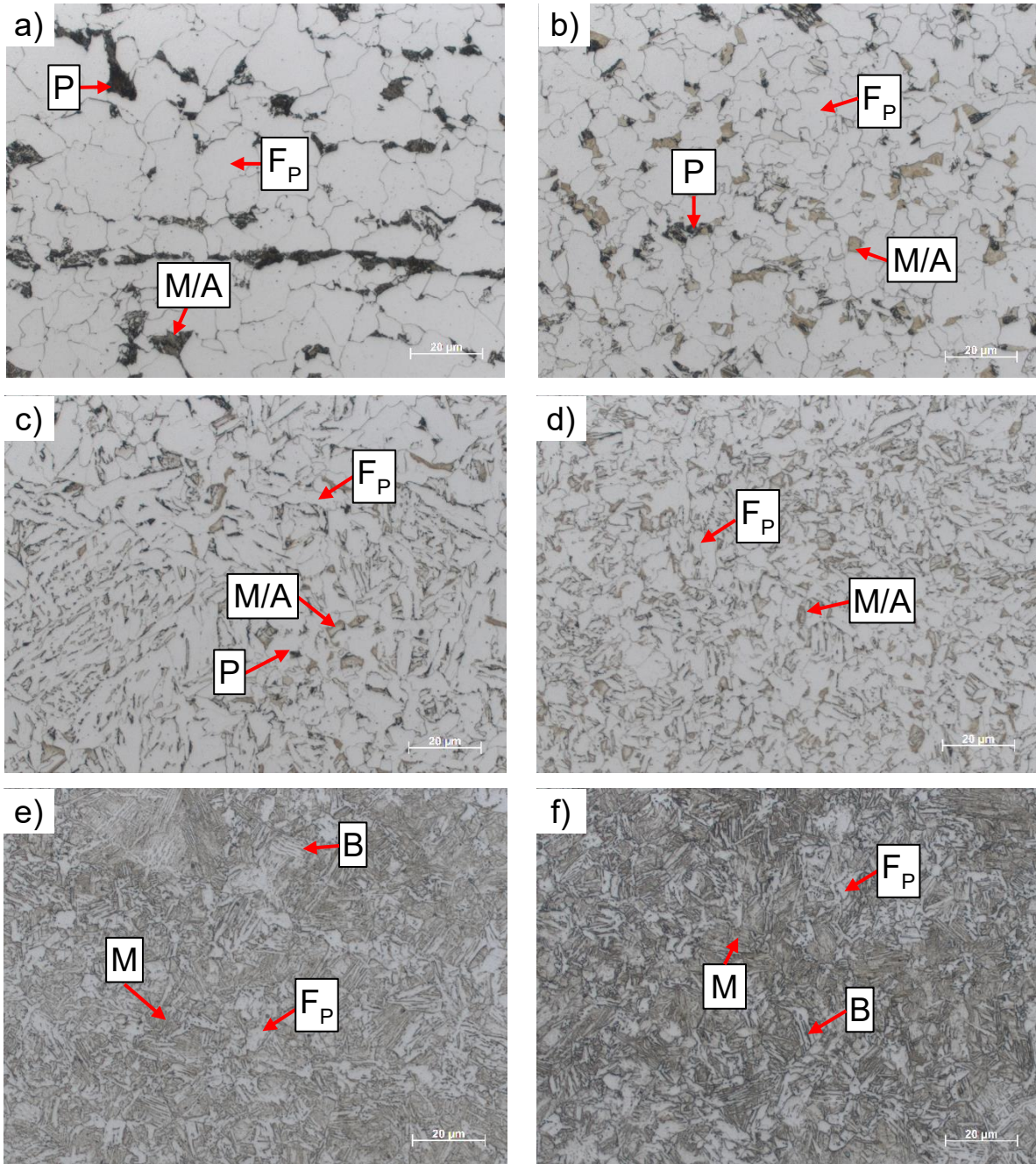


Figure 68. Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 1000 °C (1832 °F) and continuously cooled at various rates. (a-f): 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).

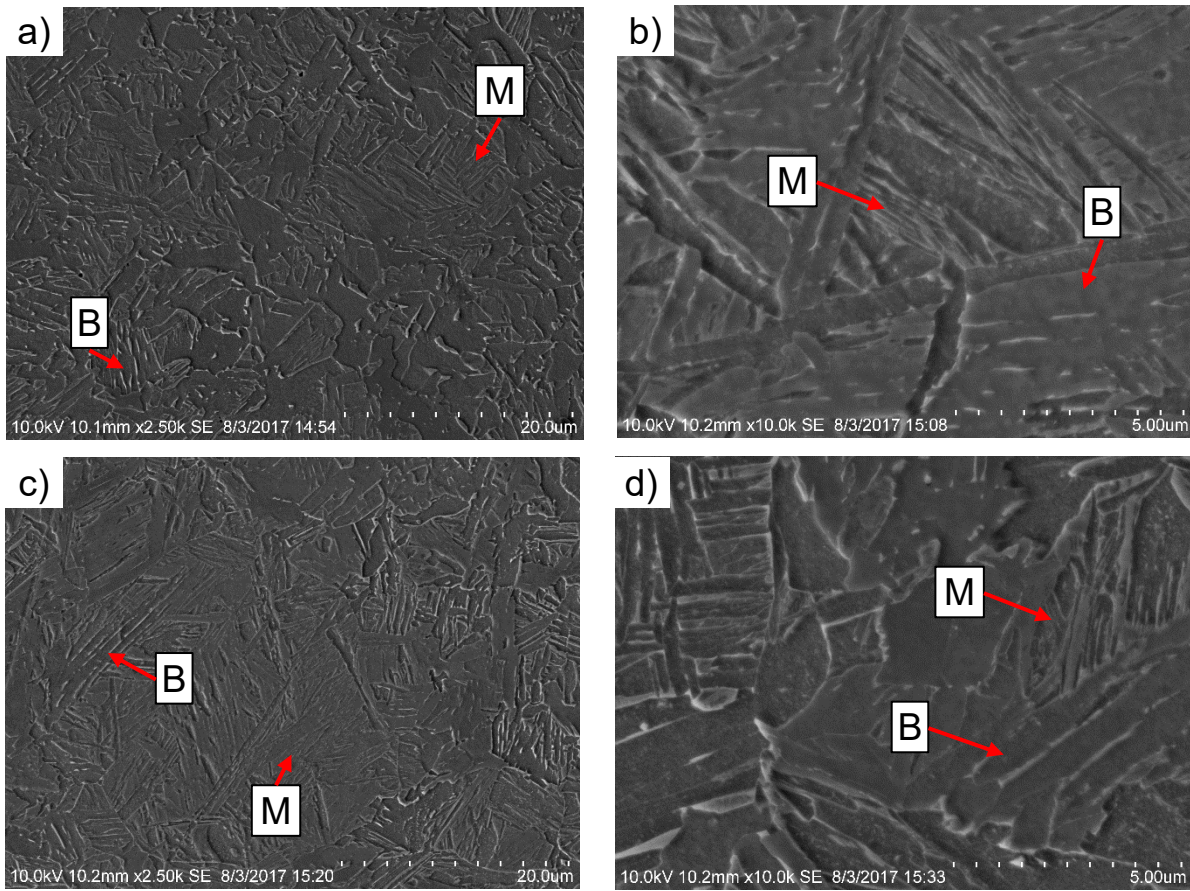


Figure 69. SEM images of the microstructure of a HSLA-65 dilatometry specimen heated to a peak temperature of 1000 °C (1832 °F) and cooled at 100 and 200 °C/s (180 and 360 °F/s) showing islands of martensite interspersed with bainite.

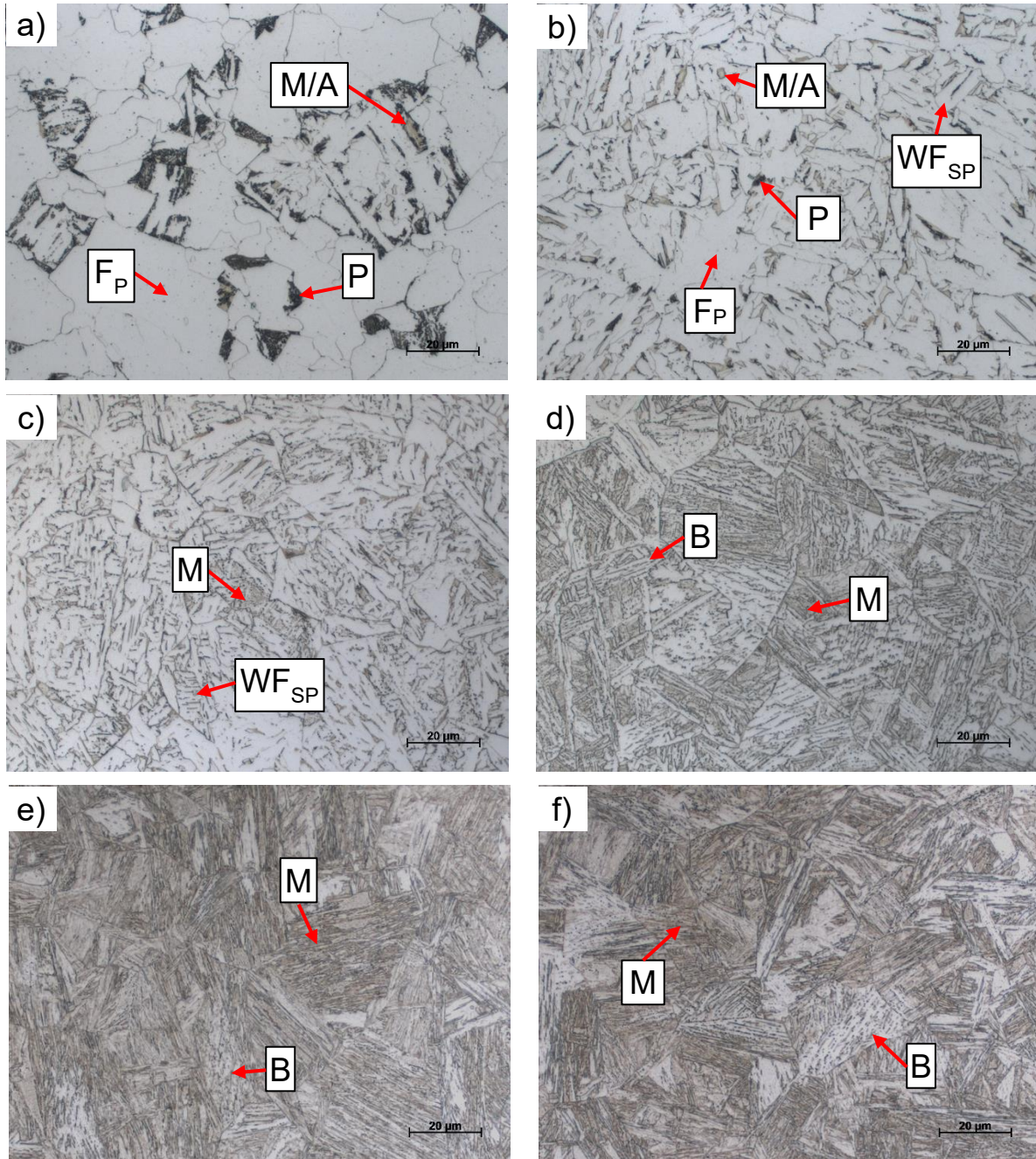


Figure 70. Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 1150 °C (2102 °F) and continuously cooled at various rates. (a-f): 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).

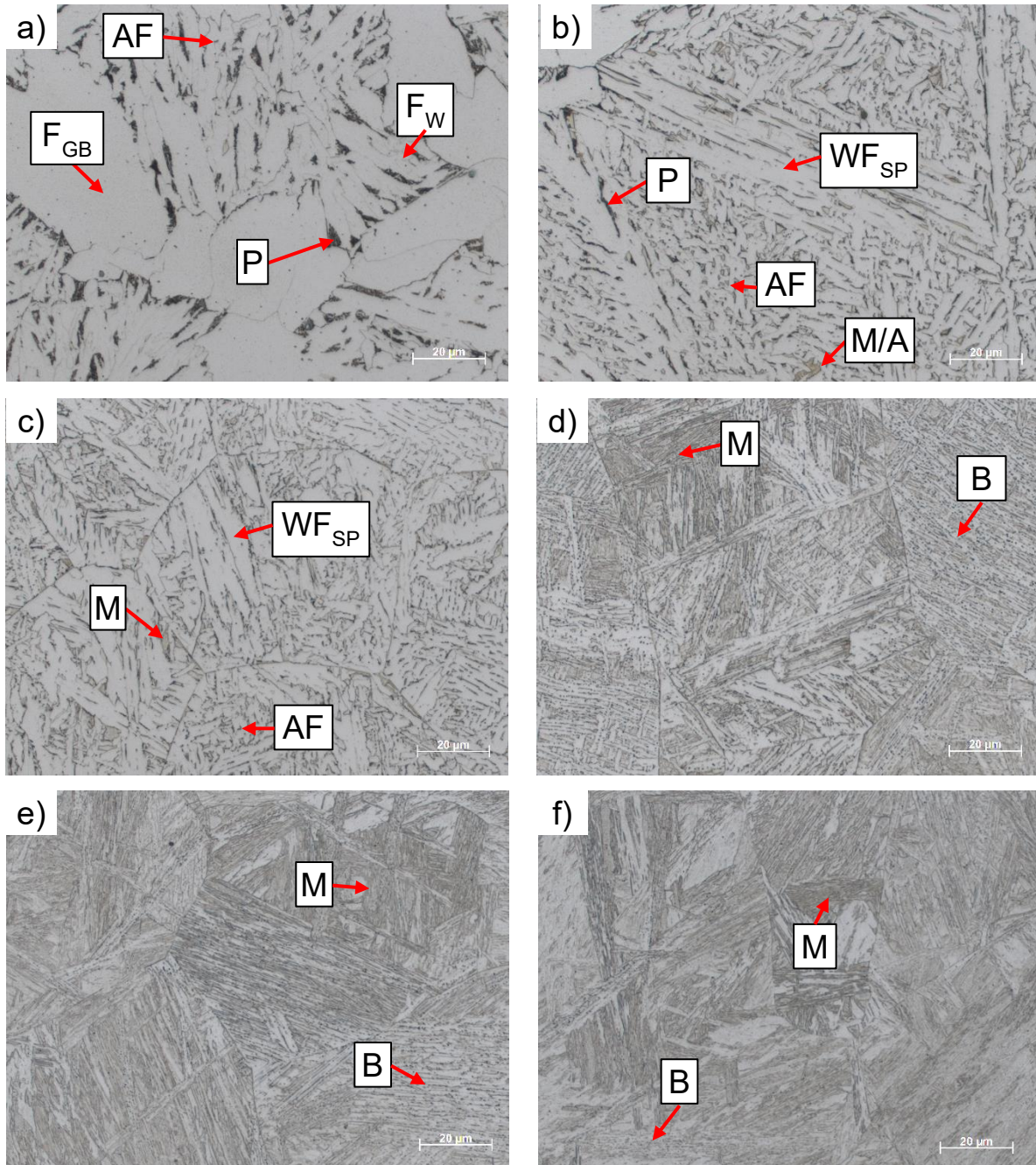


Figure 71. Representative microstructures from HSLA-65 dilatometry specimens heated to a peak temperature of 1350 °C (2462 °F) and continuously cooled at various rates. (a-f): 1, 5, 10, 25, 100, and 200 °C/s (1.8, 9, 18, 45, 180, and 360 °F/s).

Appendix D: Microhardness Measurements**Table 15.** Vickers microhardness of HSLA-65 as a function of peak temperature and cooling rate. Errors are one standard deviation. Values in parenthesis are the number of indents measured.

Cooling Rate		Peak Temperature				
		°C	825	1000	1150	1350
°C/s	°F/s	°F	1517	1832	2102	2432
1	1.8	179 ± 11 (10)		194 ± 7 (90)	194 ± 14 (10)	219 ± 5 (63)
5	9	199 ± 13 (10)		214 ± 10 (90)	226 ± 5 (60)	235 ± 6 (60)
10	18	203 ± 10 (10)		221 ± 5 (63)	248 ± 7 (63)	248 ± 10 (80)
25	45	208 ± 10 (10)		231 ± 5 (60)	300 ± 16 (60)	301 ± 20 (80)
100	180	225 ± 21 (10)		275 ± 29 (10)	354 ± 10 (10)	343 ± 14 (10)
200	360	238 ± 28 (10)		295 ± 21 (10)	339 ± 20 (10)	337 ± 12 (10)

Table 16. Vickers microhardness measurements across the experimental tee joint weldment illustrated in **Figure 20**. For each pattern the column on the left corresponds to the row closest to the plate surface.

Pass 1			Pass 2		
	Distance from Fusion Line (μm)	Hardness (HV_{200})	Distance from Fusion Line (μm)	Hardness (HV_{200})	Hardness (HV_{200})
Weld Metal	-600	212	-600	229	227
	-500	190	-500	238	232
	-400	192	-400	228	222
	-300	218	-300	199	198
	-200	211	-200	201	194
	-100	186	-100	207	212
	0	228	0	191	187
CGHAZ	100	209	100	215	209
	200	221	200	200	204
	300	227	300	212	210
	401	231	400	211	207
	500	228	500	200	214
FGHAZ	600	227	600	206	204
	700	238	700	203	201
	800	233	800	216	205
	900	227	900	224	223
	1000	222	1000	208	207
ICHAZ	1100	226	1100	213	200
	1200	222	1200	201	193
	1300	225	1300	208	194
	1400	210	1400	211	200
	1500	202	1500	206	198
Base Metal	1600	201	1600	202	206
	1700	204	1700	198	207
	1800	201	1800	208	200
	1900	217	1900	222	215
	2000	212	2000	198	208
	2100	211	2100	209	212
	2200	218	2200	205	206
	2300	206	2300	206	
	2400	221			

Table 17. Vickers microhardness measurements across the experimental butt joint weldment illustrated in **Figure 21**. Distance refers to distance from the fusion line, parallel to the indent pattern. For each pattern the column on the left corresponds to the row closest to the plate surface. Extensive reheating of Traverse 2 obscured delineation between traditional heat-affected zone sections.

Traverse 1 (Pass 1)						
	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)
Weld Metal	-300	199	-600	206	-860	204
	0	203	-300	206	-560	193
CGHAZ	300	195	0	225	-260	201
	600	200	300	198	40	225
	900	207	600	219	340	204
FGHAZ	1200	214	900	218	640	229
	1500	193	1200	217	940	222
	1800	197	1500	201	1240	201
	2100	191	1800	189	1540	197
ICHAZ	2400	187	2100	179	1840	183
	2700	175	2400	177	2139	178
	3000	176	2700	184	2440	176
	3300	188	3000	190	2740	200
Base Metal	3600	181	3300	186	3040	200
	3900	188	3600	191	3340	196

Traverse 2 (Pass 1)						
	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)
	152	206	0	198	-270	194
	452	201	300	200	30	191
	752	201	599	201	330	206
	1052	204	899	201	630	208
	1352	190	1199	199	929	207
	1652	196	1499	191	1230	198
	1952	181	1799	189	1530	179
	2252	186	2100	183	1830	174
	2552	191	2399	176	2130	181
	2852	183	2699	191	2430	181
	3152	183	2999	190	2729	173
	3452	176	3299	187	3030	193
	3752	170	3599	190	3330	196
	4052	183	3899	197	3630	205
	4352	187	4199	192	3930	198

Traverse 3 (Pass 2)

	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)
Weld Metal	-483	221	-600	218	-765	218
	-183	209	-300	214	-465	204
CGHAZ	117	216	0	228	-165	206
	417	206	299	209	135	208
	717	206	600	203	435	201
FGHAZ	1016	200	900	205	735	205
	1317	207	1200	202	1034	188
	1617	194	1500	188	1335	190
ICHAZ	1917	189	1799	183	1635	195
	2217	203	2100	202	1935	198
	2516	210	2400	195	2235	203
	2817	197	2700	206	2534	207
	3117	204	2999	194	2835	200
	3417	211	3299	200	3135	211
	3716	207	3600	208	3435	207
Base Metal						

Traverse 4 (Pass 2)

	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)	Distance (μm)	Hardness (HV ₂₀₀)
Weld Metal	-299	228	-600	220	-900	207
	0	233	-300	222	-600	228
	300	202	0	222	-300	224
CGHAZ	600	212	300	209	0	211
	900	208	600	216	300	213
	1200	203	900	204	600	222
FGHAZ	1500	189	1200	196	900	193
	1800	180	1500	184	1200	191
	2100	181	1800	178	1500	183
ICHAZ	2400	188	2100	179	1800	186
	2700	192	2400	189	2100	190
	3000	183	2700	183	2400	175
	3300	180	3000	176	2700	181
Base Metal	3600	183	3300	178	3000	182
	3900	186	3600	193	3300	194

Appendix E: Thermo-Physical Property Datasets for HSLA-65**Table 18.** Experimentally measured specific heat of HSLA-65.

Temperature (°C)	c_p (J/g-°C)	(°C)	c_p (J/g-°C)	(°C)	c_p (J/g-°C)	(°C)	c_p (J/g-°C)
23	0.4574	320	0.5731	620	0.7510	920	0.6061
30	0.4602	330	0.5771	630	0.7676	930	0.6023
40	0.4643	340	0.5811	640	0.7826	940	0.6154
50	0.4685	350	0.5853	650	0.7997	950	0.6202
60	0.4725	360	0.5896	660	0.8153	960	0.6195
70	0.4772	370	0.5938	670	0.8367	970	0.6126
80	0.4811	380	0.5984	680	0.8486	980	0.6202
90	0.4852	390	0.6030	690	0.8699	990	0.6201
100	0.4891	400	0.6078	700	0.8989	1000	0.6268
110	0.4929	410	0.6127	710	0.9200	1010	0.6288
120	0.4969	420	0.6180	720	0.9443	1020	0.6305
130	0.5008	430	0.6234	730	0.9823	1030	0.6321
140	0.5048	440	0.6289	740	1.0524	1040	0.6337
150	0.5086	450	0.6345	750	1.2567	1050	0.6354
160	0.5124	460	0.6401	760	1.2659	1060	0.6370
170	0.5163	470	0.6462	770	1.0920	1070	0.6387
180	0.5199	480	0.6524	780	0.9750	1080	0.6403
190	0.5234	490	0.6573	790	0.9068	1090	0.6420
200	0.5272	500	0.6640	800	0.8741	1100	0.6434
210	0.5310	510	0.6694	810	0.8504	1110	0.6447
220	0.5352	520	0.6793	820	0.8370	1120	0.6459
230	0.5393	530	0.6841	830	0.8444	1130	0.6472
240	0.5436	540	0.6879	840	0.8434	1140	0.6485
250	0.5470	550	0.6947	850	0.8490	1150	0.6497
260	0.5506	560	0.7030	860	0.8334	1160	0.6510
270	0.5542	570	0.7054	870	0.7757	1170	0.6523
280	0.5579	580	0.7145	880	0.7227	1180	0.6535
290	0.5615	590	0.7206	890	0.6865	1190	0.6548
300	0.5653	600	0.7312	900	0.6450	1200	0.6557
310	0.5692	610	0.7413	910	0.6110		

Table 19. Experimentally measured thermal properties of HSLA-65.

Temperature (°C)	Thermal Diffusivity (cm²/sec)	Density (g/cm³)	Thermal Conductivity (W/cm-°C)
23	0.11403	7.79	0.406
50	0.11258	7.78	0.416
100	0.10964	7.77	0.426
200	0.10445	7.73	0.418
300	0.09601	7.70	0.403
400	0.08642	7.66	0.391
500	0.07727	7.63	0.374
600	0.06492	7.59	0.355
700	0.05231	7.56	0.221
800	0.0336	7.53	0.269
900	0.05547	7.53	0.406
1000	0.06007	7.49	0.282
1100	0.06441	7.46	0.309
1200	0.06727	7.41	0.327

Appendix F: Thermo-Mechanical Properties

Table 20. Assumed elevated temperature elastic modulus for HSLA-65 base material, as shown in Figure 29. Data are based on an assumed room temperature modulus of 210 GPa (30.5 Msi).

Temperature (°C [°F])	Fraction of Room Temperature Modulus	Elastic Modulus (GPa [Msi])
23 (73)	1.0000	210 (30.5)
100 (212)	1.0000	210 (30.5)
200 (392)	0.9000	189 (27.4)
300 (572)	0.8000	168 (24.4)
400 (752)	0.7000	147 (21.3)
500 (932)	0.6000	126 (18.3)
600 (1112)	0.3100	65 (9.4)
700 (1292)	0.1300	27 (4.0)
800 (1472)	0.0900	19 (2.7)
900 (1652)	0.0675	14 (2.1)
1000 (1832)	0.0450	9 (1.4)
1100 (2012)	0.0225	5 (0.7)
1200 (2192)	0.0000	0 (0)

Table 21. Elevated temperature mechanical properties of HSLA-65 base material, as shown in Figures 30-31.

Specimen ID	Test Temperature (°C [°F])	Yield Strength (MPa [ksi])	Tensile Strength (MPa [ksi])	Elongation in 2 in. (%)
4230-2FTT	23 (74)	470 (68.2)	680 (98.6)	24
4230-2ETT1	200 (392)	400 (58.0)	570 (82.7)	23
4230-2ETT2A	400 (752)	320 (46.4)	530 (76.9)	27
4230-2ETT3A	600 (1112)	180 (26.1)	225 (32.6)	37.5
4230-2ETT4	700 (1292)	85 (12.3)	112 (16.2)	43
4230-2ETT5	800 (1472)	38 (5.5)	45 (6.5)	83.5
4230-2ETT6	900 (1652)	34 (4.9)	37 (5.4)	56
4230-2ETT7	1000 (1832)	20.1 (2.9)	27 (3.9)	41
4230-2ETT8B	1100 (2012)	8.8 (1.3)	12 (1.8)	20

Table 22. Yield strength of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at different rates, as illustrated in **Figure 32**.

Temperature (°C) [°F]	Yield Strength (MPa)			
	1 °C/s Cooling	10 °C/s Cooling	25 °C/s Cooling	100 °C/s Cooling
25 [77]	470	610	561	724
200 [392]	454	589	656	712
400 [752]	381	502	561	637
600 [1112]	268	339	283	242
700 [1292]	176	194	175	163

Table 23. On-heating flow behavior of HSLA-65 base material, as shown in **Figures 33-34**.

22 °C (72 °F)		200 °C (392 °F)		400 °C (752 °F)		600 °C (1112 °F)		700 °C (1292 °F)	
$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)
0.0000	485	0.0000	407	0.0000	300	0.0000	162	0.0000	81
0.0020	485	0.0020	440	0.0020	366	0.0020	210	0.0020	105
0.0050	486	0.0050	470	0.0050	401	0.0050	222	0.0050	110
0.0100	490	0.0100	490	0.0100	429	0.0072	225	0.0100	111
0.0200	524	0.0192	525	0.0112	435			0.0200	113
0.0365	570							0.0263	113

800 °C (1472 °F)		900 °C (1652 °F)		1000 °C (1832 °F)		1100 °C (2012 °F)	
$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)
0.0000	20	0.0000	20	0.0000	21	0.0000	9.3
0.0020	38	0.0020	39.2	0.0020	22.5	0.0020	9.4
0.0050	44.6	0.0028	40.4	0.0050	22	0.0050	10.3
0.0100	46.5			0.0100	22.5	0.0100	10.4
0.0200	47.5			0.0200	24.4	0.0200	11.5
				0.0332	25.7	0.0269	11.9

Table 24. Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 1 °C/s (1.8 °F/s), as illustrated in **Figure 35**. Terminal values are at the UTS.

25 °C (77 °F)		200 °C (392 °F)		400 °C (752 °F)		600 °C (1112 °F)		700 °C (1292 °F)	
$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)
0.000	470	0.000	454	0.000	381	0.000	268	0.000	176
0.005	509	0.005	517	0.005	445	0.005	293	0.004	187
0.010	528	0.010	542	0.010	472	0.010	297		
0.020	562	0.020	580	0.020	504	0.020	299		
0.030	590	0.030	610	0.030	525	0.029	300		
0.040	610	0.040	627	0.040	540				
0.050	626	0.050	638	0.050	552				
0.100	672	0.100	676	0.100	584				
0.156	691	0.182	696	0.108	589				

Table 25. Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 10 °C/s (18 °F/s), as illustrated in **Figure 36**. Terminal values are at the UTS.

25 °C (77 °F)		200 °C (392 °F)		400 °C (752 °F)		600 °C (1112 °F)		700 °C (1292 °F)	
$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)
0.000	610	0.000	589	0.000	502	0.000	339	0.000	194
0.005	672	0.005	682	0.005	591	0.005	372	0.005	225
0.010	694	0.010	715	0.010	613	0.007	373	0.006	226
0.020	724	0.020	753	0.020	638				
0.030	744	0.030	779	0.030	651				
0.040	758	0.040	795	0.040	657				
0.050	769	0.050	809	0.050	661				
0.100	787	0.100	826	0.100	670				
0.105	788	0.128	827	0.104	672				

Table 26. Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 25 °C/s (45 °F/s), as illustrated in **Figure 37**. Terminal values are at the UTS.

25 °C (77 °F)		200 °C (392 °F)		400 °C (752 °F)		600 °C (1112 °F)		700 °C (1292 °F)	
$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)
0.000	561	0.000	656	0.000	561	0.000	283	0.000	175
0.005	728	0.005	799	0.005	648	0.005	356	0.005	209
0.010	777	0.010	852	0.010	675	0.010	366	0.0055	210
0.020	828	0.020	891	0.020	693	0.010	367		
0.030	858	0.030	914	0.030	699				
0.040	877	0.040	933	0.040	702				
0.050	886	0.050	943	0.050	703				
0.096	905	0.100	979	0.064	703				
		0.132	986						

Table 27. Flow stress of simulated HSLA-65 CGHAZs after heating to 1350 °C (2462 °F) and cooling at 100 °C/s (180 °F/s), as illustrated in **Figure 38**. Terminal values are at the UTS.

25 °C (77 °F)		200 °C (392 °F)		400 °C (752 °F)		600 °C (1112 °F)		700 °C (1292 °F)	
$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)	$\epsilon_{p, \text{true}}$	σ (MPa)
0.000	724	0.000	712	0.000	637	0.000	242	0.000	163
0.005	833	0.005	868	0.005	732	0.005	328	0.004	185
0.010	873	0.010	927	0.010	757	0.010	340		
0.020	912	0.020	978	0.020	766	0.011	341		
0.030	934	0.030	1000	0.023	767				
0.040	948	0.040	1014						
0.050	958	0.050	1025						
0.088	971	0.100	1051						
		0.105	1052						

Table 28. Engineering fracture strain for the specimens shown in **Figures 35-37**.

Temperature (°C) [°F]	Fracture Strain, $\epsilon_{f, \text{eng}}$ (mm/mm)			
	1 °C/s Cooling	10 °C/s Cooling	25 °C/s Cooling	100 °C/s Cooling
25 [77]	0.415	0.393	0.362	0.355
200 [392]	0.416	0.270	0.245	0.181
400 [752]	0.361	0.682	0.208	0.091
600 [1112]	0.281	0.066	0.066	0.052
700 [1292]	0.235	0.056	0.054	0.061

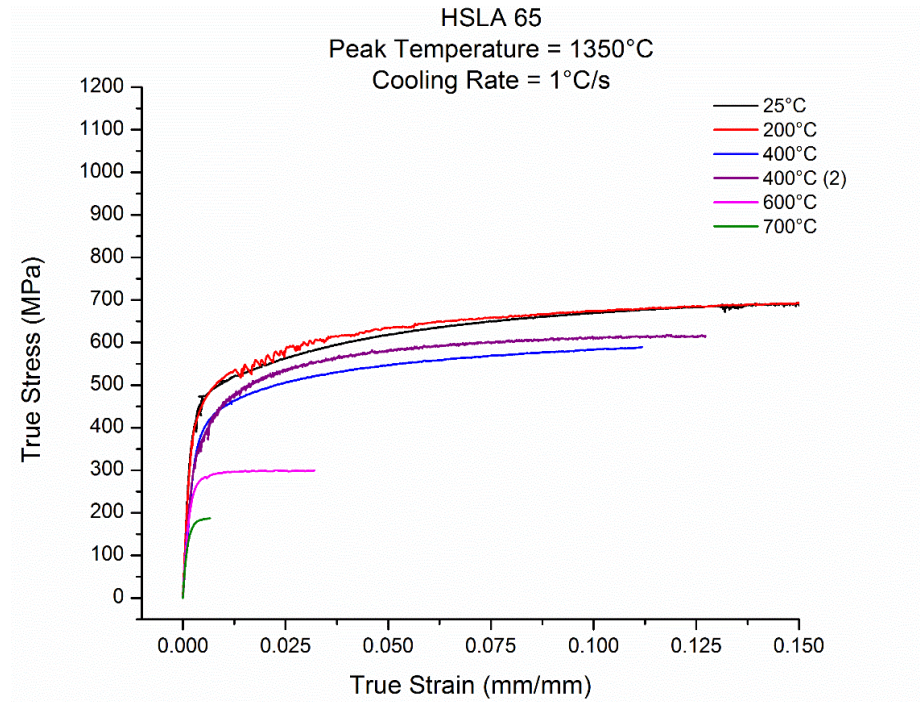


Figure 72. Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 1 °C/s (1.8 °F/s), then reheated to the test temperature.

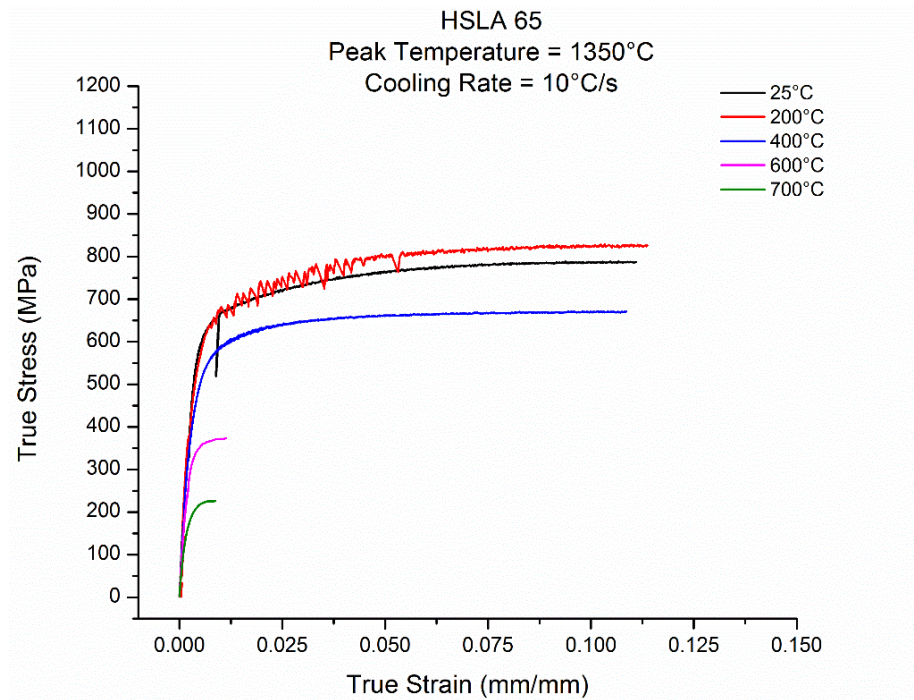


Figure 73. Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 10 °C/s (18 °F/s), then reheated to the test temperature.

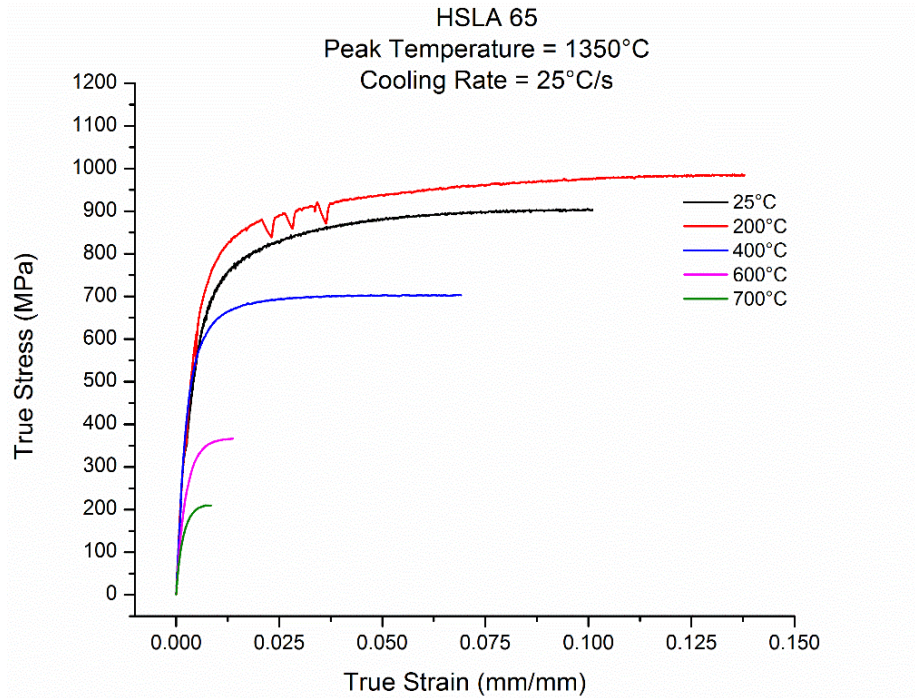


Figure 74. Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 100 °C/s (180 °F/s), then reheated to the test temperature.

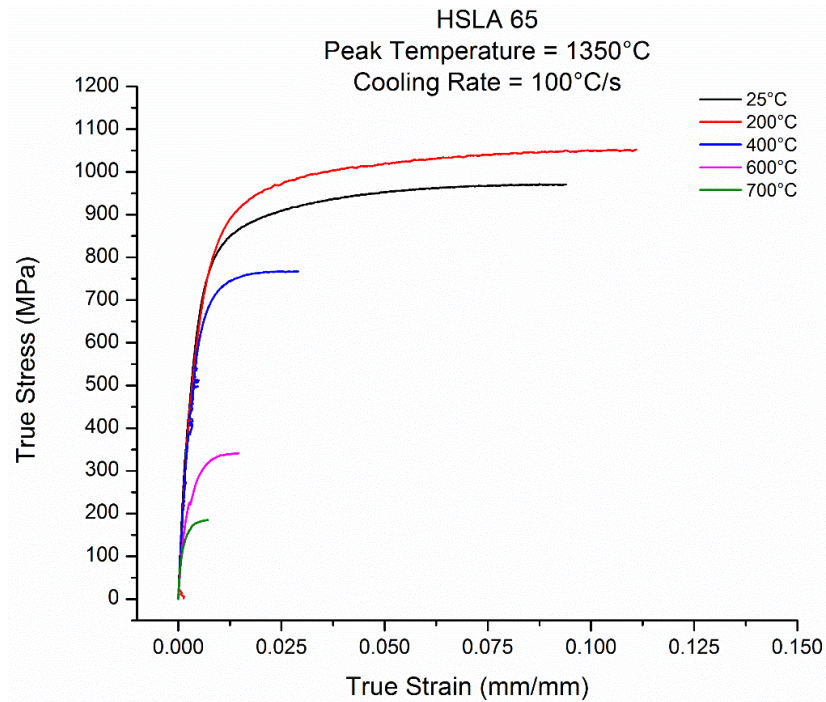


Figure 75. Uniaxial tension stress-strain curves from Gleeble tensile samples thermally cycled to a peak temperature of 1350 °C (2462 °F), cooled at 100 °C/s (180 °F/s), then reheated to the test temperature.

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