

**Modeling Long-term Creep Performance for
Welded Nickel-base Superalloy Structures for
Power Generation Systems**

Final Report

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ABSTRACT

The goal of this project is to model long-term creep performance for nickel-base superalloy weldments in high temperature power generation systems. The project uses physics-based modeling methodologies and algorithms for predicting alloy properties in heterogeneous material structures. The modeling methodology will be demonstrated on a gas turbine combustor liner weldment of Haynes 282 precipitate-strengthened nickel-base superalloy. The major developments are:

- microstructure-property relationships under creep conditions and microstructure characterization
- modeling inhomogeneous microstructure in superalloy weld
- modeling mesoscale plastic deformation in superalloy weld
- a constitutive creep model that accounts for weld and base metal microstructure and their long term evolution

The developed modeling technology is aimed to provide a more efficient and accurate assessment of a material's long-term performance compared with current testing and extrapolation methods. This modeling technology will also accelerate development and qualification of new materials in advanced power generation systems.

This document is a final technical report for the project, covering efforts conducted from October 2014 to December 2016.

EXECUTIVE SUMMARY

This section provides a high level summary of some of the technical highlights and accomplishments achieved during this reporting period, in addition to a review of program schedule and financial status. A more detailed review can be found in the “Results and Discussions” section of the report.

Experimental testing and characterization provided crucial inputs for calibration and validation of microstructure and property models in Task 3, 4, 5. The welded and 100% weldment HA282 materials, after a standard post-weld heat treatment PWHT2 (2075°F/0.5hr + 1850°F/2hr + 1450°F/8hr), did not show significant debit in tensile strength at 1500°F and 1600°F. Notably, tensile strength for welded, and 100% weldment materials dropped at 1700°F compared to base metal. Ductility (% elongation) increased with the test temperature in all four types of materials, and decreased from base metal to 100% weldment to welded HA282 materials, with as-welded being the least ductile among all test temperatures. Creep tests showed no significant difference in rupture times among base-metal, welded, and 100% weldment HA282 materials, but a reduced ductility (% elongation) in the welded HA282. This work also provided first application of both pair-wise PD (volumetric) and ex-situ DIC (full-field surface) methods for measurements of creep strain across heterogeneous gauge section in welded material. The local strain measurements provided data for comparison with model predictions.

Optical and SEM imaging confirmed that all welded specimens failed along high-angle grain boundaries in the weld zone. Characterization of grain boundary phases and damage in ruptured and interrupted creep specimens showed common features regardless of testing temperature and stress. Weld zone consistently showed more grain boundary cavities with larger sizes, compared to base metal. Substantial γ' denuded zone were found near most cavities in weld zone, but rarely seen in base metal. Some cavities clearly grew into the γ' denuded zone, suggesting that the latter acted as a weak region against cavity growth. The reduced ductility in welded HA282 and its intergranular failure were attributed to an enhanced cavity growth in the γ' denuded zone in weld zone at a late stage of creep. The γ' denuded zone was absent in the static exposed specimens. Its formation was associated with localized strain that significantly sheared γ' particles, but not due to microsegregation. Under PWHT2, γ' size and fraction were found similar in base metal and weld zone, which explained a comparable creep life in base metal, welded, weldment HA282. Deformation mechanisms were characterized with TEM on interrupted creep specimens. Besides a relatively higher dislocation density was observed in weld zone, similar dislocation operation mechanisms were found in base metal and weld zone.

To bridge the processing-property relationship, three types of microstructure models were applied to welded HA282. For the as-weld microstructural state,

Scheil model and moving-boundary diffusion (DICTRA) model were applied to simulate solidification path for HA282. The predicted solidification phases and element microsegregation (except for Al) were found in a reasonable agreement with experimental observations in this study and in the literature. For post-weld heat treatments, a 2D homogenization model was applied to simulate the evolution of microsegregation through different PWHTs. The results were compared well with the EPMA quantitative map measurements from the actual heat treated samples. Finally, the residual microsegregations were used to simulate γ' microstructures. It was found that the primary effect of the composition inhomogeneity was a variation of local γ' volume fraction from dendrite core to interdendritic region in the weldment, but was negligible on the sizes of γ' even under the least effective homogenization condition PWHT3a (1850°F/1hr + 1450°F/8hr) of this study. Predicted long-term γ' coarsening between 1500-1700°F for up to 10,000 hours was found in an excellent agreement with experimental data. The simulated microstructures and spatial variation of γ' were provided as inputs to mechanical property models in Task 4 and Task 5. Some simulation results were also used to explain experimental findings in Task 2.

In Task 4, full-field image-based crystal plasticity model was developed and validated against the experimental testing data on HA282 base metal from Task 2. Key features of the model include: (i) dislocation-based, (ii) HA282 specific dislocation-precipitate interactions, and (iii) FFT-based micromechanical solver. The model has been applied to study the influence of structural heterogeneities in weld, such as microsegregation and grain boundary denuded zone found by experiments and simulations in Tasks 2 and 3, on the tensile and creep performance. The degree of structural inhomogeneity under PWHT3a was found to bring no significant changes of the mechanical properties of HA282 weld according to our simulations. However, if the processing conditions allowed even larger degree of structural heterogeneities in the weld, our simulation results suggested that tensile and creep strengths would be reduced significantly. To study the evolution of grain boundary creep cavities as observed in Task 2, we have developed a phase-field model that incorporates multiple diffusion mechanisms and inhomogeneous elasticity. The cavity dynamics for simple bicrystal grain boundary cavity configuration has been simulated using the phase-field model and the cavity kinetics and morphological change have been predicted. Simulation results from the simplified configuration have provided Task 5 some useful insight into the physical assumptions made by the analytical cavity growth model.

A continuum damage creep model was developed for HA282 base metal and its weld for a broad range of temperatures (1400F-1700°F) and stress levels (5-35Ksi). Two simulation methods were proposed for solving the model equations capturing the possible differences between the HA282 base metal and HA282 weld: (i) material point approach, which is very efficient from computational point

of view but limited to uniform stress and strain states, and (ii) finite element approach, which is more appropriate for non-uniform stress distributions. The model predictions were compared against both short and longer term (5000-7000 hours) experimental data for base metal, 100% weld and cross-welded creep specimens.

RESULTS AND DISCUSSIONS

This section of the report provides a review of each of the areas of the project that were active during the quarter. In this quarter the following technical tasks were active:

- Task 2 Development of Structure-Property Relationships
- Task 3 Modeling of the Microstructure of a Superalloy Weld
- Task 4 Mesoscale Modeling of Plastic Deformation in a Superalloy Weld
- Task 5 Development of a Constitutive Creep Model that accounts for Weld and Base Metal Microstructure and their Long Term Evolution

Task 2: Development of Structure-Property Relationships

Task 2 provides correlations between microstructural evolution of welded Haynes 282 (HA282) during thermal exposure and mechanical properties. Outcomes of this task have been used for calibration and validation of the models developed in Tasks 3, 4 and 5. Microstructural characterization of the samples taken from base metal, weld zone, and heat affected zone (HAZ) regions was performed to quantitatively evaluate the evolution of grain structure, residual stress, texture, microsegregation, porosity, γ' precipitate size and morphology during thermal exposure at 1500, 1600 and 1700°F. Tensile and creep tests were performed at 1500, 1600 and 1700°F using three types of materials: HA282 base metal, HA282 weldment (100% weld), and welded HA282 consisting of the base metal, HAZ and weldment. Table 1 shows the test matrix for evaluating the evolution of microstructure and tensile properties. Table 2 shows the creep test conditions. In-situ local creep strain measurements were conducted at selected creep test conditions using digital image correlation (DIC) and potential drop (PD) approaches. The in-situ measurement results were provided for comparison with the finite element creep simulations in Task 5.

Table 1: Test matrix for evaluating microstructure evolution and tensile properties. HA282 base metal, HA282 weldment, and welded HA282 were used.

		As-received	As-welded	Welded + PWHT	Weldment + PWHT	Static thermal exposure (no stress)											
						1500°F				1600°F				1700°F			
						500 h	1000 h	5000 h	10000 h	500 h	1000 h	5000 h	10000 h	500 h	1000 h	5000 h	10000 h
Microstructure		X	X	X	X	X	X	X	X	X	X	X	X	X	X	X	X
Tensile test	1500°F	X	X	X	X			X									
	1600°F	X	X	X	X							X					
	1700°F	X	X	X	X											X	

Table 2: Creep test matrix for HA282 weldment and welded HA282. Completed test conditions are marked as “X”. Additional specimens, marked with “I” below, were interrupted for microstructure characterization.

		Conventional Creep						PD/DIC Creep		
		Welded HA282			100% Weldment			Welded, 100% Weldment		
	Temp	1500°F	1600°F	1700°F	1500°F	1600°F	1700°F	1500°F	1600°F	1700°F
Stress	45							X		
	30	X			X					
	25	X	X			X			X	
	15	X	X	X	X	X	X			X
	10			X, I			X			
	8			X, I			X			
	6			X						

2.1. Material preparation

Plates and weld wires of HA282 were acquired from Haynes International, Inc. Table 3 shows compositions of the plate and wires in comparison with the nominal composition. Welded HA282 and HA282 weldment materials were prepared at GE Power. All the materials were welded within the ranges stated for the Manual TIG as outlined on Page 19 of the Haynes Brochure ¹. The welded and weldment plates are shown in Figure 1 and Figure 2, respectively.

Table 3: Compositions of HA282 plate and weld wires.

	Composition (wt%)										
	Ni	Cr	Co	Mo	Ti	Al	Fe	Mn	Si	C	B
Nominal ¹	Bal.	20	10	8.5	2.1	1.5	<1.5	<0.3	<0.15	0.06	0.005
Plate	Bal.	19.52	10.33	8.8	2.12	1.39	0.49	0.04	<0.05	0.066	0.005
Wire	Bal.	19.43	10.42	8.55	2.16	1.5	0.68	0.05	<0.05	0.064	0.005

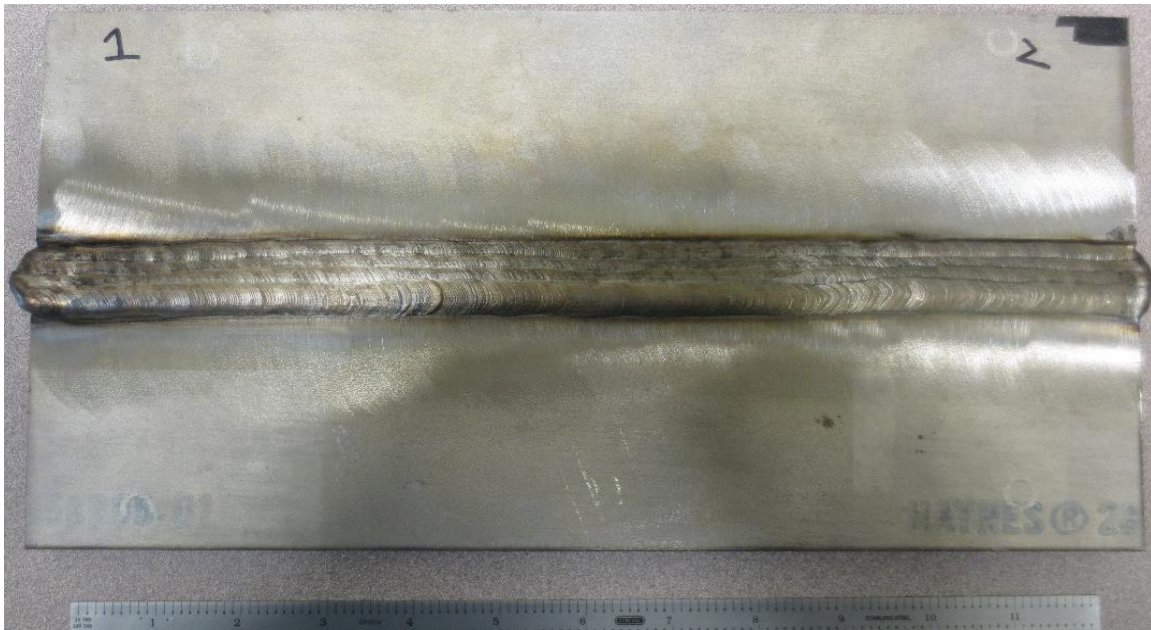


Figure 1: The welded HA282 plate.

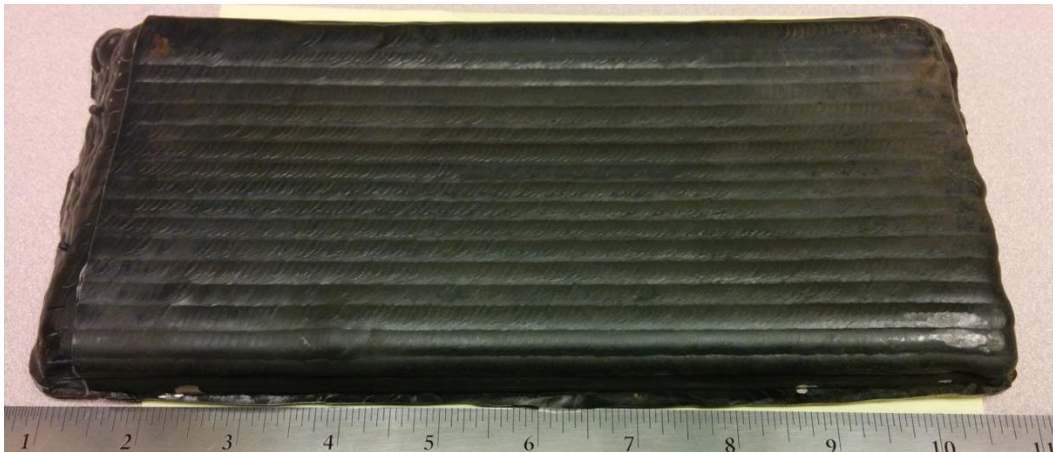


Figure 2: The HA282 weldment (100% weld) plate (9" L x 4" W x 0.5" H). Welding path was along the 9" side.

A post-weld heat treatment (PWHT) was conducted after welding. Table 4 summarizes various PWHTs that have been tried in the study. Among them, the PWHT2 is a full post-weld heat treatment specified in Haynes Brochure ¹. It is a standard PWHT for the combustion liner application, and has been applied to the majority of the specimens in the current study. The PWHT3b is an alternative PWHT condition without the solution step. It has been applied to some specimens together with PWHT3a and PWHT1 to achieve different levels of inhomogeneity in weld material. Among the mechanical testing specimens, PWHT2 was applied to all tensile and most creep specimens. PWHT3a was

applied to a few creep tests for local strain measurement (PD and DIC techniques) in order to compare the effect on mechanical properties.

Table 4: Post-weld heat treatment conditions.

Designation	Condition	
PWHT2	Solution	2075°F (1135°C) / 30min → Air cool
	Primary age	1850°F (1010°C) / 2h → Air cool
	Secondary age	1450°F (788°C) / 8h → Air cool
PWHT3a	Primary age	1850°F (1010°C) / 1h → Air cool
	Secondary age	1450°F (788°C) / 8h → Air cool
PWHT3b	Primary age	1850°F (1010°C) / 2h → Air cool
	Secondary age	1450°F (788°C) / 8h → Air cool
PWHT1	Solution	1895°F (1035°C) / 30min → Air cool
	Primary age	1850°F (1010°C) / 2h → Air cool
	Secondary age	1450°F (788°C) / 8h → Air cool

After PWHT, tensile and creep specimens were extracted from the welded HA282, in transverse direction, and HA282 weldment plates, in transverse and longitudinal directions, with respect to the weld direction. Specimens for static thermal exposure heat treatments (listed in Table 1) were encapsulated in silica tubes back filled with Argon gas.

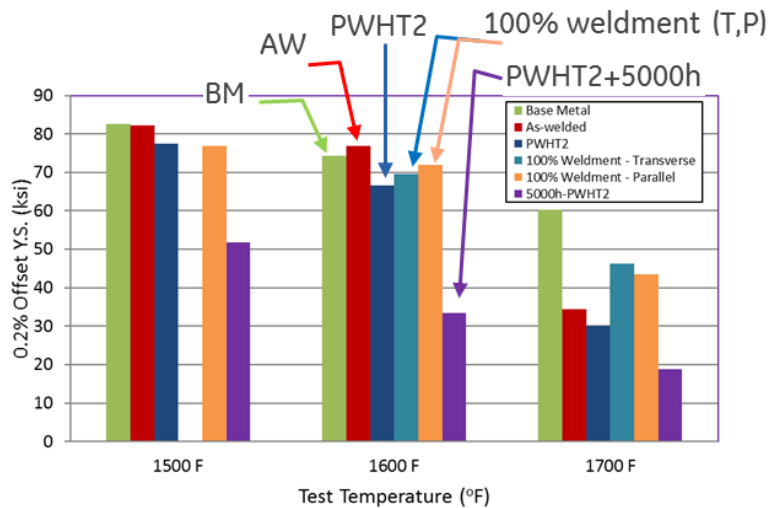
2.2 Mechanical properties

2.2.1. Tensile properties

The tensile properties of the HA282 base metal, welded HA282 (as-weld and PWHT2), and HA282 weldment (PWHT2), measured at 1500°F, 1600°F and 1700°F, are summarized in Table 5 and Figure 3. Comparison charts between literature data and measured properties are also shown in Figure 4. As expected, ductility (% elongation or % strain to failure) increased with increase in the test temperature in all four types of materials. Tensile ductility decreased from base metal to 100% weldment to welded HA282 materials, with as-welded being the least ductile among all test temperatures. Welded and 100% weldment (PWHT2) HA282 materials did not show significant debit in yield strength (YS) and ultimate tensile strength (UTS) as compared to base metal at 1500°F and 1600°F, however, showed the largest drop at 1700°F compared to base metal. No difference in YS and UTS between two orientations (transverse vs. parallel) was observed for 100% weldment. A significant reduction in tensile strength in 5000 hrs exposure specimens at all temperatures, was also seen likely due to degraded strengthening microstructure due to long thermal exposure. Microstructure evaluation was carried out to understand the lower ductility in 5000 hrs exposure specimens.

Table 5: Experimentally measured tensile properties of base metal, as-welded and PWHT2 welded, and 100% weldment HA282 materials, including 5000 hrs exposure specimens, at 1500, 1600 and 1700°F.

Material	Orientation	Test Temperature	UTS (ksi)	0.02% Offset Y.S. (ksi)	0.2% Offset Y.S. (ksi)	Modulus (ksi)	Elongation (%)
Base Metal HA282	Transverse	RT	168	93.5	101.6	30310	29.6
		1500°F	101.1	77.9	82.7	22610	24.2
		1600°F	79.8	70.7	74.4	20730	34.8
		1700°F	60.3	58.7	60.2	20280	47
	Longitudinal/Parallel	1500°F	100.3	77.9	83.4	22070	27.3
		1600°F	81.2	71.6	74.7	20080	35.5
Welded HA282 (as-weld)	Transverse	1500°F	88.6	75.7	82.3	18300	3.3
		1600°F	84.3	67.7	76.9	19600	5.7
		1700°F	38.4	27.2	34.4	17000	12.1
Welded HA282 (PWHT2)	Transverse	1500°F	87	69.8	77.5	15700	10.1
		1600°F	76.5	66.5	66.6	28700	10
		1700°F	41.2	24.8	30.1	22900	23
100% Weldment HA282	Transverse	1600°F	69.6	61.07	69.53	17850	22.1
		1700°F	46.4	39.8	46.34	16220	21.5
	Longitudinal/Parallel	1500°F	89.7	79	76.81	15520	15.9
		1600°F	72.3	58.4	71.99	18230	31.1
		1700°F	43.6	37.75	43.5	15430	33.5
Welded HA282 5000 hr exposure	Transverse	1500°F	52.4	45.9	51.8	18000	3.7
		1600°F	41	27.6	33.4	19200	17.1
		1700°F	23.8	14.7	18.9	16300	18.5



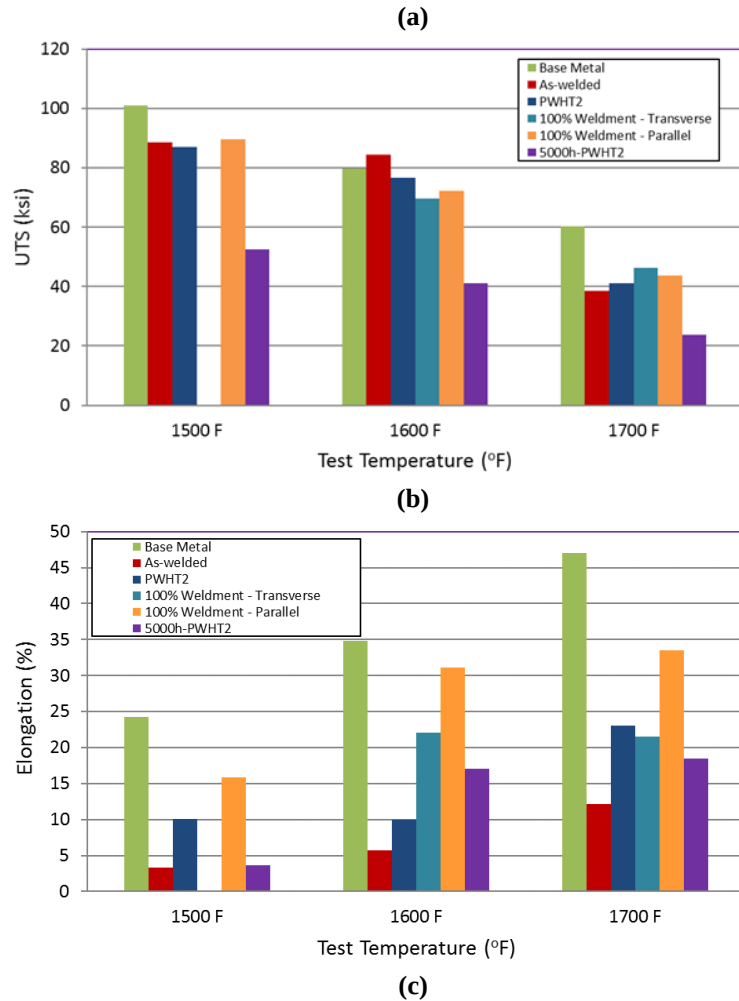
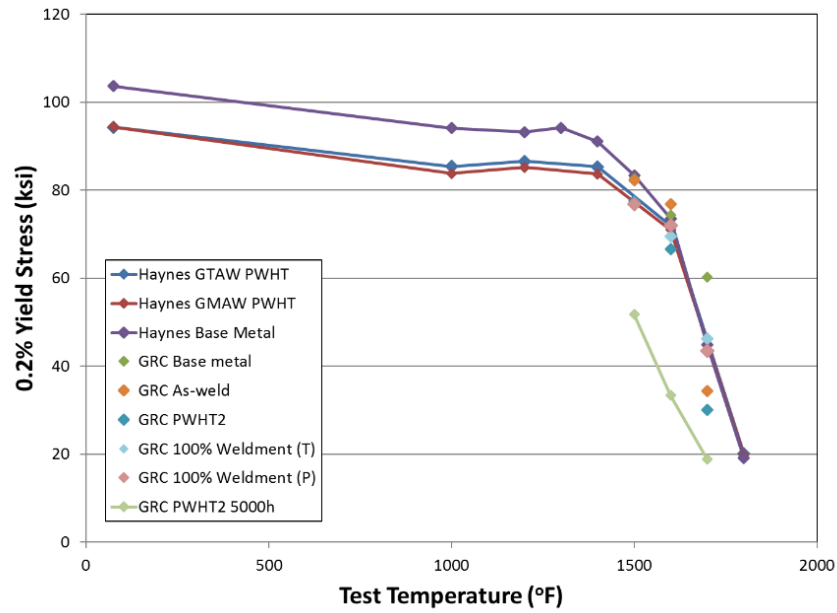
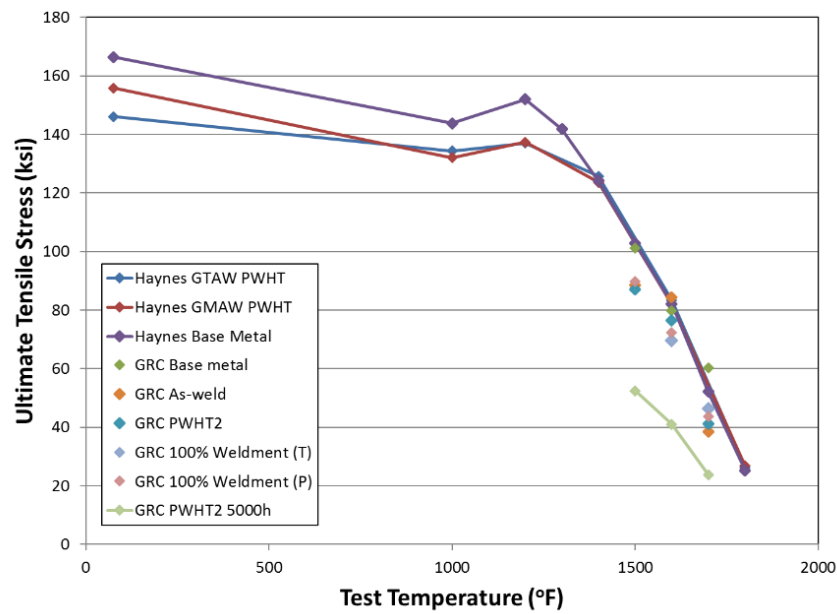


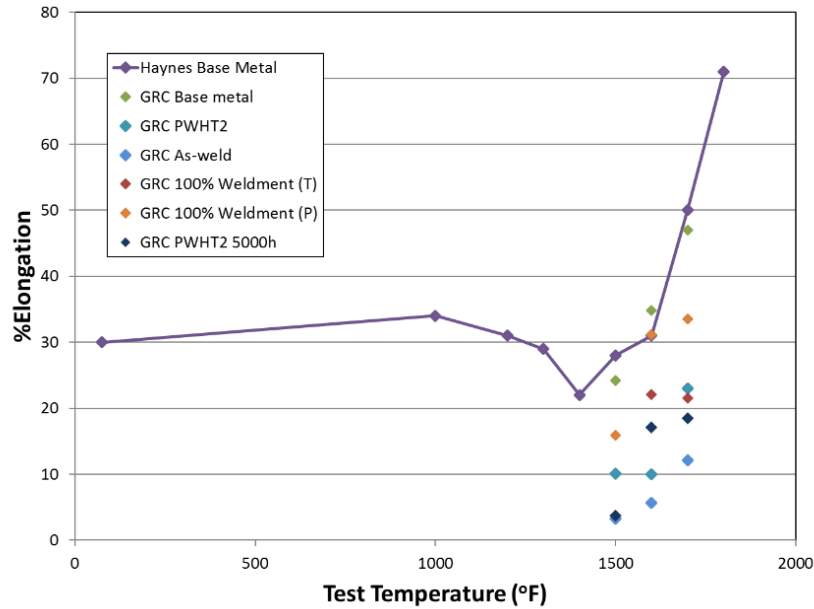
Figure 3: Tensile properties of base metal, as-welded and PWHT2 welded, 100% weldment HA282 materials, including 5000 hrs exposure specimens, at 1500°, 1600° and 1700°F: (a) 0.2% offset yield strength (0.2% Y.S.), (b) ultimate tensile strength (UTS), and (c) % elongation/estimated strain-to-failure.



(a)



(b)



(c)

Figure 4: Comparison of experimentally measured tensile properties with literature data, of base metal, as-welded and PWHT2 welded, 100% weldment HA282 materials, including 5000 hrs exposure specimens, at 1500°, 1600° and 1700°F: (a) 0.2% offset yield strength (0.2%Y.S.), (b) ultimate tensile strength (UTS), and (c) ductility (% elongation or % strain-to-failure).

2.2.2. Creep properties

Creep test results of welded (standard PWHT2, and non-standard PWHT3a) HA282 materials, consisting of the base metal, HAZ and weldment, at 1500°F, 1600°F and 1700°F, are summarized below in Table 6, Table 7, and Figure 5. Creep testing was performed at several combinations of stress levels (6 to 45 ksi) and temperatures (1500 to 1700°F) to obtain creep properties at a variety of test conditions and provide a rich dataset for creep model development in Task 5. Note that the columns “Final creep reading (hrs)” and “Deformation (%)” report the last recorded time and strain readings, respectively, during the creep test. Elongation (%) was measured from fractured specimens according to the ASTM standard.

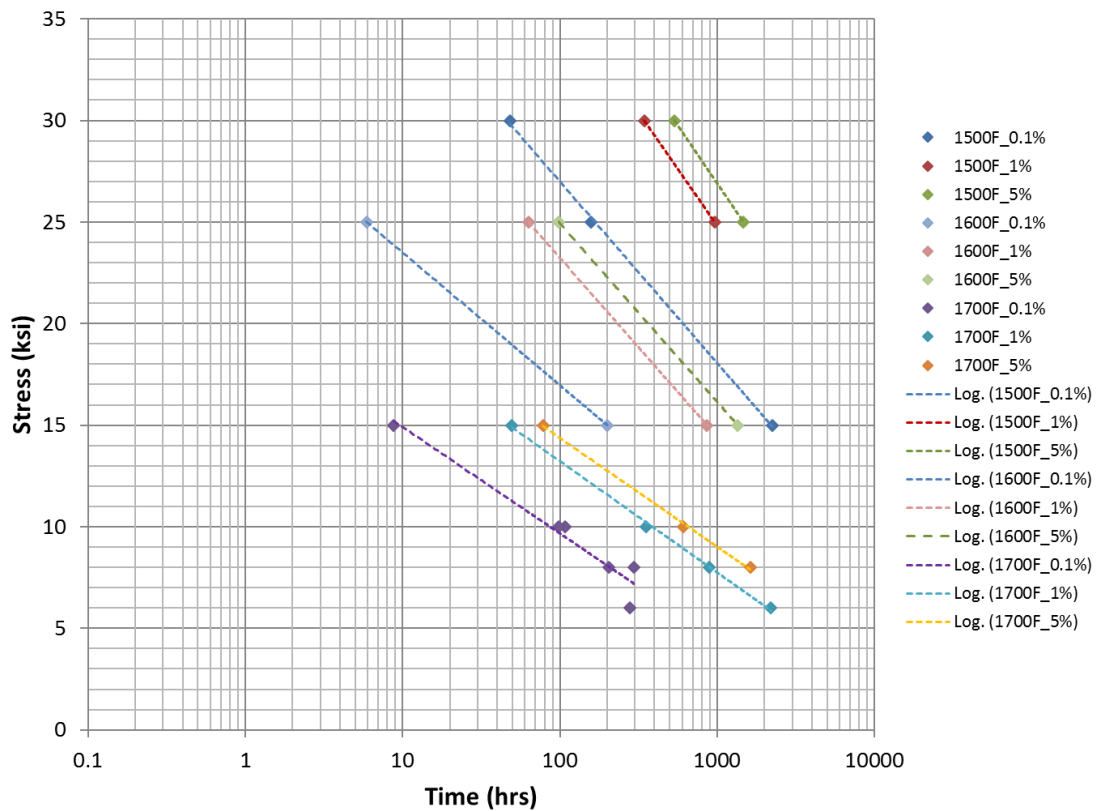
Table 6: Experimentally measured creep properties of PWHT2 HA282 material, tested at 1500°, 1600° and 1700°F. Test-3, Test-7 and Test-9 were interrupted at 7884, 200 and 1000 hrs, respectively.

Test ID	Temperature (°F)	Stress (ksi)	Rupture Time (hrs)	Elongation (%)	Final creep reading (hrs)	Deformation (%)
Test-1	1500	30	563.3	7.4	562.6	6.511
Test-2	1500	25	1606.0	9.4	1604.9	8.010
Test-3	1500	15	-	-	7883.6	0.440
Test-4	1600	25	124.7	19.2	124.0	16.550
Test-5	1600	15	1677.8	15.0	1676.5	13.775
Test-6	1700	15	106.6	21.3	105.7	16.703
Test-7	1700	10	-	-	200.0	0.285

Test-8	1700	10	653.7	7.0	652.7	6.581
Test-9	1700	8	-	-	1000.0	0.967
Test-10	1700	8	1637.2	6.5	1636.4	5.299
Test-11	1700	6	3604.1	6.3	3603.7	4.469

Table 7: Experimentally measured time to achieve x% (0.1, 0.2, 0.5, 1, 2, 5%) creep strain values in PWHT2 welded HA282 material, tested at 1500, 1600, and 1700°F. Test-3, Test-7 and Test-9 were interrupted at 7884, 200 and 1000 hrs, respectively.

Test-ID	Temperature °F	Stress ksi	Time to achieve x% strain						
			0.1% (h)	0.2% (h)	0.5% (h)	1.0% (h)	2.0% (h)	5.0% (h)	Rupture (h)
Test-1	1500	30	48.2	120.5	248.8	343.5	426.5	534.8	563.3
Test-2	1500	25	157.6	396.2	727.7	964.8	1184.1	1471.2	1606.0
Test-3	1500	15	2243	4631.3	-	-	-	-	Interrupted
Test-4	1600	25	5.9	21.8	46.7	63.1	78.3	98.3	124.7
Test-5	1600	15	200.2	417.8	672.8	862.6	1059.8	1352	1677.8
Test-6	1700	15	8.8	21.3	38.8	49.6	61.2	78.2	106.6
Test-7	1700	10	107.4	165.5	-	-	-	-	Interrupted
Test-8	1700	10	98.3	153	267.8	355.1	457.5	609.4	653.7
Test-9	1700	8	295.7	470.9	749.7	-	-	-	Interrupted
Test-10	1700	8	205.9	370.5	637.9	894.2	1199.9	1624	1637.2
Test-11	1700	6	279.8	809.6	1509	2197.3	2903.2	-	3604.1



(a)

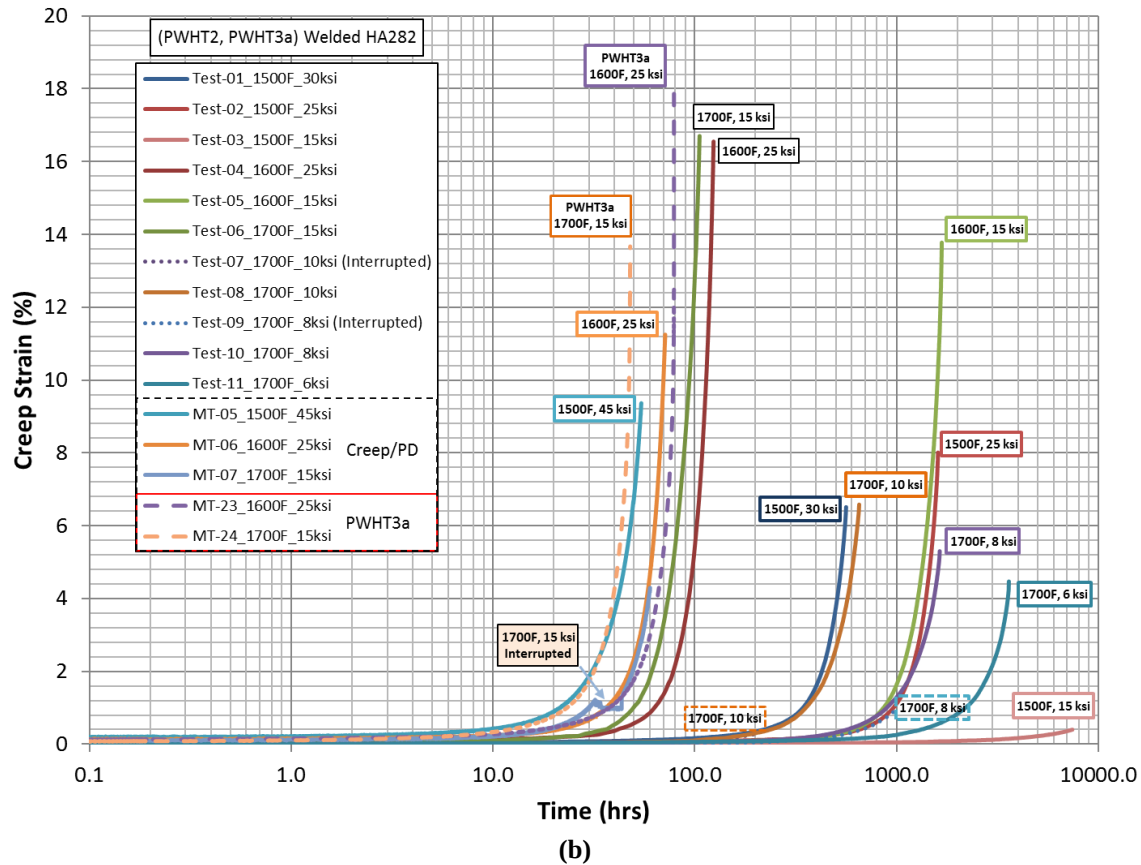


Figure 5: Experimentally measured creep properties of welded HA282 material (after PWHT2 or PWHT3a), tested at 1500, 1600, and 1700°F: (a) semi-log plot of Stress (ksi) vs. Time (hrs), and (b) semi-log plot of Creep strain (%) vs. Time (hrs).

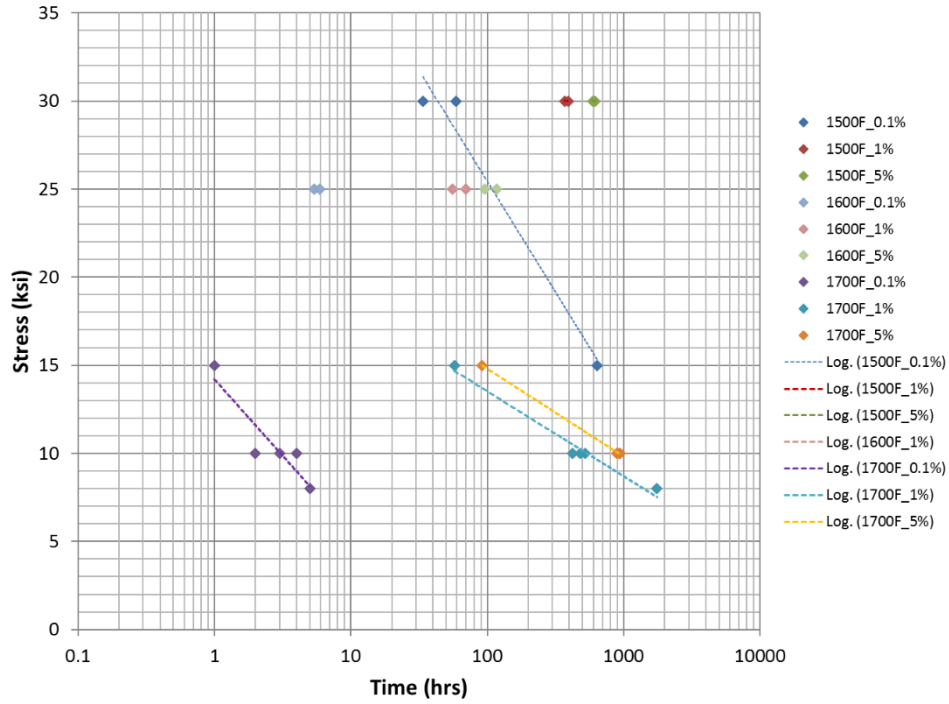
Tensile and creep specimens, of 100% weldment HA282 material, were machined in two orthogonal orientations termed as transverse (T) and parallel (P) directions, which are normal and parallel to the welding directions, respectively (see 100% weld plate in Figure 2). The tensile and creep testing plan for 100% weldment is shown in Table 5 and Table 8, respectively. Creep test conditions for weldment specimens were similar to welded (PWHT2) specimens, however, some combinations were eliminated due to limited number of available specimens. Creep test results from weldment specimens are shown below in Table 8, Table 9 and Figure 6. Moreover, creep properties between welded and 100% weldment materials are also compared in Figure 7 and Figure 8. Moreover, some of creep specimens were tested with 3D DIC based local strain measurement technique (Table 10).

Table 8: Experimentally measured creep properties of 100% weldment HA282 material. tested at 1500°, 1600° and 1700°F. Tests T2 and T7 were interrupted at 5000 and 2000 hrs, respectively.

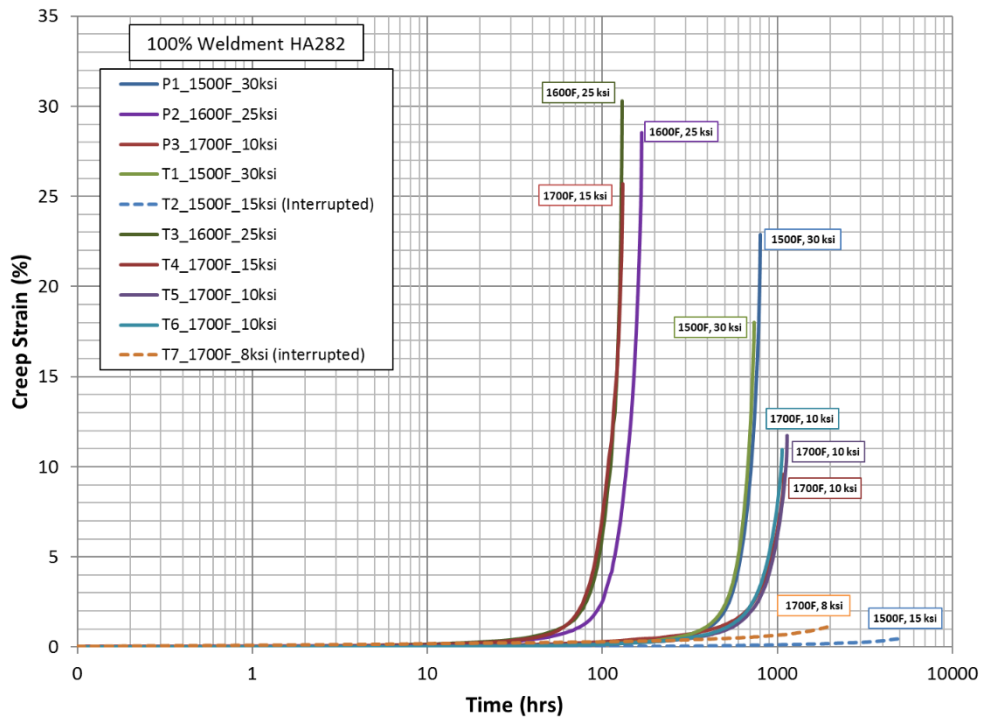
Test ID	Temperature (°F)	Stress (ksi)	Rupture time (h)	Elongation (%)	Final creep reading (hrs)	Deformation (%)
T1	1500	30	734.6	24.3	733.5	18.021
P1	1500	30	802.3	27.1	802.2	22.902
T2	1500	15	-	-	5000.0	0.455
T3	1600	25	129.6	36.2	129.6	30.274
P2	1600	25	169.3	29.4	168.5	28.554
T4	1700	15	131.0	32.8	130.5	25.688
T5	1700	15	1142.9	14.5	1141.4	11.721
T6	1700	10	1066.0	14.1	1064.7	10.944
P3	1700	10	1087.4	11.5	1086.9	9.604
T7	1700	8	-	-	2000.0	1.154

Table 9: Experimentally measured time to achieve x% creep strain values of 100% weldment HA282 material, tested at 1500, 1600, and 1700°F. Tests T2 and T7 were interrupted at 5000 and 2000 hrs, respectively.

Test-ID	Temperature °F	Stress ksi	Time to achieve x% strain						
			0.1%	0.2%	0.5%	1.0%	2.0%	5.0%	Rupture
			(hrs)	(hrs)	(hrs)	(hrs)	(hrs)	(hrs)	(hrs)
T1	1500	30	33.8	103.7	245.9	368.5	476.8	594.6	734.6
P1	1500	30	59.2	125.4	270.4	392	500.5	617.2	802.3
T2	1500	15	639.40	2058.7	-	-	-	-	5000.0
T3	1600	25	5.4	12.5	35.1	55.9	75.1	95.6	129.6
P2	1600	25	5.9	19.2	44.7	69.9	93	117.7	169.3
T4	1700	15	6.2	17.2	40.1	57.6	72.9	91.6	131.0
T5	1700	10	37.3	120.6	317	521.1	709.8	950.3	1142.9
T6	1700	10	47.5	121.7	300.5	483.6	668.3	888.1	1066.0
P3	1700	10	17.7	59.6	198.1	421.1	662.6	922.3	1087.4
T7	1700	8	1.4	16.7	605.0	1758.7	-	-	2000.0



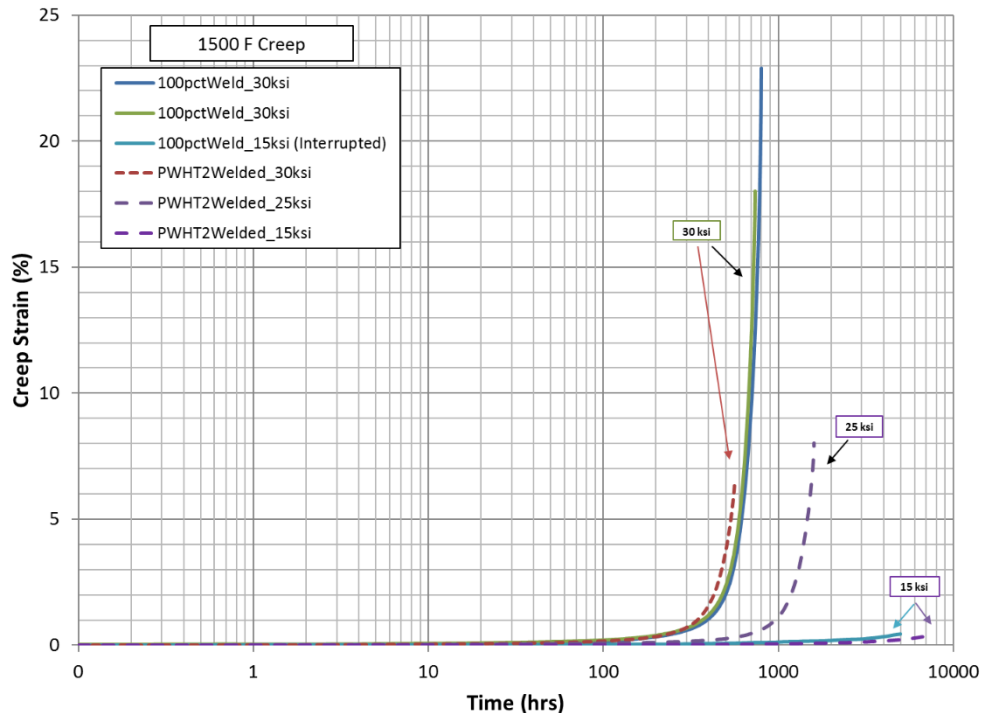
(a)



(b)

Figure 6: Experimentally measured creep properties of 100% weldment HA282 material (after PWHT2), tested at 1500, 1600 and 1700°F: (a) semi-log plot of Stress (ksi) vs. Time (hrs), and (b) semi-log plot of Creep strain (%) vs. Time (hrs).

In order to examine the differences in creep strain values between 100% weldment and PWHT2 welded specimens, separate charts which compare creep data at different stresses at 1500°F, 1600°F and 1700°F are shown in Figure 7. Notably, for same applied stress (at any temperature), welded HA282 shows higher strain at a given time in the tertiary regime, and shorter creep rupture life. However, with limited stress conditions for comparison, it is difficult to say if there is a trend or it's within scatter.



(a)

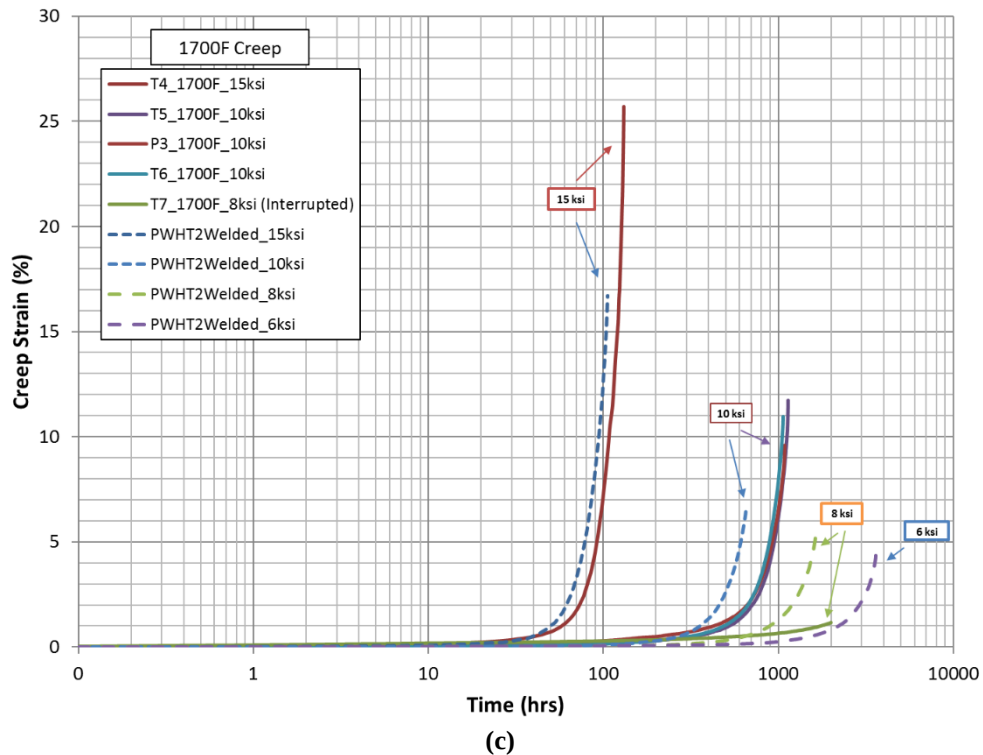
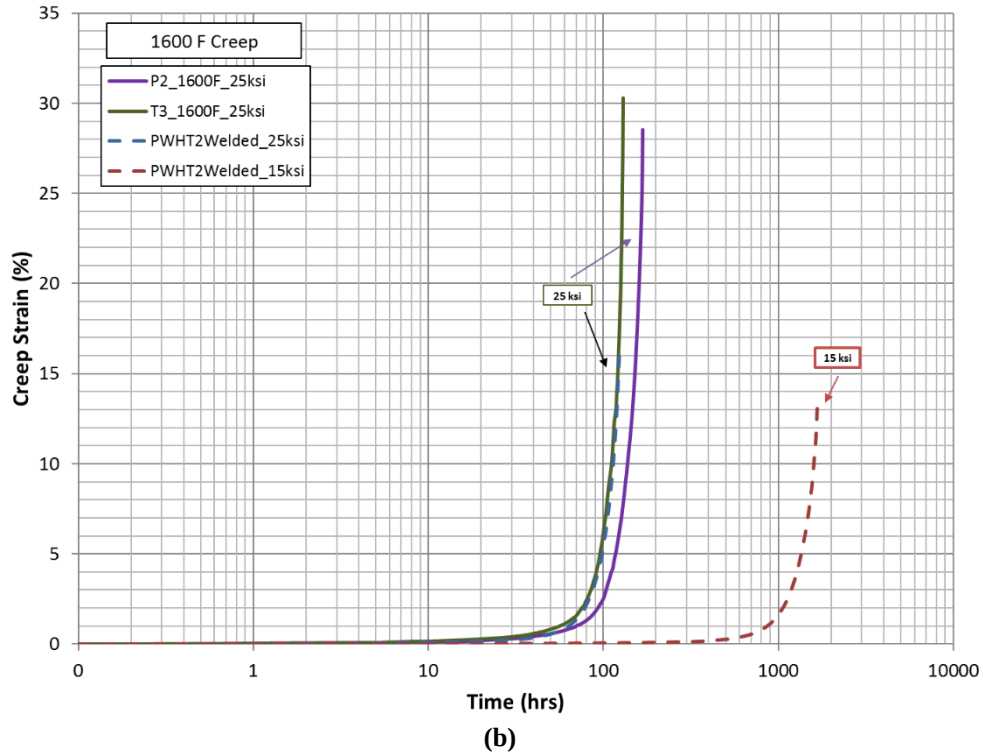
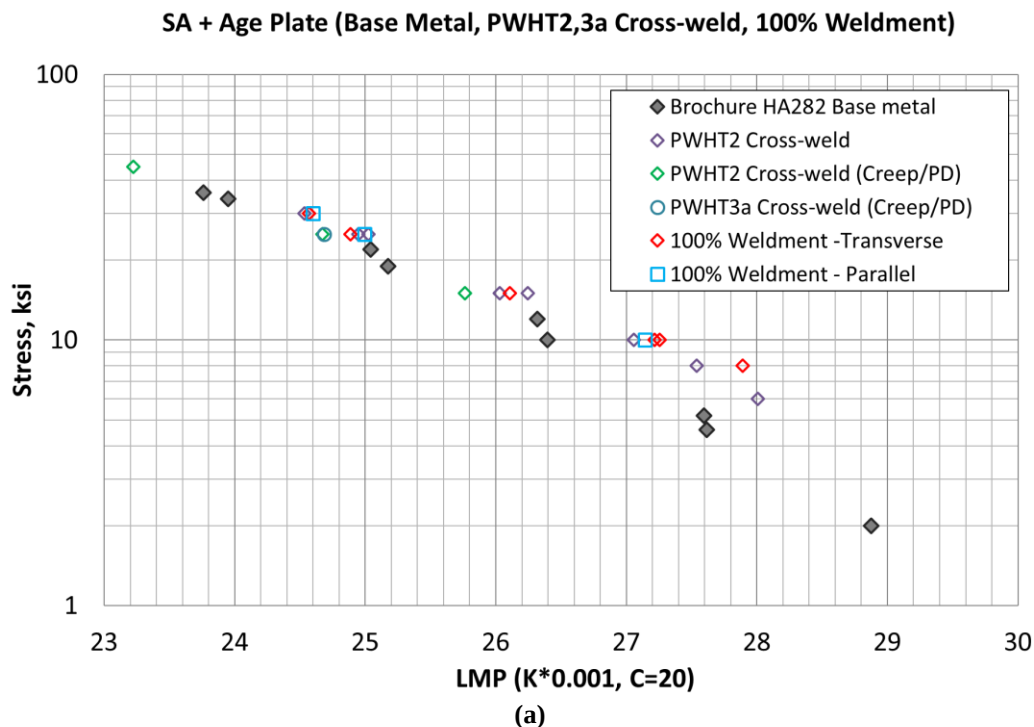


Figure 7: Comparison of creep strain (%) vs. time (hrs) between PWHT2 welded and 100% weldment HA282 materials, tested at (a) 1500°F, (b) 1600°F, and (c) 1700°F. Creep specimens from PWHT2 welded HA282 material are shown by dotted lines.

In addition to the creep curves (creep strain vs. time) shown above for welded (in PWHT2, PWHT3a) and 100% weldment HA282 material, creep test results are also plotted in terms of Larson-Miller parameters (LMP) for comparison with existing literature data from base metal and welded HA282 material. For example,

Figure 8a and b compare GRC data with HA282 base metal creep data from Haynes data sheet ¹ for 1% creep strain and rupture, respectively. Apparently, no debit in creep properties was observed between base metal and welded HA282 materials. Welded material with non-standard PWHT condition (PWHT3a) that was applied to generate inhomogeneous γ' microstructure, didn't show noticeable effect on creep properties and compared well with welded material with standard PWHT2. Moreover, 100% weldment specimens obtained from transverse and parallel/longitudinal directions showed similar creep properties.

Figure 8c compares GRC data with a broader creep database on HA282 material (literature + Haynes + ORNL + GE DoE) for creep rupture. All the creep data obtained from both PWHT2 and PWHT3a, as well as from 100% weldment, collapsed on existing trend on base metal and welded HA282 material. It may appear that there is difference in creep properties between base metal and weld materials at higher side of LMP parameter (base metal and welded/weldment curves diverging to the right side in the plots). However, when compared with proprietary GE data (not shown here) for base metal, no difference between welded material and base metal was seen. Thus, it can be concluded that welded HA282 material is not worse or better than base metal.



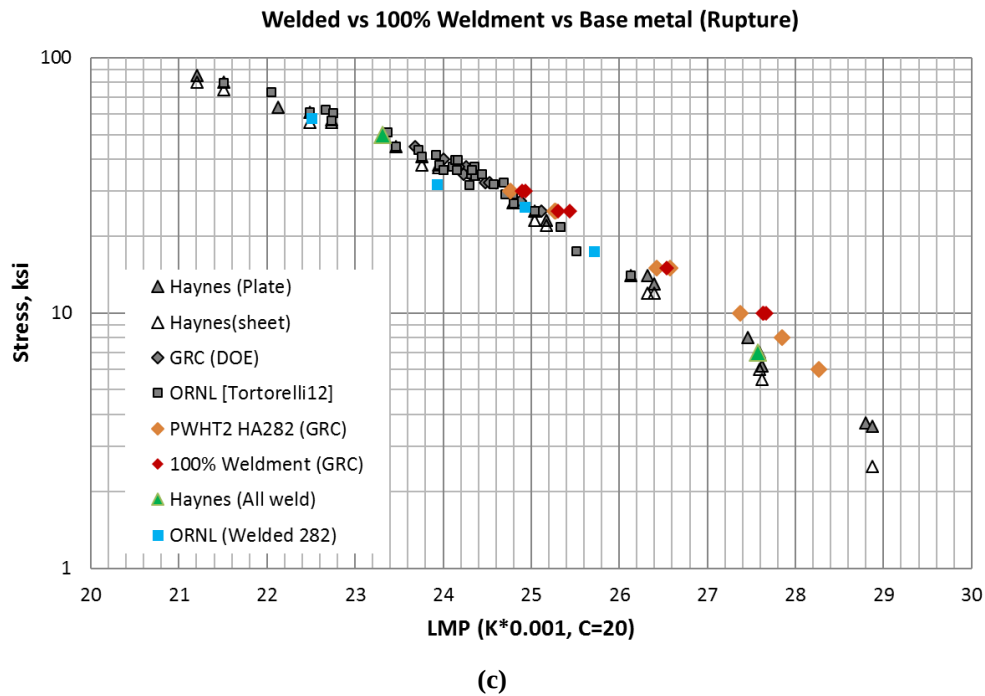
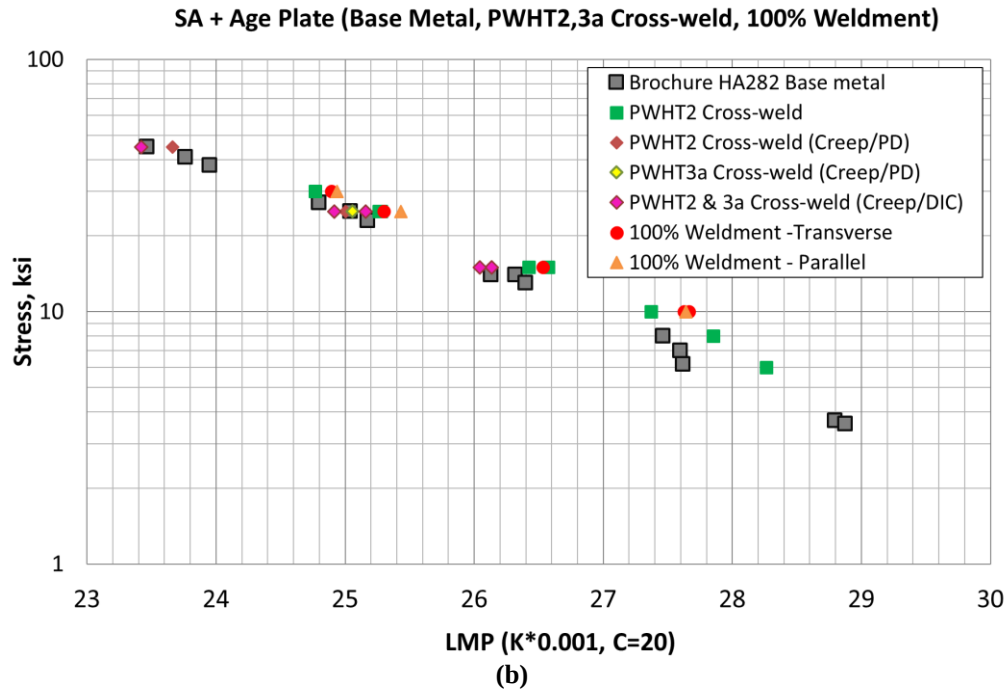
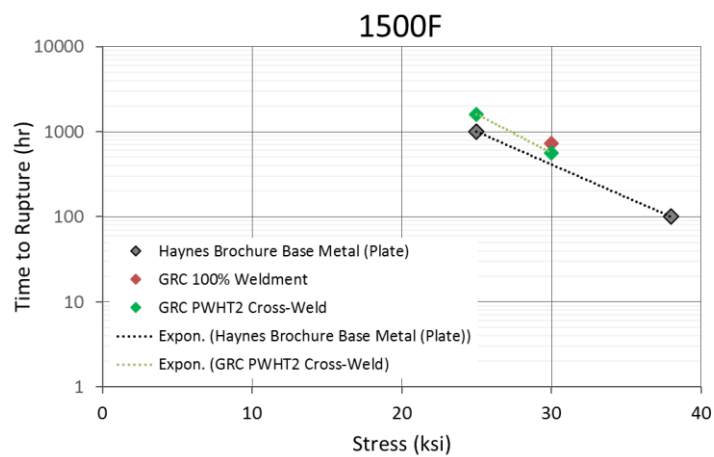


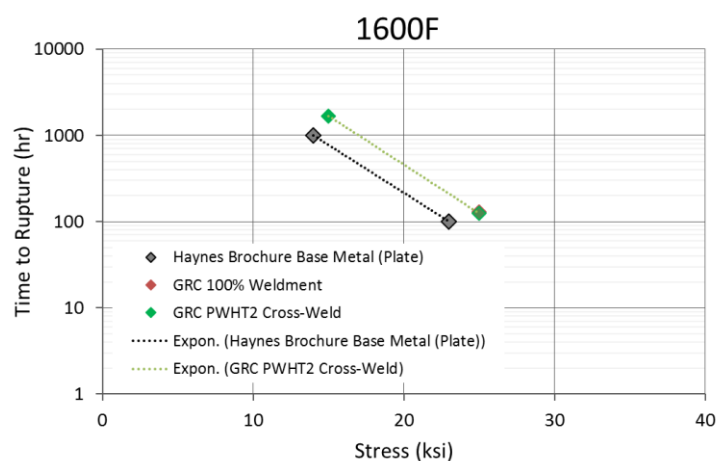
Figure 8: Comparison between experimentally measured creep properties (at GRC), for base metal, 100% weldment, PWHT2 and PWHT3a welded HA282 materials, tested at 1500°F, 1600°F, and 1700°F, and literature data: (a) 1 % creep strain LMP plot for GRC and Haynes Brochure data¹, (b) creep rupture LMP plot for GRC and Haynes Brochure data¹, and (c) creep rupture LMP plot compares GRC data with broader dataset taken from literature, Haynes Brochure¹, ORNL, and previous GE/DoE program. Creep rupture data obtained from creep tests that were run along with potential drop (PD) and DIC measurements, are also included.

Figure 9 shows the rupture time for base metal ¹, 100% weldment, and welded (PWHT2) HA282 materials. Although the data points were taken from different sources, different heat of materials and possibly different specimen geometries, a clear general trend can be seen. The rupture time for 100% weldment and PWHT2 welded materials appear to be longer than the rupture time for the HA282 base metal.

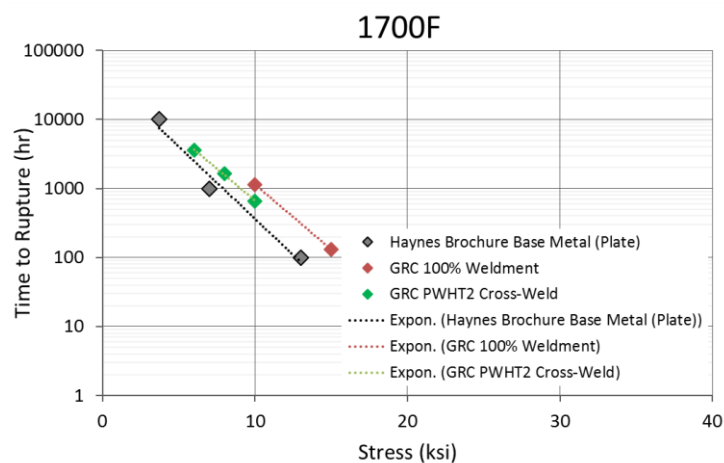
Figure 10 shows the time to reach 1% strain vs. stress plots for the three types of materials. This data confirms again that the 100% weldment and PWHT2 welded materials have higher strength than the base metal. The trend in strain accumulation (based on time to 1% strain through rupture) suggests that the base metal strengthens less than the weld during the creep process. There are only a couple of conditions where 100% weldment can be compared directly with PWHT2 welded. Some of these data point either nearly coincide, or 100% weldment shows longer time to reach 1% strain or rupture than PWHT2 welded material, suggesting superior strength than base metal as well as PWHT2 welded materials.



(a)



(b)



(c)

Figure 9: Rupture time (hours) for various stress values (ksi) for HA282 base metal, 100% weldment, and welded (PWHT2) specimens, (a) 1500°F, (b) 1600°F, and (c) 1700°F.

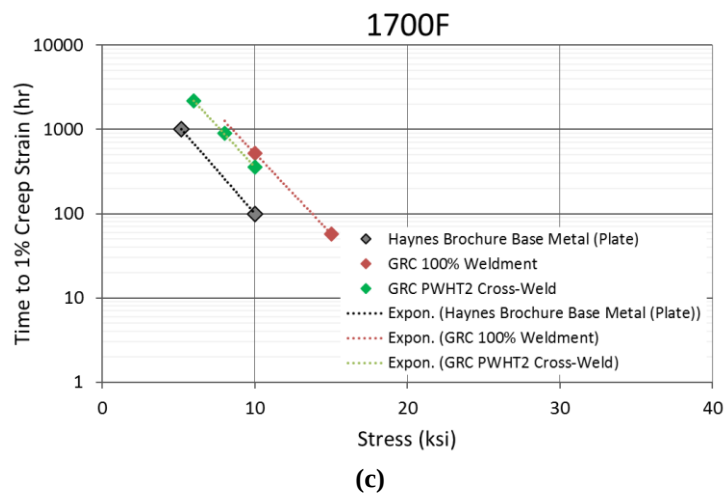
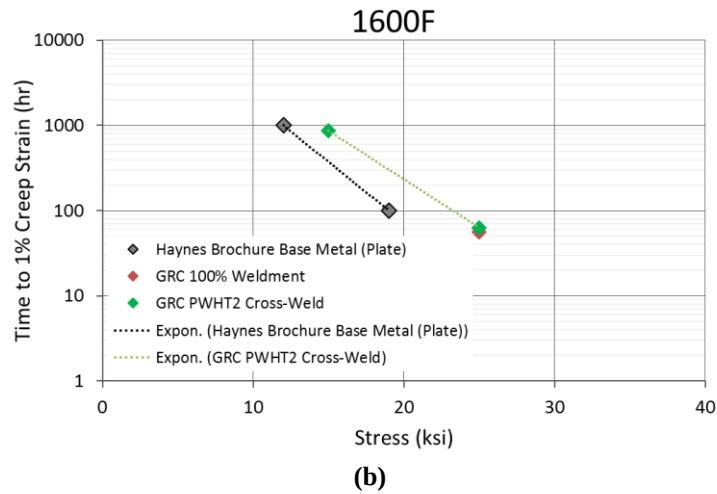
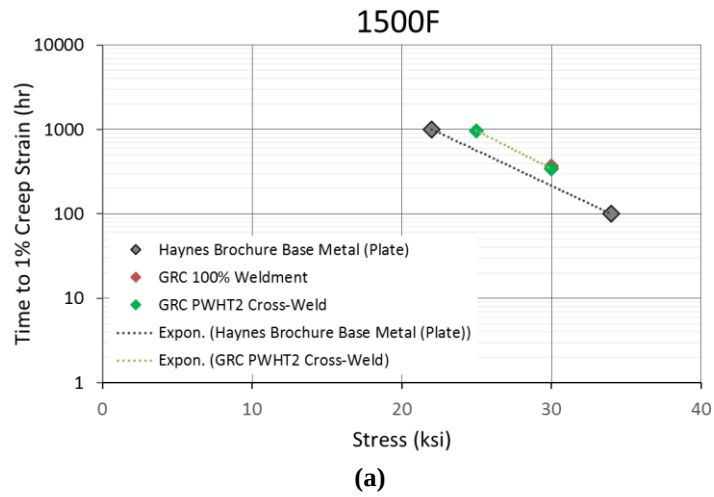


Figure 10: Time to reach 1% strain for various stress values (ksi) for HA282 base metal, 100% weldment, and welded (PWHT2) specimens, (a) 1500°F, (b) 1600°F, and (c) 1700°F.

2.2.3. Local creep strain measurements

In-situ local creep strain measurements were performed at selected creep test conditions (see below in Table 10) using digital image correlation (DIC) or potential drop (PD) based techniques. These DIC or PD -based creep tests were not performed in a conventional creep test setting, and used servo-hydraulic frame (due to availability and requirements of the PD-based set-up). The results provided data for validation of the creep model.

Table 10: Creep testing matrix, along with 2-D DIC or PD for the local strain measurements, for PWHT2, PWH3a welded and 100% weldment HA282 materials. Tests highlighted in grey are completed. For 100% weldment materials - “T” – Transverse, and “P” – Parallel/Longitudinal.

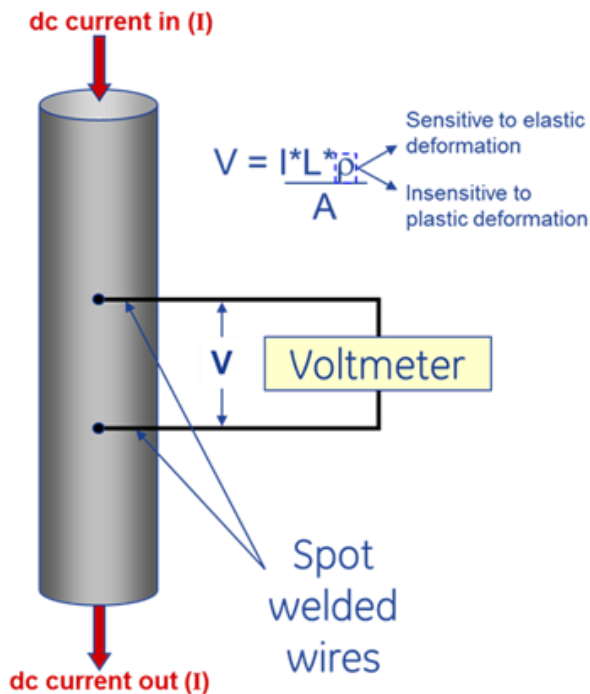
Material	Temperature (°F)	Stress (ksi)	PD/Creep Testing	DIC/Creep Testing
PWHT-2 Welded HA282	1500	45	MT-05	-
	1600	25	MT-06	-
	1700	15	MT-07	-
	1500	45	-	MT-08
	1600	25	-	MT-09
	1700	15	-	MT-10
PWHT-3a welded HA282	1600	25	MT-23	
	1700	15	MT-24	
	1600	25	-	MT-21
	1700	15	-	MT-22
PWHT-2 100% Weldment HA282 (Transverse)	1600	25		MT-16
	1700	15		MT-17
PWHT-2 100% Weldment HA282 (Parallel)	1500	45		MT-18
	1600	25		MT-19
	1700	15		MT-20

For local creep strain measurements, pair-wise potential drop (PD) based method was employed. PD measurements were made in the gauge section of a specimen taken from the welded HA282 material. For this potential drop based approach, it was assumed that the resistivity is invariant with plastic deformation. Potential drop was measured during the test to estimate local “volumetric” strain evolution within each pair during creep loading. In contrast to digital image correlation (DIC)-based full-field measurements of local creep strain (discussed later), PD measurements provide “volumetric” creep strain at different locations within the gauge section (see schematic in Figure 11). According to the presented PD approach, during plastic deformation, for each pair, strain can be estimated from:

$$\text{Strain} = 0.5 * \ln V/V_0$$

where, V is measured potential during the test, and V_0 is an initial potential when no load was applied (baseline) for each pair.

Damage monitor concept



$$\Rightarrow V = I * R$$

Resistance can be related to resistivity through

$$\Rightarrow R = \rho * (L/A), \text{ and therefore:}$$

$$\Rightarrow V = I * \rho * (L/A)$$

Normalizing by an initial voltage reading

$$\Rightarrow V/V_0 = [I * \rho * (L/A)] / [I_0 * \rho_0 * (L_0/A_0)]$$

Assuming constant volume deformation

$$\Rightarrow L * A = L_0 * A_0, \text{ and therefore:}$$

$$\Rightarrow A_0 = L * A / L_0$$

Assuming constant current is maintained ($I = I_0$), and resistivity is invariant with plastic deformation ($\rho = \rho_0$):

$$\Rightarrow V/V_0 = (L/A) / (L_0/A_0)$$

Substituting the constant volume assumption:

$$\Rightarrow V/V_0 = L^2 / L_0^2$$

During plastic deformation

$$\Rightarrow \ln(V/V_0) = 2 * \ln(L/L_0) = 2 * \epsilon_{\text{true}}$$

Figure 11: Schematic illustration of in-situ creep strain measurement methodology using a potential drop (PD) technique.

In the beginning of test campaign, an exploratory PD test was performed at 1600°F in air. The specimen was loaded under displacement control at a rate of 0.002"/min until failure. Pairs of conductive wires were tack-welded within the gauge section of the specimen to measure potential drop for each pair as illustrated in the Figure 12. A constant current of 2.8 A was maintained throughout the test. The results from the PD test are shown in the Figure 13. The strain estimated from the PD approach (Pair-B) compares well with one measured using 0.5" extensometer. In this test, voltage measured from the Pair-C showed some error, possibly due to some cross-connections, and therefore, data from Pair-C is excluded from the plot in Figure 13a. The significant differences in strains for different Pairs (A, B, and D) at higher strain values suggest the localized strain evolution within the gauge section during the tensile loading. Notably, within the elastic region (Figure 13b), PD-based strain measurements show significant differences (as well as larger noise) in strain

values (and resulting slope, i.e. Modulus). This may be due to changes in resistivity which is sensitive to elastic deformation.

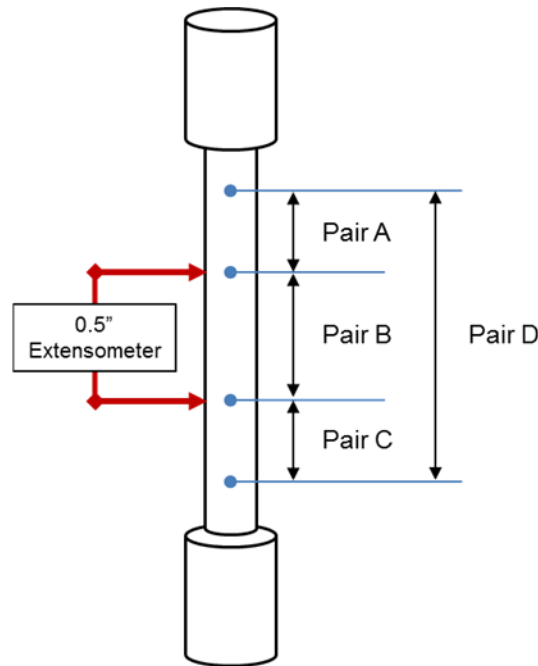


Figure 12: Schematic illustration of in-situ creep strain measurement using a potential drop (PD) approach. Voltage measurements were performed between pairs, e.g., pairs A, B, C and D. Potential drop between pair B was used to verify strain measurements by using a physical extensometer (0.5").

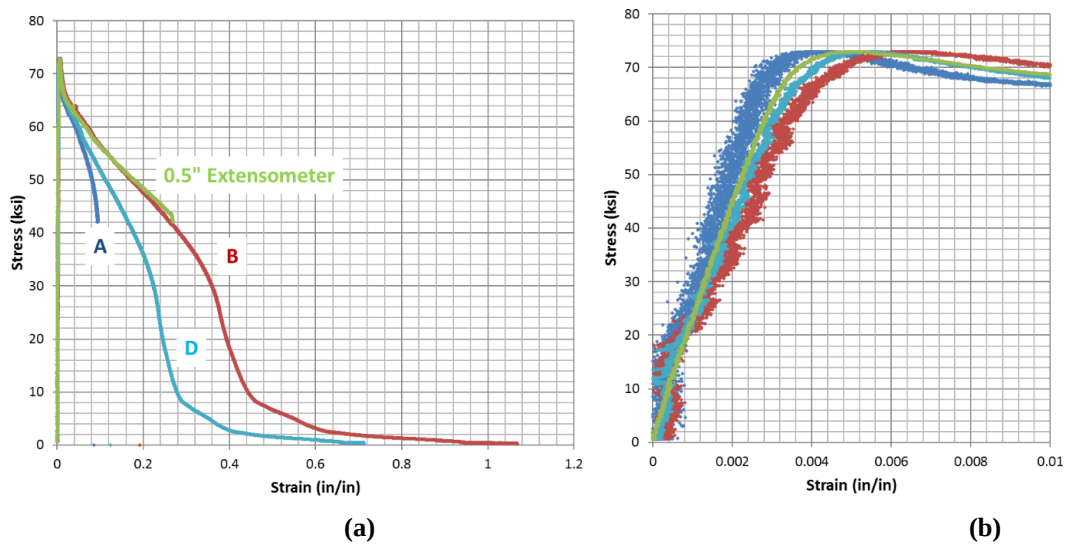


Figure 13: Experimental engineering stress-strain curves of tensile tests of HA282 base metal at 1600°F. A 0.5" extensometer and potential drop between three pairs A, B and D were used to measure strain within the gauge section.

The round bar creep specimens were machined from V-joint welded plate of HA282 material (Figure 14a). This resulted in non-uniform gauge section of the welded specimen consists of weld and base metal. PD probes were tack-welded at 6 locations shown in Figure 14b (1 and 6 in base metal, 2 and 5 near base-metal and weld interface, and 3 and 4 within the middle section of the weld). Two representative PD-based creep measurements, performed on welded (PWHT2) HA282 materials, are presented below (Figure 15, Figure 16).

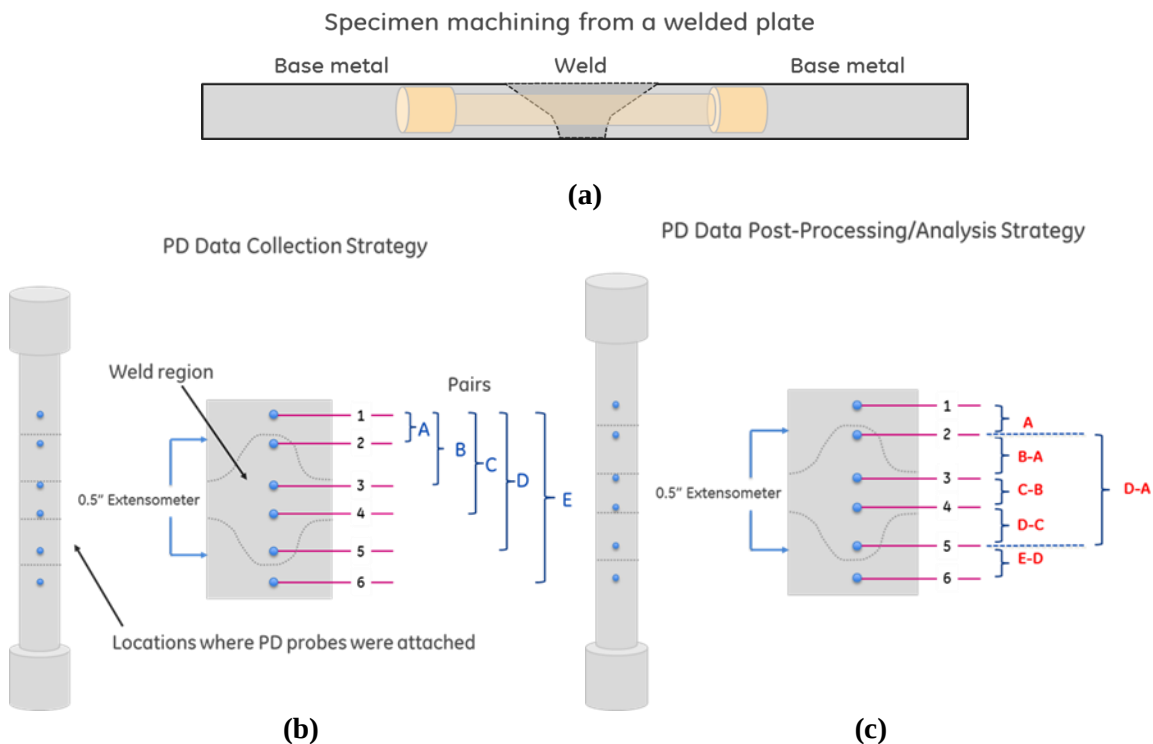
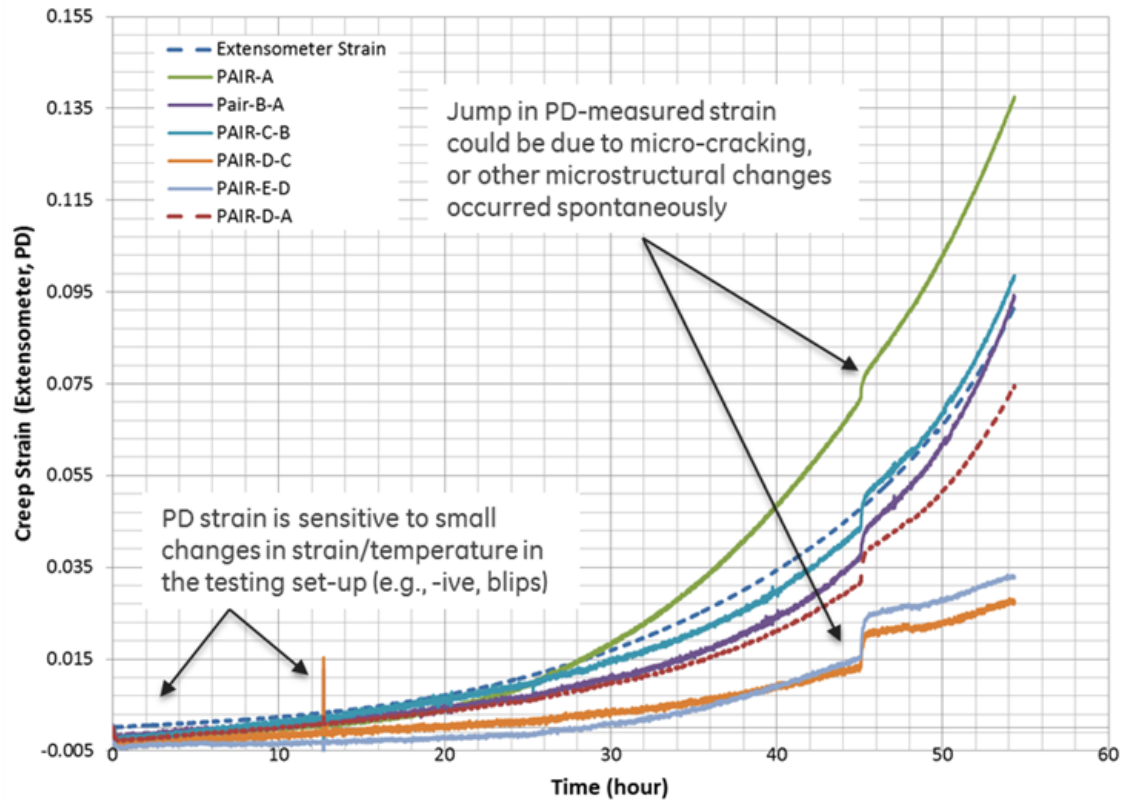


Figure 14: (a) Schematic illustrates non-uniform weld region within the gauge section of the specimen, since the specimen was machined at the half depth of the V-joint welded plate. Schematic illustration of in-situ creep strain measurement using a potential drop (PD) approach: (b) PD data collection strategy where voltage measurements were performed between pairs - A, B, C, D and E. (c) PD data post-processing and analysis strategy where potential drop between consecutive pairs was calculated. A physical extensometer (0.5" gauge length) was used to compare with PD-based strain between pair D-A.

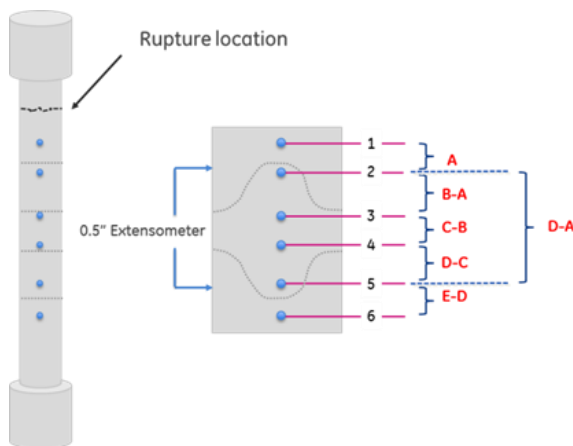
Figure 15 shows experimental creep strain (in/in, normalized) vs. time (hrs) plots for various pairs in welded (PWHT2) HA282, creep tested at 1500°F, 45 ksi. As shown in schematic below (Figure 15b), specimen failed in the gauge section above the location-1 (within the base metal regime). A large-permanent deformation was seen within the regions of pairs A, B-A and C-B as compared to other locations on the specimen, which is also evident in the creep curves for pairs A, B-A and C-B that showed larger creep strains at any time as compared to D-C and E-D pairs. Due to such inhomogeneous deformation in the gauge section, PD probe-measured strain for pair D-A showed lower strain compared to extensometer strain within the same gauge length. Notably, PD-based strain

measurements were also influenced by formation of micro-cracks (such as shown in Figure 15) that formed during early stages of creep deformation and grew throughout the creep test duration. In addition, PD-based strain measurements were also found to be highly sensitive to fluctuations in current and furnace temperature, as well as ambient environment.

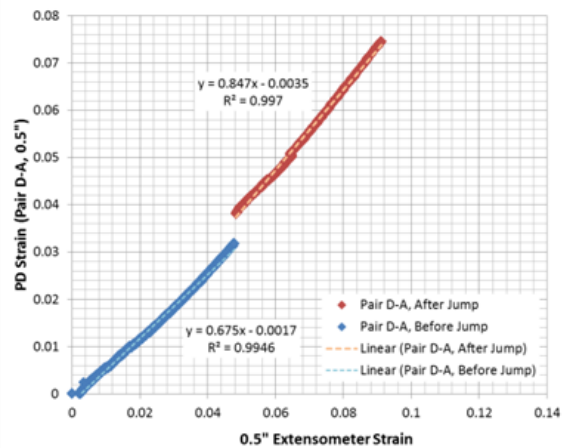
Similarly, Figure 16 shows experimental creep strain (in/in, normalized) vs. time (hrs) curves for various pairs in welded (PWHT2) HA282 material, tested at 1600°F, 25 ksi. As shown in the schematic in Figure 16b, the specimen failed in the gauge section below the location-6 (i.e. larger deformation in the lower half of the specimen as opposed to the upper half of the specimen). A large-permanent deformation was seen within the regions of pairs C-B and D-C as compared to other locations. Other than the necking region, most deformation within the gauge section was localized in regions of pairs C-B and D-C, resulting in to relatively better correlation between PD probe-measured strain for pair D-A and the extensometer strain within the same gauge length at lower strain values. In both cases, shown below, Pair C-B (weld zone) has always shown slightly higher strain than Pair B-A and Pair D-C (consist of base-metal and weld zones) irrespective of the rupture location. This suggests that the weld zone might have accumulated slightly larger creep strain than the base metal for a given total strain across the entire gauge length.



(a)

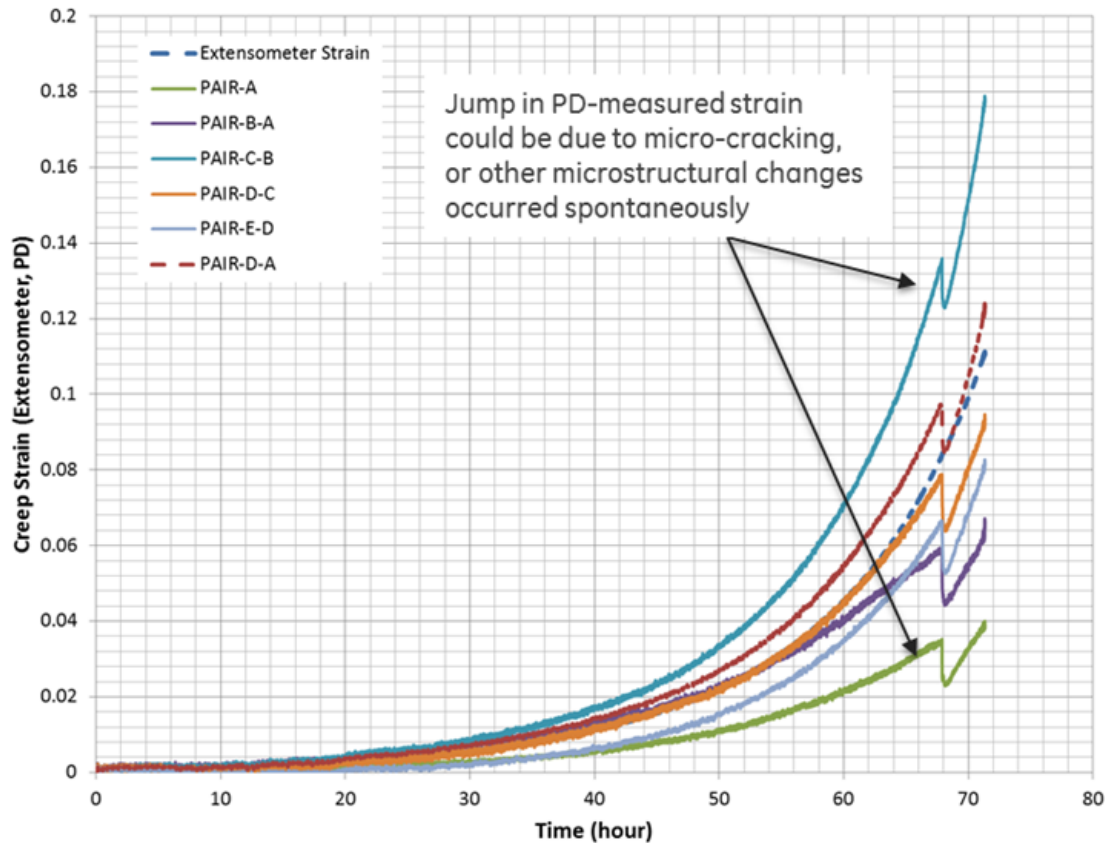


(b)

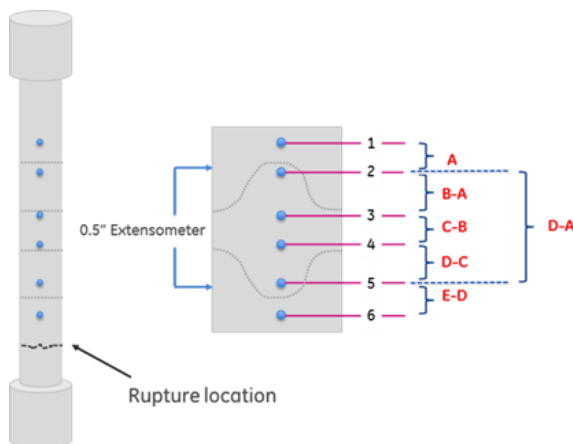


(c)

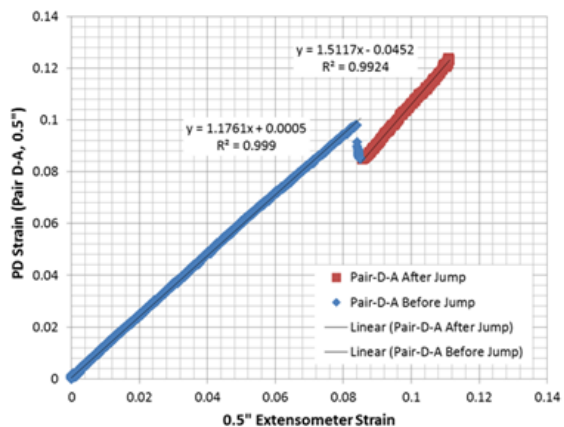
Figure 15: (a) Experimental creep strain vs. time curves for various pairs in welded (PWHT2) HA282 material, tested at 1500°F, 45 ksi (MT-05), (b) schematic illustration of pair locations, and location of creep rupture, and (c) plot between PD-based strain between pair D-A and 0.5" physical extensometer, which suggests that the former lags in strain within the 0.5 gauge.



(a)



(b)



(c)

Figure 16: (a) Experimental creep strain vs. time curves for various pairs in welded (PWHT2) HA282 material, tested at 1600°F, 25 ksi (MT-06), (b) schematic illustration of pair locations, and location of creep rupture, and (c) plot between PD-based strain between pair D-A and 0.5" physical extensometer, which suggests that the latter lags in strain within the 0.5 gauge.

2.2.4. In-situ local creep strain measurement using DIC technique

In addition to pair-wise potential drop (PD) based method, 2D/3D DIC based technique was also employed to measure localized strain within the welded specimens. A generic set-up for DIC testing is illustrated in a schematic in Figure 17. The DIC method uses an initial guess and finds the displacements by tracking unique individual subsets and determining the best correlation between a deformed image and an un-deformed reference image. For this purpose, a high contrast and non-uniform pattern is required on the sample surface. For example, speckle pattern on a test specimen (shown in Figure 17b), with a non-uniform distribution of gray scale, was obtained by painting the surface with a thin layer of white spray paint and then applying a mist of black spray paint to create the black speckles.

The challenges associated with the application of digital image correlation (DIC) technique at high temperatures include: image distortion due to convective currents, surface oxidation, inadequate lighting, degradation of speckles, etc. In the current work, a high-temperature automotive paint (rated for use up to 2000°F) was used for applying speckle patterns on the specimens for DIC measurements. Moreover, to overcome challenges of the application of high-temperature DIC to the creep test specimens, DIC was performed at room temperature to measure accumulated strain from a specimen that has undergone a creep test at high-temperature.

An exploratory DIC test, to measure creep strain within the gauge section of a specimen from HA282 base metal, was carried out. Applied stress levels of 14 and 16 ksi were chosen to achieve approximately 5-10% strain for about 45-50 hours (creep) hold at 1700°F (based on creep data shown in Figure 18a). The room temperature DIC measurements were performed by interrupting a creep test 3 times (see schematic in Figure 18b), and accumulated creep strains were measured (e.g., between each consecutive creep cycles, 0 and 1, 1 and 2, 2 and 3). Although the surface quality and image contrast of the spray-painted surface degraded over time when exposed at high-temperatures for long periods, full-field DIC measurements appeared to be unaffected at larger displacements. At smaller displacements, DIC measurements were affected by poor signal-to-noise response due to deteriorated speckle quality and loss of black-white contrast.

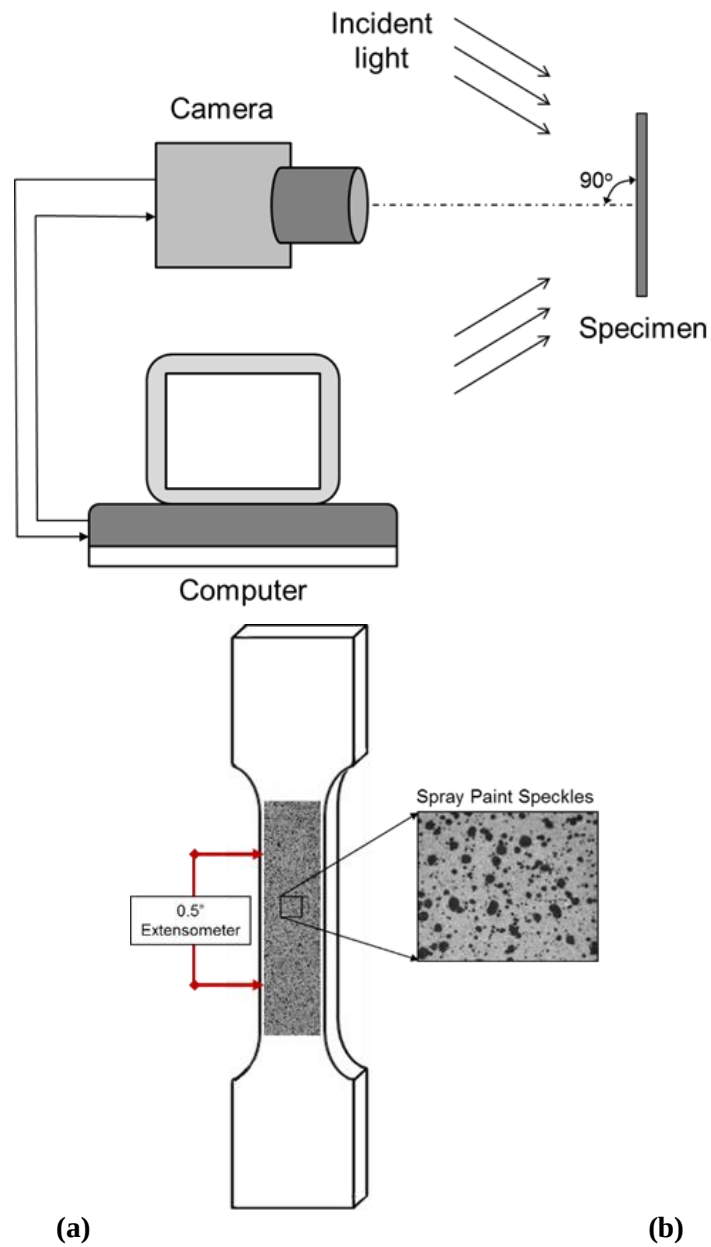


Figure 17: Schematic diagrams of in-situ load creep strain measurement method using a full-field digital image correlation (DIC) approach: (a) a generic 2D-DIC experimental set-up, and (b) speckle patterned specimen used for the DIC method.

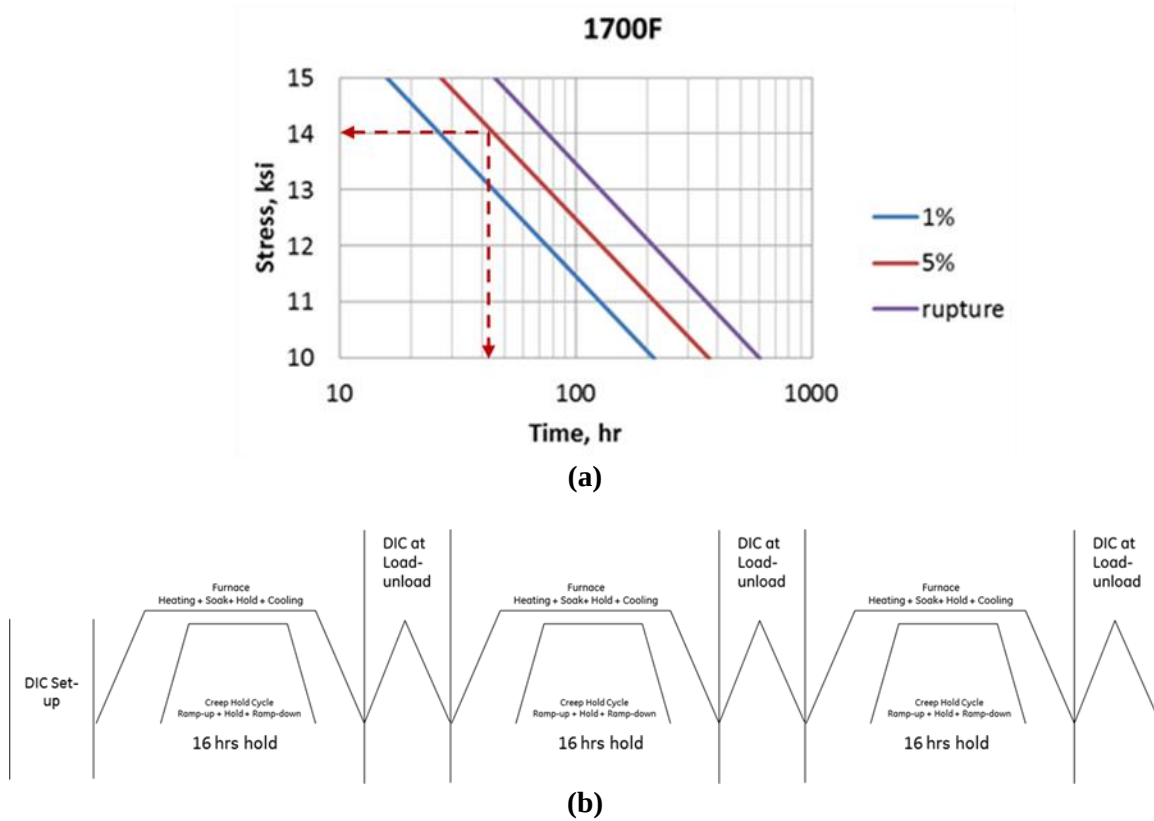


Figure 18: (a) Stress vs. time plot reduced from 1700°F creep data for HA282¹. Applied stress levels of 14 and 16 ksi were chosen to obtain approximately 5-10% strain for about 45-50 hours hold at 1700°F. (b) Schematic representation of creep strain measurements during intermittent creep loading cycles on a speckle patterned HA282 base metal specimen using the 2D-DIC method.

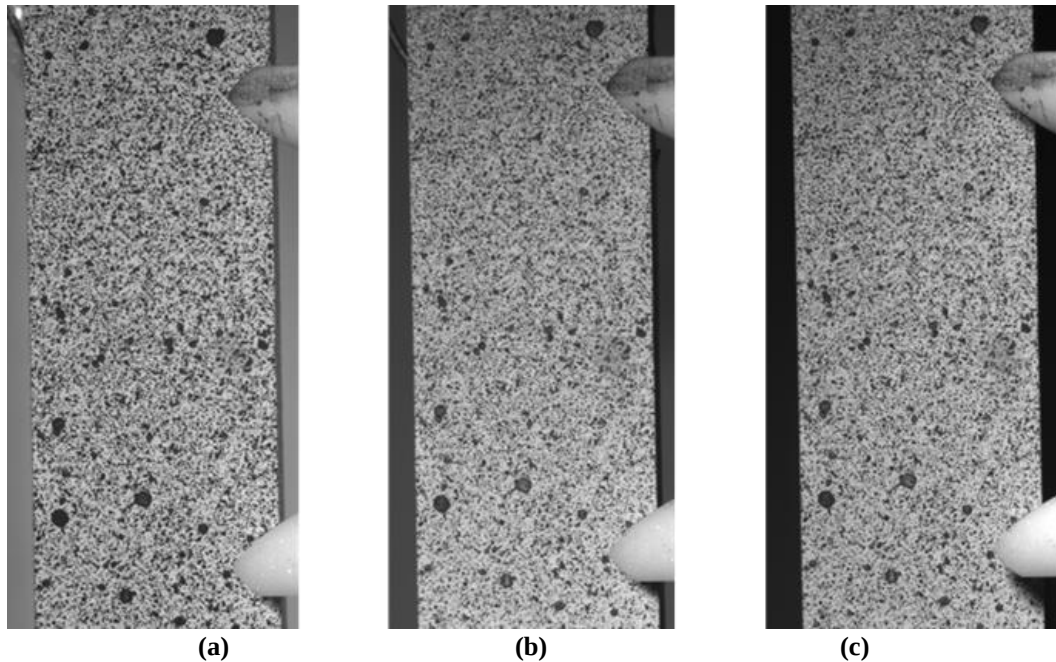
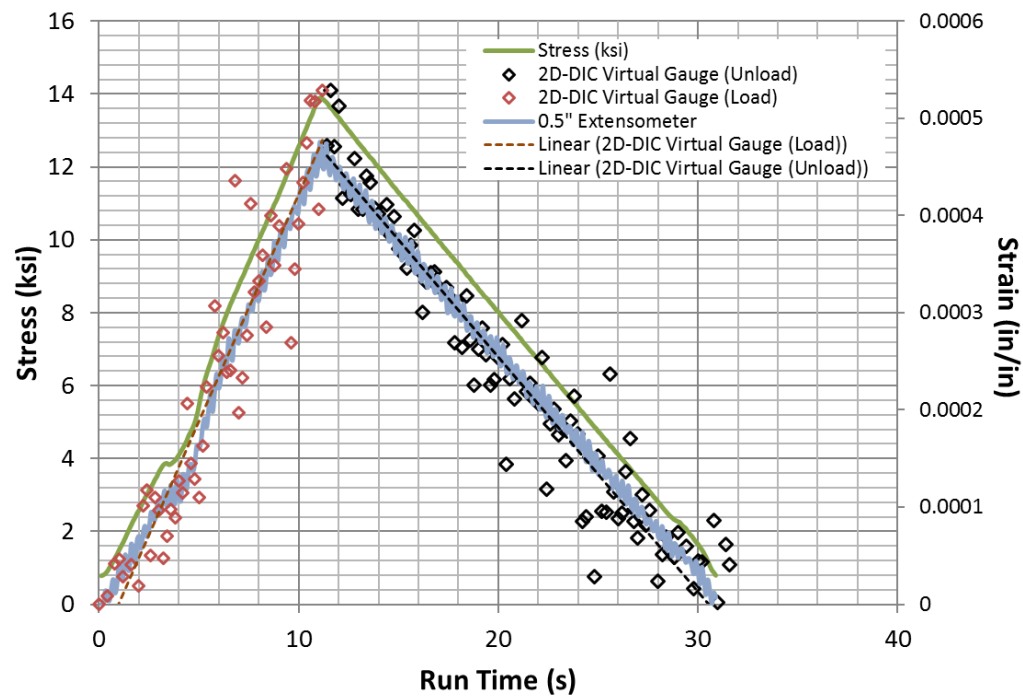
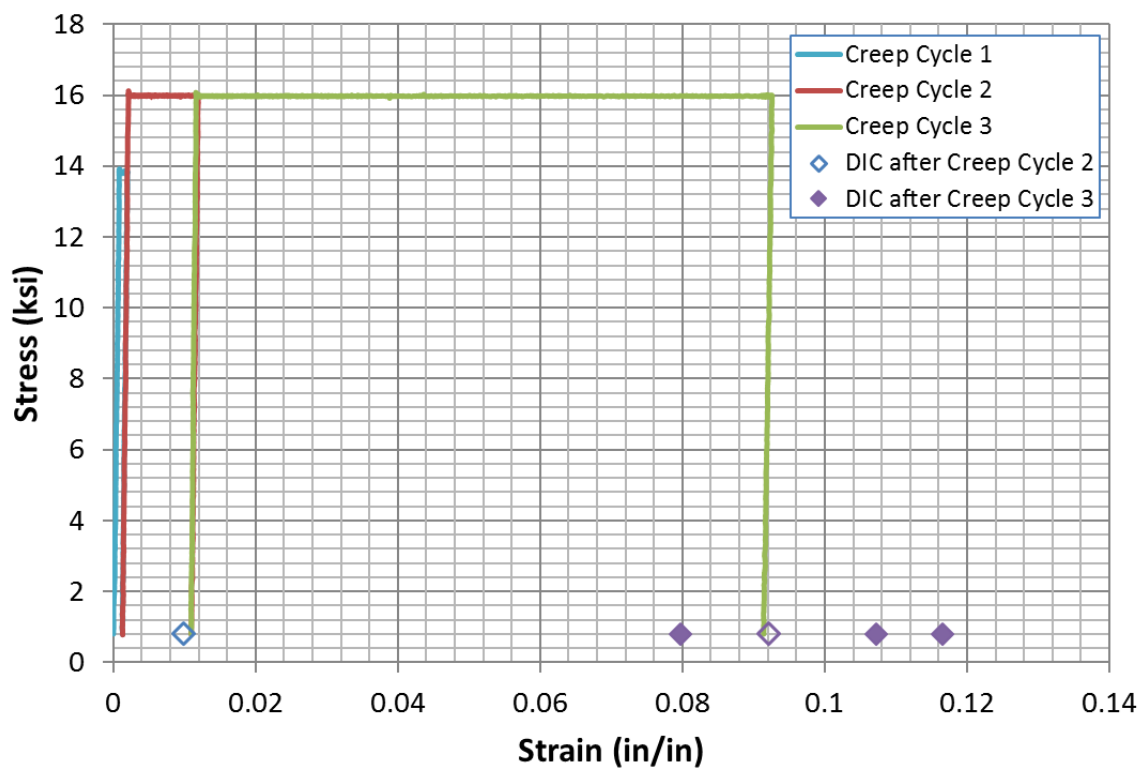


Figure 19: Series of images illustrate that the quality and image contrast of speckle painted surface continued to degrade over time under high temperature exposure: (a) pre-test room temperature image, (b) post first creep cycle (32 hrs exposure at 1700°F) room temperature image, and (c) post second creep cycle (total 48 hrs exposure at 1700°F) room temperature image.

The strain values measured from a 0.5" physical extensometer and a 2D-DIC virtual extensometer compared well (Figure 20a). At applied stress of 14 ksi, no significant creep strain was accumulated for the creep cycle 1 and therefore, stress was raised to 16 ksi for remainder of the creep test. For cycles 2 and 3, creep strains of ~1 and 8%, respectively, were achieved as measured by 0.5" extensometer (Figure 20b). These strain values were further confirmed by DIC measurements. Moreover, increase in creep strain rate can be seen on the strain vs. time plot, for cycles 2 and 3 (Figure 21). After 3rd creep cycle, an average strain of ~8% was measured by a 0.5" extensometer within the gauge section. Several virtual extensometers were drawn on the gauge section (using the 2D DIC software) that showed regions of varied accumulated creep strains (7% to 11%). Full-field 2D-DIC measurements also confirmed the presence of strain inhomogeneity within the gauge section at higher strains (see Figure 22).



(a)



(b)

Figure 20: (a) Experimental results from the first load-unload cycle (obtained after the first creep cycle at 1700°F) of HA282 base metal. A 0.5" extensometer was used to measure and verify strain

measurements (a 0.5" long 2D-DIC virtual extensometer was drawn on a speckled image to measure the strain) within the gauge section. (b) Stress vs. strain plot showed no significant creep strain was accumulated during the first creep cycle (~0.13%), followed by significant strain accumulation of ~1% (cycle 2) and ~8% (cycle 3).

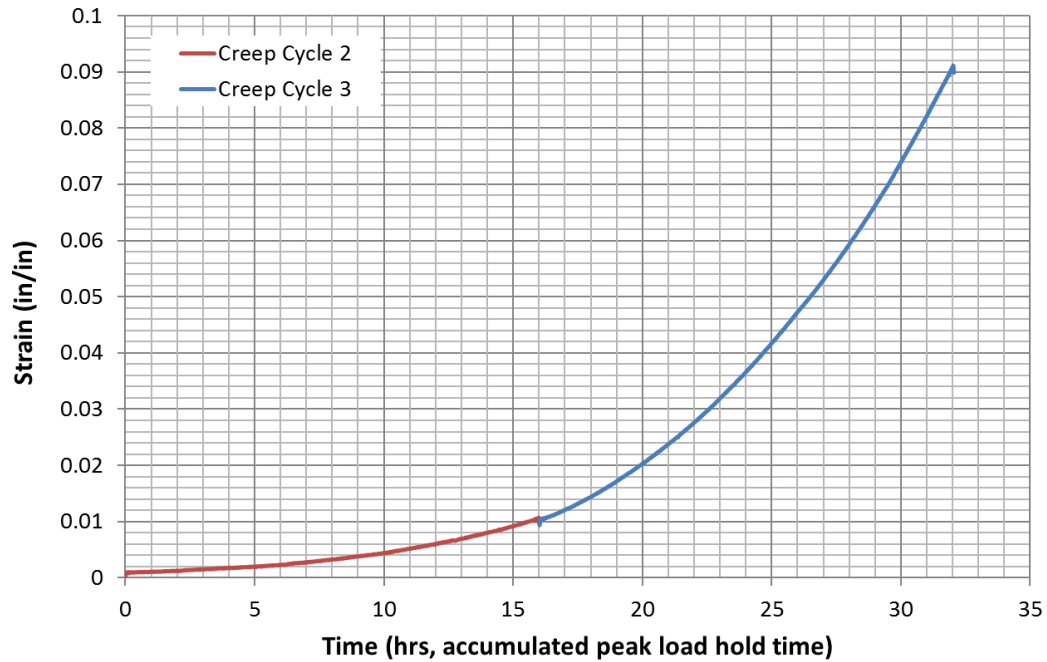


Figure 21: Accumulated creep strain vs. hold time plot from a HA282 base metal specimen, tested at 1700°F at 16 ksi for total 32 hours.

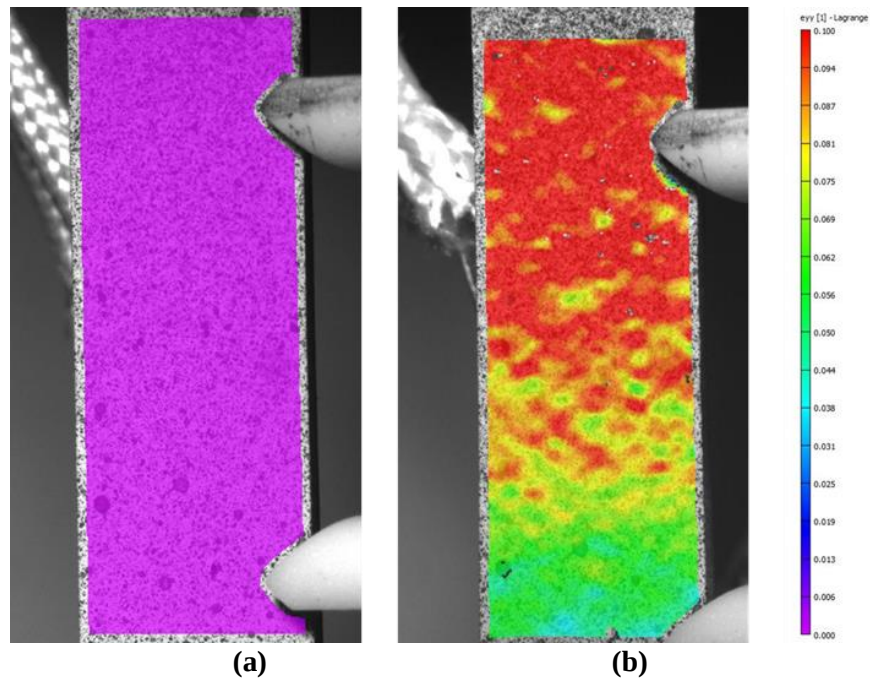


Figure 22: Full field longitudinal strain (ϵ_{yy}) maps: (a) taken after creep cycle 2 (16 ksi, 16 hours exposure at 1700°F), and (b) taken after creep cycle 3 (16 ksi, 16 hours exposure at 1700°F). Longitudinal strain (ϵ_{yy}) maps illustrate inhomogeneity in accumulated creep strain during cycle 3 (average 8% within 0.5" gauge section, but varies between 7 to 11% within the entire gauge section).

This exploratory test established the test procedure for Creep testing in conjunction with ex-situ DIC method, post-processing parameters, and feasibility of employing the technique for local creep strain in welded HA282 material. A more detailed experimental set-up for 2D and 3D DIC testing is shown below in Figure 23, which illustrates the orientation of cameras with respect the test specimen, extensometer location, furnace, etc. For the current DIC/Creep study, creep tests were performed on an MTS servo-hydraulic frame. Ex-situ 2D and 3D DIC were performed to measure accumulated strain at a room temperature from a specimen that has undergone a creep loading at high-temperature (as depicted in schematic in Figure 24). 2D DIC method was employed to capture more detailed information from the gauge section at a higher magnification, in combination with 3D DIC method which captured the entire specimen length. Between two consecutive creep cycles (e.g., between Cycle 1 and Cycle 2 in Figure 24), accumulated creep strain was measured by means of DIC method. Test conditions and relevant details are summarized in the Table 11.

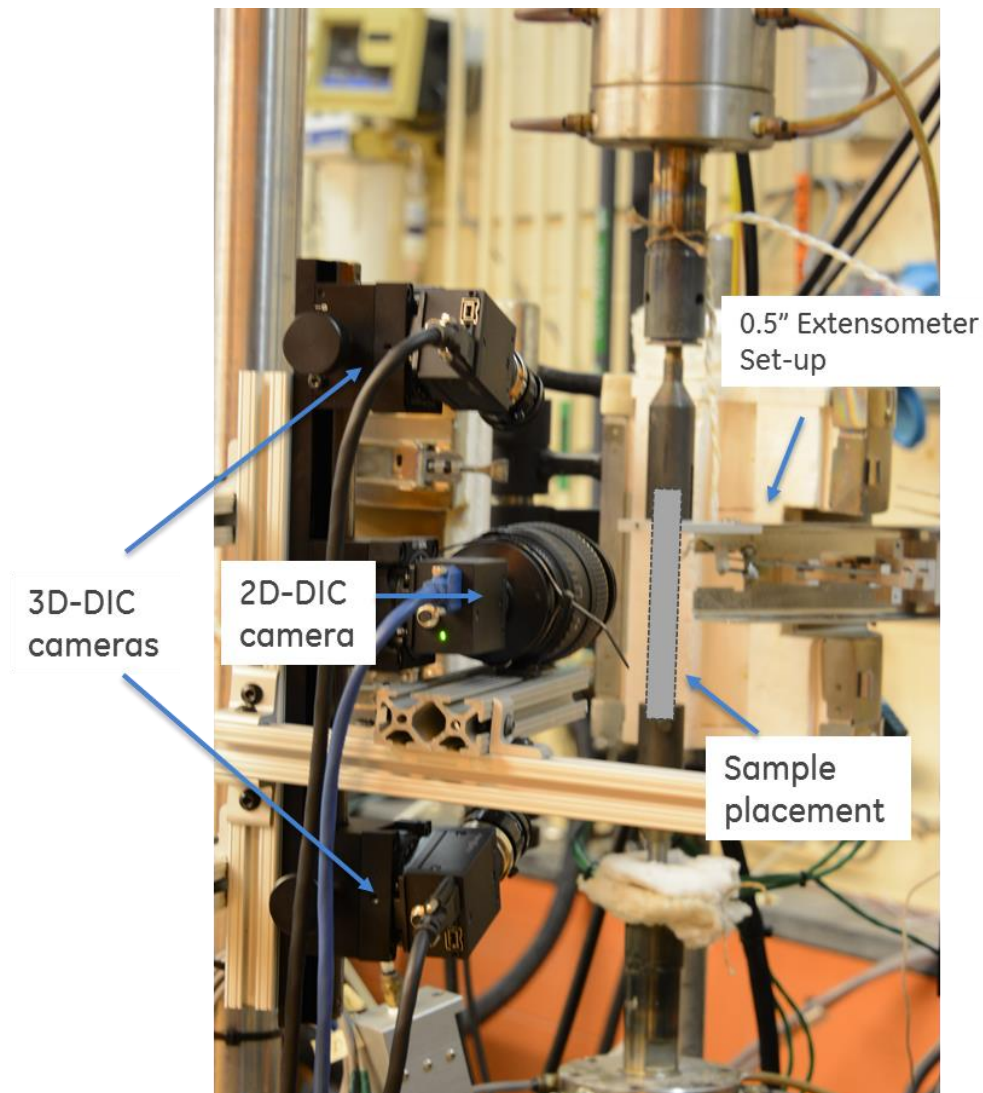


Figure 23: Experimental set-up for ex-situ local creep strain measurements using 2-D and 3-D full-field digital image correlation (DIC) methods.

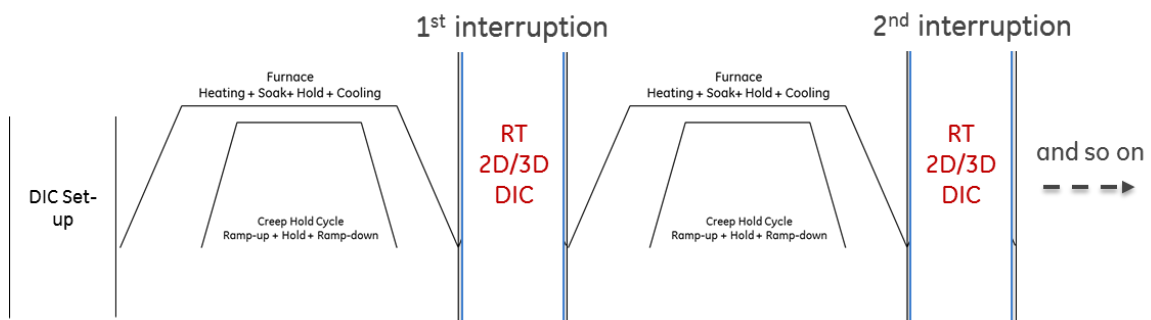


Figure 24: Schematic illustrates ex-situ creep strain measurement approach using DIC methods. 2D and 3D DIC measurements were performed at RT during each creep test interruption.

Table 11: Creep test interruptions for DIC measurements. Extensometer measured strain and crosshead displacement values at each interruption are reported.

Specimen ID	Stress (ksi)	Temperature (°F)	Status	Cycle #	Time (hrs)	Extensometer Strain (%)	Crosshead Displacement (in)
MT-08	45	1500		0	0	0	0
			Interrupted	1	22.9	0.63	0.0274
			Failed	2	32.6	-	-
MT-09	25	1600		0	0	0	0
			Interrupted	1	45.3	0.1025	0.0076
			Interrupted	2	50.9	0.142	0.0091
			Interrupted	3	56.2	0.1598	0.01146
			Failed	4	97.4	-	-
MT-10	15	1700		0	0	0	0
			Interrupted	1	18.1	0.083	0.0105
			Interrupted	2	41.8	0.988	0.0503
			Interrupted	3	47.2	1.509	0.074
			Failed	4	60.8	-	-
MT-18	45	1500		0	0	0	0
			Interrupted	1	43.2	0.3119	0.0136
			Interrupted	2	85.2	1.099	0.0685
			Failed	3	92.0	-	-
MT-20	15	1700		0	0	0	0
			Interrupted	1	22.37	0.522	0.0096
			Failed	2	38.02	-	-
MT-21	25	1600		0	0	0	0
			Interrupted	1	23.99	0.625	0.00699
			Interrupted	2	42.758	1.91	0.022
			Failed	3	59.484	-	-
MT-22	15	1700		0	0	0	0
			Interrupted	1	40.56	1.364	0.0106
			Failed	2	50.71	-	-

Although the same procedure was followed for preparing speckle patterned specimens as the exploratory test, first few specimens showed speckle paint cracking and spallation after exposure to high temperature. For example, Figure 25a shows surface of the specimen after 24 hrs exposure at 1600°F. DIC measurements could not be performed on such specimens due to deteriorated surface quality. Thereafter, a few modifications were made in spray paint speckling process which yielded some successful DIC measurements. Figure 25b illustrates a series of images taken at 0, 45 and 56 hrs exposure at 1600°F, which show that the speckle paint remained intact, however, quality and image contrast of speckle painted surface continued to degrade over time.

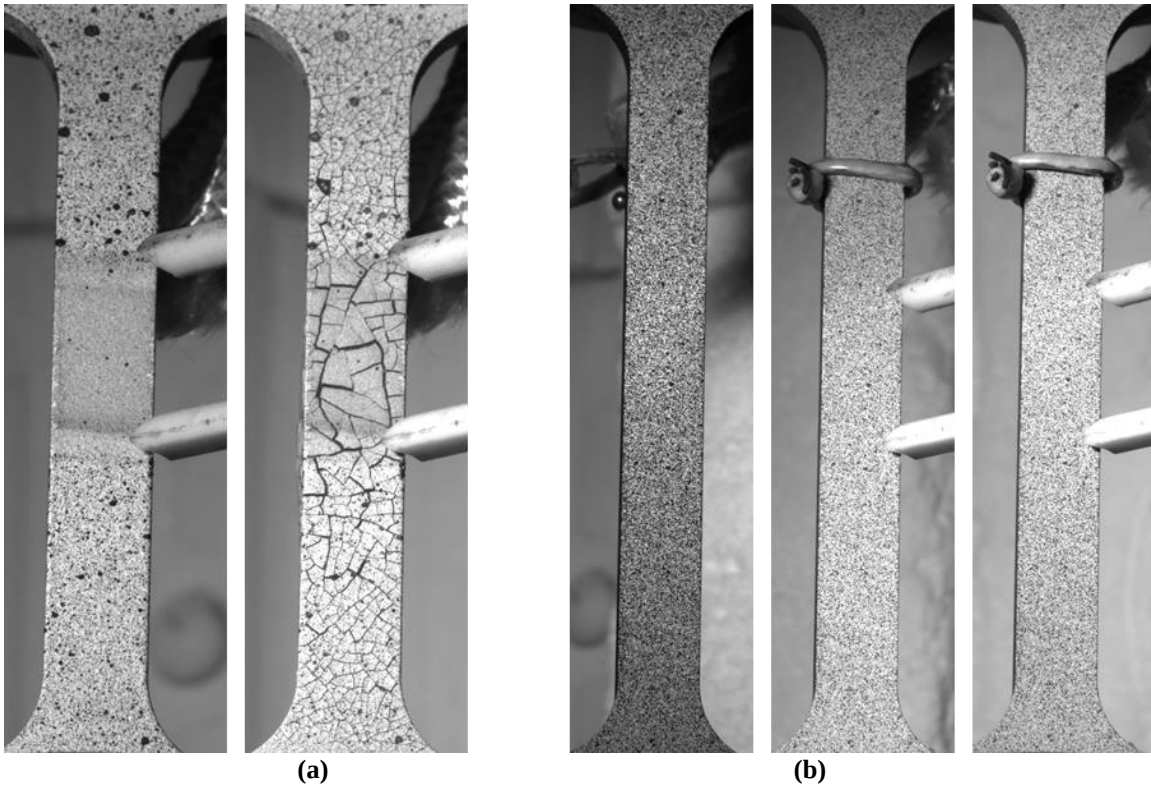


Figure 25: (a) First set of images shows speckle paint cracking and spallation after 24 hrs exposure at 1600°F. DIC measurements cannot be performed on a few test specimens due to inadequate sample preparation. (b) Second set of images (taken at 0, 45 and 56 hrs exposure at 1600°F) illustrate that the quality and image contrast of speckle painted surface continued to degrade over time under 1600°F exposure. A few modifications were made in spray paint speckling process which yielded some successful DIC measurements.

The DIC experiments were performed by utilizing commercial VIC-Snap-8® data acquisition (for capturing images), VIC-2D-2009® (for 2D DIC) and VIC-3D-7® (for 3D DIC) post-processing and analysis software (Correlated Solutions Inc., Columbia, SC). The 3D DIC system was calibrated only once (in the beginning of the test campaign) due to time constraints on the frame and technician's availability. Post-test campaign, DIC calibration was checked and no variations were observed that would have affected the quality of measurements. The calibration process involves the acquisition of a series of images (usually 25-30) of a calibration target in different orientations (translations/tilts around all three axes). From the images of the calibration target, calibration parameters for the two cameras as well as their relative orientation are determined automatically using VIC3D® software, which are used for test analysis.

Images for 3D DIC were acquired using two 5 mega-pixel PointGrey Grasshopper3 cameras, and the camera frame acquisition rate was set to 1 frame per sec. Total 20 set of images (2 images for 3D DIC, 1 image for 2D DIC per sec) were taken at RT from the specimen (held at 20 lbf load) during each creep test interruption. 2D and 3D DIC cameras were in sync and captured

images at the same time. The recorded image had dimensions of 2448 x 2048 pixels, corresponding to a resolution of 0.111 pixel/ μm for 2D DIC camera set-up and 0.026 pixel/ μm for 3D DIC camera set-up.

The first two images captured under 20 lbf load (referred as un-deformed condition) were used as the reference images for the full-field 3D DIC analysis, irrespective of the interruption cycle. During the post-test analysis, combinations of subset sizes and step sizes were evaluated for a characteristic spray-paint speckle pattern (e.g., shown in Figure 25b). Based on the preliminary analysis and observations of corresponding correlation error values, the maximum subset size of 23 x 23 pixels and a step size of 4 pixels were chosen for a representative 3D DIC image shown in Figure 25b. At each correlated point, the horizontal, Δx , vertical, Δy , and out of plane, Δz , displacements were calculated and differentiated to obtain Lagrange strains assuming a small strain approximation. Similarly, a subset size of 29 x 29 pixels and step size of 5 pixels were used for 2D DIC analysis. At each location, the horizontal, Δx , and vertical, Δy , displacements were calculated. Note that out-of-plane displacement, Δz , information was not obtained through 2D DIC method with a single camera. After post-processing 2D DIC images, we found out that the method didn't provide any additional insight in creep strain accumulations as expected, and therefore, 2D DIC results are not presented here.

It is important to note that the weld zone shape and size within the gauge (as well as front to back) varied significantly from sample-to-sample, depending upon the location within the V-joint welded plate from where the specimen was machined (Figure 26).

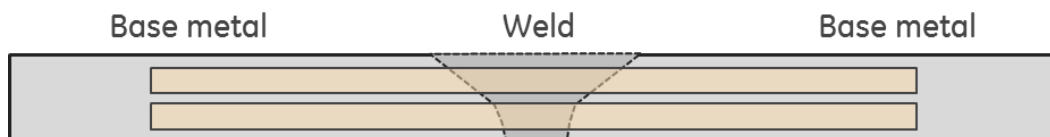


Figure 26: Schematic illustrates the apparent shape of weld zone within a flat creep test specimen. Specimen taken from the top contains wider weld zone as compared to the one taken from the bottom. Moreover, weld zone shape is not uniform through thickness within the specimen.

Figure 27a, Figure 28a and Figure 29a show 3D-DIC measured full-field strain (vertical strain, ϵ_{xx}) maps, as obtained from the straight section of flat creep specimens held under 20 lbf load at room temperature (RT) after “x” hrs of creep testing (x = creep hours at the final interruption, as indicated in plots). These full-field strain maps (Figure 28a and Figure 29a) illustrate heterogeneous distribution of strain within the straight gauge section consists of base-metal, HAZ and weld. Notably, in almost all cases creep strain first decreased and then increased from base-metal to transition (and/or HAZ-weld) zone, and decreased again within weld zone. Some of the tests were interrupted 2-3 times to capture the evolution of creep strain within different zones. Creep strain vs. time plots in

Figure 27b, Figure 28b and Figure 29b, show DIC measured strain within different zones of the straight gauge section at various creep loading times. Each point on the plot shows average of strain within a rectangular area in the full-field strain map. Different colors indicate measurements taken from different locations within the straight gauge section. As shown in these plots, average strain within each identified section show differences in accumulated creep strain over time.

Overall, these quantitative measurements suggest that there is relatively higher strain within the transition zone between the base metal and weld zone, and the width of transition zone is varied from specimen to specimen depending on where it machined from V-joint weld plate. Most of the specimens dominantly failed within base metal away from the middle section. Therefore, strain values taken from base metal regions appeared to be very high in the plots shown below. A few possible causes of dominant failure below radii were identified, such as machining defects, misalignment, temperature non-uniformity, but could not be confirmed. In the full-field strain maps (shown below), there are green-to-red colored islands near edges or at extensometer tips. These are artefact of DIC analysis due to missing correlations near edges or at locations speckle paint spalled off. Moreover, strain measured using physical extensometer, across 0.5" gauge compares well with DIC virtual extensometer measured strain within DIC measurement error (there appears to be a systematic difference of 0.0005-0.0007 in/in between DIC and physical extensometer). These local strain measurements using 3D DIC provided data for comparison with model predictions (discussed later in detail in Task-5).

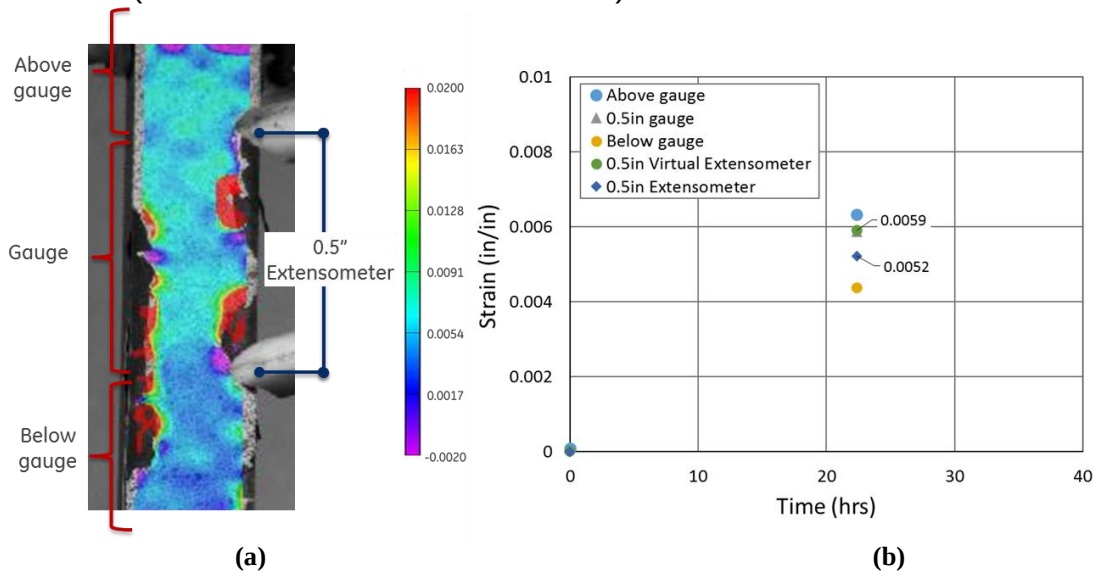


Figure 27: (a) Full-field strain (vertical, ϵ_{xx}) measured by 3D DIC in a flat creep specimen of 100% weldment HA282 material, tested at 15 ksi, 1700°F (MT-20), and (b) DIC virtual extensometer measured creep strain vs. time plot illustrating variability (or lack thereof) in creep strain in various sections of the specimen. Strain measured using physical extensometer, across 0.5" gauge compares well with DIC virtual extensometer measured strain within DIC measurement error. Specimen failed in 38.02 hrs.

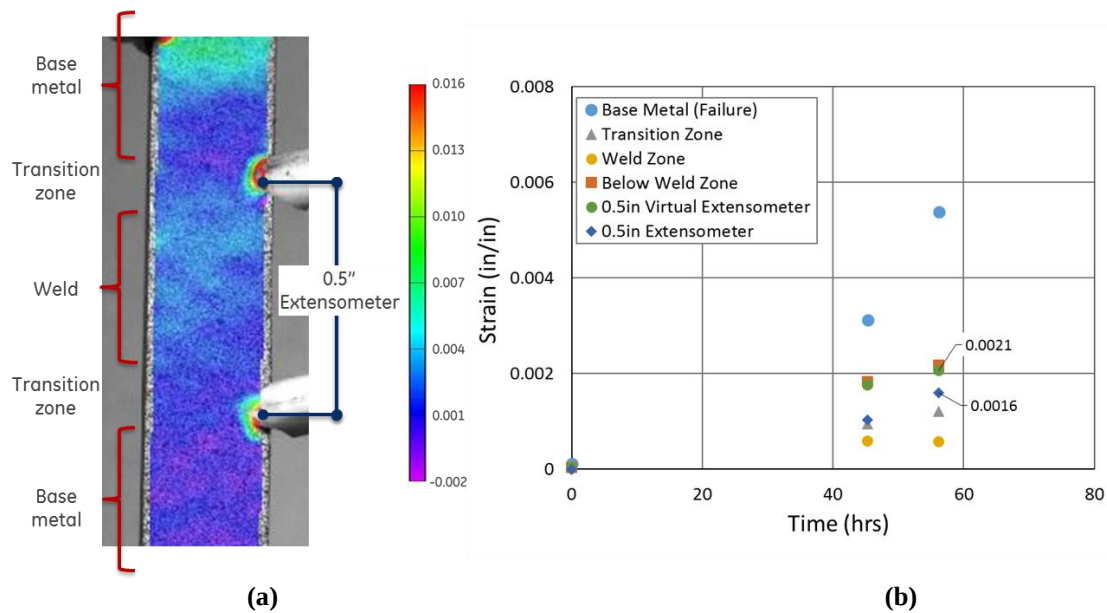


Figure 28: (a) Full-field strain (vertical, ϵ_{xx}) measured by 3D DIC in the flat creep specimens of welded (PWHT2) HA282 material, tested at 25 ksi, 1600°F (MT-09), and (b) DIC virtual extensometer measured creep strain vs. time illustrating variability (or lack thereof) in base metal, weld and HAZ-weld transition zones. Strain measured using 0.5" physical extensometer, across weld + HAZ-weld transition zone compares well with DIC virtual extensometer measured strain within DIC measurement error. Specimen failed in 97.36 hrs.

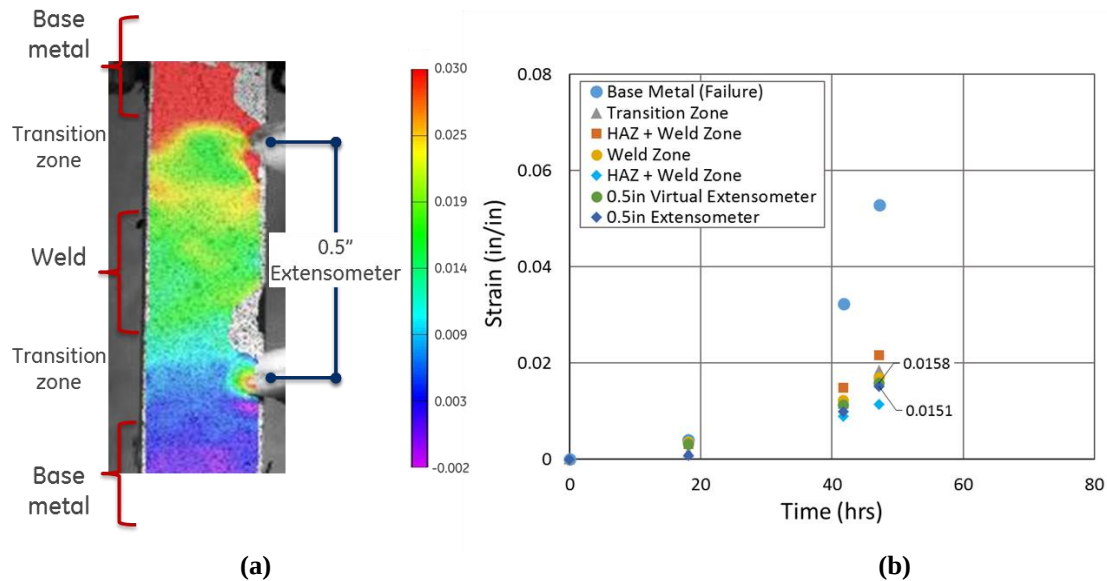


Figure 29: (a) Full-field strain (vertical, ϵ_{xx}) measured by 3D DIC in the flat creep specimens of welded (PWHT2) HA282 material, tested at 15 ksi, 1700°F (MT-10), and (b) DIC virtual extensometer measured creep strain vs. time plot illustrating variability (or lack thereof) in base metal, weld and HAZ-weld transition zones. Strain measured using 0.5" physical extensometer, across weld + HAZ-weld transition zone compares well with DIC virtual extensometer measured strain within DIC measurement error. Specimen failed in 60.78 hrs.

2.2.5. Summary of mechanical properties

All tensile and creep tests on welded and 100% weldment (PWHT2) HA282 materials, planned for Task 2, have been completed. The welded and 100% weldment HA282 materials, after standard post-weld heat treatment PWHT2 (2075°F/0.5hr+1850°F/2hr+1450°F/8hr), did not show significant debit in 0.2%Y.S. and UTS as compared to base metal at temperatures 1500°F and 1600°F. Notably, both 0.2%Y.S. and UTS for welded, and 100% weldment materials dropped significantly at 1700°F compared to base metal. Tensile ductility (% elongation) increased with the test temperature in all four types of materials. Moreover, ductility decreased from base metal to 100% weldment to welded HA282 materials, with as-welded being the least ductile among all test temperatures. The specimens after 5000hr furnace aging showed a noticeable reduction in 0.2%Y.S. and UTS.

Creep data did not show significant differences in creep properties from base-metal, welded and 100% weldment (PWHT2) HA282 materials. Rupture life was found very similar between welded HA282 and base metal. Some difference in creep strain and elongation at rupture between welded HA282 and 100% weldment was observed, but the difference is small compared to test data scatter and is difficult to be separated.

To investigate the possible cause for differences in ductility between 100% weldment, welded (PWHT2) specimens, and base metal, selected creep specimens of 100% weldment and welded HA282 were characterized by SEM, EBSD, and TEM to compare the difference of grain boundary microstructure, creep cavities, γ' coarsening, creep deformation mechanisms, and dislocation density in base metal vs. weld zone (discussed in detail in the following sections). Microstructure evaluation suggested that the reduced ductility in welded HA282 can be attributed to grain-boundary weakening (γ' denuded zone) and enhanced cavity growth in weld zone.

This work also provided first application of both pair-wise PD (volumetric) and ex-situ DIC (full-field surface) methods for measurements of creep strain across heterogeneous gauge section in welded material. Within the scope of this task, pair-wise PD-based measurements of local strain in a few creep specimens were also completed. Although the PD method showed heterogeneity in strain in various regions such as base-metal and weld zone, measurements were found to be highly influenced by micro-crack/cavity formation during early-to-late stages of creep deformation and the location of necking. Additionally, ex-situ DIC tests for full-field measurements of local creep strain were also completed. Strain measured across the 0.5" gauge section of the sample using 2D/3D DIC method agreed with extensometer data within reasonable measurement errors. DIC results, albeit limited, suggested that there are subtle differences in accumulated creep strain from base metal to weld-zone, with relatively higher accumulated

creep strain in base-metal to weld-zone transition regions. These local strain measurements provided data for comparison with model predictions.

2.3. Microstructure characterization

The tensile and creep tests suggested that after the standard post-weld heat treatment PWHT2 the welded HA282 had a similar bulk strength (tensile and creep) to the base metal, but a noticeable reduction in ductility. Detailed microstructure characterizations before and after tensile or creep deformation were performed using optical microscopy (OM), scanning electron microscopy (SEM), electron backscatter diffraction (EBSD), electron-probe microanalysis (EPMA), and transmission electron microscopy (TEM). The primary focus was on the differences between weldment and base metal, in order to understand the major factors that differentiated the mechanical properties in the welded structure. The objectives included (1) compare welded microstructure with base metal, (2) study the microstructure inhomogeneity introduced by various post-weld heat treatment, (3) determine the tensile and creep failure locations and mechanisms, (4) clarify the microstructure evolution during static exposure and creep deformation, (5) identify difference in base metal and weld zone that leads to creep properties. Finally, since our initial study indicated clearly an intergranular failure mode within weld zone of HA282, substantial amount of effort was made on comparison of creep damage at grain boundaries in weld zone versus base metal. The results provided inputs for microstructure model (Task 3), mesoscale deformation model (Task 4), and constitutive creep model (Task 5), as well as an improved understanding of relationship between microstructure and creep behavior for welded HA282. Table 12 summarizes major microstructural characteristics prior to mechanical testing.

Table 12: Summary of microstructure before mechanical testing

Condition	Grain Shape	Texture	Micro Segregation	γ'	Secondary Phases
Base Metal	Equiaxial	None	None	Spherical γ' with ~100um diameter; intragranular and intergranular	Intragranular carbides: Ti, Mo-rich MC, Cr-rich $M_{23}C_6$ Mo-rich M_6C
As-Welded	Columnar	<001> fiber texture along plate normal direction	Strong micro-segregation across dendrites	γ' not observed by high-resolution SEM	Fine inter-dendritic MC and M_6C , also at high angle boundaries
Welded + PWHT	Columnar; some recrystallization	Same as As-weld	Decreased in the order of PWHT3a, PWHT3b, PWHT1, PWHT2	After PWHT2, homogeneous distribution of γ' in weld zone; similar to base metal	After PWHT2, similar to base metal

2.3.1. Initial microstructures

HA282 Base metal

Figure 30 shows microstructure of HA282 base metal after full heat treatment cycle specified in Ref¹:

Solution	2075°F (1135°C) / 30min → Air cool
Primary age	1850°F (1010°C) / 2h → Air cool
Secondary age	1450°F (788°C) / 8h → Air cool

Equiaxial grain structure with average grain size of ~140μm is seen in base metal. MC carbide particles, which are enriched with Ti and Mo, showed distribution along the rolling direction (normal to the plate thickness) (Figure 30a), and the size is up to 20μm (Figure 30b). M₂₃C₆ and M₆C carbides were observed in both grain interiors and grain boundaries. EDS analysis showed M₆C was rich in Mo, and M₂₃C₆ was rich in Cr and Mo. Fine, spherical γ' precipitates were observed in the grain interiors (Figure 30d).

Grain boundary carbides were identified by combining Energy Dispersive Spectroscopy (EDS) and Electron Backscatter Diffraction (EBSD) techniques. Figure 31 shows the back scattered electron (BSE) and secondary electron (SE) images taken from the same location of a specimen, together with EDS spot spectra on the grain boundary phases. The phase with bright contrast in the BSE image (red circle in Figure 31) is rich in Mo and carbon, while the phase with gray contrast in the BSE image (green circle in Figure 31) is rich in Cr, Mo, and carbon. EBSD Kikuchi patterns collected on the two phases were analyzed and checked against possible phases including carbides (MC, M₂C, M₇C₃, M₂₃C₆, M₆C, M₁₂C), borides (MB₂, MB, M₅B₃, M₂B), P phase, Sigma phase, and Laves phase. The EBSD patterns matched well with M₆C for the Mo rich carbide and M₂₃C₆ for the Cr rich carbide.

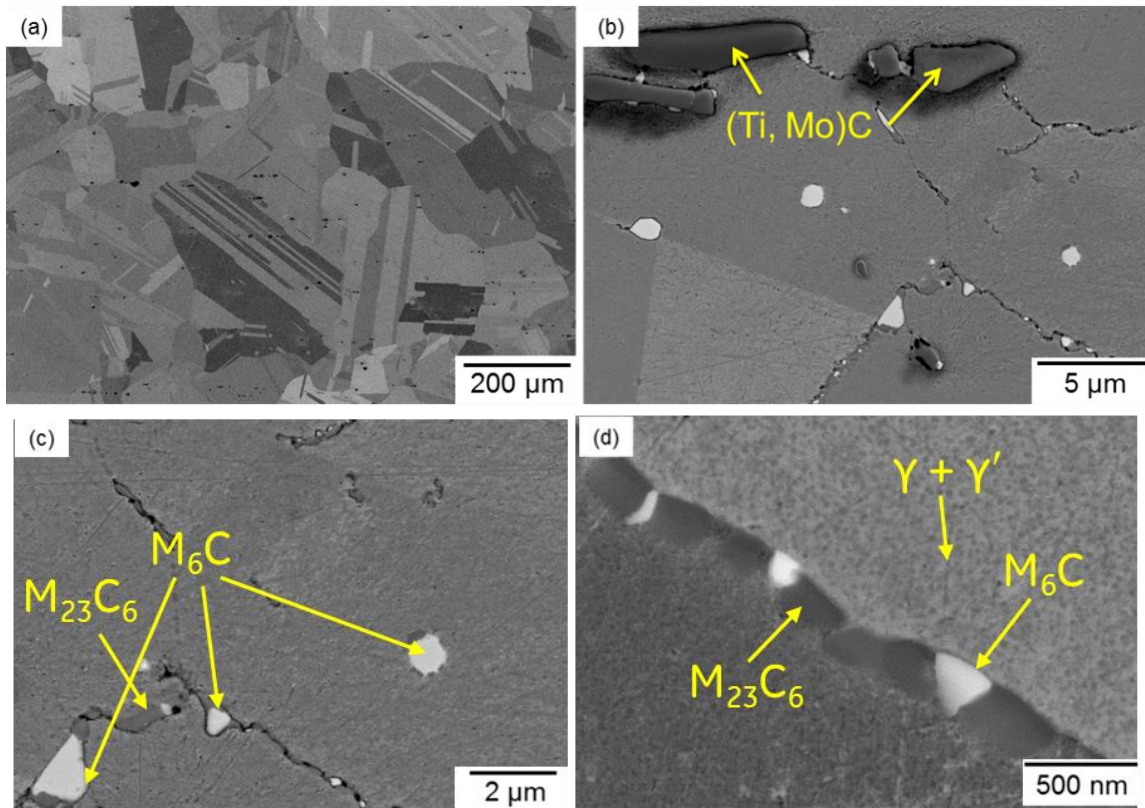


Figure 30: Backscattered electron images of fully heat treated HA282 base metal.

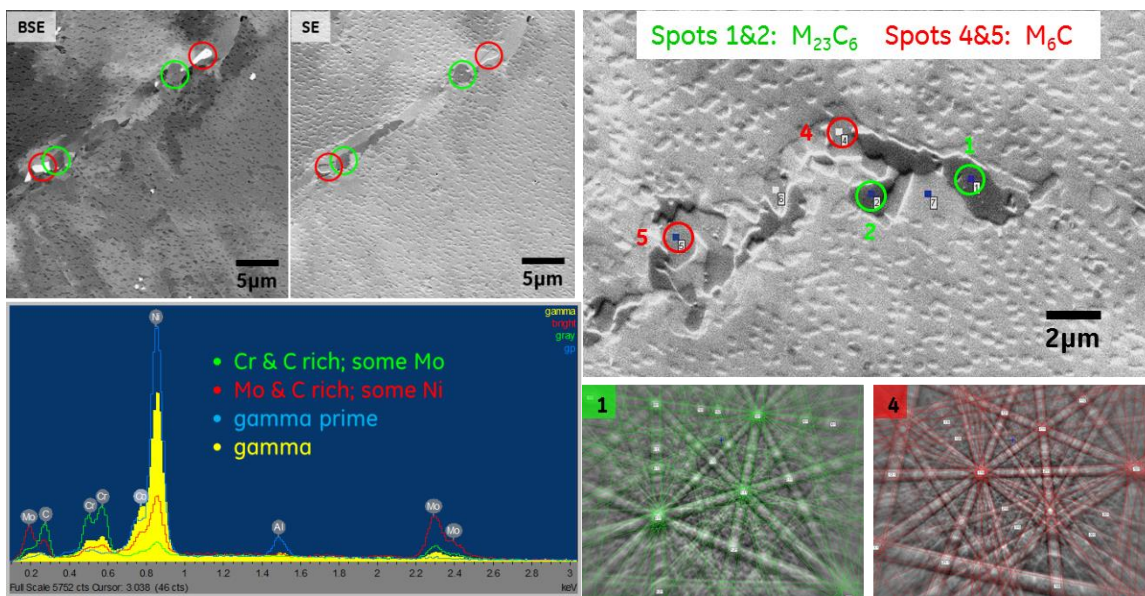


Figure 31: Grain boundary phases are identified as Mo rich M_6C carbides and Cr, Mo rich $M_{23}C_6$ carbides by EDS spectra and EBSD patterns.

Welded HA282

Figure 32 shows a low-magnification image of a cross section of welded HA282, which clearly shows weld, HAZ, and base metal regions. Figure 33 shows optical micrographs taken from the weld, HAZ, and base metal. The weld region showed fine dendritic structure with primary dendrite arm spacing about 20–30 μm (Figure 33a). Both HAZ and base metal exhibited alternating structure consisting of a region with relatively large grains and a region with finer grains associated with the band containing high density of carbide particles (Figure 33b and c). The region with finer grains in HAZ (Figure 33b) showed larger grain size than the base metal (Figure 33c), possibly due to the grain coarsening during welding.

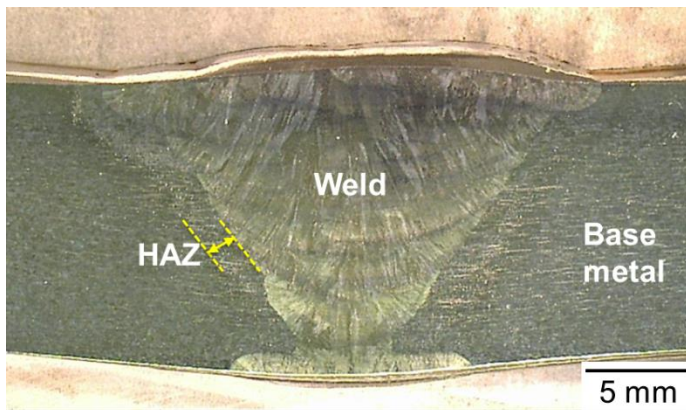


Figure 32: Macroscopic image of a cross section of TIG welded HA282.

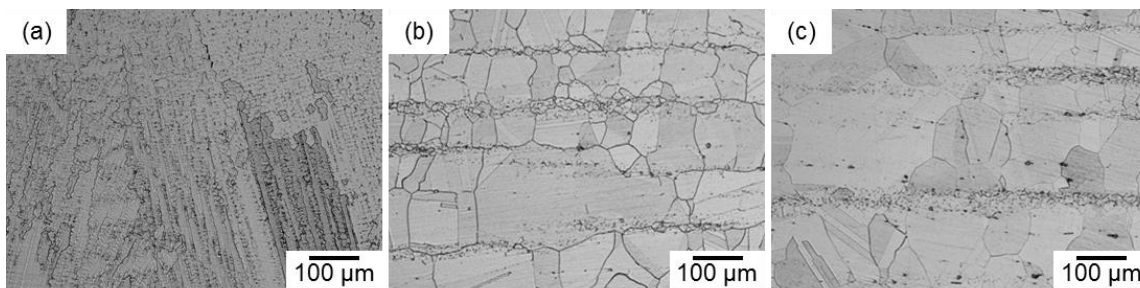


Figure 33: Optical images of welded HA282: (a) weld, (b) heat affected zone and (c) base metal.

The backscattered electron image showed that the interdendritic region exhibited presence of fine particles with dark and bright contrast (Figure 34(a)). EDS analyses indicated the bright contrast particles were possibly M_6C carbide, and the dark contrast particles were MC phase in the forms of particle (Figure 34(b,c)). Due to fast cooling rate from welding process, these particles are much smaller compared to those found in base metal after full heat treatment.

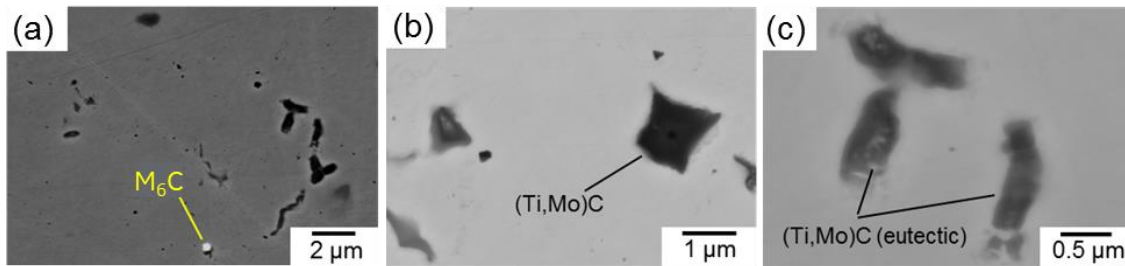


Figure 34: Microstructure of the welded HA282: backscattered electron (BSE) image of the weld region highlighting particles observed in the interdendritic region.

Grain texture in base metal and weld zone was characterized by EBSD. In Figure 35, base metal is shown in a random texture while weld zone has $\langle 100 \rangle$ fiber texture along sample y axis, which is the growth direction of the primary dendrite arms ($\langle 100 \rangle$ orientation) during solidification. In the transverse (x-z) section, the grains in the weld zone are shown in a random texture. Both the columnar grain shape and texture in weld zone may cause higher grain boundary incompatibility to accommodate deformation strain and grain boundary damage.

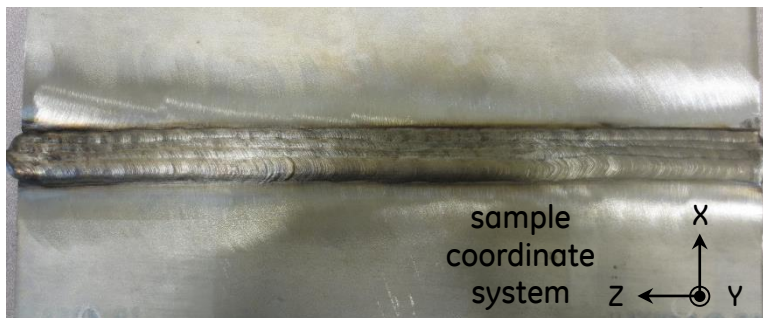
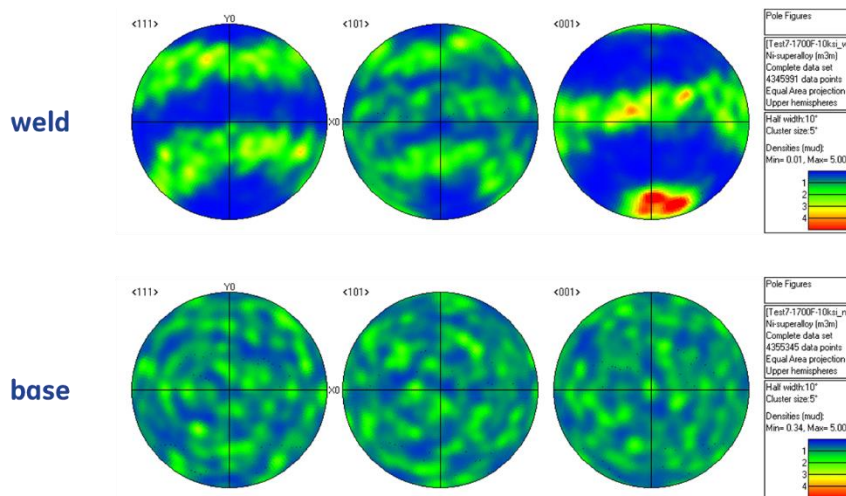


Figure 35: Pole figures for base metal and weld zone in welded HA282 with PWHT2 treatment. Samples x direction is cross weld direction or tensile/creep specimen loading direction, z direction is welding direction, and y direction is the V-groove direction (or grain growth direction).

As-weld microsegregation

Electron-probe microanalysis (EPMA) was applied to measure dendritic composition segregations (microsegregations) in the weld zone of HA282. All elements except for the interstitial C and B in the HA282 composition (Table 3) were measured. Figure 36 shows, for example, a quantitative EPMA map of element Ti taken from the middle of the weld zone. The color contrast shows variation of compositions in different interdendritic regions. Ti shows a strong partition to the interdendritic regions.

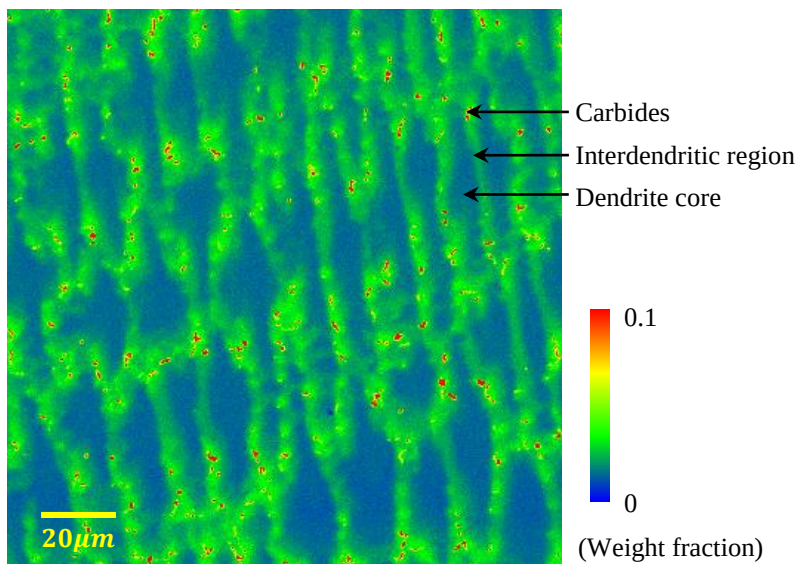


Figure 36: EPMA map of Ti element in as-welded HA282.

To quantify microsegregation for each element, line segments were manually selected on the EPMA map. As shown in Figure 37, each line segment was made perpendicularly across a dendrite core (from interdendritic to interdendritic). All segments avoided interdendritic particles. The composition was then extracted along the line segment. There were totally 150 line segments made in both cases in Figure 37.

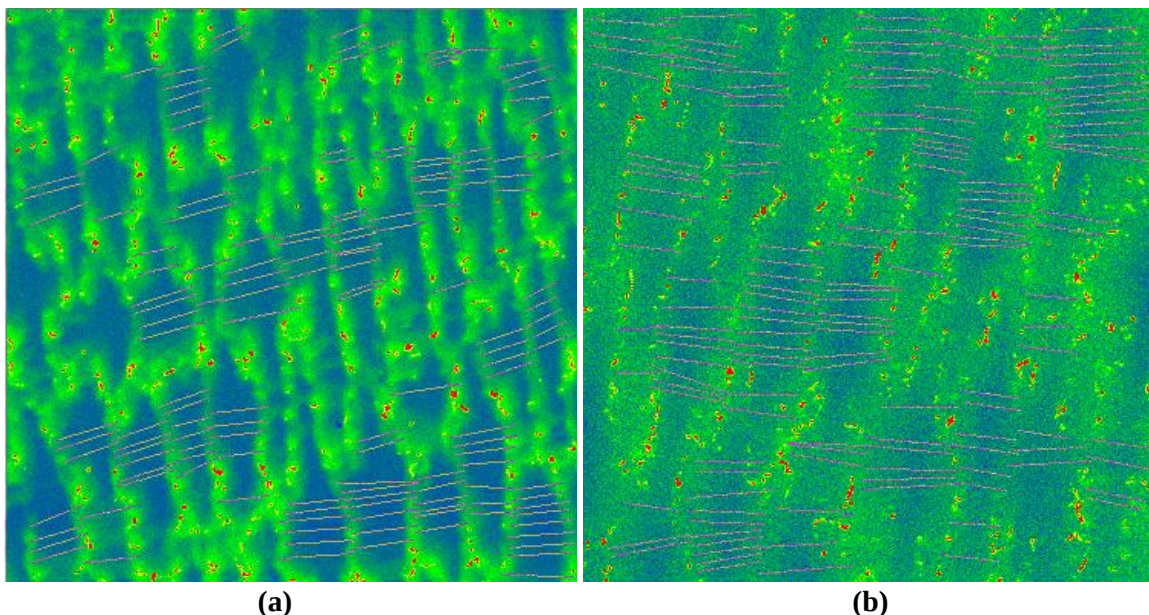


Figure 37: Manually selected line segments across dendrite core for microsegregation analysis. (a) As-welded, (b) Heat treated. Interdendritic phases were avoided. Ti maps are shown.

Composition profiles along these line segments are shown in Figure 38(a) for Ti. Since these line segments pass different dendrite core size, the distance of each segment is normalized to $[-1,1]$, where 0 corresponds to dendrite core, -1 and 1 to interdendritic region. Average and standard deviation are then plotted in Figure 38(b). It is clearly seen that Ti segregates to the interdendritic regions. Figure 39 summarizes the microsegregation profiles in the as-welded HA282 for all elements, including the trace elements Mn, Si, and Fe. With standard derivation, it can be seen that elements Cr, Mn, Si, and Fe have very little microsegregation in the as-welded HA282. While Al is on the border line, it shows slight segregation to interdendritic regions, as opposed to the Scheil calculation with TTNi7 Ni database² (see Section 3.1). To compare more clearly the degree of microsegregation among all elements, we normalize the composition profile of each element by its average composition. Figure 40 shows a descending order: $\text{Ti} > \text{Mo} > \text{Co} > \text{Ni}$, $\text{Al} > \text{Cr}$, with Co and Ni segregation to dendrite core, Ti, Mo, and Al to interdendritic region.

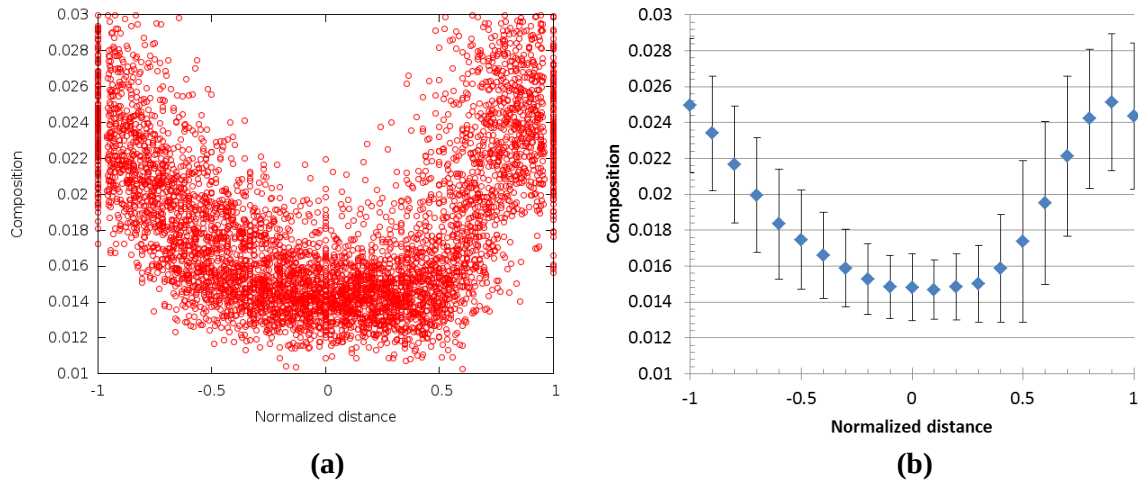


Figure 38: Ti composition (in weight fraction) across dendrite core in as-welded HA282. (a) Overlay of the 150 line segments, and (b) the average profile and standard deviation. The distance is normalized (0 is dendrite core, -1 and 1 are in interdendritic region).

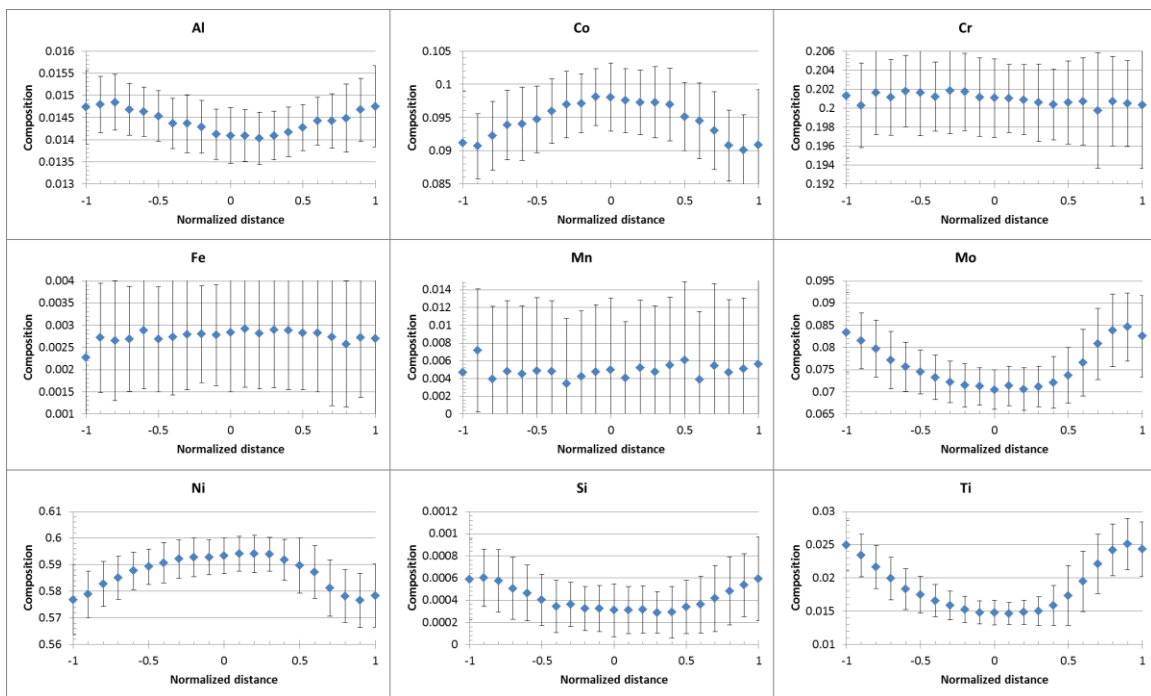


Figure 39: Composition profiles (in weight fraction) in as-welded HA282. The distance is normalized (0 is dendrite core, -1 and 1 are in interdendritic region).

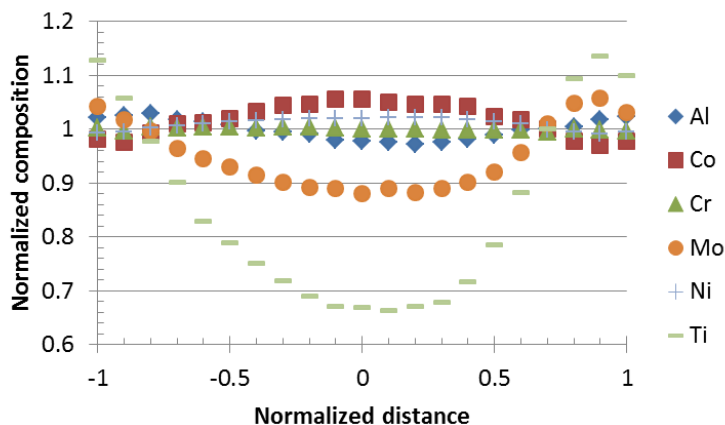


Figure 40: A comparison of microsegregation of each element after normalizing each composition by the respective average composition. (As-weld HA282, the normalized distance 0 is dendrite core, -1 and 1 are in interdendritic region)

2.3.2 Microstructures after PWHTs

Welded HA282

Figure 41 shows the SEM backscattered electron (BSE) images of the welded HA282 after the PWHT. The base metal and HAZ regions exhibited equiaxed grains (Figure 41a and c). Grain boundaries were decorated with Cr-rich $M_{23}C_6$ carbides (dark contrast) and Mo-rich M_6C carbides (bright contrast) (Figure 41b and d). Large Ti, Mo-rich MC carbides were observed throughout the specimen. In the grain interiors, M_6C carbides were observed together with smaller $M_{23}C_6$ carbides next to them (Figure 41d). The weld region exhibited a columnar grain structure (Figure 41e). The (Ti,Mo)C carbides were observed in the former interdendritic regions. Along the grain boundaries, both $M_{23}C_6$ and M_6C carbides were present (Figure 41f).

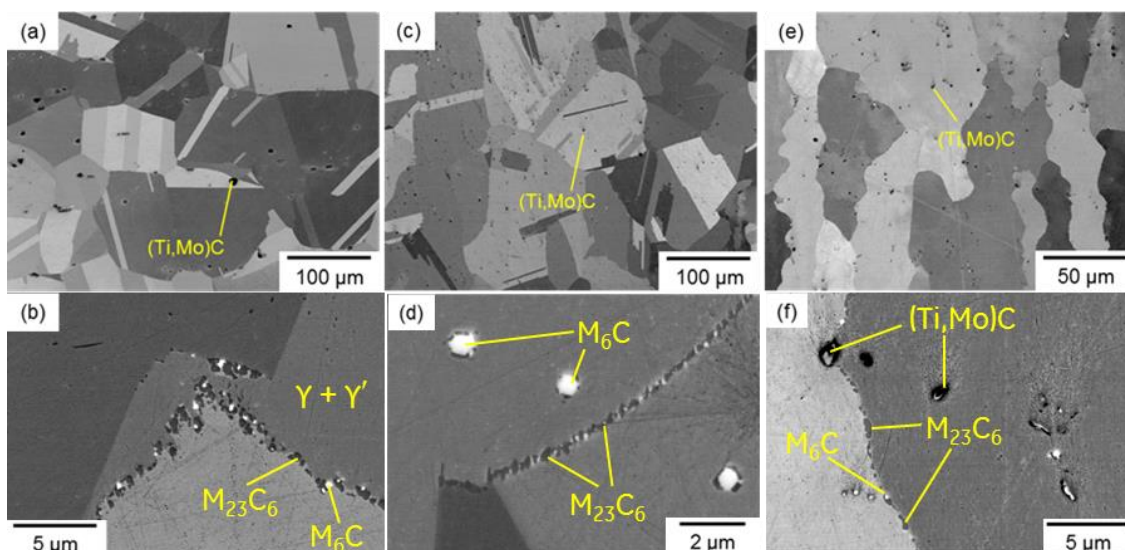


Figure 41: Microstructure of welded HA282 after the PWHT: (a-b) base metal, (c-d) HAZ, (e-f) weld.

HA282 weldment

Figure 42 shows the SEM backscattered electron (BSE) and secondary electron (SE) images of the 100% weldment after the stress-relief (solution anneal) heat treatment. In comparison to the welded HA282 microstructure after PWHT2, similar columnar grain structure containing a few prior dendrites and (Ti,Mo)C carbide particles decorating the prior interdendritic regions were observed in the weld zone (Figure 42a, b, c). Images taken at multiple locations suggest no location-specific microstructure variation. Solidification voids from welding process were rarely seen. One instance is shown in Figure 42d and e. Overall the weldment material is homogeneous and similar to the microstructure in the welded HA282.

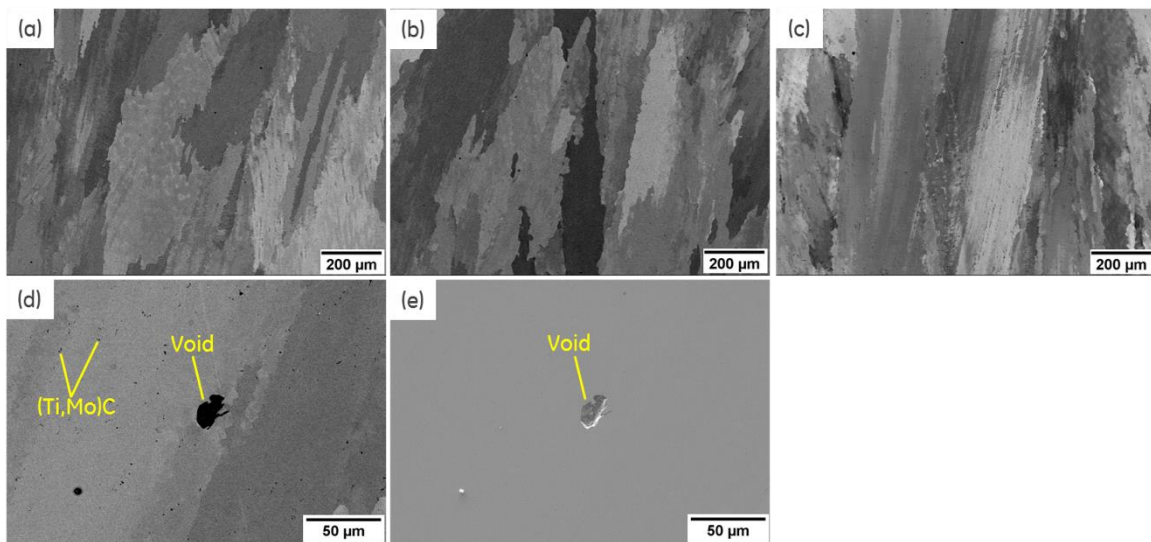


Figure 42: Microstructure of (a) longitudinal and (b) transverse cross section of the 100% weldment after stress relief (solution treatment), compared to (c) welded material after PWHT. (d) and (e) show BSE and SE images that highlight a solidification void and (Ti,Mo)C dark particles distributed at the prior interdendritic regions.

Alternate PWHTs

Two post-weld heat treatment conditions without the solution anneal step, PWHT3a and PWHT3b (see Table 4), were applied to as-welded HA282 material to introduce inhomogeneity in the weld zone. The purpose was to study the effect of homogenization on microstructure and creep properties of welded HA282. Unlike in the PWHT2 that included a 2075°F solution anneal, coarse γ' particles about 100~200nm in size were seen in some interdendritic regions in the weld zone (Figure 110). Since the nominal γ' solvus temperature of alloy HA282 was reported as 1827°F³, which is below the aging temperature 1850°F, the coarse γ' observed in the inter-dendritic region after PWHT3a indicates that local γ' solvus temperature in the interdendritic region may be higher than 1850°F. Due to Al

and Ti segregation at inter-dendritic region (Figure 39), the γ' solvus temperature and equilibrium fraction are expected to be higher than that in the dendritic core region. Therefore, the prior interdendritic γ' particles can remain undissolved during the 1850°F/1hr step in PWHT3a. On the other hand, a close examination of the fine γ' in dendritic core and interdendritic regions showed similar size of the precipitates. More discussions on the results of microstructure model in comparison with experimental data are given in Section 3 of this report.

Residual microsegregations after PWHT

The standard post-weld heat treat PWHT2 (Table 4) yielded a very homogeneous composition distribution. Figure 43 shows composition the EPMA maps taken at a location near the center of weld zone. It is clearly seen that even the slowest diffusion element, Mo, has become very uniform in the grain interiors.

Composition variations are seen mostly as grain boundary segregation (Cr, Mo enrichment, Al, Ni, Co depletion), as intra-granular (Ti, Mo-rich) MC carbides, and as grain boundary Ni, Co, Mo-rich phase and Cr-rich phase particles.

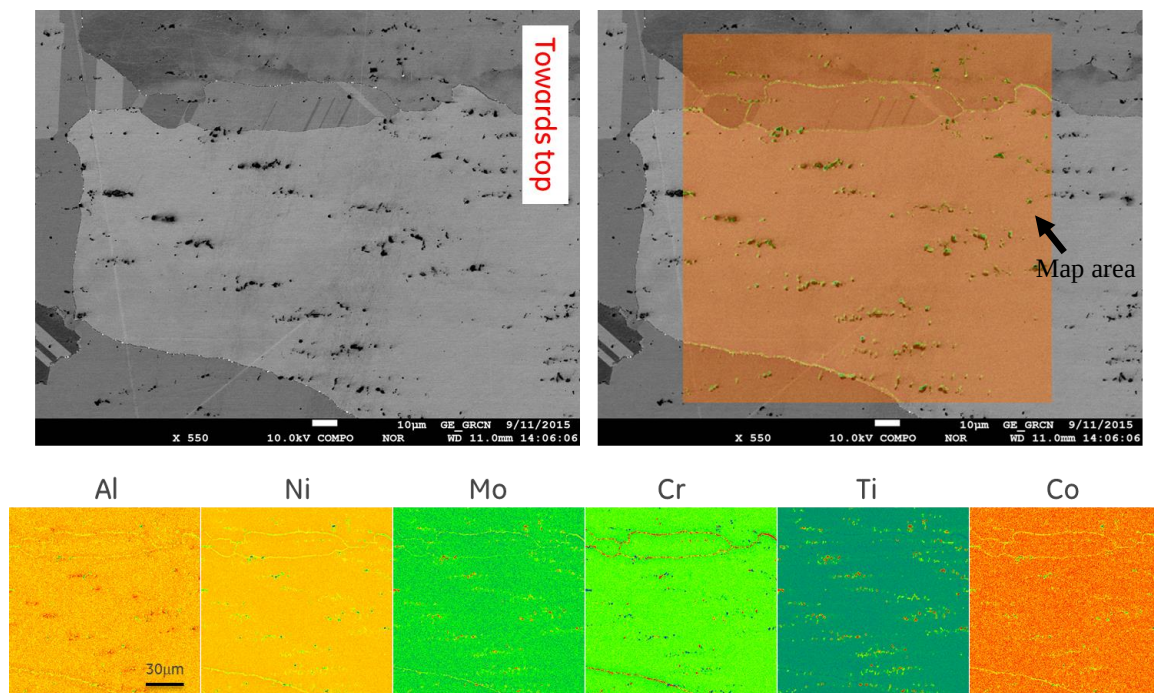


Figure 43: EPMA analysis of HA282 after PWHT2. The specimen was taken near the center of weld zone. Composition maps are shown at the bottom.

The alternate condition suggested in the Haynes 282 Brochure¹ that does not include the 2075°F solution step (PWHT3b in Table 4) produced a much less homogeneous composition distribution in the material. Figure 44 compares the EPMA maps after PWHT3b, as well as PWHT3a (less homogenization time). Compared with PWHT2, these heat treatments resulted in noticeable residual microsegregation of Mo, Ti, Al, and Co.

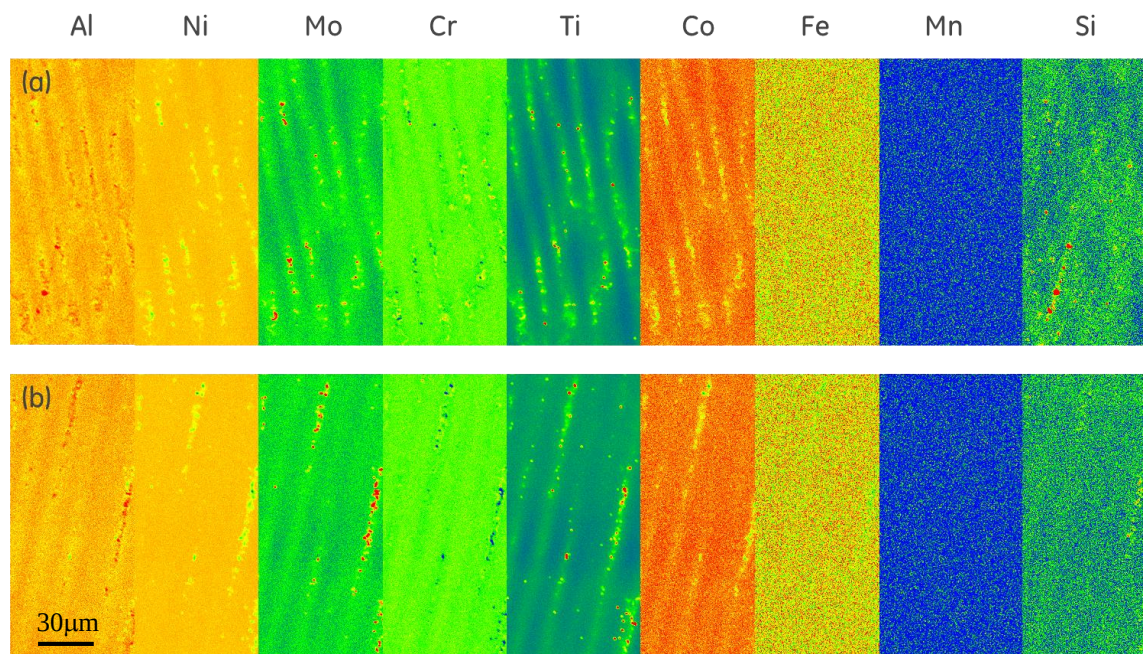


Figure 44: EPMA element maps of HA282 after post-weld heat treatment (a) PWHT3a, (b) PWHT3b. The heat treatment conditions are listed in Table 4.

To make a quantitative comparison, 1D composition profiles across dendrites were extracted from the EPMA map data. About 100 profiles were selected from each map. Each profile passed a single dendrite in the perpendicular direction and avoided interdendritic particles. The profiles were averaged and plotted for each chemical element in Figure 45. The profiles from the prior PWHT2 and as-welded conditions were also included in the figures for comparison.

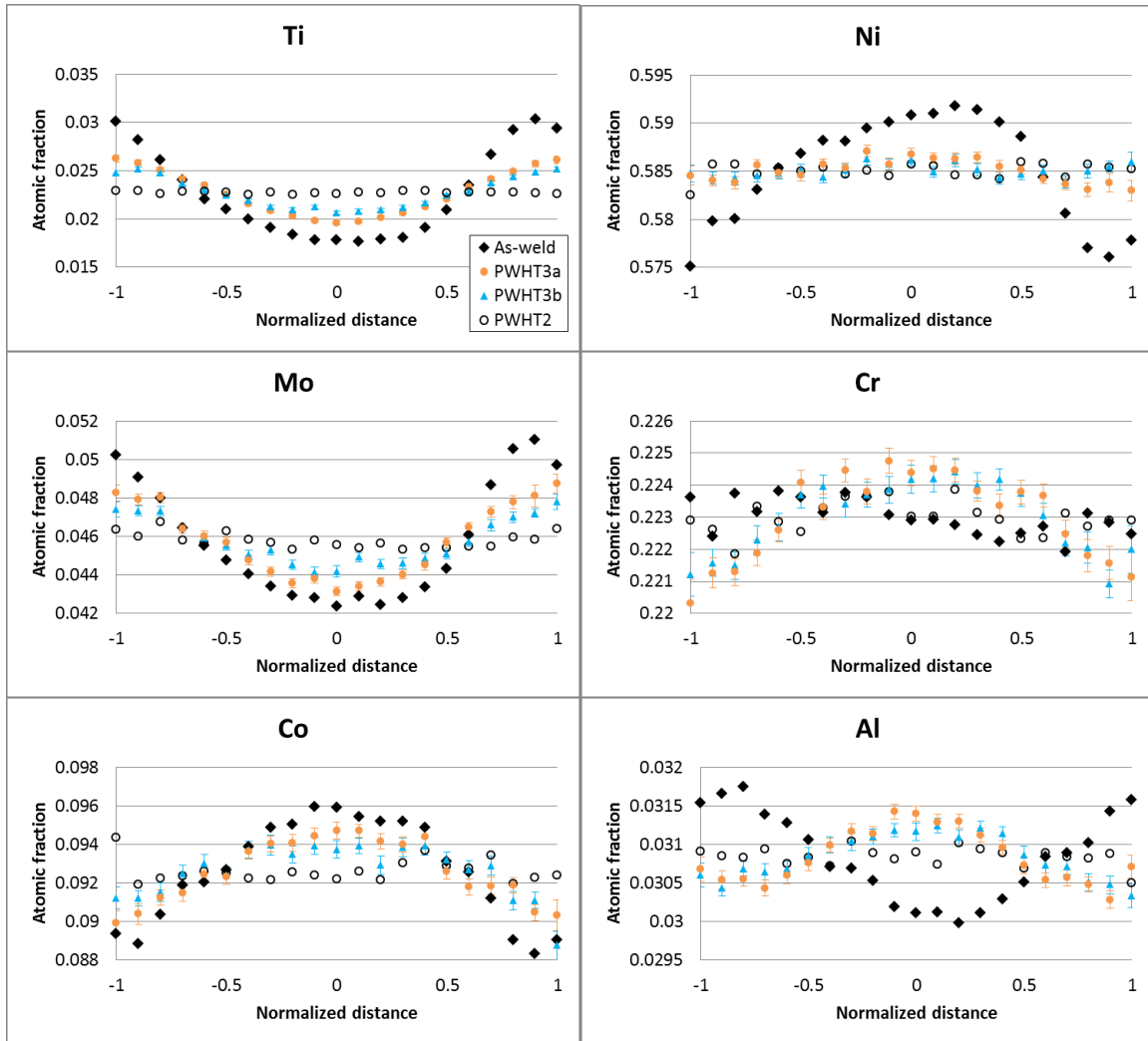


Figure 45: Comparison of microsegregation of as-weld, PWHT#3a, #3b, and PWHT#2. Dendrite core is located in the middle (0) of the normalized distance, interdendritic regions are at ± 1 . The error bars are the standard deviation of mean from about 100 selected line profiles across dendrites.

The decrease of microsegregation with respect to heat treatment condition can be clearly seen in Figure 45 and Figure 46. In Figure 46, an average composition is taken from the dendrite core and interdendritic region, respectively. In the order of heat treatment temperature and time combination, the residual microsegregation of the slow diffusion element Mo and the γ' forming elements (Ti + Al) are both progressively reduced, with the latter (Ti + Al) almost indistinguishable after PWHT2.

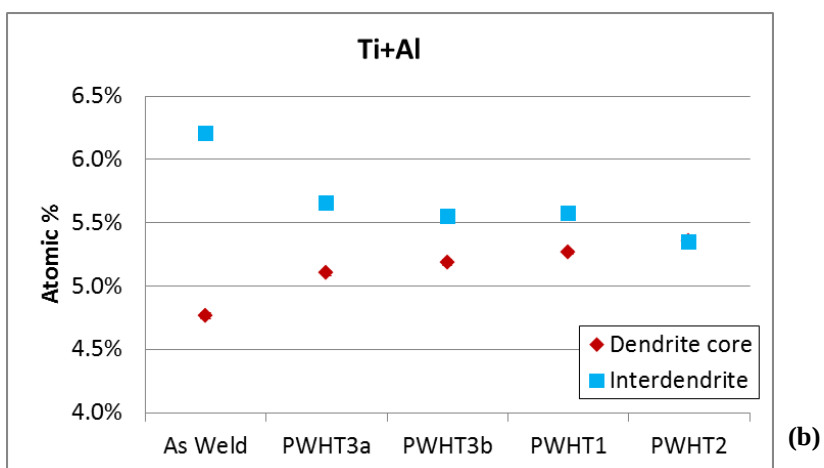
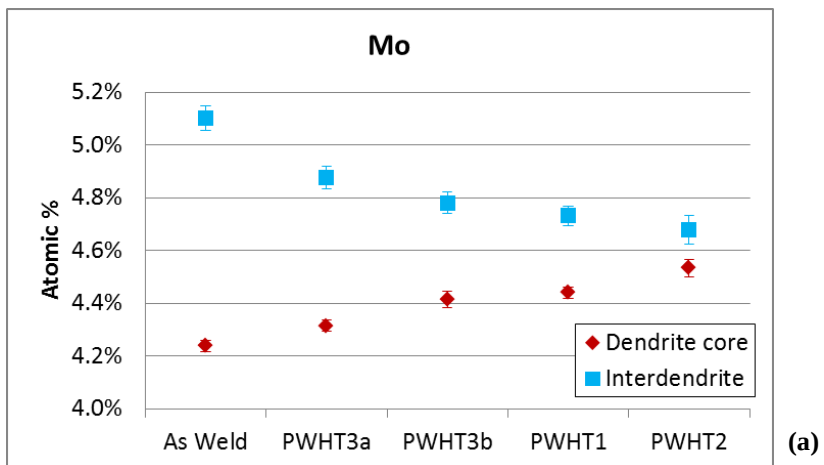


Figure 46: Comparison of the average compositions of dendrite core vs. interdendritic region among different post-weld heat treatments.

2.3.3. Microstructures stability during static thermal exposure

1000 hour

The microstructures of welded HA282 after 1000 hours of static exposures at 1500, 1600, and 1700°F are shown in Figure 47 to Figure 49. After the 1000 hour exposure at three temperatures, γ' coarsening and morphological change from spherical to cuboidal shapes were found in the base metal, HAZ, and weld zone. There was no significant difference in γ' size, morphology and distribution between the center of the grains and the areas adjacent to the grain boundaries (Figure 47c, f, i, Figure 48b, e, h, and Figure 49b, e, h). The grain boundaries were decorated with $M_{23}C_6$ and M_6C carbides and γ' precipitates in the forms of particles or films (Figure 47c). (Ti,Mo)C carbide particles were located intra-granularly in most cases, while a few were observed on grain boundaries.

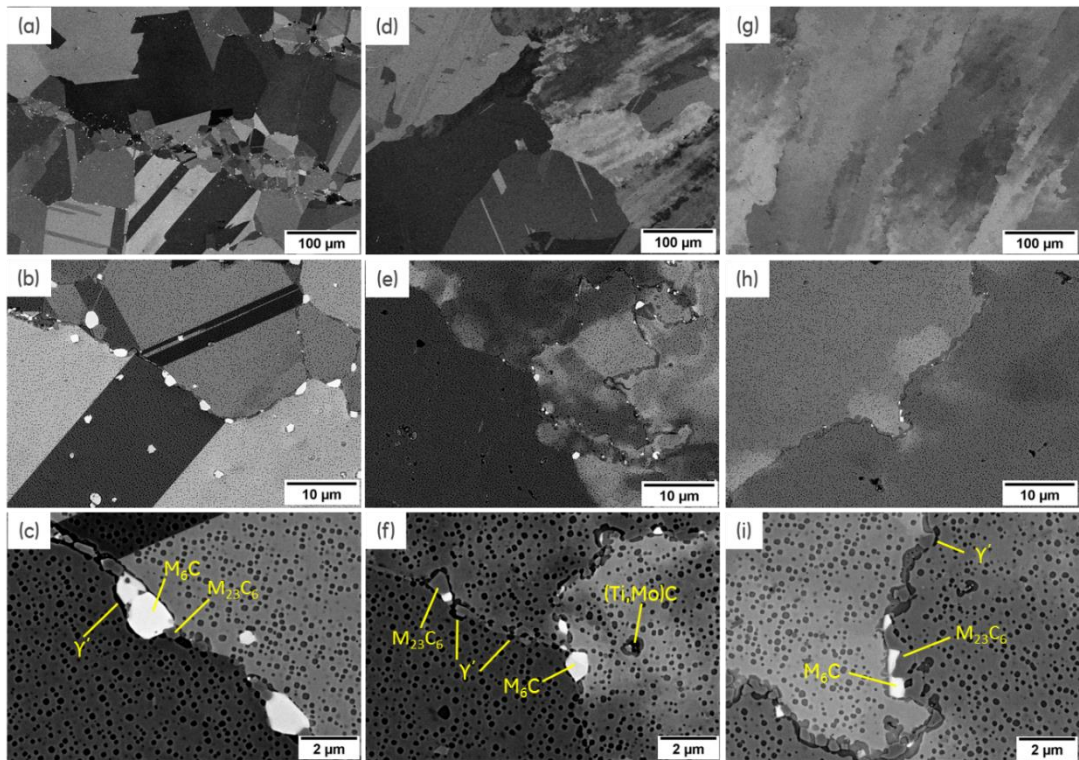


Figure 47: Microstructure of the welded HA282 after the static exposure at 1500°F for 1000 hours: (a)-(c) base metal, (d)-(f) HAZ, and (g)-(i) weld zone.

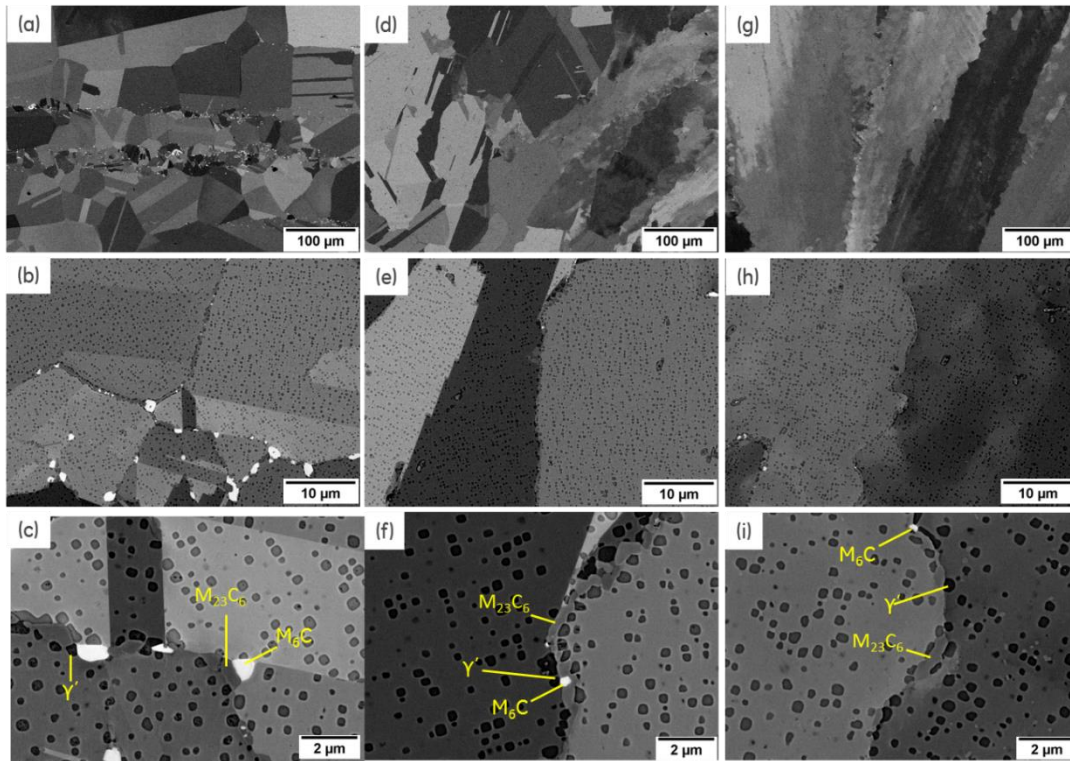


Figure 48: Microstructure of the welded HA282 after the static exposure at 1600°F for 1000 hours: (a)-(c) base metal, (d)-(f) HAZ, and (g)-(i) weld zone.

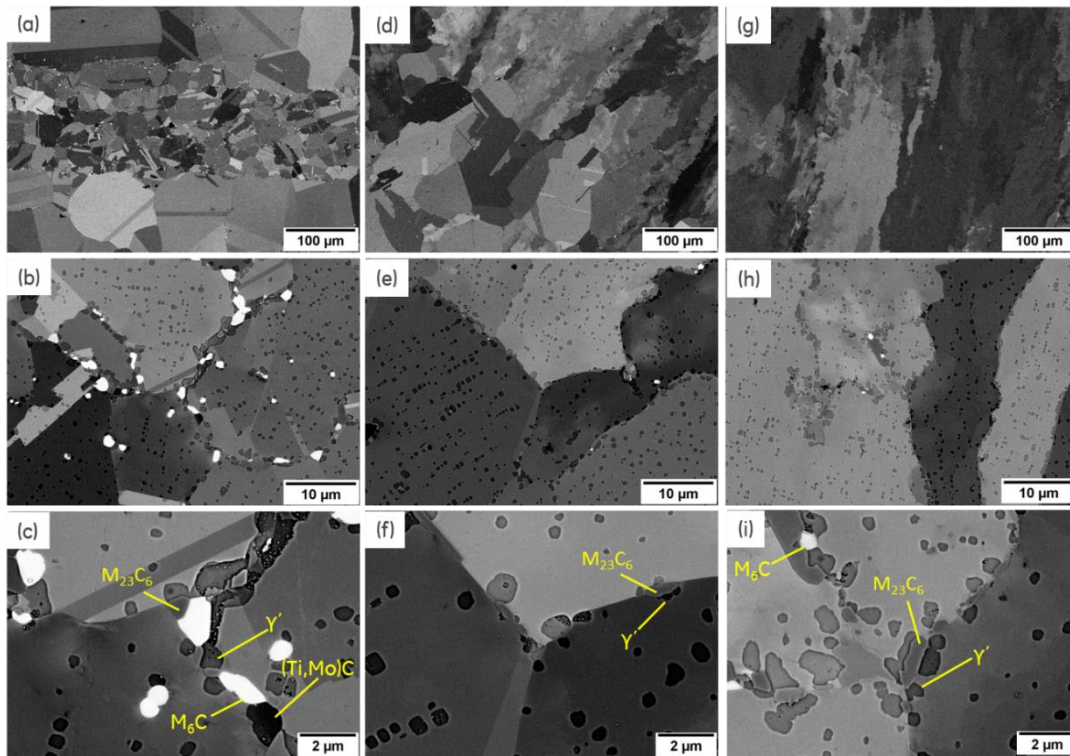


Figure 49: Microstructure of the welded HA282 after the static exposure at 1700°F for 1000 hours: (a)-(c) base metal, (d)-(f) HAZ, and (g)-(i) weld zone.

5000 hours

The microstructures of welded HA282 after 5000 hours of static exposures at 1500, 1600, and 1700°F are shown in Figure 50 to Figure 52. Compared to 1000-hour static exposure at the three temperatures, γ' and Mo-rich M_6C carbide coarsening were observed in the base metal, HAZ, and weld zone. Similar grain boundary carbides were observed after 5000 hours of exposure in comparison to 1000 hours exposure. γ' precipitates in the forms of films (Figure 52d) can sometimes be seen on a few grain boundaries.

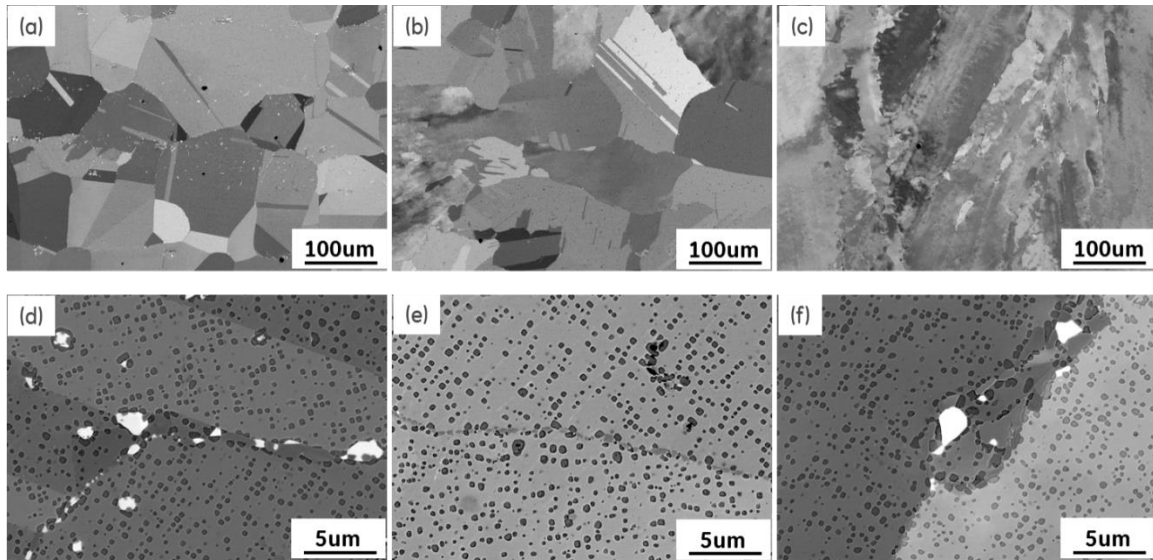


Figure 50: Microstructure of the welded HA282 after the static exposure at 1500°F for 5000 hours: (a, d) base metal, (b, e) HAZ, and (c, f) weld zone.

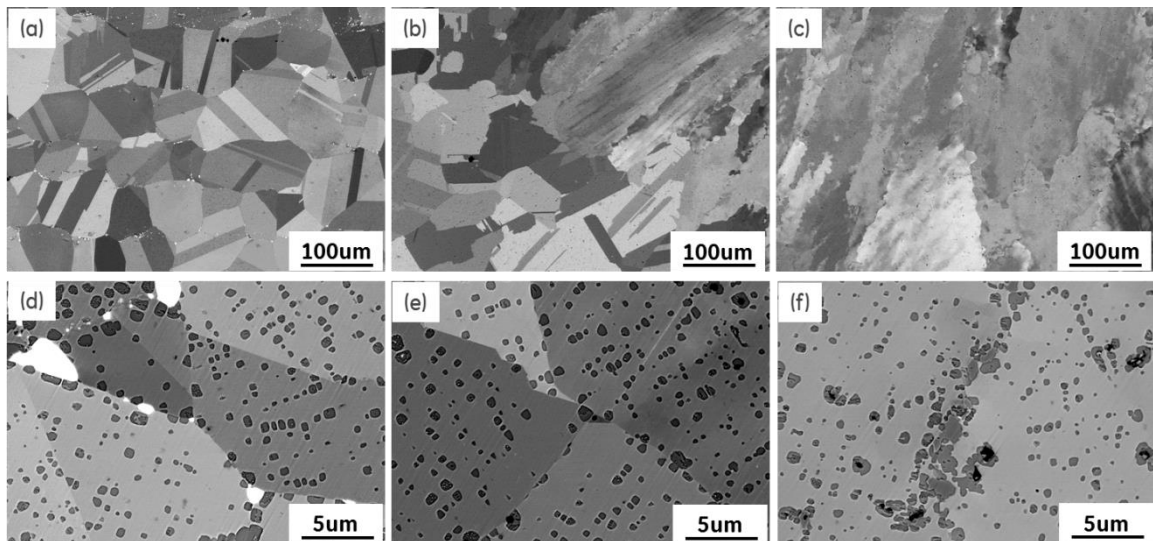


Figure 51: Microstructure of the welded HA282 after the static exposure at 1600°F for 5000 hours: (a, d) base metal, (b, e) HAZ, and (c, f) weld zone.

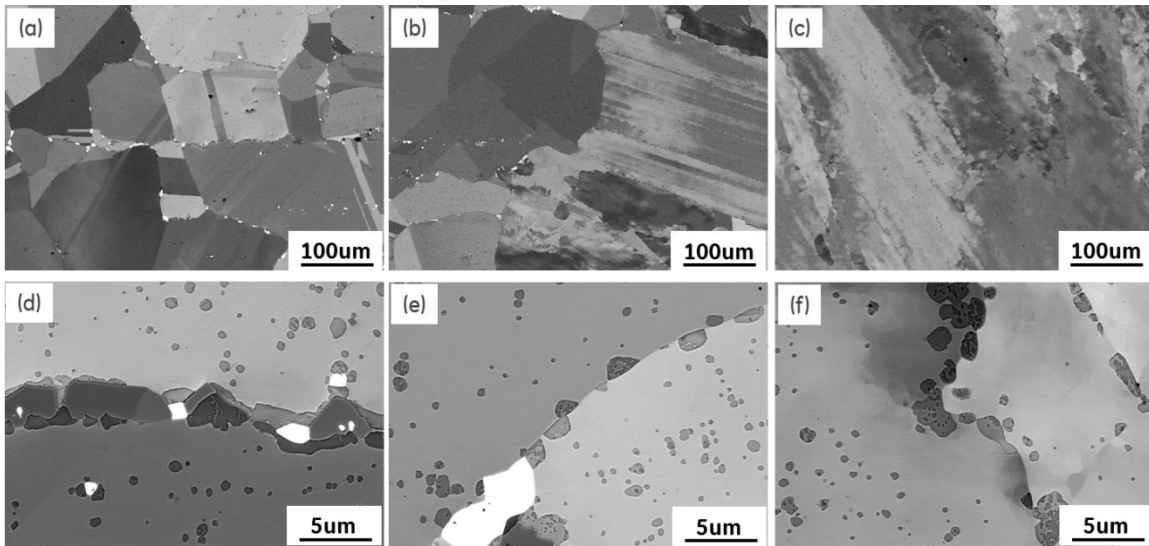


Figure 52: Microstructure of the welded HA282 after the static exposure at 1700°F for 5000 hours: (a, d) base metal, (b, e) HAZ, and (c, f) weld zone.

10,000 hours

Microstructure evolution after the 10,000 hours of static exposure was similar to 5000 hour condition. Figure 53 shows intragranular and intergranular γ' precipitates at a low magnification in welded HA282 after static exposures at 1500, 1600, and 1700°F for 10,000 hours. No difference was found in intragranular and intergranular γ' precipitates across different regions in each specimen.

In addition, a needle-shape phase as shown in Figure 54 was observed after annealing at 1600 and 1700°F for 10,000 hours. The phase was also observed in the 1500°F 10,000-hr specimen, but at a much lesser amount. None of the 5,000-hour specimens exposed at 1500-1700°F, however, showed this phase in SEM images. Due to the small size, EDS spectra only provided the phase chemistry qualitatively. The needle phase was found rich in Mo and Cr compared to the matrix. The calculated equilibrium phases suggested a possible Mo, Co-rich μ phase or σ phase in the temperature range (Figure 55). There is no evidence, however, that this phase is related to creep rupture in the present study. It is noted that Fu et al. ⁴ reported acicular Mo-rich μ phase in HA282 after 3,000 hours exposure at 1400 and 1482°F. The μ phase particles in their observation appeared to be in much larger size.

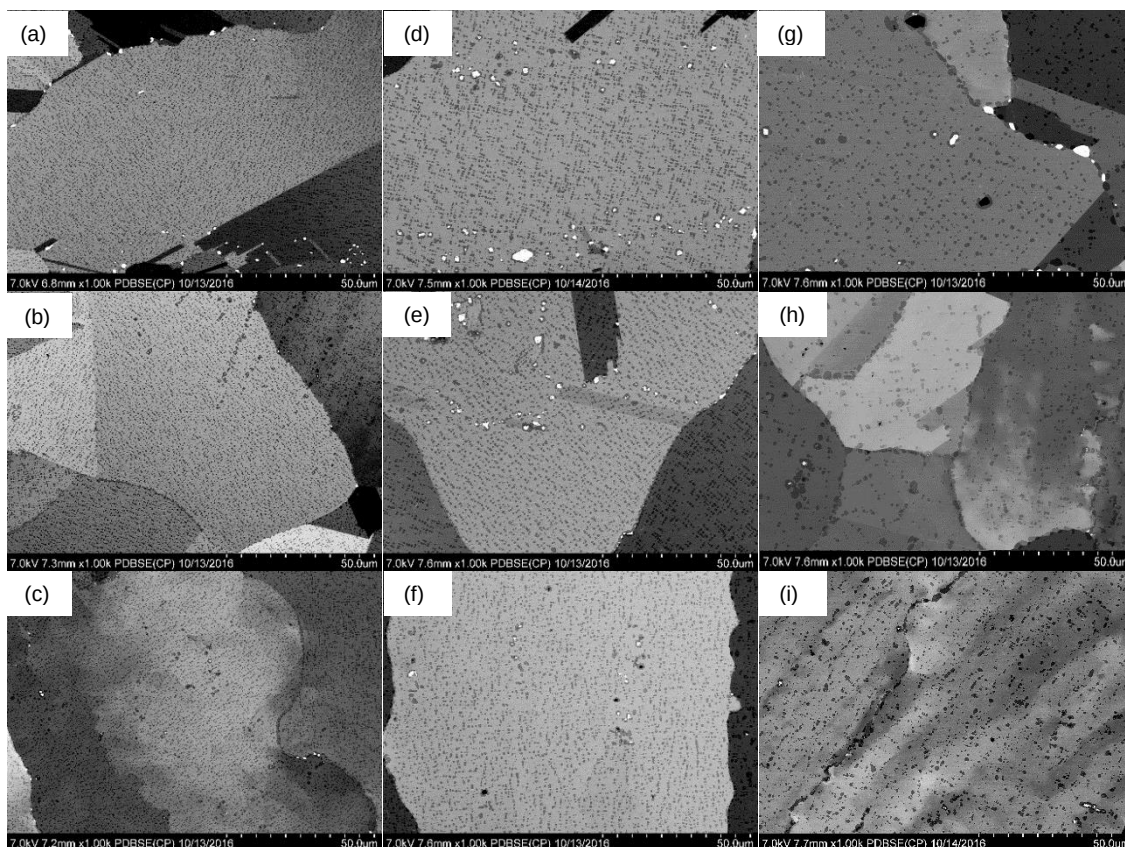
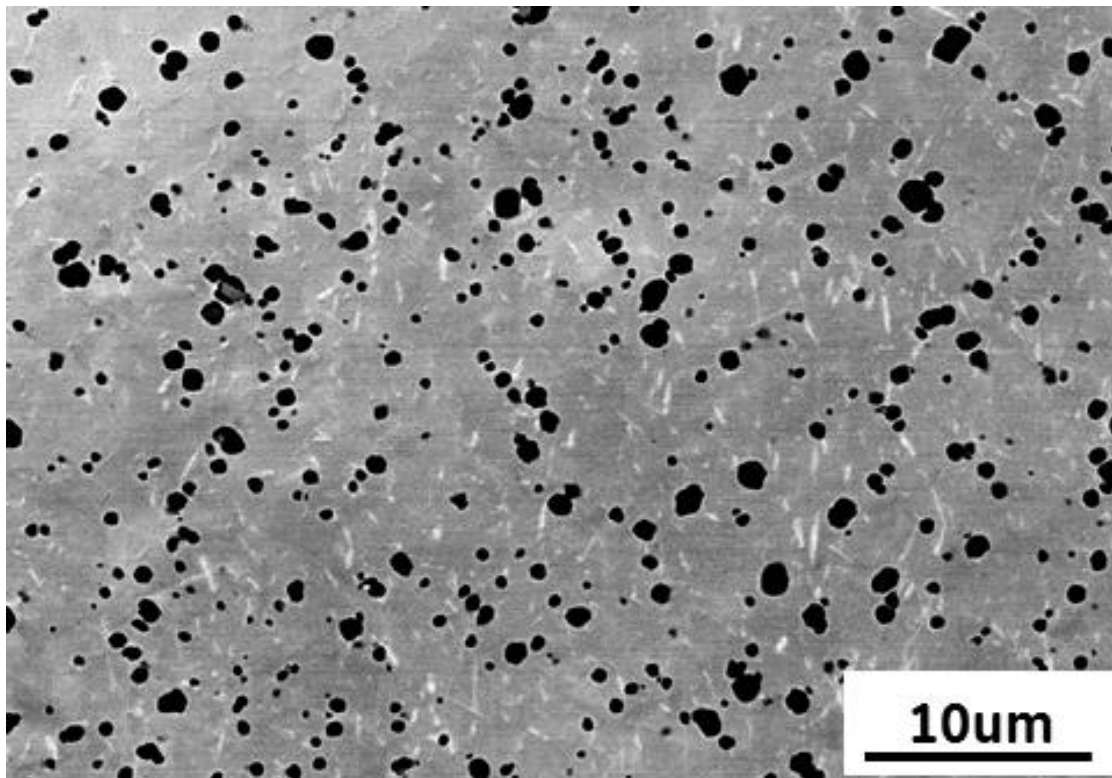
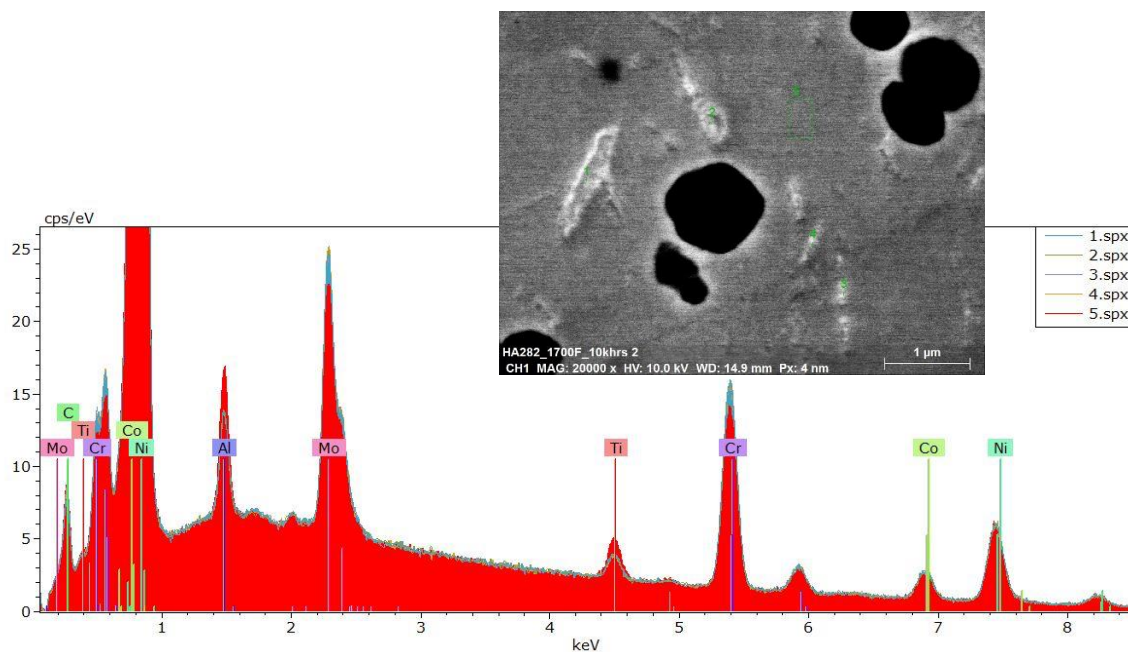


Figure 53: BSE images of γ' precipitates after static exposure of 10,000 hours: (a, b, c) 1500°F, (d, e, f) 1600°F, and (g, h, i) 1700°F 10,000 hours. (Top to bottom: base metal, HAZ, and weld zone, respectively).



(a)



(b)

Figure 54: (a) Bright contrast needle-shape phase seen after 1600°F and 1700°F 10,000 hours of exposure. (b) EDS spot spectra 1~4 suggest that this phase is enriched in Mo and Cr compared to the matrix (box spectrum 5).

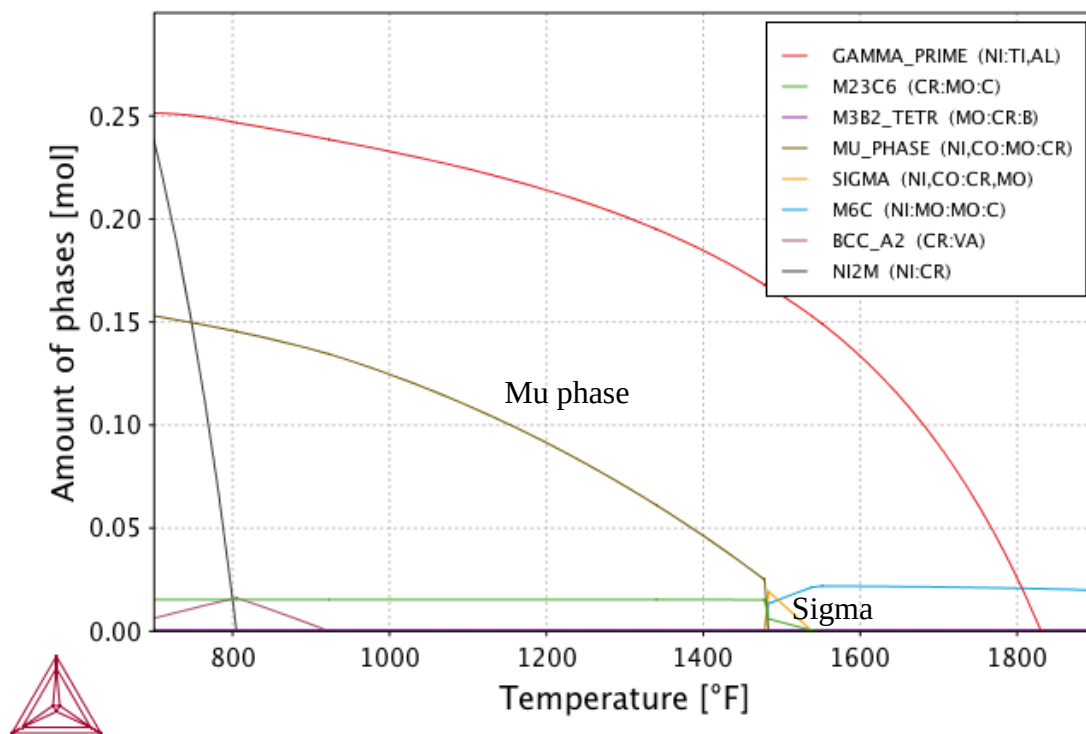


Figure 55: Calculated equilibrium phases in Haynes 282 with composition of the base metal material. TTNi7 database was used.

Gamma prime coarsening

Intragranular γ' precipitates in welded HA282 after static exposures at 1500, 1600, and 1700°F up to 10,000 hours are compared in Figure 56 to Figure 58. In general, γ' sizes increased with exposure time and temperature, and the area fractions reduced with temperature. Under most conditions, except for 1700°F, little difference of γ' size could be seen between the base metal, HAZ, and weld zone. The specimen after 1700°F 5000 hours of exposure however showed some degree of variation in γ' size, with the largest size in HAZ and the smallest in weld zone. Additional SEM images were taken from multiple locations in each region, in order to give a better examination of the microstructure variation in the 1700F/5000hr specimen. As shown in Figure 59, the γ' size distribution could be different: the weld zone appeared to have more small γ' than in the HAZ, which possibly led to a smaller average size in the former region. Some of the dark contrast particles could have been actually MC carbides, and were not separated from γ' in image analysis.

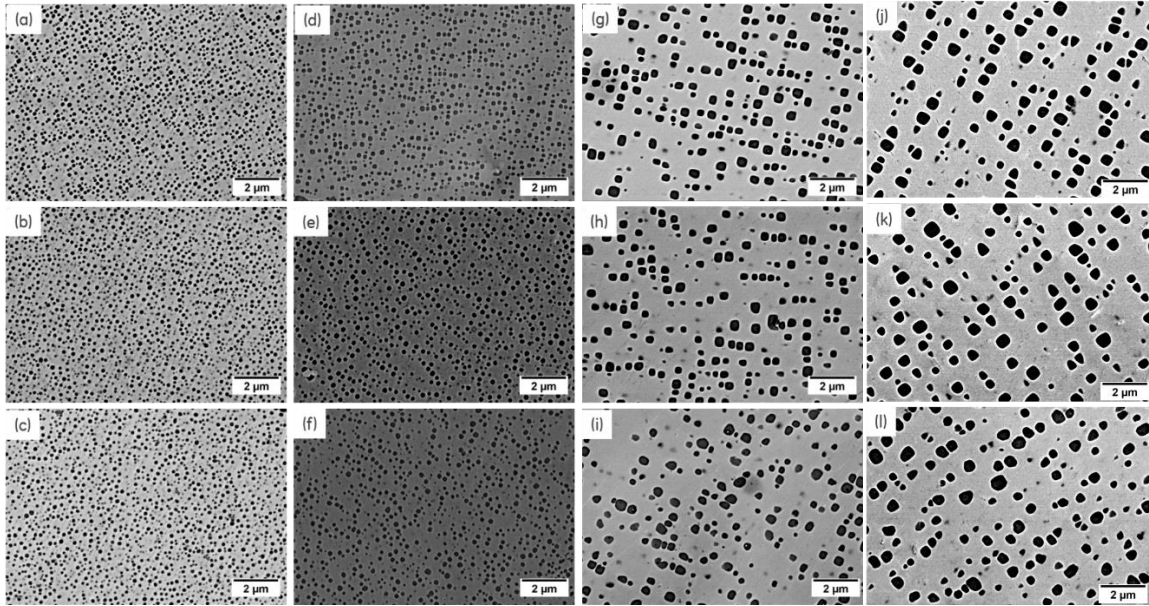


Figure 56: BSE images of intragranular γ' precipitates (from top to bottom: base metal, HAZ, weld zone) after 1500°F static exposure: (a, b, c) 500 hours, (d, e, f) 1000 hours, (g, h, i) 5000 hours, (j, k, l) 10000 hours.

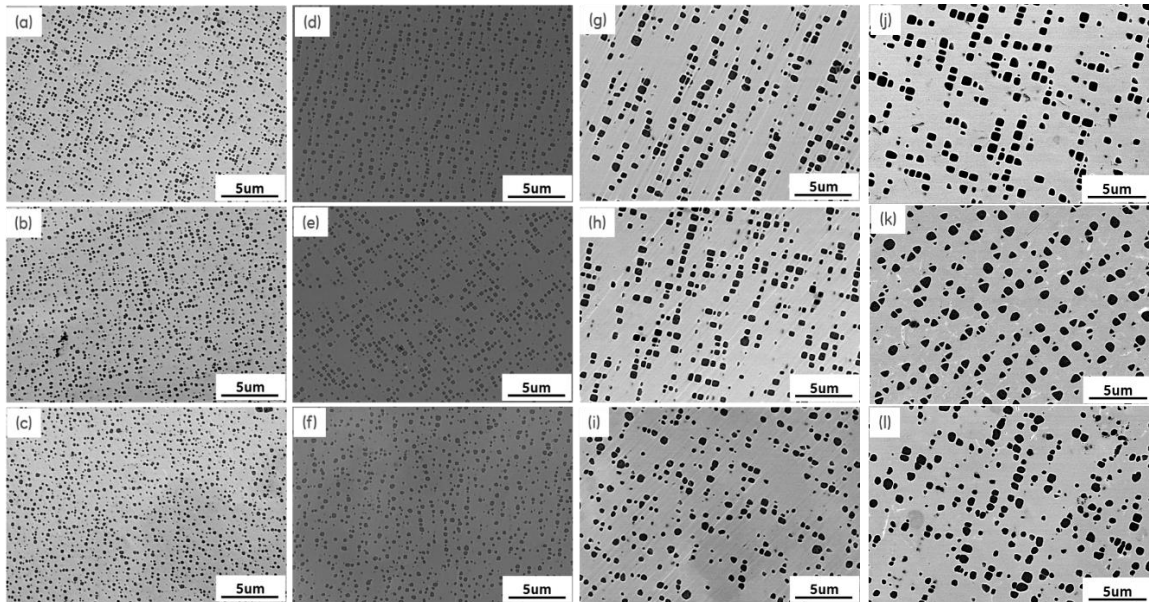


Figure 57: BSE images of intragranular γ' precipitates (from top to bottom: base metal, HAZ, weld zone) after 1600°F static exposure: (a, b, c) 500 hours, (d, e, f) 1000 hours, (g, h, i) 5000 hours, (j, k, l) 10000 hours.

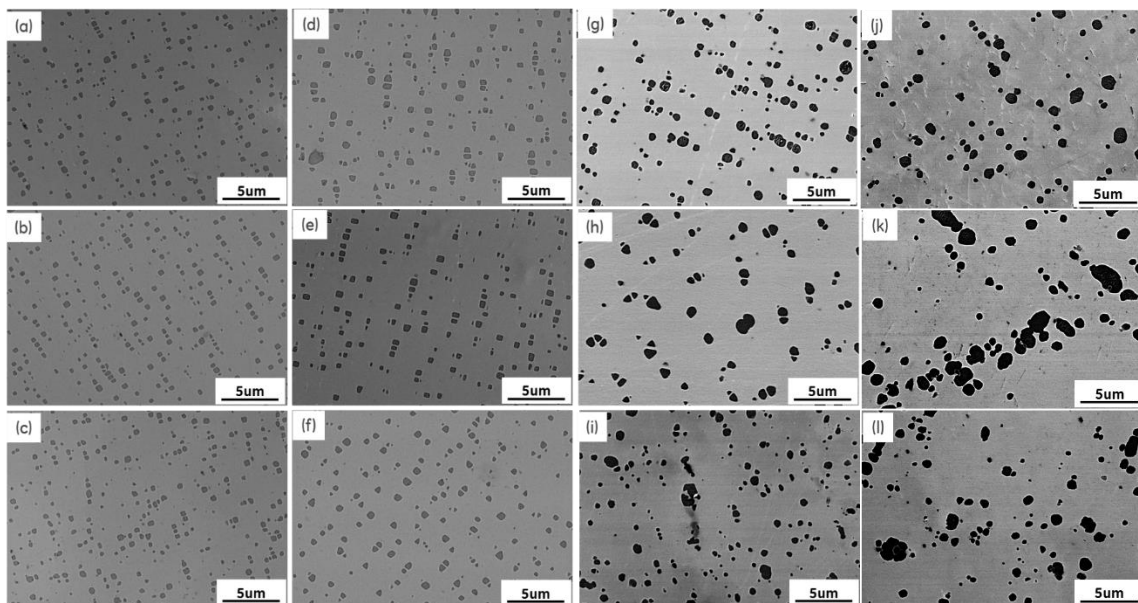


Figure 58: BSE images of intragranular γ' precipitates (from top to bottom: base metal, HAZ, weld zone) after 1700°F static exposure: (a, b, c) 500 hours, (d, e, f) 1000 hours, (g, h, i) 5000 hours, (j, k, l) 10000 hours.

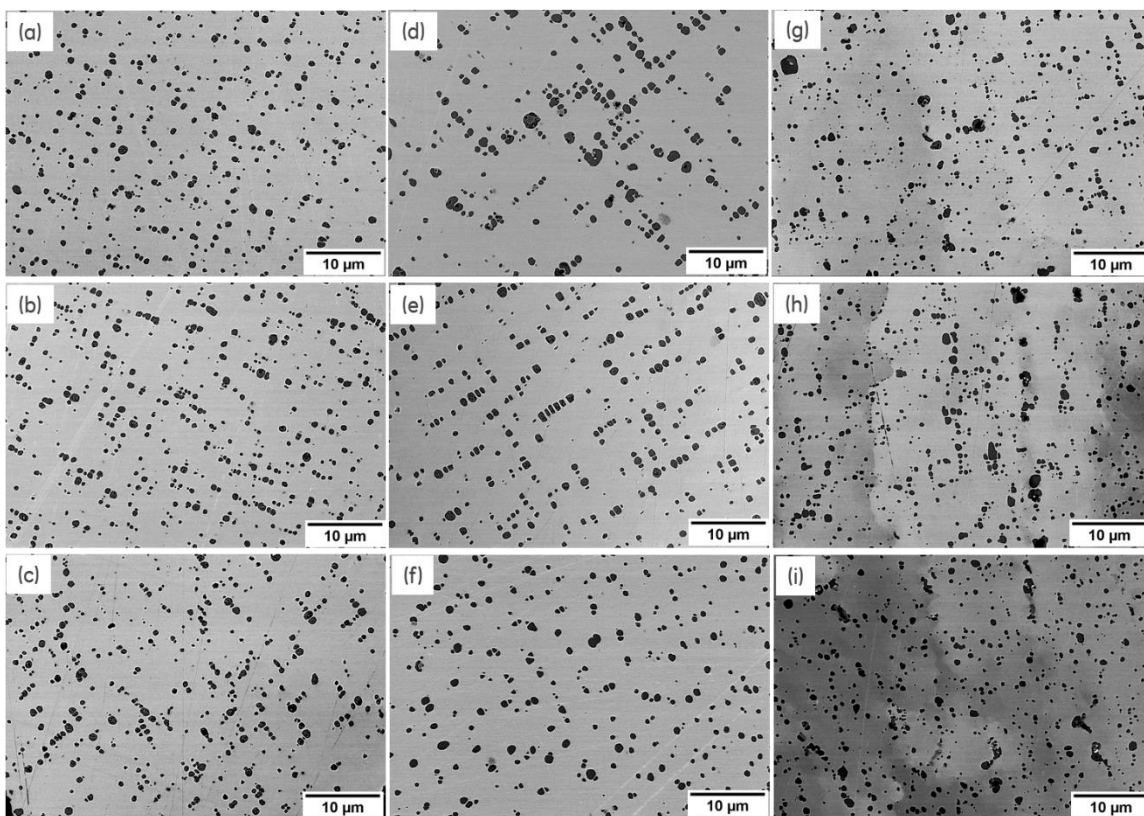


Figure 59: BSE images of intragranular γ' precipitates after 1700°F 5000 hours static exposure from three random locations in (a, b, c) base metal, (d, e, f) HAZ, and (g, h, i) weld zone.

Similar to the 5,000-hour specimen, the specimen after 1700°F 10,000 hours of exposure showed some degree of variation in γ' size, with the largest size in HAZ and the smallest in weld zone. This seems to be related to the difference in γ' size distribution: the weld zone appeared to have more fine γ' than in the HAZ, which possibly led to a smaller average size in the former region.

The size and area fraction of γ' precipitates were quantified by image analysis. Contrast threshold was applied on back scattered electron (BSE) images to highlight the γ' particles, and then binary image was created to convert γ and γ' phases to white and black regions. Additional corrections were made on the binary images to improve accuracy of measurement, including removing particles embedded under the surface and breaking up connected particles. The measured size was converted to an equivalent spherical diameter for all γ' particles, regardless of cuboidal or spherical shape. Four images at random locations were taken for quantification in order to improve statistics. Average γ' area fraction and spherical diameter are plotted in Figure 60. Each data point was averaged from multiple images. The standard deviations are also plotted in Figure 60. These data have been used to validate precipitation modeling in Task 3 (see Section 3.3 of this report).

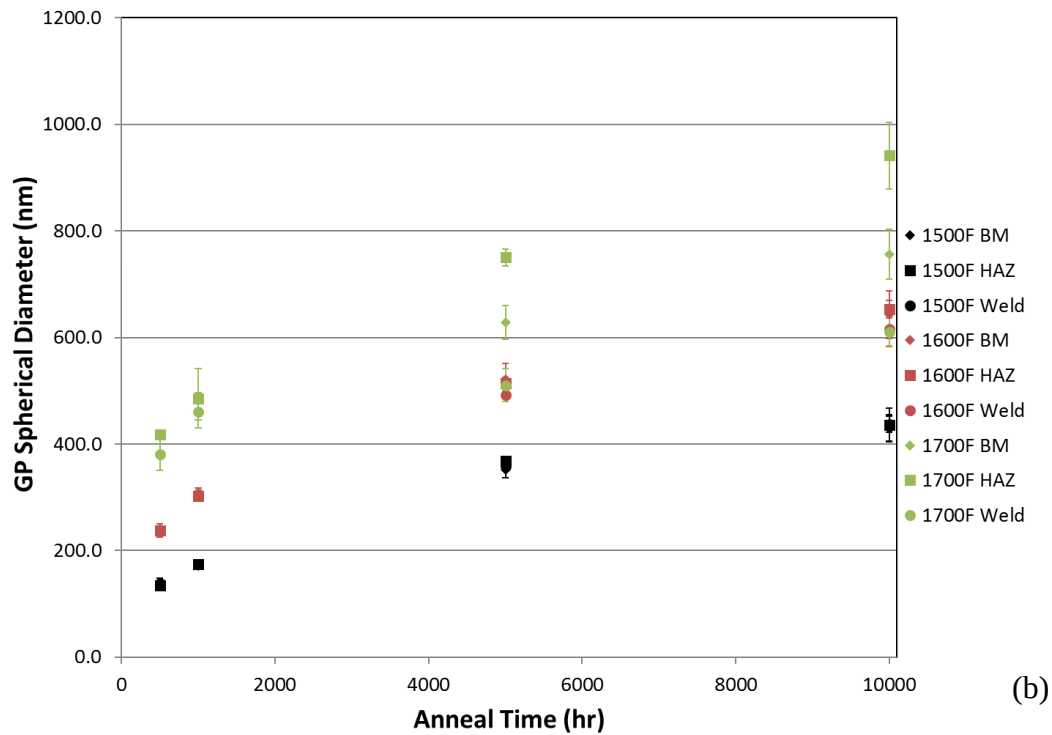
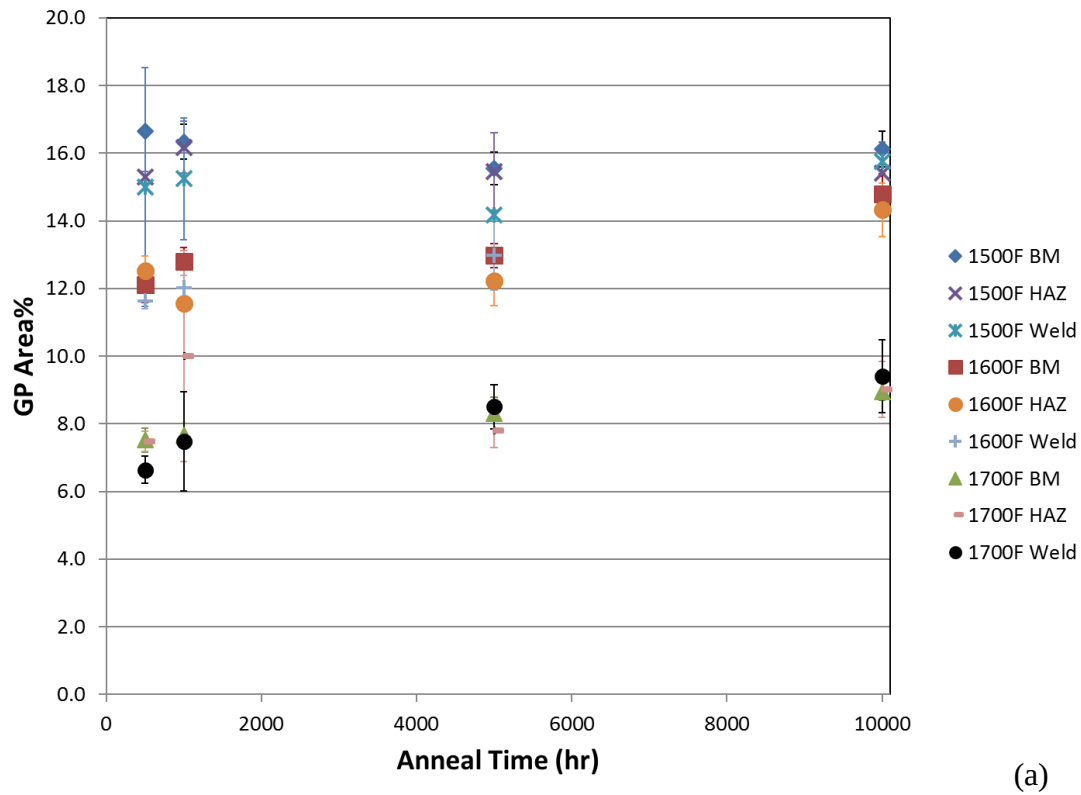


Figure 60: (a) Area fractions and (b) sizes of γ' precipitates in base metal, HAZ, and weld after static exposure. The error bars represent standard deviation among multiple images.

2.3.4. Failure location and mechanism

To examine the failure location and mechanisms in the failed creep and tensile specimens, of as-welded, heat treated welded (PWHT2, PWHT3a) and 100% weldment HA282 materials, post-test characterization was performed (see summary in Table 13). Some representative micrographs are shown in this section.

Table 13: Post-test characterization matrix for examining failure locations and mechanisms in the failed specimens, of as-welded, heat treated welded (PWHT2 and PWHT3a), and 100% weldment HA282 materials, that have undergone tensile and creep loading at 1500, 1600, and 1700°F.

Material	Test	Temp	Stress	Failure Location	Failure mechanism
As-welded HA282	Tensile	1500°F	-	within weld zone	combination of both inter-dendritic and inter-granular
		1600°F			
		1700°F			
PWHT2 HA282		1500°F		within weld zone	inter-granular
		1600°F			
		1700°F			
100% Weldment		1700°F		-	inter-granular
PWHT2 HA282	Creep	1500°F	30 ksi	within weld zone (in some cases, closer to weld and base-metal interface)	inter-granular
			25 ksi		
			15 ksi		
		1600°F	25 ksi		
			15 ksi		
			15 ksi		
		1700°F	15 ksi		
			10 ksi		
			10 ksi		
			8 ksi		
			8 ksi		
			6 ksi		
100% Weldment		1600°F	25 ksi	gauge section	inter-granular
		1700°F	15 ksi		

Tensile specimens: PWHT2 welded, 100% weldment HA282

Series of representative micrographs, shown in Figure 61 through Figure 65, illustrate the failure location and mechanism(s) in the as-welded, PWHT2 welded and 100% weldment HA282 materials, tensile tested at various temperatures. All of the tensile specimens failed within the weld zone in the gauge section. Several inter-granular micro-cracks were observed near the fracture surface in the cross-sectioned, polished and etched gauge sections. In the as-welded specimens, fracture surface showed combination of inter-dendritic and inter-granular fracture, as compared to dominant inter-granular fracture in PWHT2 and 100% weldment materials. Moreover, failure locations and mechanism(s) did not

appear to change at different temperatures, as well as between PWHT2 welded and 100% weldment material.

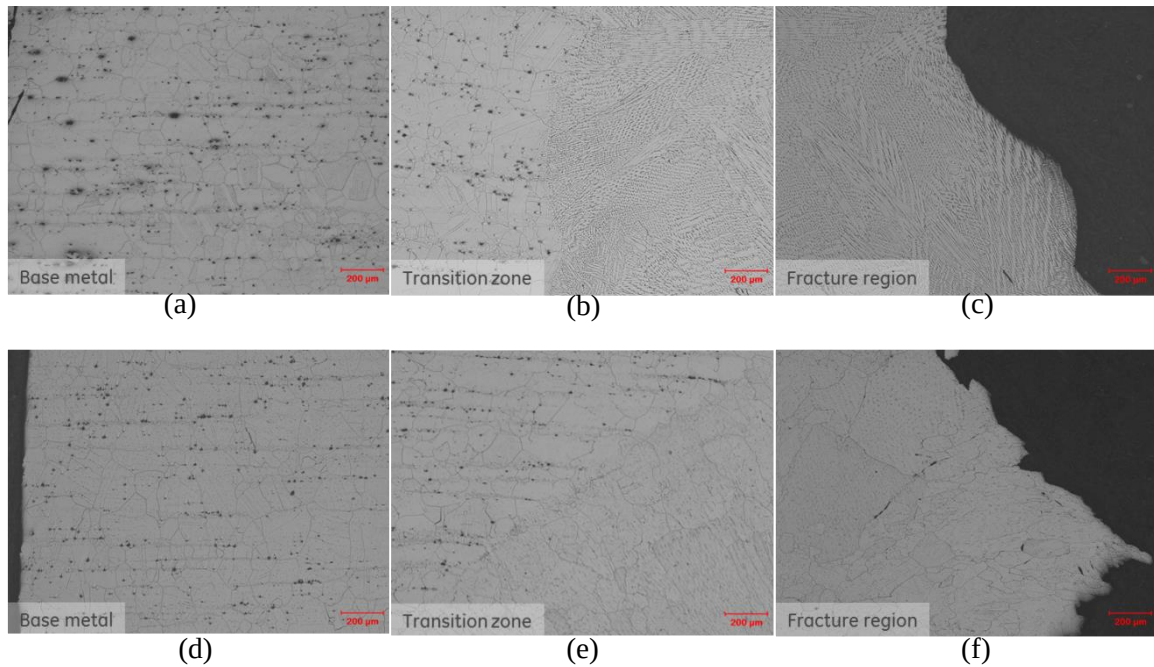


Figure 61: Optical micrographs taken from tested tensile specimens at 1600°F: (a, c) as-welded, and (d, f) PWHT2 welded HA282 material.

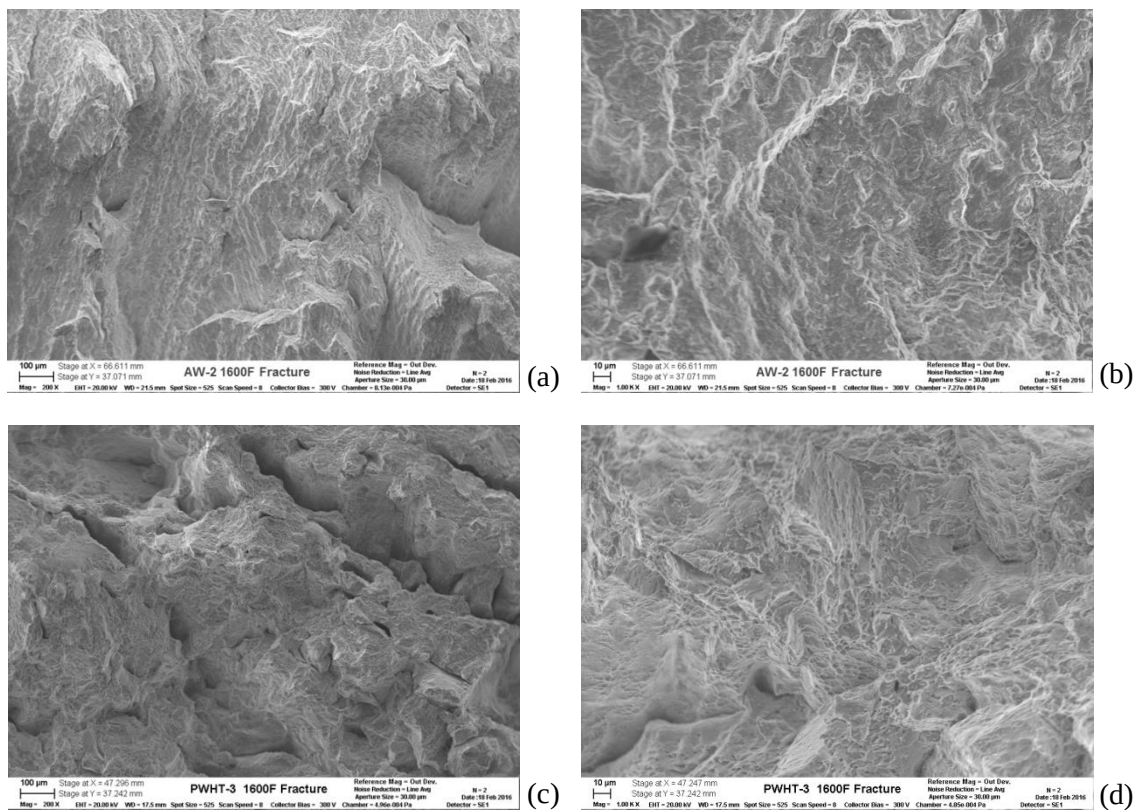


Figure 62: SEM fractographs taken from tested tensile specimens at 1600°F: (a, b) as-welded, and (c, d) PWHT2 welded HA282 material.

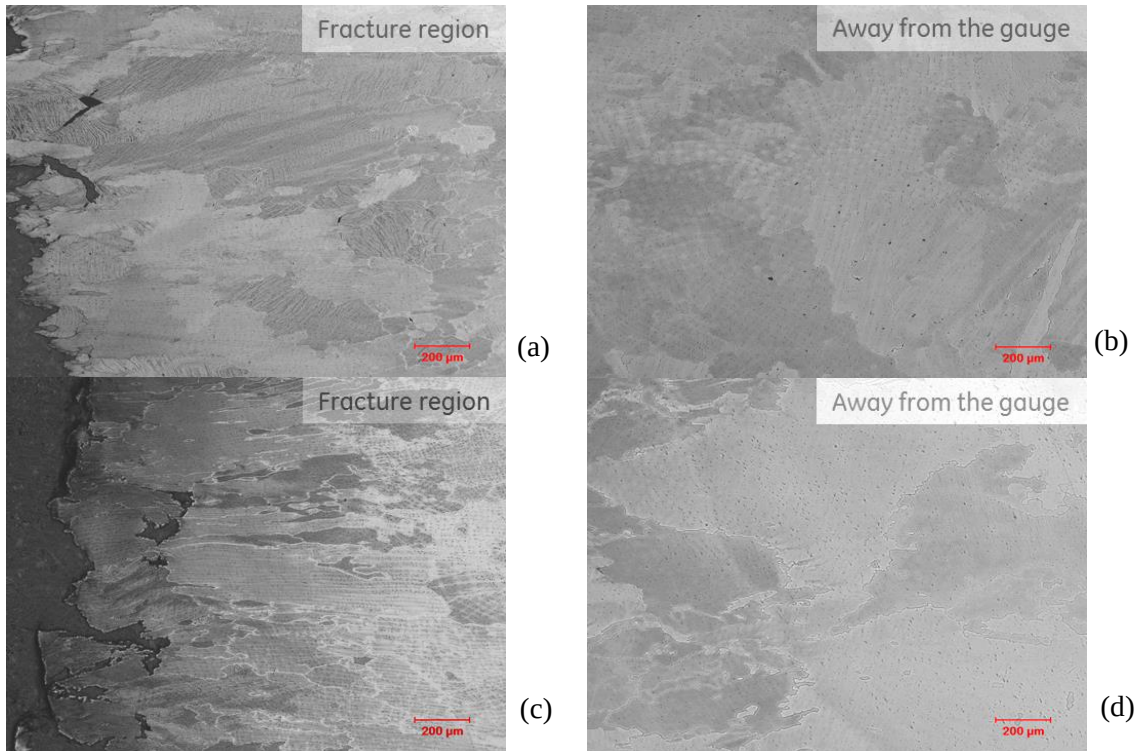


Figure 63: Optical micrographs taken from tested tensile specimens at 1700°F: (a, b) 100% weldment (transverse direction) HA282, and (c, d) 100% weldment (parallel direction) HA282.

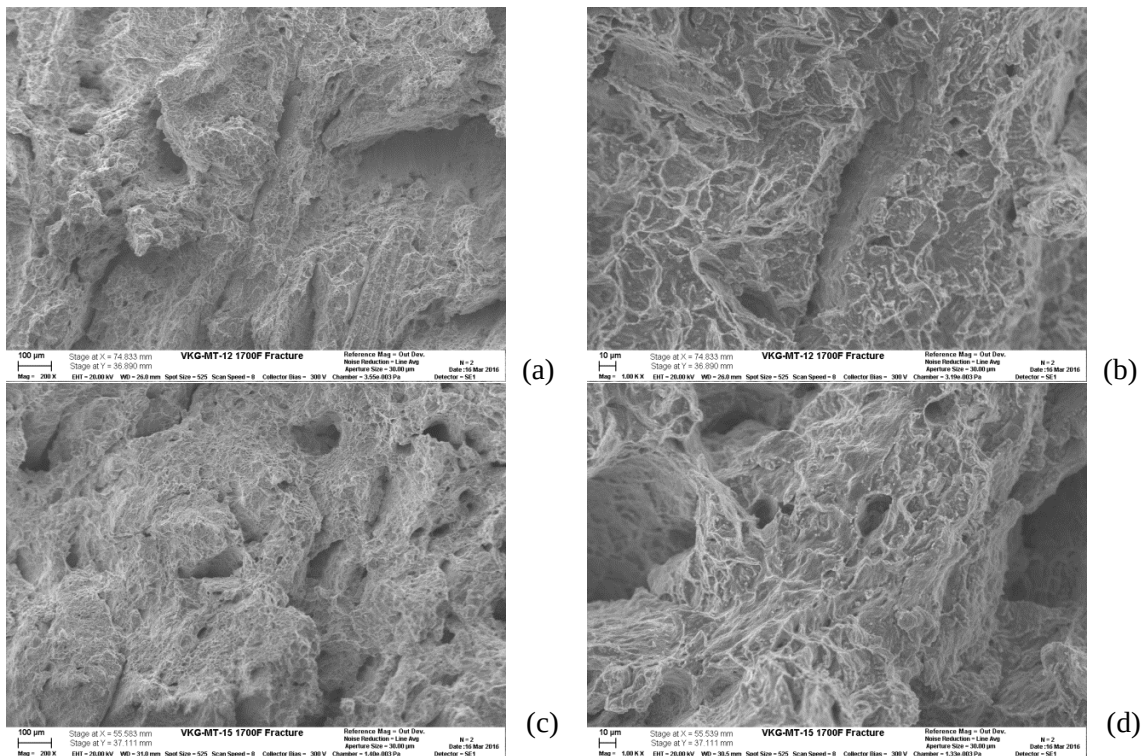


Figure 64: SEM fractographs taken from tested tensile specimens at 1700°F: (a, b) 100% weldment (transverse direction), and (c, d) 100% weldment (parallel direction) HA282.

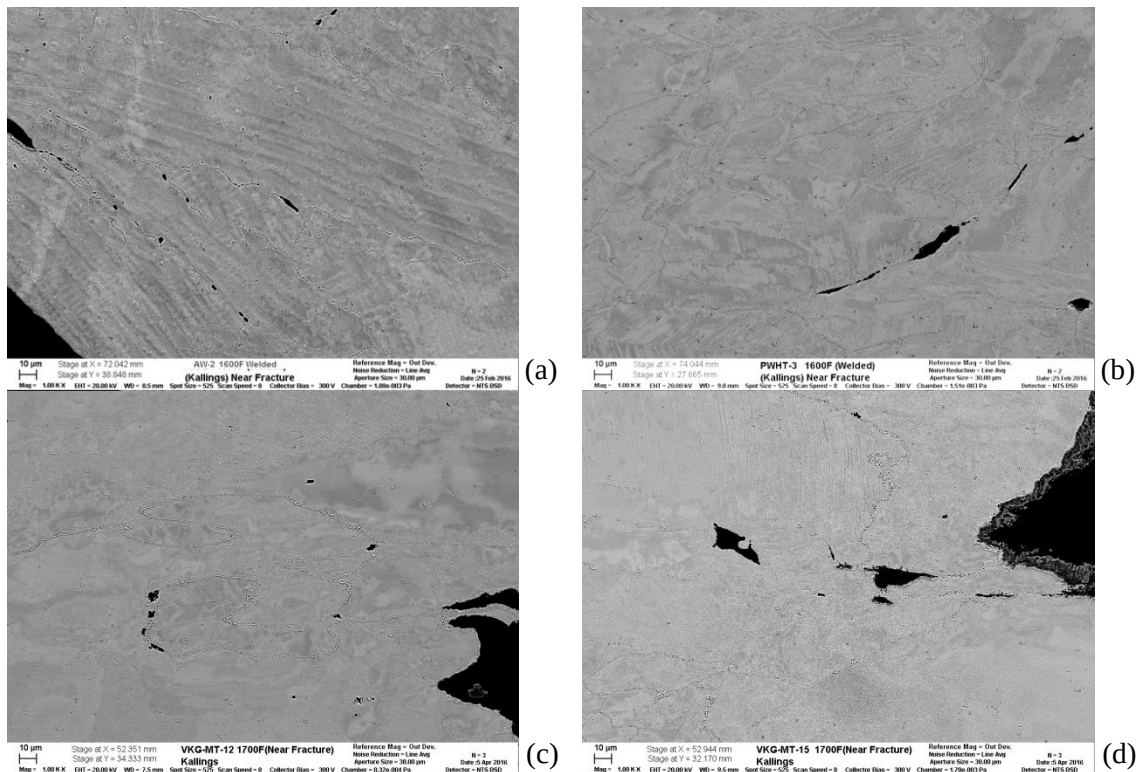


Figure 65: SEM micrographs taken from tested tensile specimens: (a) as-welded HA282, 1600°F, (b) PWHT2 welded HA282, 1600°F, (c) 100% weldment (transverse) HA282, 1700°F, and (d) 100% weldment (parallel) HA282, 1700°F.

Creep specimens: PWHT2 welded HA282

Series of representative micrographs, shown in Figure 66 through Figure 68, illustrate the failure location and mechanism(s) in the PWHT2 welded HA282 material, creep tested at various temperatures at different stresses. All the creep specimens failed within the weld zone in the gauge section. SEM fractographs suggested inter-granular failure mechanism, which was also confirmed by optical/SEM images taken from polished/etched cross-section of gauge sections. Large inter-granular cracks (also function of large creep strains) were seen near the fracture surface within the weld-regions.

Further EBSD analysis (Figure 70) confirms that these cracks are inter-granular cracks rather than inter-dendritic cracks, since they all appear on high-angle grain boundaries while low-angle grain boundaries (inter-dendritic boundaries) remain intact. Figure 71 also shows that the inter-granular cracks also propagate along the interface between matrix and grain boundary carbides rather than cracking through the carbides.

Additionally, surface cracks were present in most of the creep specimens subjected to higher temperatures/lower stresses, which also resulted into large creep strains. For example, series of SEM images (Figure 69) taken from an

interrupted bar show surface cracks may have formed during early stages of creep deformation, but have only grown to longer cracks at last stage of the creep rupture (e.g., during necking). Notably, failure locations and mechanisms did not appear to change at different temperatures.

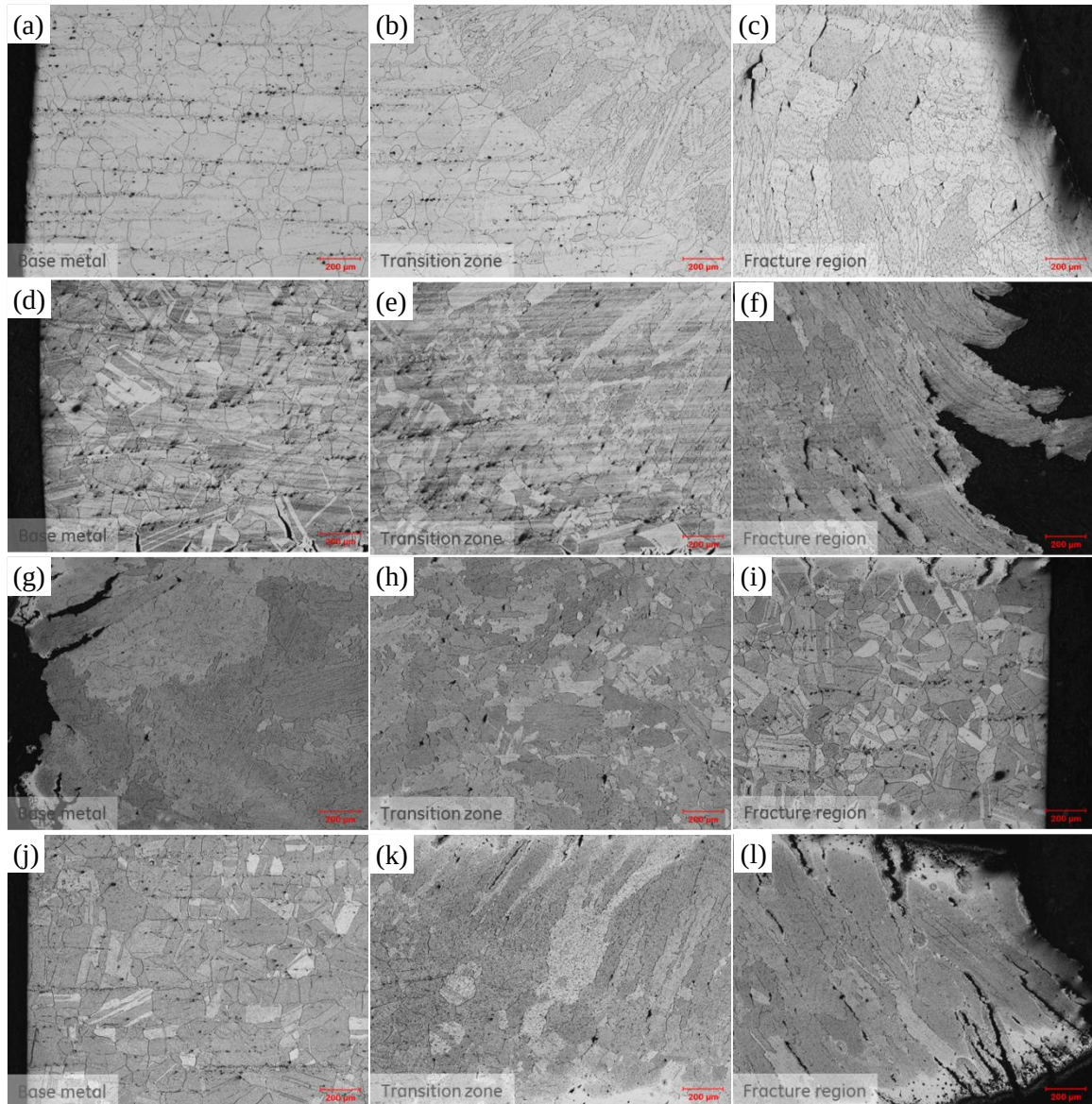


Figure 66: Optical micrographs taken from failed specimens of PWHT2 welded HA282 material, creep tested at: (a-c) 1500°F, 30ksi, (d-f) 1700°F, 15ksi, (g-i) 1700°F, 10ksi, and (j-l) 1700°F, 8ksi.

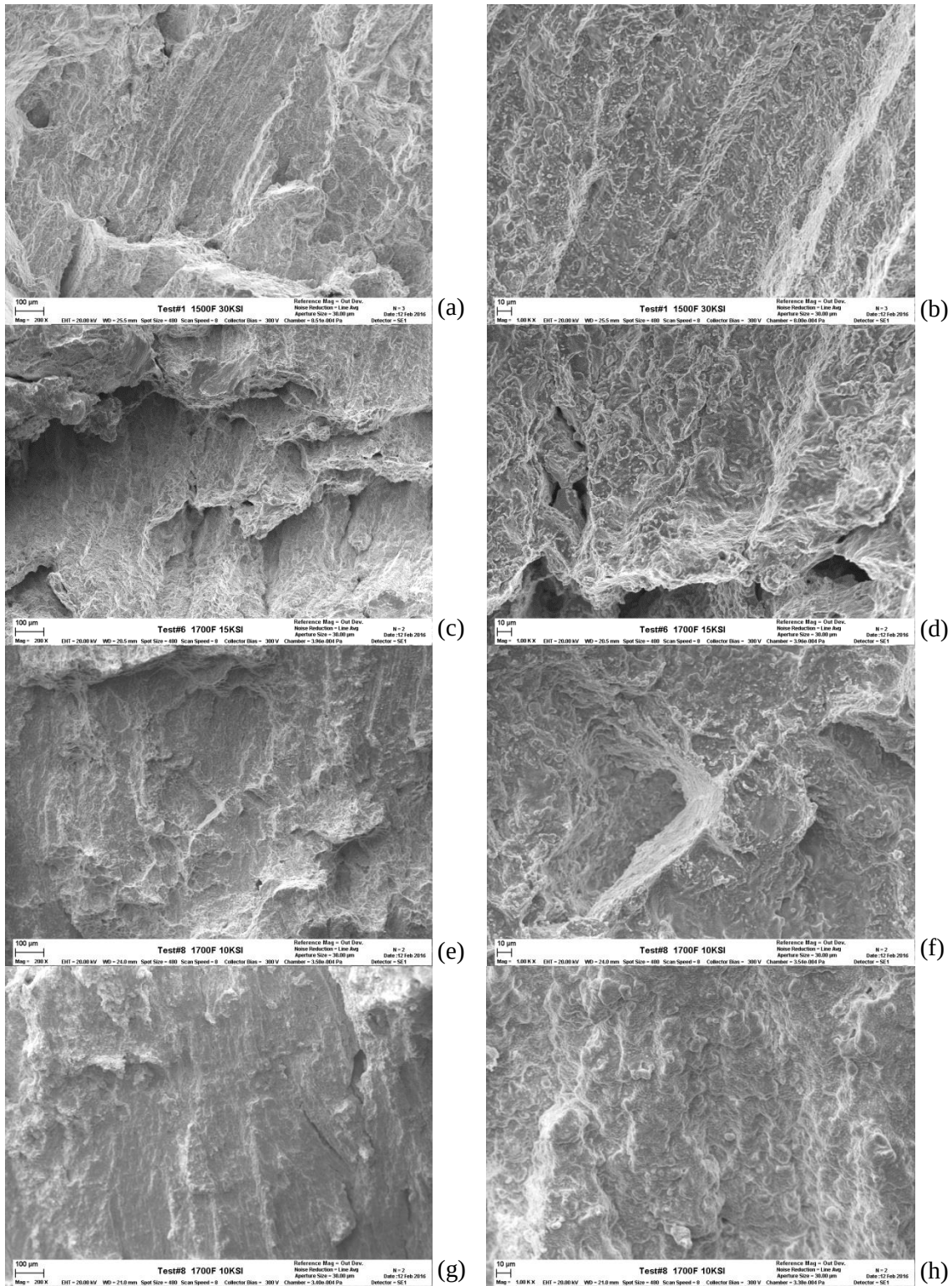


Figure 67: SEM fractographs taken from failed specimens of PWHT2 welded HA282 material, creep tested at: (a, b) 1500°F, 30ksi, (c, d) 1700°F, 15ksi, (e, f) 1700°F, 10ksi, and (g, h) 1700°F, 8ksi.

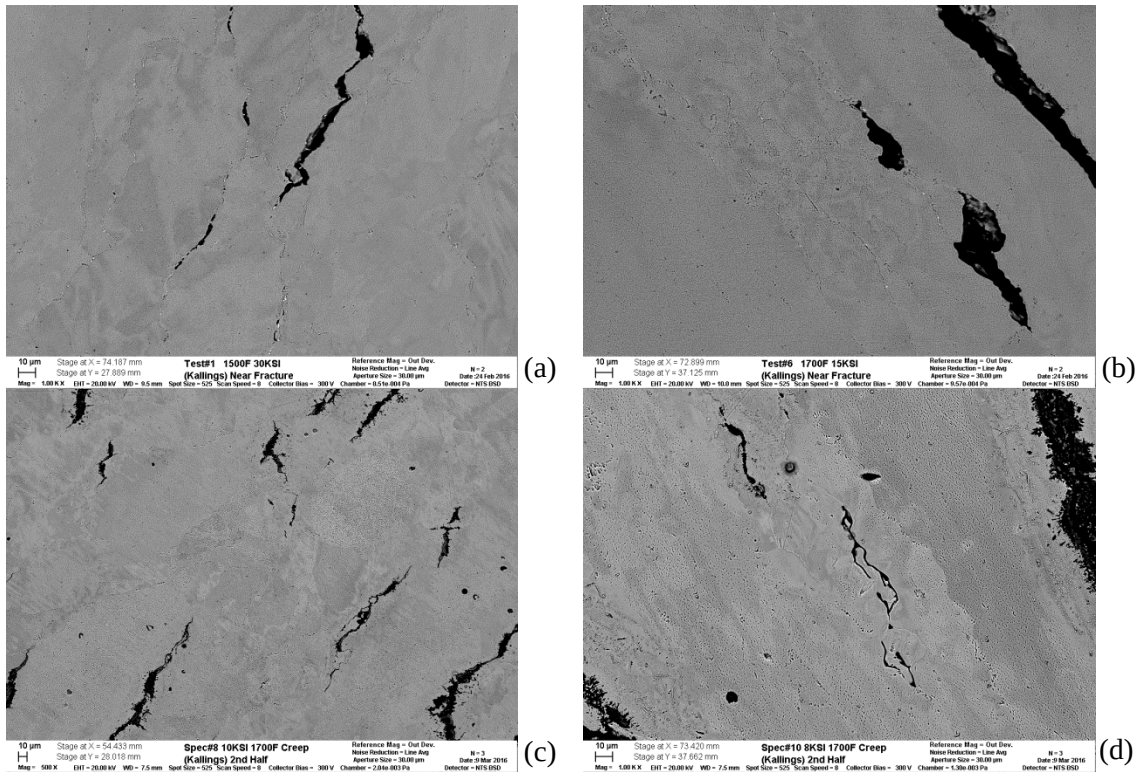


Figure 68: SEM micrographs taken from failed specimens of PWHT2 welded HA282 material, creep tested at: (a) 1500°F, 30 ksi, (b) 1700°F, 15 ksi, (c) 1700°F, 10 ksi, and (d) 1700°F, 8ksi.

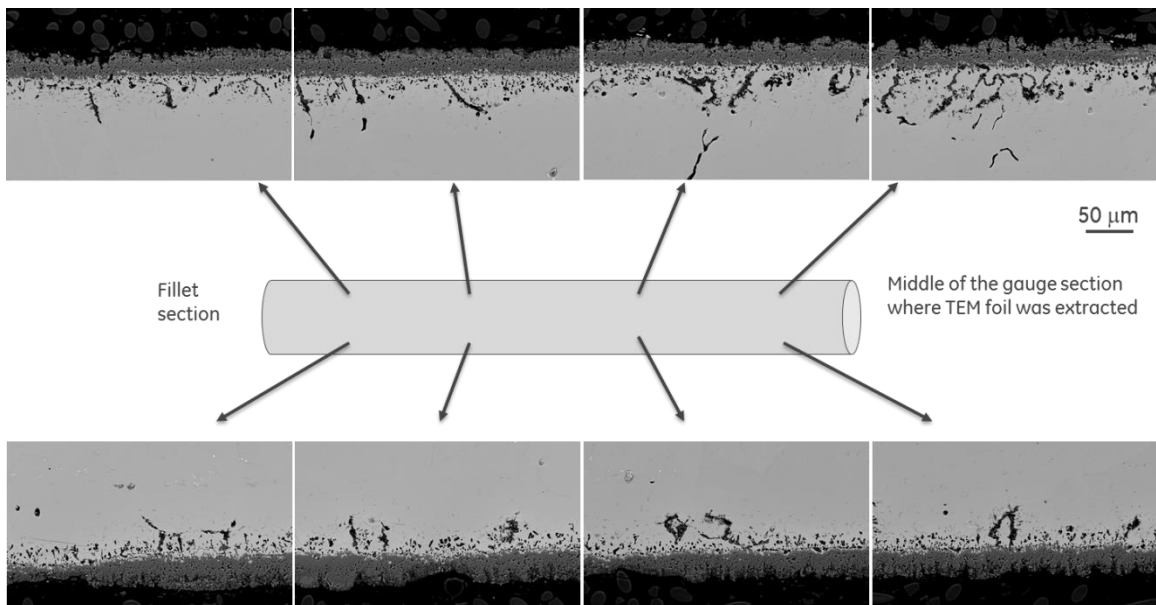


Figure 69: Series of SEM micrographs taken from an interrupted (1000 hrs) specimen of PWHT2 HA282 material, creep tested at 1700°F, 8ksi.

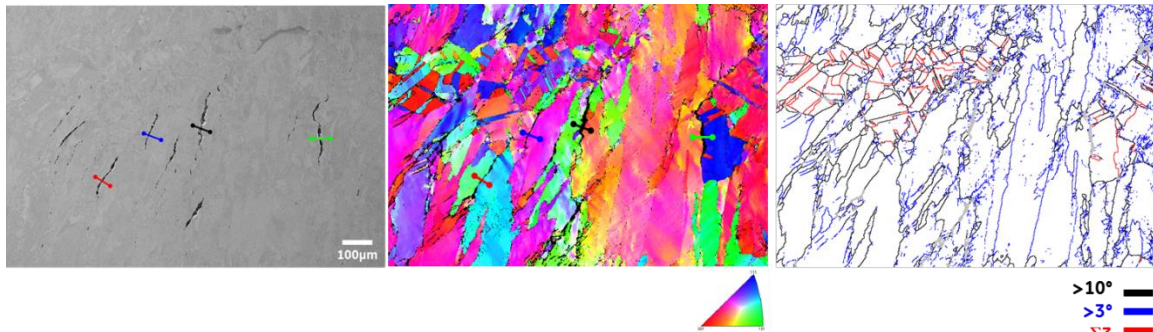


Figure 70: SE image, inverse pole figure (IPF) map, and grain boundary (GB) map of bulk cracks close to fracture surface in PWHT2 welded HA282 samples after creep rupture. All cracks appear to be on high angle grain boundaries (i.e., inter-granular cracks). Notably, low angle grain boundaries (inter-dendritic regions) remain intact.

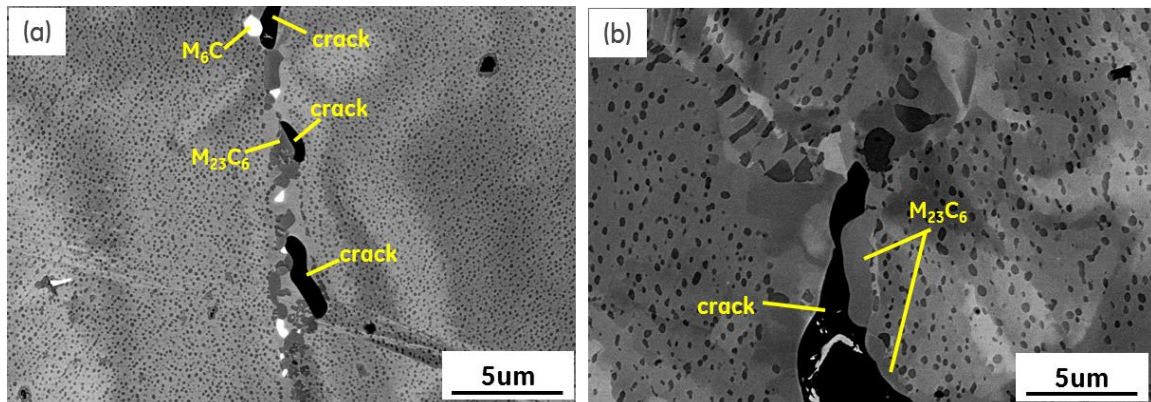


Figure 71: Inter-granular cracks close to fracture surface in the creep specimen of PWHT2 welded HA282 material, tested at (a) 1500°F, 30 ksi; (b) 1600°F, 15 ksi.

Creep specimens: PWHT3a welded HA282

A small set of PWHT3a welded HA282 specimens were also creep tested to evaluate the differences (or lack thereof) between PWHT2 and PWHT3a induced microstructural differences which might influence creep properties. SEM/optical micrographs (Figure 72 through Figure 74) taken from failed specimens did not show any difference in failure location and mechanism between PWHT2 (discussed above) and PWHT3a welded HA282 specimens.

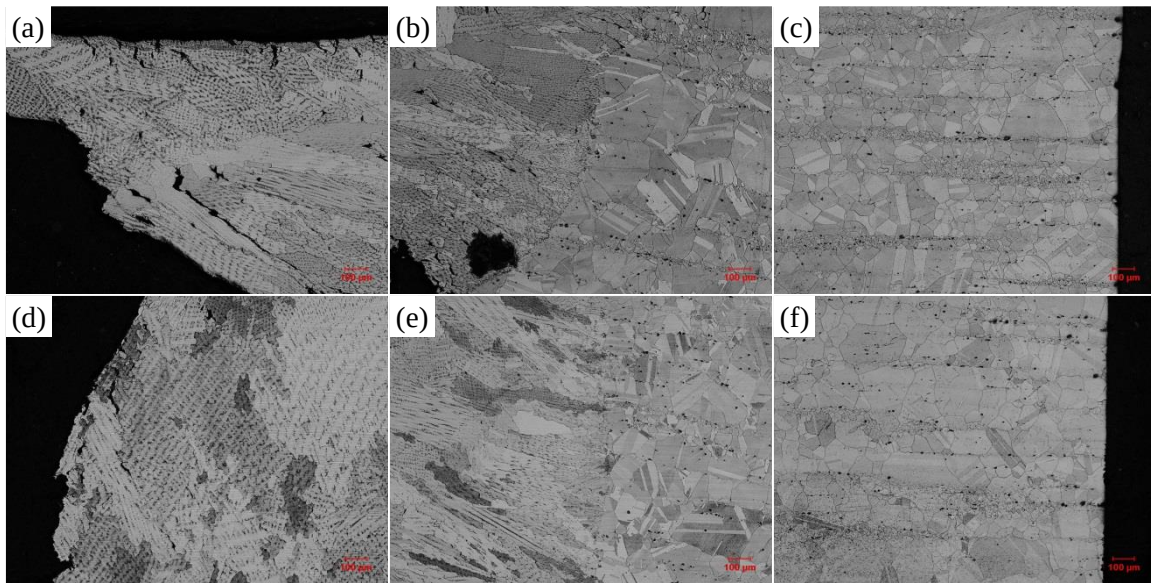


Figure 72: Optical micrographs taken from failed specimens of PWHT3a welded HA282 material, creep tested at: (a-c) 1600°F, 25ksi, and (d-f) 1700°F, 15ksi. Left to right – fracture (weld) region, (weld-base metal) transition zone, base metal.

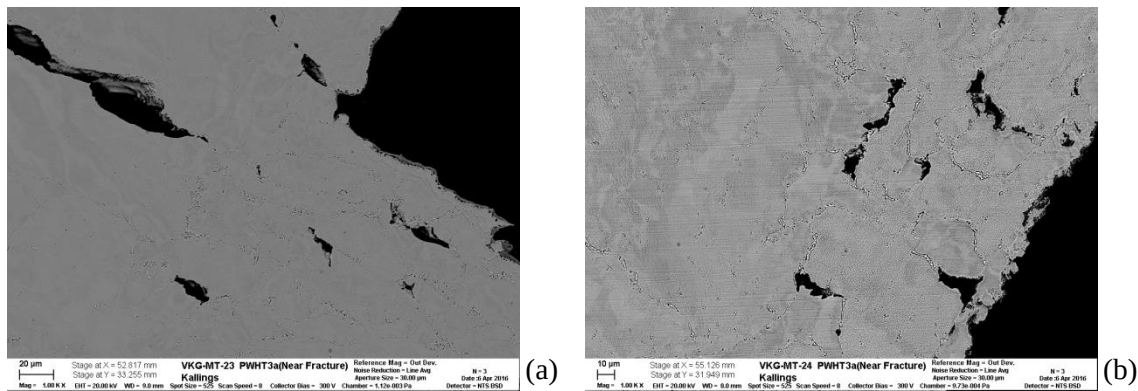


Figure 73: SEM micrographs taken from failed specimens of PWHT3a welded HA282 material, creep tested at: (a) 1600°F, 25ksi, and (b) 1700°F, 15ksi.

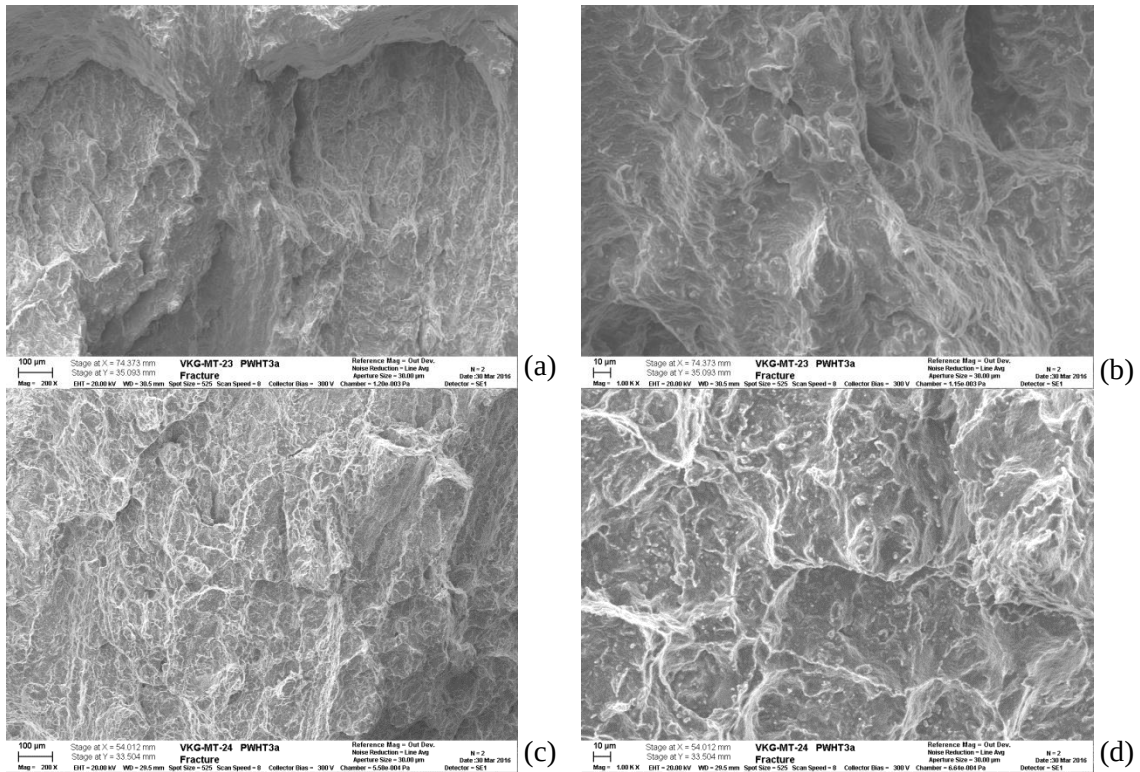


Figure 74: SEM fractographs taken from failed specimens of PWHT3a welded HA282 material, creep tested at: (a-b) 1600°F, 25 ksi, and (c-d) 1700°F, 15 ksi.

Creep specimens: 100% weldment (parallel vs. transverse) HA282

Combination of SEM/optical micrographs and fractographs (shown in Figure 75 through Figure 77) of failed specimens of 100% weldment material suggested inter-granular failure mechanism. Additionally, failure locations and mechanisms did not appear to change at different temperatures, as well as between transverse and parallel orientations.

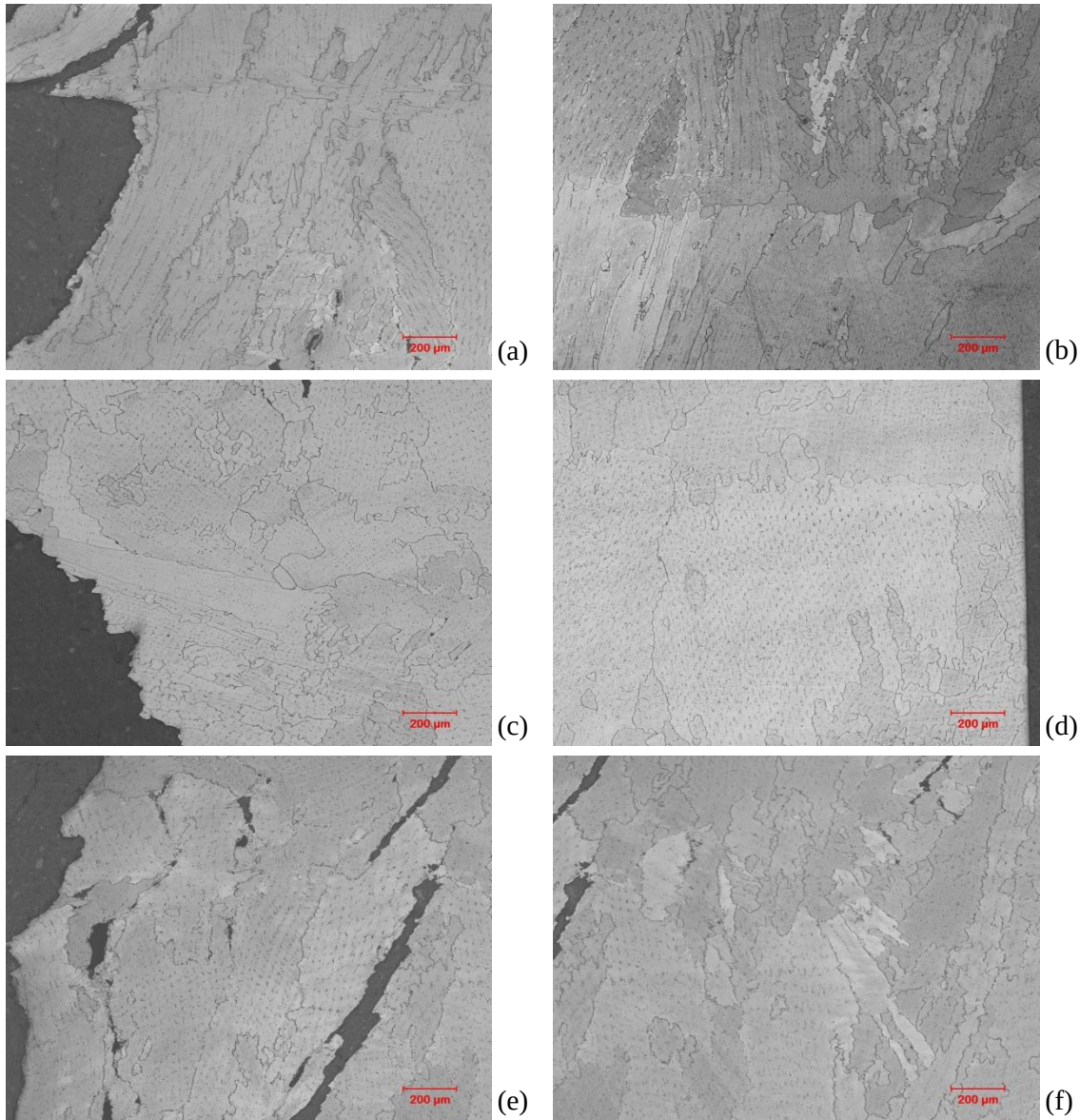


Figure 75: Optical micrographs taken from failed specimens of 100% weldment HA282 material, creep tested at: (a) parallel, 1600°F, 25ksi, (c-d) transverse, 1600°F, 25ksi, and (e-f) transverse, 1700°F, 15ksi.

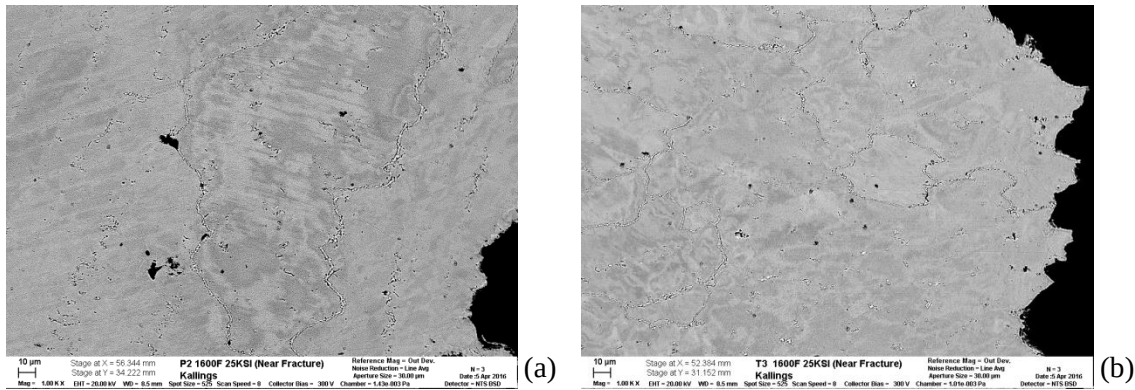


Figure 76: SEM micrographs taken from failed specimens of 100% weldment HA282 material, creep tested at: (a) parallel, 1600°F, 25ksi, and (b) transverse, 1600°F, 25ksi.

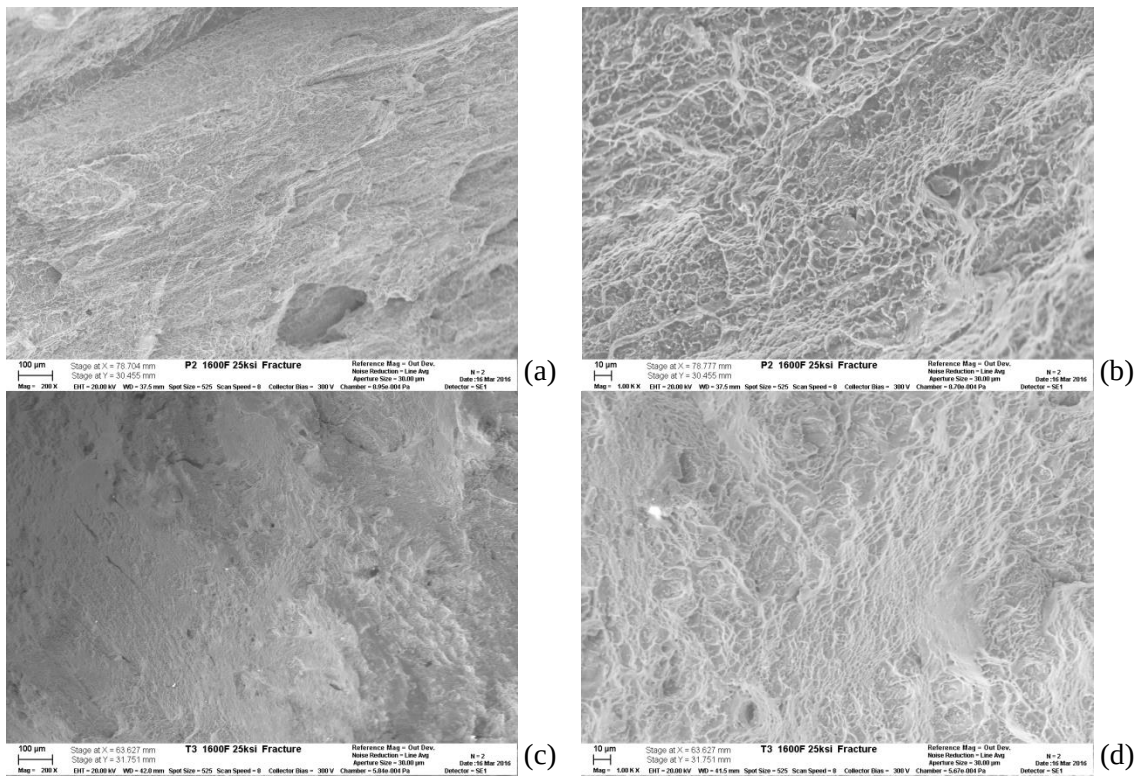


Figure 77: SEM fractographs taken from failed specimens of 100% weldment HA282 material, creep tested at: (a-b) parallel, 1600°F, 25ksi, and (c-d) transverse, 1600°F, 25ksi.

2.3.5. Microstructure changes and damage evolution in creep

Detailed characterization was performed on creep specimens to isolate key microstructural features that led to difference in creep properties in welded HA282. In terms of bulk strength, intragranular γ' size, morphology, and fraction have great contribution. Therefore, characterization was focused on γ' evolution with temperature/time and creep stress effect on γ' in base metal and weld region. In terms of lower creep ductility and intergranular failure in welded HA282, characterization was focused on difference of grain boundary microstructure in base metal vs. weld region and grain boundary damage evolution. To clarify the creep deformation mechanisms, dislocation structure of selected specimens was characterized as well. Creep specimens at different temperature, stress levels, and creep strain were selected, including a pair of interrupted and ruptured specimens at the same test condition. Their conditions are listed in Table 14.

Table 14: Welded HA282 creep specimens selected for post-creep microstructure characterization.

Specimen ID	Condition	Temperature, °F	Stress, ksi	Time, hrs
Test 2	Welded + PWHT2	1500	25	1606 (ruptured)
Test 3	Welded + PWHT2	1500	15	7884 (interrupted)
Test 5	Welded + PWHT2	1600	15	1678 (ruptured)
Test 7	Welded + PWHT2	1700	10	200 (interrupted)
Test 9	Welded + PWHT2	1700	8	1000 (interrupted)
Test 10	Welded + PWHT2	1700	8	1637 (ruptured)

Intragranular γ'

To examine the effect of applied stress and creep deformation on γ' , size and area fraction of γ' precipitates were quantified for the specimens of Test 5 and 9. The micrographs of the creep and static exposed specimens are compared in Figure 78 for Test 5, and in Figure 79 for Test 9. Unlike cuboidal γ' commonly seen after static exposure, the γ' in creep specimens all showed sheared morphologies along with slip bands throughout the matrix. The degree of shearing varied from grain to grain, depending on the local crystal orientation with respect to the applied load. Table 15 compares the γ' sizes and area fractions for Test 5 and 9 specimens. Base metal and weld zone show similar values. As will be further described in Section Task 3, there is little difference in γ' coarsening kinetics between the creep and static exposed specimens. Stress effect on γ' coarsening is thus not evident.

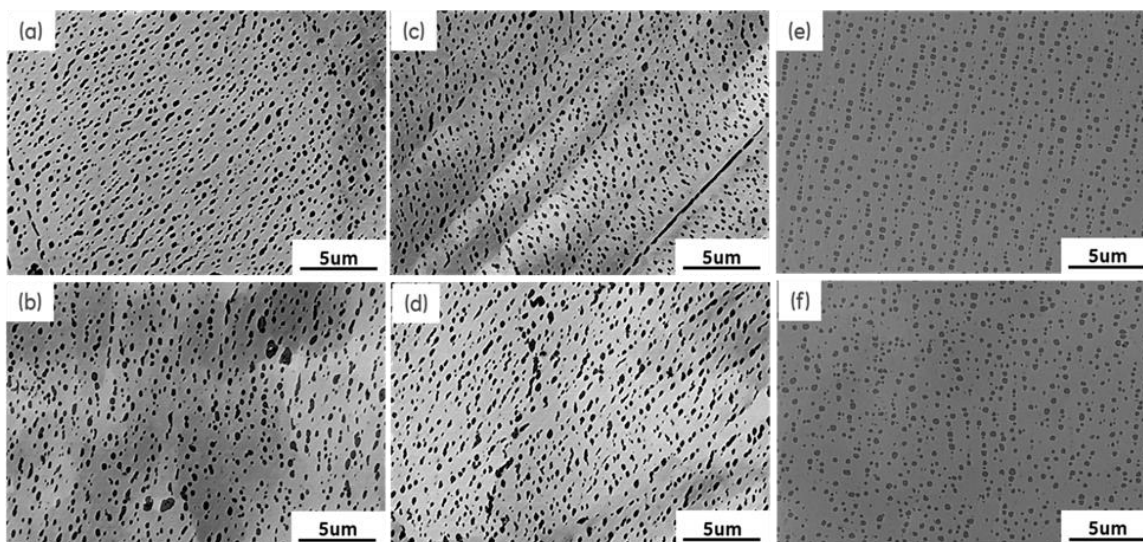


Figure 78: BSE images of precipitates (a, c) in base metal of Test 5; (b, d) in weld zone of Test 5; (e) the base metal and (f) the weld zone of 1600°F 1000 hrs static exposed sample.

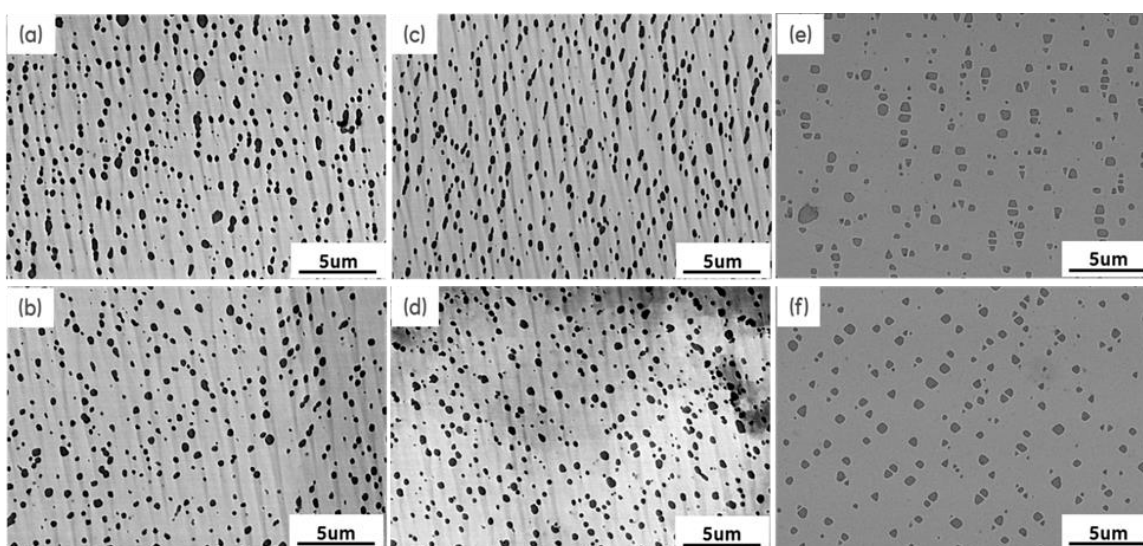


Figure 79: BSE images of precipitates (a, c) in base metal of Test 9; (b, d) in weld zone of Test 9; (e) the base metal and (f) the weld zone of 1700°F 1000 hrs static exposed sample.

Table 15: Quantified γ' size and area fraction. Average and standard deviation were obtained from four images.

Specimen ID	Area%		Equivalent spherical diameter	
	Average	Std. dev.	Average, nm	Std. dev., nm
Test 5, Base metal	14.5	0.9	295.7	10.9
Test 5, Weld zone	15.2	0.7	314.1	18.8
Test 9, Base metal	11.8	0.5	422.7	10.4
Test 9, Weld zone	10.9	0.6	396.1	18.5

Creep cavities

Creep cavitation has been characterized for specimens of Test 2, 5, 9, 10. It was noticed that carbides and cavities were best differentiated in the secondary electron (SE) imaging condition, in which a cavity would appear in a dark contrast and with a bright edge. For each specimen, multiple SE images were taken at random locations. The number density and size of creep cavitation were compared qualitatively in base metal and weld zone.

Figure 80 compares creep cavities in base metal and weld zone for each specimen. The cavities are marked by red circles. In most specimens including Test 2, 5, 10, higher number density and larger size of cavities were seen in the weld zone than in the base metal region. In specimen Test 9, similar density of creep cavitation was observed in base metal and weld zone, but the cavities were relatively larger in the weld zone. The majority of the cavities were found along grain boundaries, with only very few seen in the grain interior in the base metal region. Furthermore, the difference between Test 9 (interrupted at 1000 hours) and Test 10 (ruptured at 1637 hours), which had the same temperature and applied stress, suggests that the cavities may have evolved faster in the weld zone in the later stage of creep.

A closer examination of randomly selected ten cavitation locations in weld zone revealed that almost all the cavities was located at grain boundaries. Similarly, in base metal majority of the cavities was found at grain boundaries, with only a few inside grains. Cavitation inside carbide was occasionally seen. Some examples are shown in Figure 81 and Figure 82, such as grain boundary cavities (Figure 81a, d, e, h), intra-granular cavities (Figure 81c, f, g), and cavitation inside carbide particles (Figure 81b).

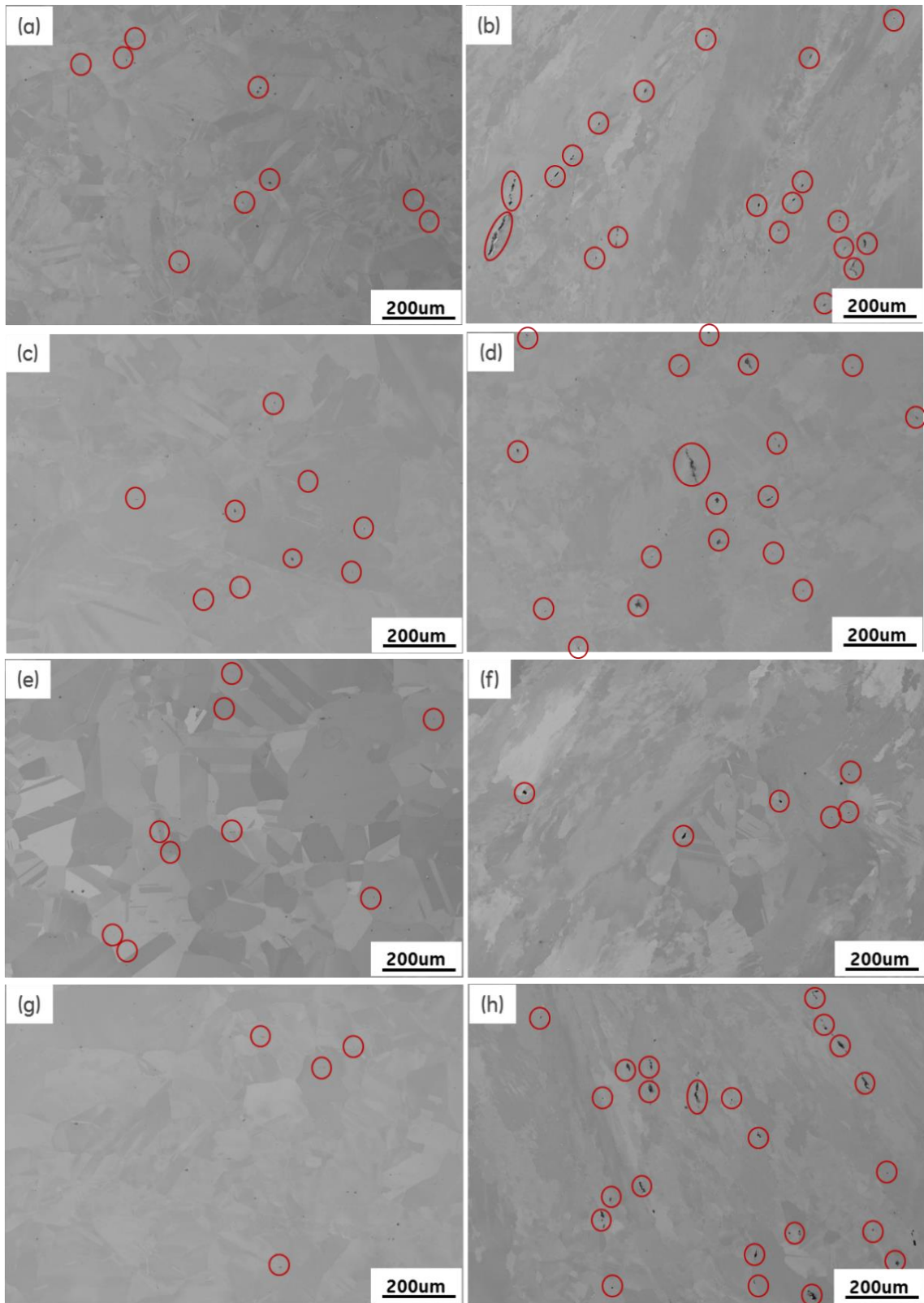


Figure 80: SE images highlighting creep cavities by red circles (left: base metal, right: weld zone) in (a, b) Test 2, (c, d) Test 5, (e, f) Test 9, and (g, h) Test 10.

Grain boundary γ' denuded structure

A major difference in grain boundary microstructure between weld zone and base metal in creep specimens was found to be associated with formation of a narrow band of coarse γ' or γ' -free structure, generally referred to as a γ' denuded zone. Such a structure is much more frequently observed in weld zone than in base metal of the same specimen (see Figure 81 and Figure 82). It is also noted that the γ' denuded structure was not observed in specimens under static heat treatment for the same durations, regardless weld zone or base metal. Clearly formation of γ' denuded structure is induced by localized plastic deformation. In the analyzed creep specimens (Test 2, 5, 9, 10) almost all cavities in weld zone were found next to the γ' denuded structure. Some examples are shown in Figure 82. For grain boundaries without cavities in weld zone, γ' denuded zone was also often seen. It suggests that the formation of γ' denuded zones at grain boundaries may have facilitated creep cavitation development and microcrack propagation in weld zone, as they are generally less creep resistant than the regions strengthened by γ' precipitates or carbides.

Electron backscatter diffraction (EBSD) analysis was performed in multiple regions that contain γ' denuded structure for the specimen Test 10. In each case, grain boundary was found to be at one side, instead of the center, of the γ' denuded structure. One example is shown in Figure 83. Such an asymmetric grain boundary configuration suggests that formation of the γ' denuded structure may be associated with discontinuous γ' coarsening at grain boundary, as found in a similar Ni-base alloys IN740H by Bechetti *et al.*⁵

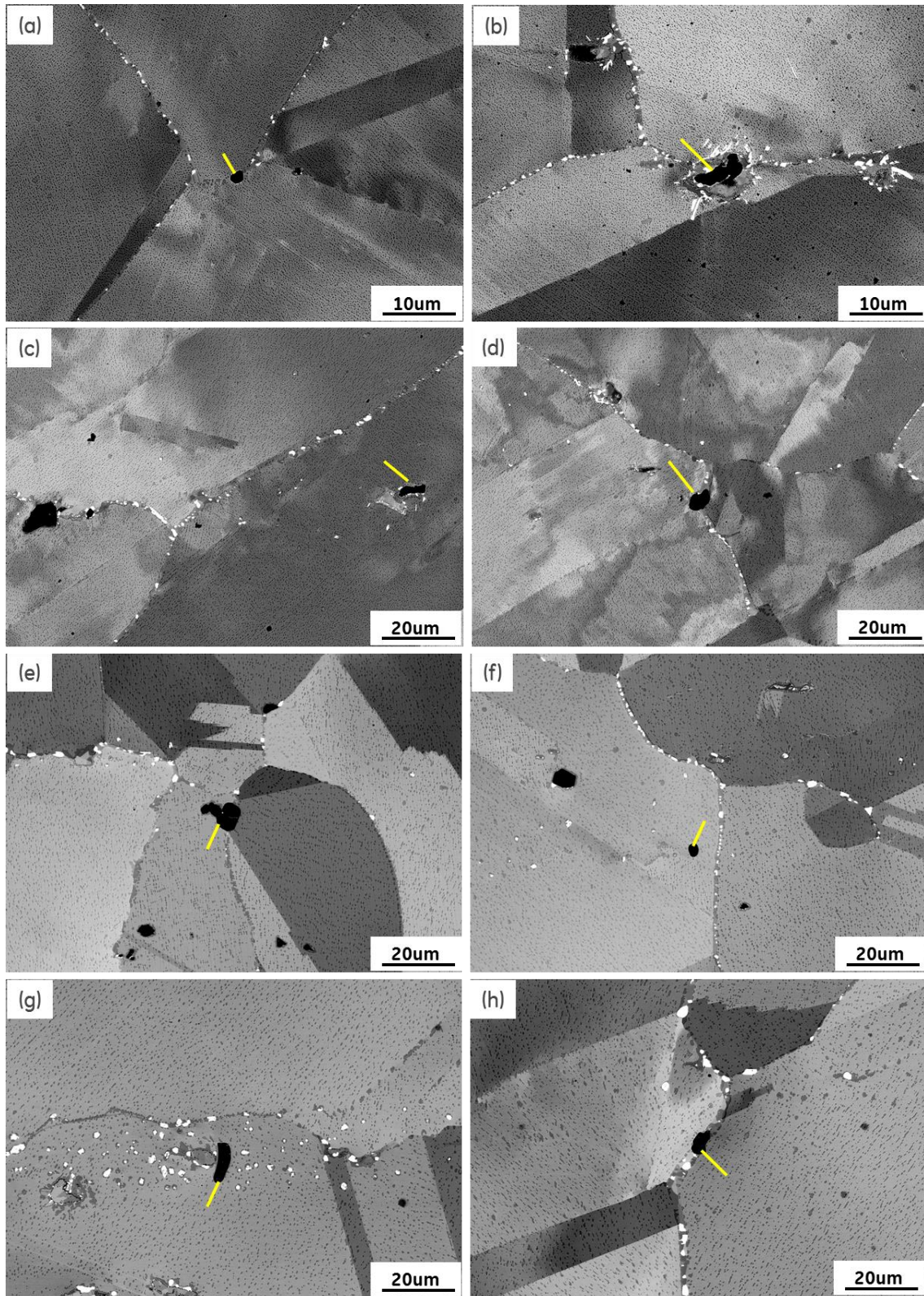


Figure 81: Examples of cavities in base metal region of creep specimens: (a, b) Test 2, (c, d) Test 5, (e, f) Test 9, (g, h) Test 10. Cavities are pointed by yellow line.

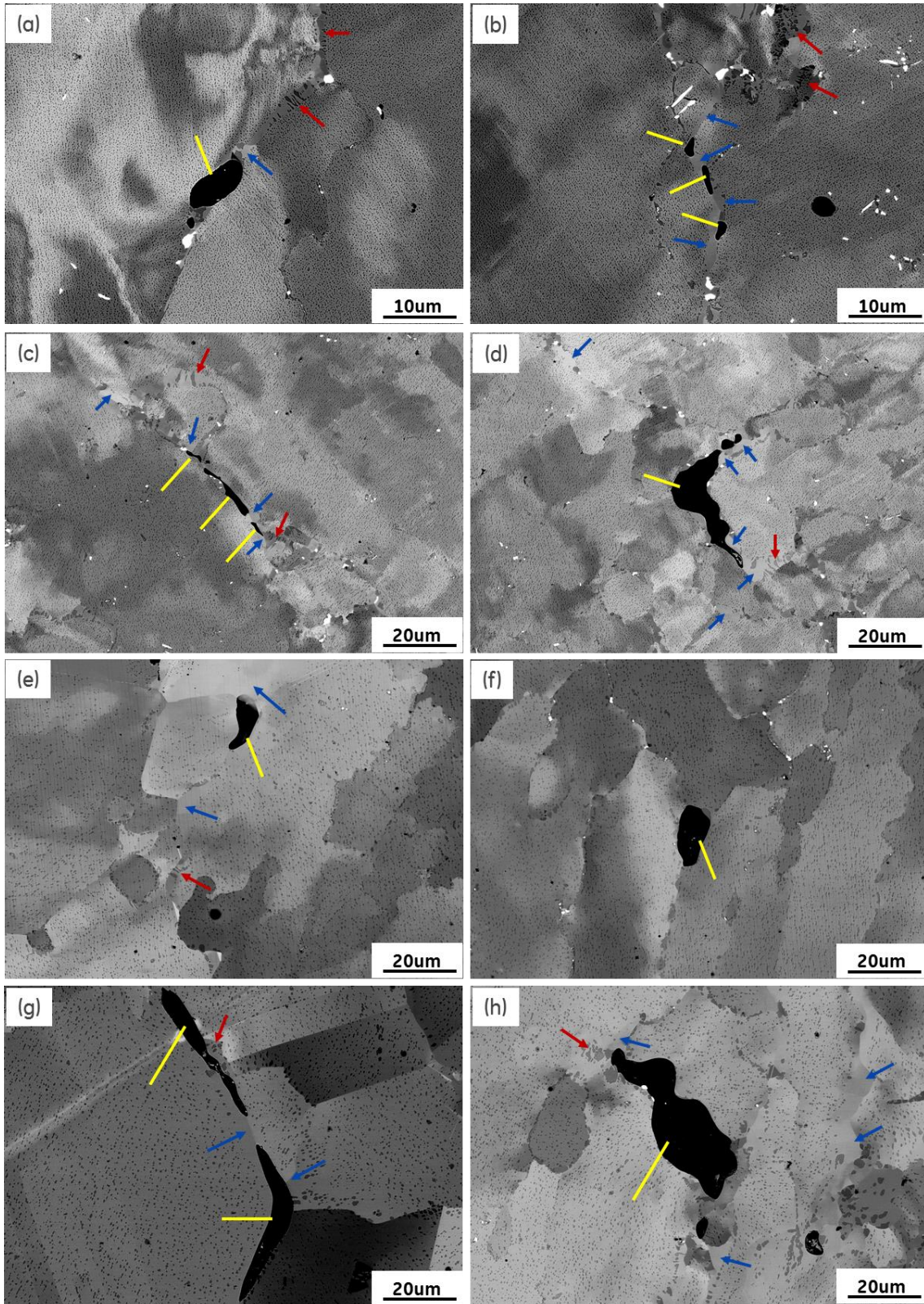


Figure 82: Examples of cavities in weld zone of creep specimens: (a, b) Test 2, (c, d) Test 5, (e, f) Test 9, (g, h) Test 10. Cavities are pointed by yellow line, γ' denuded zones are pointed by blue or red arrow.

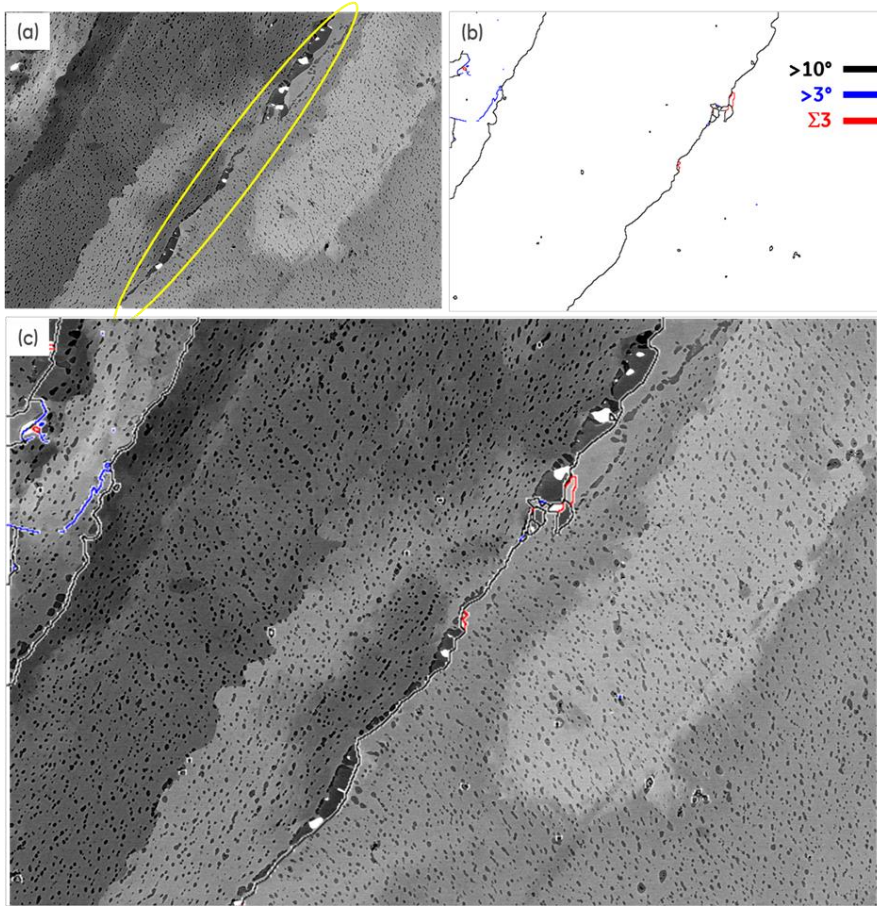


Figure 83: BSE image of γ' denuded zone (a) in weld region of Test 10 and the corresponding grain boundary map (b). The γ' denuded zone is shown in the yellow circle. (c) overlay of (a) and (b).

HA282 weldment (100% weld)

Two creep tested samples, T5 and T2, made of 100% weldment were characterized by SEM. Their conditions are listed in Table 16. The purpose is to compare post-creep microstructure with the welded samples. The 100% weldment specimen T5 is compared to the welded specimen Test 9 in Figure 84, as they were tested at the same temperature, similar creep time, but different stress levels (10ksi vs. 8ksi). It is found that microstructure is very similar in terms of γ' size and the presence of γ' denuded zones and voids at grain boundaries. The only significant difference is that the 100% weldment sample shows a higher fraction of intragranular voids, while the welded sample shows a relatively higher level of voids on grain boundaries. Note that the lower density of grain boundary voids should not be considered as a difference, because T5 was creep ruptured with extensive voids formation right before rupture while Test 9 was creep interrupted.

Another pair for comparison are 100% weldment specimen T2 (Table 16) and welded specimen Test 3 (Table 14). Microstructure features in T2 and Test 3 are displayed in Figure 85. Microstructure is very similar in terms of the presence of γ' denuded zones and voids at grain boundaries. Grain boundary γ' denuded zone is more often seen in Test 3 than T2, probably due to longer creep time in Test 3. In both samples, both intragranular and intergranular cavities were observed, although intergranular cavities still held higher fraction. Since as-weld and weldment materials before PWHT2 do not show intragranular voids, the cavities we observed in post-creep samples are most likely formed during creep deformation. Also, bright-contrast acicular-shape phases are present in addition to the bright-contrast Mo-rich M_6C carbides on grain boundaries, although their size is bigger than the needle phases in the static-exposed specimens. Such phase is not seen in specimen T5 crept at higher temperature but with shorter time.

In summary, comparisons have been made in the post-creep microstructure of welded and weldment specimens. Post-creep welded and weldment specimens show similar microstructure features including γ' size, intergranular voids, γ' denuded zones at grain boundaries. The major difference identified so far is the higher fraction of intragranular voids in the weldment in contrast to very few intragranular voids in the welded samples.

Table 16: 100% weldment creep specimens selected for post-creep microstructure characterization.

Specimen ID	Condition	Temperature, °F	Stress, ksi	Time, hrs
T5	100% weldment + PWHT	1700	10	1100 (ruptured)
T2	100% weldment + PWHT	1500	15	5000 (interrupted)

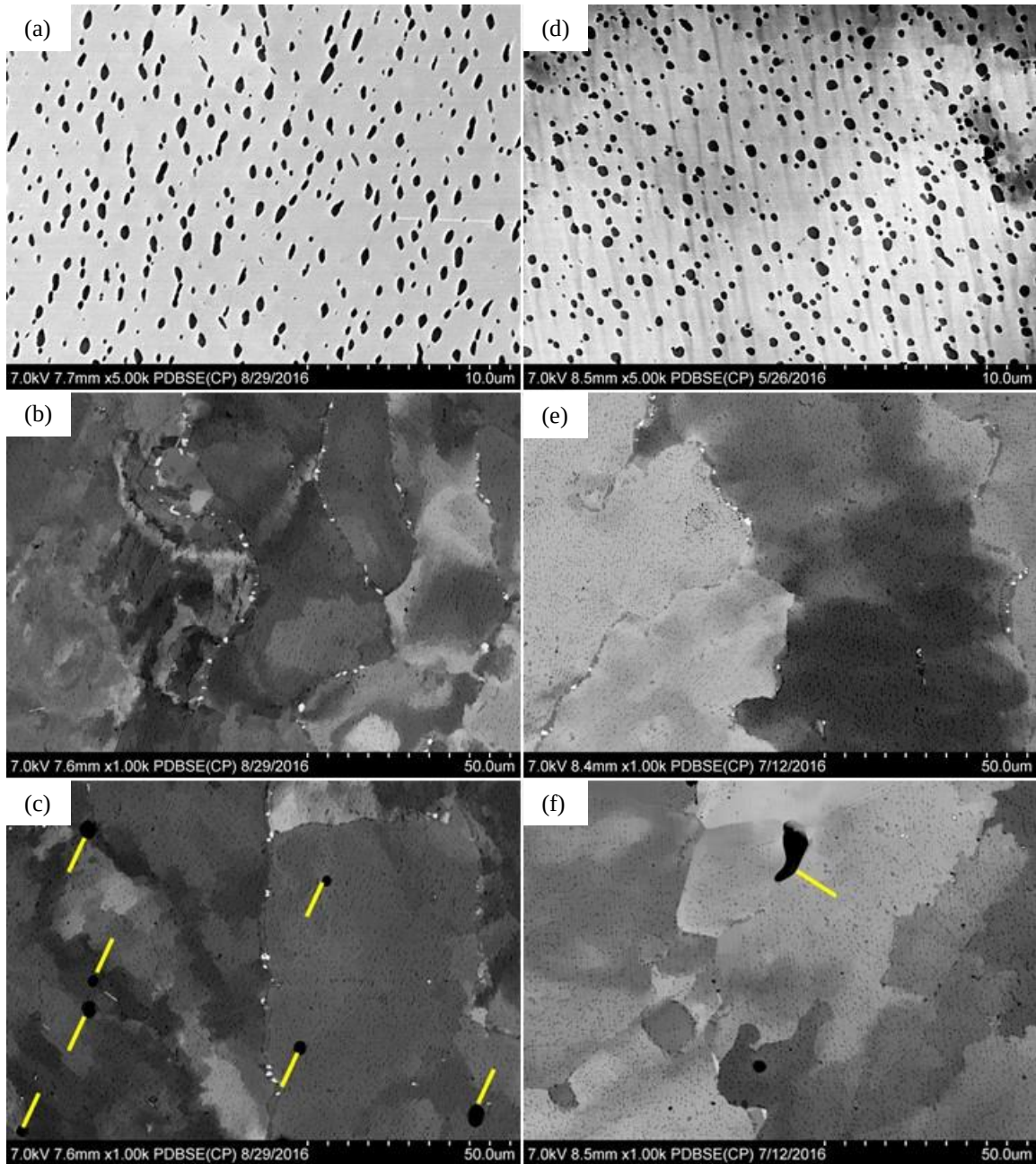


Figure 84: Post-creep microstructure in (a, b, c) 100% weldment specimen T5 and (d, e, f) weld zone of welded HA282 creep specimen Test 9. Yellow lines in (c, f) highlight intergranular and intragranular voids.

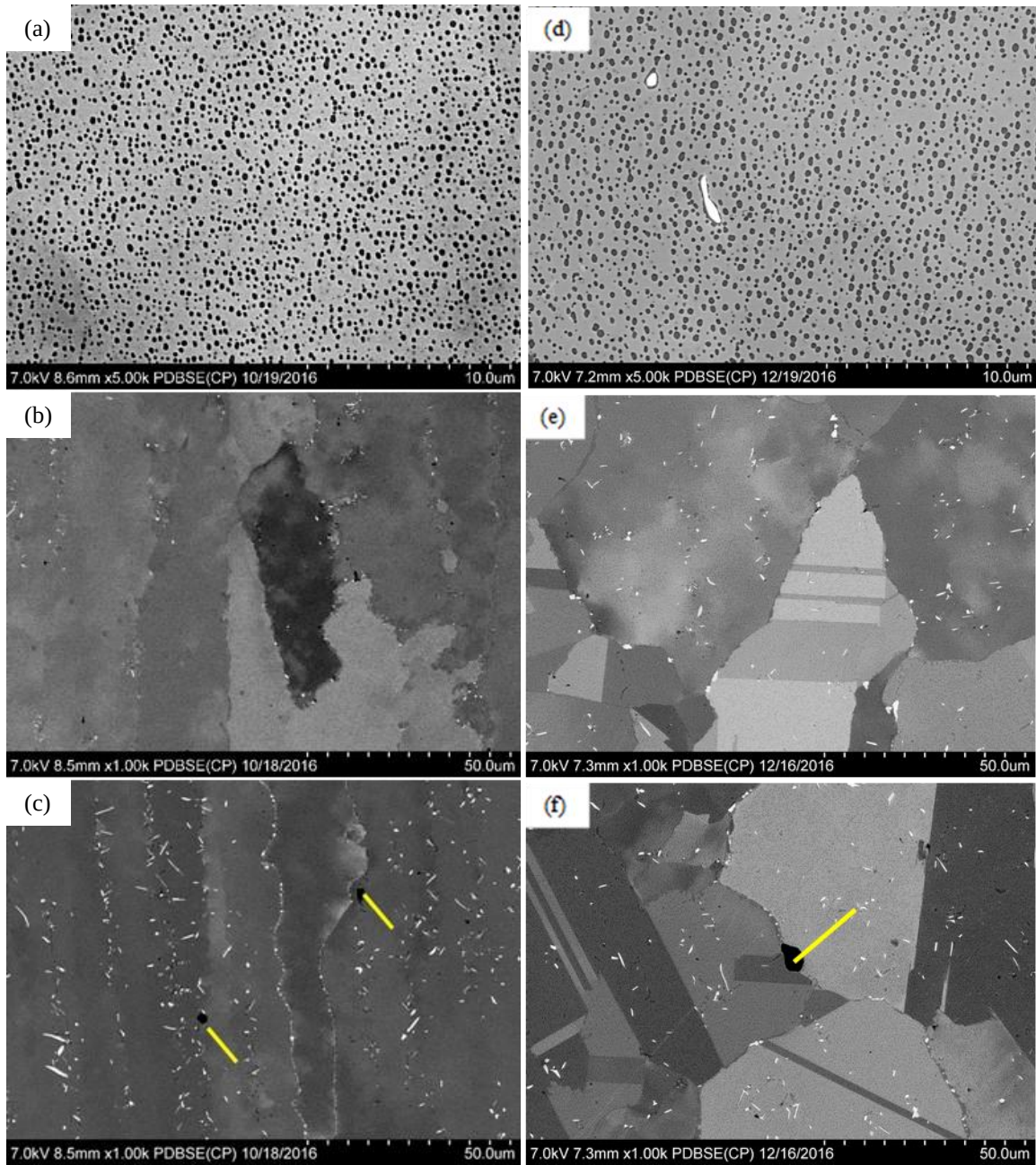


Figure 85: Post-creep microstructure in (a, b, c) 100% weldment specimen T2 and (d, e, f) weld zone of welded HA282 creep specimen Test 3. Yellow lines in (c, f) highlight intergranular and intragranular voids.

Deformation mechanisms and dislocation structures

Transmission electron microscopy (TEM) characterization was performed on creep specimens Test 7 and 9 to determine operative deformation mechanisms and dislocation density in base metal and weld zone. TEM foils were prepared from weld zone far away from rupture surface and from base metal in the gage section, respectively. Figure 86 and Figure 87 display examples of dislocation activities and interactions with γ' precipitates in the two regions. For both Test 7 and 9, a relatively higher dislocation density was observed in weld zone than base metal. In Test 9, similar deformation mechanisms are dominant in both regions, including $\frac{1}{2}\langle 110 \rangle$ dislocations gliding on octahedral slip planes and bypass (looping), dislocation networks around γ' precipitates, dislocation cross slip and climb. Some slip bands or octahedral planar activity were observed as well. In the base metal of Test 7, the dominant deformation mechanisms were dislocation octahedral gliding and $\frac{1}{2}\langle 110 \rangle$ dislocation pairs shearing precipitates, although some dislocation looping around γ' was also seen. Slip bands, dislocation glide and planar activities on octahedral planes are observed. In the weld zone of Test 7, dislocation looping, networks, cross slip and climb appear to be major interactions.

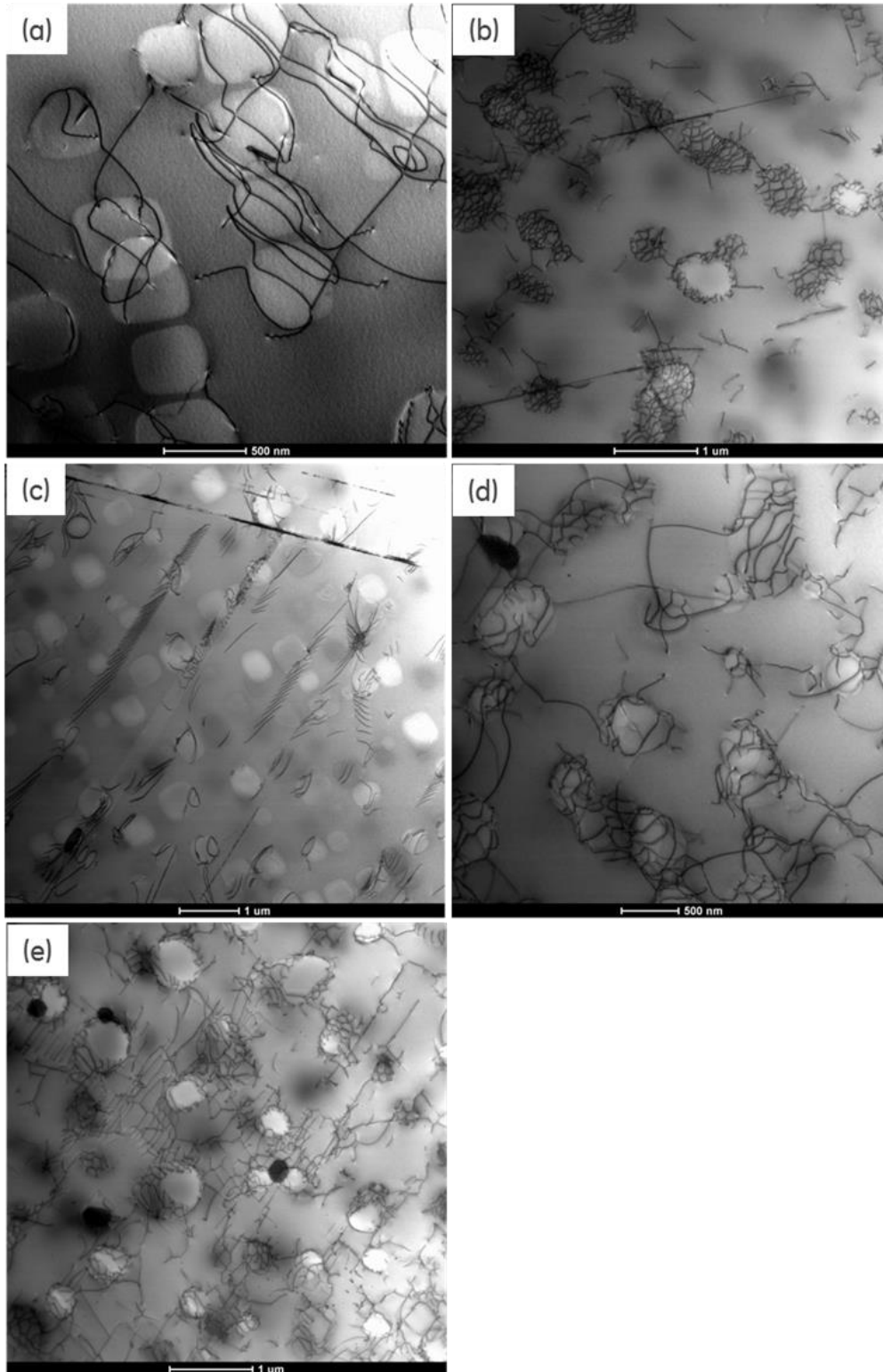


Figure 86: $\frac{1}{2}\langle 110 \rangle$ dislocation activities in specimen Test 9. In base metal: (a) looping and cross slip/climb, (b) dislocation networks around γ' precipitates, (c) octahedral planar activities. In weld zone: (d) looping and cross slip/climb, (e) dislocation networks around γ' precipitates.

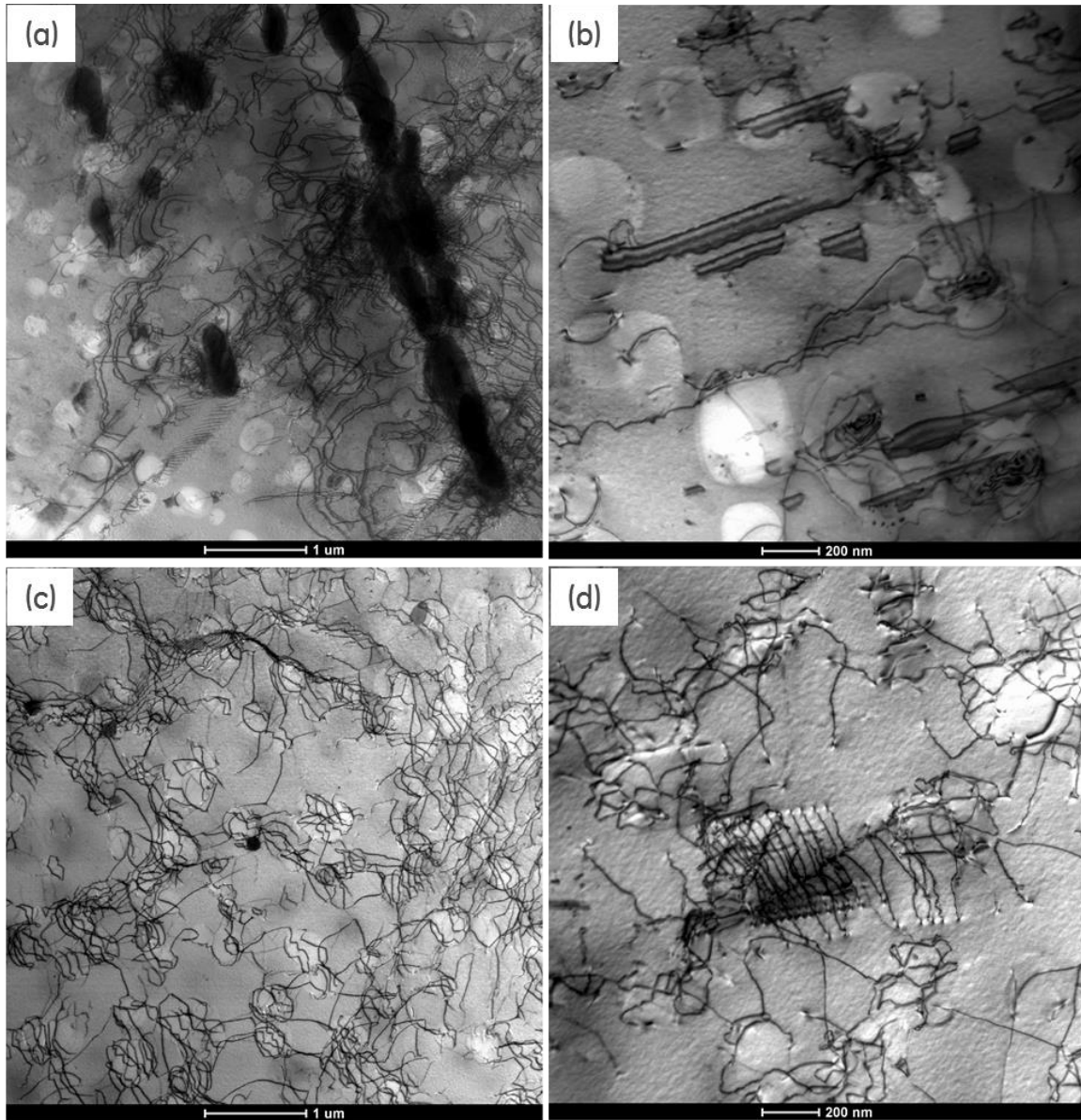


Figure 87: Dislocation activities in specimen Test 7. In base metal: (a) $\frac{1}{2}\langle 110 \rangle$ octahedral planar activities, slip bands, $\frac{1}{2}\langle 110 \rangle$ dislocation pairs, (b) planar activity with dissociated dislocations. In weld zone: (c) $\frac{1}{2}\langle 110 \rangle$ dislocation looping, networks, cross slip/climb, (d) slip bands.

Task 3: Modeling of the Microstructure of a Superalloy Weld

This task develops a modeling linkage between processing and microstructures, and provides inputs to mechanical property modeling, including mesoscale crystal plasticity modeling in Task 4 and continuum creep modeling in Task 5.

Incorporating microstructural changes introduced by welding process is a fundamental step for generalizing our creep model from base metal to a welded structure. The welding not only changes average properties, but also introduces new spatial non-uniformities across fusion zone and base metal, and between dendrite cores and interdendritic regions in the fusion zone. These effects evolve with the underlying microstructures over time during post-weld heat treatment and creep testing. Modeling the initial (as-weld) and the evolution of microstructure serves as an important input to predicting long-term creep behaviors. Overall, Task 3 attempts to address the following three questions:

1. What are the microstructural changes in the weldment, relative to the base metal of HA282, that can impact mechanical properties?
2. How are they affected by post-weld heat treatment?
3. How do they continue to evolve during creep?

3.1. Modeling solidification path of HA282 weldment

The primary microstructural factors under consideration here are the weld-zone microsegregation and interdendritic phases developed during solidification at the end of welding. The compositional non-uniformity then affects the strengthening gamma prime (γ') precipitates. Minor phases formed in the interdendritic regions, particularly carbides and topologically closed packed (TCP) phases, due to segregation of refractory elements, are of importance. Average grain size is coarse in base metal HA282 and does not contribute significantly to the strength. The grains in welded zone in HA282 are generally coarser than that in the base metal⁶, due to the loss of γ' pinning in post-weld heat treatments. Once formed, the grain structure is constrained by γ' microstructure and remains stable with time in creep. The difference of the initial grain size between weld zone and base metal, however, has been taken into account in the constitutive creep model of Task 5.

3.1.1. Equilibrium and Scheil simulations

We started with a thermodynamic calculation for equilibrium phases in HA282, with the nominal HA282 composition given in Table 3. ThermoCalc software⁷ and TTNi7 Ni-alloy thermodynamic database² were used. As shown in Figure 88, the predicted liquidus and solidus temperatures of HA282 are respectively 2453°F and 2295°F. The first solid phase to appear is gamma (γ) phase, which is the

primary solid (matrix) phase of the alloy. The strengthening phase gamma prime (γ') solvus is around 1840°F. At temperatures below 1650°F, some TCP phases, sigma (σ) and mu (μ) (rich in Cr, Mo, Ni, Co), were predicted to form. Minor phases including (Ti-rich, Mo, Cr-) MC, (Mo, Ni, Cr-) M₆C, (Cr, Mo-) M₂₃C₆ carbide phases, and (Mo, Cr-) M₃B₂ boride phase were predicted to form as temperature decreased.

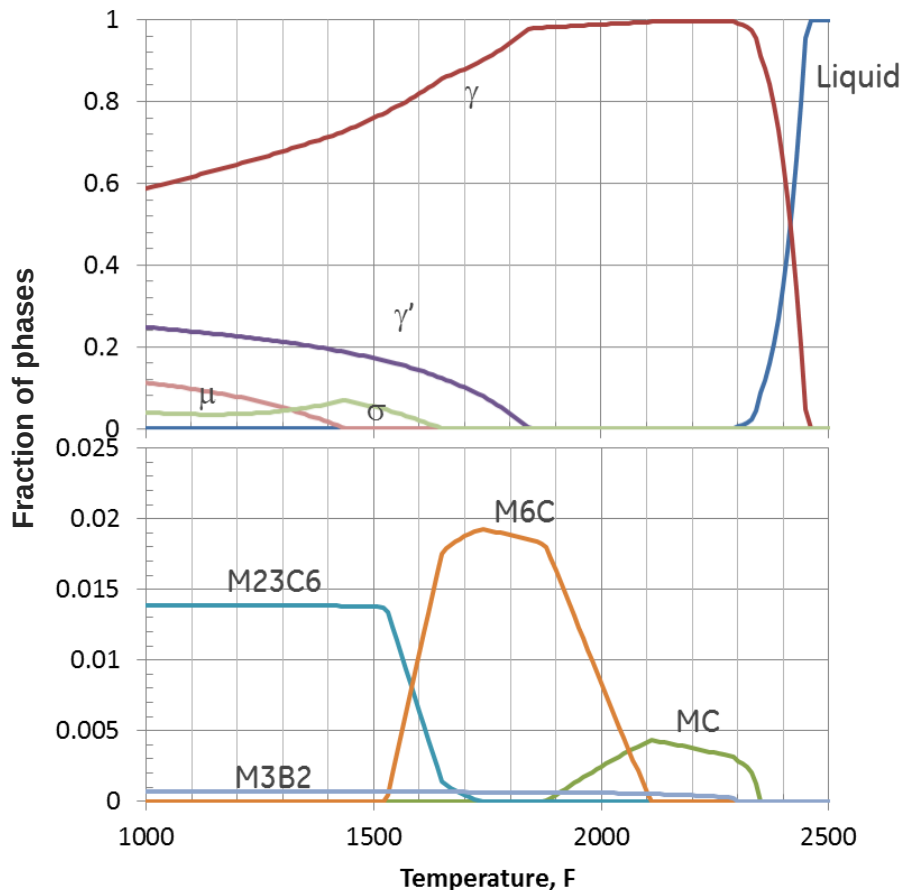


Figure 88: Calculated equilibrium phases in HA282 alloy.

In comparison, because of the rapid cooling during welding, the actual formation of solid phases does not follow the equilibrium path. With the same composition, a Scheil simulation showed that the overall solidification range, including the (non-equilibrium) solidus, was extended to a significantly lower temperature at about 2130°F (Figure 89), as compared with the equilibrium solidus 2295°F. The calculation of the solidification path here used a traditional Scheil model, in which partition of elements between liquid and solid follows phase diagram during cooling, implying an equilibrium condition at solid-liquid interface at all temperatures. Diffusion in liquid is infinite, and is forbidden in solid phases. Fast solid diffusion of interstitial elements, C and B, however, was enabled. Figure 89 labels solid phases that form during solidification. The initial solid phases, γ and

MC, M₆C carbides, formed in a same sequence as in Figure 89 but were extended to lower temperatures. γ' was found to form after solidification completes, thus reduced the possibility of forming γ' eutectic in the interdendritic regions. Some phases, however, such as σ phase, was predicted to form at a higher temperature than in the equilibrium condition. This was found to correlate to enrichment of elements such as Mo in the liquid phase due to the solidification partitioning behavior.

Figure 90 shows a similar Scheil calculation but with the actual HA282 weld wire composition in Table 3. Compared with Figure 89, the sequence of solid phase formation remained the same, except that the liquidus temperature was slightly increased from 2453°F to 2463°F. The difference between the two compositions (Table 3) is mainly on the trace elements Fe, Mn, and Si. Thermodynamic equilibrium phase calculations with alternated Fe, Mn, and Si compositions suggested that the increase in the liquidus was due to both Mn and Si, but not Fe (Figure 91).

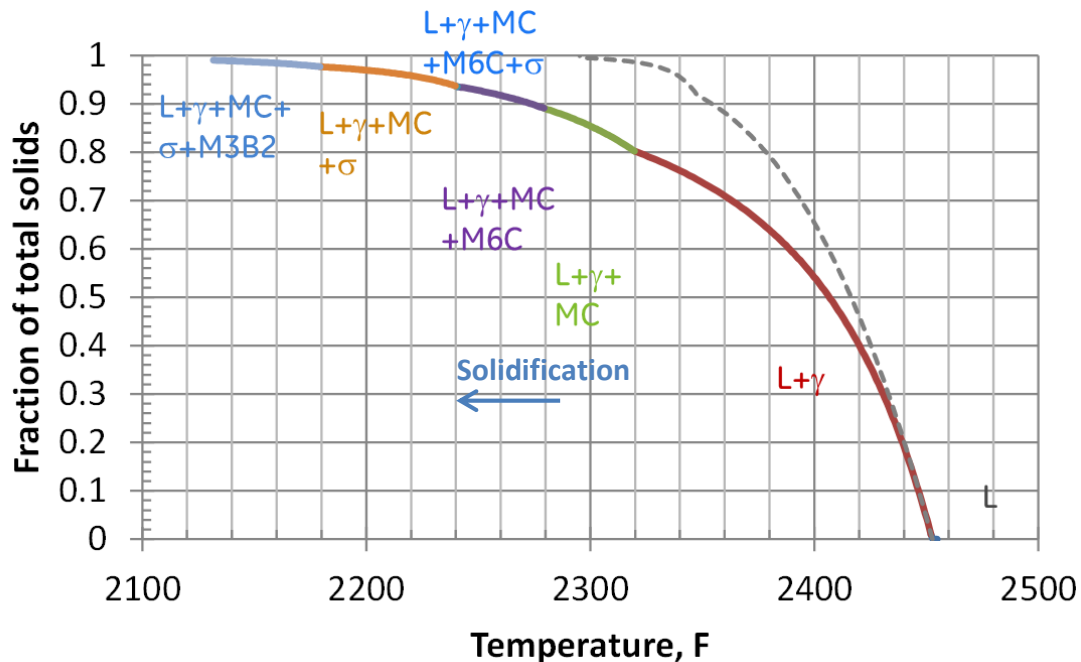


Figure 89: Calculated phase formation during solidification in HA282 alloy. The nominal HA282 composition in Table 3 is used. The dashed line is the equilibrium fraction of solids, close to the γ line in Figure 88.

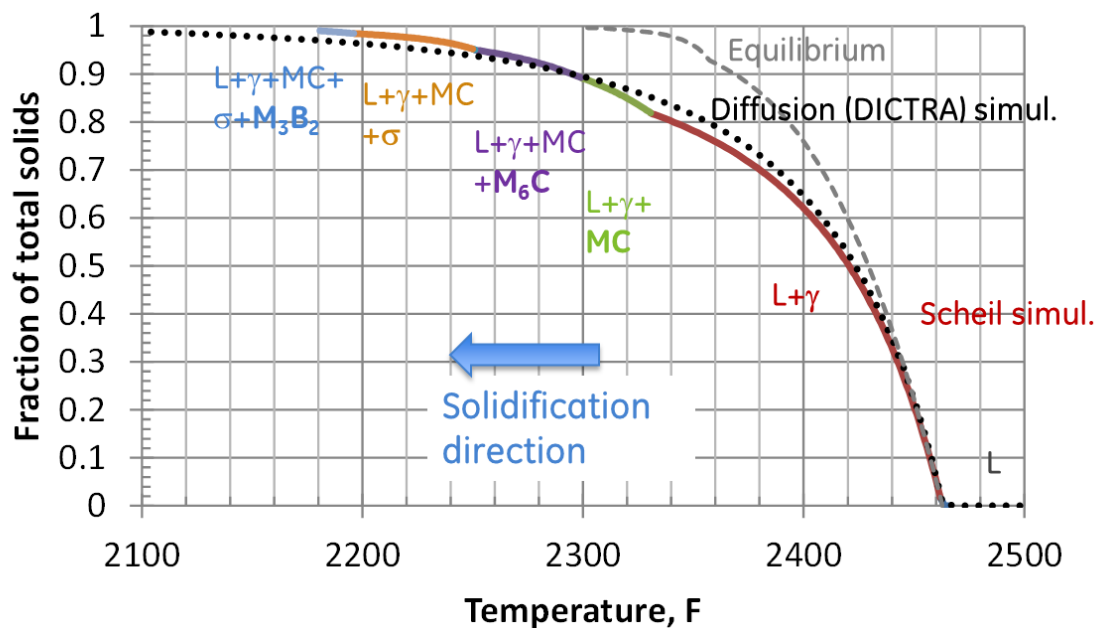


Figure 90: Calculated phase formation during solidification in HA282 alloy. The HA282 weld wire composition in Table 3 is used. The dashed line is the equilibrium fraction of solids, close to the γ line in Figure 88. The dotted line is from DICTRA simulation.

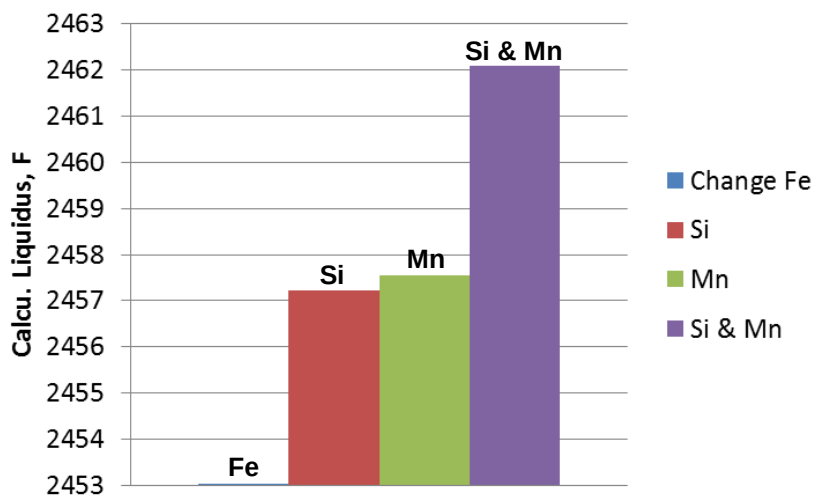


Figure 91: Calculated liquidus temperature by changing Mn, Si, and Fe respectively from the nominal composition to the wire composition in Table 3.

Alloy element partition was derived from the calculation results. Figure 92 shows the composition of each element in the remained liquid and in the major solid phase γ , versus temperature along Scheil solidification path. It can be seen that Ni, Al, Fe are preferentially partitioned to the solid (γ) phase in HA282, while Mn, Mo, Si, Ti, B, C are partitioned to the liquid phase. Cr and Co do not show a strong preference. These partitioning behaviors influence phase formation during solidification. For example, as discussed earlier, segregation of Mo, Ti, and C in the remained liquid would promote formation of σ phase and Ti-rich MC carbide, respectively.

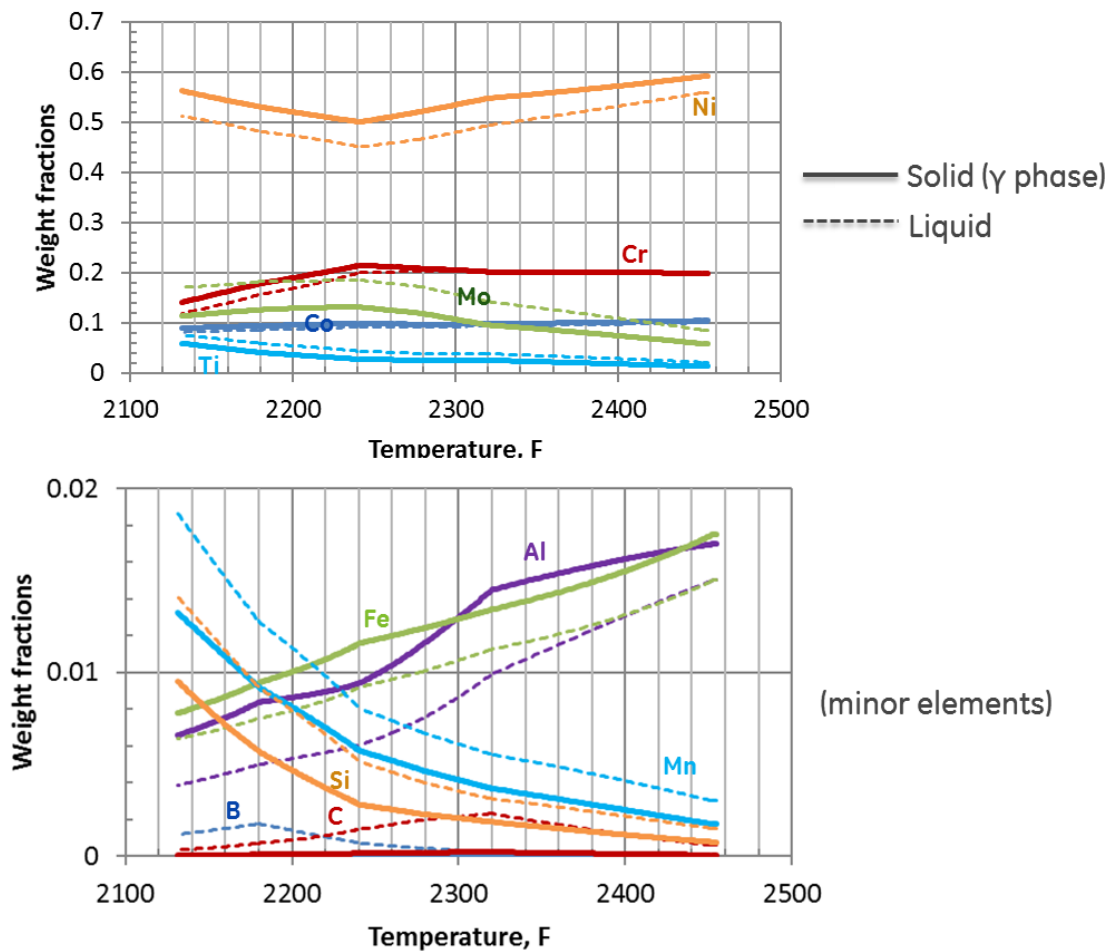


Figure 92: Calculated element partition during solidification in HA282: (a) and (c) in liquid, (b) and (d) in γ (major solid matrix phase).

3.1.2. DICTRA simulation

In general, Scheil model considers that the solid-liquid boundary moves much faster than diffusion in solid. Thus except for fast-diffusing interstitial elements such as C, B, the solid formed behind the solid-liquid front is practically frozen. Additionally, the alloy elements' partition at the solid-liquid boundary is considered always at local equilibrium. As a result, the solidification path predicted by the Scheil model is independent of cooling rate and interdendritic spacing. However, rapid cooling in welding processes typically produces fine dendrite arm spacing (Figure 33a). Consider that diffusion time reduces with the square of the diffusion distance (here, the dendrite arm spacing), the effect of back diffusion in solid at later stages of welding needed some further assessment.

To study this effect, we treated the solidification process as a moving boundary diffusion problem with DICTRA simulation^{8,9}. In order to make a close comparison with the Scheil predictions, the actual HA282 wire composition (in Table 3) excluding the trace amount B was used. The face-centered cubic γ is the major solid phase. From the Scheil prediction (Figure 90), γ is the first solid phase to form and remains as the only solid phase up to 80% of the total solid formation until the second solid phase (the MC carbide) appears. Therefore, we only considered two phase, liquid and γ , in the calculation. Furthermore, by considering only diffusion along the principal direction normal to the solid-liquid moving boundary, the calculation was reduced to a one-dimensional (1D) problem, as shown in Figure 93. The dendrite arm spacing depends on solidification cooling rate (Figure 94). Using the primary dendrite spacing in the as-weld HA282 specimen (Figure 33a), in the calculation we took the cooling rate at about 90°F/s and the secondary dendrite spacing approximately as 8 μm .

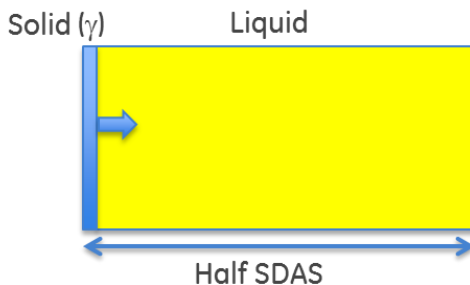


Figure 93: A schematic 1-D configuration of simulation. The solid phase (γ) grows into the liquid phase during solidification. The diffusion distance is measured from dendrite core (left side) to the center of interdendritic region (right side).

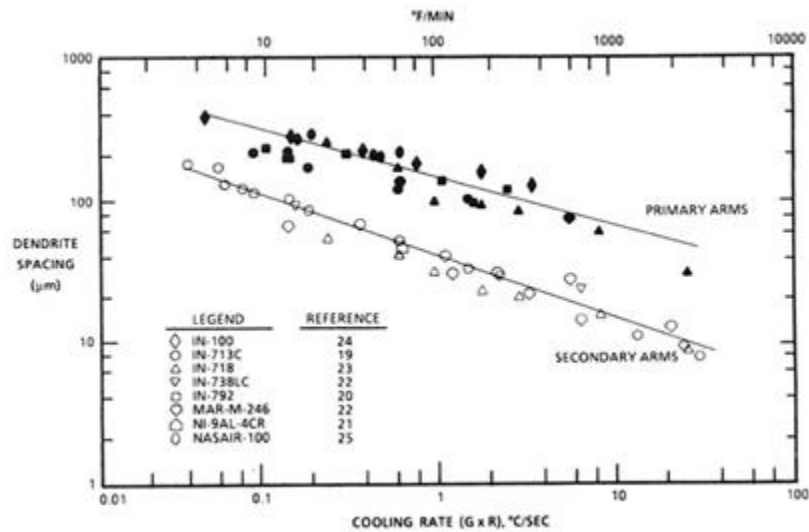


Figure 94: Dependence of dendrite arm spacing on cooling rate for a wide range of nickel-based superalloys (from Ref.¹⁰).

Figure 95 shows the simulated composition profiles at 10%, 50%, 90%, and 100% of solidification, in which a constant cooling rate of 90°F/s was applied. The solid-liquid boundary is indicated by the sharp change in each composition profile. The liquid profiles, at the right side of the boundary, maintain mostly flat because of the much greater diffusivities in liquid. The change in liquid composition as a result of solid-liquid element partition is clearly seen, and becomes more significant toward the end of solidification. While the liquid composition determines what solid phase will form next, the solid composition (on the left side of the boundary) outlines what forms behind the solid-liquid front, which is eventually retained as a non-uniform distribution of elements, or microsegregation. As discussed earlier, solid diffusion is usually much slower. A close look at the composition profiles in Figure 95 suggests a minor solid diffusion effect on Cr, Ti, and Co, but almost negligible on Mo.

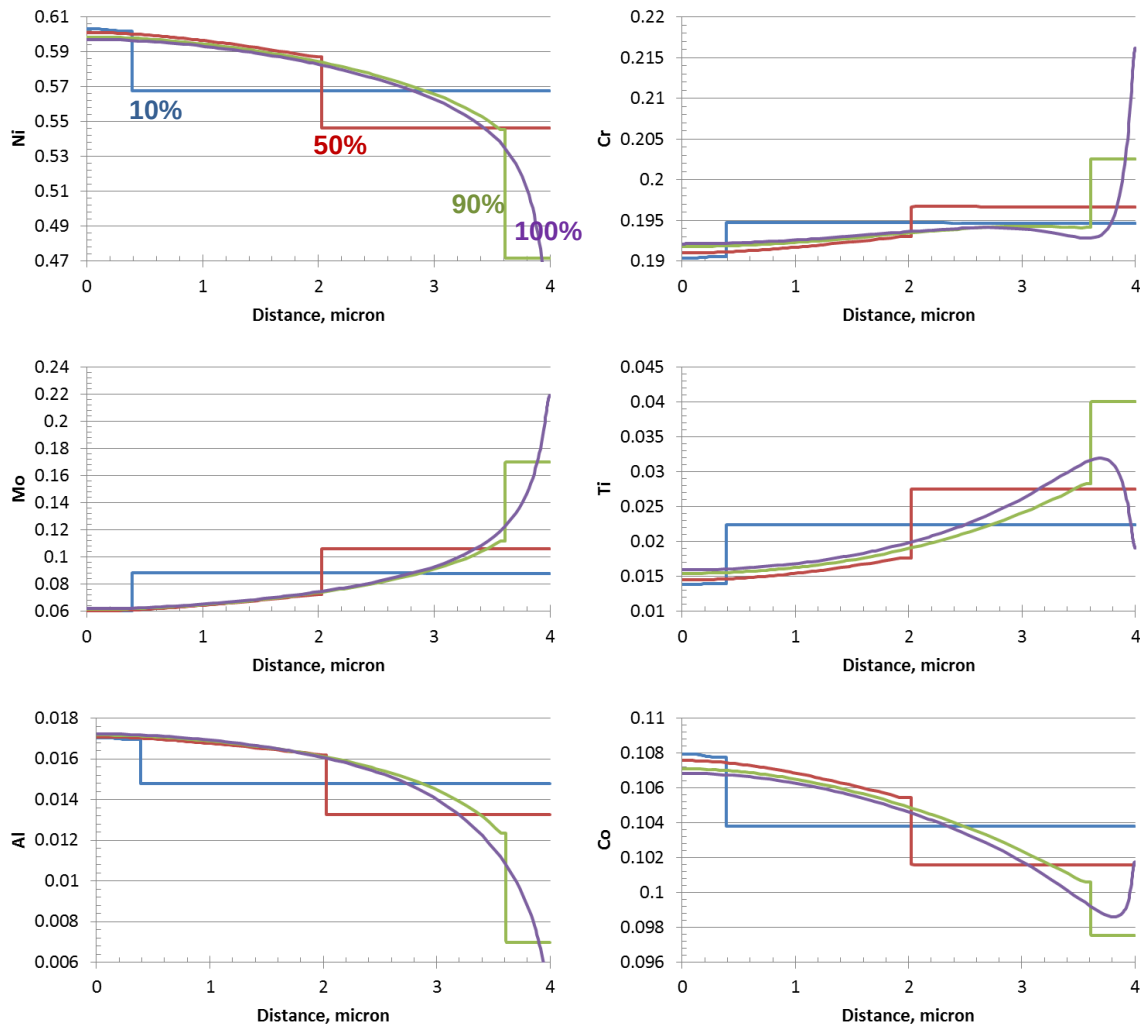


Figure 95: Snapshots of composition profiles at 10, 50, 90, and 100% of solidification. The distance is measured from dendrite core (left) to interdendritic spacing (right), equal to the half of SDAS.

The predicted solid (γ phase) fraction from the diffusion simulation is compared with that from Scheil calculations in Figure 90. Because the same composition was used and that γ phase dominated the solid phase, the predicted solidification curves (including the liquidus temperatures) by the two methods agreed well with each other from the beginning. At a later stage (2330~2420°F), diffusion simulation predicted a slightly higher solid amount than did the Scheil method. Since in this range γ still remained as the only solid phase, the difference was most likely caused by solid diffusion, which was absent in the Scheil model. At even lower temperatures, additional solid phases, such as MC and M₆C carbides, started to form, and the predictions from the two methods further departed. In particular, the completion of solidification (i.e., the solidus temperature) in the diffusion simulation was extended to a lower temperature. Figure 96 compares additionally the cooling rate effect on solidification shown by the diffusion simulation. Here four different cooling rates were applied, with different dendrite

spacing SDAS chosen according to Figure 94: 1.8°F/s (36 μ m), 14.4°F/s (16 μ m), 36°F/s (11.2 μ m), and 90°F/s (8 μ m). It is seen that as the cooling rate increased, the solidification completion temperature was moderately extended toward lower temperatures.

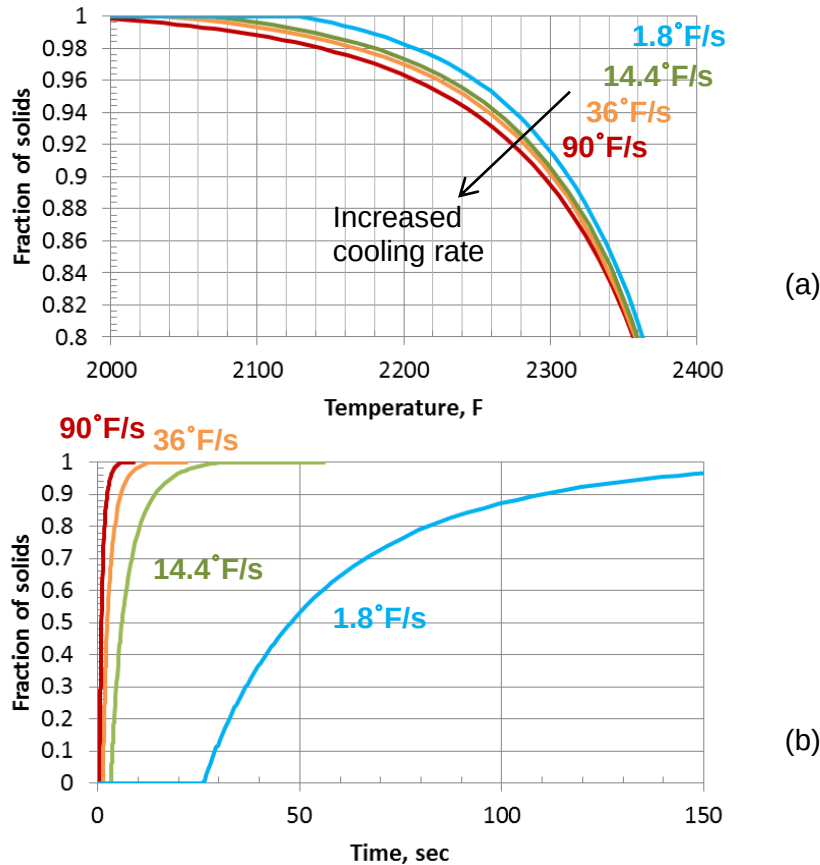


Figure 96: Comparison of cooling rate effect on solidification in diffusion simulation.

3.1.3. Comparison with experimental observation

Matysiak *et al*¹¹ characterized the phases in as-cast HA282 material that was prepared with vacuum induction melting and investment casting. The identified phases are summarized in Table 17. The fact that γ' was observed in dendrite core regions was because the cooling rate from casting was slow. Since Scheil model assumes no back diffusion in solid, the simulated condition is close to the conditions of fast cooling. Thus we shall only make a qualitative comparison with their results. The observed Ti-rich MC carbide and σ phase, and an absence of μ phase, were also predicted by the calculations. $M_{23}C_6$, M_6C , γ' were considered to form by solid-state reaction after solidification¹¹. Since the trace elements S and N are not included in the calculations, carbosulphides and nitrides phase are not predicted by the model.

Table 17: Phase characterization in as-cast HA282 (Ref ¹¹).

Phase identified	Description	Spatial characteristics
γ'	Primary strengthening precipitates, $\text{Ni}_3(\text{Al,Ti})$ type	Dendrite core: spherical Interdendritic: coarser, cuboidal
$(\text{TiMoCr})\text{C}$	Primary carbides (MC)	Preferentially interdendritic
TiN	Primary nitrides (MN)	Homogeneously dispersed
σ	TCP phase, rich in Cr, Mo, Ni Detrimental intermetallic Depletes refractory elements from γ	Globular, Preferentially interdendritic
$(\text{TiMo})_2\text{SC}$	Carbosulphide	Adjacent to σ
Mo,Cr-carbides	Lamellar g/carbide, Secondary carbides M_{23}C_6 , M_6C	Adjacent to σ

The characterization of as-weld HA282 microstructure (Section 2.3.1) confirmed the formation of Ti, Mo-rich MC carbide phase. The predicted Mo-rich M_6C carbide was also observed in the interdendritic regions. The γ' phase did not present in SEM images. Overall the Scheil calculation predicted reasonably well the phase formation in as-weld HA282.

On microsegregation, a comparison was initially made with the experimental data reported by Osoba *et al*¹² with electron probe micro-analysis in laser weld fusion zone of HA282. The rapid cooling rate (about 1800°F/s) in laser welding is desirable for the comparison since the element partitioning during welding was mostly frozen-in due to lack of time for diffusion in solid. However, as seen in Figure 92, element partitioning between solid and liquid was continuously varied during cooling. This suggests that a composition gradient existed even inside a dendrite core. Thus we only took the average values for the dendrite core composition. Also, since γ was the first and the major solid phase, we used γ composition to represent the overall solid composition. To reduce the influence from other solid phases, we chose the temperature range from liquidus down to 2320°F, where γ was the only solid phase yet formed. At this temperature, the solid fraction was about 80% according to the calculation (Figure 89) and should capture well the dendrite core region. The calculated average composition of dendrite core compared fairly well with the measurement data by Osoba *et al*¹² (Table 18). Taking the alloy's nominal composition as the reference, the near equal partition of Co and Cr between liquid and solid, and the deficiency of Ti and Mo in solid were seen to be consistent with the measured data. Partition of Al, however, was found opposite between the calculation and experiment. The range of variation of Al (and also Mo, Ti) with temperature in the solid (γ) phase was predicted fairly large.

Table 18: Comparison of calculated dendrite core compositions with measured in laser welded HA282.

(weight percent)	Al	Cr	Mo	Ti	Co	Ni
Scheil calculation: high-low (average)	1.7-1.45 (1.6)	19.8-20.1 (20.0)	5.83-9.5 (7.7)	1.28-2.5 (1.9)	10.4-9.8 (10.1)	59.0-54.7 (57.0)
Measured in dendrite core ¹²	1.47	19.89	7.90	1.71	10.58	58.59
Alloy nominal composition ¹	1.5	20	8.5	2.1	10	57

In the current study, EPMA quantitative maps of as-welded HA282 material were acquired over large areas. The maps shown in in Figure 97, for example, were taken in an area of $300 \times 300 \mu\text{m}$ with a step size of $1 \mu\text{m}$. More details have been reported in Section 2.3.1. As shown by Figure 97, Ti and Mo were partitioned strongly to interdendritic regions, while Ni and Co moderately to dendrite cores, and Cr, Al, Fe, Mn, and Si did not exhibit a strong preference. These partitioning behaviors are comparable to the observations in the laser welded HA282 by Osoba et al¹².

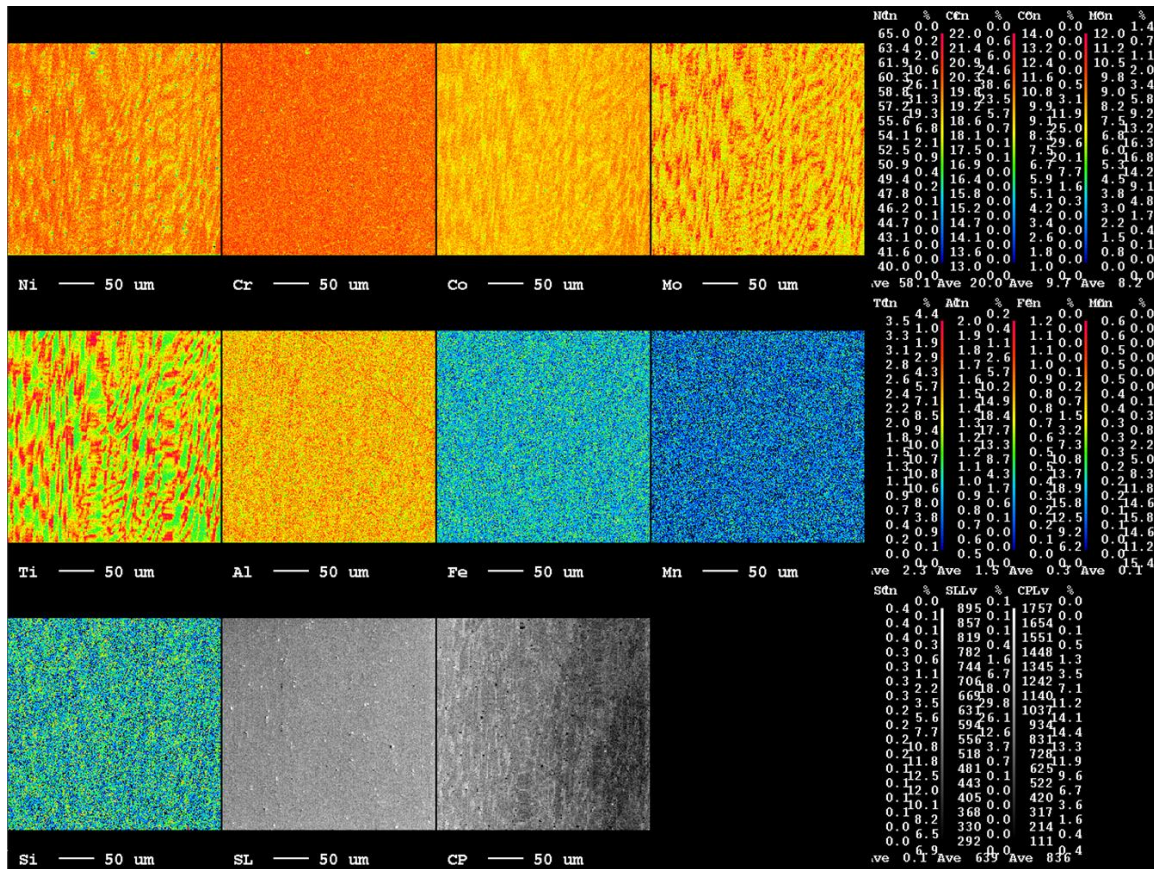


Figure 97: EPMA quantitative map ($300 \times 300 \mu\text{m}$) of as-weld HA282.

To make an approximate comparison with the predicted microsegregation profiles from Scheil simulation and diffusion (DICTRA) simulation, we assumed (1) there was negligible back diffusion in solid, and (2) element segregations varied monotonically from dendrite core to interdendritic region (either increase or decrease). For each element we then converted the EPMA map composition data to a cumulative distribution plot with the y-axis as the composition, and the x-axis as the cumulative probability. We further approximately let the probability be equal to the fraction of solid. It should be noted, however, that the assumption (2) above was not rigorous. It essentially let us to correlate the maximum and minimum ranges of compositions in the data set to either dendrite core or interdendritic region, and was mostly applicable to the situations where only one solid phase existed. In reality secondary solid phases such as interdendritic carbides could alter the maximum/minimum partition of some elements to some extent. The data from Scheil simulation was taken directly from each element's composition in solid (the major γ phase), and listed with respect to the fraction of total solid. From the diffusion simulation, we took the solid and liquid compositions right at the moving solid-liquid boundary, and then converted the boundary coordinate to the fraction of total solid.

Figure 98 compares the converted composition profiles between the EPMA data and the predicted data from Scheil and diffusion simulations. In Scheil model, the partition of each element is described by

$$C_s(f_s) = kC_0(1 - f_s)^{k-1} \quad (3.1)$$

where C_s is its composition in solid at the solid fraction of f_s , C_0 is the alloy (or average) composition, and the constant k is the partition coefficient of the element. It is easy to see that $k = C_s(f_s = 0)/C_0$, thus k describes the initial partition of the element in the solid. For comparison purpose, we also fit Equation (3.1) to the EPMA data to determine the value of k for each element. The fit curves for Ni, Ti, Mo are also plotted in Figure 98 for comparison. Although the comparison is only approximate because of several assumptions, we can see a fairly good agreement between the simulation (Scheil and diffusion modeling) and the EPMA data.

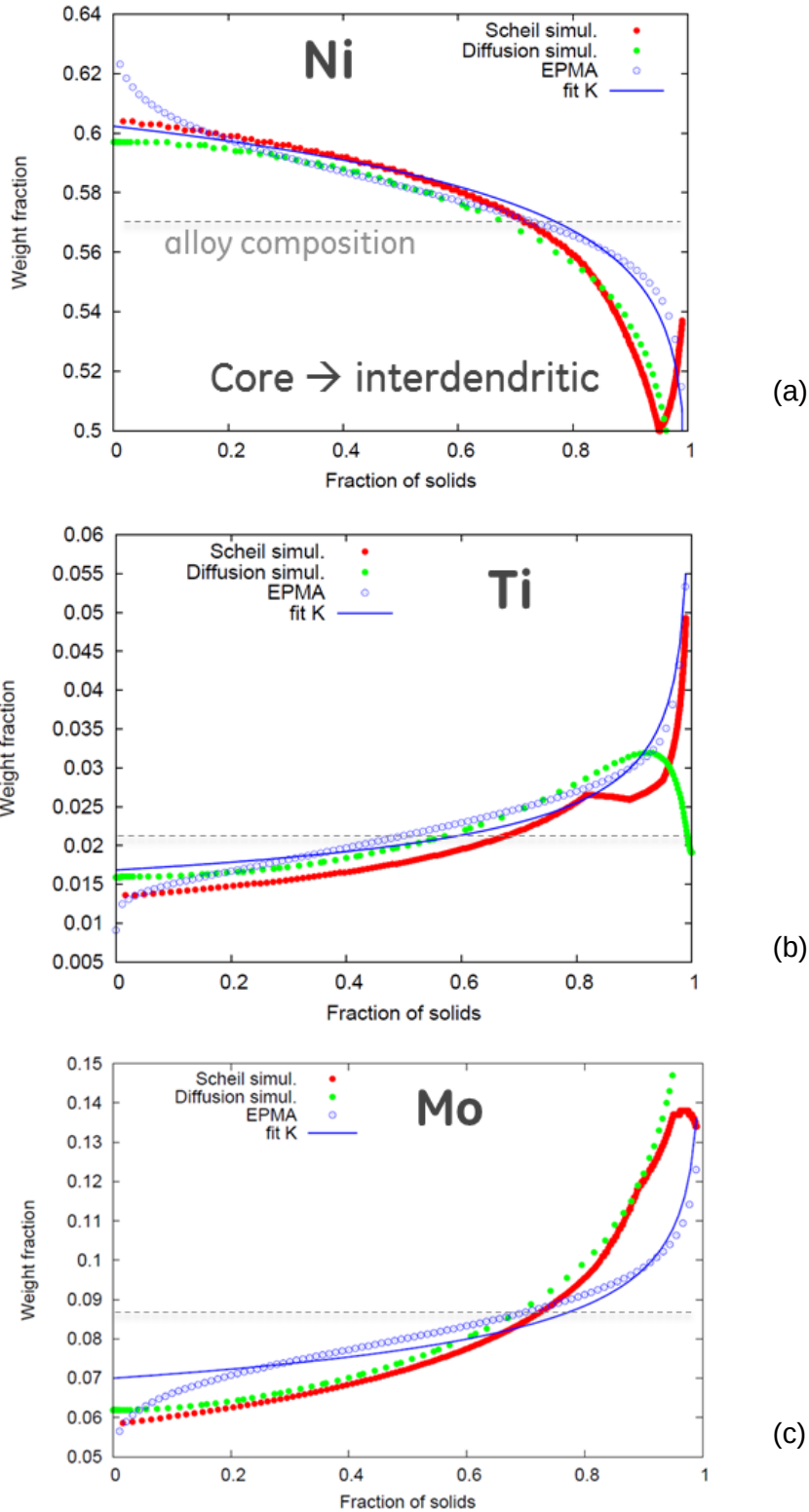


Figure 98: Comparison of microsegregation from EPMA data, Scheil simulation, and diffusion simulation: (a) Ni, (b) Ti, (c) Mo. The fit K curves are from Equation (3.1) fit to EPMA data. Horizontal dashed lines indicate the nominal composition of each element if the alloy is uniform.

Table 19 compares the partition coefficient k of each element in the HA282 weldment. The experimental k values were obtained by fitting Equation (3.1) to the EPMA data. Overall the Scheil predictions are close to the experimental values. The partition coefficient for Ni and Co are greater than one, indicating the elements partition more to solid, and Cr, Mo, Ti are less than one, indicating they enter more preferentially to liquid during solidification. The element Al, however, was predicted in an opposite trend as compared with the EPMA data. As discussed earlier, we have found Scheil model predicted correctly Al segregation to liquid among other superalloys. This wrong trend for HA282 is probably caused by the TTNi7 thermodynamic database.

Table 19: Partition coefficient k fit to EPMA data and predicted by Scheil simulation

	Ni	Cr	Co	Mo	Ti	Al
EPMA fit k	1.04	0.97	1.08	0.86	0.74	0.90
Scheil simul.	1.06	0.98	1.04	0.68	0.62	1.14

Overall, the solidification phases and element partitioning predicted by Scheil and diffusion simulations for HA282 were found in a reasonable agreement with EPMA measurement from as-weld HA282 and with various experimental observations from published work on HA282 alloy. The partitioning trend of element Al, however, was found to be opposite to the experiment observations.

3.2. Modeling homogenization in post-weld heat treatment

3.2.1. Homogenization modeling

In the standard heat treatment PWHT2 (Table 4), the 2075°F solution step is intended to homogenize the element microsegregations from the as-weld condition. The temperature is well above the solvus temperatures of γ' and other low-temperature carbides (M_6C , $M_{23}C_6$). Solid-state diffusion is the principal physical process at this stage.

A main challenge to modeling diffusion processes for superalloys is the large number of alloy elements. In HA282 there are 11 elements (Table 3). In general, diffusion not only depends on the spatial gradients of each element, but also their mutual interactions. In some cases, the interaction can be strong enough to cause ‘up-hill’ diffusion, i.e., an element diffuses against its own gradient. Secondly, multiple phases could be present in the material. In HA282, due to microsegregations, γ' and MC carbide can be present in the interdendritic regions, and influence the overall diffusion (homogenization) kinetics. Treating multiple phases increases considerably the complexity in a diffusion problem. Lastly, for practical interest, a model needs to take temperature as an explicit variable so that it can be applied to a realistic (non-isothermal) heat treatment. In the current study, we adopt a general multi-component diffusion model of

Ågren¹³, Andersson and Ågren¹⁴, and a multi-phase treatment by Larsson *et al*^{15,16} based on a homogenization approximation. An in-house code developed previously has been applied to HA282, with temperature as an explicit variable.

For HA282, we excluded all trace elements Fe, Mn, Si, and interstitial elements C and B. These elements were not expected to have a major effect on the diffusion process due to low concentration. The element C was later introduced in a separate set of simulations for studying the effect of carbides in homogenization process. Homogenization was simulated at several different temperatures, including one at the standard solution temperature 2075°F (PWHT2), and the other at a lower 1895°F (PWHT1). Note the γ' solvus temperature of HA282 is about 1840°F.

The diffusion simulations were performed on two-dimensional space using an in-house code developed previously for single crystal alloys¹⁷. This approach removes various microstructural and compositional assumptions made typically in one-dimensional methods, and allows greater fidelity of treating realistic dendritic microstructures. All simulations started with an actual as-welded HA282 composition distribution measured by EPMA. This included 6 element maps (Ni, Cr, Co, Mo, Ti, Al) in an area of $150 \times 150 \mu m$. The Ti map has been shown in Figure 36.

Selected time snapshots, at 100s, 6min, 15min, and 30min, are shown in Figure 99 for the 1895°F simulation. The top row (time zero) is same as the EPMA map data. During homogenization, all elements show continuous reduction of spatial non-uniformity. The element Mo, as the slowest diffusing species, shows the greatest residual segregation in the end. Ti is also seen to have noticeable non-uniformity. Since Ti is a γ' forming element, a spatial variation of γ' volume fraction is expected. To confirm this, the residual segregation of all elements was taken to a post calculation for γ' volume fraction at 1450°F. The result shown in Figure 100 suggests a $\pm 4\%$ variation around an average 18% γ' volume fraction.

Homogenization simulation at the higher temperature 2075°F is shown in Figure 101. After the same heat treatment times, even the slow diffusion Mo can be seen to become much more uniform.

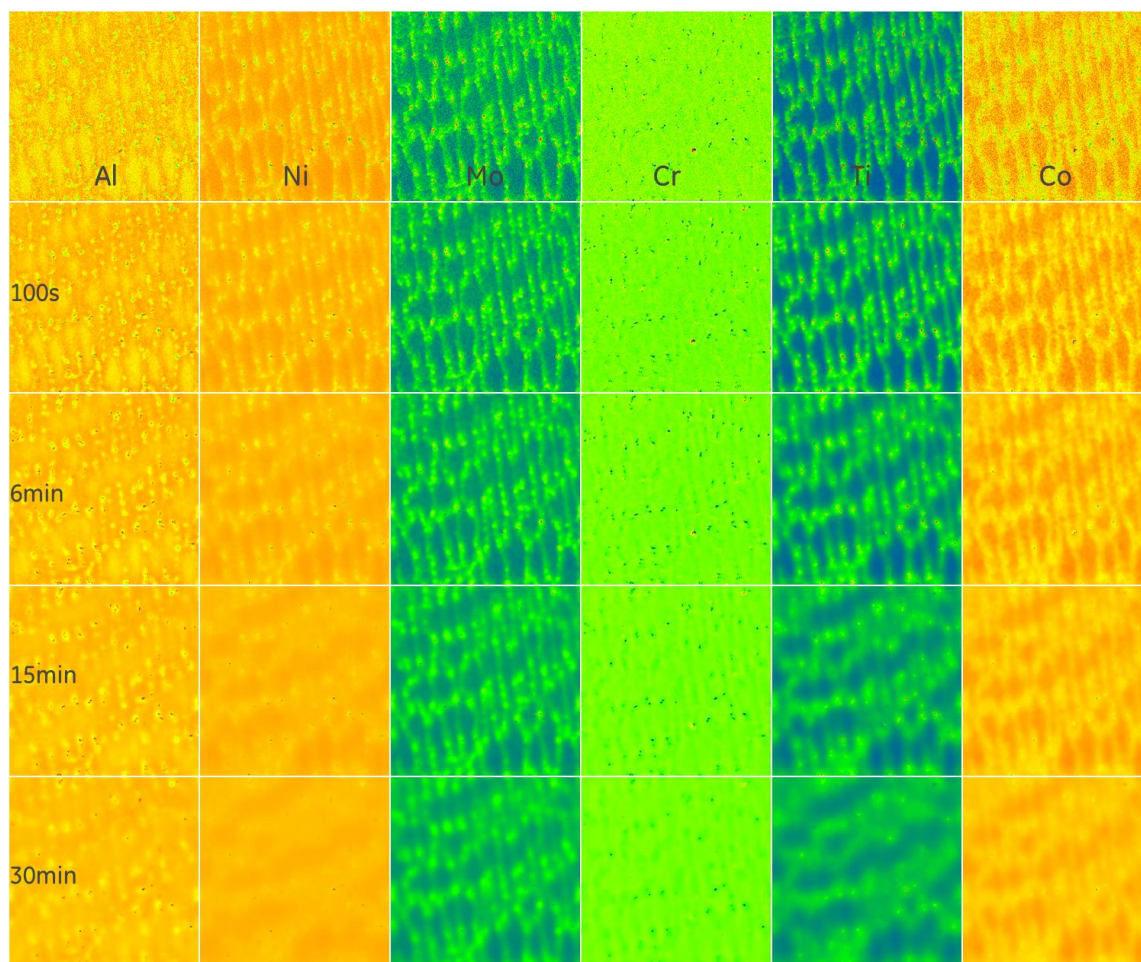


Figure 99: Simulated homogenization of HA282 at 1895°F. The top row is the starting as-welded condition from quantitative EPMA measurement.

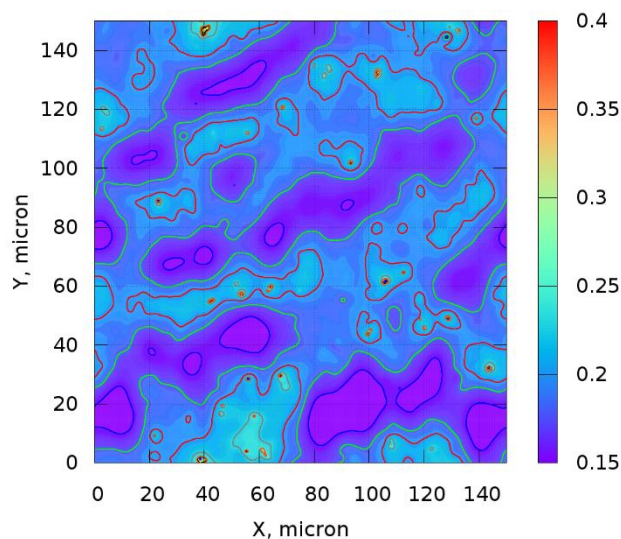


Figure 100: A contour of γ' volume fraction at 1450°F. The non-uniform distribution is due to the residual microsegregation at the end of the 1895F/30min heat treatment shown in Figure 99.

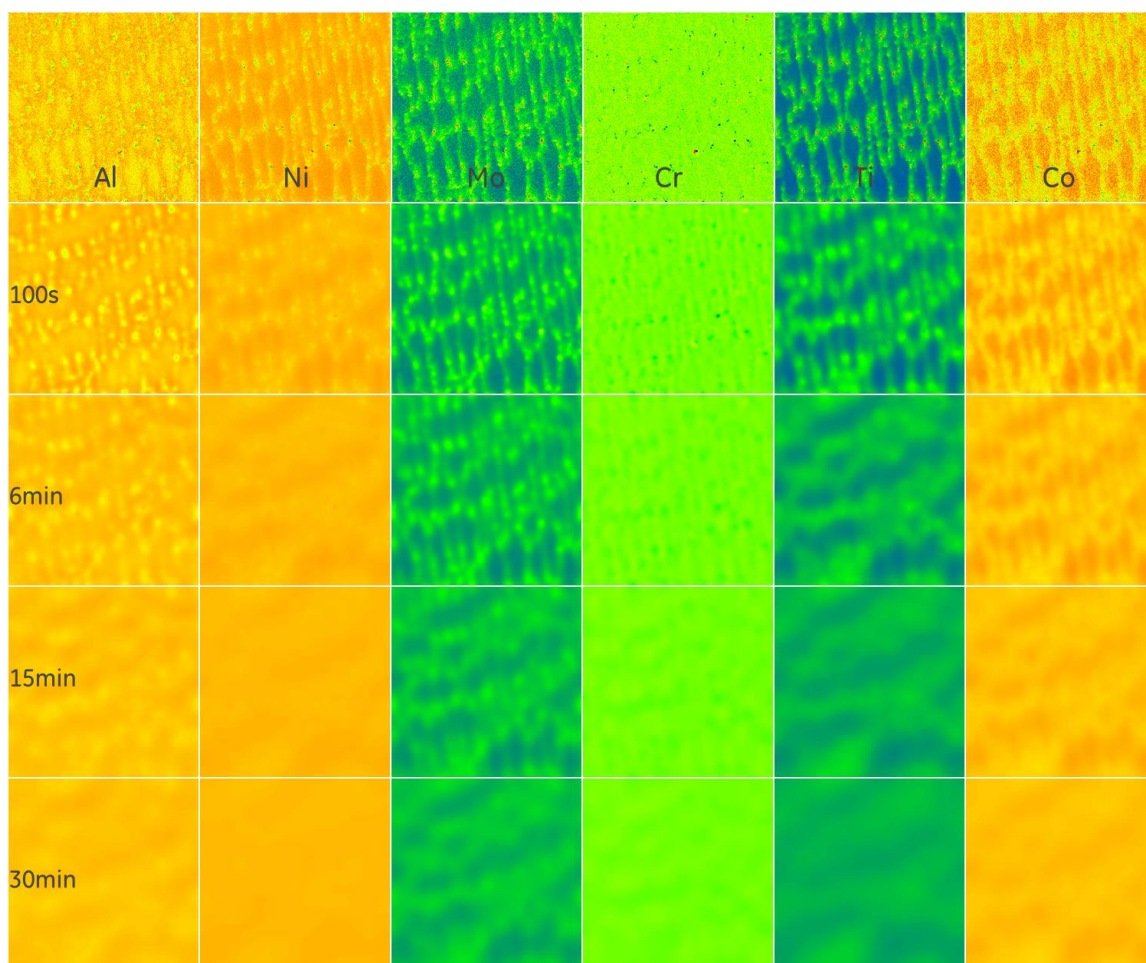


Figure 101: Simulated homogenization of HA282 at 2075°F. The top row is the starting as-welded condition from quantitative EPMA measurement.

An exploratory simulation was performed to examine how carbon could affect homogenization kinetics, primarily under the consideration that carbide phases in the interdendritic regions may lower the chemical potential of some carbide forming elements, and thus alter their diffusion driving force. In the current multi-component diffusion model, including carbon element as a fast-diffusing species causes certain numerical stiffness challenges. As a result, the numerical time step has been reduced greatly. A better solution would have been to use an implicit integration solver. At the moment, we only performed the simulation for a short period of time, just to examine the idea. Figure 102 shows a snapshot at 14 seconds in the 2075°F homogenization heat treatment. Carbon is seen to enrich in interdendritic regions (Figure 102(a)), which leads to stabilization of MC carbides in these regions at about 1.5% volume fraction. $M_{23}C_6$ carbides are found to form in dendrite cores, with volume fraction about 2.5%. They however do not overlap with the regions where MC carbides form.

A comparison of microsegregation between two parallel simulation cases, one with carbon and the other without carbon, shows only small difference over the short time of 14 second during the 2075°F heat treatment.

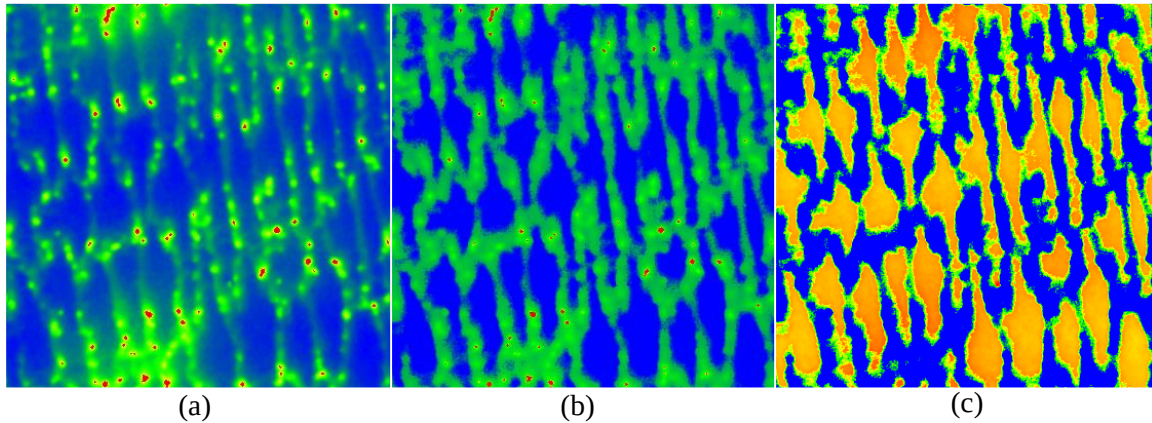


Figure 102: Simulated distributions of (a) carbon composition, and carbide phases: (b) MC, (c) $M_{23}C_6$ after very short time 14 seconds of 2075°F homogenization.

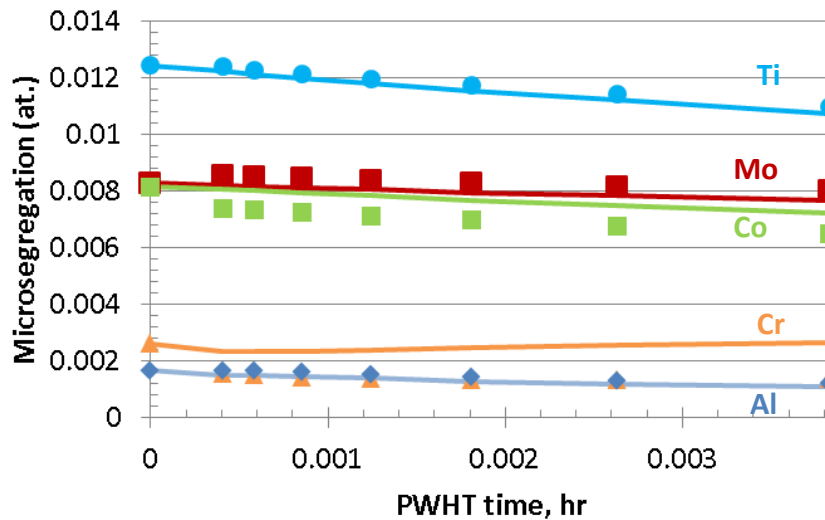


Figure 103: Comparison of decrease of microsegregation in 2075°F homogenization indicates no substantial difference over short time when carbon diffusion and carbide phases are taken into account: (Discrete symbols) with carbon, (lines) without carbon.

3.2.2. Comparison with experiment

For a quick glance of experimental data, Figure 104 shows the EPMA composition maps of the specimen heat treated at 1895°F. This has been compared with the simulated results in Figure 99.

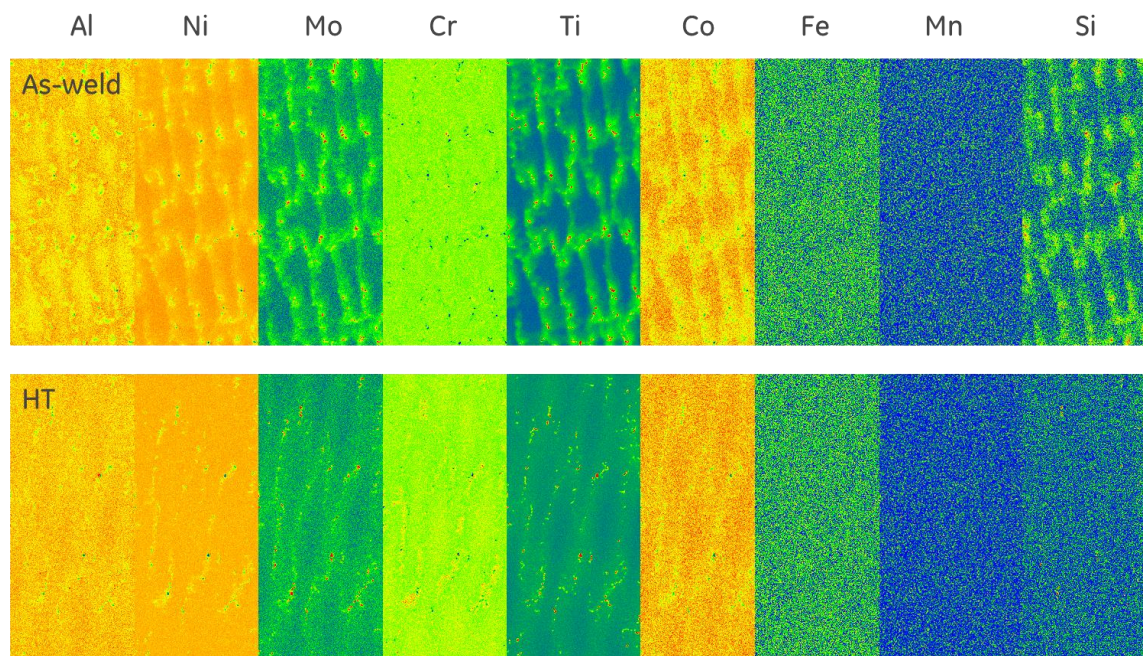


Figure 104: Element distributions from quantitative EPMA measurement: (top row) as-welded HA282, (bottom row) after heat treatment 1895°F/30min + 1850°F/2hr + 1450°F/8hr.

In order to make a quantitative comparison, we applied the same analysis method discussed in Section 2.3.1 to both experimental and simulation data. Similar to Figure 39, the quantified microsegregation profiles across dendrite cores are plotted for each element in Figure 105. Here both the experimental EPMA data of as-welded and heat treated (1895°F) and the simulated snapshots at 1895°F/100s, 6min, 15min, and 30min are compared. The simulation result at the shortest time 100s is close to the as-weld EPMA data, with the latter being also used as the initial state of the simulation. The simulation result at 1895°F/30min (thick blue curves) was found very close to the experimental data (open circles) of 1895°F heat treatment for all the 6 elements. The agreement is an encouraging indication of the model's predictive capability despite the complexity of the simulation, since there is no calibration or fitting parameter in all the simulations. Finally, it is interesting to note that for the controversial Al profile, although the magnitude of segregation keeps decreasing with time, the simulation shows a slight reversal of the Al partition between dendrite core and interdendritic region. The same behavior was also clearly observed in the experimental data.

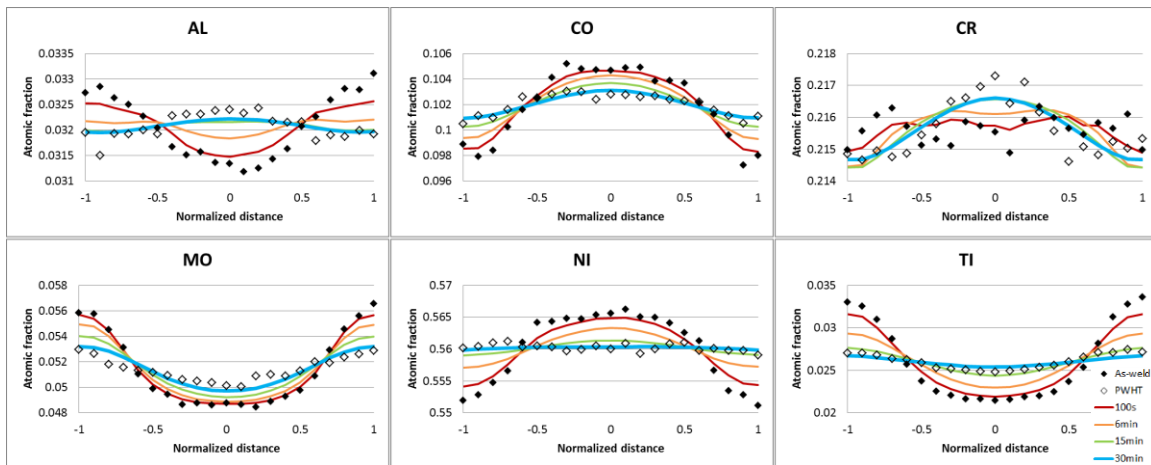


Figure 105: Change of microsegregation profiles during homogenization at 1895°F. The lines are from simulation. The discrete symbols are from EPMA measurement.

The comparison of microsegregation can be made in a further step, by taking the difference between the maximum and minimum values of a composition profile. This single number, reduced from an entire spatial map, characterizes a microstructurally averaged compositional difference between dendrite core and interdendritic region for an element. Plotting this number with respect to time gives an averaged look how the microsegregation of an element is reduced during the heat treatment. Figure 106 compares all the 6 elements with such a construction based on the simulation data. The microsegregations of Ni and Ti, being the highest at beginning, decrease quickly with time, while Mo becomes the most dominant one. The simulation result is seen fairly close to the EPMA data of the specimen after 1895°F solution and 2-step aging heat treatment.

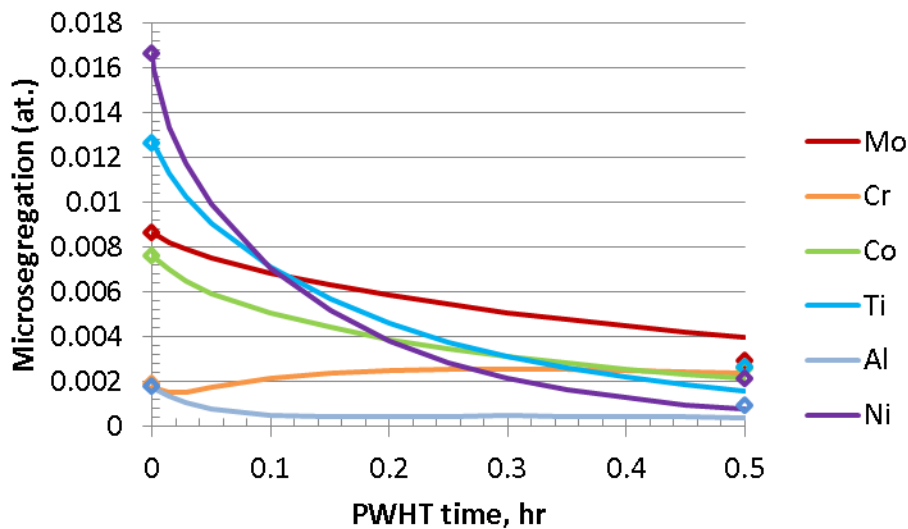


Figure 106: Decrease of microsegregation with time in a heat treatment at 1895°F. Lines are from simulation, symbols are from EPMA data.

3.3. Modeling gamma prime microstructure

3.3.1. The effects of microsegregation on γ' microstructure in weld zone

The creep strength of HA282 is primarily controlled by volume fraction and size of the coherent gamma prime (γ') precipitates. A direct impact of welding to γ' microstructures is from solution and re-precipitation of γ' , as well as the compositional variation due to microsegregation. After a post-weld heat treatment that contains a γ' solution step, such as the four heat treatments in Table 4, the leading factor that affects γ' is the residual microsegregation, in part, of the γ' forming elements Ti and Al. Among the four conditions, PWHT2 achieved a most homogeneous composition distribution in the weld zone, while PWHT3a and 3b retained various degrees of residual microsegregations of Ti and Al. The resulting effect is a spatial variation of γ' in the weld zone, from dendrite cores to interdendritic regions.

An estimate of γ' variation as a result of residual microsegregation is given in Figure 107. The calculation of γ' fractions took local compositions along the same line profile of the microsegregation analysis. At each coordinate (normalized distance) on the profile, the composition of all alloying elements was taken to make an equilibrium phase calculation, at a given temperature, with ThermoCalc software.

Compared with PWHT2, γ' fractions after PWHT3a and 3b show noticeable variation between dendrite core and interdendritic region, at about 4% and 3% volume fraction difference, respectively. The effect of this variation is expected to be more significant for mechanical properties at 1700°F than at 1500°F, because of the lower average γ' fraction 8% at 1700°F, compared with 16% at 1500°F.

Figure 108 compares the effect over a range of temperatures from 1400 to 1800°F. γ' fraction with an ideal homogeneous composition of HA282 is included in the plot. Note the alternate heat treatment PWHT2 (with a 2075°F solution anneal) is essentially same as the homogeneous case. Further increase of the microsegregation beyond PWHT3a leads to a greater difference of γ' between dendrite core and interdendritic region. Set3-4 in Figure 108 represents approximately the as-weld condition, while Set3-7 has an even higher (and unrealistic) degree of microsegregation.

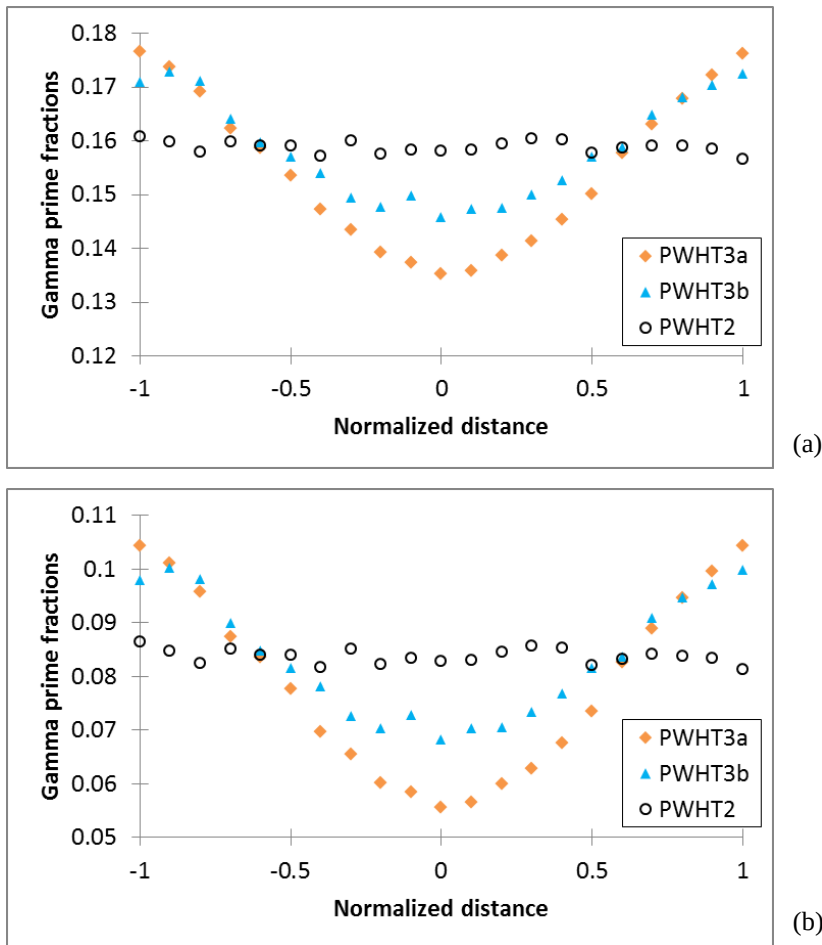


Figure 107: Simulated variation of gamma prime fraction at (a) 1500°F, (b) 1700°F, across a dendrite (core = 0, interdendritic region = ± 1) after different post-weld heat treatments.

Taking into account the microsegregation in a kinetic modeling of γ' coarsening with the precipitation model, we found that in the range of 1500-1700°F, even in the extreme case Set3-7 the γ' sizes evolution did not show a significant difference from the homogeneous case (Figure 109). The similarity of γ' sizes between dendrite core and interdendritic region after PWHT3a is also seen in the SEM images in Figure 110(c) and (d).

In Figure 110(b), some coarse γ' particles appeared in the interdendritic region after PWHT3a. Normally the γ' solvus temperature of HA282 is below the highest temperature 1850°F of PWHT3a. However, using the actual microsegregation from measurement, we calculated the γ' solvus in dendrite core and interdendritic region. The interdendritic solvus in PWHT3a was found at 1868°F (Table 20). The calculation result is consistent with the observation of un-solutioned γ' .

It thus appears that the main effect of residual microsegregation in weld zone of HA282 is primarily on γ' volume fraction, solvus temperature, and insignificantly on intragranular γ' size.

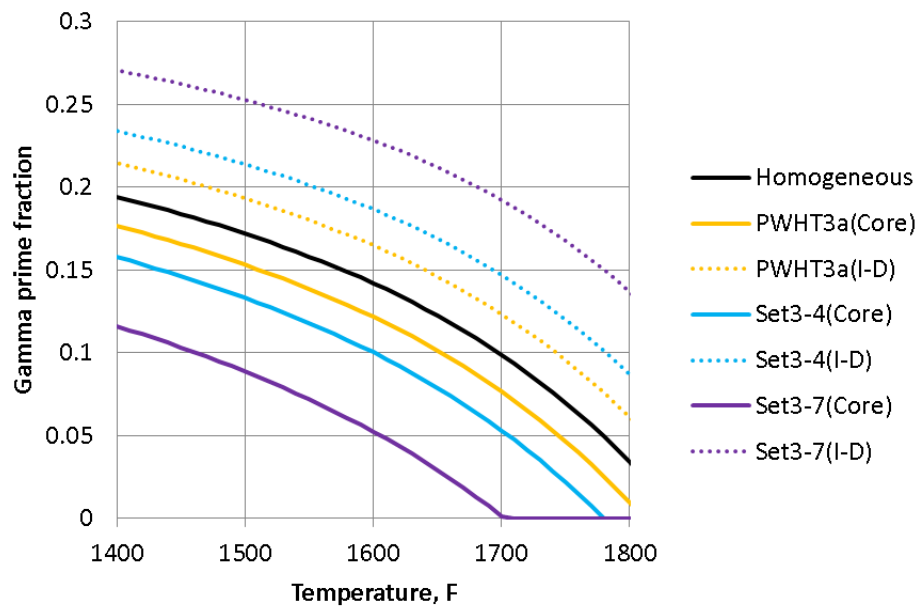


Figure 108: Simulated γ' volume fractions under different conditions of residual microsegregation.

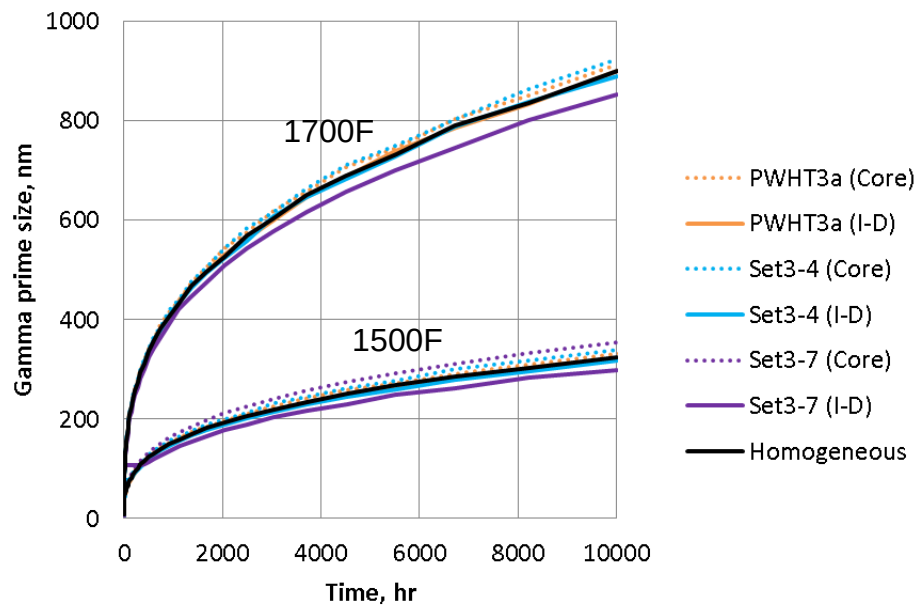


Figure 109: Simulated γ' sizes (coarsening) under different conditions of residual microsegregation.

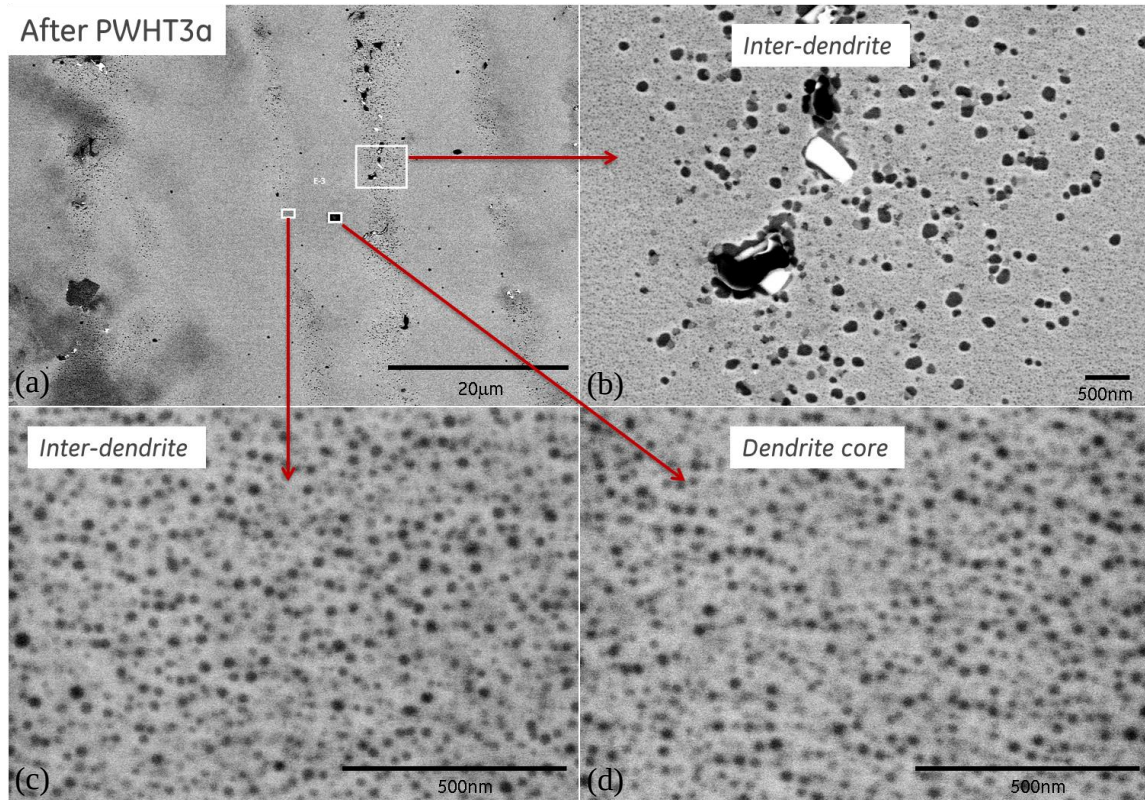


Figure 110: Backscattered electron SEM images of HA282 weld zone after post-weld heat treatment PWHT3a. (a) dendrites shown in a low magnification, the typical γ' precipitates in (c) interdendritic region and (d) dendrite core. Some un-solutioned γ' are seen in the interdendritic region as coarse particles (b).

Table 20: Calculated γ' solvus temperatures with different residual microsegregation.

γ' solvus ($^{\circ}\text{F}$)	Core	Interdendritic region
Homogeneous	1839	
PWHT3a	1811	1868
Set3-4 (As-weld)	1780	1894
Set3-7	1702	1938

The above analyses assumed that the microsegregation remained unchanged. This seems to be a reasonable assumption, considering that further thermal exposure in creep tests at 1500-1700 $^{\circ}\text{F}$ is significantly lower than the homogenization temperature 1850-2075 $^{\circ}\text{F}$. However, over a long period of time for creep, the effect can be accumulated. To evaluate this impact, we performed a set of diffusion simulation with both γ and γ' phases using the model that we had applied previously to model homogenization heat treatments. Here PWHT3a was taken as the initial condition. The time evolution of γ' variation is compared in Figure 111 for different exposure temperatures.

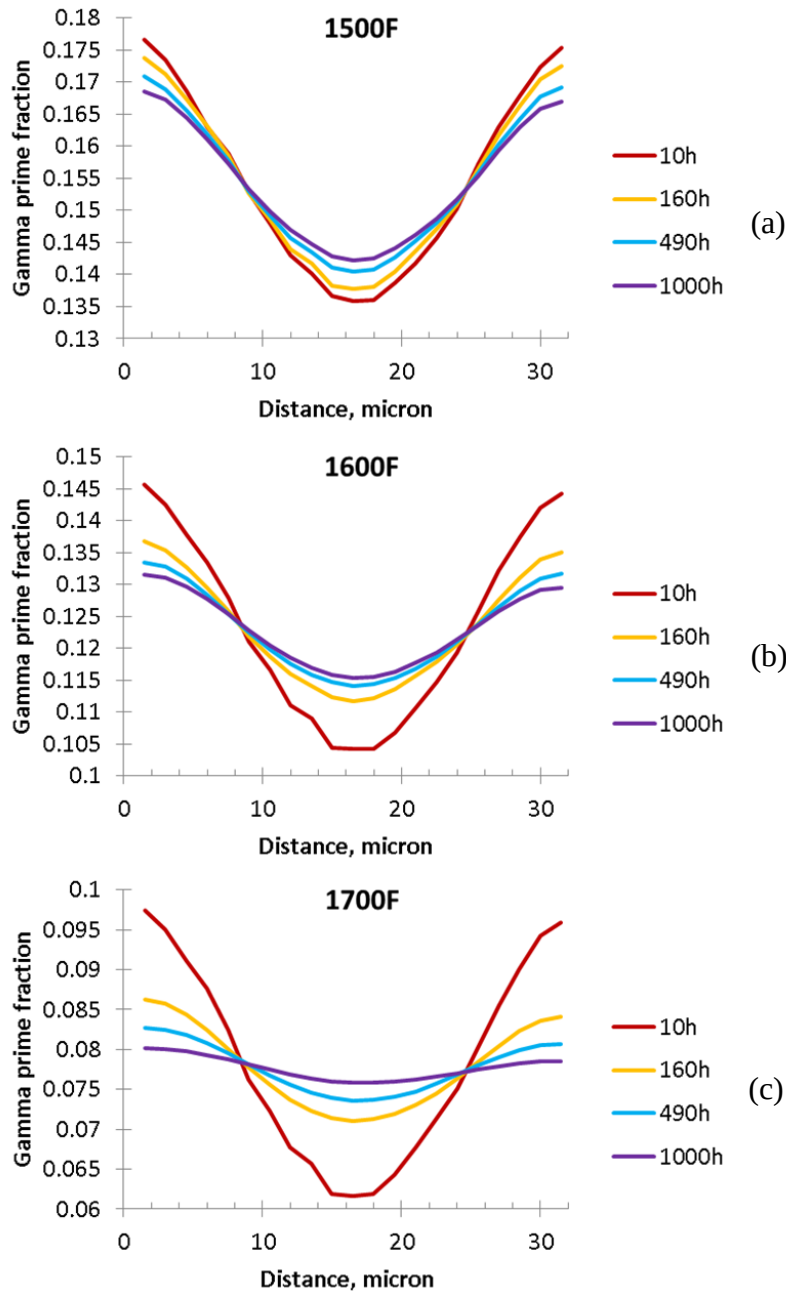


Figure 111: γ' volume fraction becomes more homogeneous between dendrite core (middle) and interdendritic region during extended exposure at the creep test temperatures (a) 1500°F, (b) 1600°F, (c) 1700°F, as suggested by diffusion simulations.

The comparison in Figure 111 shows a clear acceleration effect as the temperature increases. To show it more clearly, the amplitude of γ' variation (the difference between dendrite core and interdendritic region, or max minus min) is compared in Figure 112. Overall, the acceleration is a result of both an Arrhenius relation of diffusivity to temperature and a decrease of γ' volume fraction as temperature increases. The simulation results indicate that γ' will actually

become more homogeneous after a few hundred hours at 1600-1700°F, while much less so at 1500°F. Therefore, the conclusions drawn from Figure 108 and Figure 109 should be considered as an upper-bound of the effect of microsegregation on γ' microstructures in the weld zone of HA282 during creep.

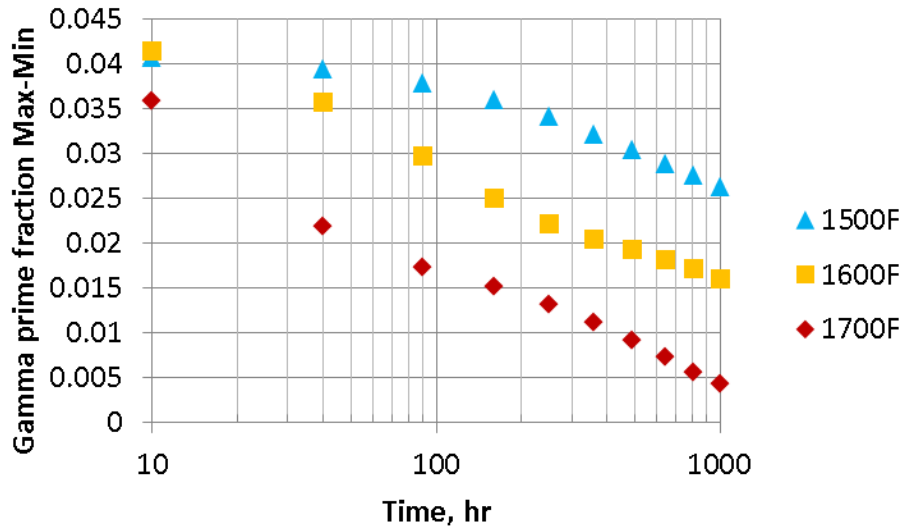


Figure 112: Comparison of homogenization of γ' distribution in weld zone between 1500-1700°F. The initial γ' variation is taken as PWHT3a. The profiles are shown in Figure 111.

3.3.2. Construction of inhomogeneous γ' microstructures for crystal plasticity modeling

Unlike the coupling of γ' model and the constitutive creep model in Task 5, where a γ' microstructure needs to be reduced to a volume fraction and an average (time-dependent) γ' size, the crystal plasticity model in Task 4 uses directly spatial γ' microstructures. Thus the effects of spatial inhomogeneity of γ' in the weld zone can be better explored without over simplification. The knowledge derived from there can be applied to better coupling of the γ' model and the constitutive creep model for weldment.

Figure 113 shows our first try on constructing a non-uniform microstructure to represent γ' in a half width of a dendrite arm (from dendrite core to interdendritic region). Instead of arbitrarily assuming a certain size of the particles in a composition gradient, we used a phase field precipitation model to actually simulate the process of γ' nucleation and growth. As a result, the variation of γ' size in addition to volume fraction was produced on a better physics basis. In this first try, the initial non-uniform composition was taken in a linear gradient from dendrite core to interdendritic region that approximately corresponded to the residual microsegregation from PWHT3a. Then the system was quenched from the calculated solvus (1840°F) rapidly at 200°F/min to a lower temperature at

1500°F, and was stabilized for various amount of time. The microstructure shown in Figure 113 is after 400 hours. The bottom left inset verifies the increase of local fraction of γ' , taken as the area fraction on successive lattice planes, along the composition gradient.

In addition to γ' gradients, we applied the same procedure to generate uniform γ' microstructures at different volume fractions (Figure 114). These 'digital' microstructures have been used in the crystal plasticity model as the baseline cases for studying the effects of inhomogeneous γ' in Task 4.

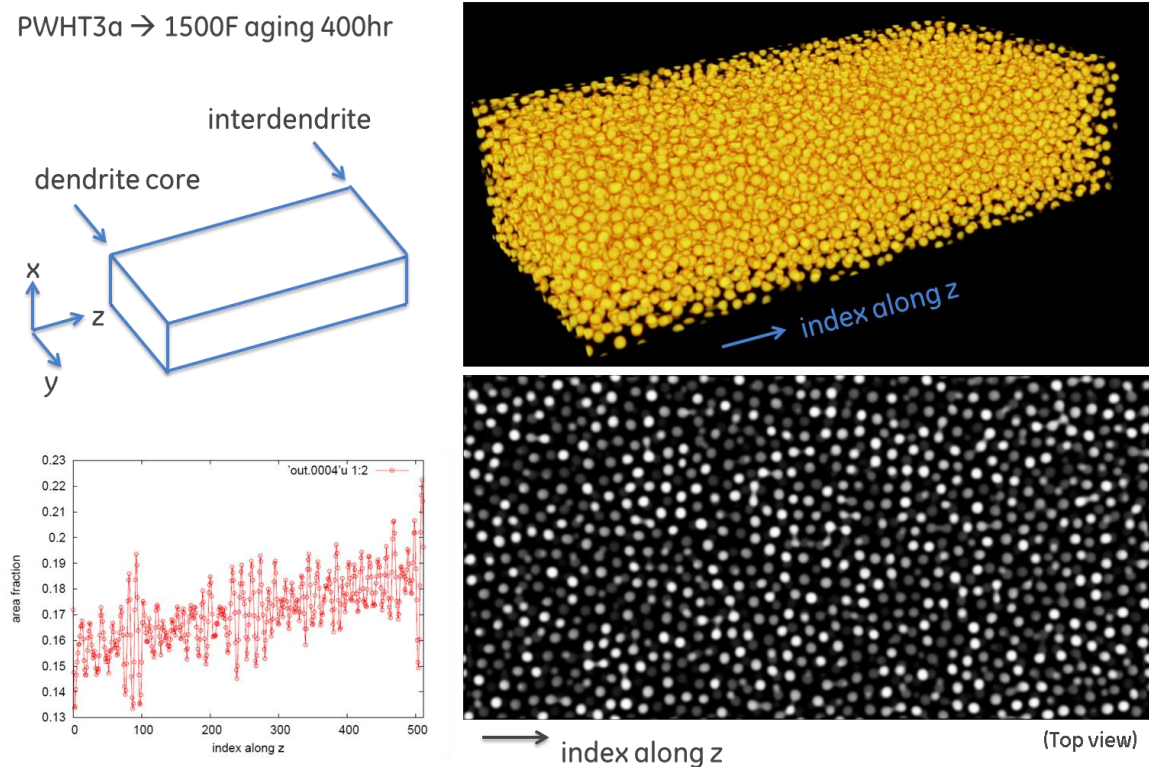


Figure 113: Construction of a 3D γ' microstructure with the phase field precipitation model. A linear gradient of volume fraction is assigned along the 'z' axis with the distance (from dendrite core to interdendritic region) 10 μm . The volume fraction variation represents approximately the condition of PWHT3a followed by a short 400 hours aging at 1500°F.

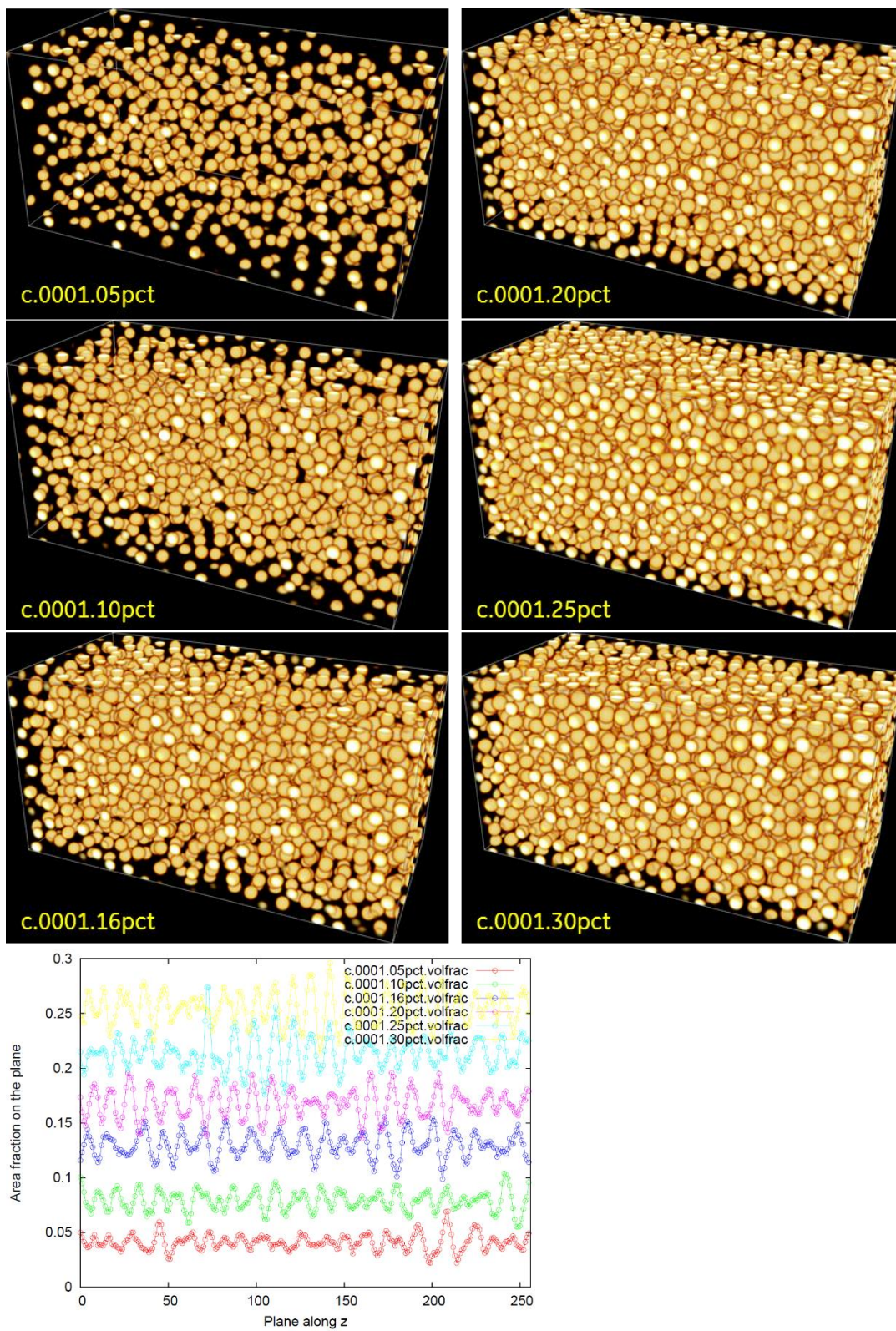


Figure 114: Simulated 3D γ' microstructures with different volume fraction. In each case the γ' has a spatially uniform distribution, as shown by the local volume fraction profiles at the bottom.

In the subsequent tries, we expanded the microstructure to a full dendrite arm width, with dendrite core in the middle (Figure 115). The linear composition gradient was replaced by a sinusoidal gradient that represented better the actual microsegregation profiles across a dendrite arm (see, for example, Figure 39). With the average composition (thus the average volume fraction) fixed at a constant value of HA282, the amplitude of the sinusoidal profile is successively increased from Set3-1 to Set3-7 (Figure 115), among which Set3-4 represents approximately the as-weld condition. The PWHT3a condition is about between Set3-2 and Set3-3. Additional non-realistic cases beyond the as-weld condition (Set3-5 to 3-7) were generated for the purpose of numerical study. While the spatial variation of γ' is not so obvious in a 3D microstructure of Set3-4 (as-weld), it is clearly noticeable in the extreme case Set3-7. Additionally, in Set3-7 a gradient of the size of γ' is also clearly seen, with a finer size in dendrite core (in the middle) than in the interdendritic region (at two sides).

Figure 116 and Figure 117 show the constructed microstructures for PWHT3a, by interpolating the model settings between Set3-2 and Set3-3. Upon formation of γ' precipitates from cooling, the microstructure was stabilized at 1500°F (Figure 116) and 1700°F (Figure 117), respectively. A spatial variation can be seen in the top view of each 3D microstructure and is more obvious at 1700°F because of the reduced average volume fraction at the higher temperature. By plotting the profiles of the local γ' fraction, the spatial variations are clearly seen at both 1500 and 1700°F.

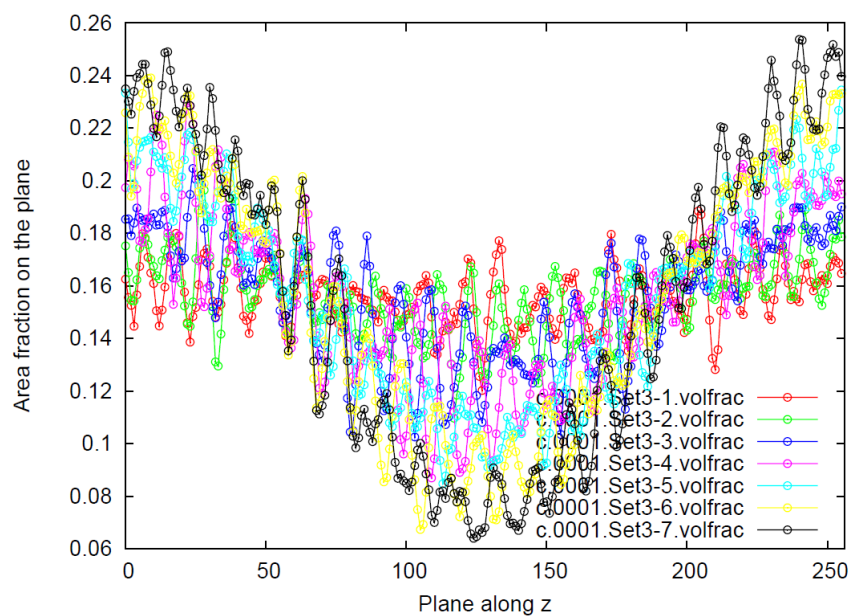
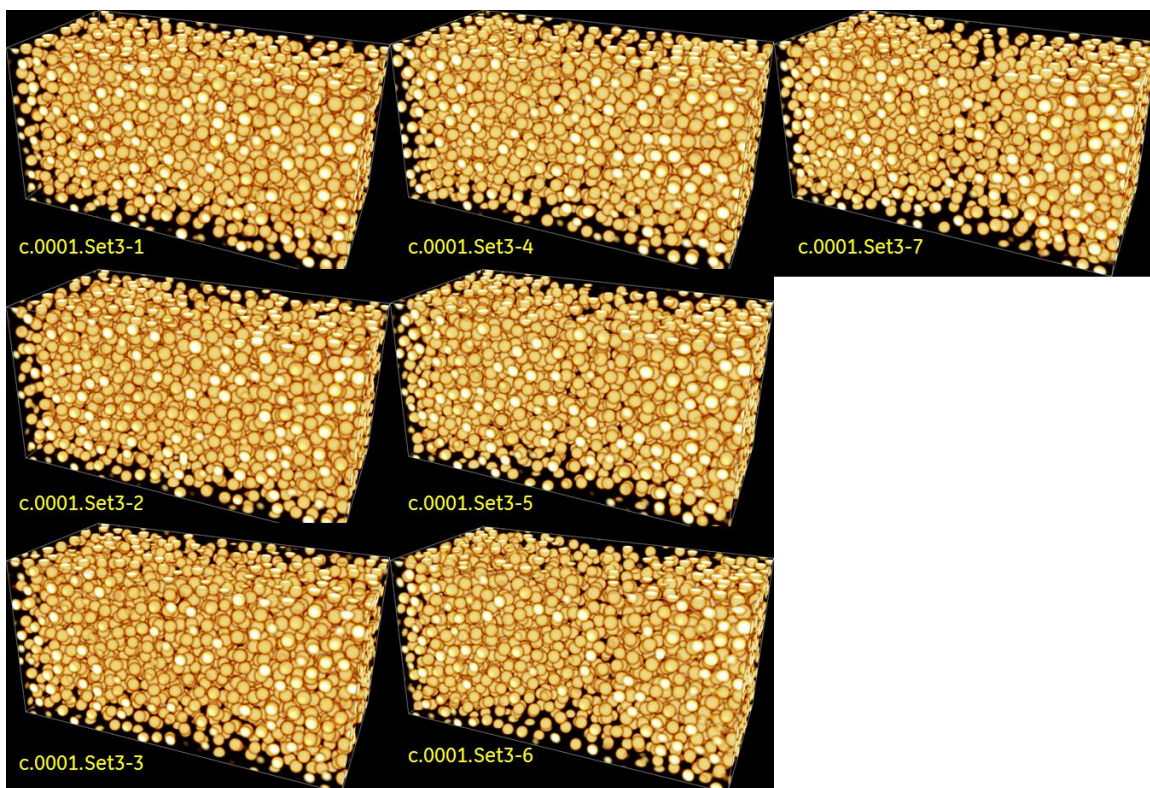


Figure 115: Simulated 3D γ' microstructures with a sinusoidal distribution of volume fraction that represents the variation across a dendrite. From Set3-1 to Set3-7, the amplitude of variation increases, but the average is kept same, as shown by the local volume fraction profiles in the inset.

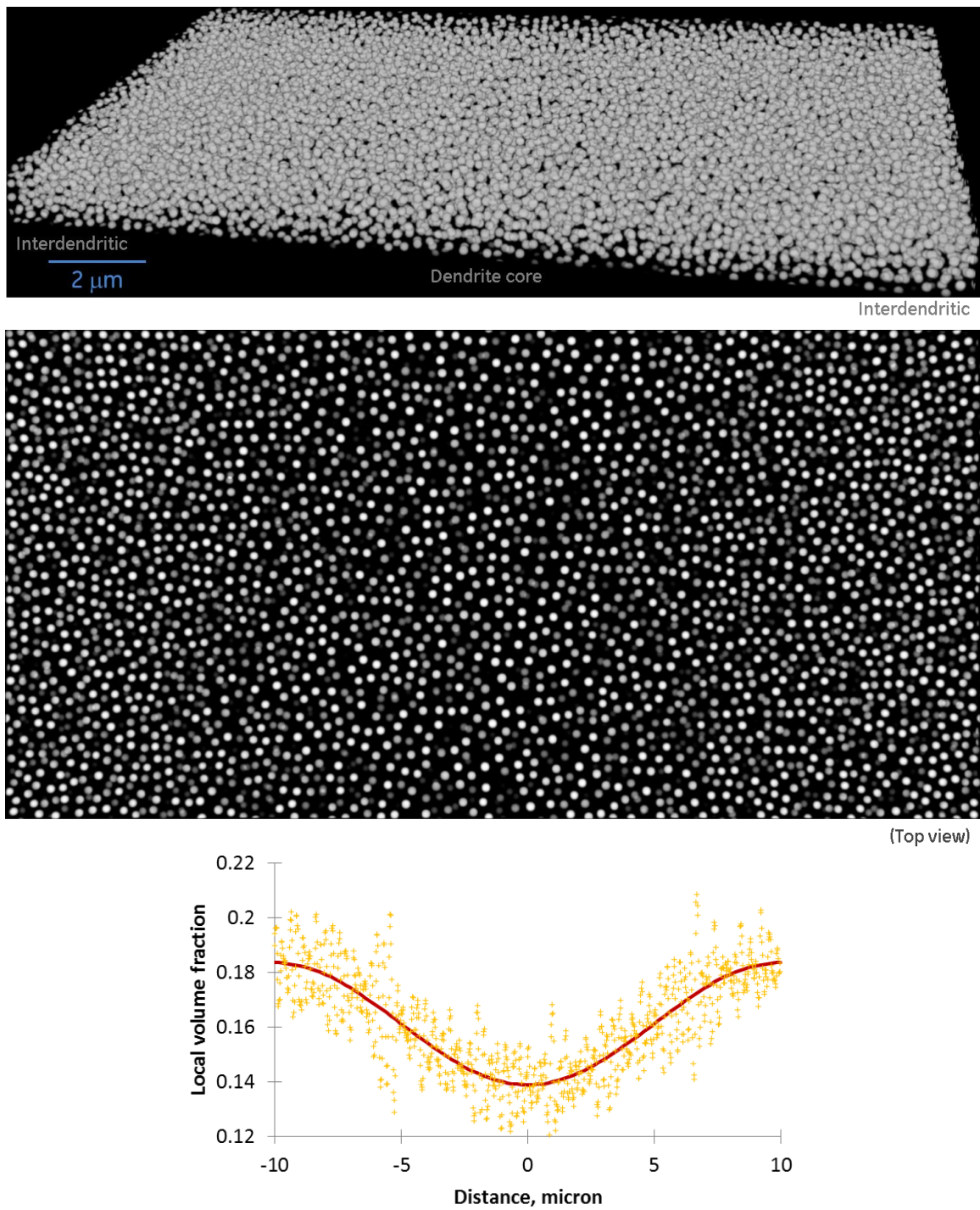


Figure 116: A simulated 3D γ' microstructure at 1500°F with the residual microsegregation of PWHT3a.

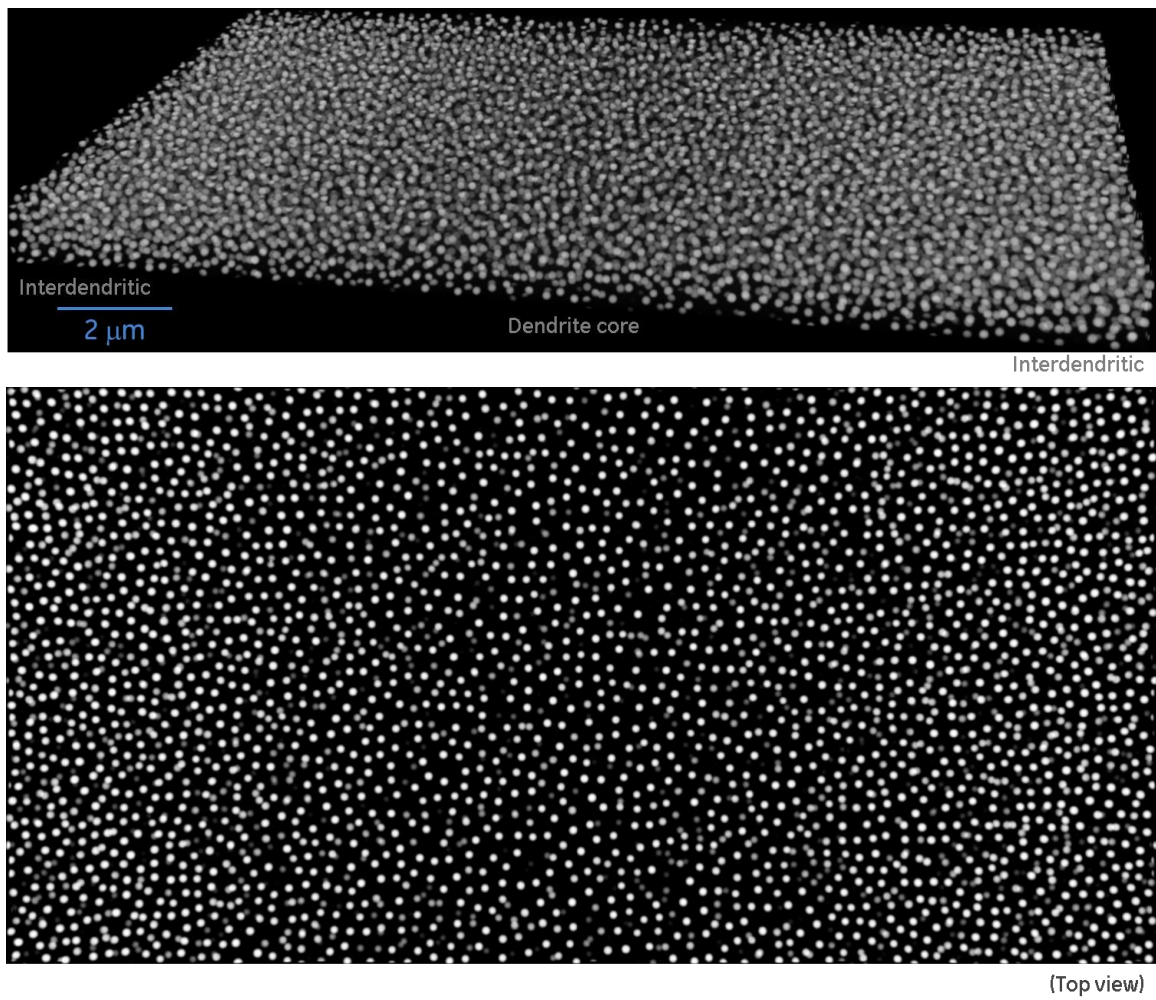


Figure 117: A simulated 3D γ' microstructure at 1700°F with the residual microsegregation of PWHT3a.

3.3.3. Comparison of γ' prediction with long-term heat treatment experiment

The predicted γ' sizes are compared with experimental data of furnace heat treatment for up to 10,000 hours (see Section 2.3.3) in Figure 118. As discussed in a Quarterly Report (2016Q2), originally the three heat treatments, designated as HT19391, HT19398, and HT19390, were intended for isothermal aging at 1500, 1600, and 1700°F, respectively. However, during long exposure time the heat treatment furnaces had experienced large temperature fluctuations. Therefore, instead of using the nominal heat treatment temperatures in the simulation, we applied the actual time-temperature data from thermocouple record for each specimen. The effects of temperature variation are easily seen in Figure 118 between the solid lines and the dashed lines.

The comparison with the experimental measurement suggests an excellent predictability of the model at 1500°F (HT19391) and 1600°F (HT19398). At 1700°F (HT19390), the measured γ' sizes (red open symbols in Figure 118) were found generally smaller than the model prediction. The reason for the discrepancy, however, is not clear. Unlike at 1500°F and 1600°F, the experimental data at 1700°F are quite scattered among base metal, HAZ, and weld zone. Note the current measured data at the nominal 1700°F heat treatment fall approximately around the previous 1650°F data.

Previously¹⁸, the same model has been compared well with heat treatment data at 1650°F up to 3000 hours (Figure 120). In the current project, we also compared the model prediction with long-term heat treatment data of Oak Ridge National Lab¹⁹ in a temperature range between 1300-1500°F (Figure 119). The comparison showed an excellent agreement. It thus seems that the best way to investigate the above discrepancy in HT19390 is either to look for additional 1700°F coarsening data in the literature, or to repeat the 1700°F long-term heat treatment. The latter option, however, was not feasible given the schedule constraint in the project.

Overall, we found that the HA282 Precipitation Model performed very well over long term exposures at 1300-1650°F (1500-1600°F in the current study, in addition to 1300-1650°F in the previous studies). The model was calibrated only to short-term heat treatment data within a few hours, and was able to predict accurately the evolution of γ' sizes over thousands of hours at various high temperatures. The discrepancy at 1700°F remains as a subject for a future study when additional 1700°F coarsening data are available.

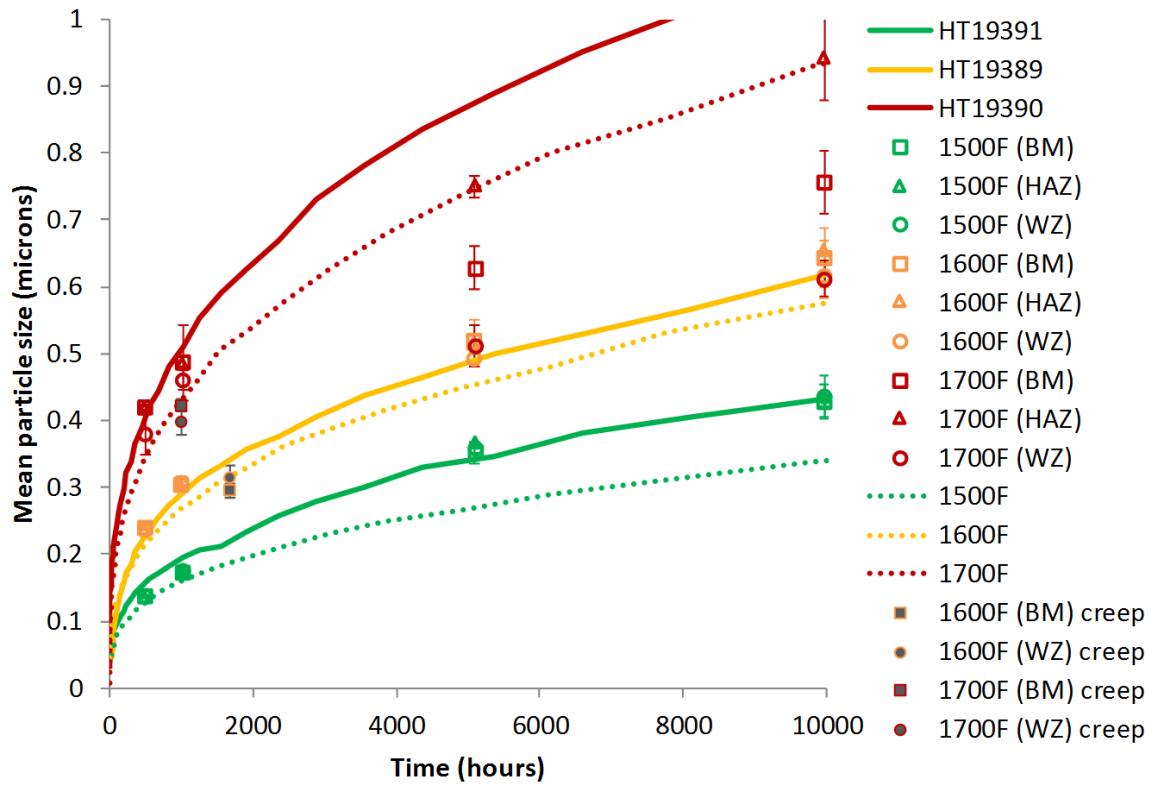


Figure 118: Comparison of γ' sizes from precipitation modeling (lines) and experimental data (open symbols, same data as in Figure 60) up to 10,000hrs. Among the model predictions, dotted lines used nominal (target) temperatures, solid lines used the actual data from thermocouple record. Additional data measured from creep specimens (gray filled symbols) are also compared.

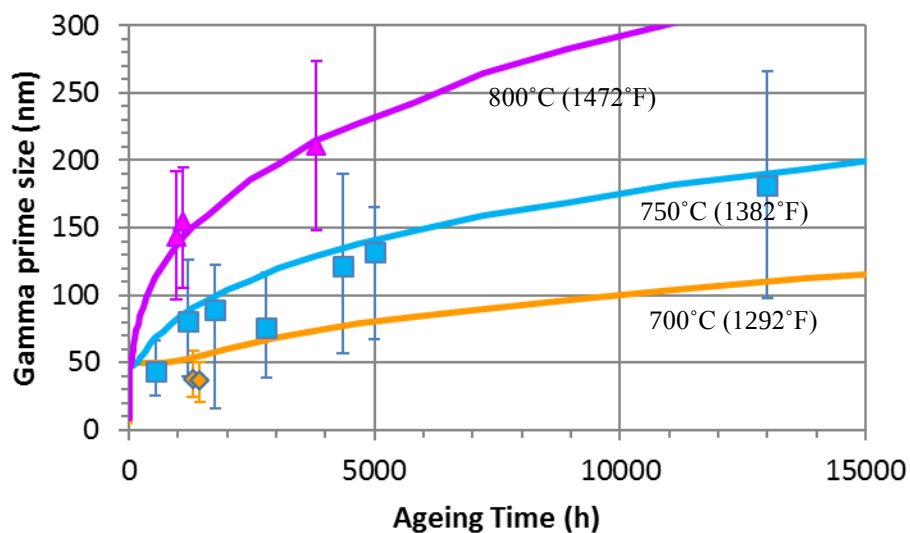


Figure 119: Validation of HA282 precipitation model predictions (lines) with long-term heat treatment experimental data (symbols, courtesy Drs. P. Tortorelli and K. Unocic, ORNL, Ref. ¹⁹).

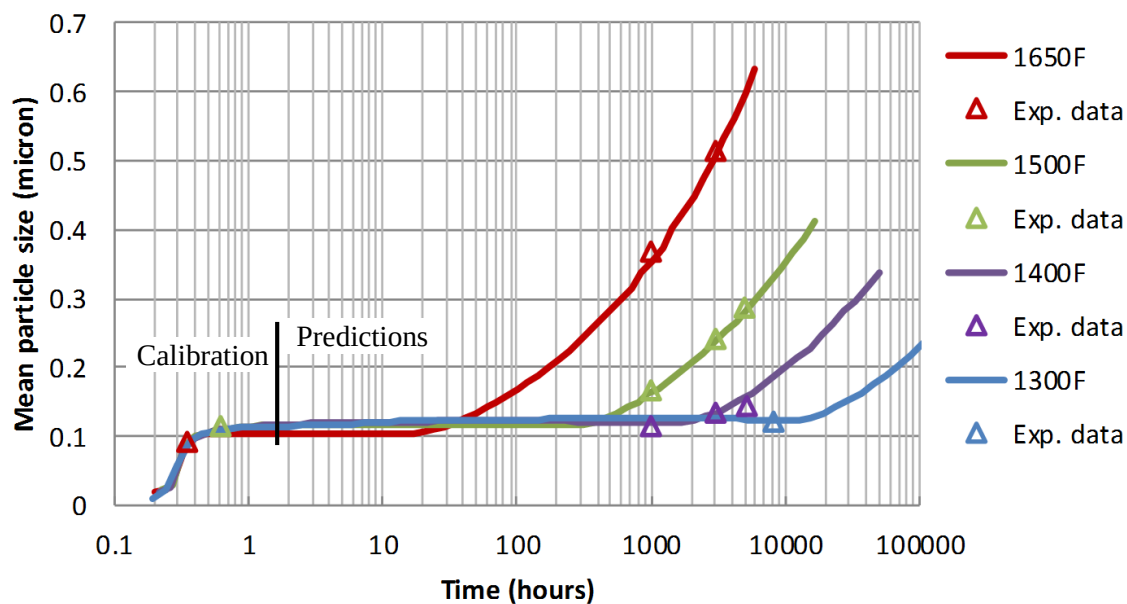


Figure 120: Previous model validation between 1300-1650°F (Ref. ¹⁸). Solid lines are from simulation.

3.3.4. Comparison with γ' in creep specimens

In the earlier section (Sec.2.3.5) we have discussed morphological difference of γ' in the tested creep specimens as compared with the furnace heat treatment specimens that had no stress load. Clearly the stress changed the shape of γ' particles. On the other hand, comparing the measured γ' sizes from two creep specimens, 1600°F/15ksi (run to rupture at 1678 hours) and 1700°F/8ksi (interrupted at 1000 hours), we did not see a significant difference from the γ' sizes predicted without stress load at 1600°F and 1700°F, respectively (Figure 118). Additionally, the measured γ' volume fractions from the two creep specimens did not differ much from the static heat treatment specimens or from the model predicted curve in Figure 121. Thus, we conclude that although creep stress changed γ' morphology through plastic deformation, it had a negligible effect on γ' volume fraction and coarsening kinetics in the range of current study.

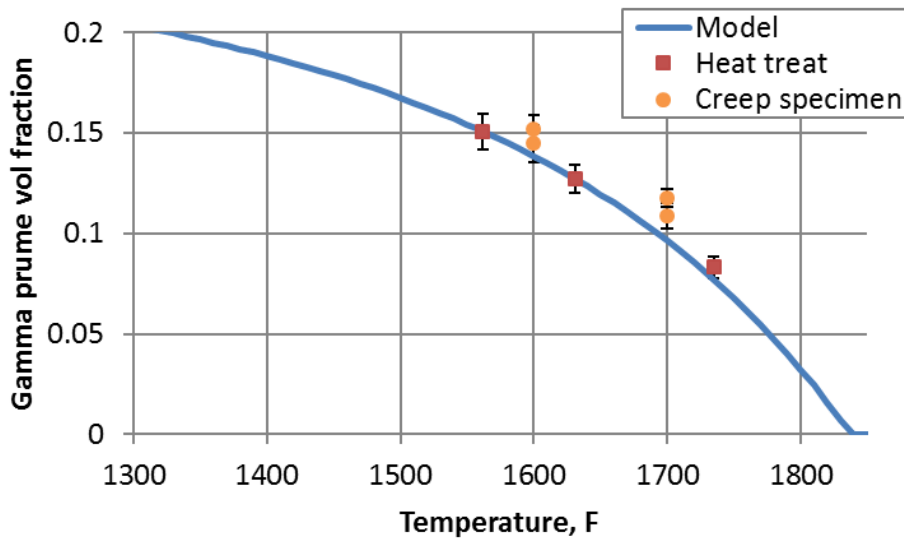


Figure 121: Comparison of γ' volume fractions from precipitation modeling (line), measurement from furnace heat treated specimens (square symbols), and from creep specimens (circular symbols).

Task 4: Mesoscale Modeling of Plastic Deformation in a Superalloy weld

The objective of Task 4 is twofold: (1) modeling the plastic deformation of HA282 at microstructural level by incorporating experimentally revealed dislocation mechanisms, and (2) simulating the grain boundary cavity evolution subject to creep condition. The tensile and creep tests on HA282 base and weld samples in Task 2 provided dataset that was used to calibrate the crystal plasticity model. The microstructure study in Task 3 also provided 3D γ/γ' microstructural dataset that contained the experimentally observed structural non-uniformities produced during the welding and/or post-weld heat treatments. Those results from previous tasks served as the direct inputs for the current task, allowing an experiments-informed simulation study and assessment of the microstructural influence on creep performance of HA282.

In this task, two proposed milestones have been completed: (1) the development of full-field crystal plasticity model for inhomogeneous weld microstructure was completed in 2016Q1, and (2) the mesoscale modeling of creep cavity evolution under creep condition was completed in 2016Q4. The deliverable of this task include: (1) simulation of microstructure effect on plastic deformation in weld HA282 and (2) grain boundary cavity evolution under constant stress. Those results provide useful data that can be effectively used in Task 5 for the long-term macroscopic constitutive creep modeling.

4.1. Mechanisms-informed, microstructure-sensitive crystal plasticity modeling

The goals of this subtask include:

- (1) develop a micromechanical simulator that can use complex γ/γ' two-phase mixture microstructures and account for the dislocation mechanisms,
- (2) calibrate the model developed in (1) against HA282 base metal using both tensile and creep test data from Task 2,
- (3) explore the effect of structural heterogeneities of HA282 weld such as micro-segregation and grain boundary denuded zone.

To this end, some critical challenges have been addressed, including the incorporation of complex geometry of γ/γ' microstructures containing different structural non-uniformities and different dislocation-precipitate interaction for tensile and creep conditions.

4.1.1. Model development of FFT-EVP

Traditionally, crystal plasticity (CP) models for various engineering metals and alloys have been developed within the framework of finite-element method (FEM)

^{20,21}. In FEM-based CP (FEM-CP) models, continuum equations (e.g., stress equilibrium equation) are solved in the weak form (i.e., an integral form) using a class of numerical methods (e.g., Galerkin method) that convert the original problem to a discrete problem where the solution is approximated by a set of so-called shape-functions that are defined locally in the real-space. To complete the discrete problem, the real-space domain needs be discretized into a collection of small elements, which is a procedure usually called “meshing”. For problems with a simple geometry, meshing is usually straightforward and there are standard methods to proceed.

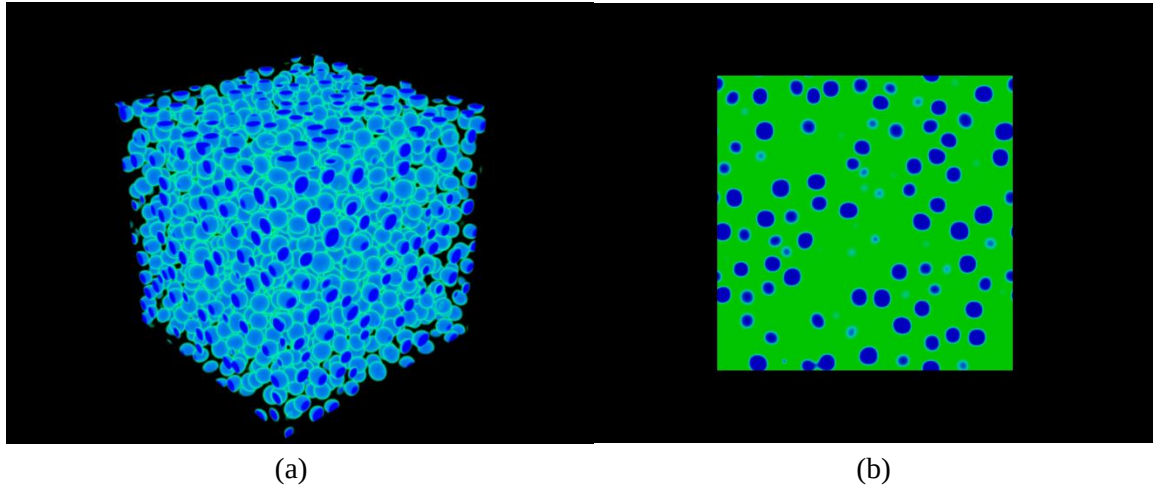


Figure 122: The γ/γ' microstructure of HA282 generated by phase-field simulation in Task 3. The γ' particles are mostly spherical as shown in (a) 3D view and (b) 2D cross-section, with a volume fraction of 18.4%. The average particle spacing and particle size (3D equivalent sphere diameter) are 38 nm and 48 nm, respectively.

In the current study, however, the complex γ/γ' two-phase microstructure of HA282, shown in Figure 122, raises a significant challenge to meshing if FEM-CP model is to be used. In addition, when structural non-uniformities at γ' precipitate level is further included, the size of a typical representative volume element (RVE) may quickly exceed the maximum affordable RVE in FEM-CP in practice. In order to overcome these difficulties, a spectral method framework was used instead of the traditional FEM. In particular, a fast Fourier transform based elasto-viscoplastic (FFT-EVP) framework²² was employed to develop a CP model for HA282. In FFT-EVP, continuum equations are solved in the *strong* form (i.e., a differential form) and the solution is approximated using the Fourier series that have global support. The advantage of using FFT-EVP is that it is an “image-based” method, meaning that no meshing is required to approximate the real-space domain but instead direct sampling on the microstructure images will be used. In addition, the used FFT algorithm provides a faster convergence compared to the FEM method²³. Finally, there is a potential advantage that FFT-EVP can be readily integrated with phase-field model, another image-based model that also utilizes the solution in Fourier space²⁴.

Kinematics

The deformation kinematics in FFT-EVP follows small-strain approximation, which is particularly valid in creep conditions. At any given time t with all equilibrium fields have been obtained, the equilibrium field at time $t + \Delta t$ is given as

$$\boldsymbol{\sigma}^{t+\Delta t}(\mathbf{x}) = \mathbf{C}(\mathbf{x}) : \boldsymbol{\varepsilon}^{e,t+\Delta t}(\mathbf{x}) = \mathbf{C}(\mathbf{x}) : [\boldsymbol{\varepsilon}^{t+\Delta t}(\mathbf{x}) - \boldsymbol{\varepsilon}^p(\mathbf{x}) - \dot{\boldsymbol{\varepsilon}}^p(\mathbf{x}, \boldsymbol{\sigma}^{t+\Delta t})\Delta t] \quad (4.1)$$

where $\boldsymbol{\sigma}(\mathbf{x})$ is the stress field, $\boldsymbol{\varepsilon}(\mathbf{x})$ is the total strain field that is equal to the sum of plastic strain $\boldsymbol{\varepsilon}^p(\mathbf{x})$ and elastic strain $\boldsymbol{\varepsilon}^e(\mathbf{x})$, and $\dot{\boldsymbol{\varepsilon}}^p(\mathbf{x})$ is the plastic strain rate. The location-dependent elastic stiffness tensor $\mathbf{C}(\mathbf{x})$ suggests that the model can account for modulus mismatch between precipitate and matrix or between grains with different orientations. In Eq. (4.1), the use of $\boldsymbol{\sigma}^{t+\Delta t}$ in determining $\dot{\boldsymbol{\varepsilon}}^p(\mathbf{x})$ implies an implicit Euler scheme that ensures numerical stability but requires iterative solution. A Green's function solution utilizing the Fourier transform to Eq. (4.1) has been presented by Lebensohn et al.²² and we will follow the approach there to solve the mechanical equilibrium. It needs to be pointed out that because of the use of FFT the RVE is required to be periodic. This turns out to be not a limit to the problem under consideration, as the γ' precipitate microstructure can always be well approximated by a periodic unit at a certain length. In addition, the 3D microstructure generated by phase-field simulation in Task 3 is also periodic. Finally, we can always use some numerical treatment such as adding a layer of air to mimicking a non-periodic boundary condition.

Constitutive model

Following the conventional crystal plasticity, the plastic strain rate is given as

$$\dot{\boldsymbol{\varepsilon}}^p(\mathbf{x}) = \sum_{\alpha=1}^{\mathcal{N}} \mathbf{m}^{\alpha}(\mathbf{x}) \dot{\gamma}^{\alpha}(\mathbf{x}) \quad (4.2)$$

where $\mathcal{N}$ is the total number of slip systems (e.g., 12 for FCC), $\mathbf{m}^{\alpha}$ is the Schmid tensor and $\dot{\gamma}^{\alpha}$ is the plastic shear rate for α -th slip system. Different constitutive models differ from each other in formulating $\dot{\gamma}^{\alpha}(\mathbf{x})$, of which the physical origin can range over various mechanisms such as dislocation slip, twinning, and phase transformation. Experimental observations and physical processes are usually considered in establishing the constitutive model.

In the case of γ/γ' superalloys such as HA282, the following experimental observations have been commonly found under creep conditions: (1) full matrix dislocations serve as the dominant plasticity carriers, (2) climb/bypass seems to dominate in the current creep conditions (1400-1500°F, 20-40ksi), and (3) at tensile test conditions, matrix glide via Orowan looping is considered as the dominant deformation mode. All these findings suggest that in HA282, with a

relatively low volume fraction of γ' precipitate, it is the interaction between full matrix dislocations and the coherent γ' particles that controls the plastic deformation. Consequently, in this study it was assumed simply that γ' particles were unshearable during both creep and tensile test conditions. It is noted that the assumption is not exactly true under the tensile condition, but we only kept it for simplicity of the model. As a result, we only needed to specify a formulation of $\dot{\gamma}^\alpha(\mathbf{x})$ that took into account the dislocation-precipitate interaction at the temperature and stress regimes of interest. Based on the dislocation-based constitutive model for simple FCC metals developed by Ma et al.²⁵ and the above specific experimental findings for HA282, we proposed the following Orowan-type kinetic equations for plastic shear rate in γ matrix:

$$\dot{\gamma}^\alpha = \begin{cases} 0, & |\tau^\alpha| \leq \tau_{\text{pass}}^\alpha \\ \dot{\gamma}_0^\alpha \exp\left[-\frac{Q_{\text{slip}}}{k_B T}\right] \sinh\left[\frac{|\tau^\alpha| - \tau_{\text{pass}}^\alpha - \tau_{\text{oro}}^\alpha}{\tau_{\text{cut}}^\alpha}\right] \text{sign}(\tau^\alpha), & |\tau^\alpha| > \tau_{\text{pass}}^\alpha \end{cases} \quad (4.3)$$

where the pre-exponential factor $\dot{\gamma}_0^\alpha = \frac{2k_B T}{c_1 c_3 \mu b^2} \nu_D \sqrt{\rho_P^\alpha}$, with the attempt frequency ν_D being in the order of Debye frequency, k_B the Boltzmann constant, T the temperature, μ the shear modulus, b the Burgers vector magnitude, and c_i fitting parameters. Here Q_{slip} is the effective activation energy for gliding through distributed obstacles consisting of parallel and forest dislocations with a density of ρ_P^α and ρ_F^α , respectively, which result in a passing stress $\tau_{\text{pass}}^\alpha = c_1 \mu b \sqrt{\rho_P^\alpha}$ and a cutting stress $\tau_{\text{cut}}^\alpha = \frac{Q_{\text{slip}}}{c_2 c_3 b^2} \sqrt{\rho_F^\alpha}$. The activation volume for slip is $c_3 b^2 \lambda_\alpha$ where $\lambda_\alpha = c_2 / \sqrt{\rho_F^\alpha}$.

Compared with the original formulation, the new feature in Eq. (4.3) is the introduction of τ_{oro}^α , which accounts for the dislocation-precipitate interaction. More specifically, in tensile test conditions, τ_{oro}^α should reflect the underlying Orowan looping mechanism, and in creep test conditions it should reflect the mechanism of dislocation climb/bypass over γ' . The reason why this additional slip resistance appears as the similar role played by the passing stress $\tau_{\text{pass}}^\alpha$ is that the underlying mechanisms all fall into the so-called “diffuse obstacles” scenario^{26,27}. This is in contrast to the physical scenario represented by the forest dislocation cutting stress τ_{cut}^α (and similarly particle cutting such as APB shearing) that is related to the “strong obstacles”.

In the next subsection, we will discuss how to determine τ_{oro}^α , which is a key parameter containing the microstructure-sensitivity. To complete the formulation of constitutive model, the equations for defining and evolving dislocation densities are needed and we follow the original work by Ma et al.²⁵ that is summarized below. The parallel and forest dislocation density are given as

$$\rho_F^\alpha = \sum_{\beta}^{\mathcal{N}} \rho_{\text{SSD}}^\beta |\cos(\mathbf{n}^\alpha, \mathbf{t}^\beta)| + |\rho_{\text{GNDs}}^\beta \cos(\mathbf{n}^\alpha, \mathbf{d}^\beta)| + |\rho_{\text{GNDet}}^\beta \cos(\mathbf{n}^\alpha, \mathbf{t}^\beta)|$$

$$\rho_P^\alpha = \sum_{\beta}^{\mathcal{N}} \rho_{\text{SSD}}^\beta |\sin(\mathbf{n}^\alpha, \mathbf{t}^\beta)| + |\rho_{\text{GNDs}}^\beta \sin(\mathbf{n}^\alpha, \mathbf{d}^\beta)| + |\rho_{\text{GNDet}}^\beta \sin(\mathbf{n}^\alpha, \mathbf{t}^\beta)|$$

where the statistically-stored dislocation (SSD) density ρ_{SSD}^α and geometrically-necessary dislocation (GND) density ρ_{GND}^α for a given slip system α evolve according to

$$\begin{aligned} \dot{\rho}_{\text{SSD}}^\alpha &= c_4 \sqrt{\rho_F^\alpha} \dot{\gamma}^\alpha + c_6 d_{\text{dipole}}^\alpha \rho_M^\alpha \dot{\gamma}^\alpha - c_5 \rho_{\text{SSD}}^\alpha \dot{\gamma}^\alpha \\ &\quad - c_7 \exp \left[-\frac{Q_{\text{bulk}}}{k_B T} \right] \frac{|\sigma_{\text{VM}}^\alpha|}{k_B T} (\rho_{\text{SSD}}^\alpha)^2 (\dot{\gamma}_{\text{VM}}) c_8 \\ \dot{\rho}_{\text{GNDs}}^\alpha &= -\frac{1}{b} \nabla \dot{\gamma}^\alpha \cdot \mathbf{t}^\alpha, \quad \dot{\rho}_{\text{GNDet}}^\alpha = \frac{1}{b} \nabla \dot{\gamma}^\alpha \cdot \mathbf{d}^\alpha, \quad \dot{\rho}_{\text{GNDen}}^\alpha = 0 \end{aligned}$$

where “VM” represents von-Mises equivalent stress/strain, Q_{bulk} is the activation energy for self-diffusion, and $\mathbf{d}^\alpha$ and $\mathbf{t}^\alpha$ are respectively slip direction and sense vector of slip system α such that the GND density can be decomposed into three components under the constraint that $(\rho_{\text{GND}}^\alpha)^2 = (\rho_{\text{GNDs}}^\alpha)^2 + (\rho_{\text{GNDet}}^\alpha)^2 + (\rho_{\text{GNDen}}^\alpha)^2$. The non-local operator (gradient) for GND evolution determines the physical length of FFT-EVP simulation. The mobile dislocation density ρ_M^α , following the principle of maximum plastic dissipation during the plastic deformation, scales with immobile dislocation densities as $\rho_M^\alpha = \frac{2k_B T}{c_1 c_2 c_3 \mu b^3} \sqrt{\rho_P^\alpha \rho_F^\alpha}$. As a result, the rate equations for SSD and GND densities, together with the flow rule, complete the dislocation-based constitutive laws for our FFT-EVP model.

Inter-particle spacing and determination of τ_{oro}^α

In many existing models that consider precipitate hardening, a geometric parameter called “inter-particle” spacing, λ_p , plays a critical role in determining various threshold stresses in several strengthening mechanisms in a dispersion-hardened alloy, including Orowan looping. An analytic expression for λ_p is usually used, namely ²⁸,

$$\lambda_p = 1.6 r_p \left[\sqrt{\frac{\pi}{4f}} - 1 \right] \quad (4.4)$$

where r_p is the average particle radius, and f is the volume fraction of precipitates. While being simple and easy to use, Eq. (4.4) has a limitation when it is applied to a full-field study with structural non-uniformities, where microstructure images of spatial dispersion of particles are to be taken into

account. This is clearly seen if one notices that in Eq. (4.4) the inter-particle spacing is only determined by the average particle radius and volume fraction, regardless of how precipitates are distributed throughout the matrix. As a result, the λ_p determined via Eq. (4.4) was not suitable in our study of microstructural influence on mechanical properties for HA282 weldment.

In order to address this issue, we used a new approach of calculating λ_p so that the underlying spatial arrangement of γ' particles could be included. To this end, we employed the so-called nearest-neighbor (NN) distance $d_{NN}(\mathbf{x})$ as a location-dependent measure of *local* $\lambda_p(\mathbf{x})$, i.e.,

$$\lambda_p(\mathbf{x}) = d_{NN}(\mathbf{x}) \quad (4.5)$$

where $d_{NN}(\mathbf{x})$ can be calculated for a given γ/γ' microstructure image. As a demonstration, Figure 123 shows the calculated $d_{NN}(\mathbf{x})$ for three 3D HA282 microstructures simulated using phase-field from Task 3. Here even for the homogeneously distributed γ' microstructures, the calculated spatial distribution of $d_{NN}(\mathbf{x})$ exhibits local fluctuations. As a result, the inter-particle spacing defined in Eq. (4.5) indeed captures local microstructural information.

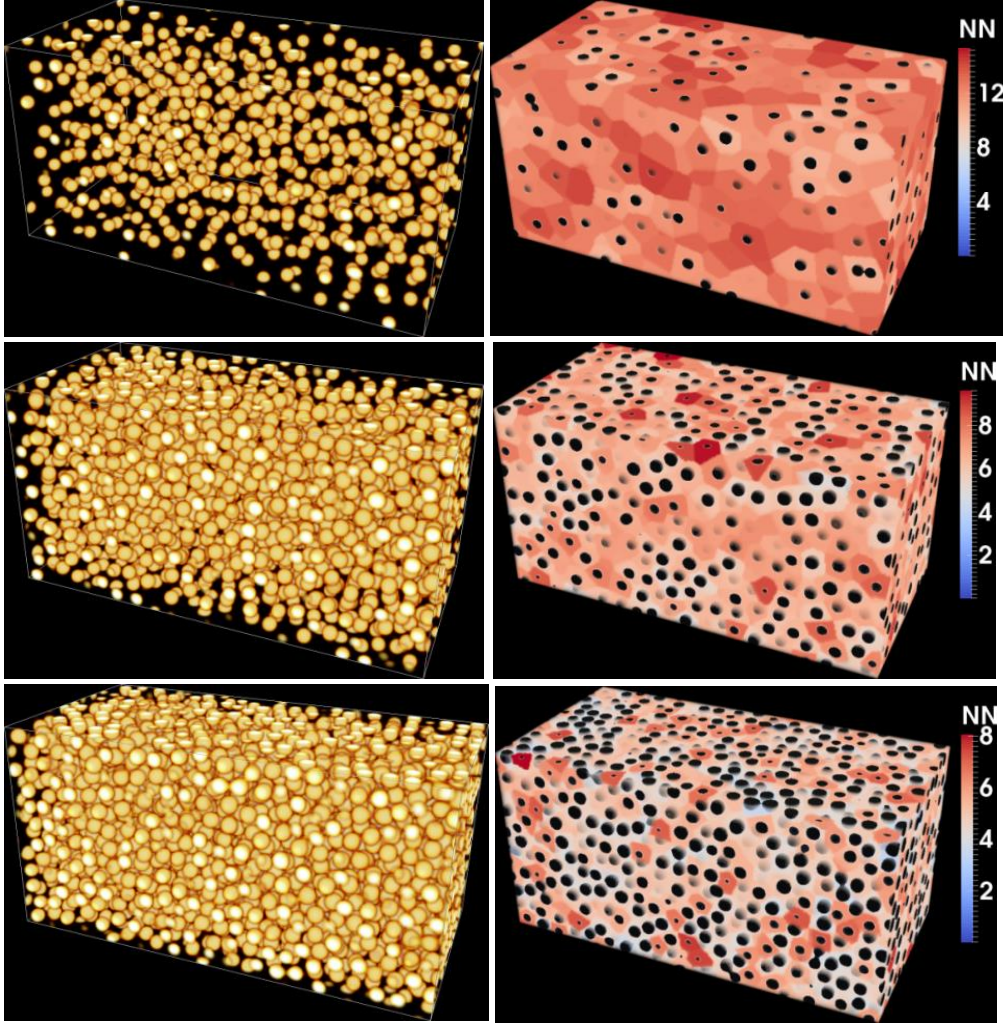


Figure 123: Phase-field generated homogeneous HA282 microstructure (left column) with different γ' volume fraction. The corresponding nearest-neighbor distance map is shown in the right column.

It is also worth pointing out that once the location-dependent $\lambda_p(\mathbf{x})$ is obtained, we can perform a volume-average to obtain an average inter-particle spacing, i.e.,

$$\bar{\lambda}_p = \langle \lambda_p(\mathbf{x}) \rangle_{\mathbf{x}} = \langle d_{\text{NN}}(\mathbf{x}) \rangle_{\mathbf{x}} \quad (4.6)$$

This microstructural descriptor can then be used for grain-level study of deformation behavior, e.g., macroscopic constitutive modeling in Task 5, instead of using the conventional Eq. (4.4). We will show later that inter-particle spacing defined by Eq. (4.6) has exhibited the desired microstructure-sensitivity, while the one defined by Eq. (4.4) was unable to account for the underlying microstructure non-uniformities.

Given a local inter-particle spacing $\lambda_p(\mathbf{x})$, the Orowan stress τ_{Oro} in Eq. (4-3) can now be obtained following previous works ²⁹:

$$\tau_{\text{Oro}}(\mathbf{x}) = \frac{0.84\mu b}{2\pi(1-\nu)^{\frac{1}{2}}} \frac{\ln\left(\frac{R_0}{r_0}\right)}{\lambda_p(\mathbf{x})} \quad (4.7)$$

where R_0 and r_0 are respectively the outer and inner cutoff radius for the elastic energy of a single dislocation ³⁰. Because of the use of $\lambda_p(\mathbf{x})$, it is now expected that regions where local γ' particles are more sparsely distributed will have less resistance to dislocation slip across the precipitates.

4.1.2. Model calibration of FFT-EVP

After the constitutive model was formulated, several unknown parameters were determined by calibrating the FFT-EVP model to experimental data. These parameters are related to specific physical processes, but can only be estimated in terms of orders of magnitude because atomic description is missing in a continuum model. Specifically, we fit the FFT-EVP model to the tensile and creep test data of HA282 base metal developed in Task 2. Upon successful calibration, one set of model parameters should be able to yield the correct temperature and stress dependence in the experimental data. It is worth mentioning that at the first glance the number of fitting parameters in the constitutive model described in Section 4.1.1 seems to be relatively large compared to some other CP models with simplified internal state variables (e.g., critical resolved shear stress). However, for real application when temperature and stress-dependence are desired, those “simple” models usually need a temperature dependence on the internal variables, leading to a significant increase of the total number of fitting parameters. In contrast, our constitutive model has the temperature and strain-rate effects automatically built in owing to the use of thermally activated deformation, and there is no need to explicitly introduce the temperature or strain-rate dependence in those parameters in Section 4.1.1.

Figure 124 compares the simulated and the experimental tensile tests on HA282 base metals at room-temperature (RT), 1500, 1600, and 1700°F. The tests were subject to strain-controlled boundary conditions with the applied strain rate being 3×10^{-4} /s. Except for the “abnormal” yield behavior at 1700°F, the temperature-dependent yield stress ranging from RT to 1600°F was correctly reproduced by the calibrated FFT-EVP model. In addition, the hardening post peak-stress was also captured quantitatively from RT to 1600°F. The deviation of the hardening behavior during the later stage at RT may have been due to the grain boundary hardening, which was not considered in the FFT-EVP model. It also needs to be pointed out that at 1500, 1600, and 1700°F, the experimental stress-strain curves

showed softening, which may be related to the dynamic recrystallization or even other microstructural effects. No further investigation has been carried out experimentally, since this softening behavior was not the relevant for creep. The agreement of the yield and early hardening between simulation and experiments suggested that the calibrated parameters in the constitutive model had captured the correct dislocation storage and annihilation dynamics in the precipitate-hardened HA282.

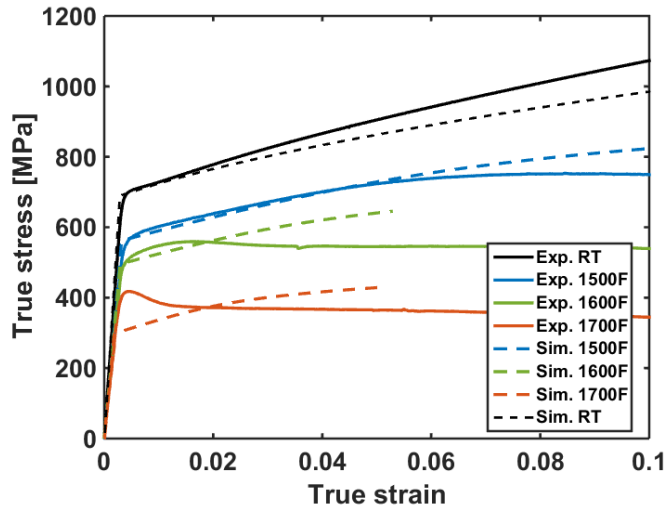


Figure 124: Comparison between experimental and simulated tensile tests on HA282 base metals at room-temperature (RT), 1500, 1600, and 1700°F. The tests are subject to strain-controlled boundary conditions with the applied strain rate being $3 \times 10^{-4}/s$

As a direct prediction, the dislocation density evolution at RT and 1500°F is shown in Figure 125, corresponding to the relevant stress-strain curves in Figure 124. It is seen clearly that at both temperatures, the mobile dislocation content is only a small fraction of the immobile dislocations that consist of SSD and GND, which is consistent with the model assumption. The spatial distribution of strain and dislocations are also direct prediction of the full-field FFT-EVP simulations. Figure 126 shows the distribution of ε_{zz} component during tensile test (snapshot at macroscopic elongation of 3.5%) at RT and the corresponding GND distribution.

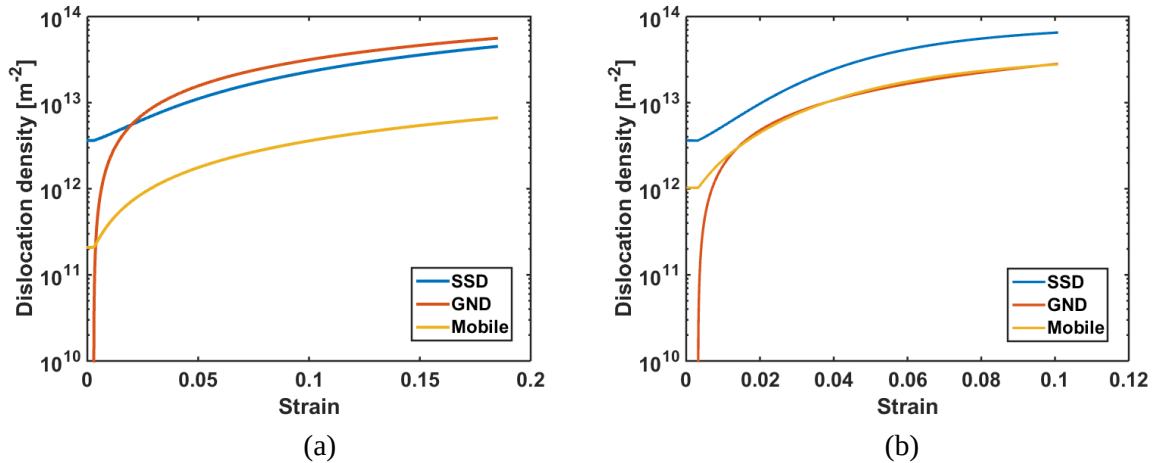


Figure 125: Simulation prediction of dislocation density evolution at (a) RT and (b) 1500F. SSD = statistically stored dislocations; GND = geometrically necessary dislocations.

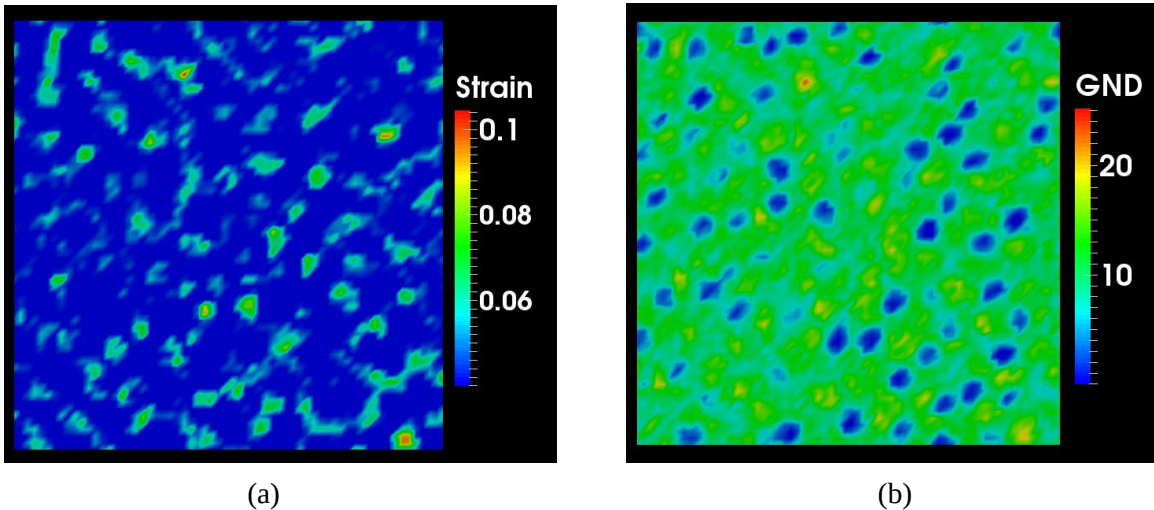


Figure 126: Simulation prediction of spatial distribution (2D cross-section) of (a) ϵ_{zz} and (b) GND at RT when the macroscopic elongation is 3.5%. The applied stress is along z-axis and is normal to the cross-section.

Using the calibrated model parameters, we carried out simulations of the creep tests at 1350, 1400, and 1450°F. The results are shown in Figure 127. Again, it is seen that the calibrated model captured well the temperature and stress dependence of creep behaviors up to ~1000 hours (or the early secondary creep). Since in the FFT-EVP the microstructures were treated to be static, the simulation could only be applied to the early stages of creep. As creep deformation continues, the underlying microstructure can undergo significant changes even in a low γ' superalloy. For example, precipitate particle coarsening will always occur at elevated temperatures; grain boundary cavities will eventually nucleate and grow into cracks, which is the dominant reason of the tertiary creep. To consider a full creep curve that is comparable with

experimental measurement, integration of FFT-EVP with phase-field microstructure model is necessary. Figure 128 shows the simulated creep strain distributions at two different stress levels at 1400°F. Compared with the strain patterns in tensile tests in Figure 126(a), the strain in creep exhibits a different vein-like pattern.

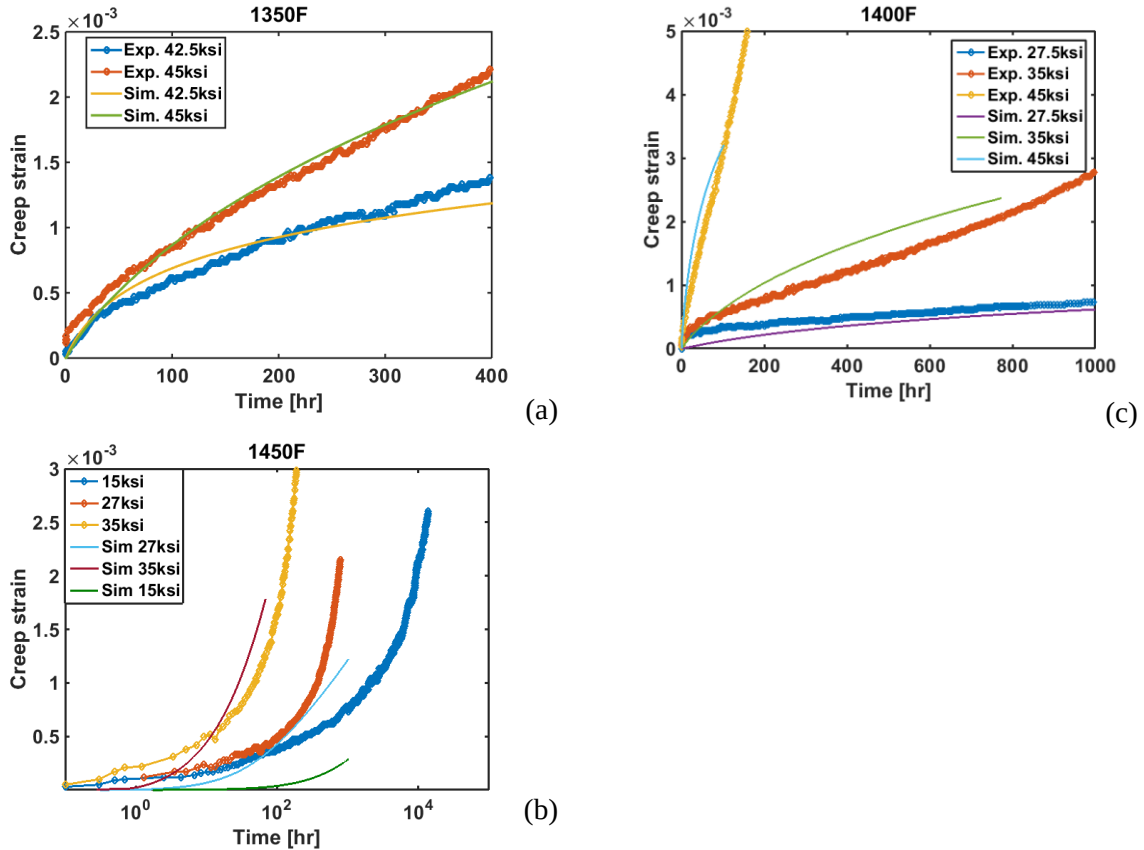


Figure 127: Comparison between simulated and experimental creep curves of HA282 base metals at (a) 1350F, (b) 1400F, and (c) 1450F at various stresses.

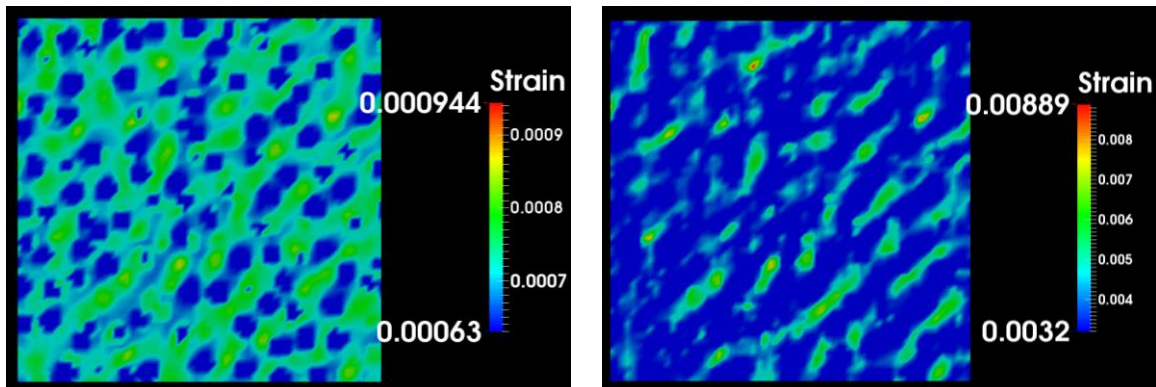


Figure 128: Simulation prediction of spatial distribution (2D cross-section) of ϵ_{zz} during creep at 1400F subject to (a) 27.5ksi and (b) 45ksi.

4.1.3. FFT-EVP model application to HA282 weldment

As has been shown by experimental and simulation studies in Tasks 2 and 3, the structural heterogeneities in the weld zone largely depend on the post-weld heat treatment procedures. Features such as microsegregation and γ' denuded zones at grain boundary (GB) were observed and simulated in Tasks 2 and 3. Nevertheless, their effects on the subsequent plastic deformation were difficult to evaluate experimentally. Using the FFT-EVP model that were successfully calibrated against both tensile and creep test data of HA282 base metals at the temperature and stress levels of interest, we explored those effects by using the simulated HA282 microstructure containing the desired microstructural non-uniformity of interest generated by the phase-field model in Task 3. In this project, we have specifically investigated the effects of two types of structural non-uniformities, namely, microsegregation and γ' denuded zones at GB. To establish a correlation between the degree of non-uniformity and the influence on the plastic deformation, parametric studies have been carried out.

Influence of microsegregation on tensile and creep properties

In Task 3, EPMA maps have shown microsegregation of different alloy elements at dendrite core and interdendritic regions. Microsegregation and subsequent γ' precipitation lead to a non-uniform distribution of γ' particles, which has also been confirmed by the microstructure simulations in Task 3. Figure 129 shows a sinusoidal-like distribution of γ' precipitates for HA282 weld sample subject to seven different levels of residual microsegregation, of which the Set3-2 corresponds to the experimental PWHT3a in Task 2. In all cases, the volume fraction of precipitate is 15%. We also define the so-called degree of microsegregation Δf as indicated in Figure 129.

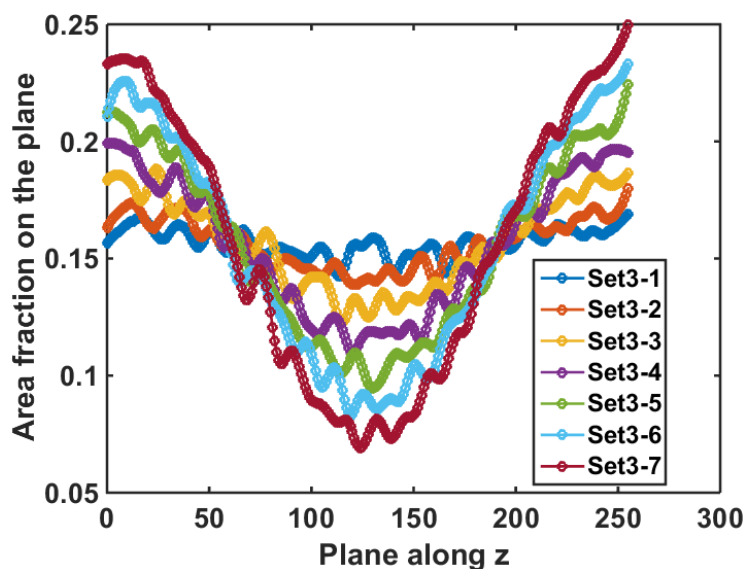


Figure 129: Variation of local γ' volume fraction (area fraction on a 2D plane) when crossing the interdendritic and dendrite core regions. Here 7 sets of phase-field simulations are carried out for the purpose of parametric study. The experimental PWHT3a corresponds to Set3-2. The degree of micro-segregation Δf is defined as an example for Set3-6.

Before carrying out the mechanical simulations, we analyzed the microstructure using the new location-dependent inter-particle spacing $\lambda_p(\mathbf{x})$ introduced in Section 4.1.1. Figure 130 shows three HA282 microstructures that contain sinusoidal distribution of γ' particles, together with the corresponding $\lambda_p(\mathbf{x})$ maps. Clearly, the $\lambda_p(\mathbf{x})$ maps reflect the underlying microsegregation and inhomogeneity of spatial γ' distribution. In addition, at the macroscopic level, we can examine the microstructure sensitivity of the average inter-particle spacing defined by Eqs. (4.4) and (4.6). The result is shown in Figure 131. It is obvious that using the analytic inter-particle spacing defined in Eq. (4.4), the calculated geometric descriptor for these 3D HA282 micro-segregated microstructures showed little variation since the volume fraction did not change much ($\sim 15\%$). On the contrary, the volume average of nearest-neighbor distance defined in Eq. (4.6) clearly showed that as the degree of microsegregation increased, γ' precipitate tended to have smaller inter-particle spacing overall, as had been expected.

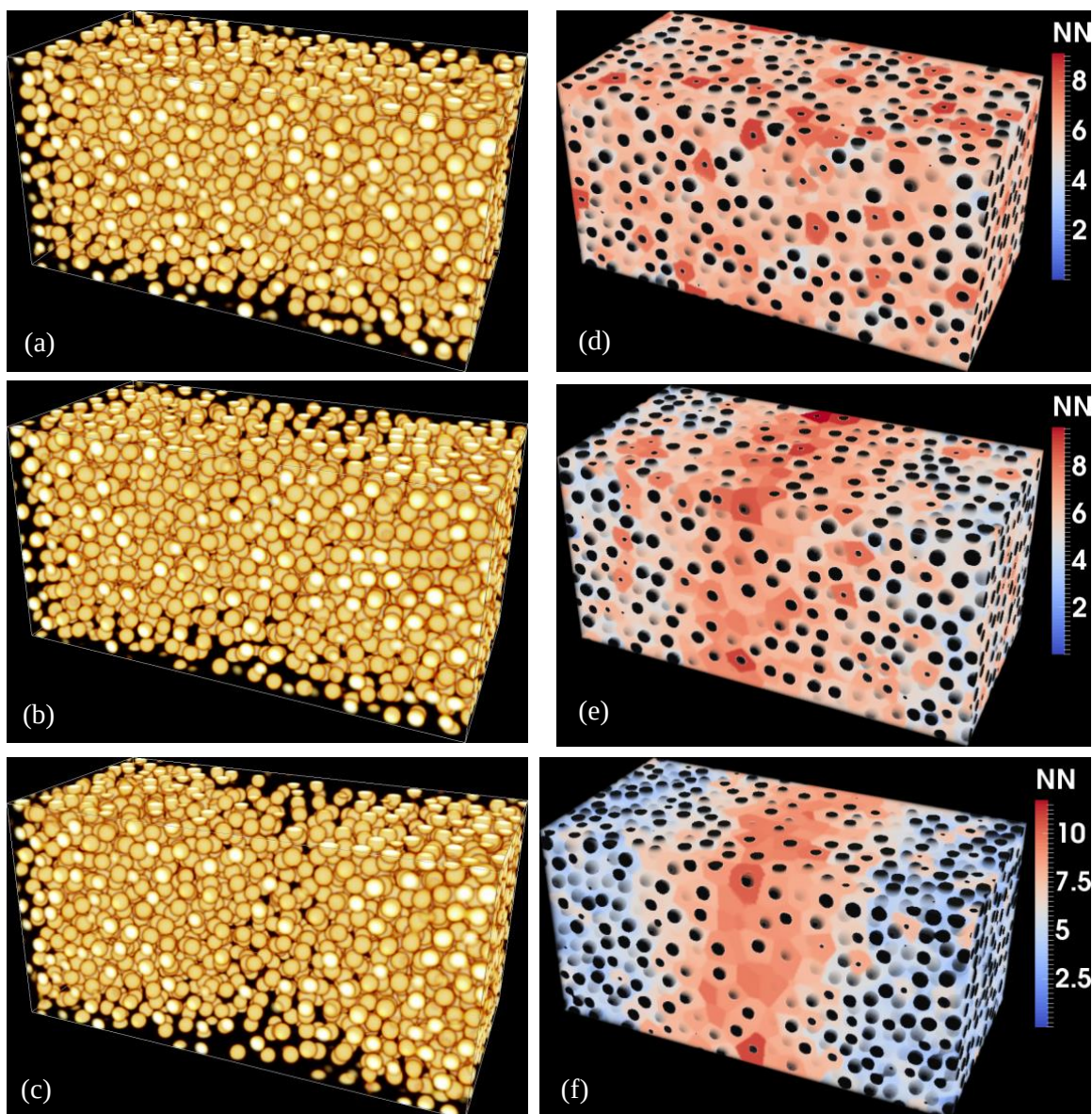


Figure 130: (a-c) Phase-field generated HA282 microstructure with different sinusoidal precipitate distribution and (e-f) the corresponding inter-particle spacing maps. The volume fraction remains the same as 15%.

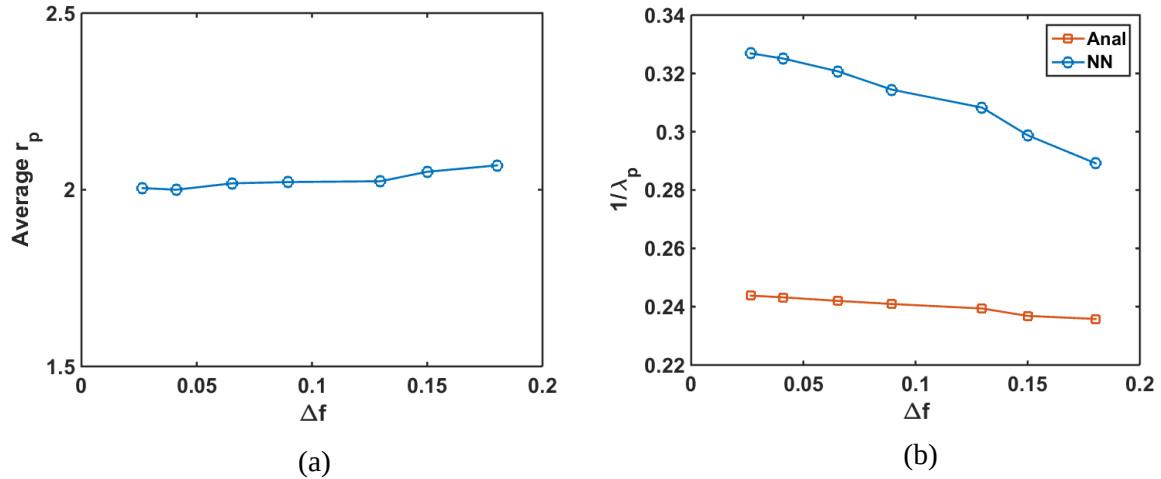


Figure 131: (a) Average γ' particle radius in simulated HA282 microstructure with different degree of micro-segregation in Fig. 8, and (b) the corresponding inverse of average inter-particle spacing.

We then carried out tensile test simulations at 1500°F using the above 3D HA282 micro-segregated microstructures. The simulated stress-strain curves are shown in Figure 132. With an increase in the degree of microsegregation, both the yield strength (YS) and flow stress decreased, but the hardening rate appeared to remain the same. Figure 133 compares the 0.2% YS versus the degree of micro-segregation. The lack of a significant degradation of YS in the predictions appeared to be consistent with the actual observation in Task 2.

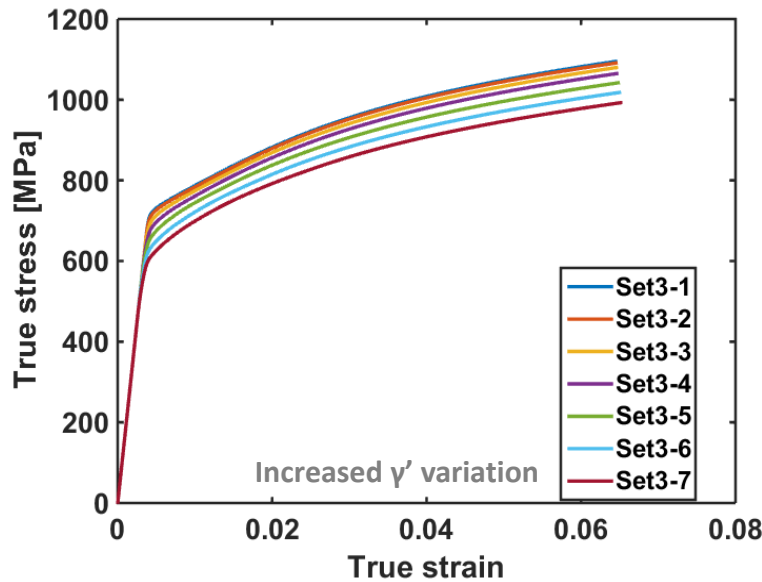


Figure 132: Simulated tensile tests at 1500°F on HA282 microstructures with different degree of microsegregation in Figure 129.

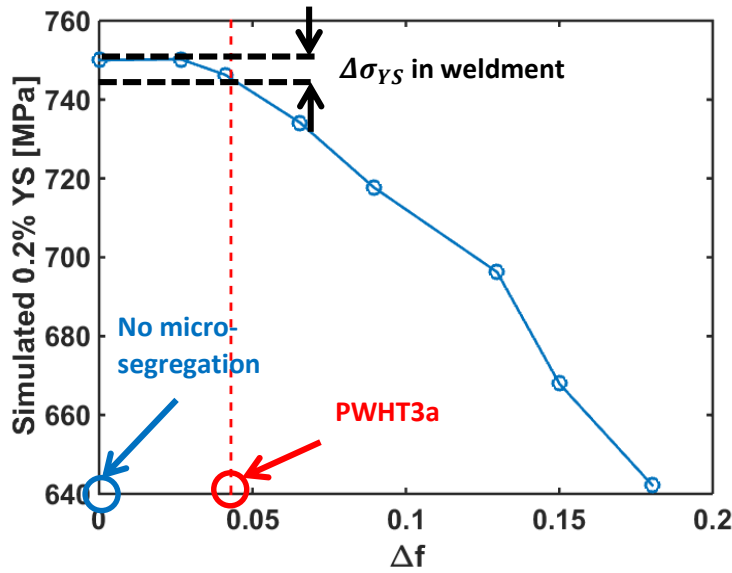


Figure 133: Simulated 0.2% yield strength versus degree of microsegregation Δf .

As another demonstration on the advantage of using the Eq.(4.6) for the microstructural descriptor, we analyzed the YS versus $1/\lambda_p$ using the simulation data for the microsegregation (Figure 132), which is shown in Figure 134. Additionally, we carried out a set of tensile test simulations using homogeneous HA282 microstructure at different volume fraction of γ' . The obtained YS versus $1/\lambda_p$ data is also included in Figure 134. In these cases, we used both formulations for λ_p , by Eqs. (4.4) and (4.6), as shown in Figure 134. It is seen that when λ_p is used as a volume average of the location-dependent nearest-neighbor distance, it can give a consistent scaling relationship between YS and $1/\lambda_p$, which however is not achieved if Eq. (4.4) is used. As a result, it was concluded that Eq. (4.6) was a better microstructural descriptor for HA282 for it captured the underlying microstructure spatial information.

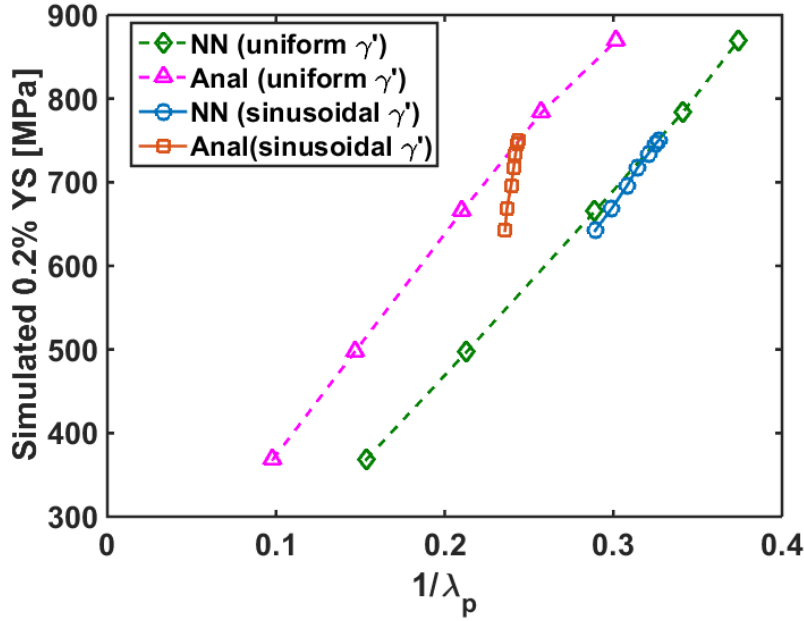


Figure 134: Simulated 0.2% yield strength versus the inverse of inter-particle spacing that is calculated using either Eq. (4.4) or Eq. (4.6).

Using the same HA282 microstructures, we then carried out a set of creep test simulations at 1500°F at a stress of 30ksi. The simulated creep curves up to 1000 hours are shown in Figure 135(a). It suggests that microsegregation accelerates the accumulation of creep strain, which is more clearly shown in Figure 135(b) where the primary and the secondary creep rates are plotted against the degree of microsegregation Δf . It is seen that the microsegregation has a significant influence on the primary creep rate while the secondary creep rate remains virtually the same regardless of the presence of micro-segregation. This is due to the fact that primary creep is closely related to the initial microstructure of the material (or the initial condition of the simulation); the presence of the non-uniformity can lead to strain localization and hence a fast accumulation of creep strain at regions where local volume fraction is significantly lower than the average. Nevertheless, as the matrix hardens during the accumulation of creep strain, creep rate will slow down. Since dislocation storage and annihilation mechanisms remain the same for all samples in spite of the non-uniformity difference, eventually the creep rate will reach a same value as strain-hardening occurs predominantly in the matrix.

Finally, in Figure 136 the spatial distribution of strain component ε_{yy} is shown for three different microstructures in Figure 129. Clearly, as the γ' variation increases, there is a strong strain localization in the region where the γ' volume fraction is low (the dendrite core).

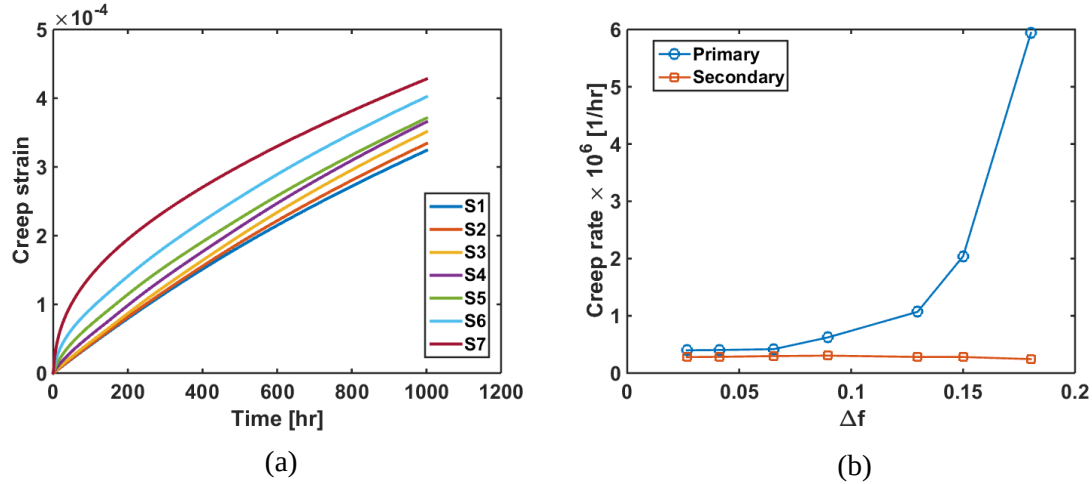


Figure 135: (a) Simulated creep curves at 1500F, 30ksi on HA282 microstructures with different degree of micro-segregation in Figure 129, and (b) the corresponding primary and secondary creep rates variation.

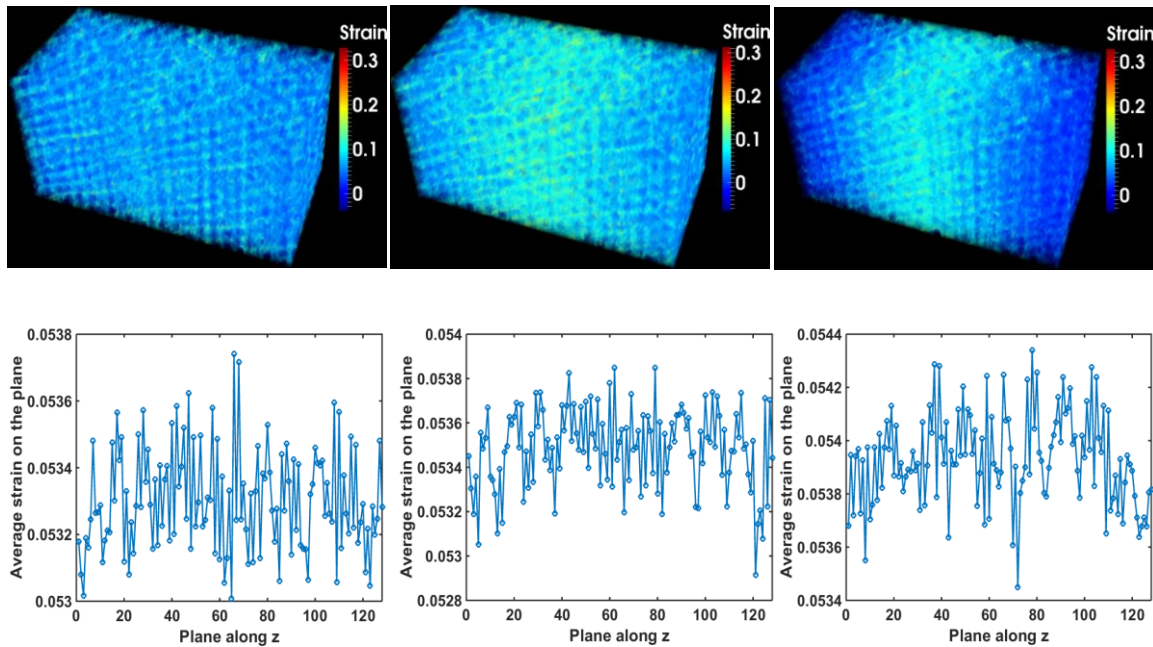


Figure 136: Spatial distribution of ϵ_{yy} and the corresponding variation of plane average along the dimension of volume fraction variation. (The uniaxial tension is along y-axis.)

Influence of grain boundary denuded zone on tensile and creep properties

In the characterizations of creep samples of welded HA282 in Task 2, the grain boundary (GB) denuded zones (DZ) were frequently observed and were found to play some important roles in influencing plastic deformation of the superalloy. In order to study and assess the effect of GBDZ, we carried out a set of parametric

simulations, following the configuration described in Figure 137. To separate the effect of DZ from that of the orientation mismatch of the two neighboring grains, we assumed that the two grains had exactly the same orientation. Additionally, based on the experimental observation, the interior grain region was kept with a constant volume fraction ($\sim 16\%$) of γ' for all microstructures but with different width of DZ. As an example, Figure 138 shows three HA282 microstructures that were generated using a statistical approach. We also need to point out that in Figure 138 the width of the DZ has been normalized by the average radius of γ' particles.

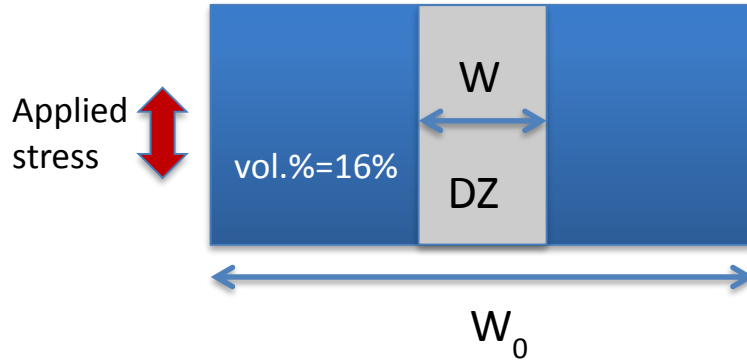


Figure 137: Simulation configuration setup considering the grain boundary denuded zone.

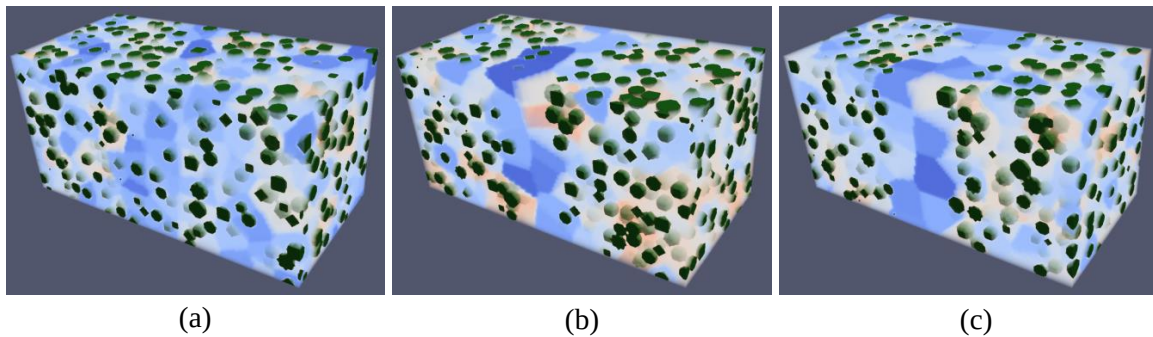


Figure 138: Synthetic bicrystal HA282 microstructure containing a DZ in the center, with a size of w_{dz}/r_p equal to (a) 2.32, (b) 6.97, and (c) 9.29. Green γ' particles are embedded in the γ matrix colored by the local inter-particle spacing $\lambda(x)$.

We first carried out a series of simulations of uniaxial tensile testing with those synthetic microstructures. The simulated stress-strain curves at 1500°F are shown in Figure 139(a) for different width of DZs, and the corresponding 0.2% YS is plotted against the width of DZ (w_{dz}/r_p) in Figure 139(b). It is seen that as the (relative) width of DZ increases, the YS and the overall flow stress decrease. In particular, the decrease of YS as a function of DZ width shows approximately a linear fashion in the range simulated here and for the largest DZ (10 times of particle radius) case, the YS is decreased by $\sim 10\%$. A direct comparison

between this simulation result and experiment is difficult because the real welded HA282 material is polycrystals and GBDZs could be present in any types of GBs.

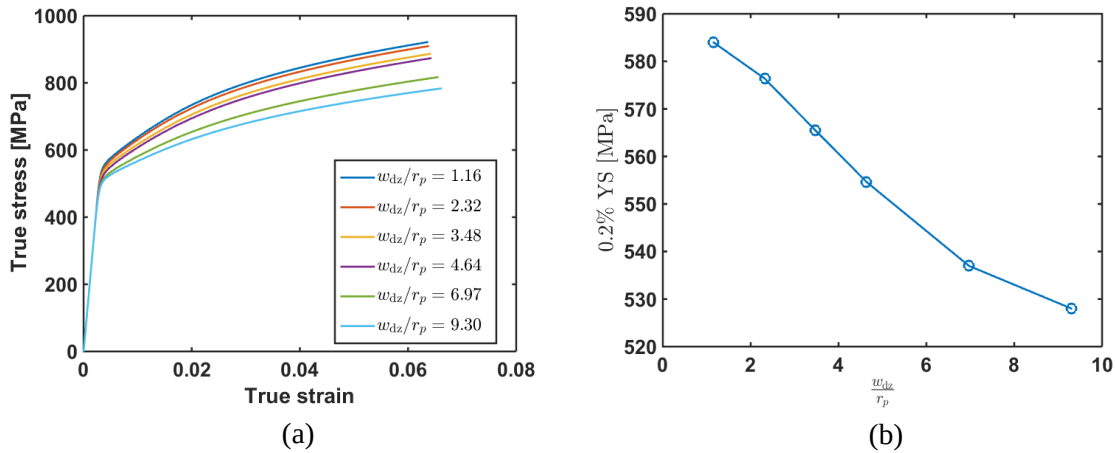


Figure 139: (a) Stress-strain curves from simulated uniaxial tensile tests at 1500F for various bicrystal HA282 containing a DZ with different width. (b) The variation of 0.2% yield stress from (a) with respect to the width of DZ.

A useful prediction from FFT-CP is the distribution of plastic deformation that is difficult to measure in experiments. In Figure 140, we plot the 3D distribution of strain component ϵ_{yy} in the uniaxial tensile test simulations, where the tensile loading is along y-axis. A 2D cross-section of the strain distribution is also plotted in Figure 140 for better visualization. It is clear that plastic deformation is more localized in GB DZ regions, as we would expect due to the lower resistance to dislocation activities in the absence of particle-strengthening. It is also interesting to see that for the microstructure with larger DZ, the plastic deformation is more heterogeneous. It needs to be pointed out that FFT-CP requires an input microstructure to be periodic in space. In the synthetic microstructure, the total size was fixed and thus changing the width of DZ also changed the overall volume fraction of γ' particles. (This could be a limit of the current simulation and larger scale simulations are required with a much more computational cost.) As larger volume fraction of DZs were available to allow more localized plastic deformation, the matrix or bulk material experienced less amount of plastic deformation, for the overall applied deformation was assumed to be same.

The creep deformation at 1500°F and 30ksi was also simulated with the above synthetic microstructures. The simulated creep-curves are shown in Figure 141(a) and the corresponding creep rate as a function of DZ width is shown in Figure 141(b). Due to the computational cost of the full-field 3D simulation, the creep deformation was only simulated up to the stage where transition to secondary creep was just finished. Clearly, the presence with larger DZ led to higher creep rate, indicating a deleterious effect of GBDZ on creep resistance of the precipitate-hardened alloy. The interesting finding here is that as the width of

GBDZ increases, the primary creep rate shows a significant increase while the secondary creep rate remains relatively unchanged. This is similar to the effect of degree of non-uniform γ' on creep rates discussed in the previous section. Since as mentioned previously, our synthetic microstructure with larger DZ had less amount of γ' particles, the increase of primary creep rate could be simply due to this reduction of average γ' volume fraction. The influence of GBDZ on secondary creep rate was not as significant as on primary creep rate, which suggested that as strain hardening occurs, the initially high local plastic deformation rate in DZ regions quickly reduced to the similar level of that in the bulk region.

Nevertheless, creep damage (e.g., cavity or void) initiations are local events and thus we should again inspect the distribution of local strain, as a driving force, to analyze the influence of GBDZ. In Figure 142, the 3D distribution of ϵ_{yy} in two simulations ($w_{dz}/r_p = 2.32$ and 4.63) of Figure 141(a) are shown together with a 2D cross-section plot. Clearly, plastic deformation is more localized in DZ regions, where the local accumulated plastic strain can be roughly three times of that in the bulk. Since larger local plastic strain can lead to a higher probability of void nucleation, the current simulation suggests that the presence of GBDZ may promote the damage formation such as GB voids. In fact, experimental characterization on various ruptured samples of both base and welded metals has suggested that most GB voids/cracks were at boundaries that DZ presented. Such statistical correction could be related to the different local plastic behavior as shown in the above simulations. However, modeling nucleation of GB void is a very complex process and requires substantial development in a future work.

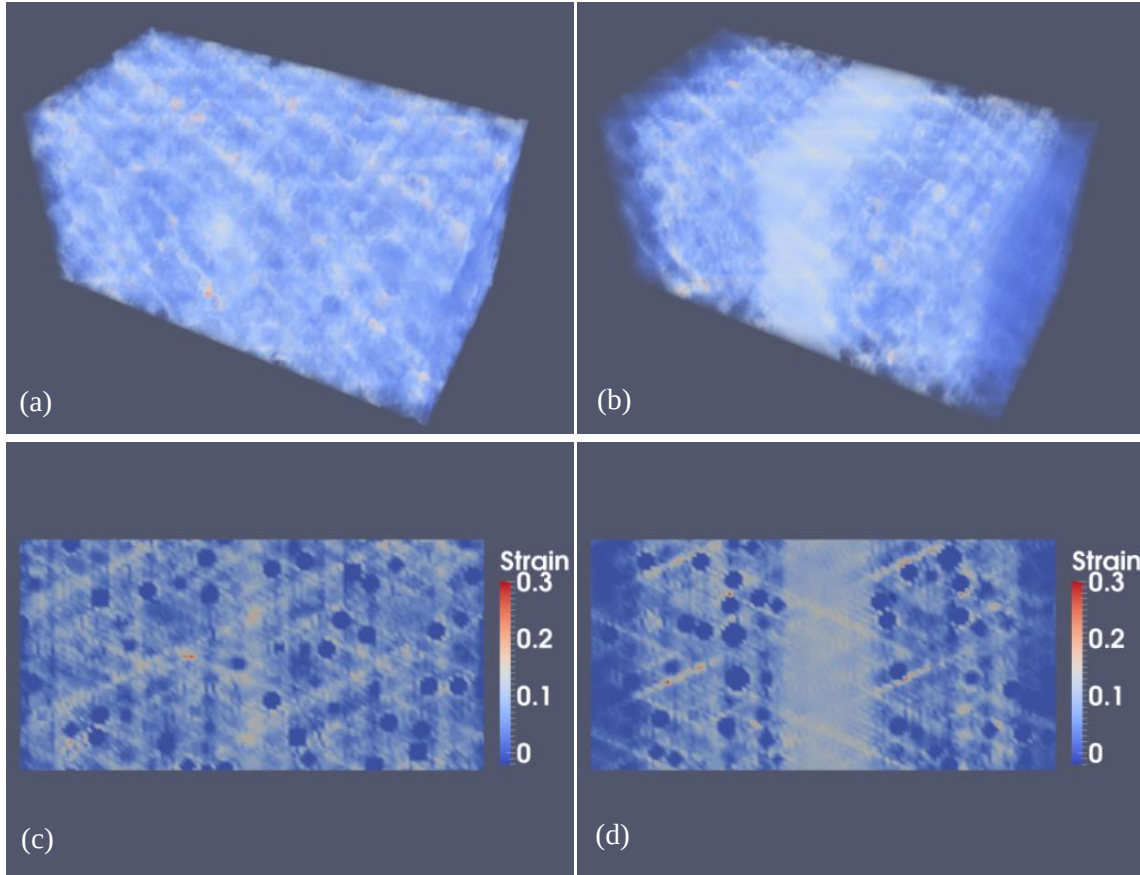


Figure 140: Three dimensional distribution of ε_{yy} for uniaxially elongated (up to ~6%) HA282 containing a DZ with a width w_{dz}/r_p of (a) 2.32 and (b) 9.29. (c) and (d) are 2D cross-section of, respectively, (a) and (b). The tensile loading is along vertical direction in (c) and (d).

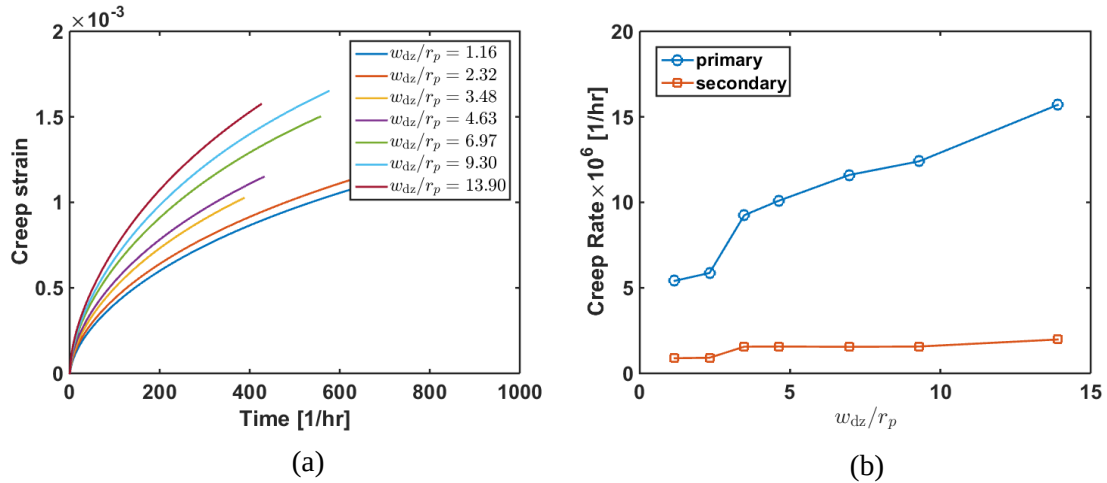


Figure 141: (a) Simulated creep curves for synthetic bicrystal HA282 containing DZ with different size, subject to 30ksi tension at 1500F. (b) Primary and secondary creep as a function of DZ width.

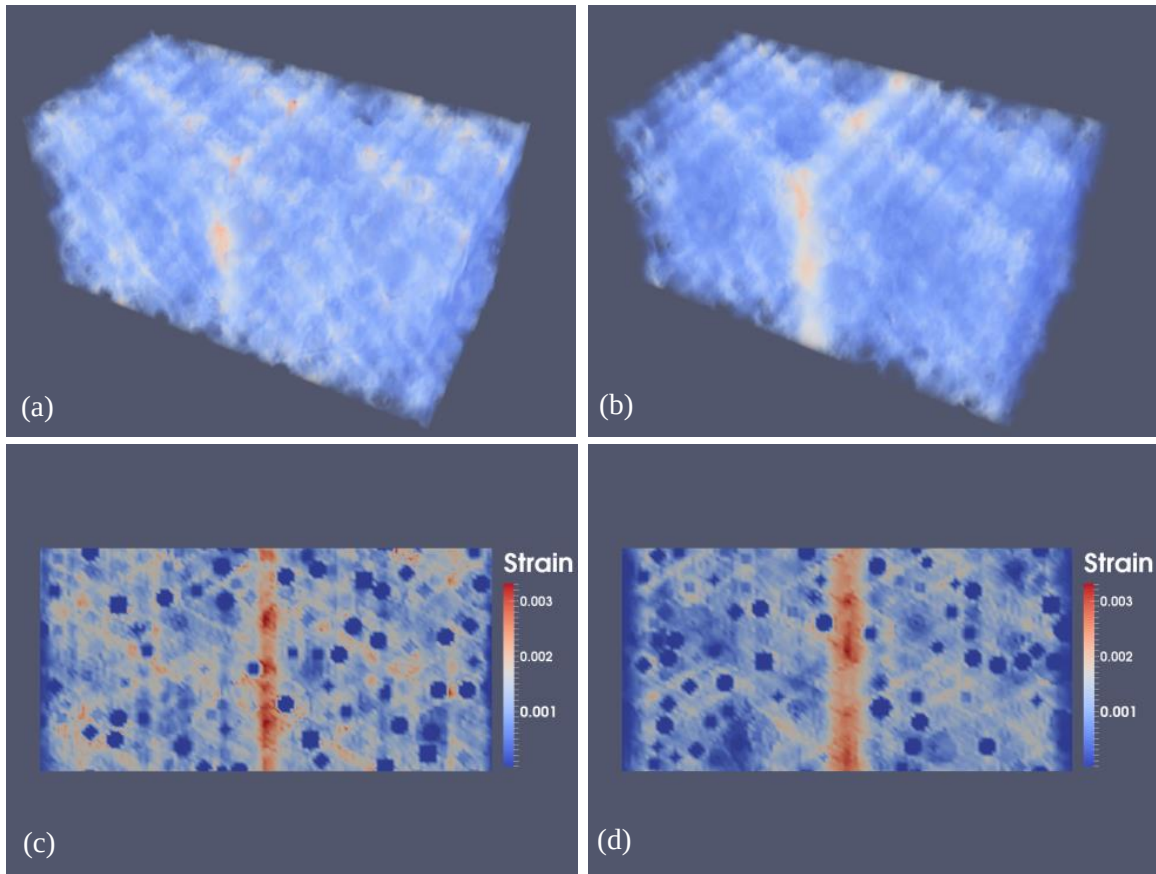


Figure 142: Three dimensional distribution of ϵ_{yy} during creep (up to $\sim 0.12\%$ total creep strain) in synthetic HA282 containing a DZ with a width w_{dz}/r_p of (a) 2.32 and (b) 4.63. (c) and (d) are 2D cross-section of, respectively, (a) and (b). The tensile loading is along vertical direction in (c) and (d).

4.2. Creep cavitation modeling using phase-field method

The creep strain evolution, especially in the tertiary stage, is largely determined by the gradually formed or accumulated damages during long-term service exposure at elevated temperatures. An accurate systematic prediction of rupture time at various testing conditions requires a clear understanding of the evolution of these damages. In metals subject to creep, cavities usually nucleate at grain boundaries and grow under certain conditions and eventually interlink and form cracks, leading to catastrophic failure³¹. The grain boundary cavitation has been found in HA282 during creep in the temperature range 1000-1700°F both by existing works³² and in the experimental studies of Task 2. The nucleation mechanism of creep cavities is still not well established, even though they are extensively found at grain boundaries, particularly on those transverse to a tensile stress, as schematically shown in Figure 143. For commercial alloys such as HA282, it is also possible that cavities are associated with second phase particles. Nevertheless, we assumed in HA282 that the nucleation of creep cavities was controlled by whatever pre-existing sources/seeds, which after

certain period of creep strain accumulation would become operative voids and would either grow or heal depending on local stress state and environment. Thus, the following was focused on the growth of pre-existing voids. Nucleation itself can actually be treated separately by some phenomenological models or explicit statistical approaches to yield quantitative input for growth modeling.

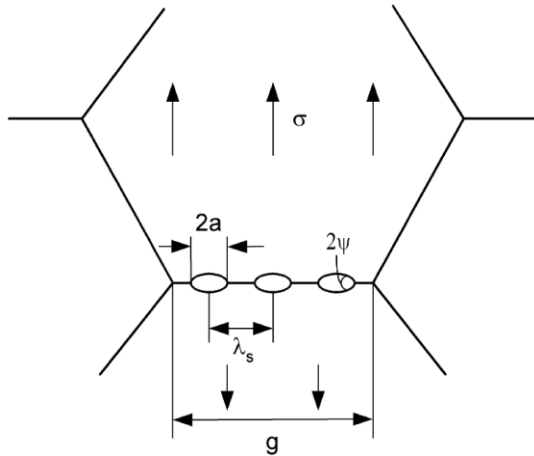


Figure 143: Void formed at a transverse grain boundary.

At elevated temperatures, the cavity growth has long been suggested to involve vacancy diffusion. The diffusion can be dominated by either grain boundary diffusion or bulk diffusion, and as temperature increases, the difference between the two may become ignorable. When the material is subject to non-vanishing far-field stress, a stress gradient is built up in the vicinity of the void surface, which is the driving force for vacancy to diffuse towards certain region of the void surface. The void/solid surface then migrates via accretion of these newly arrived vacancies; the void shape is also evolved due to the stress gradient along the surface, resulting in surface diffusion. This is the basic physical picture to be incorporated into our creep cavitation modeling, and is summarized in the schematics in Figure 144.

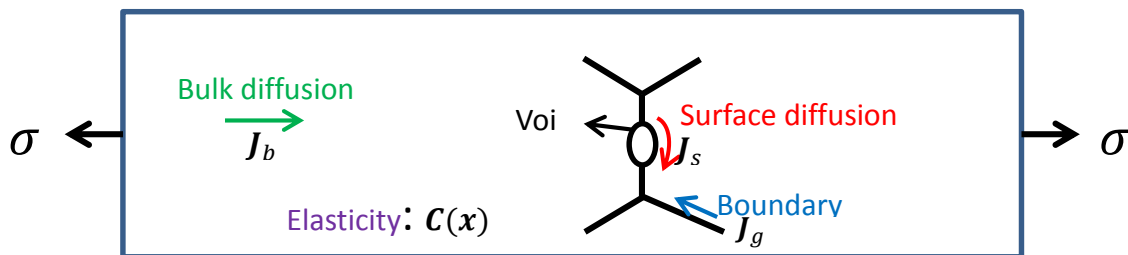


Figure 144: Illustration of vacancy diffusion controlled cavity evolution when subject to a constant.

4.2.1. Develop a phase-field model for creep cavitation

In the newly developed model, four major features have been considered. (1) Bulk diffusion, which controls the void growth/healing via diffusion at void/solid surface. (2) Surface diffusion, which controls the void shape change. (3) Grain boundary diffusion, which controls void growth/healing via diffusion at grain boundaries. (4) Inhomogeneous elasticity solver, which accounts for the high modulus contrast between solid and void phases. Here we only briefly present the total free energy functional of the phase-field model and address some key technical issues related to the model formulation.

In order to account for the solid and void phases, we use a phenomenological order parameter field $\phi(\mathbf{x})$, which takes value 1 in solid and 0 in void. In addition, the model should be able to consider a grain aggregate consisting of many grains with different orientations. This leads to a set of order parameters $\{\eta_i(\mathbf{x})\}_i^{N_g}$, where N_g is the total number of grains considered. Finally, the inhomogeneous elasticity can be solved following the phase-field microelasticity theory³³ and thus another order parameter called “virtual transformation strain field” $\epsilon_{ij}(\mathbf{x})$ is needed as well. Using these fields, the total free energy functional can be written as

$$F = \int \left(\frac{2\gamma_s}{\kappa_s \pi} \left[f(\phi, \{\eta_i\}) + \frac{1}{2} \kappa_s^2 |\nabla \phi|^2 + \frac{1}{2} \sum_i \left(\frac{\gamma_g}{\gamma_s} \right) \kappa_s \kappa_g |\nabla \eta_i|^2 \right] \right) dV + E^{\text{elast}} \quad (4.8)$$

where γ_s, γ_g are respectively the interface energy of void/solid and grain boundary, and κ_s, κ_g controls the width of these two interfaces. The bulk free energy density takes the form

$$f(\phi, \{\eta_i\}) = A\phi^2(1-\phi)^2 + B \left[\phi^2 + 6(1-\phi) \sum_i \eta_i^2 - 4(2-\phi) \sum_i \eta_i^3 + 3 \left(\sum_i \eta_i^2 \right)^2 \right] \quad (4.9)$$

where A and B are two constants. The elastic energy takes the form

$$\begin{aligned} E^{\text{elast}} = & \frac{1}{2} \int_V \left[C_{ijmn}^0 \Delta S_{mnpq}(\mathbf{x}) C_{pqkl}^0 - C_{ijkl}^0 \right] \left[\epsilon_{ij}^0(\mathbf{x}) - \epsilon_{ij}(\mathbf{x}) \right] \left[\epsilon_{kl}^0(\mathbf{x}) - \epsilon_{kl}(\mathbf{x}) \right] dV \\ & + \frac{1}{2} \int_V C_{ijkl}^0 \epsilon_{ij}^0(\mathbf{x}) \epsilon_{kl}^0(\mathbf{x}) dV - \bar{\epsilon}_{ij} \int_V C_{ijkl}^0 \epsilon_{kl}^0(\mathbf{x}) dV \\ & + \frac{V}{2} C_{ijkl}^0 \bar{\epsilon}_{ij} \bar{\epsilon}_{kl} - \frac{1}{2} \oint \frac{d^3 k}{(2\pi)^3} \tilde{\sigma}_{ij}^0(\mathbf{k}) \Gamma_{jikt}(\mathbf{n}) \tilde{\sigma}_{kl}^0(\mathbf{k})^* \end{aligned} \quad (4-10)$$

where $\Delta S_{ijkl} = \Delta C_{ijkl}^{-1}$ with $\Delta C_{ijkl} = C_{ijkl}^0 - C_{ijkl}(\mathbf{x})$, $\bar{\epsilon}_{ij}$ is the macroscopic average strain, $\sigma_{ij}^0 = C_{ijkl}^0 \epsilon_{kl}^0$, $\mathbf{k}$ is the reciprocal space vector, $\Gamma_{ijkl} = B_{ik} n_j n_l$, where $\mathbf{n} = \frac{\mathbf{k}}{|\mathbf{k}|}$ and $B_{ij}^{-1} = C_{ikjl}^0 n_k n_l$. The tilde represents the Fourier transform and the superscript asterisk the complex conjugate. The macroscopic strain is given as

$$\bar{\epsilon}_{ij} = \frac{1}{V} \int_V \epsilon_{ij}(\mathbf{x}) dV \quad (4-11)$$

where $\epsilon_{ij}(\mathbf{x})$ is the total strain field. Given a virtual transformation strain field, the functional derivative with respect to the phase-field order parameter $\phi(\mathbf{x})$ can be evaluated as

$$\begin{aligned} \frac{\delta E^{\text{elast}}}{\delta \phi} = & \frac{1}{2} C_{ijmn}^0 \frac{d\Delta S_{mnpq}(\mathbf{x})}{d\phi} C_{pqkl}^0 \left[\epsilon_{ij}^0(\mathbf{x}) - \epsilon_{ij}(\mathbf{x}) \right] \left[\epsilon_{kl}^0(\mathbf{x}) - \epsilon_{kl}(\mathbf{x}) \right] \\ & + \frac{1}{2} \left[C_{ijmn}^0 \Delta S_{mnpq}(\mathbf{x}) C_{pqkl}^0 - C_{ijkl}^0 \right] \frac{d\epsilon_{ij}(\mathbf{x})}{d\phi} \left[\epsilon_{kl}(\mathbf{x}) - \epsilon_{kl}^0(\mathbf{x}) \right] \\ & + \frac{1}{2} \left[C_{ijmn}^0 \Delta S_{mnpq}(\mathbf{x}) C_{pqkl}^0 - C_{ijkl}^0 \right] \left[\epsilon_{ij}(\mathbf{x}) - \epsilon_{ij}^0(\mathbf{x}) \right] \frac{d\epsilon_{kl}(\mathbf{x})}{d\phi} \end{aligned} \quad (4-12)$$

where $\epsilon_{ij}^0(\mathbf{x})$ is treated as another order parameter evolving simultaneously with $\phi(\mathbf{x})$ in a unified phase-field model. It is also found that the original governing equation for $\epsilon_{ij}^0(\mathbf{x})$ proposed may fail to converge when more than two phases are present, as has been pointed out by Shen et al.³⁴. Instead, a new way of updating the virtual strain field $\epsilon_{ij}^0(\mathbf{x})$ was proposed by Shen et al., which is found to be essentially the same algorithm used by Zhao et al.³⁵. In fact, both methods are iterative method based on the desired equivalency of the real and virtual systems in terms of mechanical equilibrium.

4.2.2. Modeling grain boundary cavity evolution during creep

When a simulation uses the configuration as in Figure 145 and allows diffusion to occur only at surface and bulk, an initially spherical void at the bicrystal boundary will evolve. Figure 146 shows the results for two cases: (i) $\sigma_{\text{appl}} = 0$, and (ii) $\sigma_{\text{appl}} = 400 \text{ MPa}$. When no stress is applied, the system is simply driven by reducing the extra surface energy due to the presence of the pore via the bulk and surface diffusion of mass. When a stress is applied, however, due to the balance between surface energy and elastic energy, the pore can be stabilized when the stress is sufficiently large. More specifically, while the existence of a pore inside the solid will increase the total energy by introducing extra free surface, the elastic energy stored in the material can be relaxed when that region is replaced by void phase. Similar energy balance effect between surface and

elastic energy has also been studied by Wang and Suo ³⁶. The relative importance of the elastic and the surface energy variation can be expressed using a dimensionless number as defined by Wang and Suo as

$$\Lambda = \frac{\sigma_{appl}^2 R}{E\gamma} \quad (4.13)$$

where R is the radius of the void, E is Young's modulus, and γ is the surface energy. Using the materials properties ($E = 203 \text{ GPa}$, $\gamma = 1.5 \text{ J/m}^2$), an applied stress (400 MPa), and the initial void radius ($0.182 \text{ }\mu\text{m}$) in the simulations, Wang and Suo has found that $\Lambda \sim 0.1$, which implied that the stress was relatively small so that the void shape could be stabilized subject to surface diffusion only. Our simulations considered more complicated situations, in which bulk diffusion was allowed to exchange mass at pore surface with the free surface (air layer). As a result, not only the shape of the void was evolved, but also the size was changed as shown in Figure 146. The irregular shape of the final free surface was due to the surface instability (and possibly periodic boundary condition (PBC) but with thin air layer) when under stress. Larger scale simulations with thicker air layer can be performed to further confirm. Nevertheless, the quantitative result obtained here should still be valid.

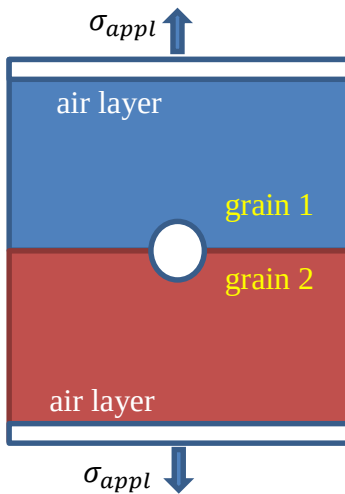


Figure 145: Simulation setup in a 2D cross-section view. A spherical void is placed at the center of the boundary between two grains. An air layer is added at the top and bottom to provide sources/sinks for void to evolve. Periodic boundary condition is then applied in all three directions in the actual 3D simulations.

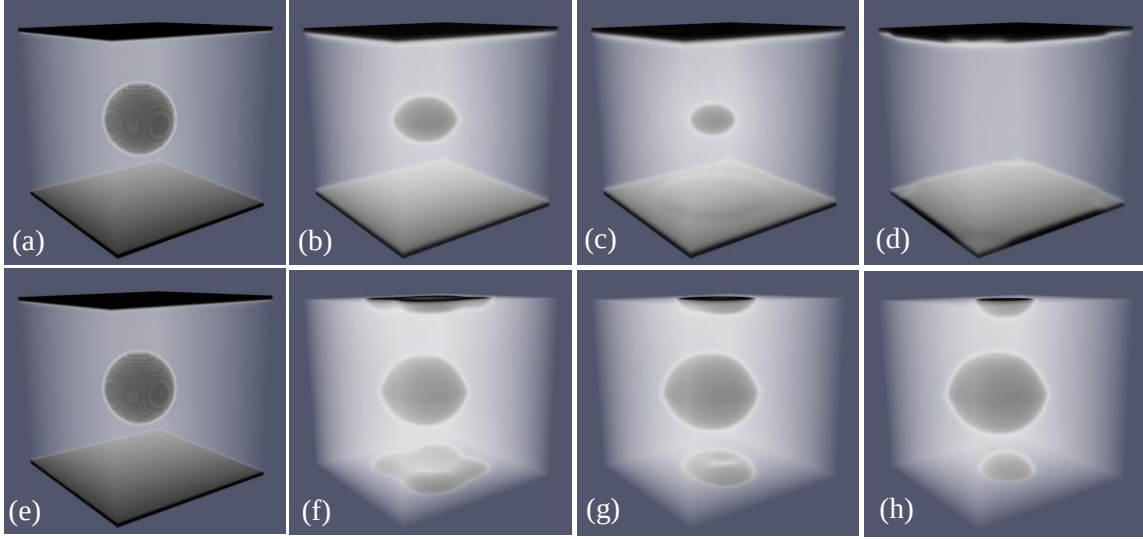


Figure 146: Evolution of a grain boundary void due to surface and bulk diffusion when (a-d) no stress is applied and (e-h) 400MPa is applied normal to the grain boundary plane. The plots are in volumetric representation and solid phase is transparent/white and void phase is black/opaque.

Figure 147 shows a 2D cross-section of von Mises stress during the void evolution in Figure 146(e-h). It is seen that the elastic field was accurately solved at every step, for the stress in the void phase was maintained virtually zero. Due to the different stress state near different surface regions, mass will diffuse from highly stressed regions to regions of nearly stress-free, which has been shown by Srolovitz³⁷ and is consistent with our simulation results in Figure 146.

Another interesting thing to check is the shape of the void during the evolution. In particular, in the absence of stress, the dihedral angle at the triple junction between two grains and void can be expressed as

$$\theta_d = 2\cos^{-1}\left(\frac{\gamma_g}{\gamma_s}\right) \quad (4.14)$$

where γ_g is the grain boundary energy, which is usually smaller than the surface energy. We carried out two simulations using the same surface energy $\gamma_s = 1.5 \text{ J/m}^2$ but different GB energy $\gamma_g = 1.2 \text{ J/m}^2$ and 0.5 J/m^2 , which corresponded to a dihedral angle of $\theta_d = 66.4^\circ$ and 80.4° , respectively. Since phase field model assumed all interfaces/boundaries are diffusive, we could only track the contour of a given order parameter value corresponding to the interface region. Consequently, the dihedral angle measured based on such contour plot introduced some uncertainty depending on threshold used for the contour plot. Nevertheless at least a qualitative comparison could be made for the two cases, as shown in Figure 148. It is obvious that during the shrinkage of the void, the simulation corresponding to $\theta_d = 80.4^\circ$ maintained the higher value as compared to the other simulation corresponding to $\theta_d = 66.4^\circ$. A crude measurement from

Figure 148 also suggested that the phase field predicted shape was approximately consistent with the analytic θ_d .

Finally, as a direction prediction, the model provided a quantitative description of the kinetics of void evolution subject to the given stress boundary condition. In Figure 149, we plot the equivalent 3D radius of void versus time. In the case of zero applied stress, the void was shrinking linearly with time when the void was large, and the shrinkage rate was accelerated at the later stage, which is consistent with an analytical result of Westwood et al.³⁸. When a stress was applied, the void was seen to grow with time. Such quantitative kinetics data can be used by high-level constitutive models as a description of the underlying damage evolution.

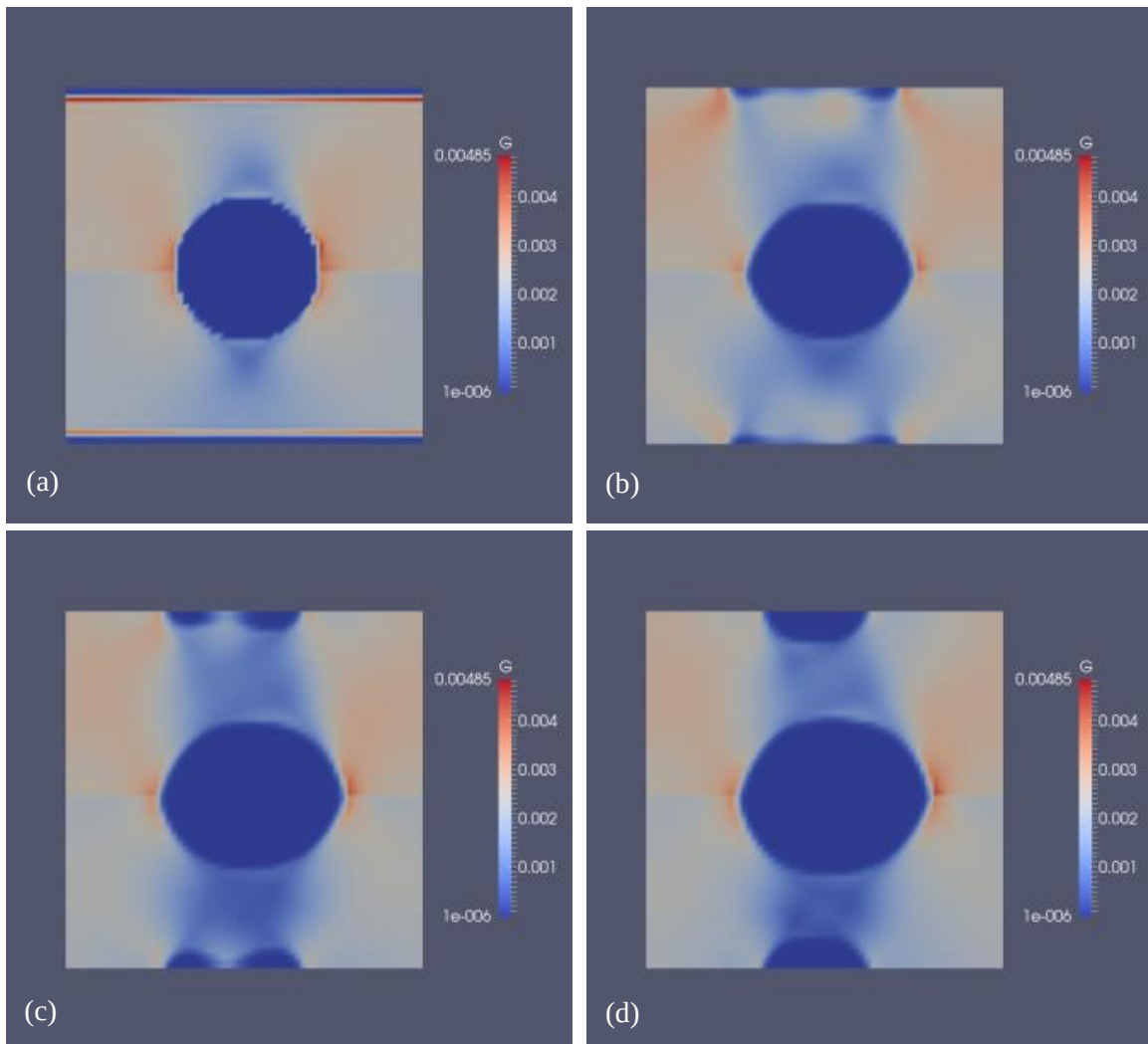


Figure 147: Von Mises stress field along 2D cross-section corresponding to Figure 123(e-h). Note that surface diffusion takes place by placing the initially highly stressed materials onto regions of nearly stress-free state. The unit corresponds to the normalization using shear modulus G .

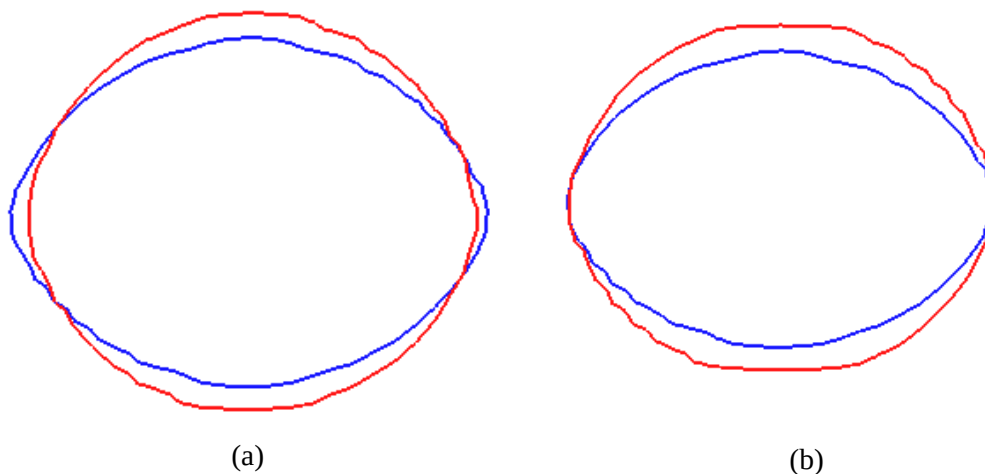


Figure 148: Evolution of void/solid boundaries as the void is shrinking due to bulk diffusion in the absence of stress. The blue curve corresponds to the case that analytic dihedral angle is 66.4° and the red one 80.4°.

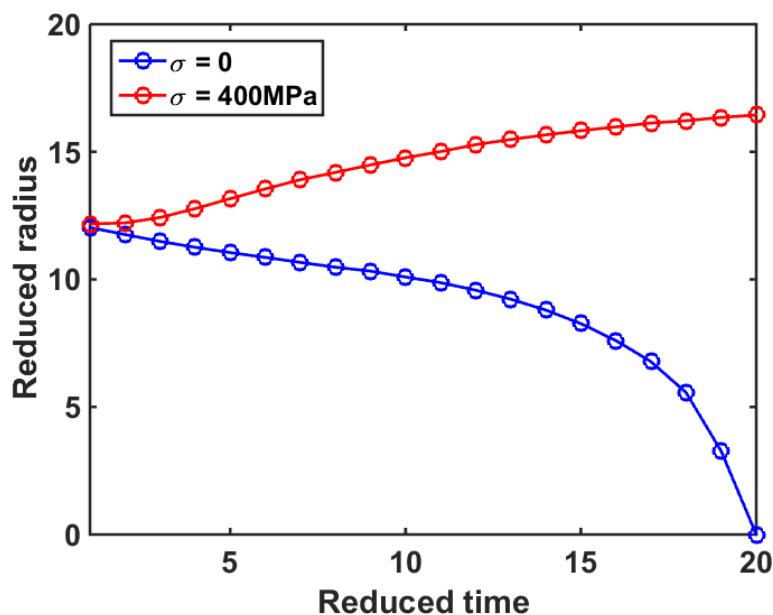


Figure 149: Grain boundary cavity kinetics predicted by the current phase-field modeling.

4.3 Summary and perspective

The creep deformation studied in Task 4 was limited to the primary creep and early secondary creep (~1000 hours) stages, while the experimental tests lasted for a longer time (over 3000 hours), as shown for example in Figure 150. In

particular, the continuously increased creep rate in the precipitate-hardened superalloy, rather than a distinctive steady-state “secondary creep”, was not captured by the FFT-EVP model developed at this time. The main reason for this is that the time evolution of underlying microstructures has not been treated sufficiently. Using a static microstructure for the limited regime under current consideration appears to be valid, but it needs to be improved for future modeling when a longer term of creep is to be captured. Specifically, two microstructural aspects need to be incorporated. First, the γ' particle coarsening is expected to influence the creep behavior and based on our current study, the resulting influence on the local inter-particle spacing may indeed accelerate creep rate as seen in experimental curves. In addition, if the “tertiary” creep is to be modeled, incorporation of the cavity nucleation and growth must be considered in order to capture the explosive growth of the creep rate.

The major outputs from Task 4 are:

- Quantitative influence of γ' structural non-uniformities on tensile and creep strength,
- Dislocation density evolution during tensile and creep deformation,
- GB cavity dynamics subject to various stresses and temperatures, and its correlation to the GB characters.

Those results actually can be effectively incorporated in the continuum constitutive modeling of the entire creep curves in Task 5. In particular, the above three major results correspond to, respectively, the following internal strain variables that have been commonly used in macroscopic crystal plasticity model:

- Tensile/creep strength (threshold stresses),
- Dislocation density,
- Damage measurement (e.g., volume fraction of cavity).
- The connection of Task 4 FFT-CP to the continuum modeling in Task 5 is indicated in Figure 150 as well.

In summary, we have developed in Task 4 a full-field image-based crystal plasticity model and have validated the model against the experimental testing data on HA282 base metal from Task 2. Key features of the model include: (i) dislocation-based, (ii) HA282 specific dislocation-precipitate interactions, and (iii) FFT-based micromechanical solver. We have also applied the model to the study of the influence of structural heterogeneities in weld, such as microsegregation and grain boundary denuded zone found by experiments and simulations in Tasks 2 and 3, on the tensile and creep performance. The degree of structural inhomogeneity under PWHT3a was found to bring no significant changes of the mechanical properties of HA282 weld according to our simulations. However, if the processing conditions allowed even larger degree of structural heterogeneities in the weld, our simulation results suggested that tensile and creep strengths would be reduced significantly. To study the evolution of grain boundary creep cavities as observed in Task 2, we have developed a phase-field

model that incorporates multiple diffusion mechanisms and inhomogeneous elasticity. The cavity dynamics for simple bicrystal grain boundary cavity configuration has been simulated using the phase-field model and the cavity kinetics and morphological change have been predicted. Results obtained from Task 4 can be effectively incorporated in the continuum modeling of the long-term creep behavior at continuum in Task 5.

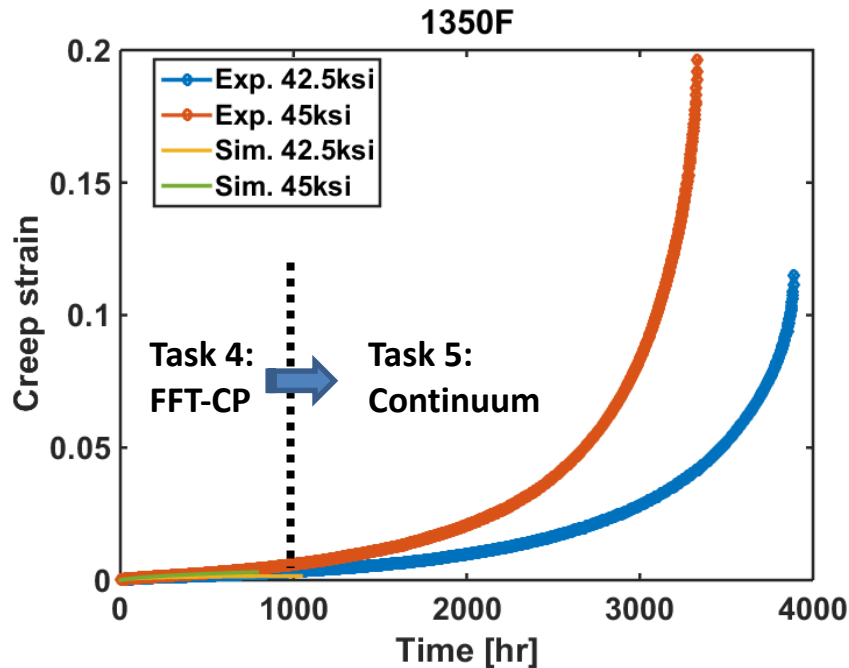


Figure 150: Connection between Task 4 and Task 5.

Task 5: Development of a Constitutive Creep Model that accounts for Weld and Base Metal Microstructure and their Long Term Evolution

A continuum damage creep model was developed under a recent DOE program³⁹ for HA282 base metal at temperatures between 1350 and 1450°F, and stress levels between 15 and 45Ksi. These ranges cover the typical temperature-stress conditions for Advanced Ultra-supercritical (A-USC) steam turbines (i.e. rotors) or A-USC boiler components (i.e, superheater or reheater outlet headers⁴⁰). Under the current program, we refined the creep model and extended its predictions capabilities to significantly higher temperatures (1400-1700°F) and to HA282 weld material. Thus, the applicability was extended to other power generation systems components (i.e. gas turbine combustion liner) which may experience more aggressive temperatures and stress levels. Within a unified formulation in term of strain rate evolution equation, the model contains explicit parameters dependent on temperature and explicit parameters capturing the possible differences between HA282 base metal and HA282 weld. The potential differences (some were found very small for HA282) between the base metal and weld were identified as:

- i) grain size (approximately 200 μ m for the weld as compared with 150 μ m for base metal)
- ii) grain boundaries cavities, which may have a different evolution during creep deformation in the weld as compared with the base metal (likely a consequence of the discontinuous coarsening mechanism at the grain boundaries in the weld).
- iii) dislocation pinning, which may be different in the weld, possibly due to the presence of third-phase pinning particles (in addition to gamma prime precipitates, as for example the carbides).

The predictions for structural components containing both base metal and weld (as for example the cross-weld specimens) could be made without additional calibration. Two methods were used for cross-weld specimens:

- a) material point approach (for uniaxial and uniform loading)
- b) finite element analysis (for complex loadings and non-uniform strain/stress spatial distribution).

The current section is structured in three parts: 1. Model formulation and the summary of the constitutive equations describing the main creep mechanisms; 2. Material point predictions, including comparison between experiments and model predictions for short and long term creep tests; and 3. Finite element predictions, containing a comparison between model predictions and creep experiments with local strain measurements.

5.1. Model formulation

5.1.1. Creep strain evolution

Using the associative flow rule^{41,42} the creep strain tensor $\boldsymbol{\varepsilon}^{creep}$ was taken parallel to the deviatoric part $\tilde{\boldsymbol{s}}$ of the effective stress tensor $\tilde{\boldsymbol{\sigma}}$. As it will be mentioned later in the current section, the effective stress is the local stress tensor corrected by the local effective element of area, accounting for the progressive damage in the material occurring during the deformation:

$$\boldsymbol{\varepsilon}^{creep} = \frac{3}{2} \frac{\varepsilon^{eq}}{\tilde{s}^{eq}} \tilde{\boldsymbol{s}} \quad (5.1)$$

where $\tilde{\boldsymbol{s}} = \tilde{\boldsymbol{\sigma}} - 1/3 \text{tr}(\tilde{\boldsymbol{\sigma}})I$ is the deviatoric part of the effective stress, $\tilde{s}^{eq} = \sqrt{\frac{3}{2} \tilde{\boldsymbol{s}} : \tilde{\boldsymbol{s}}}$ is the equivalent stress, and $\varepsilon^{eq} = \sqrt{\frac{2}{3} \boldsymbol{\varepsilon}^{creep} : \boldsymbol{\varepsilon}^{creep}}$ is the equivalent strain.

One of the main assumption of the model was that the total creep strain $\boldsymbol{\varepsilon}^{creep}$ can be represented as a superposition between the dislocation component, $\boldsymbol{\varepsilon}^{disloc}$, and the diffusion component, $\boldsymbol{\varepsilon}^{diffusion}$:

$$\boldsymbol{\varepsilon}^{creep} = \boldsymbol{\varepsilon}^{disloc} + \boldsymbol{\varepsilon}^{diff} \quad (5.2)$$

where the equivalent dislocation and diffusion strain rates are:

$$\dot{\boldsymbol{\varepsilon}}^{disloc} = \frac{3}{2} \frac{\dot{\varepsilon}^{disloc eq}}{\tilde{s}^{eq}} \tilde{\boldsymbol{s}}, \quad \dot{\boldsymbol{\varepsilon}}^{diff} = \frac{3}{2} \frac{\dot{\varepsilon}^{diff eq}}{\tilde{s}^{eq}} \tilde{\boldsymbol{s}}$$

The historical TEM analysis performed on HA282 base metal as well the current analysis performed on the HA282 weldment indicated that, at intermediate stress levels, γ' precipitate strengthening and dislocation strengthening were the dominant creep mechanism. We followed Dyson's approach^{28,43,44}, according to which the dislocation strain rate should vary as a hyperbolic-tangent function of the effective stress $\tilde{s}^{eq}$, penalized by the stress needed for the dislocation for climb around the precipitates σ_{climb} and by the stress needed for mobile dislocations to un-pin from the forest (immobile) dislocations σ_o :

$$\dot{\varepsilon}^{disloc eq} = \rho A D_V f (1 - f) \left(\sqrt{\frac{\pi}{4f}} - 1 \right) \frac{1}{M} \sinh \left(\frac{(\tilde{s}^{eq} - \sigma_{climb} - \sigma_o)}{MkT} b^2 \lambda \right) \quad (5.3)$$

where A is a constant, ρ is the density of dislocations, D_V is an effective diffusivity, f is the volume fraction of γ' precipitates (as a function of temperature), M is Taylor factor (=3.07), k is Boltzmann constant, T is the

absolute temperature, b is the Burgers vector (2.54Å), and $\lambda = 1.6r\sqrt{\pi/4f - 1}$ is the average spacing between precipitates. The notation $\langle x \rangle$ stands for the Macaulay bracket: $\langle x \rangle = x$ if $x > 0$ and 0 otherwise.

The Taylor stress is proportional with the square root of dislocation density ³¹,

$$\sigma_o = (0.45 - 0.00017T)MG(T)b\sqrt{\rho} \quad (5.4)$$

where G is the shear modulus, and ν the Poisson's ratio. The precipitation induced back stress was proposed by Dyson ^{28,43} as:

$$\sigma_{climb} = \frac{2f}{1 + 2f} \tilde{\sigma}^{eq} \left[1 - \exp \left(- \frac{1 + 2f}{2(1 - f)} E \frac{\epsilon^{disloc}}{\tilde{\sigma}^{eq}} \right) \right] \quad (5.5)$$

One of the important parameters in Eq. (5.3) is the average space between precipitates λ , expressed as a function of precipitate size and volume fraction of precipitates. Extensive analysis of precipitate evolution under long term thermal exposure has been performed for HA282 using phase field models and precipitate models previously ³⁹. An evolution equation of the form:

$$2r = \left(a + b \exp \left(- \frac{c}{RT} \right) t \right)^n \quad (5.6)$$

was used for the average precipitate size r . In the equation above, R is the gas constant, T is the absolute temperature, t is the exposure time and a, b, c and n are material constants.

A simple piecewise linear evolution was taken for the dislocation density. It was considered that the dislocation evolves linearly until it reaches a plateau:

$$\rho = \rho_i + \frac{(\rho_f - \rho_i)}{\epsilon^{crit}} * \min(\epsilon^{crit}, \epsilon) \quad (5.7)$$

with $\epsilon^{crit} = B \sigma_{applied}[MPa]$, where B is a calibration parameter; ρ_i is the initial dislocation density and ρ_f is the saturation value of dislocation density. The slope of the linear part B and an expression of the form $\varphi * \exp(\chi/kT)/\mu m^2$ were fitted for the saturation value, based on a selected set of macroscopic data.

The diffusion creep strain rate $\dot{\epsilon}^{diff^{eq}}$ was considered the cumulated effect of lattice diffusion and grain boundary diffusion:

$$\dot{\epsilon}^{diff^{eq}} = 3\pi C D_B \frac{\delta_B \Omega}{RT} \left(\frac{1}{d} \right)^3 \tilde{\sigma}^{eq} + C 3D_V \frac{\Omega}{RT} \frac{1}{d^2} \tilde{\sigma}^{eq} + \dot{\epsilon}^{cav} \quad (5.8)$$

where d is the average grain size, $D_B \delta_B$ is the grain boundaries diffusivity⁴⁵ and C is a calibrating parameter. The last term $\dot{\epsilon}^{cav}$ describes an additional strain contribution from cavity evolution, which is given later in Eq.(5.16).

5.1.2. Damage evolution

Within the CDM framework^{42,46}, the material damage due to the presence of micro-cavities was represented by a second order tensor Ω called the damage tensor. The evolution of these micro-defects (by growth and coalescence) into life limiting cavities was termed as damage evolution and represented by a change of the damage tensor Ω during the creep process. Because the optical images showed fairly random cavities distribution at the macroscopic level, we made the simplifying assumption of an isotropic damage, which reduces the number of components for Ω to one:

$$\Omega = \varpi \mathbf{e}_i \otimes \mathbf{e}_i \quad (5.9)$$

The effect of damage in creep deformation is typically captured in the literature^{42,46}, by an *effective stress tensor* $\tilde{\sigma}$, representing a transformation of the actual Cauchy stress tensor σ due to the material degradation (i.e. micro-voids growth and coalescence):

$$\tilde{\sigma} = \frac{1}{1 - \varpi} \sigma \quad (5.10)$$

Following Cocks and Ashby approach, the GE Haynes 282 creep model^{39,47} was built on the hypothesis that the micro-voids at the grain boundaries can grow by inelastic deformation of the neighboring grains (ϖ_{disloc}) or diffusion mechanisms (at the surface of the cavity or at the grain boundaries near the cavity)

$$\varpi = \varpi_{diffusion} + \varpi_{disloc} \quad (5.11)$$

Dislocation damage

In order to assess the damage component we used the evolution law, initially proposed by Budiansky et al⁴⁸ by estimating the volumetric growth rate of a spherical void in an infinite material with a power law creep.

$$\dot{\omega}_{disloc} = \left(\frac{3}{2n} \frac{\sigma_h}{\sigma_{eq}} + \frac{(n-1)(n+0.4319)}{n^2} \right)^n \dot{\epsilon}^{disloc} \omega_{disloc} \quad (5.12)$$

Where n is the power law exponent of the material around the cavity, σ_h is the applied hydrostatic stress and σ_{eq} is the applied equivalent stress.

For uniaxial loading with the applied stress σ the hydrostatic stress can be calculated as $\sigma_h = \sigma/3$ and the equivalent stress $\sigma_{eq} = |\sigma|$. Thus, the equation (5.12) reads

$$\dot{\omega}_{disloc} = K \dot{\epsilon}^{disloc} \omega_{disloc}, \quad K = \left(\frac{1}{2n} + \frac{(n-1)(n+0.4319)}{n^2} \right)^n \quad (5.13)$$

Equation (5.13) has now one material parameter n which is not fitted to creep strain versus time curve but rather informed from the material behavior in the secondary creep regime. Literature data for pure nickel and our analysis of the minimum strain rate variation with stress for HA282 reports values between 4.6 and 6.5, which gives a variation for the K parameter between 0.85 and 0.87. Equation (5.12) still makes a simplifying assumption (namely the power law creep for the grains embedding the cavity) but a large variation of the exponent n leads to a tight variation in the scaling parameter K . For very large n , the parameter K converges towards the 0.93.

As discussed in Section 2.3.5, detailed characterization of creep specimens indicated a higher number of grain boundary cavities in weld zone and that they are often found in the neighborhood of γ' denuded regions where the local material has a single γ (matrix) phase. This suggests that the material surrounding the cavity can be softer. Therefore, for weld we took K as the max value of 0.93 while for the base metal we considered 0.87

Diffusion Damage

The first model we investigated for the evolution of the diffusion damage was proposed by ^{49,50} (CA model). According to this model, the cavity can grow by both surface ω_{SD} and grain boundary diffusion ω_{BD}

$$\omega_{diffusion} = \omega_{BD} + \omega_{SD} \quad (5.14)$$

$$\dot{\omega}_{BD} = \frac{\xi}{2} \frac{\tilde{s}^{eq}}{\sqrt{\omega_{BD}} \ln\left(\frac{1}{\omega_{BD}}\right)}, \quad \dot{\omega}_{SD} = \frac{\xi}{4} \frac{1}{\sqrt{2}} \frac{l^2}{\gamma^2} \frac{\sqrt{\omega_{SD}} [\tilde{s}^{eq}]^3}{(1 - \omega_{SD})^3}, \quad \xi = C D_B \frac{\delta_B \Omega}{R T l^3} \quad (5.15)$$

The cavity growth mechanisms will add a small contribution to the diffusion strain described in Eq. (5.8). This was estimated in ^{49,50} as:

$$\dot{\epsilon}^{cav} = \xi \frac{l}{d \ln(1/\omega_{BD})} \frac{\tilde{s}^{eq}}{\omega_{BD}} + \xi \alpha \frac{\sqrt{\omega_{SD}} \tilde{s}^{eq2}}{(1 - \omega_{SD})^3} \quad (5.16)$$

A second model for diffusion damage evolution, initially developed by Van der Giessen and collaborators ⁵¹ based on finite elements simulations of cavities in power law materials was explored and compared with the CA model. The VDG model formulation and results are presented in Section 5.4. It was concluded that the VDG model cannot reduce the number of fitting parameters, thus we continued our creep behavior investigation using the constitutive Equations (5.1)-(5.16).

5.2. Material Point Predictions

The first set of predictions was done by solving directly the equations (5.1)-(5.16) for a given applied stress. While the analysis is done for a single material point, without spatial resolution and without possibility of capturing the spatial strain gradients, the predictions are representative for uniaxial loading and uniform stress. Thus, the solution should be exact for a uniform sample, made from a single material (as base metal or 100% weld) and loaded uniaxially. For bi-material (as cross-weld) specimens, where it is known that the two components do not interact significantly through field, the rule of mixture can still provide representative solution.

Figure 151 compares the model predictions with experiment for 100% weld samples in terms of creep strain evolution in time.

Figure 152-Figure 154 compare the model predictions with experiment at 1500, 1600 and 1700°F for cross-weld specimens, in terms of creep strain evolution in time, using material point model and the rule of mixture.

Figure 155-Figure 157 compare the model predictions with experiment at 1500, 1600 and 1700°F for cross-weld specimens and 100% weld, in terms of creep strain rate evolution in time, using material point model and the rule of mixture.

Figure 158 compares the LMP parameter for 100% weld and base metal. It can be observed that the LMP parameter is within the experimental limits for the entire stress range between 3Ksi and 30Ksi. The rupture times in simulation were determined as the time that the strain rate reached 0.005/hr, where the strain started to increase rapidly with time.

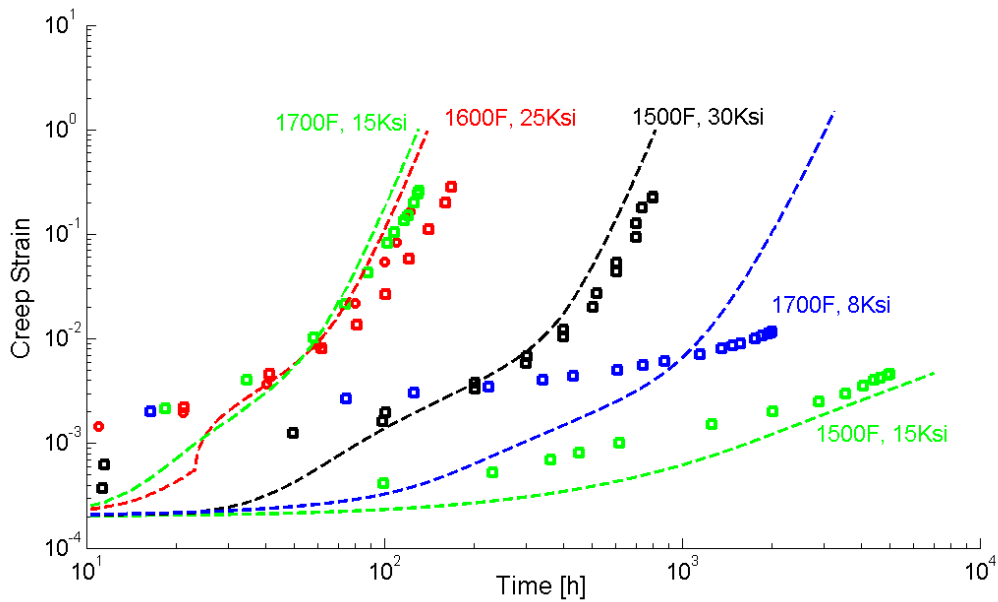


Figure 151: Experimental data for 100% weld (symbols) and corresponding model predictions using material point approach (dotted lines).

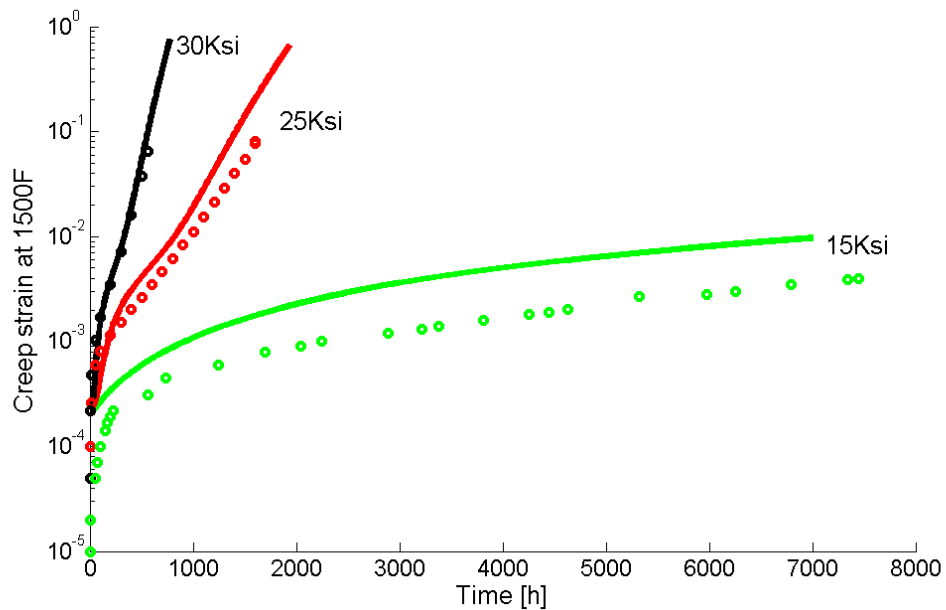


Figure 152: Creep strain evolution in time. Model predictions (continuous lines) compared with experiments (symbols) at 1500°F for cross-weld samples.

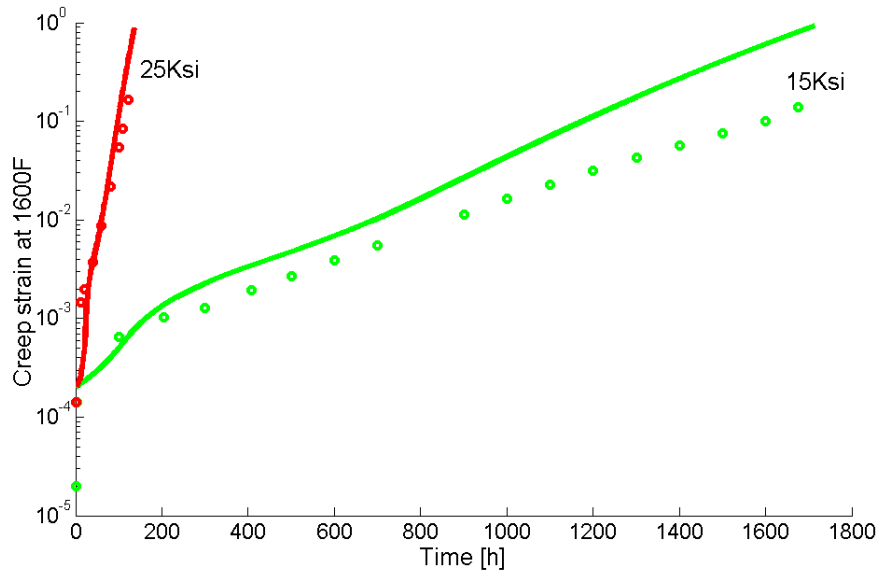


Figure 153: Creep strain evolution in time. Model predictions (continuous lines) compared with experiments (symbols) at 1600F for cross-weld samples.

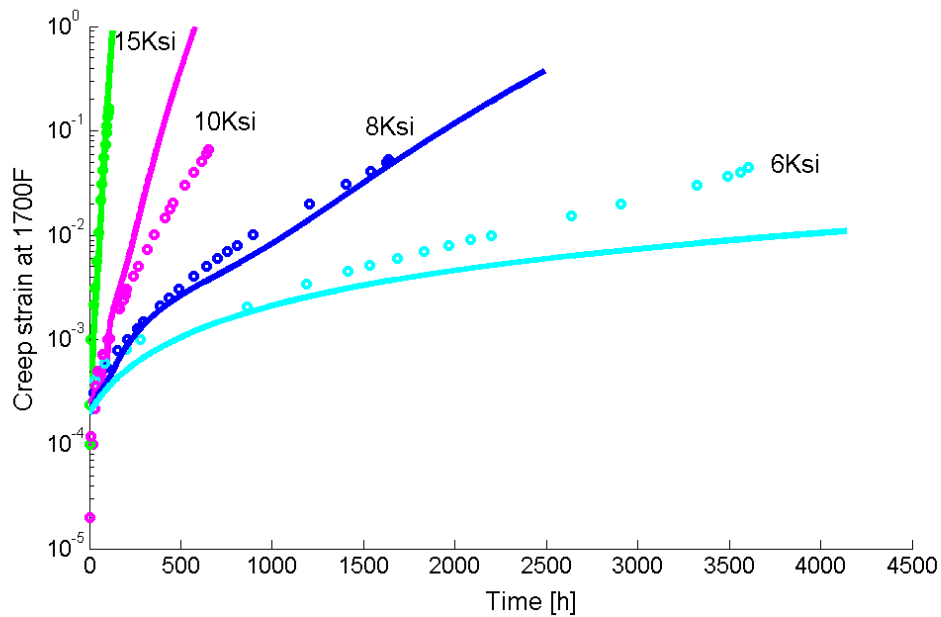


Figure 154: Creep strain evolution in time. Model predictions (continuous lines) compared with experiments (symbols) at 1700F for cross-weld samples.

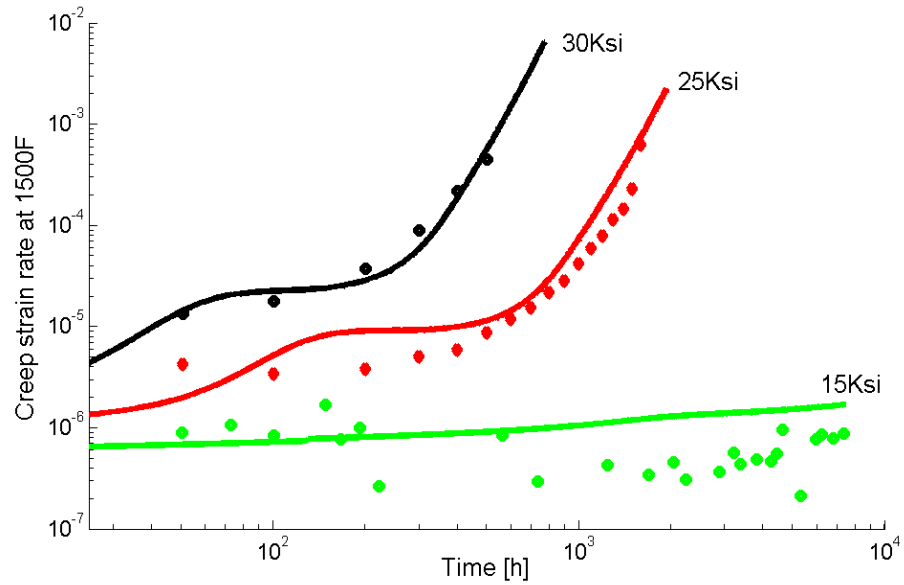


Figure 155: Creep strain rate evolution in time. Model predictions (continuous lines) compared with experiments (symbols) at 1500°F for cross-weld samples.

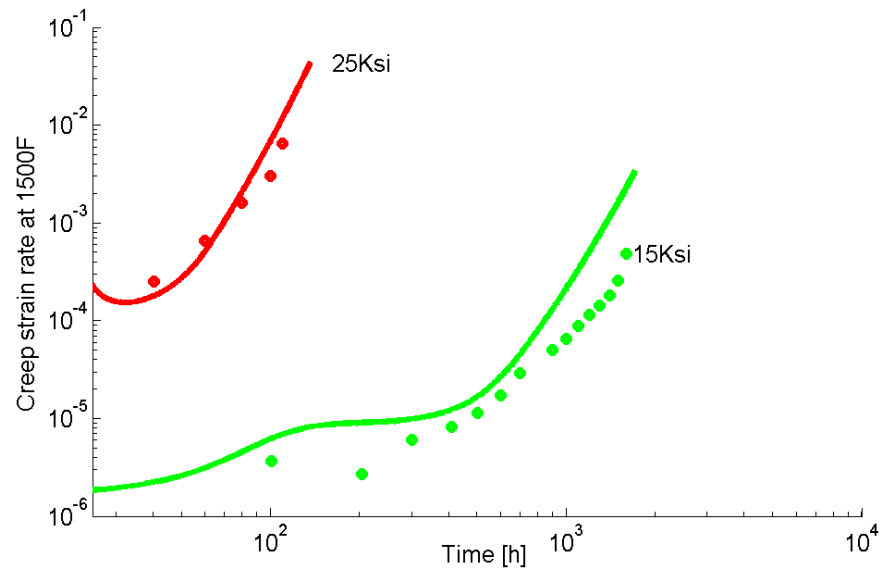


Figure 156: Creep strain rate evolution in time. Model predictions (continuous lines) compared with experiments (symbols) at 1600°F for cross-weld samples.

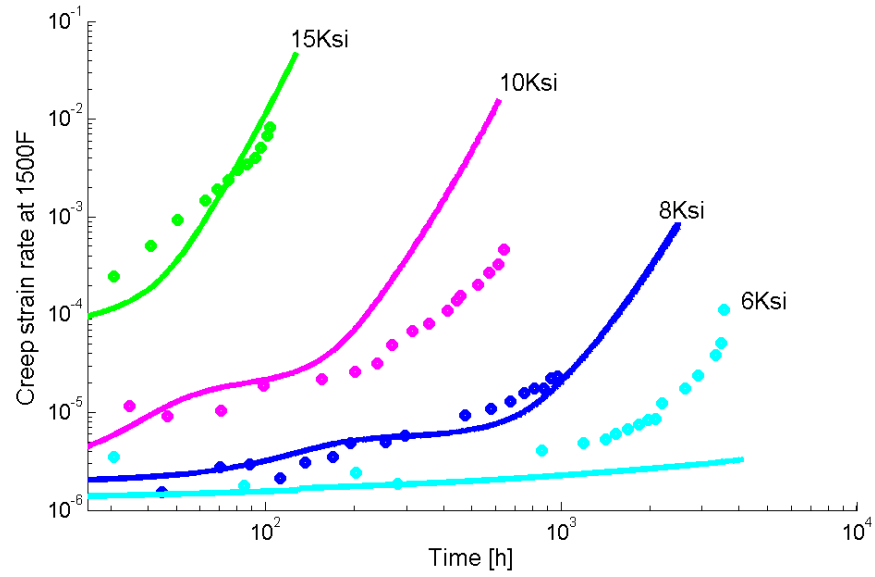


Figure 157: Creep strain rate evolution in time. Model predictions (continuous lines) compared with experiments (symbols) at 1700°F for cross-weld samples.

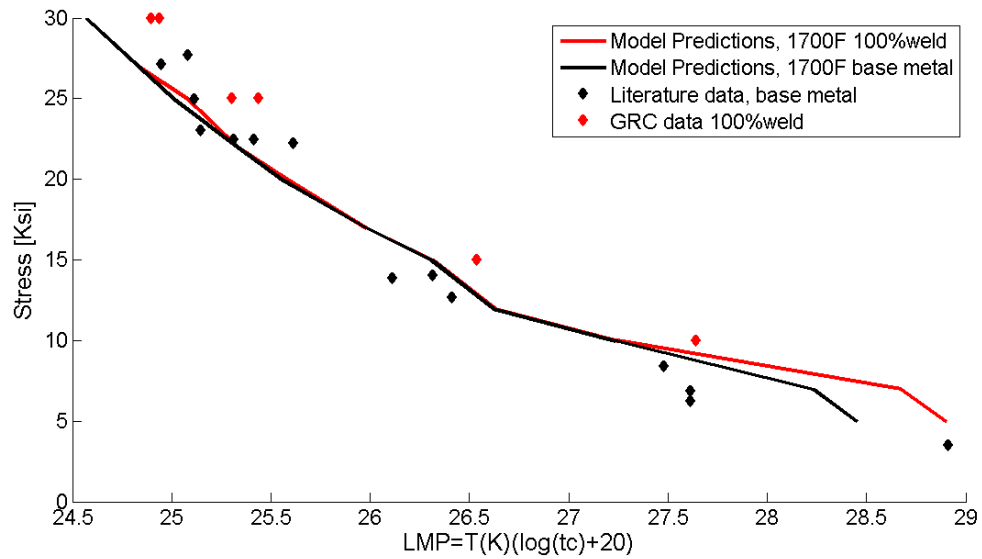


Figure 158: Stress vs Larson Miller (LMP) parameter: model predictions at 1700°F (continuous lines) for base metal and 100% weld and experiments for base metal (Literature data digitized from ⁵²) and for 100% weld.

5.3. Finite Element Predictions

In order to investigate a potential interaction through field between the base metal and weld material in a cross-weld sample, the actual three-dimensional structure of the gauge section of the specimen was simulated using a finite element analysis. An example of the simulated geometry with the symmetry boundary conditions was sketched in Figure 159. The bar is restricted to move axially on one side while the stress is applied at the other end. Symmetry boundary condition were used on the mid-section with the normal perpendicular to the loading direction. The constitutive model equations (5.1)-(5.15) were implemented in a user defined material subroutine (UMAT) in Abaqus. For the subroutine verification purpose we compared the finite element predictions with the material point predictions, for a few bars made of 100% weld. The comparison is shown in Figure 160.

Another validation of the finite element procedure was made by comparing the FE predicted creep strain with the experimental strain values obtained through DIC procedure. Figure 161(a) shows an example of creep curve obtained using the finite element procedure at 1700°F, 15Ksi for 100% weld sample. The data point marked by symbol is the strain value measured by extensometer after ~ 23 hrs. Figure 161(b) shows the uniform strain distribution at the same time, predicted by the model.

Next, we performed the finite element calculations for the cross-weld bar, seen as a composite of two materials: the 100% weld in the middle and base metal at the ends (Figure 159). The same UMAT subroutine was called by the finite element code for each one of the two materials. However, some of the input parameters were different. These are presented in Table 21.

Table 21: The model parameters that have different values for base metal and weldment.

	Grain size (microns)	Damage parameter, K	Third-phase particle volume fraction (%)
Base metal	150	0.87	0
100% weld	200	0.93	0.5

Figure 162 show the first principal creep strain and von Mises stress distribution in the bar (at 1700°F, 15Ksi, 47 hours). Although the contour colors are slightly different in the weld region as compared with weld region, the values for both stress and strain are very close. Also, the finite element program does not show any significant strain variation in the neighborhood of the interface between the weld and the base metal.

Figure 163 and Figure 164 show the results predicted by the model for two cross-weld samples tested at 1600°F, 25Ksi and 1700°F, 15Ksi. These figures are the modeling pairs of the experimental results (Figure 28 and Figure 29) presented in Section 2.2.4.

The creep strain is monitored in two locations: one selected in the weld region and one selected on the base metal (Figure 163 and Figure 165). When available, the experimental values determined by averaging the DIC measured creep strain on various regions are presented. Although the strain in the weld region seems slightly smaller than the strain in the base metal, the difference is very small and it may be within numerical errors. For both cases presented in Figure 163-Figure 166, the predicted rupture time is close to the experiment (Table 22).

Table 22: Rupture time as predicted and measured for cross-weld bars.

Condition	Time to rupture Model (hours)	Time to rupture Experiment (hours)
1600°F, 25Ksi	113	97
1700°F, 15Ksi	66	64

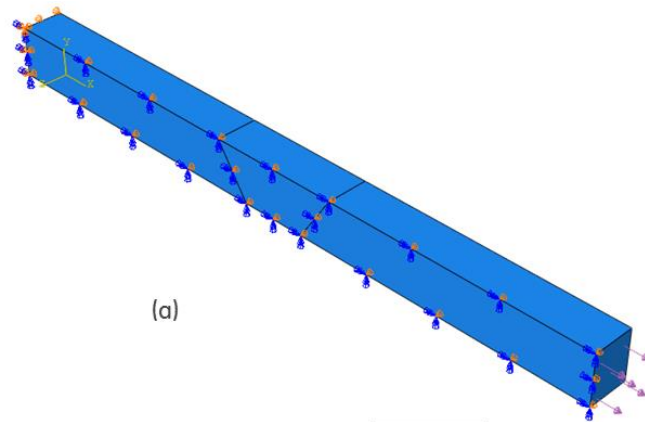


Figure 159: Sample geometry and boundary conditions for finite element simulations of cross-weld.

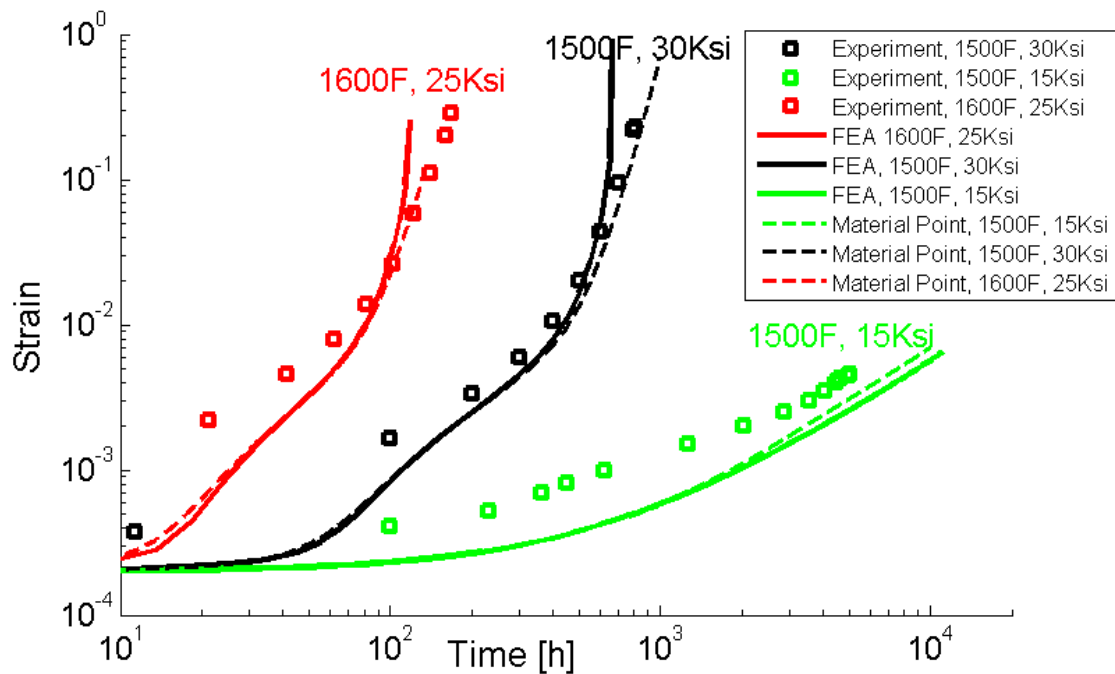


Figure 160: Comparison between material point predictions and finite element predictions for 100% weld samples.

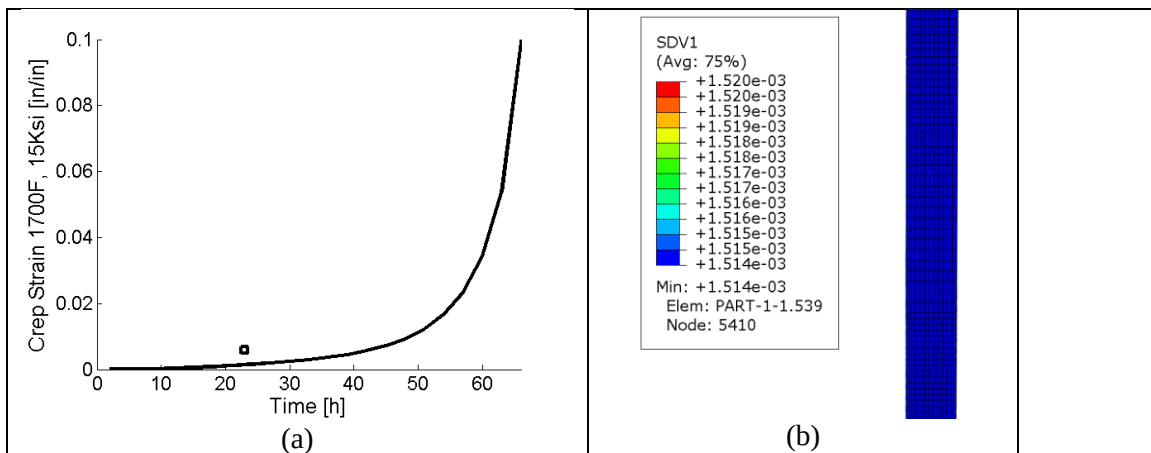


Figure 161: Sample MT 20, 1700°F, 15Ksi, 100% weld samples, standard PWHT. (a) Creep strain evolution in time for 100% weld as predicted by the finite element model (continuous line) and from experiment (at ~23hr). (b) first principal strain (constant through the gauge section) at time =23hr.

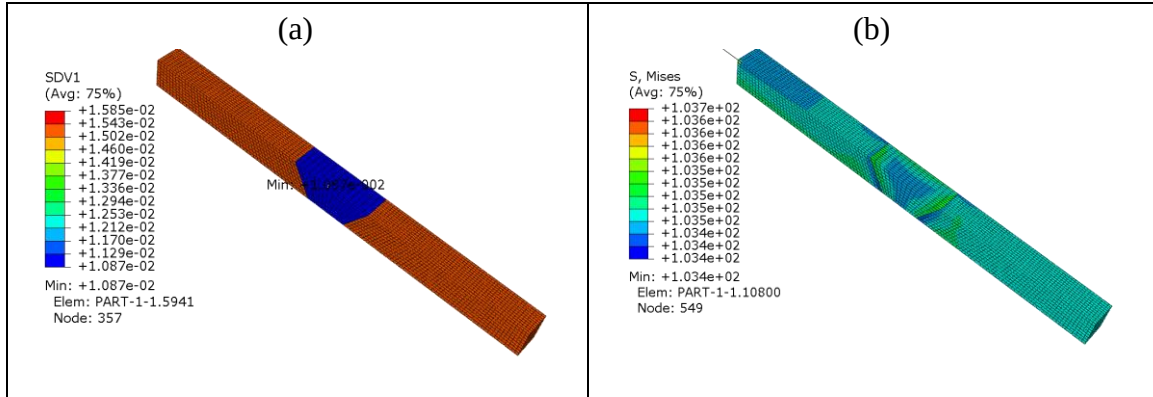


Figure 162: (a) Example of creep strain distribution and (b) Example of Mises stress distribution for cross-weld sample. The case shown corresponds to sample MT 10, 1700°F, 15Ksi, 49hr.

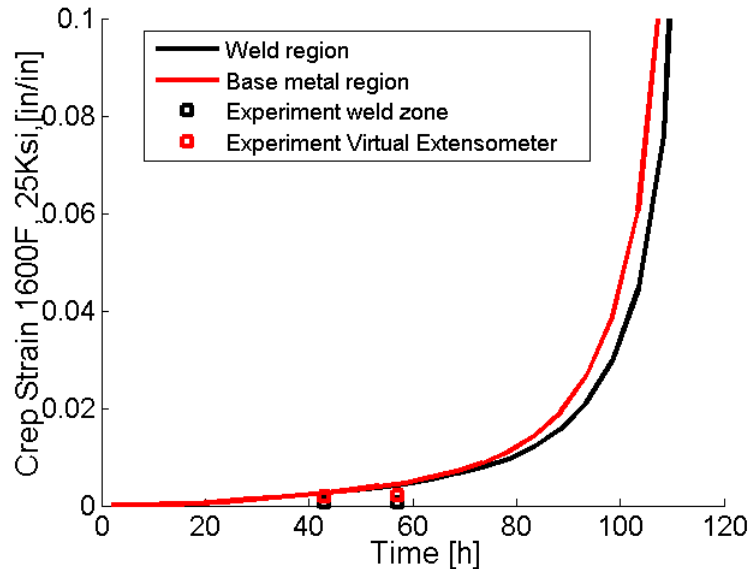


Figure 163: Cross-weld sample MT 09 tested at 1600°F, 25Ksi. Strain evolution in time as predicted through the finite element method in the weld region (black line) and in the base metal (red line). Experimental values at 43 and 57hr are represented by symbols (red - virtual extensometer and black for values obtained through DIC method by averaging the creep strain on weld region). The predicted rupture time is 113hr. The experimental rupture time is ~97hr.

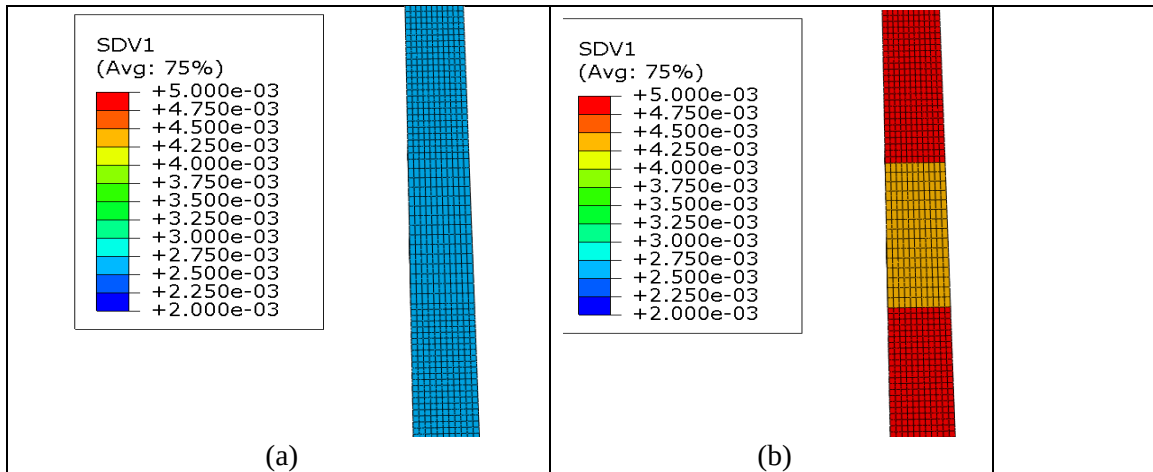


Figure 164: Cross-weld sample MT 09 tested at 1600°F, 25Ksi. Creep strain distribution in the cross-weld samples at (a) 43 hr (b) 57 hr, as predicted by the model

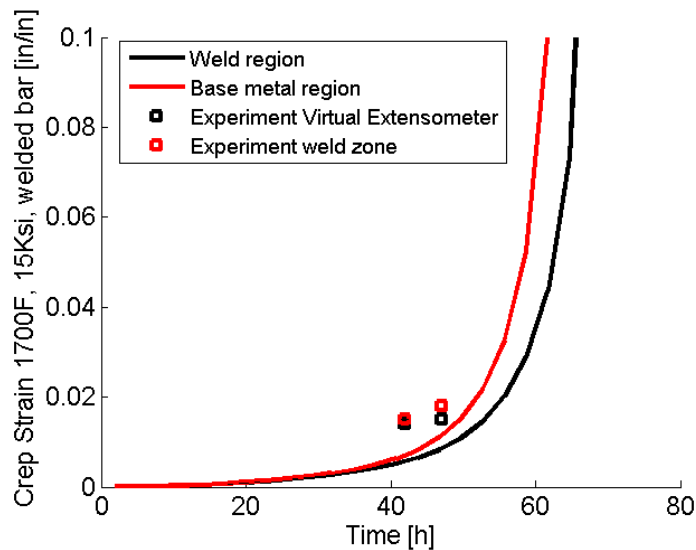


Figure 165: Cross-weld sample MT 10 tested at 1700°F, 15Ksi. Strain evolution in time as predicted through the finite element method in the weld region (black line) and in the base metal region (red line). Experimental values at 42 and 47 hr, obtained using the DIC method are represented with symbols (red – virtual extensometer and black – averaged on welded region). The predicted rupture time is 66 hr. The experimental rupture time is ~64 hr.

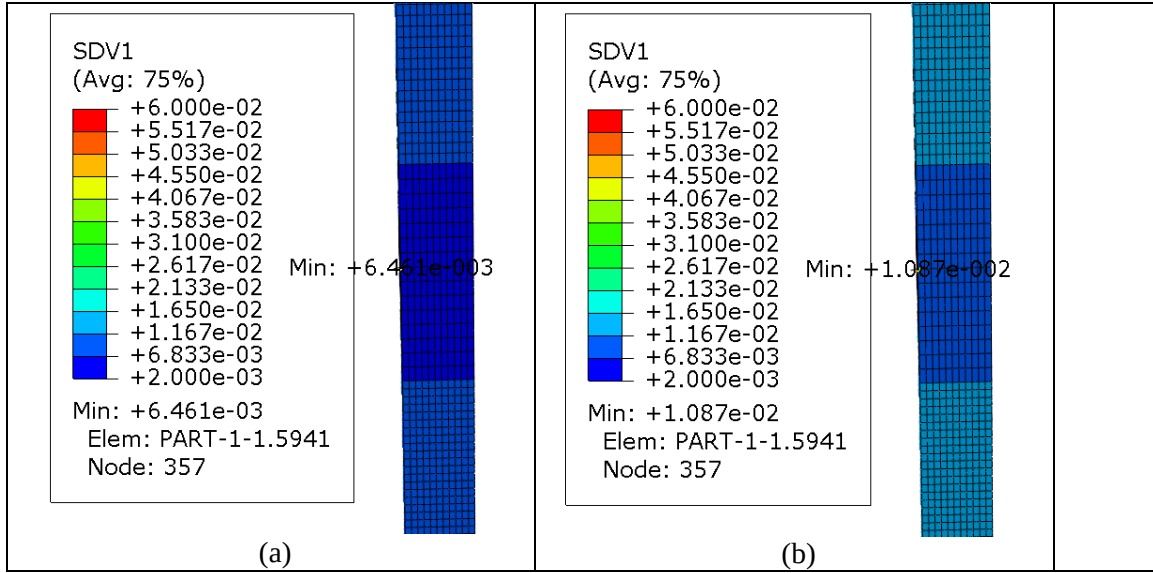


Figure 166: Cross-weld sample MT 10 tested at 1700F, 15Ksi. Creep strain distribution in the cross-weld samples at (a) 43 hr, (b) 49 hr, as predicted by the model.

5.4. Appendix: comparison of VDG vs. CA model for diffusion damage

The second model tested for the material at hand was proposed by Van der Giessen and collaborators⁵¹ (VDG). According to this model the rate of the diffusion damage is proportional to the diffusion strain rate, as:

$$\dot{\omega}_{diff} = 2 \dot{\epsilon}^{diff^{eq}} \omega_{diff} \frac{1}{\left(\frac{a}{L}\right)^3 \ln\left(\frac{1}{r}\right) - 0.5(3-r)(1-r)} \quad (5. A1)$$

Where L is a characteristic length depending on the shear modulus G , applied effective stress $\tilde{\sigma}^{eq}$, ratio between melting T_m and current temperature T ⁵³

$$L = L_0 \exp\left(\kappa \frac{T_m}{T}\right) \left(\frac{0.001G}{\tilde{\sigma}^{eq}}\right)^{3(n-1)} \quad (5. A2)$$

$L_0 = 0.00257\mu\text{m}$, $\kappa = 3.90$, $n = 4.6$ ⁵³, and r is a ratio depending on the cavity size ($2a$), inter-cavities space ($2b$) and characteristic length L :

$$r = \max\left(\left(\frac{a}{b}\right)^2, \left(\frac{a}{a + 1.5L}\right)^2\right) \quad (5. A3)$$

Both diffusion damage models (5.13), (5.A1), and (5.A2) have been tested separately (in conjunctions with Eqs. (5.1)-(5.12)). For each of the two, the same scaling parameter C was used in Eq.(5.8).

The predictions are presented in terms of the applied stress versus time to reach 0.4% strain and 1% strain for 100% weld and cross-weld in Figure 167 to Figure 170. It can be observed that, while the VDG model makes good predictions for cross weld bars under long creep test at 1500°F, 15Ksi (Figure 168), it predicts longer time to reach 0.4% strain and 1% strain at low stress (6, 8Ksi), 1700°F, than in the experiments. At all temperatures, the VDG model predicts significantly longer time to reach a certain percentage, than the CA model. A simple exercise was done by fitting the parameter C used in Eq.(5.8) for the VDG model at 10Ksi and 15Ksi, 100% weld material. With this calibrated parameter the predictions of the two models become comparable (Figure 171 and Figure 172). At lower temperatures (1500°F and 1600°F), the CA model seems to be slightly more conservative but the difference is minimal. Thus, the VDG model does not improve the creep strain predictions and does not reduce the number of parameters.

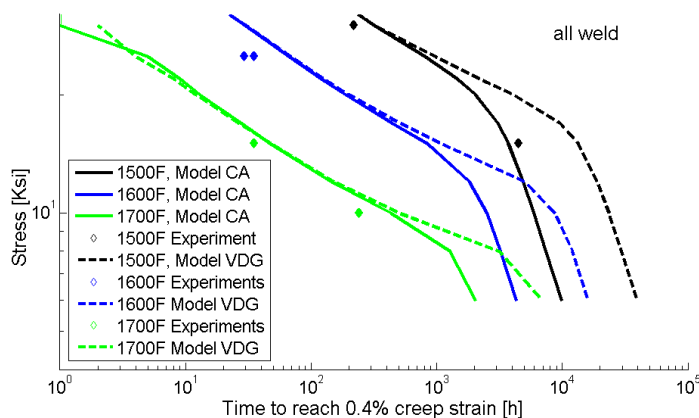


Figure 167: Stress vs time to reach 0.4% creep strain for 100% weld bars. Two models were tested for the diffusion creep: Cocks and Ashby approach (CA) and Van der Giessen approach (VDG).

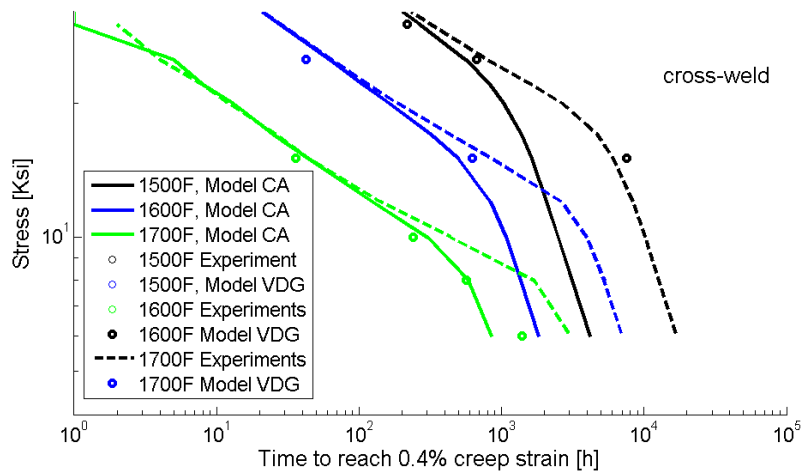


Figure 168: Stress vs time to reach 0.4% creep strain for cross-weld bars. Comparison between Cocks and Ashby approach (CA) and Van der Giessen (VDG) model predictions. The models are represented by lines, the experimental values, with symbols

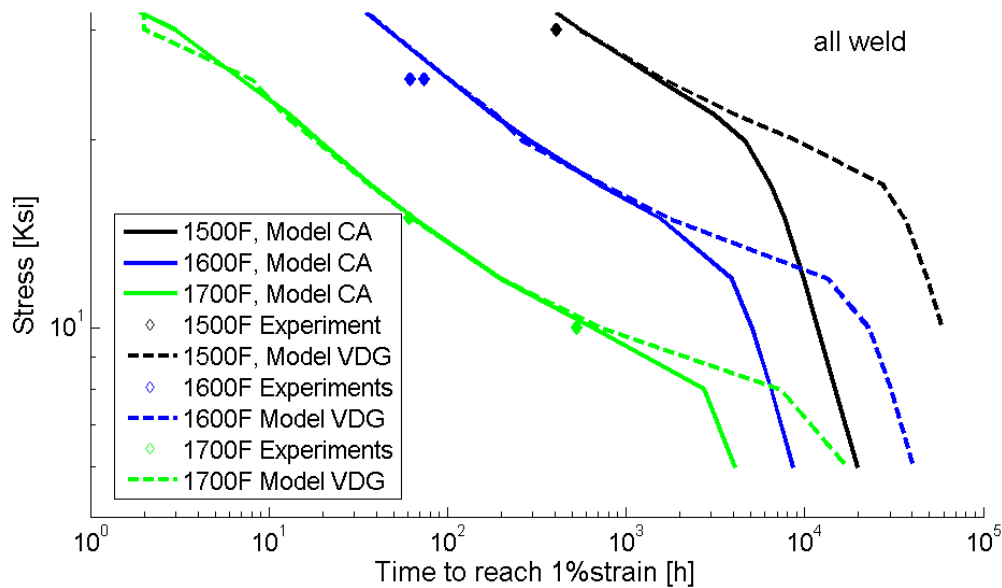


Figure 169: Stress vs time to reach 1% creep strain for 100% weld bars. Two models were tested for the diffusion creep: CA (Cocks and Ashby approach (CA) and Van der Giessen (VDG) approach. The experimental values are plotted with symbols.

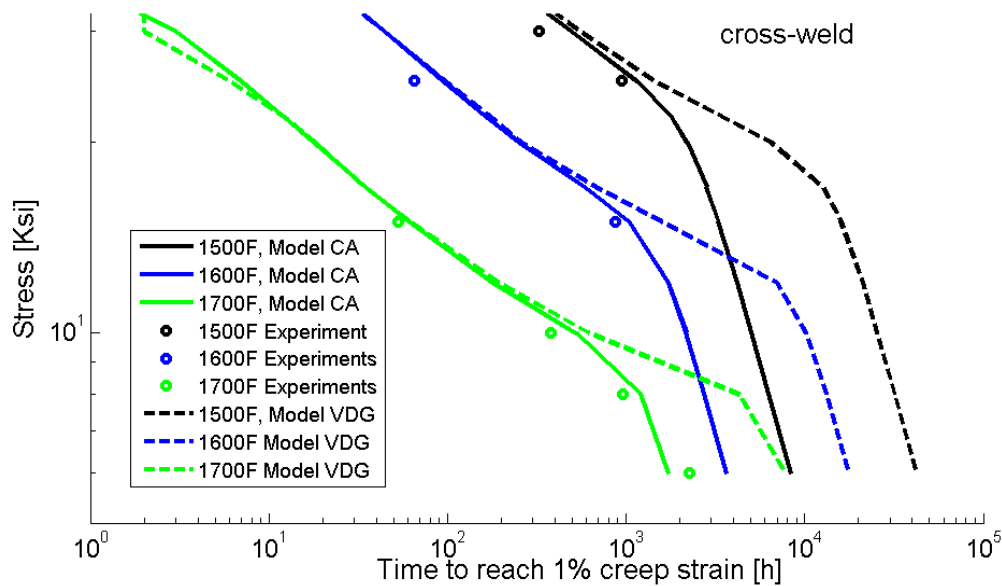


Figure 170: Stress vs time to reach 1% creep strain for cross-weld bars. Comparison between Cocks and Ashby approach (CA) and Van der Giessen (VDG) model predictions. The models are represented by lines, the experimental values, with symbols.

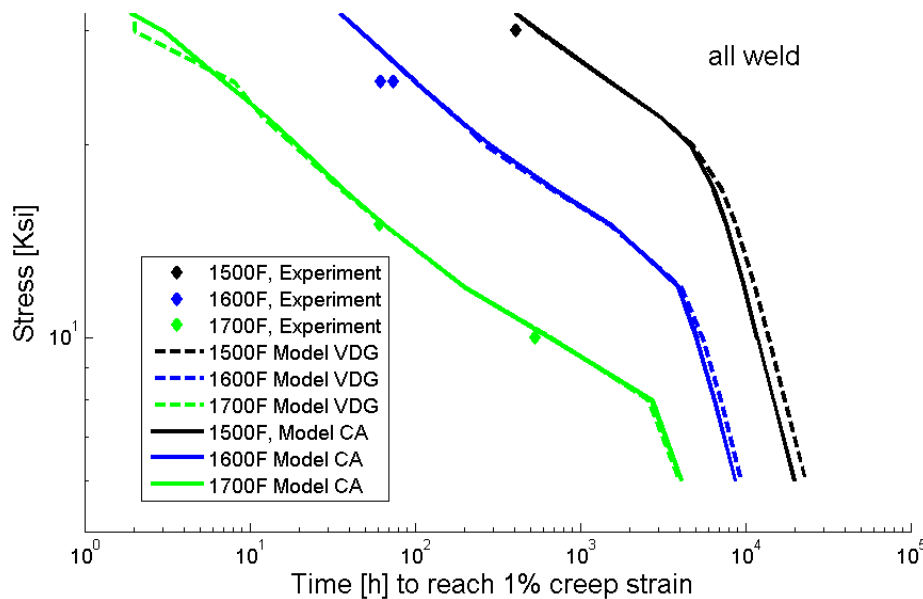


Figure 171: Stress vs time to reach 1% creep strain for 100% weld bars. Two models were tested for the diffusion creep: CA (Cocks and Ashby approach (CA) and Van der Giessen (VDG) approach – with recalibrated parameter C. The experimental values are plotted with symbols.

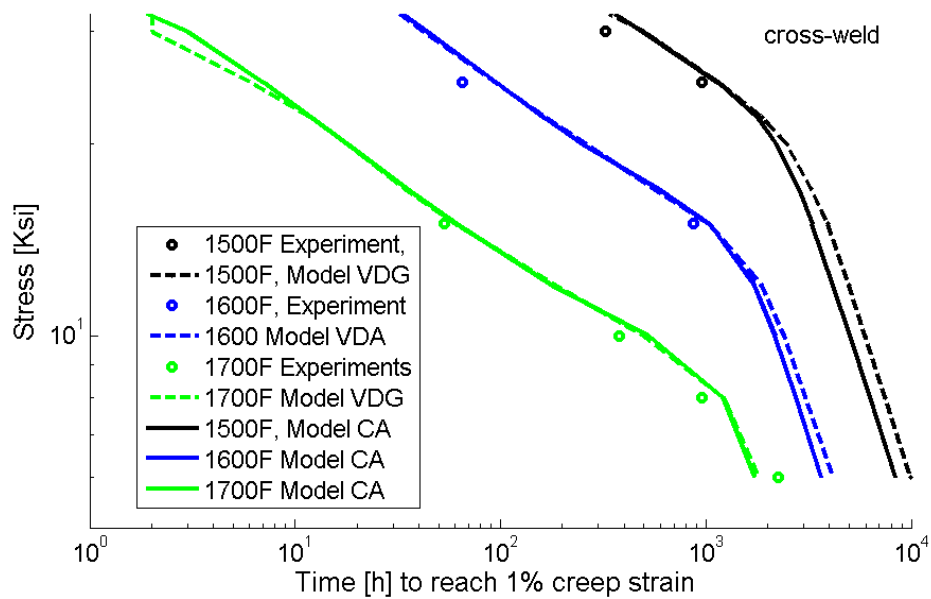


Figure 172: Stress vs time to reach 1% creep strain for cross-weld bars. Two models were tested for the diffusion creep: CA (Cocks and Ashby approach (CA)- with previously calibrated parameter C and Van der Giessen (VDG) approach – with recalibrated parameter C. The experimental values are plotted with symbols.

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