Development of High Performance Ni-base Alloys for Gas Turbine Wheels Using a Co-precipitation Approach

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Outline

- Background
- Co-precipitation Concept
- Present Status:
  - Summary of Round-2 mechanical testing results
  - Round-3 alloy design strategies and modeling
- Progress against milestones
- Summary

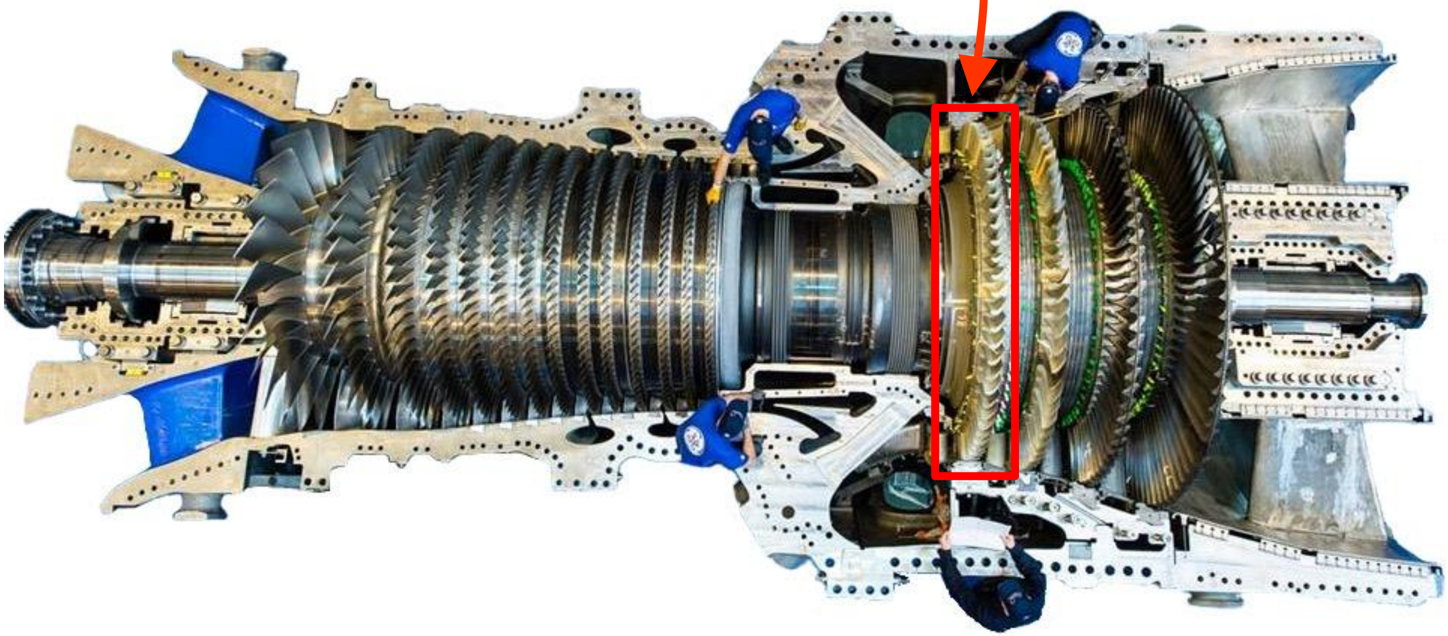
Path to Higher Efficiency Gas Turbine

Today's combined cycle efficiency is ~62%

Pressure Ratio: Higher
Firing Temperature: Higher
Sealing Flows: Lower
Cooling Flows: Lower

INCREASING WHEEL TEMPERATURE

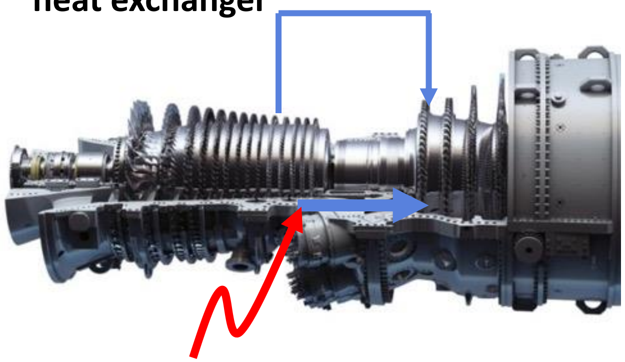
Future turbine combined cycle efficiency is ~65%



Next generation heavy duty gas turbine wheels must operate at higher temperatures to enable combined cycle efficiency improvements

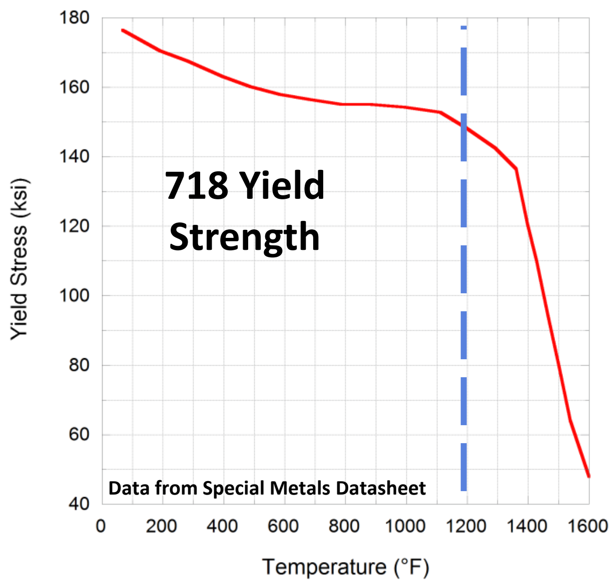
Designing a Higher Temperature Capable Wheel

Use steel and cool to lower the effective temperature



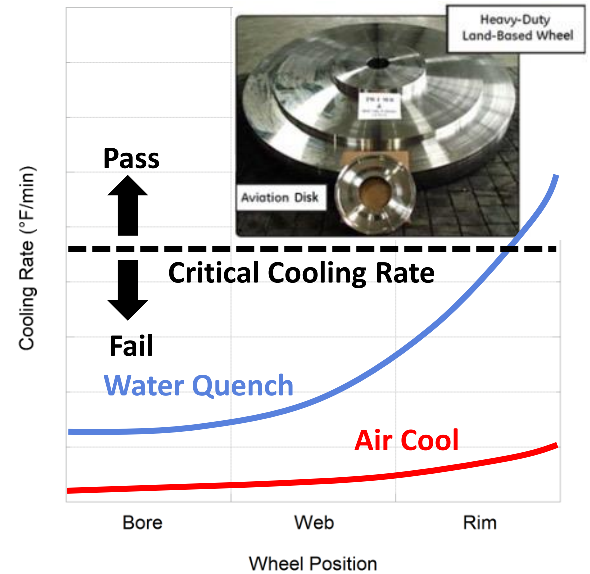
Cooling leads to reduced efficiency, increased complexity, & reliability risks.

Invent a better γ'' (Ni_3Nb) strengthened alloy



γ'' strengthening phase is unstable at temperatures >1200°F.

Use an Aviation disk alloy strengthened with γ' (Ni_3Al)



Precipitation kinetics result in severe over aging of γ', yielding poor properties.

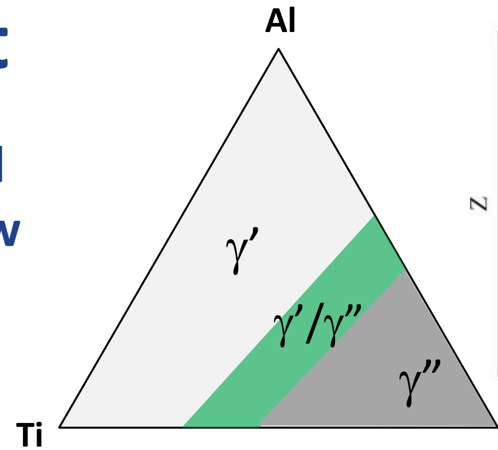
A new approach to alloy design is required to enable high temperature wheels

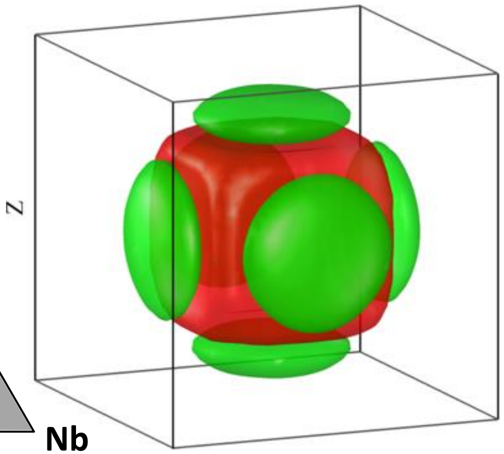
Coprecipitation Concept

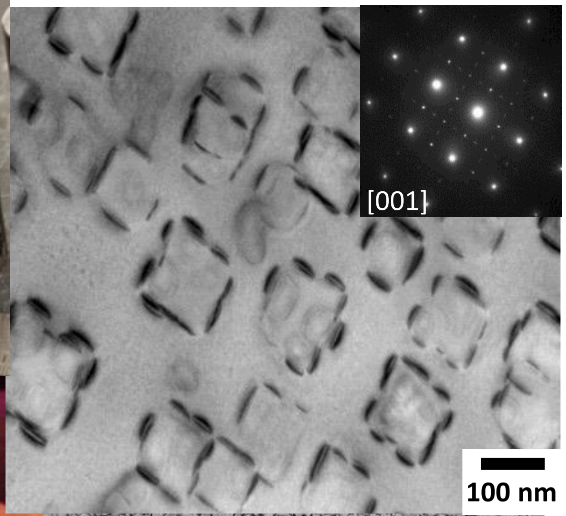
- Leverage the coprecipitation of γ’ and γ’’ to restrict γ’ coarsening during slow cooling of thick section components.
- Desire “compact” coprecipitate structure first identified by *Cozar and Pineau (1973)*

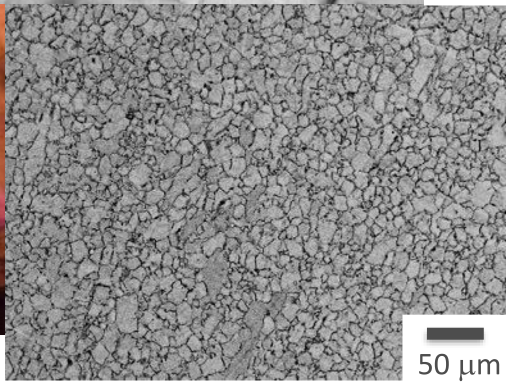
In previous NETL program (DEFE0026299):

- Developed successful sub-scale billetizing/forging procedure
- Screened tensile and hold time fatigue crack growth properties (*no creep testing or evaluation*)









Program Structure

Program Timeline

Mo. 1:
Program Initiation

Mos. 2-8:
Round 1 Alloys (2)

Mos. 9-20:
Round 2 Alloys (4)

Mos. 21-32:
Round 3 Alloys (2)

Mos. 33-34:
Program Wrap-Up

Mo. 35-36:
Documentation

1.0 Project Management [OSU]

2.0 Alloy Processing and Evaluation

- **2.1 Alloy Selection and Processing [GRC]**
- **2.2 Heat Treatment and Microstructure Characterization [GRC]**
- **2.3 Aging Informed by High Temperature Indentation [OSU]**
- **2.4 Phase Field Modeling and Characterization of Coprecipitation [OSU]**
- **2.5 Mechanical Testing [OSU/Metals Tech]**

3.0 Property Model Development

- **3.1 Deformation Mechanism Characterization [OSU]**
- **3.2 Yield Strength Model [OSU]**
- **3.3 Creep Model [OSU]**

The Ohio State University

Professor ***Michael Mills*** (PI) overall program coordination and lead on mechanical testing and deformation microstructure characterization
Professor ***Yunzhi Wang*** (co-PI) will lead the modeling efforts

GE Global Research

Rich DiDomizio and Andy Detor lead alloy preparation and processing
Reza Shargi-Moshtaghin has performed preliminary TEM microstructure characterization of as-processed alloys

Metals Technology, Inc

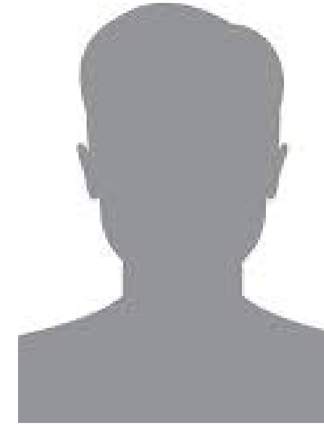
Robert Hayes will perform the creep testing as cost-share to the project

<u>PIs</u>:
Michael Mills
Yunzhi Wang

<u>Collaborators</u>:
Bob Hayes (MTI)
Rich DiDomizio (GER)
Andy Detor (GER)

<u>Experiments</u>:
Semanti Mukhopadhyay
Christopher Zenk

<u>Modeling</u>:
Kamalnath Kadrivel
Hariharan Sriram



Updates on Experimental Efforts

- **Microstructural Characterization of Proposed Round 3 Alloys**

  Three different design strategies were introduced in the previous meeting. GE has produced alloys based on these design strategies in as-cooled condition. OSU has started microstructural characterization of these alloys.

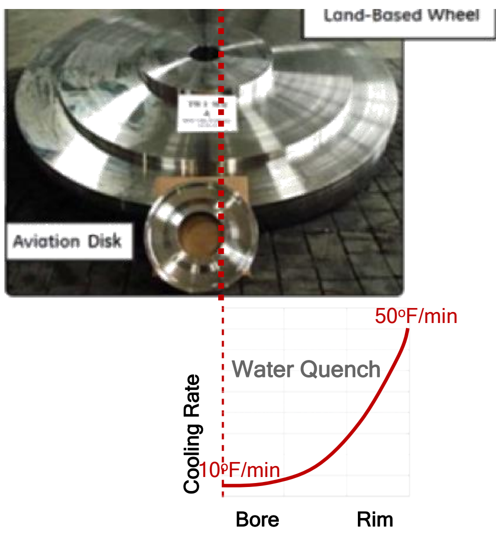
- **Microstructural Characterization of deformed alloys**

  OSU has been carrying out ECCI and STEM-HAADF to characterize and study deformation in Round1 and Round 2 alloys in as-crept condition. Two alloys (IN718-27 stabilized & IN718-50) crept at 1200^{o}F (650^{o}C) have been studied.
  Results are compared with deformation studies carried out on In-Situ tensile tested IN718-50.

Summary of Round 1 & Round 2 Alloys

- Alloys and Processing

Alloy Name (wt%)	γ' solvus	Ni	Cr	Fe	Al	Ti	Nb	Mo	C
718		52.5	19	18.9	0.5	0.9	5.13	3.05	0.02
718-011	855.8	52.5	19	18.9	1.5	---	5	3.05	0.02
718-027	830.1	52.9	18.7	18.9	1.1	0.95	4.4	3.05	0.02
718-050	854.8	54.4	17.3	17.5	1.9	---	6.2	2.8	0.02
718-051	917	56.9	14.9	15.3	2.4	---	8	2.5	0.02



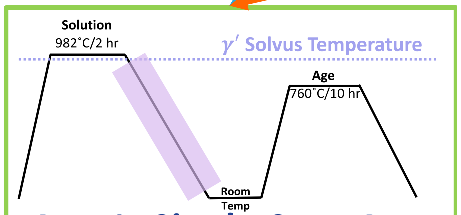
Solutionized at 982°C

Round 1 = alloy 27

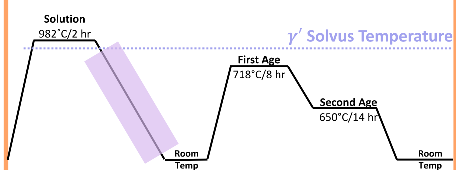
Round 2 = alloy 50

Slow Cooled 10°F/min

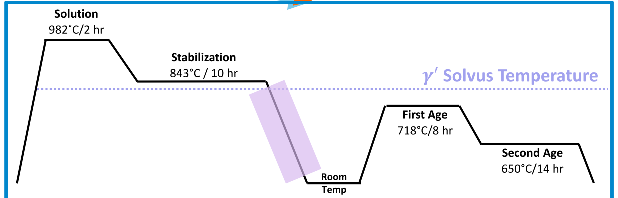
Fast Cooled 50°F/min



Age 1: Single Step Age



Age 2: Two Step Age

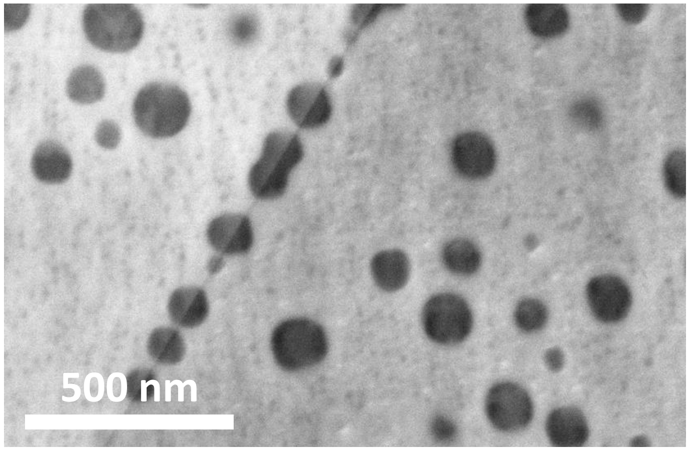


Age 3: Stabilization followed by Two Step Age

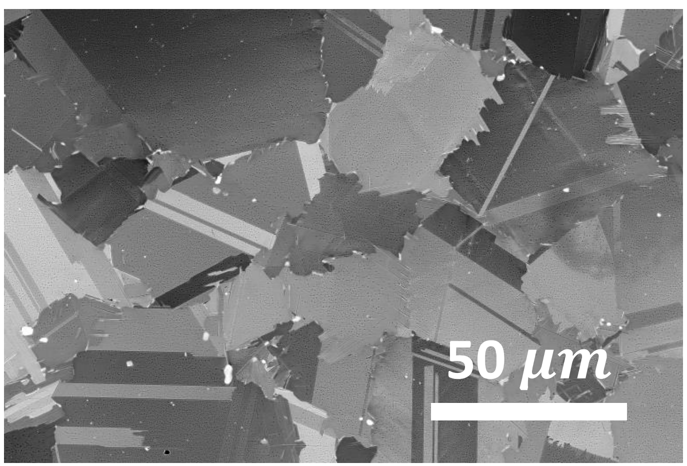
THE OHIO STATE UNIVERSITY

Summary of Round 1 & Round 2 Alloys

ROUND 1 (alloy 27)

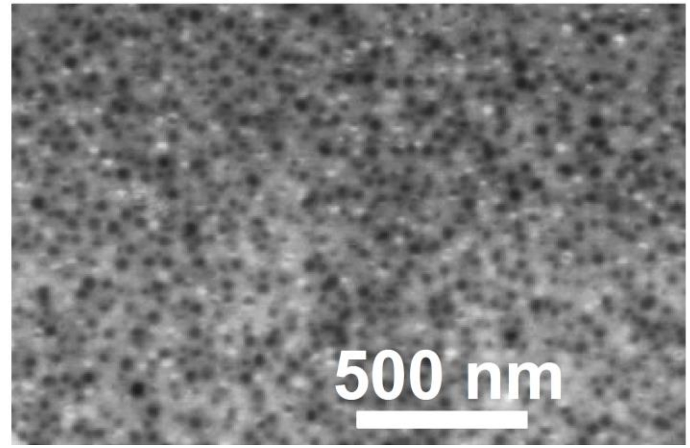


η and δ on GBs

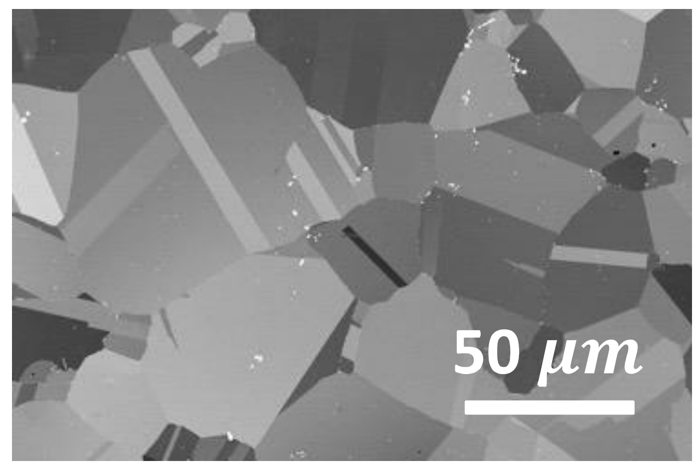


Bimodal coprecipitates
Secondary size – 150-200nm
Tertiary - 5-20nm

ROUND 2 (alloy 50)



No GB ppts

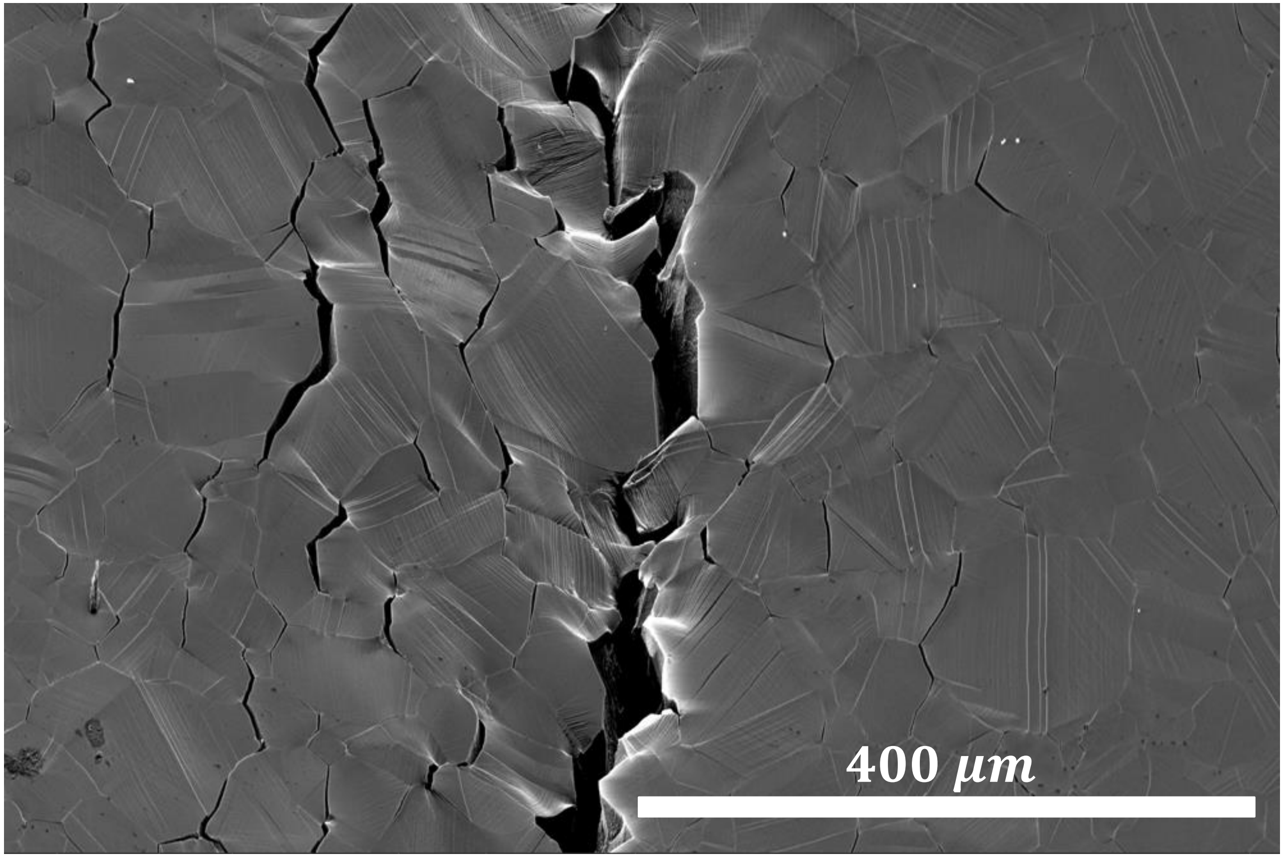


Unimodal coprecipitates
Size – 70-100nm

Summary of Round 2 Alloys (alloy 50)

FEI Thermo Fischer Quattro ESEM

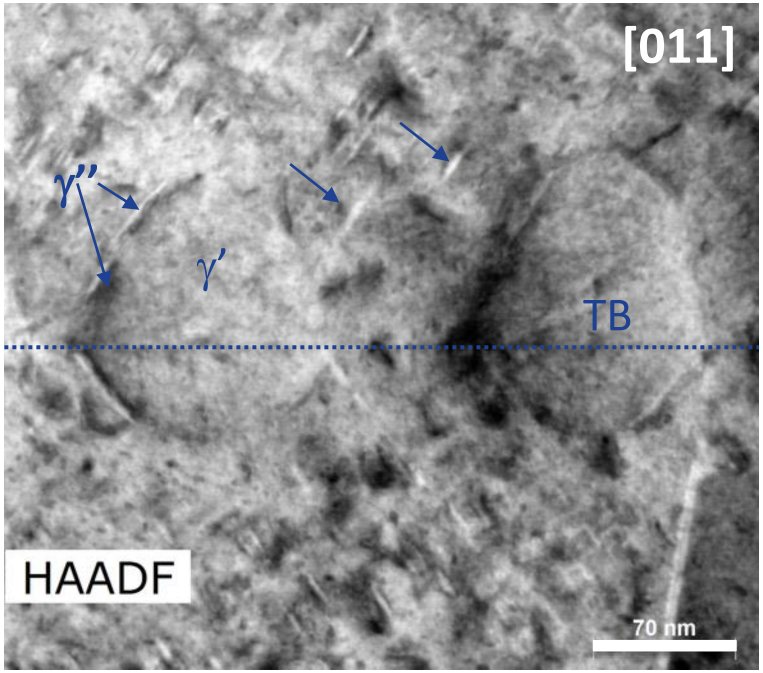
- Intergranular Failure

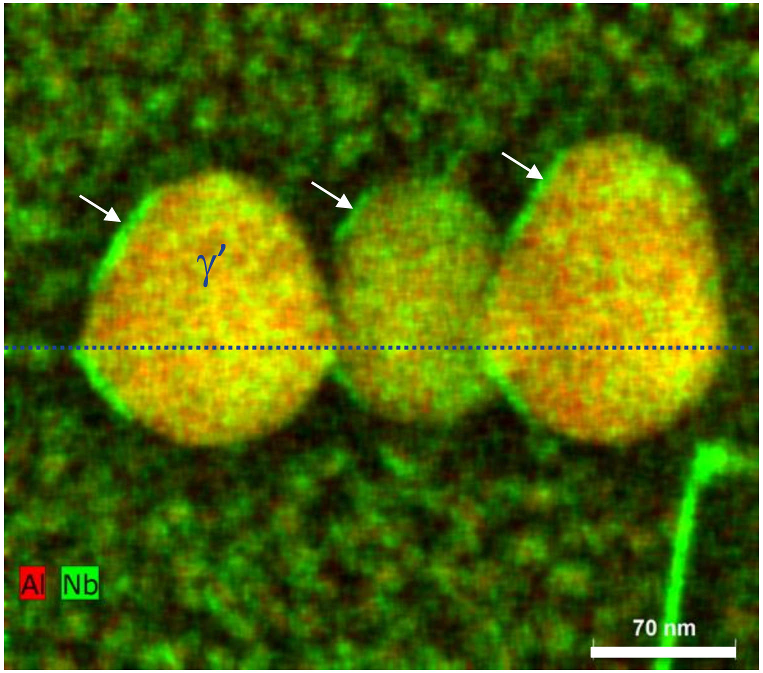


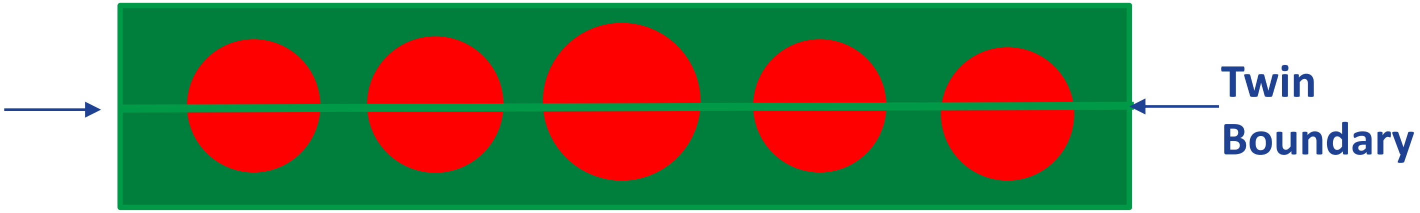
Develop understand in order to
mitigate in Round 3

Heterogeneous Precipitation at Annealing Twin Boundaries

Alloy 27



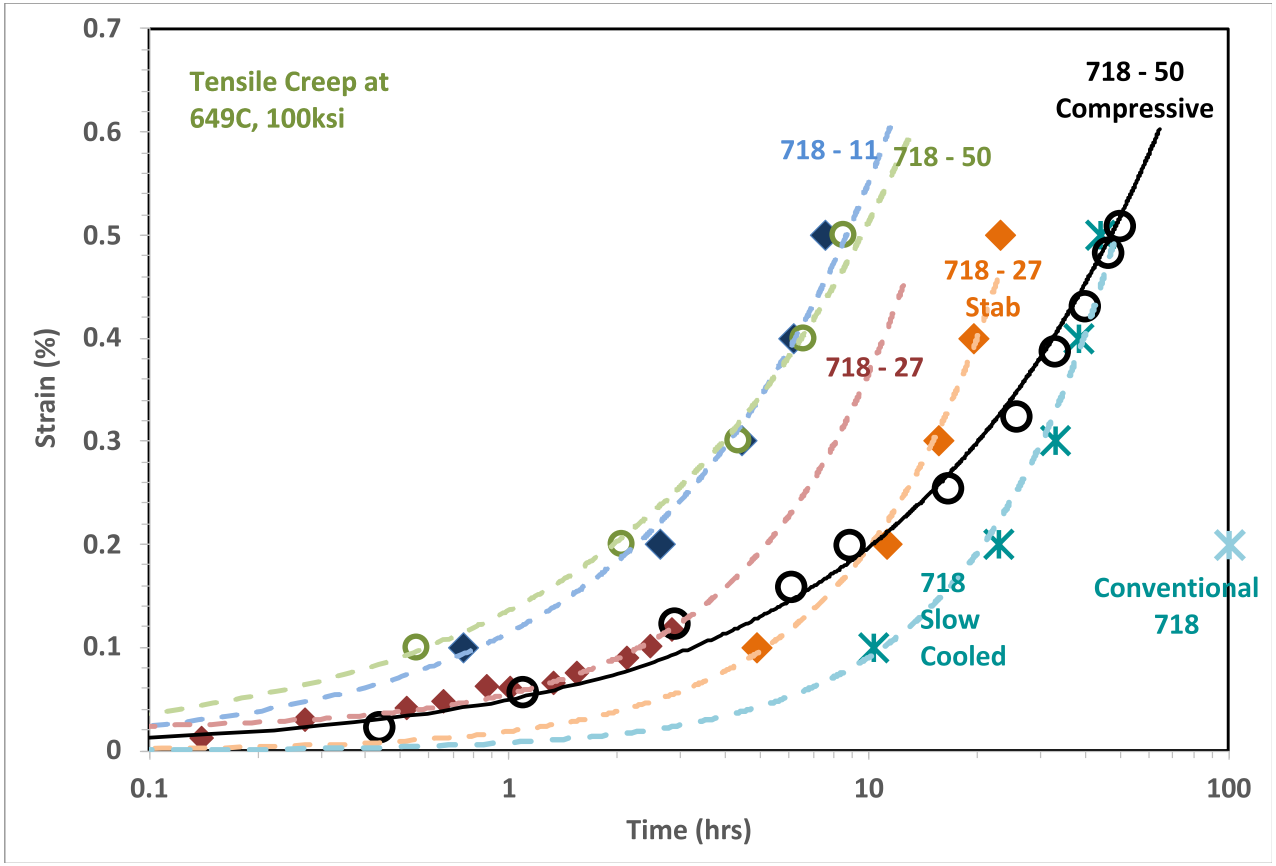




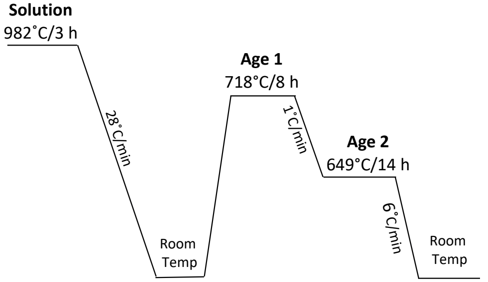
- **Not previously reported in superalloys**
- **Mechanism under investigation**

Summary of Round 1 & Round 2 Alloys

- Overall Mechanical Properties

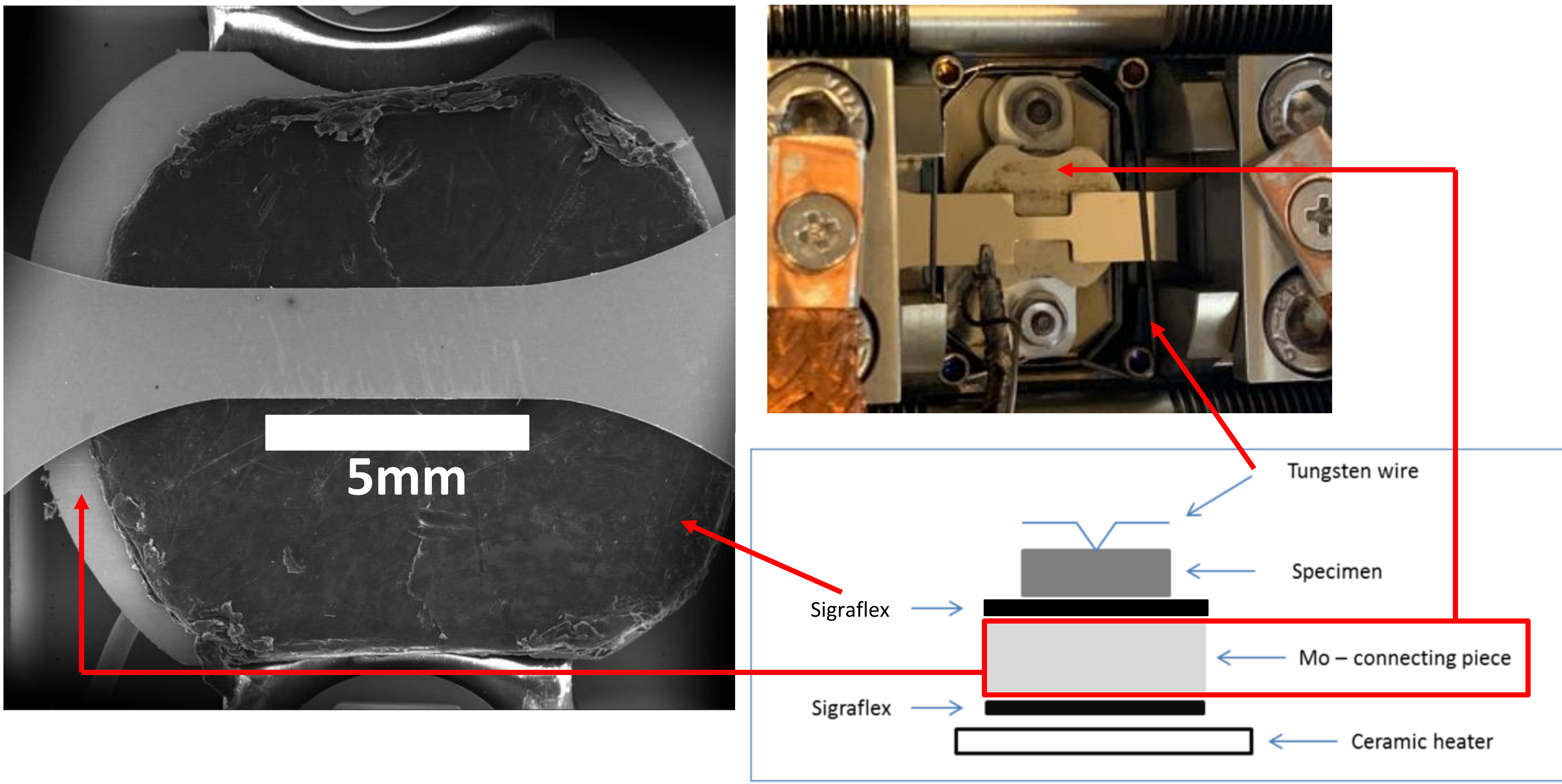


The 718 tested at 1200F, 100ksi was subjected to the following heat treatment:

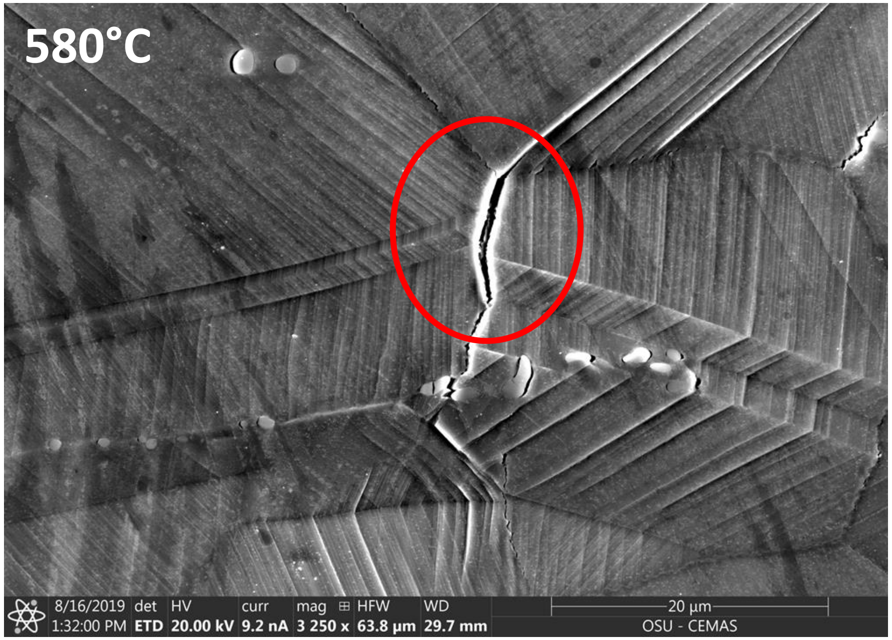


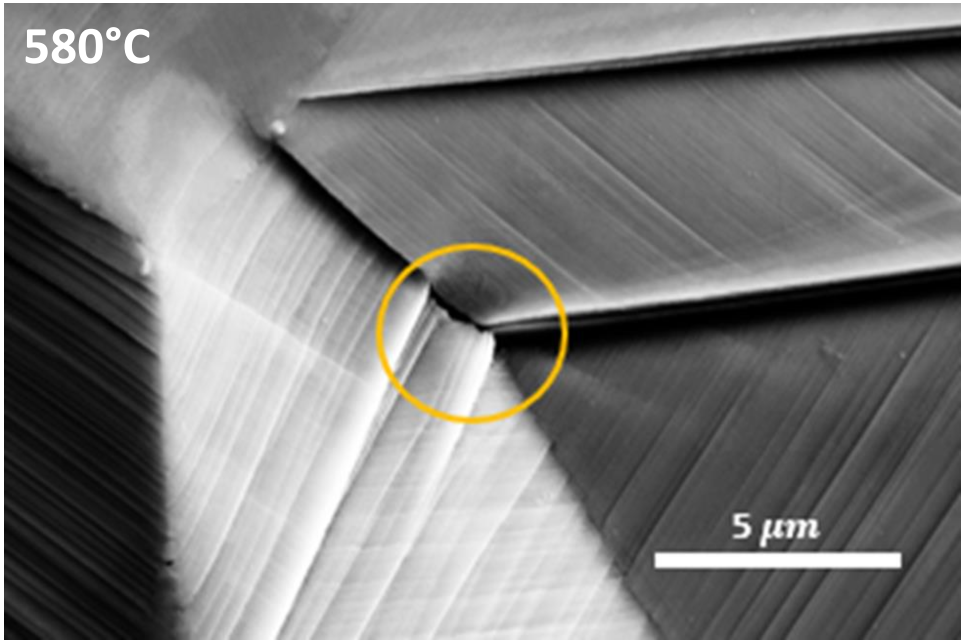
Is Situ SEM Tensile Testing

Tests at Room Temperature and 580ºC



Intergranular Crack Initiation

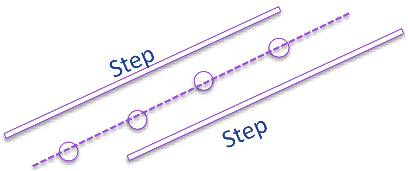


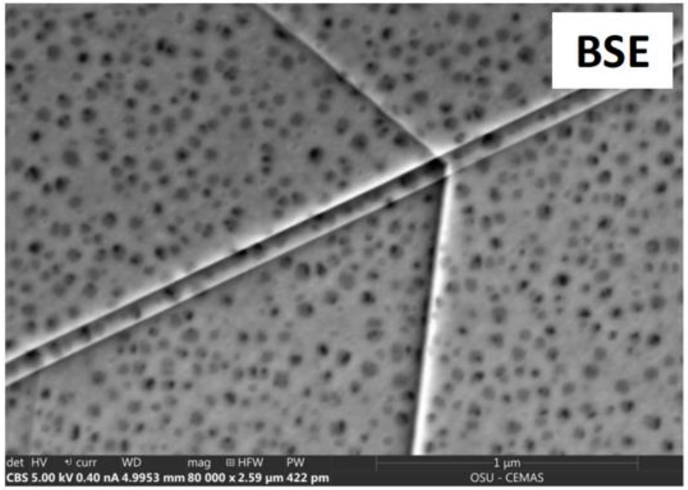


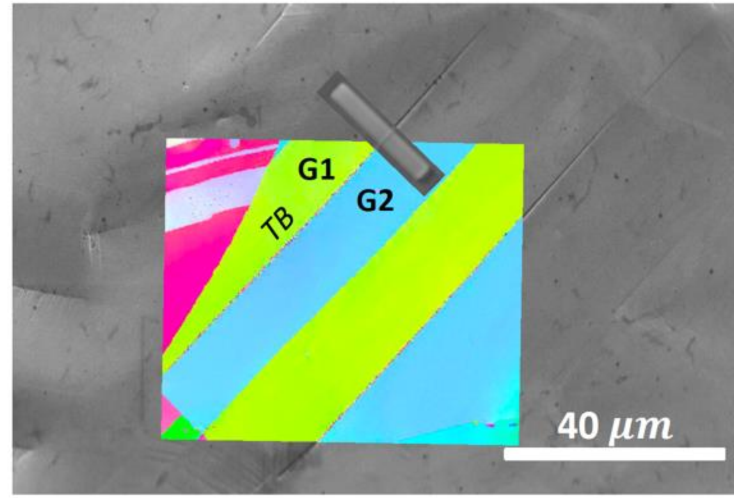
Intergranular cracks initiate frequently at the intersection of twin boundaries and grain boundaries

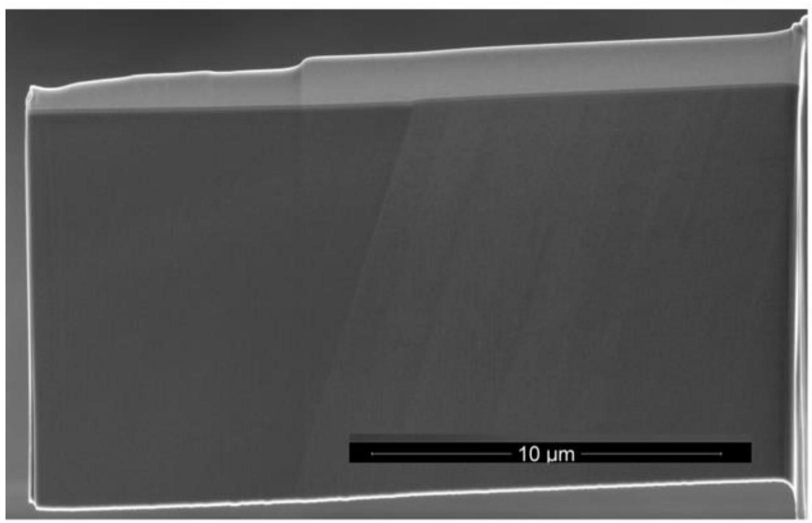
Summary of Round 2 Alloys

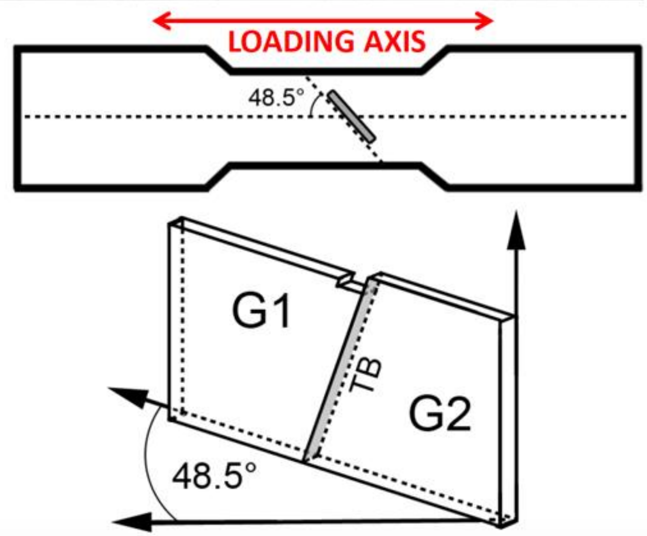
Localized and intense slip parallel to annealing twin boundaries



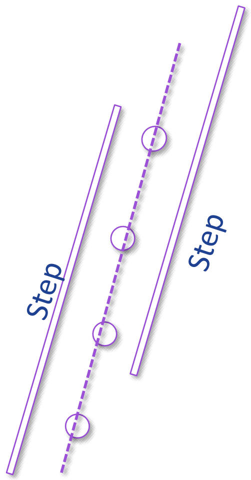


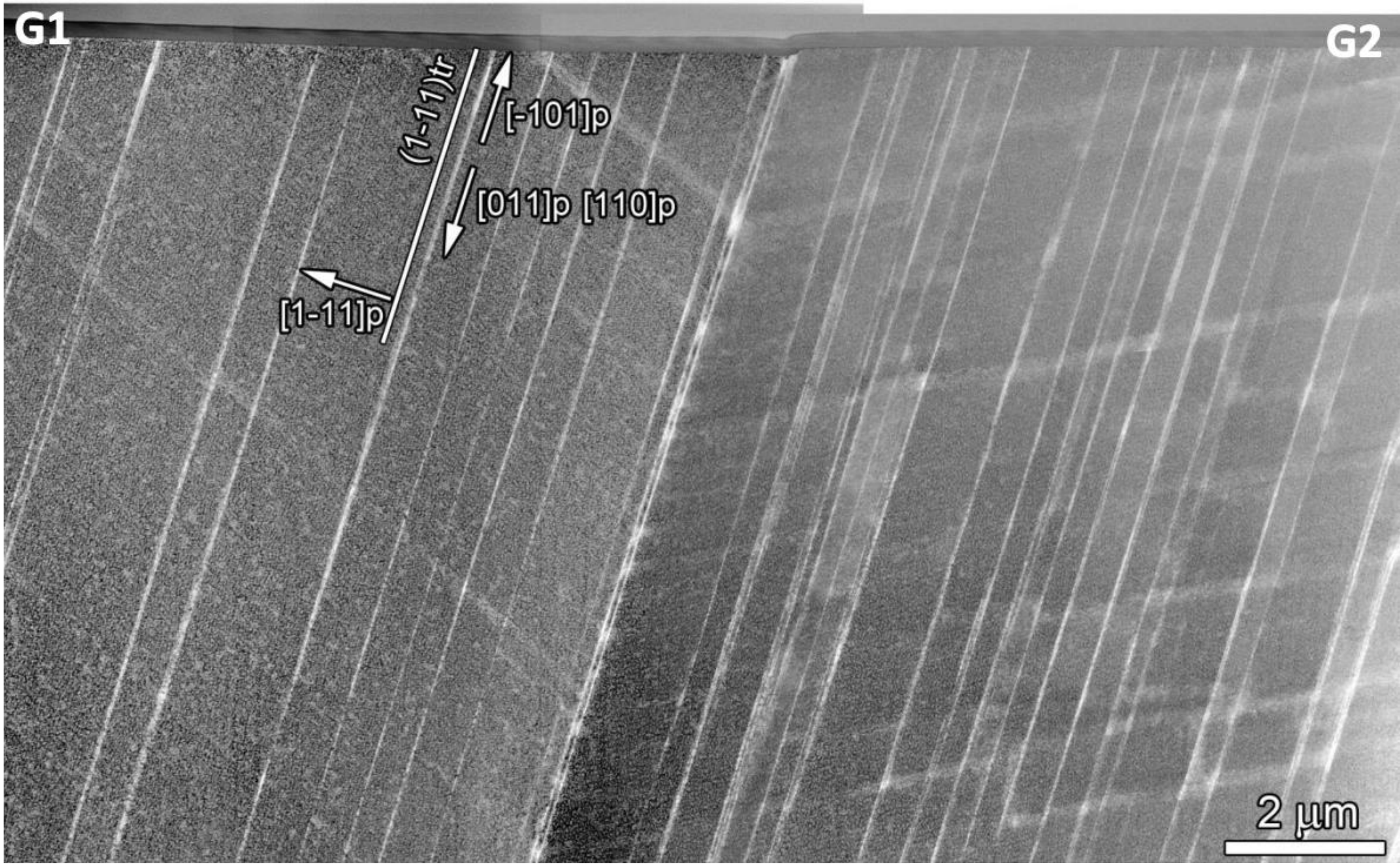




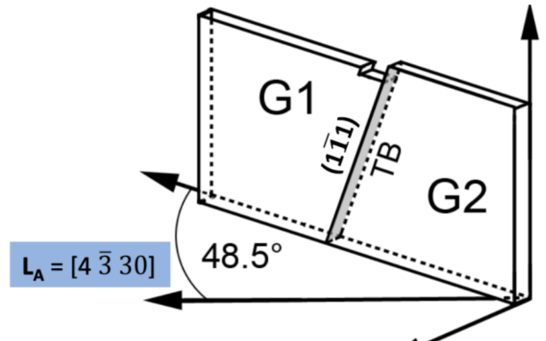


Characterization and Analysis of Slip

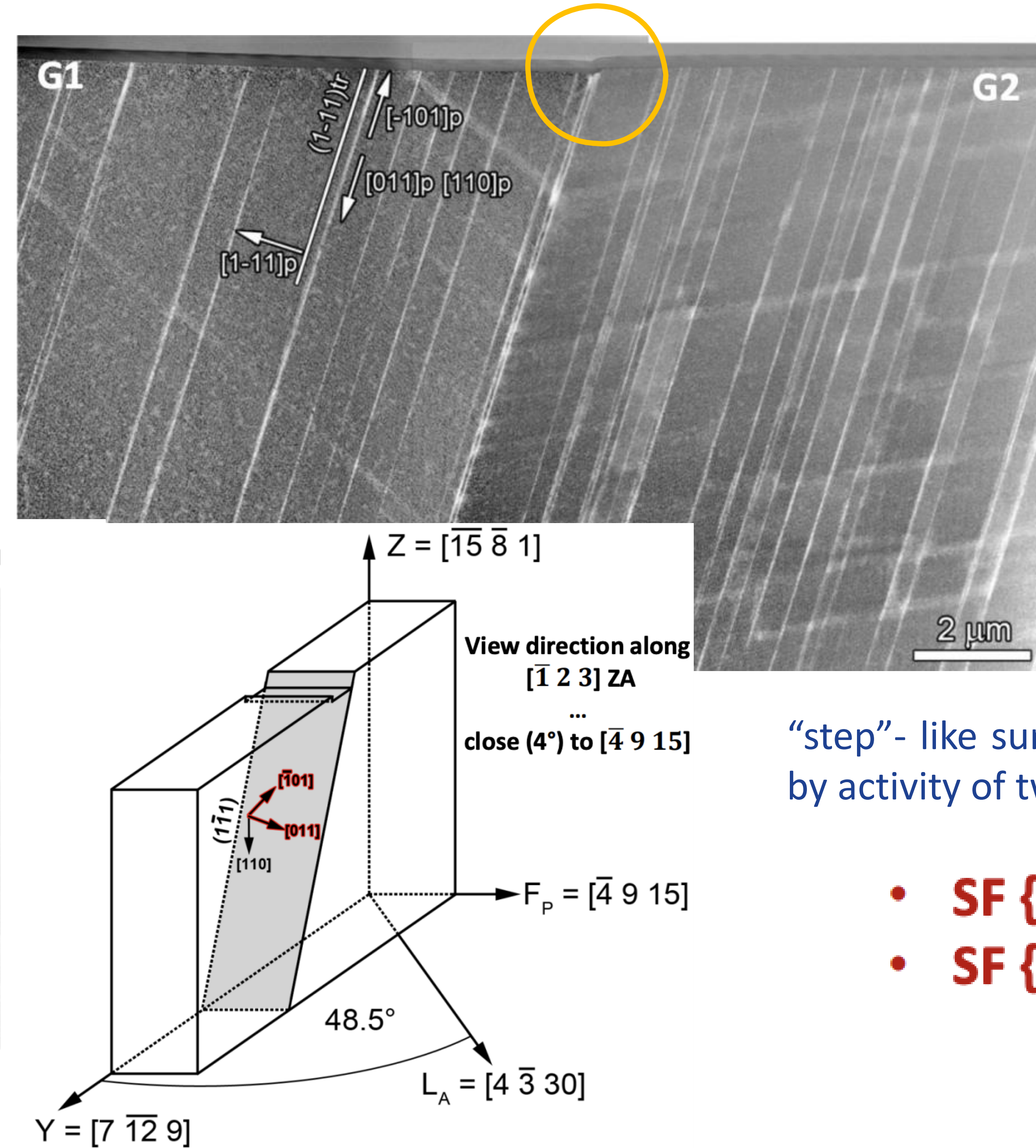




Loading axis direction	$\overrightarrow{L_A}$ = [4 $\bar{3}$ 30]
Foil plane direction	$\overrightarrow{F_P}$ = [$\bar{4}$ 9 15]
Beam directions and corresponding diffraction conditions	$\overrightarrow{B_1}$ = [$\bar{1}$23] ZA $\overrightarrow{g_{B1}}$ = [1$\bar{1}$1]
	$\overrightarrow{B_2}$ = [011] ZA



Characterization and Analysis of Slip



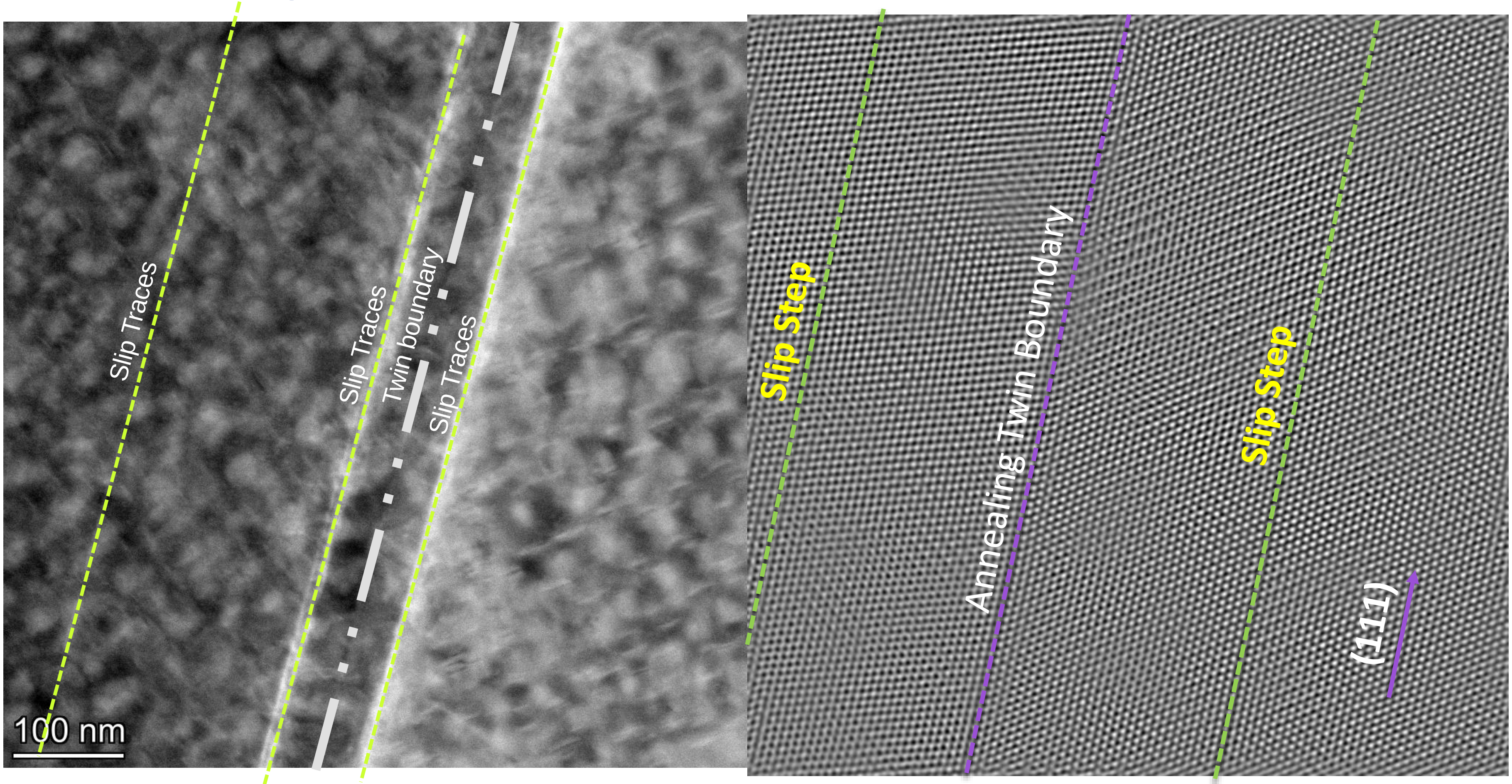
Schmid factor	Plane	Direction $\vec{b}$
0.356	(111)	[$\bar{1}$01]
0.096	(111)	[1$\bar{1}$0]
0.452	(111)	[01$\bar{1}$]
0.346	(11$\bar{1}$)	[011]
0.090	(11$\bar{1}$)	[1$\bar{1}$0]
0.435	(11$\bar{1}$)	[101]
0.345	($\bar{1}$11)	[101]
0.335	($\bar{1}$11)	[01$\bar{1}$]
0.010	($\bar{1}$11)	[110]
0.441	(1$\bar{1}$1)	[011]
0.425	(1$\bar{1}$1)	[$\bar{1}$01]
0.016	(1$\bar{1}$1)	[110]

"step"- like surface relief at $(1\bar{1}1)$ TB is formed by activity of two slip systems:

- **SF {(1$\bar{1}$1)[011]} = 0.441**
- **SF {(1$\bar{1}$1)[$\bar{1}$01]} = 0.425**

Summary of Round 2 Alloys

- High Resolution HAADF-STEM



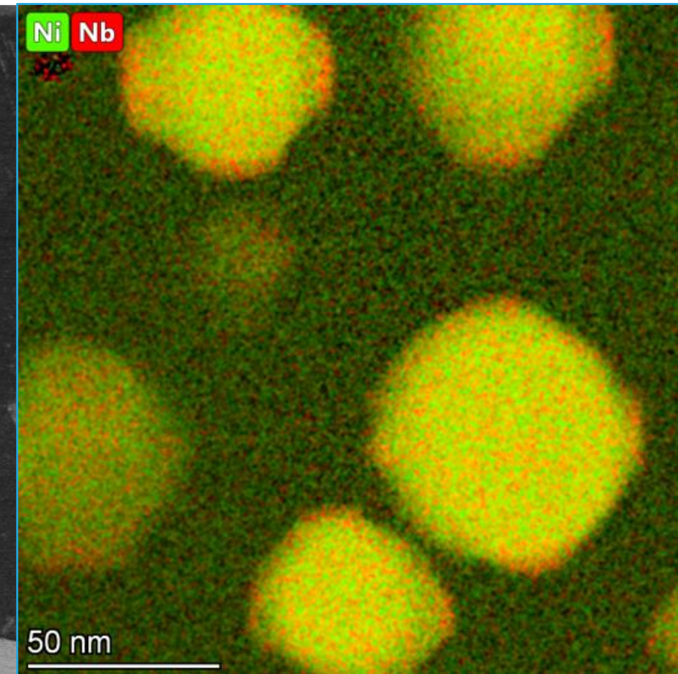
Localized and intense slip parallel to annealing twin boundaries

Characterization of Crept Sample

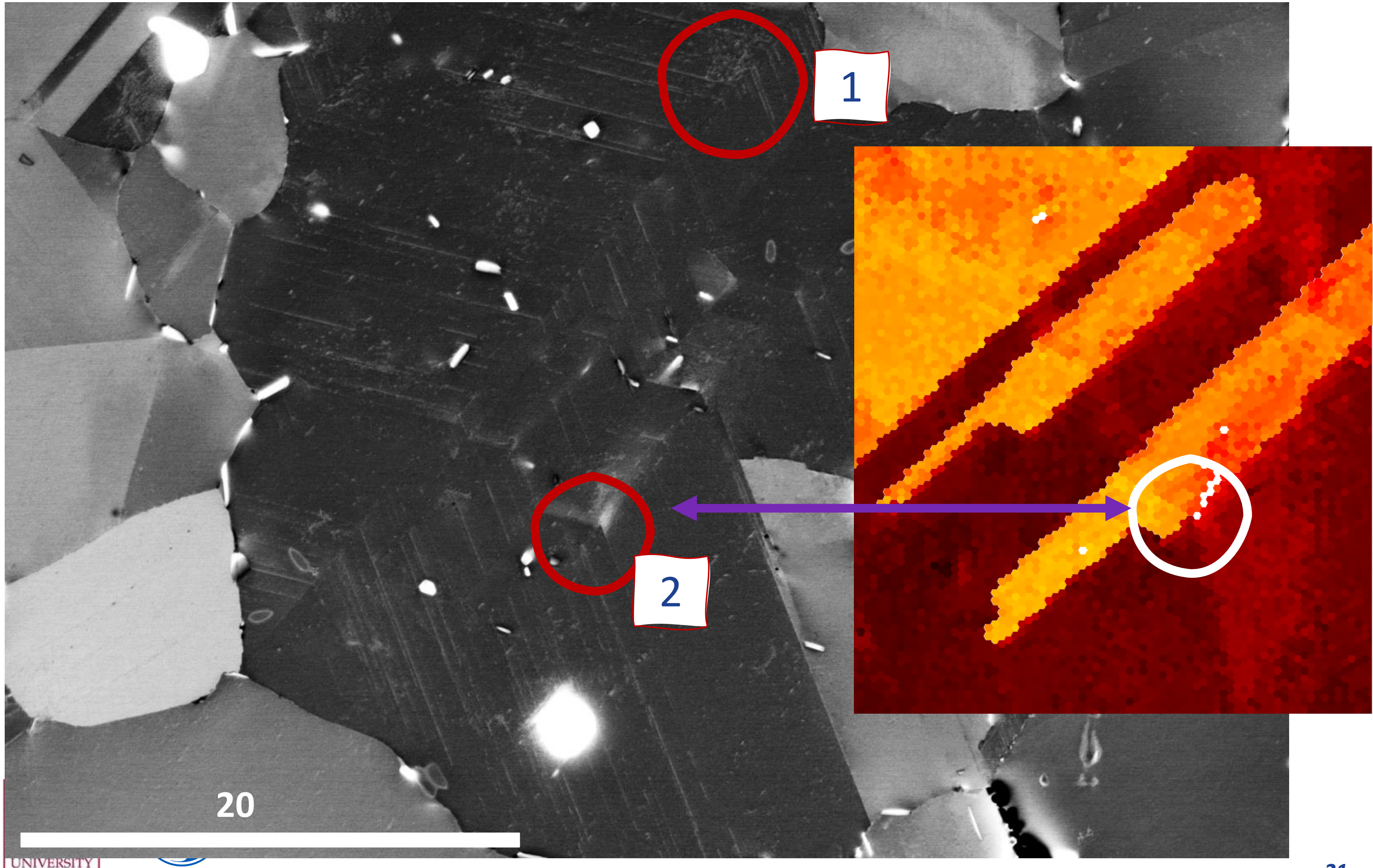
IN718-50 (HT-2)

Tension Creep at 1200°F (650°C) and 100 ksi (689 MPa)

This sample reached 0.5% strain at 8.5 hours. Sample ruptured at 10 hours & 0.6% strain

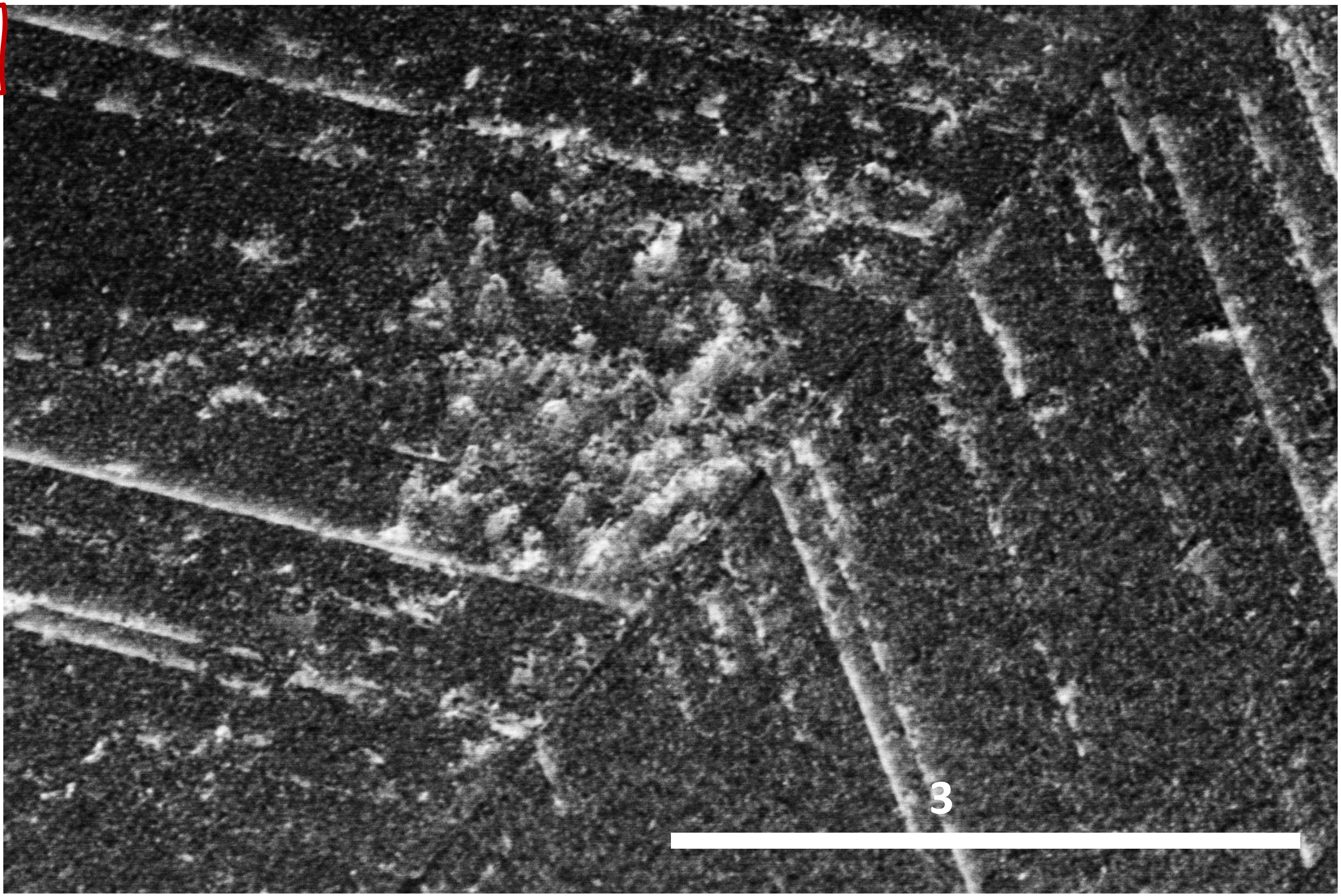


Characterization of Crept Sample

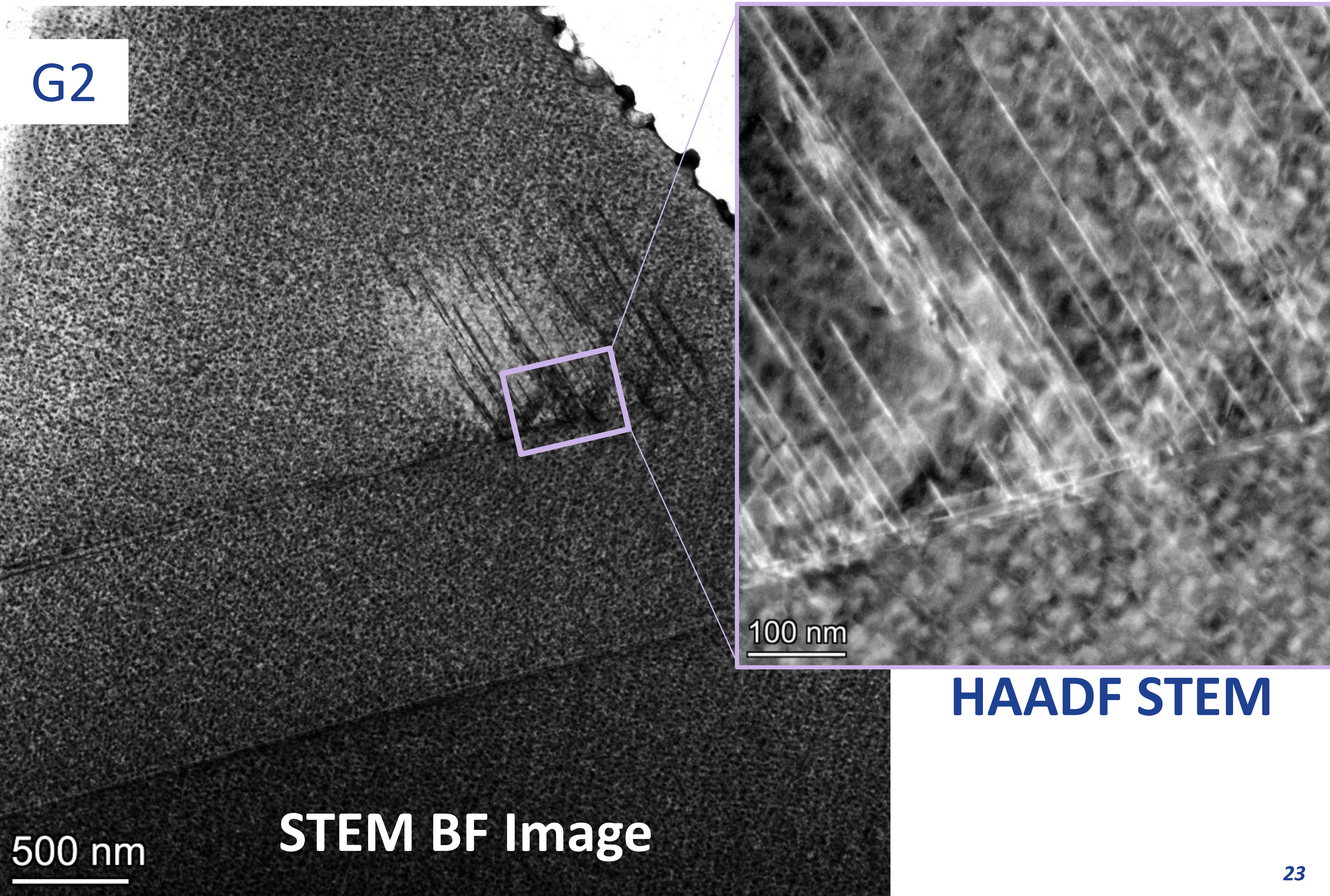


Characterization of Crept Sample

1

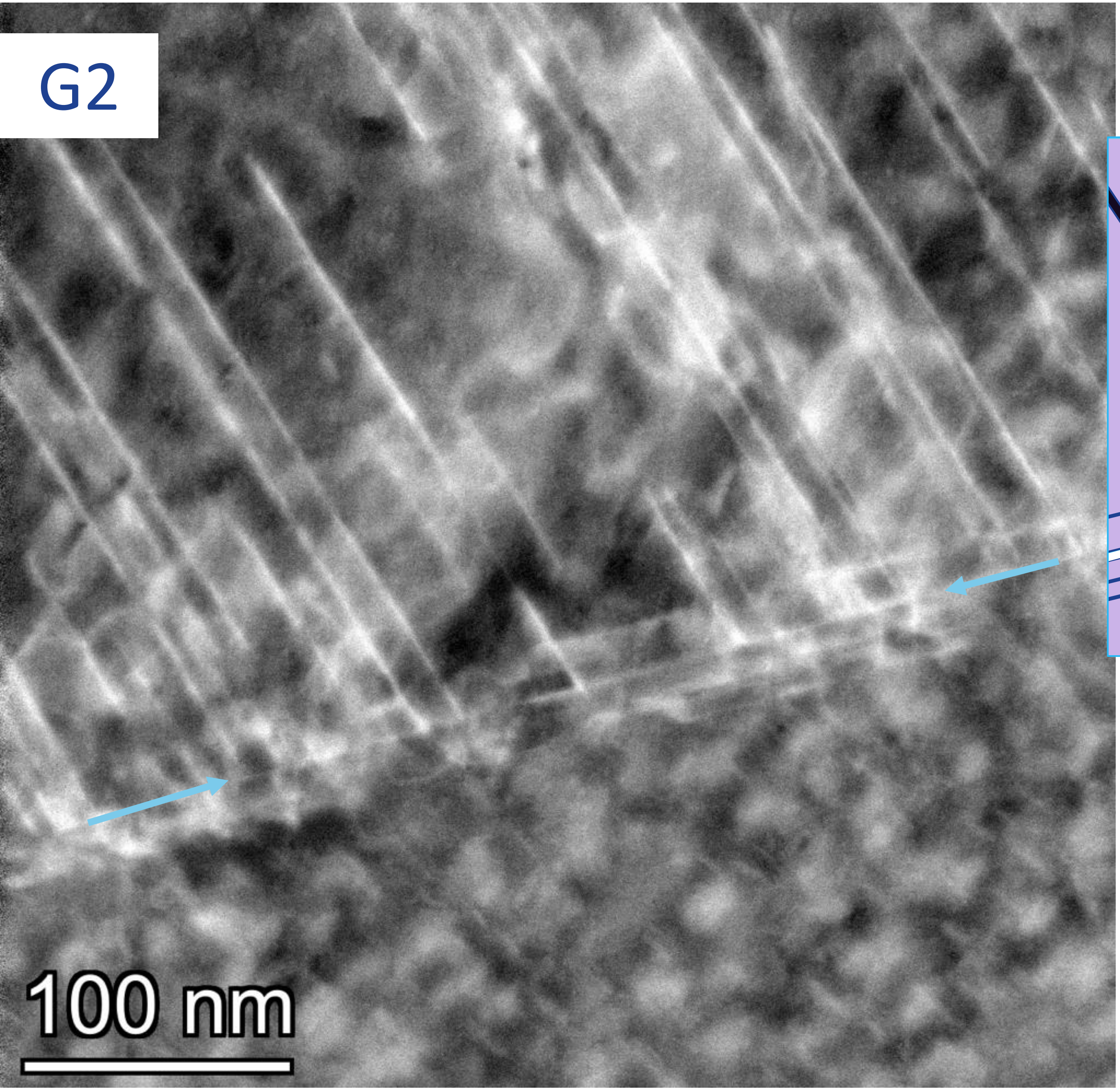


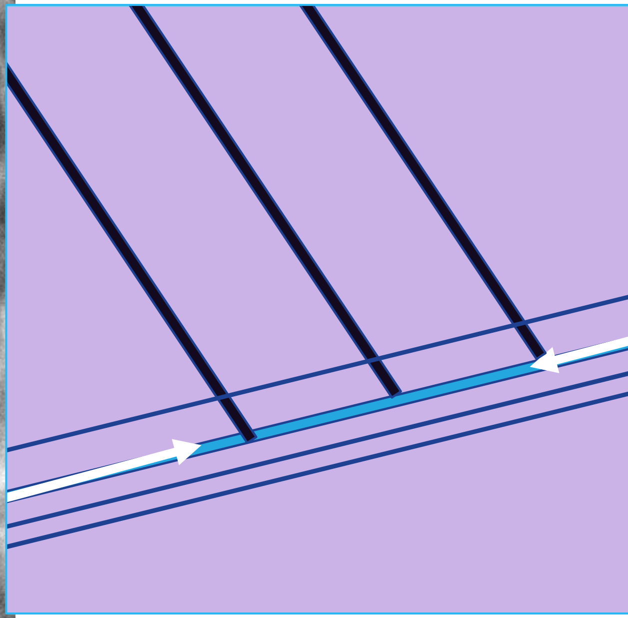
Characterization of Crept Sample



HAADF STEM

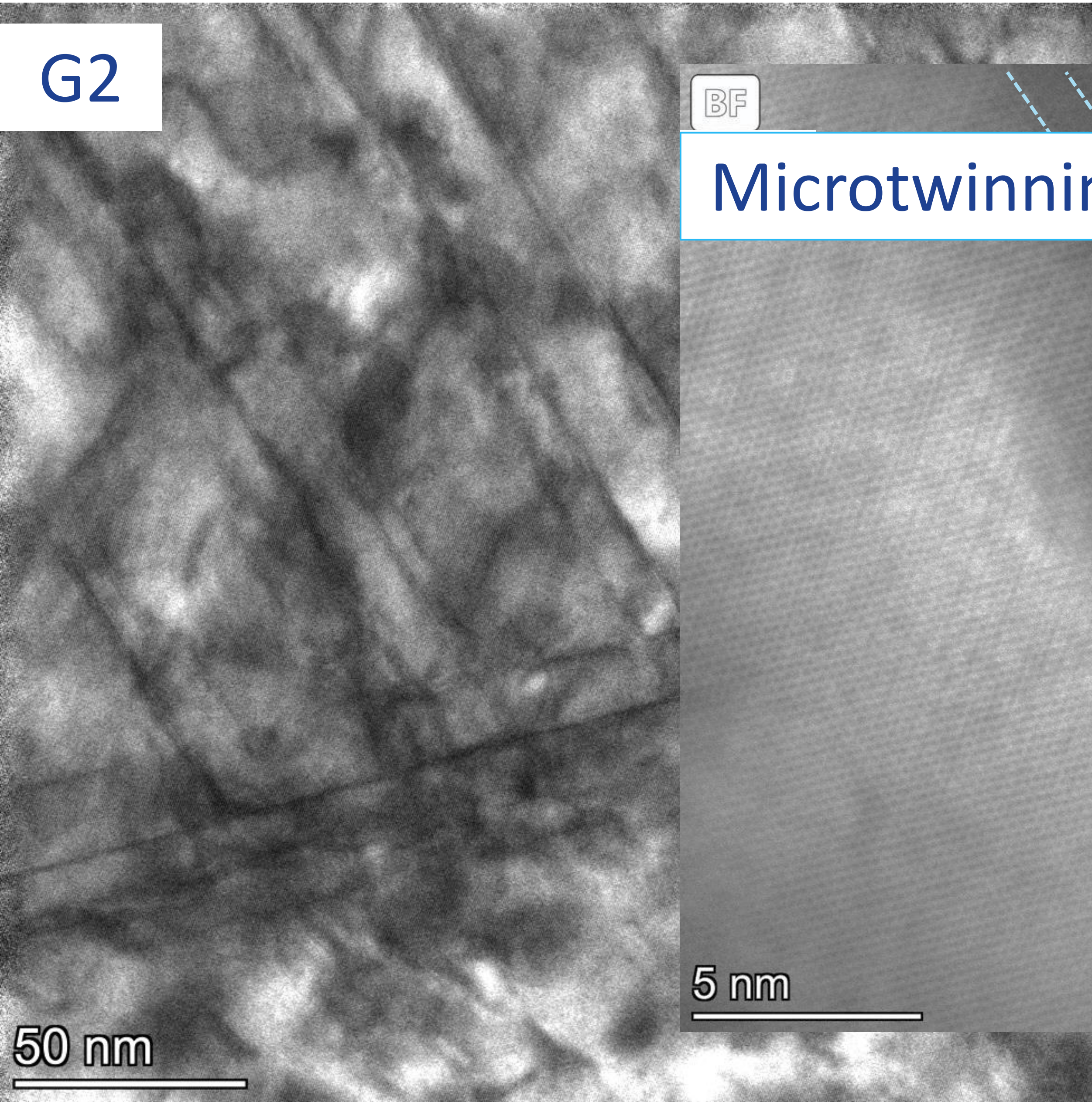
Characterization of Crept Sample

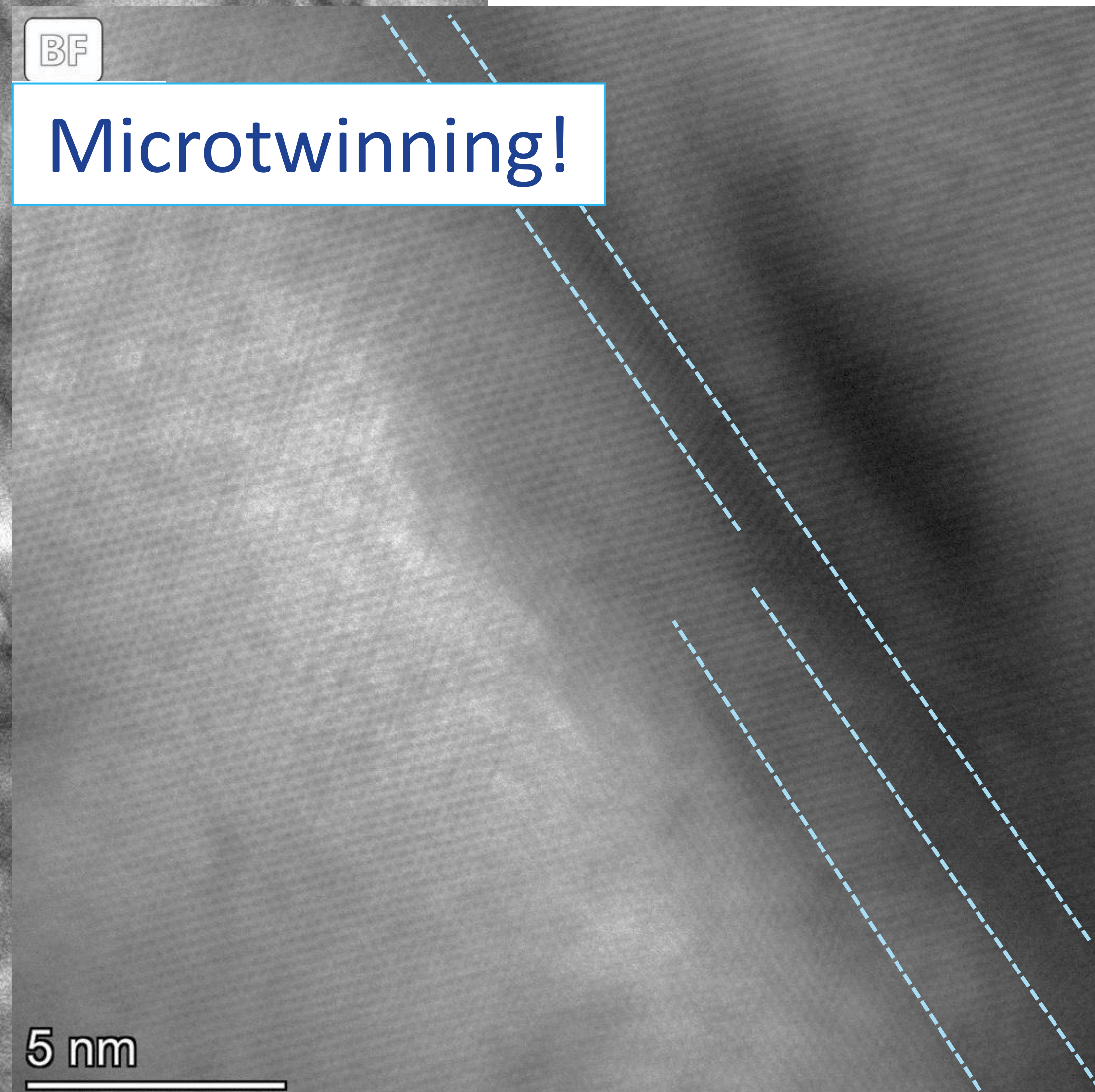




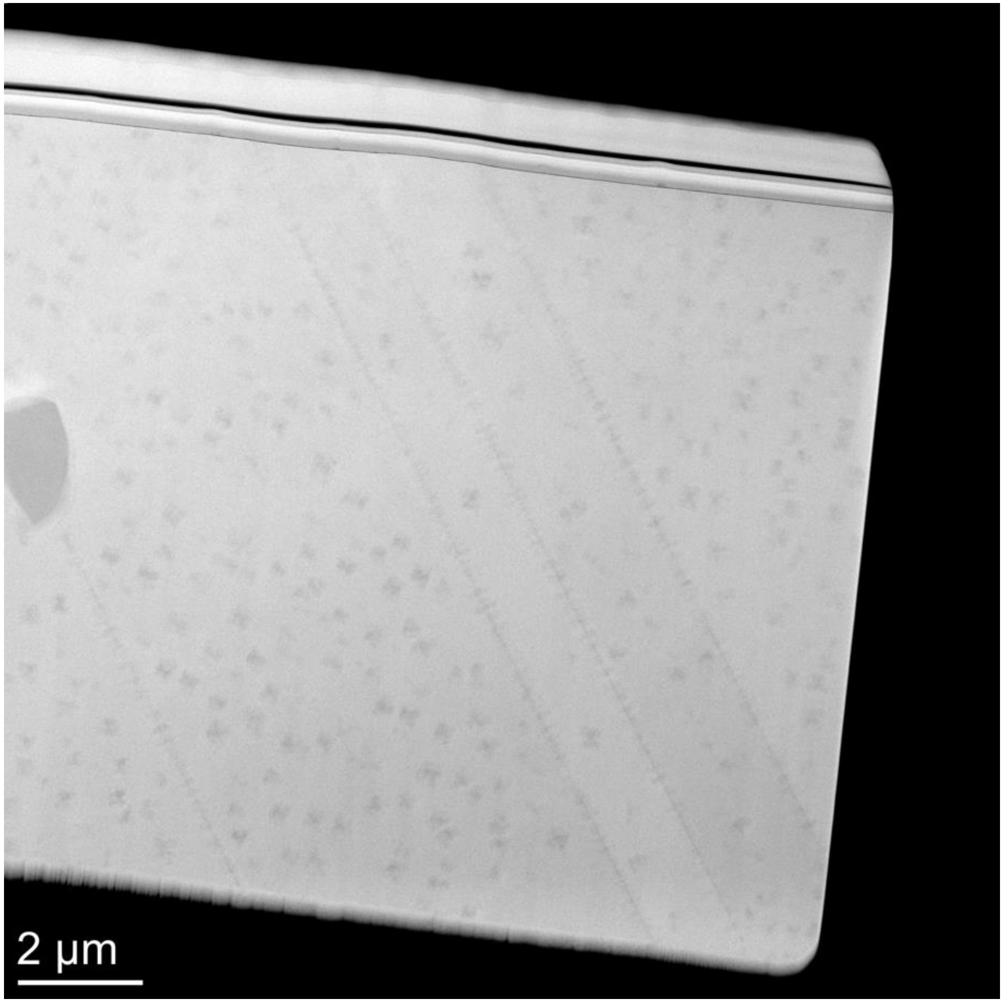
Slip parallel to Annealing twin boundary, similar to what was observed in in-situ tensile tested sample

Characterization of Crept Sample





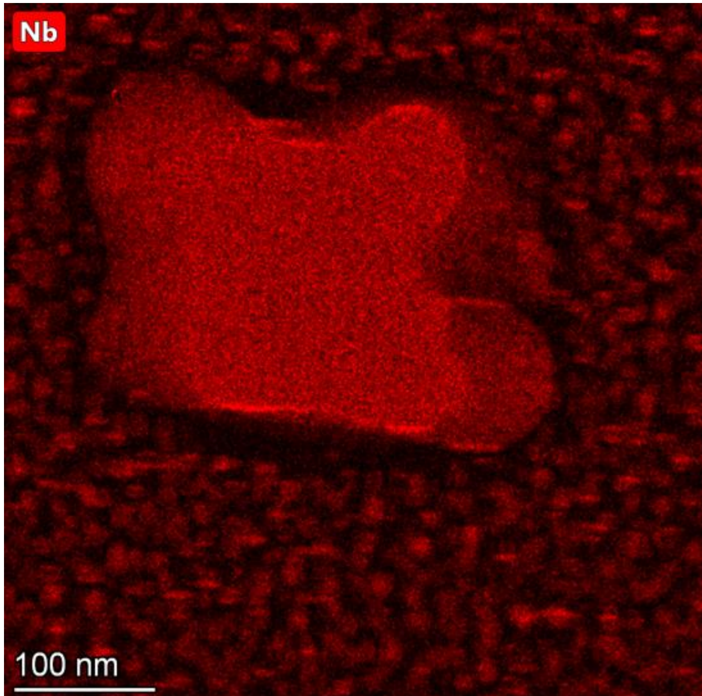
Characterization of Crept Sample



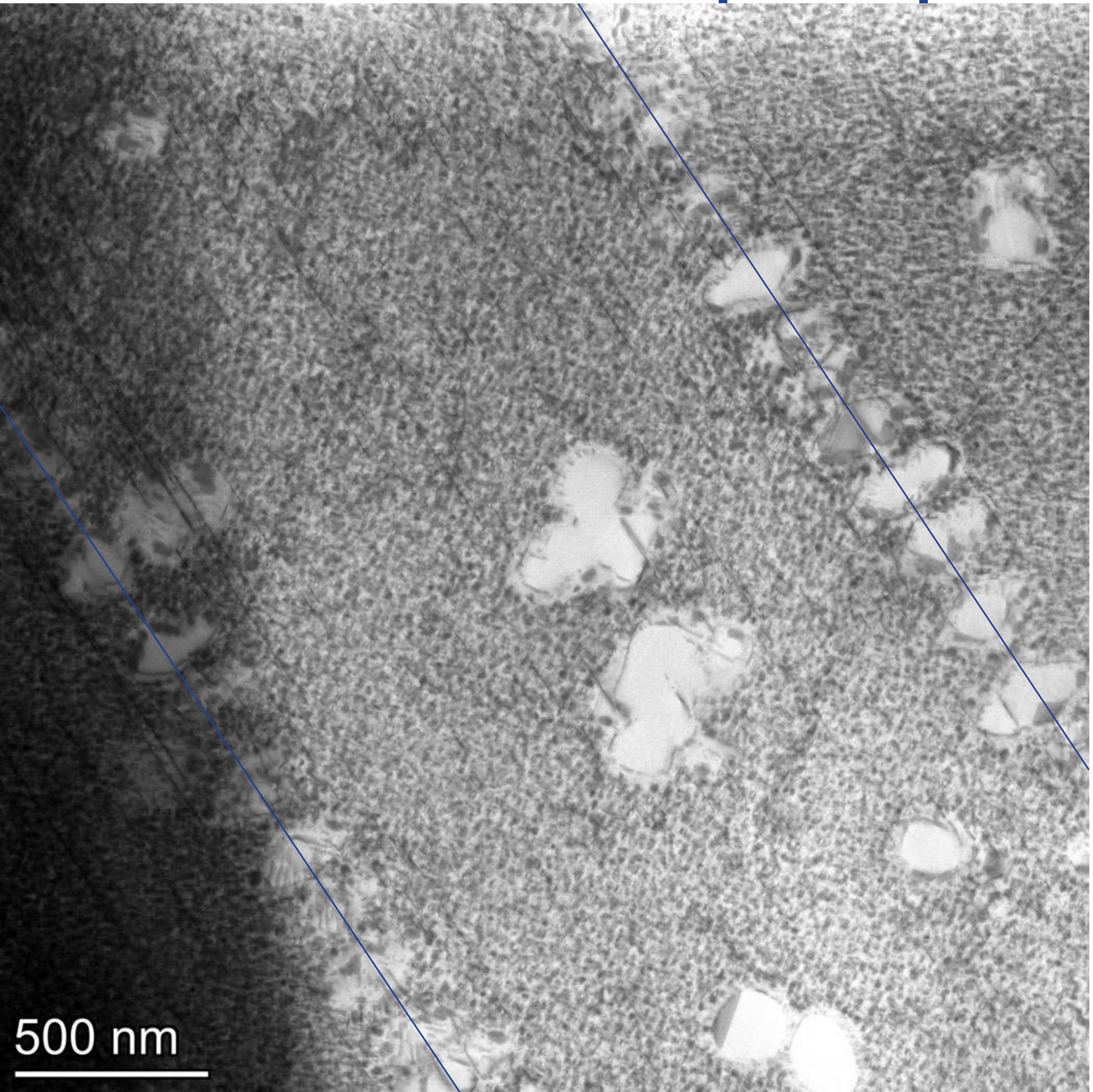
IN718-27 (HT-3)

Tension Creep at 1200°F (650°C) and 75 ksi (517 MPa)

This sample reached 0.5% strain at ~500 hours.

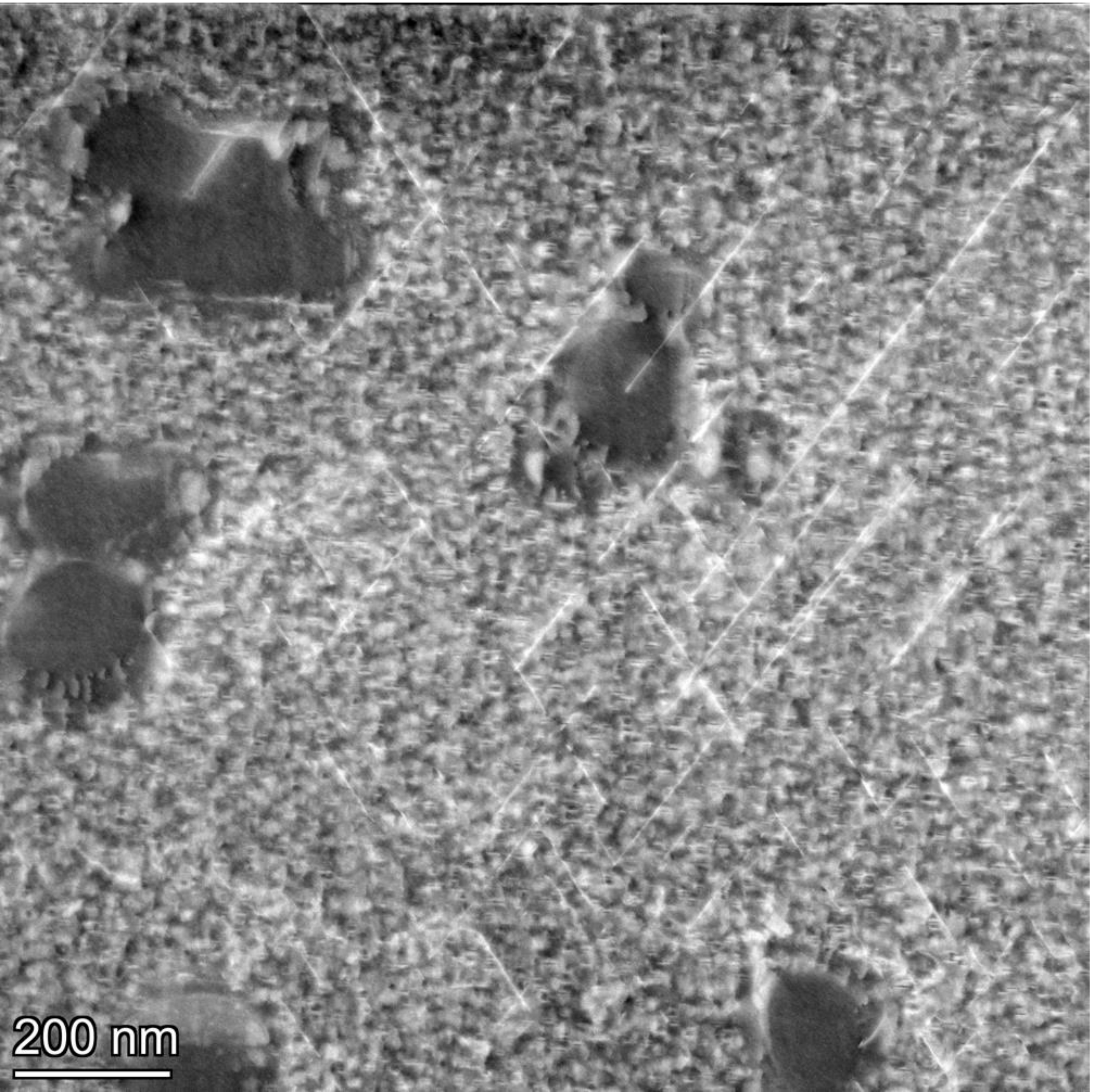


Characterization of Crept Sample

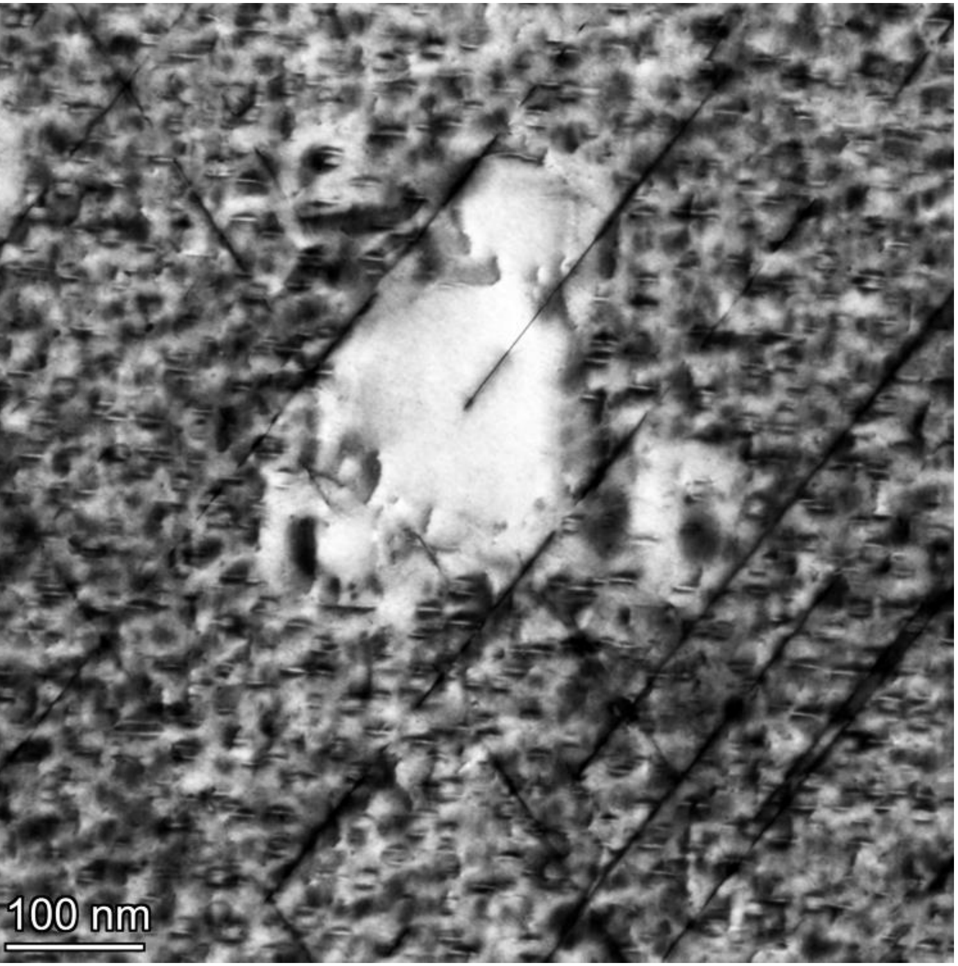


Faults parallel to annealing twin boundary

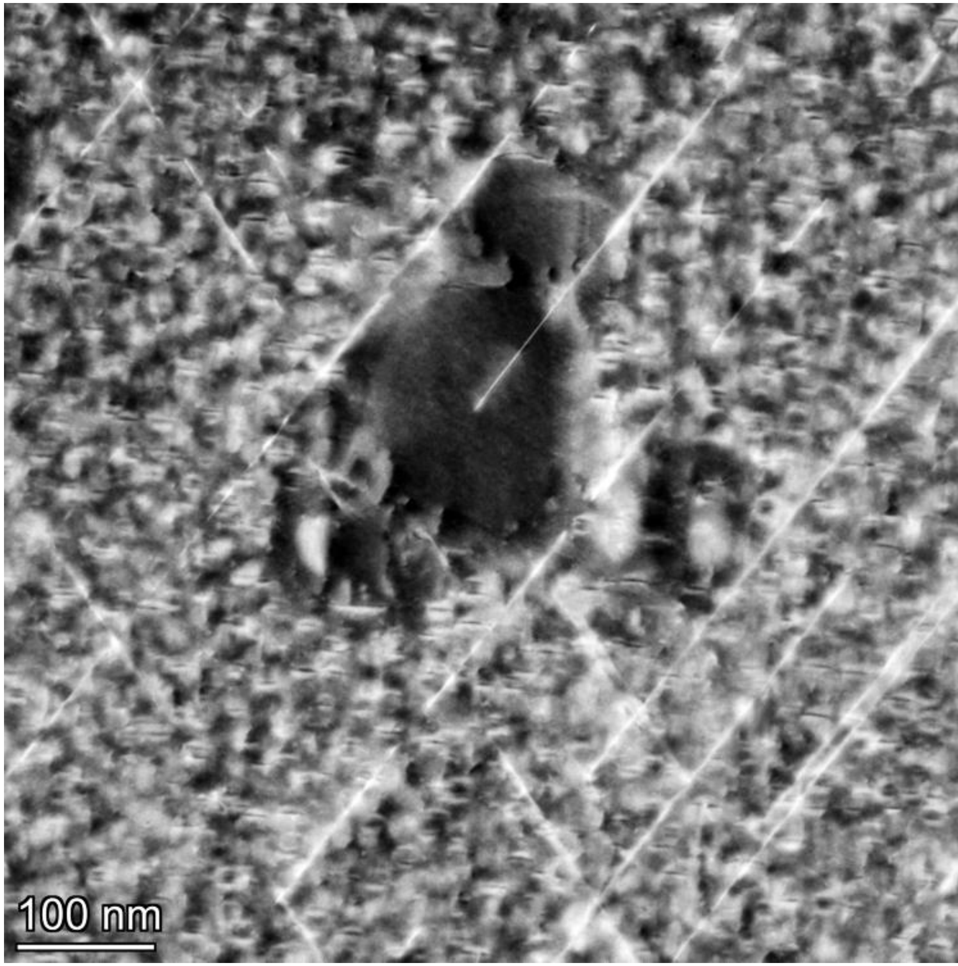
Characterization of Crept Sample



Characterization of Crept Sample



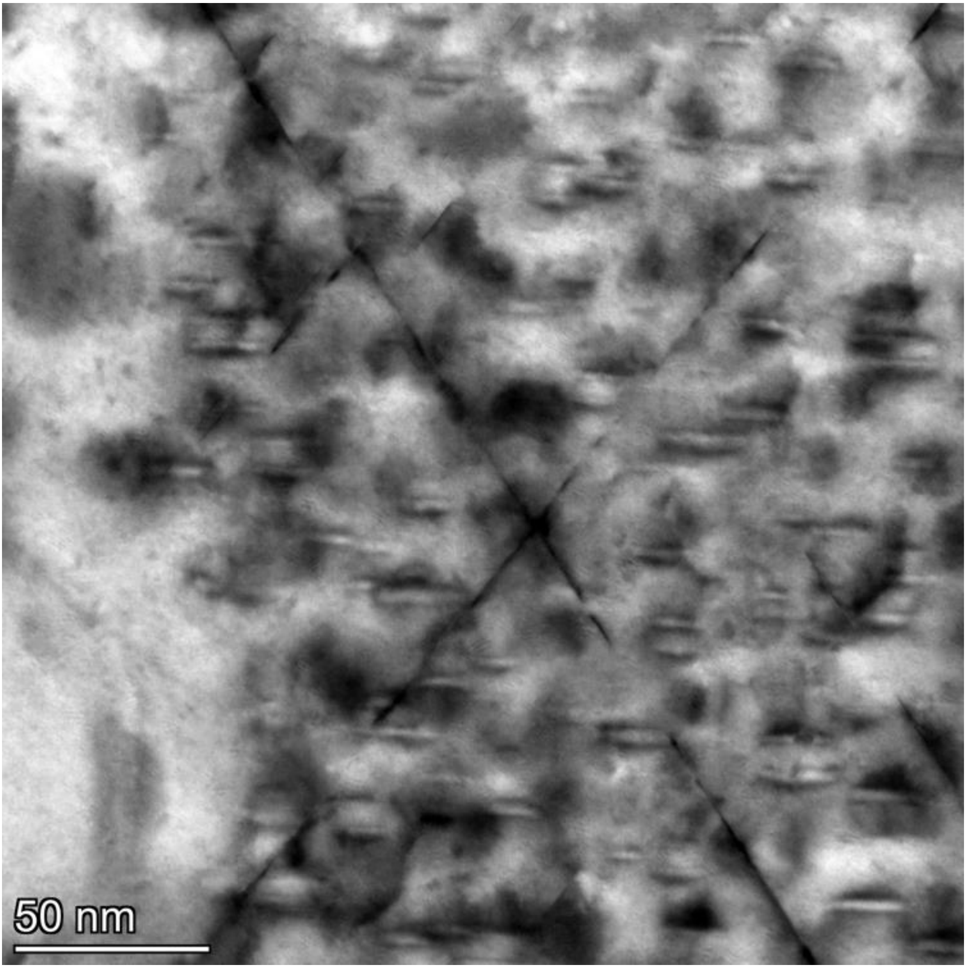
BF STEM



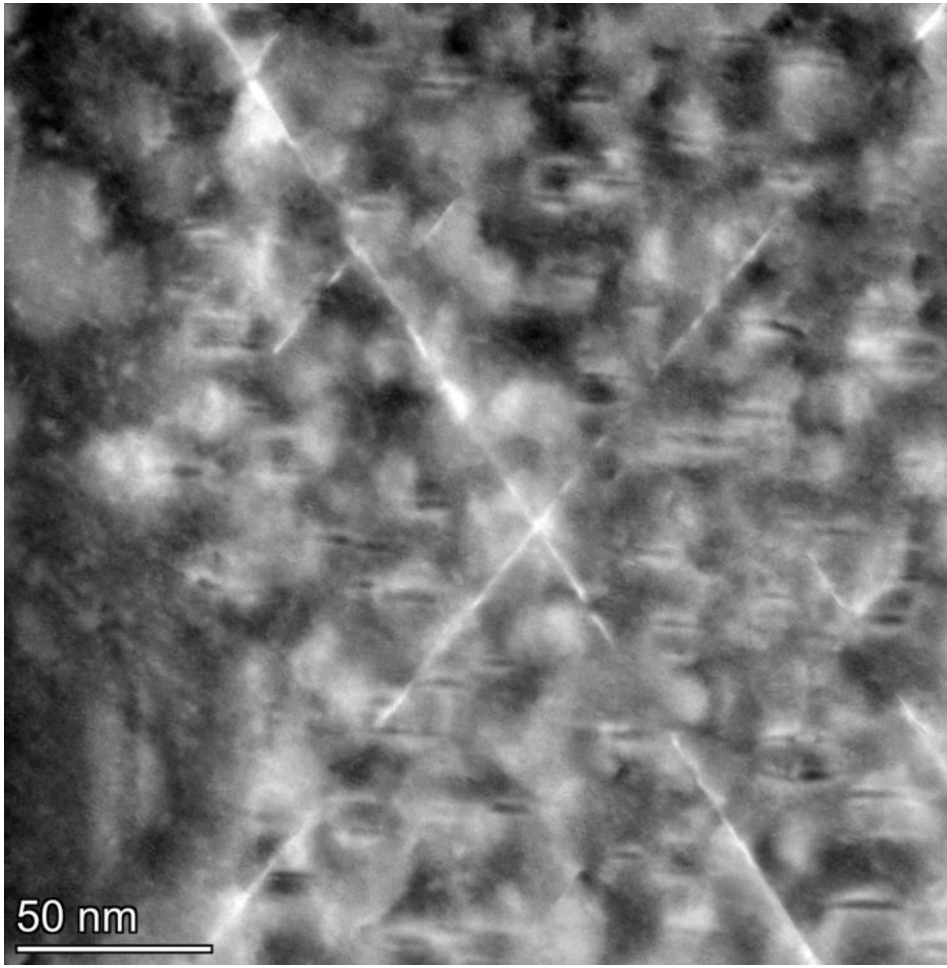
HAADF STEM

[110]

Characterization of Crept Sample



BF STEM

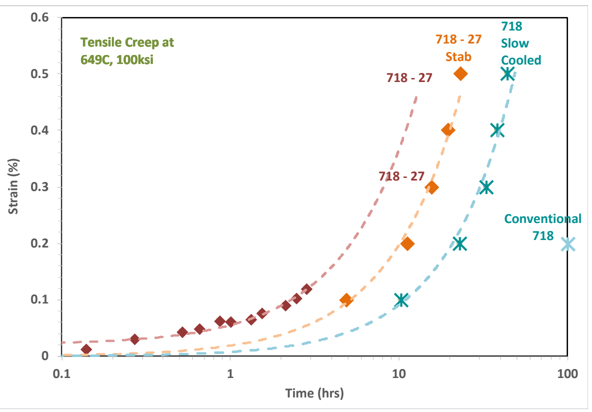


HAADF STEM

[110]

Round 1 and 2 Alloys: Key Issues

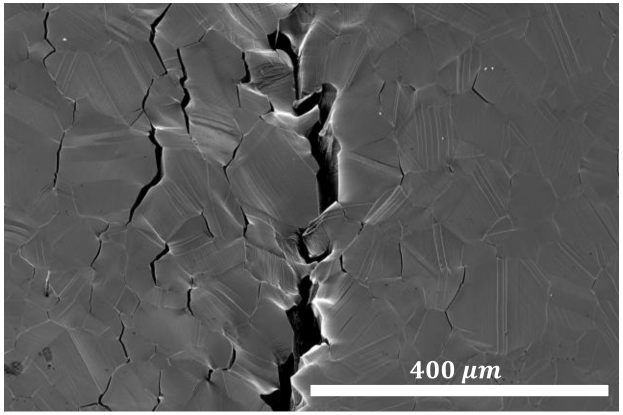
Tertiary Coprecipitates Absent



Difference in between **718-27** and **718-27 Stab** : Latter has tertiary coprecipitates

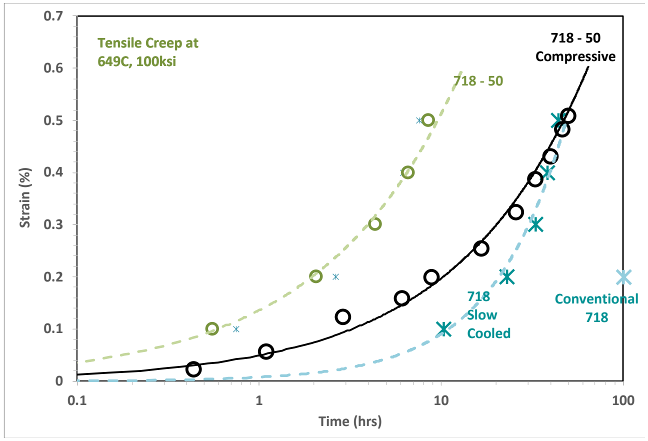
Round 2 alloy (IN718-50) does not have tertiaries

Poor **Intergranular** Strength



Round 2 alloy (IN718-50) shows intergranular failure at room temperature and high temperature

Poor **Intragranular** Strength



Even though IN718-50 performs better in compression than tension (due to grain boundaries), conventional IN718 outperforms both cases

Summary of Round 3 Alloys

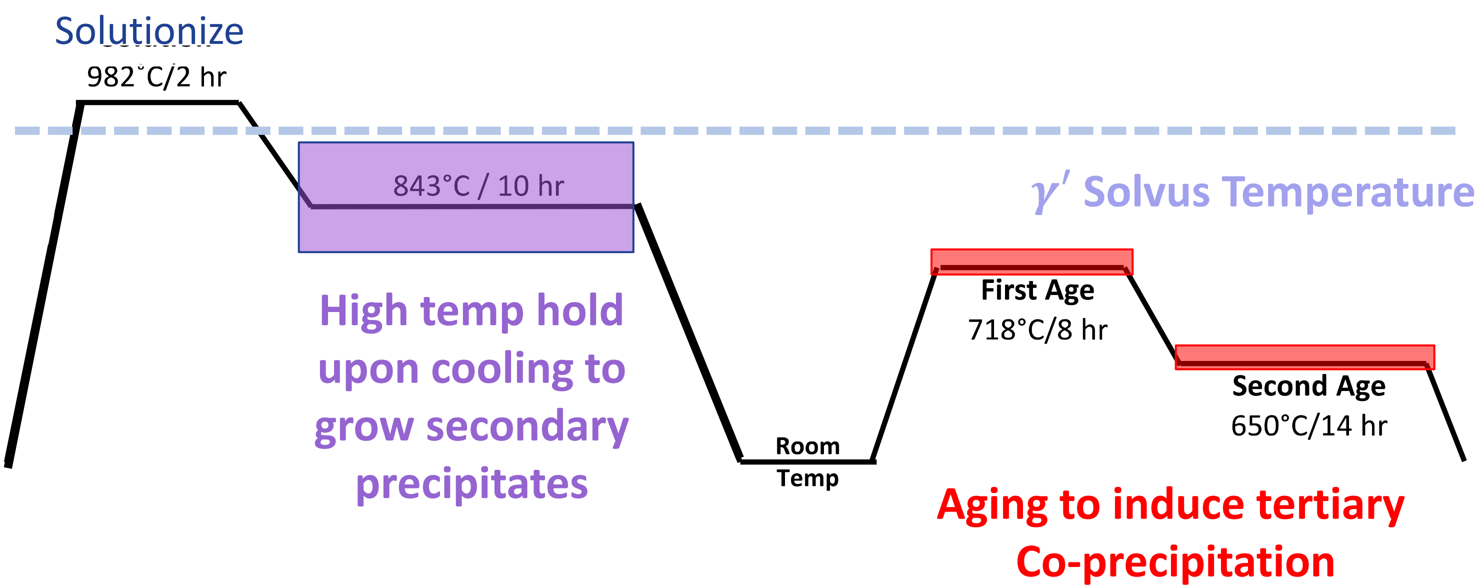
- Follow Three Distinct Design Strategies

Strategy 1

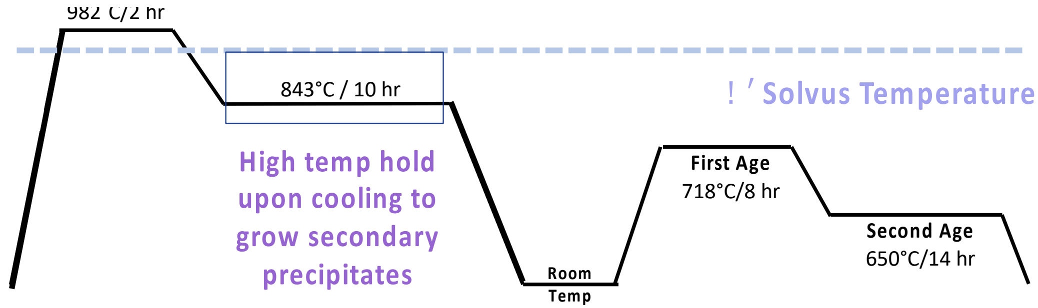
Problem:
No Tertiary Coprecipitates

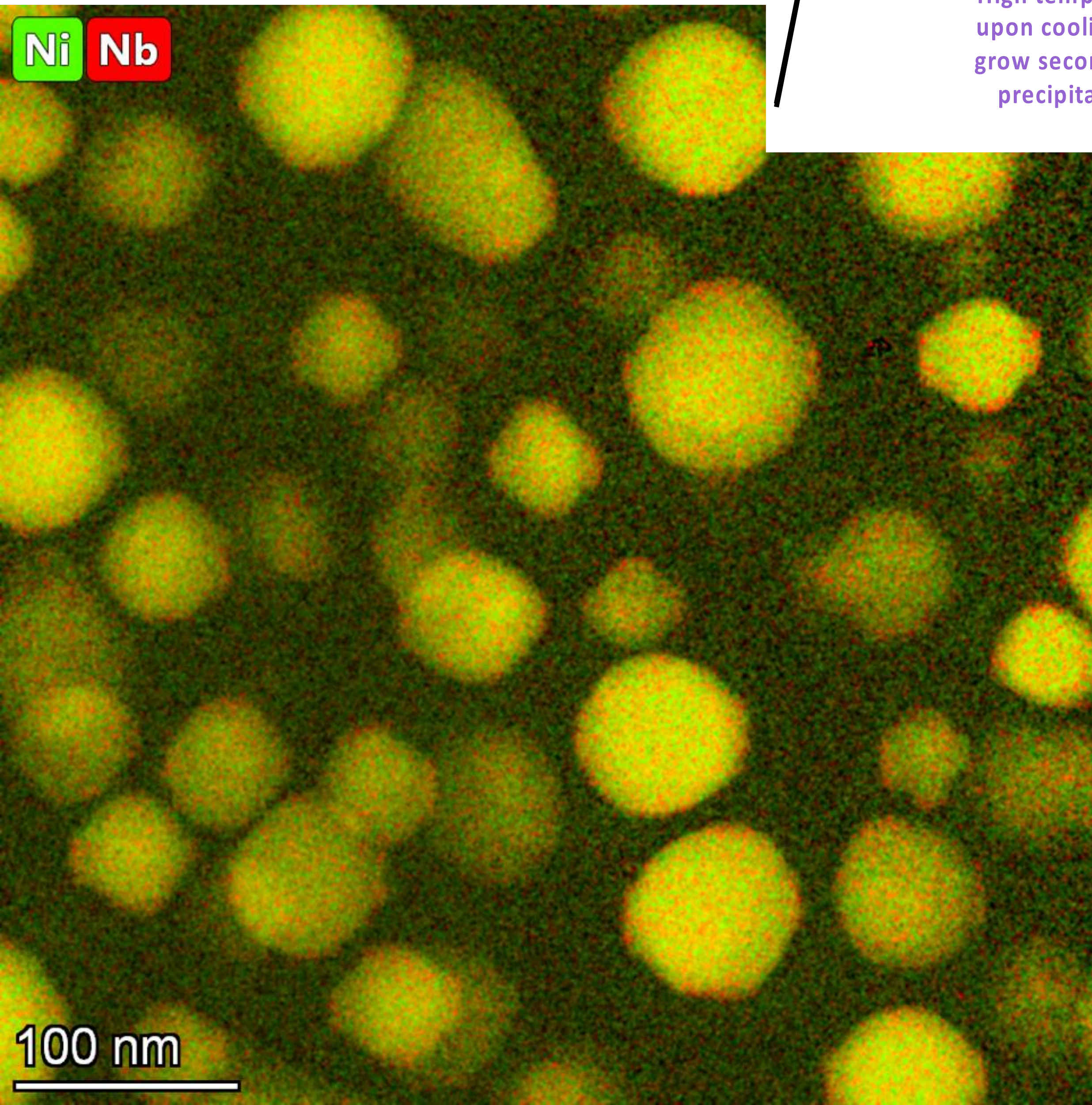
Alloy Name (wt%)	γ' solvus	Ni	Cr	Fe	Al	Ti	Nb	Mo	C
718		52.5	19	18.9	0.5	0.9	5.13	3.05	0.02
718-011	855.8	52.5	19	18.9	1.5	---	5	3.05	0.02
718-027	830.1	52.9	18.7	18.9	1.1	0.95	4.4	3.05	0.02
718-050	854.8	54.4	17.3	17.5	1.9	---	6.2	2.8	0.02
718-051	917	56.9	14.9	15.3	2.4	---	8	2.5	0.02

New heat treatment for alloy 50:



New heat treatment for alloy 50





No tertiary precipitation observed in IN718-50

Perhaps, highlights the importance of Ti in tertiary coprecipitate formation

Summary of Round 3 Alloys

- Follow Three Distinct Design Strategies

Problem:
Poor Intergranular Strength

Strategy 2

Strengthen grain boundaries
through C / B / Zr additions

Alloys produced and heat treatments by GER

***Characterize* microstructures after heat treatments and *downselect* for mechanical testing**

Alloy Name	Ni	Cr	Fe	Al	Ti	Nb	Mo	C	B	Zr
718	52.5	19.0	18.9	0.50	0.90	5.13	3.1	0.02		
718-027	52.9	18.7	18.9	1.07	0.95	4.42	3.1	0.02		
718-071	55.8	18.7	18.9	1.07	0.95	4.42	0.0	0.06	0.015	0.03
718-073	51.9	18.7	18.9	1.07	0.95	4.42	4.0	0.02	0.015	0.03
718-074	55.7	18.7	18.9	1.07	0.95	4.42	0.0	0.06	0.100	0.10
718-077	51.7	18.7	18.9	1.07	0.95	4.42	4.0	0.02	0.100	0.10
718-079	55.8	18.7	18.9	1.07	0.95	4.42	0.0	0.02	0.100	0.03
718-080	51.8	18.7	18.9	1.07	0.95	4.42	4.0	0.06	0.015	0.10
718-082	55.8	18.7	18.9	1.07	0.95	4.42	0.0	0.02	0.015	0.10

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Summary of Round 3 Alloys

- Follow Three Distinct Design Strategies

Problem:
Poor
Intragranular
Strength

Strategy 3:

Increase γ' volume fraction while maintaining co-precipitate potential

	Nominal Composition (wt%)									
Alloy Name	Ni	Co	Cr	Fe	Al	Ti	Ta	Nb	W	Mo	C
718-056	49.3		18.6	18.5	1.37		4.58	4.70		3.0	0.02
718-057	43.4		18.2	18.1	2.23		7.47	7.67		2.9	0.02
718-058	50.1		18.9	18.8	1.39	0.62	2.33	4.78		3.0	0.02

Alloys produced and heat treatments by GER

- Ta addition and balance Ta+Ti content
- Increase coprecipitate volume fraction while also maintaining the "golden ratio" of (Al+Ti+Ta)/(Nb) ~1.
- Ta a strong gamma prime stabilizer known to improve tensile strength of alloys (as in Ta718).

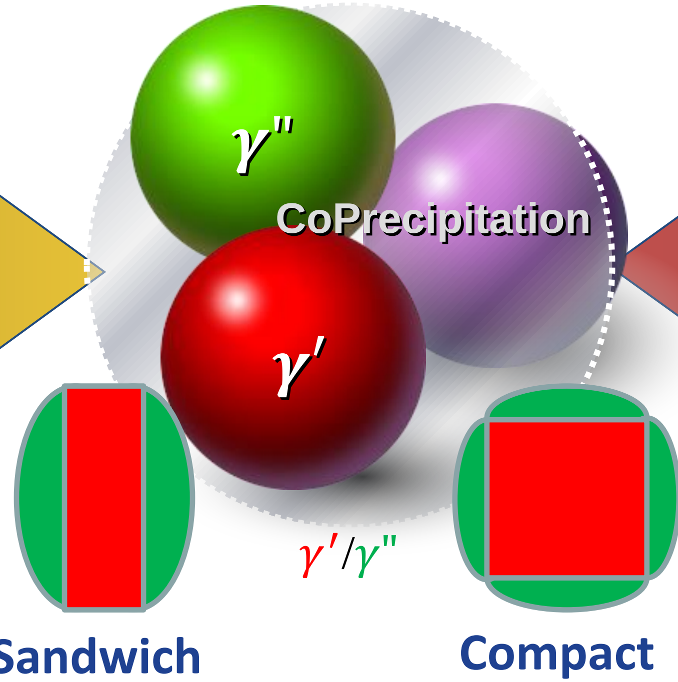
Coprecipitation Modelling Overview

Multi-Phase Field

Equilibrium Shapes
- Sandwich $\gamma''/\gamma'/\gamma''$
- Compact

Isothermal Growth
- Kinetics
- Diffusion Mechanism

Co-precipitation on continuous cooling



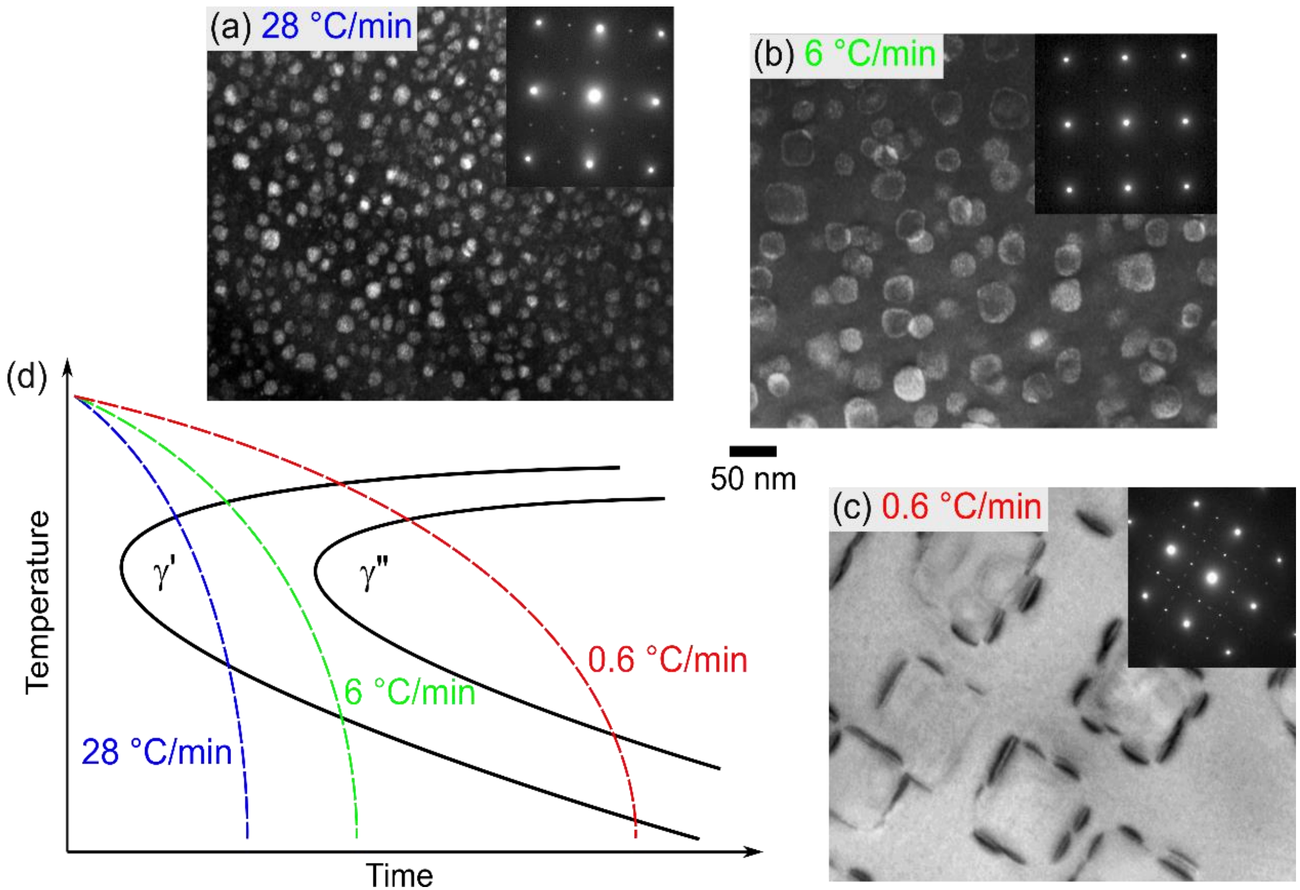
PanPrecipitation

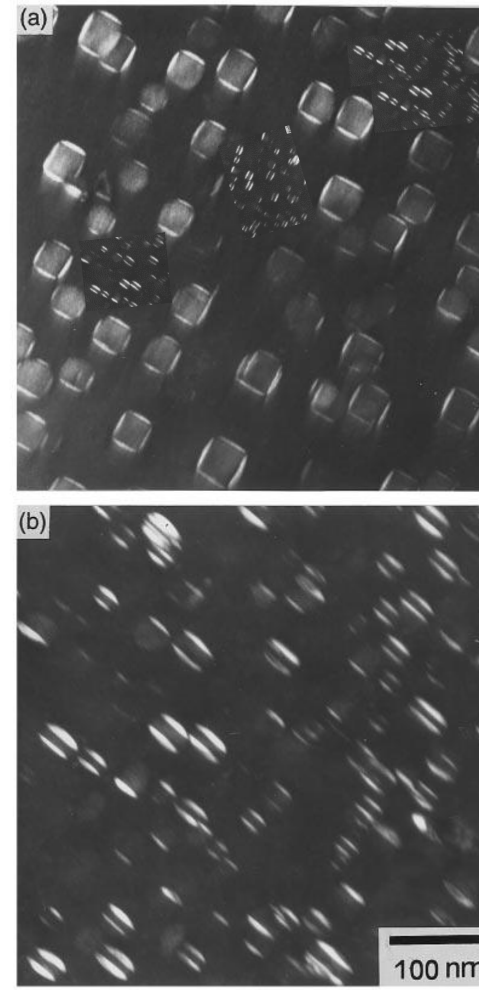
- **Slow cool kinetics**
  - Nucleation rates
  - Phase fractions
  - Average particle size

Combining CALPHAD and phase field modeling to develop quantitative prediction of the coprecipitation phenomena

Coprecipitation Concept

- Leverage the coprecipitation of γ′ and γ′′ to restrict γ′ coarsening during slow cooling of thick section components.





He, J., et al. Acta Mater 46 (1998): 215-223.

Desired: form γ′ then rapidly nucleate γ′′

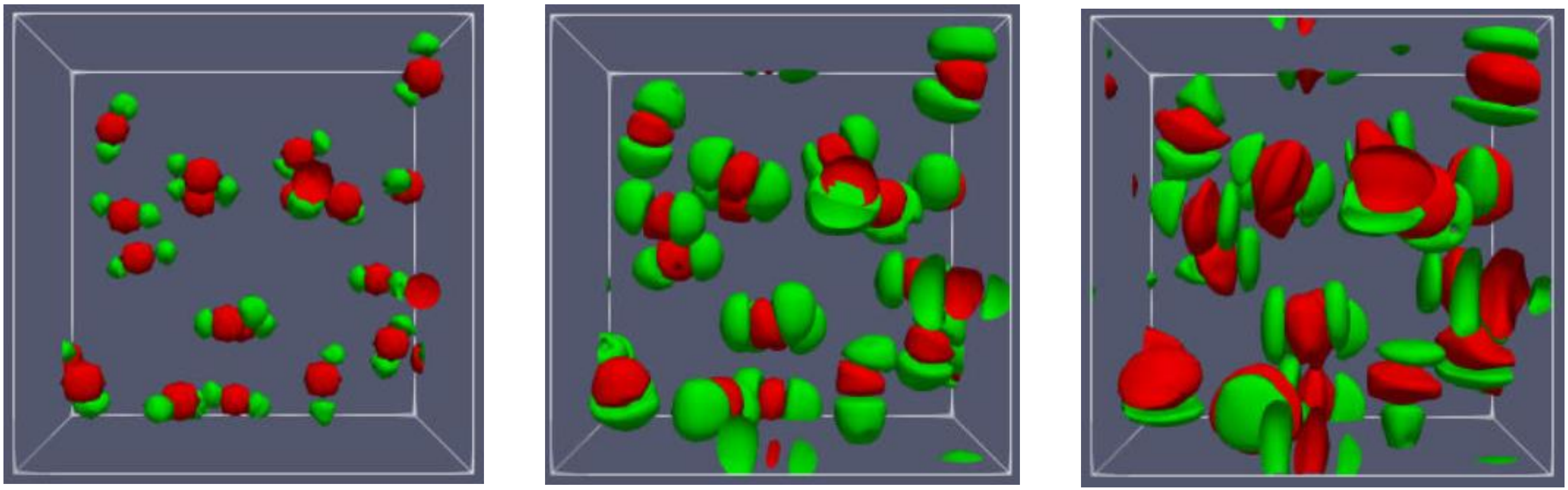
A. Detor, et al, Metall. Mater. Trans. (2017)

Updates on Modeling Efforts

- **Phase field modeling to understand coprecipitation mechanisms**
  - We are exploring the roles played by both chemical and mechanical effects, e.g., the concentration and stress fields created by precipitation of γ' phase, on precipitation of γ'' precipitates and the development of compact coprecipitates.
- **Design of Round-3 alloys and optimization of heat treatment**
  - High throughput calculations have been performed to aid design of Round-3 alloy. These calculations have helped in establishing trends of volume fractions of γ' and γ'' as function of Al and Nb contents observed. Effect of Ta in IN 718 system has also been explored.
- **Development and calibration of creep model**
  - A fast-acting creep model considering dislocation and diffusion creep has been developed and calibrated with creep data from Round 2 alloys.

Phase Field Modeling Microstructural Evolution of Coprecipitates

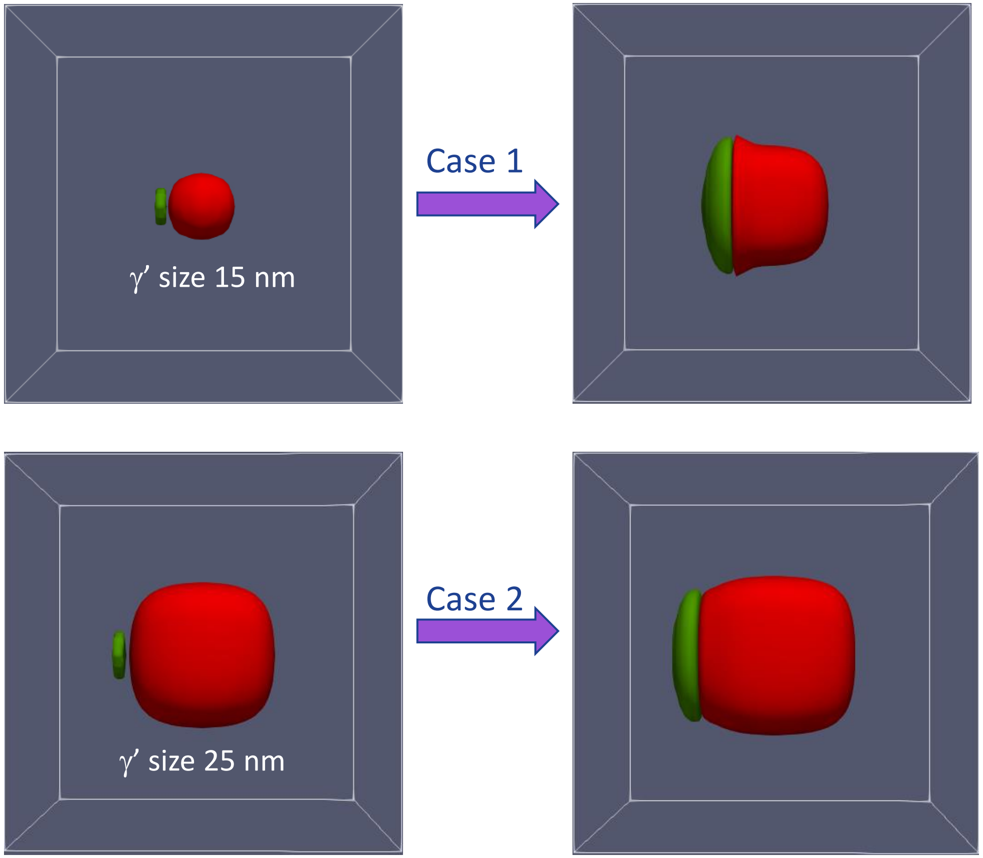
T=790°C

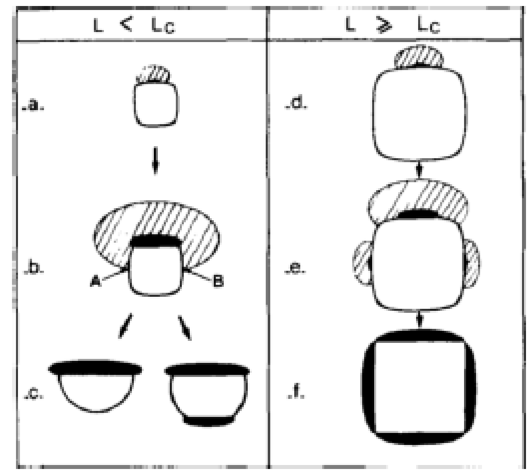


a)t=0sec b)t=2.25sec c)t=15sec

- Using pseudo-ternary database to represent multicomponent system
- For each new alloys, a pseudo-ternary database need to be developed
- Semi-quantitative in nature and diffusion in a ternary system may not be able to represent diffusion in multi-component system

Coprecipitate Formation

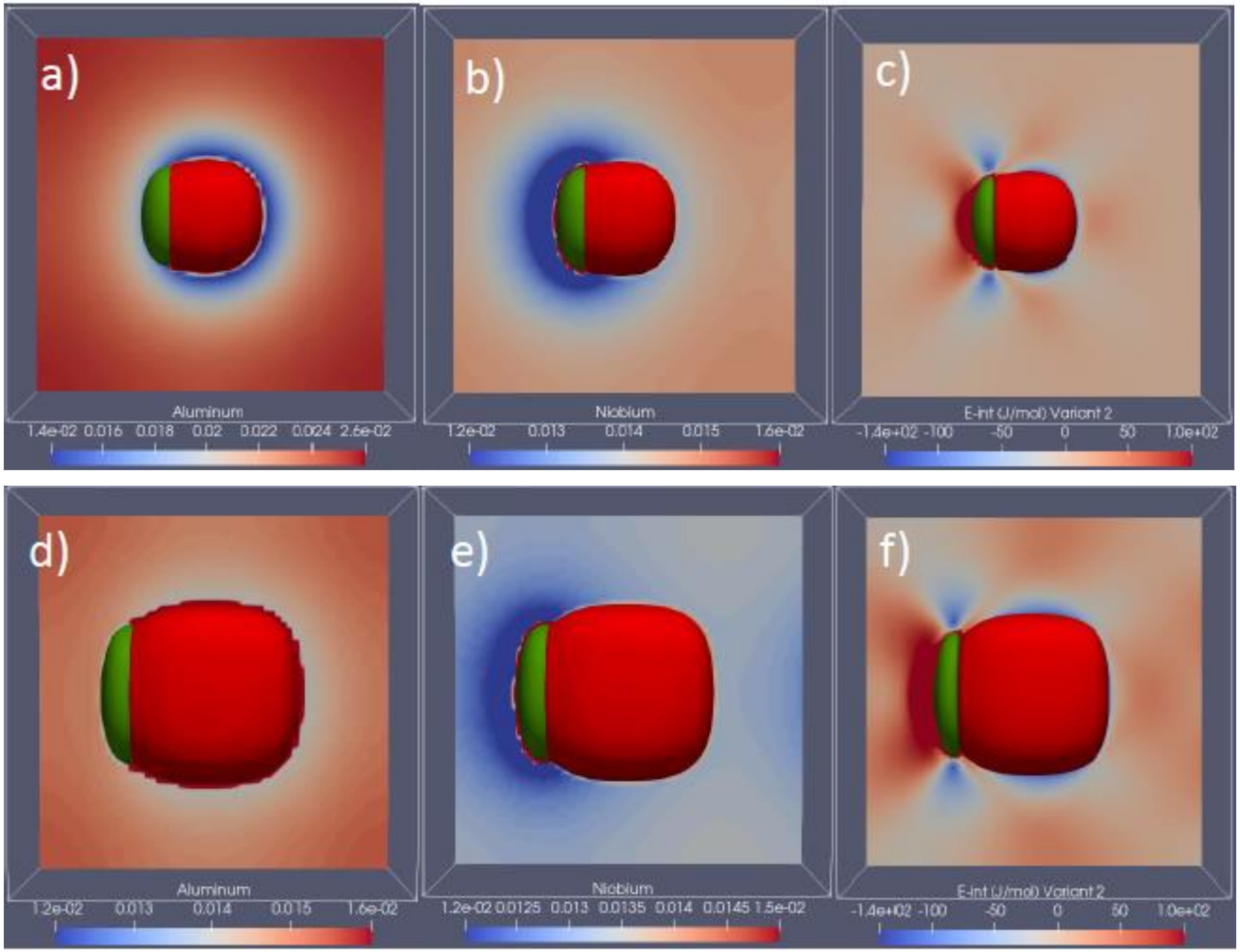




[1] . Cozar, R., Pineau, A. Metall Mater Trans B 4, 47–59 (1973).

- Influence of nucleation of γ'' on a facet of γ' of two different sizes is investigated.
- Nucleation of γ'' on a smaller γ' influences the shape evolution of γ' significantly.

Efforts of Concentration and Stress fields



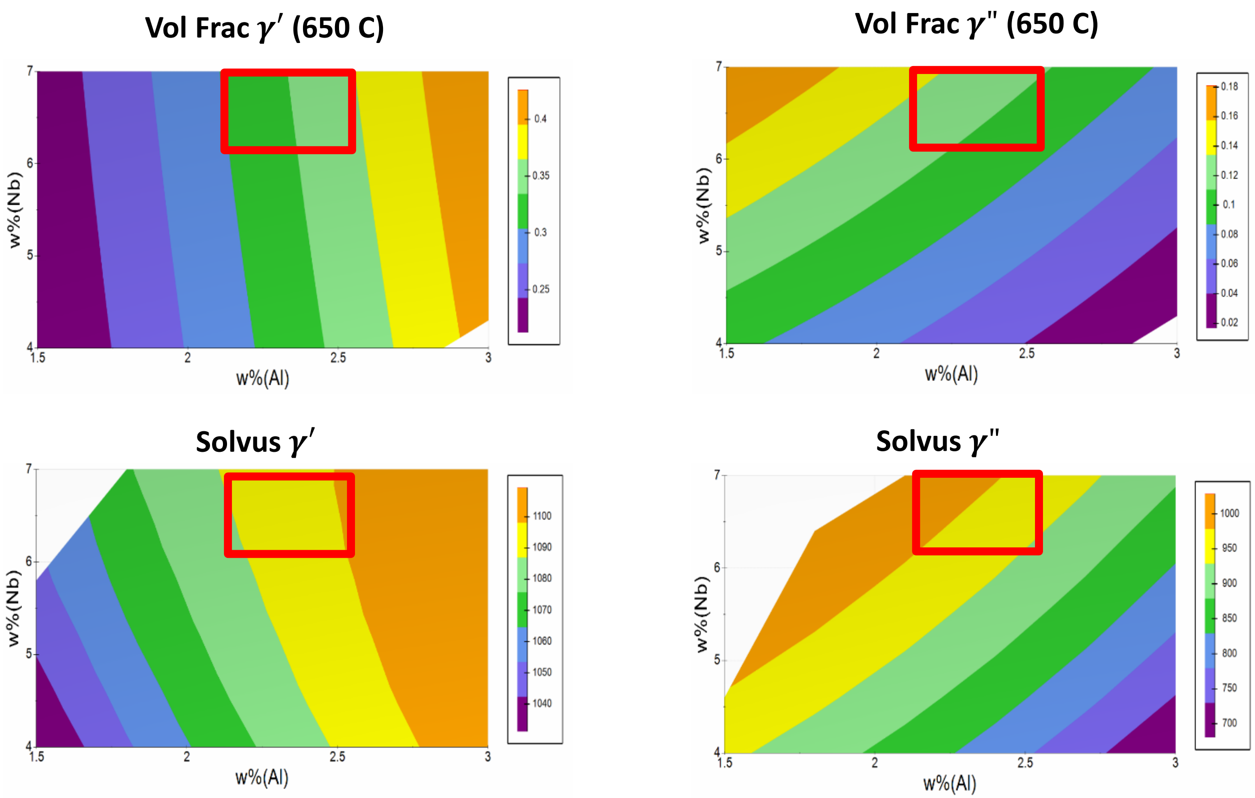
Al concentration (at%) Nb concentration (at%) Elastic interaction energy (J/mol)

- Nucleation of γ'' on a small γ' results in a large Nb depletion zone (dark blue region in b)). Nucleation of γ'' on adjacent faces is unlikely because of solute depletion, leading to sandwich coprecipitate.

- Nucleation of γ'' on a larger γ' does not deplete Nb solute on adjacent faces and the larger negative elastic Interaction energy observed on adjacent faces will assist the formation of compact coprecipitate.

Calculations to quantify the critical γ' size for the formation of compact vs. sandwich coprecipitates are underway.

High throughput calculations to aid Round-3 alloy design



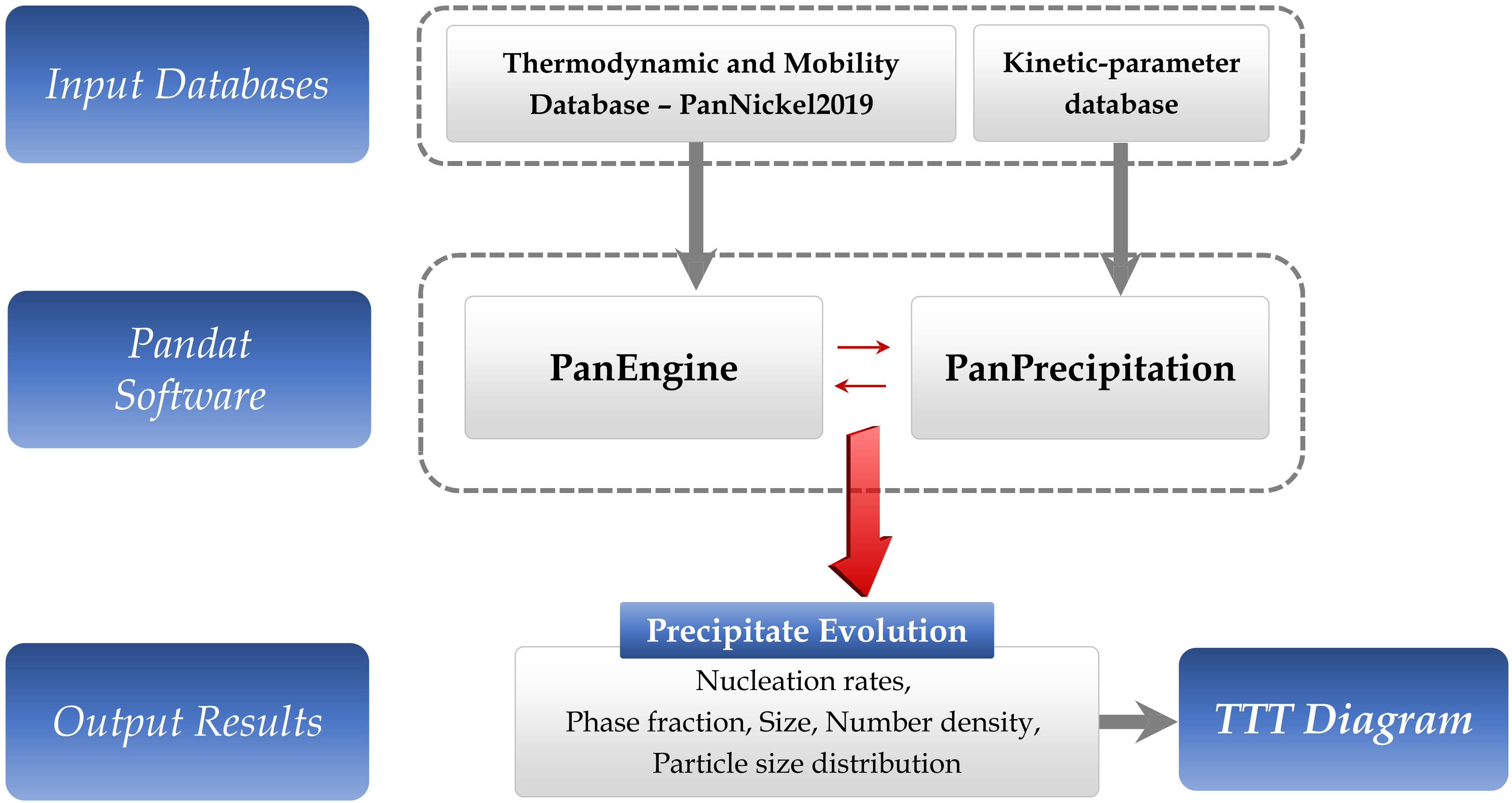
Elements	Round 3 Alloy (Wt%)
Ni	Balance
Cr	17.3
Fe	9.5
Mo	2.8
Al	1.5-3
Nb	4-7
Ta	4
Co	4

High throughput calculations using full database from CompuTherm are carried out to aid in design of Round 3 alloy. Concentrations of Al and Nb are varied from 1.5 to 3 and 4-7 (wt%), respectively

Design considerations:

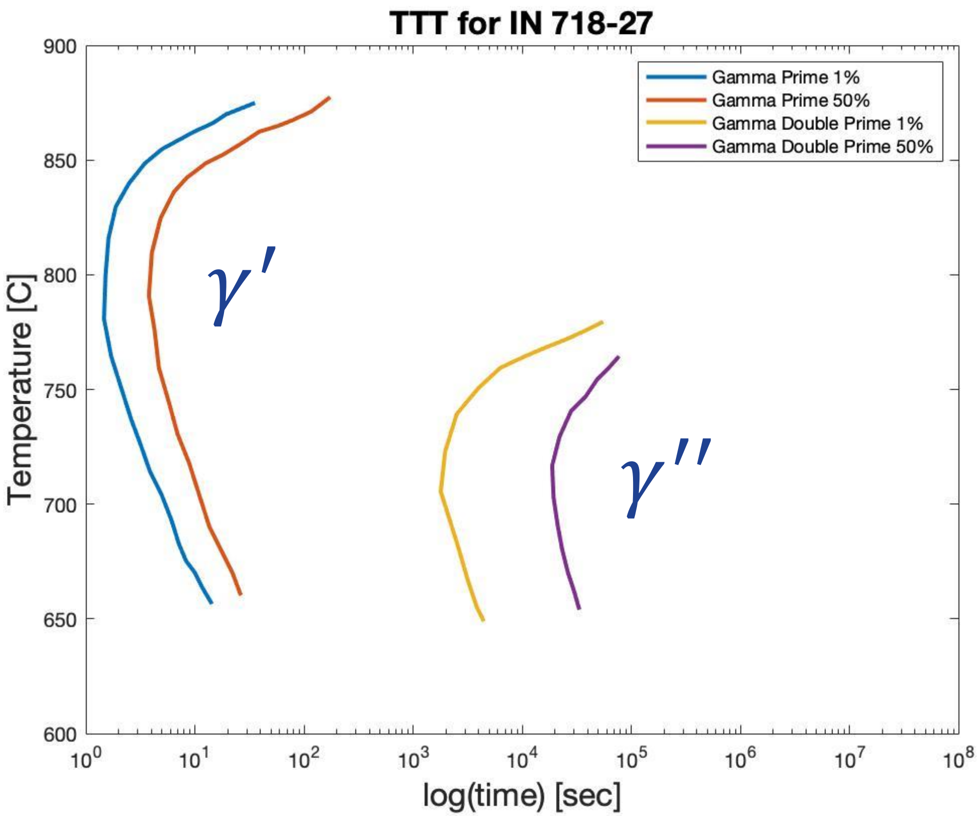
- High γ' volume fraction and satisfying stability ratio ($f_{\gamma'}/f_{\gamma''}$) predicted by phase field study
- Small difference in solvus temperatures of γ' and γ" phases to reduce growth and coarsening of γ' under slow cooling before γ" precipitates out
- Follow (Ti+Ta+Al)/Nb ratio to form coprecipitates
- Red Box indicates potential Rond-3 alloy compositions

TTT Diagram Calculation – PanPrecipitation



TTT Diagram Calculation – IN 718-27

- KWN (Kampmann-Wagner Numerical) model for precipitation
- Homogenous Nucleation of both γ' and γ'' for now
- Expanding to heterogeneous nucleation of γ''on γ' is underway



- Experimentally, compact co-precipitates were observed in IN 718-27
- Separation between γ' and γ'' TTT curves allow γ' core to grow to certain sizes and then be “coated” by γ'' precipitates

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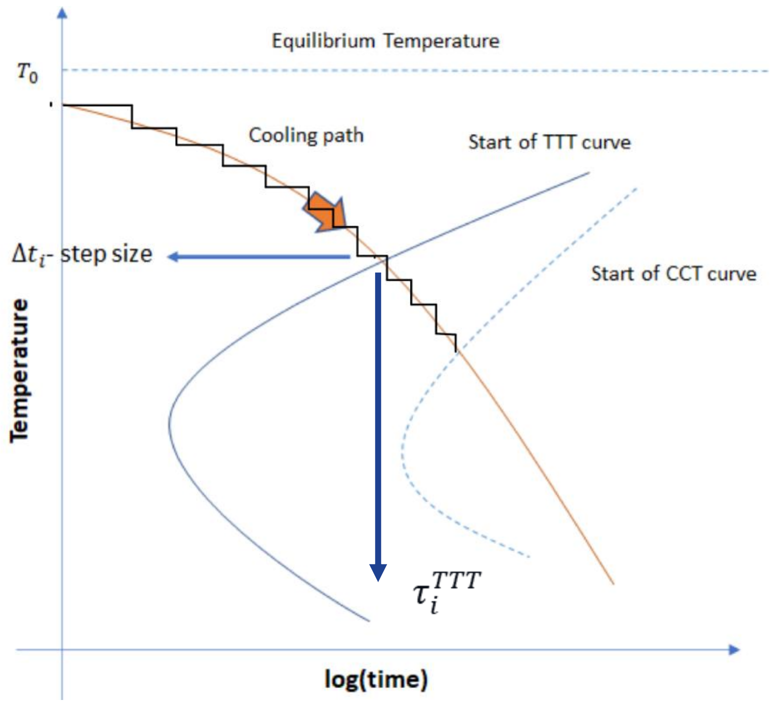
Calculating CCT diagram from TTT Diagram

Schematic for Rule of Additivity



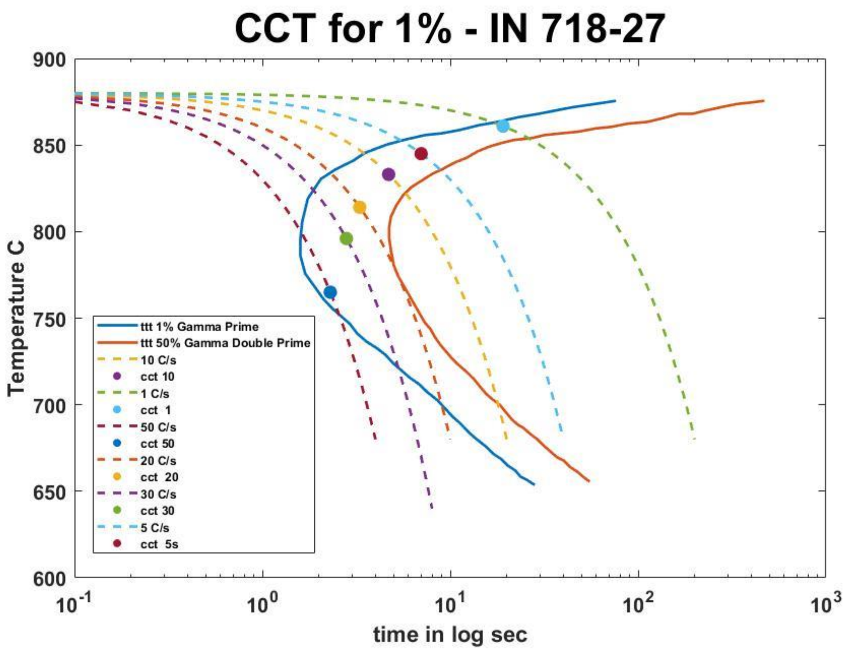
- The rule of additivity discretizes the continuous cooling curve into steps.
- For each step, time taken to reach transformation of x_0%(say 1%) transformation at that step's temperature is obtained from the TTT curve and the time spent on that step is found from the discretization.
- The fractions obtained from each step are added until the sum is unity.
- The temperature and time on the cooling curve where the sum of fractions is unity is the calculated CCT point for that cooling curve.

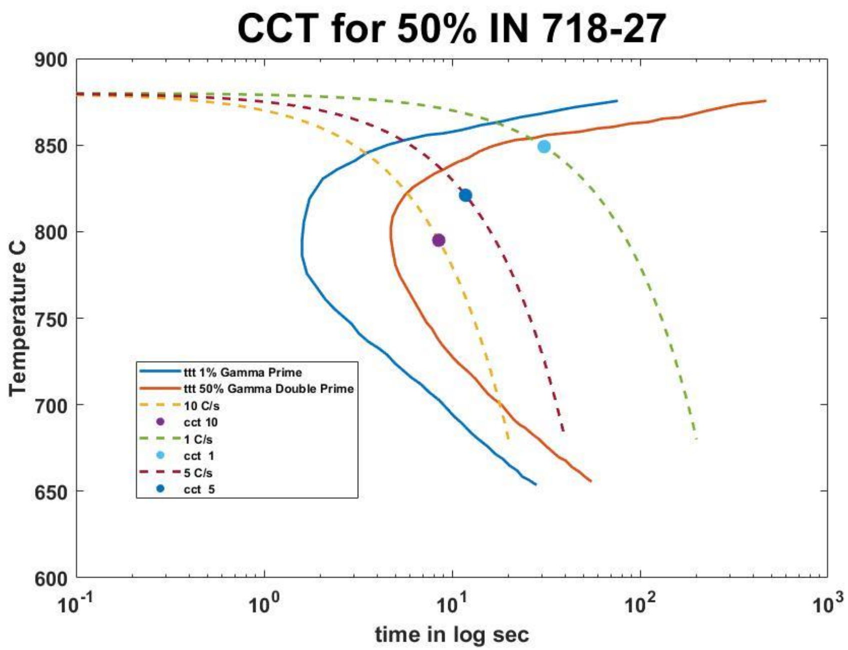
One can use the rule of additivity [1] is used to convert TTT curves into CCT curves: $\sum_{T_0}^{T_{CCT}} \frac{\Delta t_i}{\tau_i^{TTT}} = 1$

τ_i^{TTT} *—Time taken at temperature* $\mathrm{T_i}$ *to reach* $\mathrm{x_0}\%$ *transformation under isothermal condition*

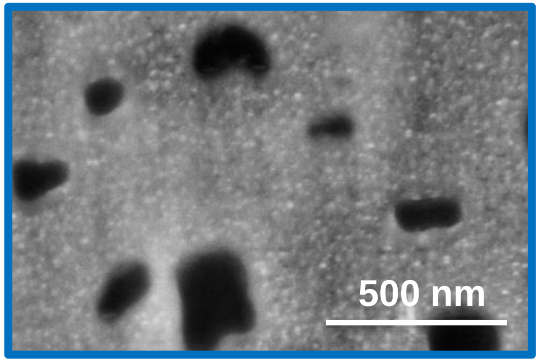
[1] *J.S. Kirkaldy, R.C. Sharma, Scripta Metallurgica,Vol 16, Issue 10,1982*

TTT to CCT conversion – IN 718-27

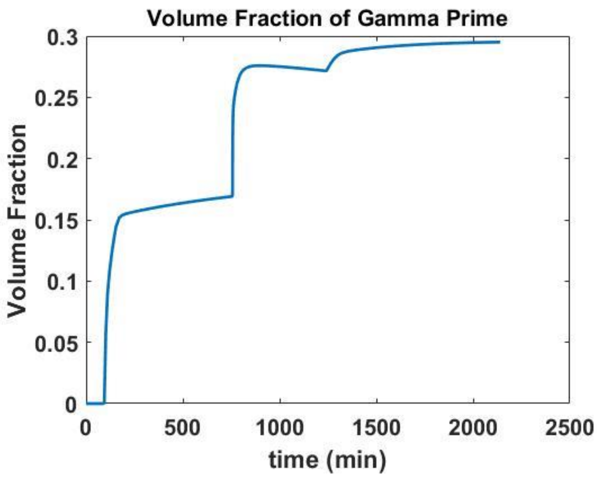


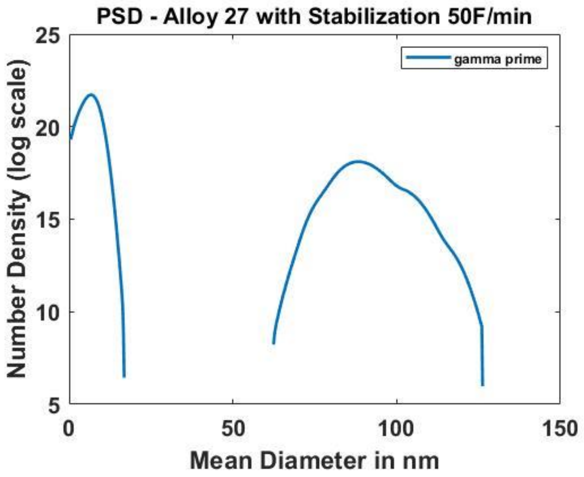


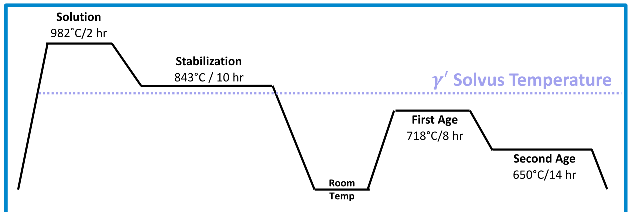
Precipitation Calculation – IN 718-27



Bimodal microstructure
Secondary coprecipitates: 150-200 nm
Tertiary "hamburger" γ'/γ'': 10-20 nm
Secondary area fraction: 10.4%
Tertiary area fraction: 17.2%



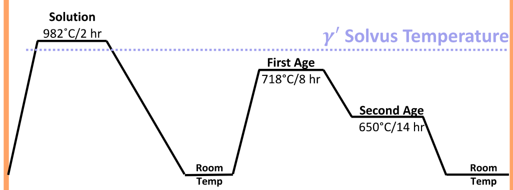




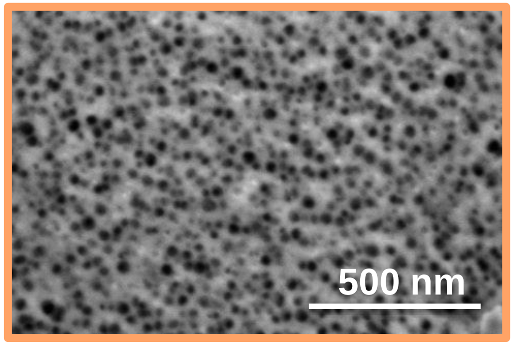
Age 3: Stabilization followed by Two Step Age

- Bimodal distribution of γ' is predicted by the simulation.
- The sizes predicted for the secondary and tertiary precipitates are reasonable
- The calibrated PanPrecipitation Module is being used to optimize heat treatment schedules for Round-2 alloys and design new heat treatment schedules for Round 3 alloys

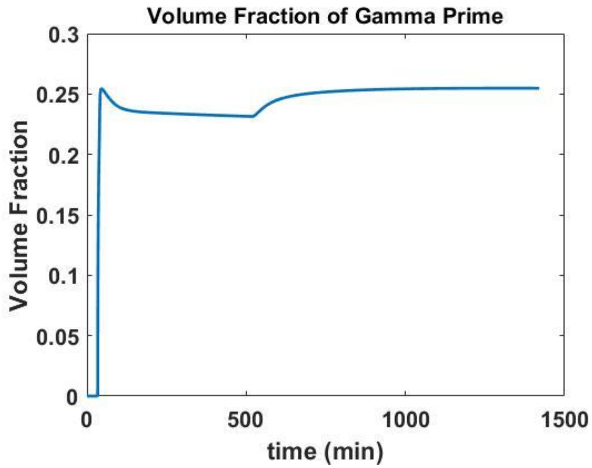
Precipitation Calculation – IN 718-50

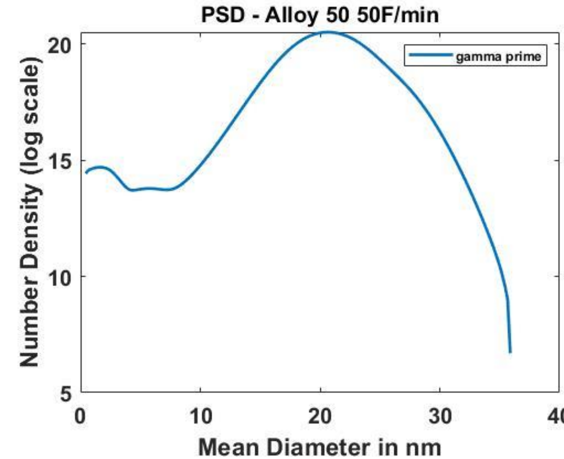


Age 2: Two Step Age

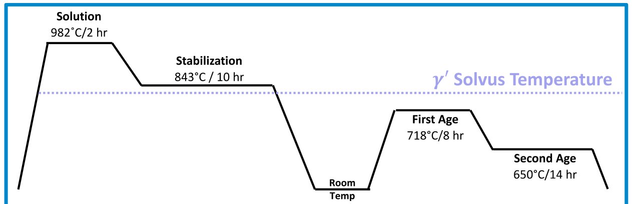


Monomodal microstructure
γ' Size: 30 ± 10 nm
Area Fraction: 27%

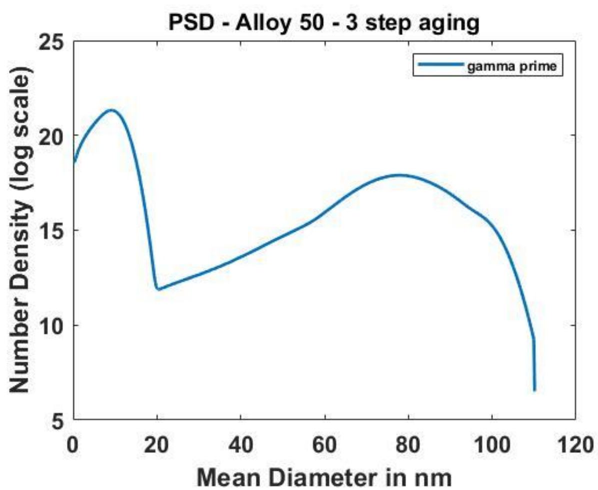




- Unimodal distribution of γ' is predicted by the simulation.
- The particle size distribution of γ' is accurate.



Age 3: Stabilization followed by Two Step Age



Bi-modal distribution obtained utilizing a 3-step sub solvus heat treatment on Alloy 50

Volume fraction predicted: secondary γ' precipitates ~ 11 %, tertiary ~ 18 %

MPF simulations with direct coupling with PanDat

- This approach has been tested by simulating phase separation in a four-component system (Ti-Zr-Nb-Ta).
- Compared to classical approach (MPFM↔PE), our new approach (MPFM ↔ PDN↔PE) is **55 times faster**.

Multi-Phase-Field Model (MPFM)

$$\frac{\partial \phi_\alpha}{\partial t} = -\sum_{\beta=1}^{\tilde{N}} \frac{L_{\alpha\beta}}{\tilde{N}} \left(\frac{\delta F}{\delta \phi_\alpha} - \frac{\delta F}{\delta \phi_\beta} \right)$$

$$\frac{\partial c_i}{\partial t} = \nabla \left(\sum_{j=1}^{m-1} M^{ij} \, \nabla \left(\frac{\delta F}{\delta c_j} \right) \right)$$

↕

PanDataNet (PDN) : Data Management system

Input: $\{c_j\}, \{\phi_\alpha\}$

Output:

$$\frac{\partial f^{chem}}{\partial c_j}, \frac{\partial f^{chem}}{\partial \phi_\alpha}, M^{ij}$$

- Uses interpolation on previous calculations to estimate the data.
- New calculation is performed only when necessary.

↕

PanEngine (PE)

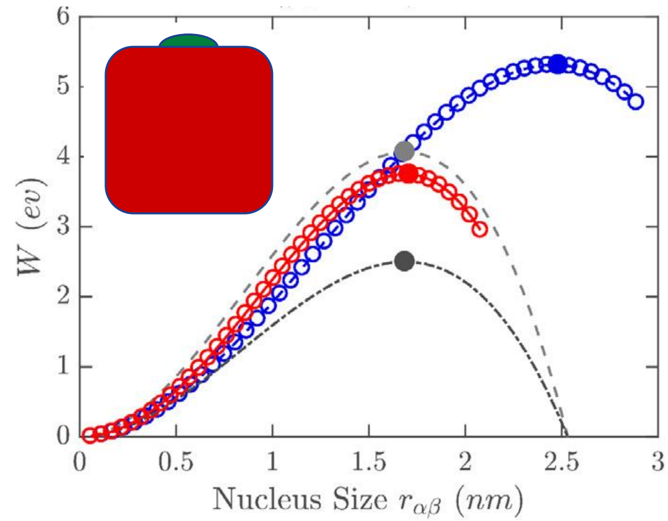
- Reads thermodynamic and kinetic database (*.pdb)
- Performs equilibrium and quasi-equilibrium calculation.

- This PanDat + MPF simulations can be used to benchmark the TTT/CCT diagram obtained from PanPrecipitation that considers heterogeneous nucleation

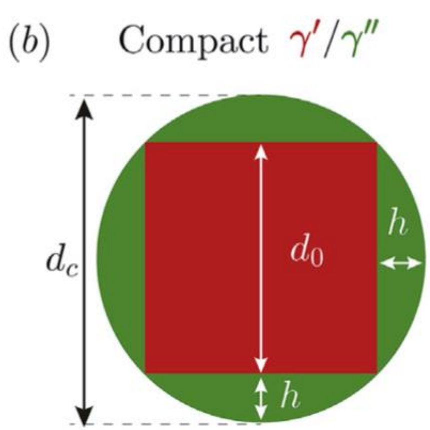
Linking Phase field to PanPrecipitation

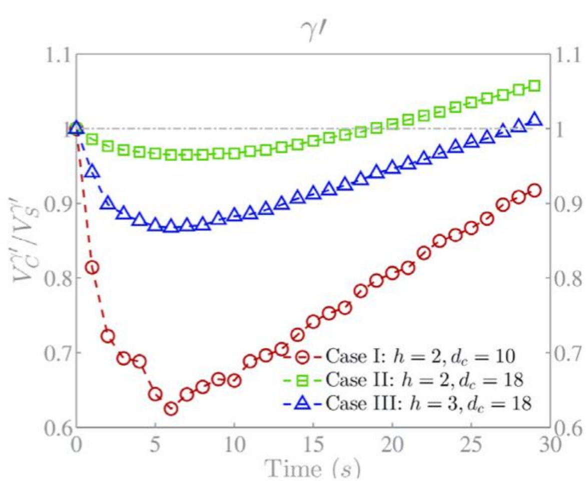
- Phase field model + NEB will be used to obtain activation energy barrier and critical nucleus size for heterogeneous nucleation of γ” on γ’.
- The phase field model will be used to compare growth rates between monolith precipitates and that of coprecipitates.
- The nucleation and growth rate equations in PanPrecipitation will be informed these phase field simulation results.

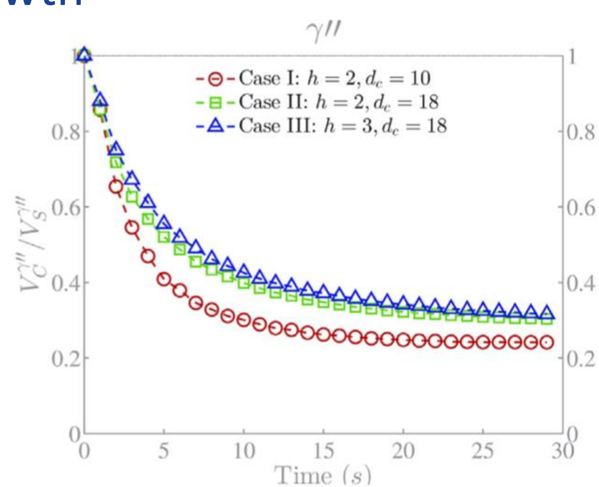
Nucleation



Growth







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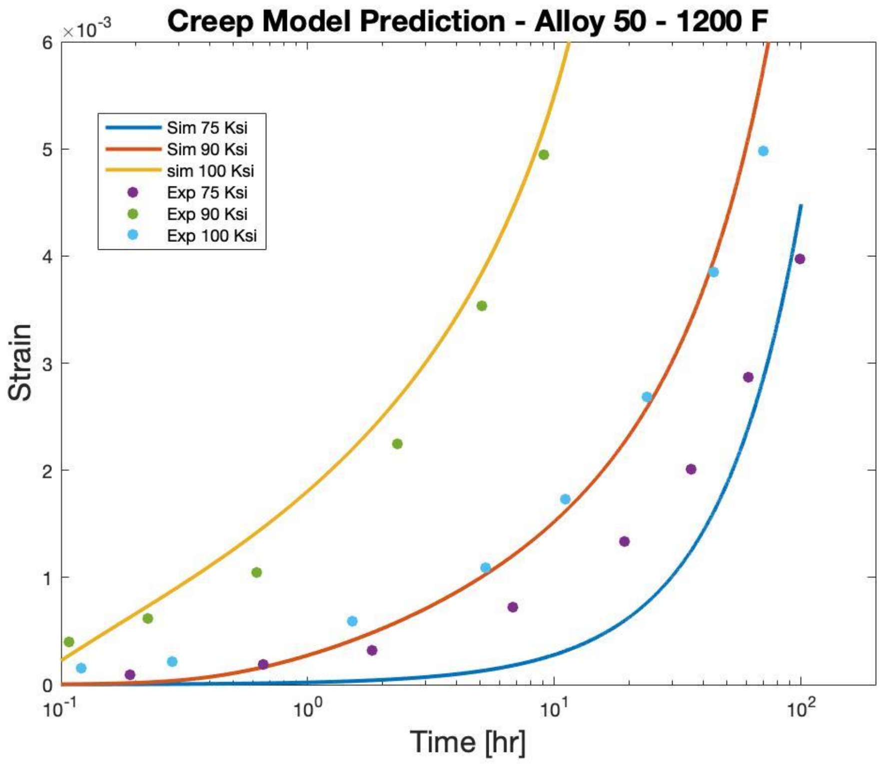
Creep Model - Results

- A mean field constitutive creep model considering dislocation creep and diffusion creep has been developed and calibrated
- The use of mean field model is justified as the microstructure is homogenous
- Deformation mechanisms considered are climb bypass and particle shearing
- Diffusion creep in the model accounts for grain boundary effects on damage accumulation.
- A MATLAB code has been developed for the fast-acting model.



Summary of Modeling Efforts

- Phase field model has been used to study coprecipitate formation mechanisms and to determine critical size on achieving compact coprecipitates
- High throughput thermodynamic calculations are carried out to aid experimental design of Round-3 alloy compositions
- Heterogeneous nucleation will be implemented in PanPrecipitation to model heterogeneous nucleation of γ'' on γ'. Direct coupling between multi-phase field model and PanDat thermodynamic database has also been explored with great promises
- Creep model framework has been developed and calibrated against creep data of Round 2 alloys.

THANK YOU
Dziękuję
감사합니다
Köszönöm
謝謝
Děkuji
Mulțumesc
Спасибо
Merci
Danke
Gracias
Obrigado
Terima kasih
ありがとう
Kiitos
Merci
Salamat
Takk
Teşekkür ederim
Dank U
Tack
Tak

Equations and Inputs for Creep Model

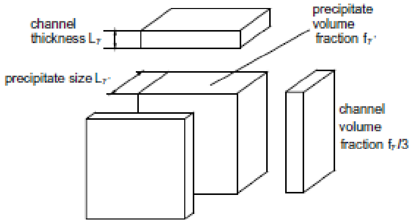
In γ matrix, plastic shear rate is given as:

$$\dot{\gamma}^{\alpha}_{\mathrm{FCC}} = 2\rho_{\mathrm{FCC}} b \lambda^{\alpha}_{\mathrm{FCC}} exp\left\{-\frac{Q^{\gamma}_{\mathrm{act}}}{k_B T}\right\} sinh\left\{\frac{\tau_{\mathrm{eff}} V_c}{k_B T}\right\}$$

$$\tau_{\mathrm{eff}} = |\tau + \tau_{\mathrm{mis}}| - \tau_{\mathrm{pass}} - \tau_{\mathrm{fric}}$$

- $Q^{\gamma}_{\mathrm{act}}$ is an *effective* energy barrier for rate-limiting process in the matrix
- τ_{eff} is the effective stress driving dislocation glide and is related to the strength model developed previously.

The composite compact and sandwich co-precipitate would be reduced to simplified γ' precipitate.



Representative microstructure used for mean field model

In γ' (similar for γ'') particles, plastic shear rate is given as

$$\dot{\gamma}^{\alpha}_{\mathrm{L1_2}} = 2\rho_{\mathrm{L1_2}} b \lambda^{\alpha}_{\mathrm{L1_2}} \nu_0 \sum_i P_i exp\left\{-\frac{Q^{i}_{\mathrm{act}}}{k_B T}\right\} sinh\left\{\frac{\tau_{\mathrm{eff}} V^{i}_{\mathrm{act}}}{k_B T}\right\}$$

$$\tau_{\mathrm{eff}} = \tau - \tau_{\mathrm{pass}}$$

- P_i − Percentage of mechanism i (obtained via Phase-field or DAD)
- $\rho^{\alpha}_{\mathrm{L1_2}}$ − dislocation density
- b – Burgers vector
- $\lambda^{\alpha}_{\mathrm{L1_2}}$ − channel width
- ν_0 − Attempt frequency
- V^{i}_{act} −Activation volume for mechanism i
- τ^{α} − Resolved shear stress for slip system α
- $\tau^{\alpha}_{\mathrm{pass}}$ − Passing stress for slip system α (threshold $stress$)

Key Inputs

- **Volume fraction, γ channel width**
- **Active deformation mechanism**
- **Experimental creep curve**