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Thermo-mechanical behavior of 410 stainless steel over a wide range of strain rates, temperatures, and loading conditions

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Dissertation

Presented in Partial Fulfillment of the Requirements for the Degree Doctor of Philosophy
in the Graduate School of The Ohio State University

By

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Graduate Program in Mechanical Engineering

The Ohio State University

2026

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Chon-Wei Ho

2026

Dedication

This document is dedicated to my family and close friends.

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C.-W. Ho, "A Discussion of Uncertainty in Helium Mass Spectrometer Testing for Electro-Explosive," in Proc. 23rd Nat. Conf. on Science and Technology of National Defense, Taoyuan, Taiwan, Nov. 2014.

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Abstract

This research investigates the thermomechanical behavior of 410 stainless steel under tension, compression, and torsion across a wide strain rate range from 10^{-4} s^{-1} to 5000 s^{-1} . The material reveals identical yielding stresses across all loading modes, however, the material exhibits more pronounced hardening under tensile loading, followed by compression, with minimal hardening observed under torsion. The compressive loading analysis accounted for physical barreling by utilizing analytical methods and reverse engineering to correct the flow stress for frictional effects. Anisotropy testing shows Lankford coefficients close to 1.0 in tension.

At high strain rates, the material exhibits adiabatic heating across all loading modes, while quasi-static and intermediate rates are characterized by significant heat loss. Under adiabatic conditions, temperature increase maintains a linear relationship with plastic strain in all modes. In contrast, heat loss patterns vary by loading mode depending on specimen geometry. The Taylor-Quinney Coefficient (β) revealed a distinct three-stage evolution: an initial minimum at the onset of plastic deformation, a gradual rise during deformation progress, and stabilization as the balance between stored energy and dissipated thermal energy is reached.

TQC analysis demonstrated a clear loading mode dependency, where torsion exhibited the highest values (0.86 to 0.94), followed by tension and compression, respectively. Fourier number (Fo) calculations and strain-temperature distributions were utilized to verify adiabatic conditions and validate the TQC results. High-rate tests consistently displayed adiabatic behavior, ensuring valid TQC measurements. At the intermediate strain

rate of 1.0 s^{-1} , heat loss was observed across all loading modes. TQC decayed less in tension due to shorter test durations and longer thermal paths. In comparison, compression and torsion specimens experienced greater decay due to longer test durations, shorter thermal paths, and larger conductive surface areas. These intermediate strain rate results identified the threshold for the transition between adiabatic heating and heat-loss-dominated behavior.

To model these complex responses, a modified Johnson–Cook (J–C) constitutive equation was developed. This improved formulation addresses the limitations of the traditional J–C model by replacing the linear logarithmic strain rate term with a non-linear exponential formulation and integrating a polynomial temperature term. The modified model achieved high predictive accuracy across all loading modes, with R-squared values above 0.93 and Mean Absolute Percentage Errors (MAPE) below 1.3%.

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Chapter 1 Introduction

1.1 Significance of the Problem

Understanding deformation-induced heating during plastic deformation of metals is essential for predictive numerical simulations of industrial applications and processes for improved product and process design. When materials deform rapidly, heat generation due to plastic deformation can affect their mechanical properties [1][2]. This can influence strength, ductility, damage evolution and fracture onset [3][4]. If not well understood or ignored in the numerical modeling and design process, these effects may lead to faulty engineered component or system failures.

Traditionally, thermal mechanical behavior is determined mainly from mechanical tests conducted at isothermal elevated temperatures. These tests use thermocouples to measure the temperature. They give insight into the temperature dependent material behavior but cannot capture real-time temperature changes due to deformation. Even when infrared sensors are used, single-point sensors rarely capture the peak temperature because localization zones cannot be predicted in advance [5][6]. As a result, sensor placement may miss critical regions of heating.

Another important issue is how to model the thermomechanical behavior of metals. Most models, such as the Johnson–Cook model [7], are based on two separate test series: isothermal, elevated-temperature tests at quasi-static rates and room-temperature tests at various strain rates. This approach treats strain rate and temperature effects independently. In reality, both effects are coupled. For some metals, this interaction is not easy to predict.

Dynamic heating can also complicate the result [6][8]. Measuring the temperature rise during deformation gives a direct way to capture both effects at the same time. Moreover, comparing the thermal softening parameter calibrated from isothermal tests with one obtained from adiabatic heating in high-rate experiments could reveal whether infrared measurements alone are sufficient. If the two approaches agree, full field IR-based measurements may provide a more efficient pathway to reliable constitutive modeling.

Most studies focus only on uniaxial tension or compression loadings. They do not consider the full range of stress states that a material may experience. Energy partition is influenced not only by strain and strain rate but also by the geometry of deformation [8]. Different loading modes can change how plastic work is stored or dissipated as heat. Ignoring these effects gives only a partial view of how stress state controls temperature rise and thermal softening. Expanding the research to a wider range of loading conditions may show whether the Taylor–Quinney coefficient is stress-state dependent in 410ss. Such knowledge would improve constitutive models and support more reliable predictions of material behavior under extreme conditions.

1.2 Literature Review

Deformation-induced heating arises from the partial conversion of plastic work into heat during irreversible deformation. Early studies by Farren and Taylor [10] and Taylor and Quinney [11] established that a large fraction of mechanical work is dissipated as heat rather than stored in the material. These pioneering studies motivated extensive research to quantify the Taylor–Quinney coefficient and understand how strain, strain rate, temperature, and loading condition affect the conversion of plastic work into heat. As

experimental techniques advanced, researchers began investigating deformation-induced heating under complex loading conditions and over wider ranges of strain rates.

During the 1970s and 1980s, researchers began to study how plastic work turns into heat during deformation. Korhonen and Kleemola [12] tested sheet steels at strain rates between 10^{-4} and 10^{-1} s^{-1} . They used heat conduction models and conducted tensile tests. Their results showed that heating reduced uniform strain during deformation. They also showed that heat transfer conditions, such as in air, forced convection, or immersion, controlled whether the test behaved close to isothermal or adiabatic. Lin and Wagoner [13] extended this work. They tested interstitial-free steel and 310 stainless steels at strain rates from 10^{-4} to 10^{-1} s^{-1} . Thermocouples were placed along the gauge length, and three cooling conditions: air, insulation, and stirred water, were applied. Their results showed that a uniaxial tensile test at room temperature was neither isothermal nor adiabatic. The neck region heated by as much as $118 \text{ }^{\circ}\text{C}$. Uniform heating lowered the work-hardening rate. Local heating along the axis created gradients that pushed the strain to localize. Their work gave a clear path for later finite element models. Ayres [14] studied ductility in 1008 AK steel using several thermocouples along tensile specimens. His tests showed temperature gradients up to $110 \text{ }^{\circ}\text{C}$ in air at strain rate of 10^{-1} s^{-1} , the total elongation stayed near 40%. When steel was tested in water baths to remove temperature gradients, elongation increased to 54%. Thermal gradients reduced ductility and strain-rate sensitivity increased ductility. These two effects balanced each other.

By the 1990s, advances in experimental techniques enabled more accurate measurements. Mason et al. [15] used a split Hopkinson bar with high-speed infrared

detectors to measure stress, strain, and temperature during dynamic deformation of 4340 steel, 2024 aluminum, and Ti-6Al-4V titanium. They provided strong evidence that the Taylor–Quinney coefficient is strain-dependent rather than constant. For 2024 aluminum and 4340 steel, β increased from approximately 0.5 to 0.85 with increasing strain and exhibited little strain-rate dependence up to about 3000 s^{-1} . Kapoor and Nemat-Nasser [5] refined this approach. They used both infrared detectors and recovery Hopkinson bar method. They studied Ta alloys, Ti, Cu, Al, and steels, and found that nearly all plastic work turned into heat during high strain rate deformation. They also found that infrared detectors alone gave lower values because of calibration errors. Their method of loading, unloading, and reloading samples stated that the true conversion ratio close to 100% during high strain rate deformation.

In the early 2000s, researchers moved from simple thermocouples toward calibrated infrared detection and more refined models to study how plastic work converts into heat. Rusinek and Klepaczko [1] investigated the Taylor–Quinney coefficient (β) in TRIP steels and mild steels using hydraulic machines under nearly adiabatic conditions. Their results showed that β increased with plastic strain, rising from approximately 0.82 to nearly 1.0 at large strains. TRIP steels generated more heat than mild steels because phase transformations released latent heat. Their work emphasized that β is not constant and that microstructural processes, such as transformations, directly contribute to heat generation. Nemat-Nasser and Guo [16] investigated DH-36 structural steel over a wide range of strain rates, from 10^{-3} s^{-1} to 8000 s^{-1} , and temperatures from 77 K to 1000 K. While their study focused on mechanical response, they also documented significant temperature rises during

plastic deformation. At high strain rates, the heating was essentially adiabatic, and local temperature rises were linked with the onset of shear bands. This showed that deformation-induced heating, not just external test temperature, controlled when and where localization occurred. Their results connected adiabatic heating to failure mechanisms in structural steels under dynamic loading.

Macdougall [2] advanced the measurement of temperature rise by refining infrared radiometry. He investigated aluminum 2024 and Ti-6Al-4V in a split Hopkinson bar under compression and torsion, supported by fast thermocouples for calibration. The results showed that β increased with strain, beginning near 0.5 and rising toward 0.9~1.0. This confirmed that early deformation stores more energy in the structure, while larger strains convert nearly all plastic work into heat. Calibration reduced uncertainty from surface emissivity and showed that infrared radiometry could capture microsecond-scale thermal signals during high-rate tests. Hodowany et al. [6] analyzed the partition of plastic work into heat and stored energy in more detail. They tested aluminum 2024 and α -titanium using both servo hydraulic frames and Kolsky bars with high-speed HgCdTe infrared detectors. The results showed that β depended on both strain and strain rate. For aluminum, β at small strains was as low as 0.3, but it increased toward unity at larger strains. For α -titanium, the rate of increase in β was faster at higher strain rates, meaning that less energy was stored as defects when loading was rapid. Their study demonstrated that energy storage is transient and that adiabatic conditions drive most plastic work into heat at large strains.

In the 2010s, advances in imaging and diagnostics shifted deformation-induced heating research from isolated measurements toward full-field and high-speed observations.

Gilat et al. [17] studied stainless steel 316L using digital image correlation combined with infrared cameras in both quasi-static tests at about 10^{-3} s^{-1} and dynamic Hopkinson bar tests near 10^3 s^{-1} . Their measurements showed that strain localization coincided with thermal hot spots, and that instability at both low and high rates was linked to adiabatic heating. This work provided one of the clearest demonstrations that strain and temperature fields could be measured simultaneously during plastic deformation. Zhang and Di [18] studied Ti-15-3 β -titanium alloy at strain rates of $10^2 \sim 10^3 \text{ s}^{-1}$. They reported strong adiabatic heating during high-rate deformation, which triggered localized flow instabilities and shear bands. Although the material is different from stainless steel, their findings emphasized that heating is a direct driver of localization, a mechanism that also governs plastic instability in other alloys. Rahmaan et al. [19] investigated AA7075-T6 aluminum in shear across strain rates from 10^{-3} up to 10^3 s^{-1} . They used a high-speed infrared camera to track temperature rise during deformation. Their results showed that at higher rates, heating produced strong thermal softening that lowered flow stress and accelerated localization. Bragov et al. [20] performed compression tests at very high strain rates of $10^3 \sim 10^4 \text{ s}^{-1}$ using split Hopkinson pressure bars. They measured temperature rise through a combination of infrared detectors and embedded thermocouples. The specimens heated significantly within milliseconds of loading, which altered the stress-strain response compared to quasi-static cases. Their work confirmed that rapid self-heating is unavoidable in dynamic plasticity and depends strongly on the alloy system. Rittel et al. [8] measured the Taylor-Quinney coefficient under tension, compression, and shear at dynamic strain rates of $10^2 \sim 10^3 \text{ s}^{-1}$. Using high-speed infrared detectors. They found that β depended on loading mode. In shear, β approached unity at small strains, while in tension it evolved more

gradually. Their results confirmed that energy partition is influenced not only by strain and strain rate but also by the stress state.

During the same period, researchers expanded the application of deformation-induced heating measurements beyond fundamental material characterization. Pieczyska et al. [3] studied Gum Metal, a β -titanium alloy, under quasi-static to medium strain rates ($10^{-3} \sim 10 \text{ s}^{-1}$). They used infrared cameras to map thermal signatures during yielding and strain bursts. Their results highlighted how defect activity and phase transitions are captured as measurable temperature spikes, demonstrating that thermography is sensitive to microstructural events even in unconventional alloys. Moyseychik et al. [21] applied infrared thermography to steel structures at low strain rates ($10^{-3} \sim 10 \text{ s}^{-1}$). They showed that localized plastic zones and early cracks generated detectable heat bursts, extending deformation heating analysis from laboratory specimens to nondestructive evaluation of engineering components. Juan et al. [22] used infrared thermography in forming processes at quasi-static speeds ($10^{-3} \sim 10 \text{ s}^{-1}$). They reported that faster forming increased local heating and demonstrated that thermal imaging could map heterogeneous temperature fields across entire workpieces. These application-oriented studies demonstrated how deformation-induced heating measurements could support both industrial forming analysis and structural health monitoring. These studies demonstrated that deformation-induced heating measurements are valuable not only for investigating plastic deformation mechanisms but also for industrial forming analysis and structural health monitoring.

In the 2020s, deformation-induced heating research entered a period of rapid development, driven by synchronized optical diagnostics, high-speed thermal imaging, the

use of machine learning, and multiphysics modeling. A key advancement was the integration of full-field digital image correlation with infrared thermography, allowing strain and temperature fields to be measured simultaneously during dynamic deformation. Soares et al. [23]~[26] conducted a series of studies on steels, commercially pure titanium, and stainless steel 316, with tests ranging from quasi-static (10^{-3} s^{-1}) to Hopkinson-bar conditions (10^3 s^{-1}). Using synchronized DIC and high-speed infrared cameras, they showed that strain localization was consistently accompanied by thermal hot spots. Their studies published between 2021 and 2022 showed that full-field imaging can be used to estimate the Taylor–Quinney coefficient. These studies established full-field strain and temperature mapping as a reference method for analyzing adiabatic heating.

Recent studies have used synchronized thermomechanical measurements to investigate how material composition and loading history influence deformation-induced heating. Žaba et al. [27][28] extended this approach to nickel-based superalloys and stainless steels. Their results showed that both alloy composition and loading rate affected thermal softening and the coupled thermomechanical response. Musiał et al. [29] applied similar analysis to 310S stainless steel, quantifying how mechanical work was partitioned between stored energy and heat at strain rates up to 10 s^{-1} . Vázquez-Fernández et al. [30] investigated stainless steels subjected to sudden strain-rate jumps and found that temperature increased sharply following the rate change, altering the strain-hardening behavior. High-speed infrared imaging captured these transient thermal responses and demonstrated the sensitivity of stainless steels to dynamic rate shifts. The relationship between deformation-induced heating and fracture also received attention. Lee et al. [4]

studied steels at strain rates up to 10^3s^{-1} and found that temperature rise followed closely with fracture initiation. Rittel et al. [31] analyzed dynamic shear failure, demonstrating that adiabatic heating is not merely a by-product of deformation but a key factor governing plastic instability and failure.

The 2020s also saw the expansion of deformation-induced heating research from experimental observations to comprehensive reviews and advanced modeling approaches. Jabłońska [32] reviewed TWIP steels and compiled data on Taylor–Quinney values, strain-rate sensitivity, and thermal dissipation. Karantza and Manolakos [33] provided a broader review of adiabatic shear banding, emphasizing the relationship between deformation-induced heating, microstructural evolution, shear localization, and fracture. These studies highlighted the importance of understanding energy conversion mechanisms across different material systems. At the same time, computational and artificial intelligence-based methods became more common. Churyumov et al. [34] used artificial neural networks to model high-temperature steel deformation, aiming to predict plastic flow without relying on fixed constitutive parameters. Hartmann and Obermeyer [35] revisited variational models of plasticity-induced heating by incorporating both stored and dissipated energy into an energy-based framework. Smith et al. [36] further advanced experimental methodology by measuring Taylor–Quinney coefficients in both uniform and localized deformation during tensile tests. Using synchronized cameras and infrared imaging, they showed how β evolves differently in homogeneous versus necking regions, providing a more spatially resolved description of energy partition during plastic deformation.

Over the past several decades, studies of deformation-induced heating have progressed from thermocouple measurements to infrared and DIC-based full-field techniques. These developments have improved the ability to capture the local temperature evolution during plastic deformation and demonstrated that a significant portion of plastic work is converted into heat. Previous studies have shown that the Taylor–Quinney coefficient depends on strain, strain rate, and material type. However, most investigations have focused on a single loading mode, making systematic comparisons among different stress states difficult. The influence of stress state on heat generation and thermal softening therefore remains unclear. To address these gaps, this dissertation investigates the temperature rise and Taylor–Quinney coefficient of 410 stainless steel under tension, compression, and torsion across a wide range of strain rates. The resulting experimental data provide a unified basis for evaluating the effects of loading mode on thermomechanical behavior and for developing the constitutive model.

Chapter 2 Experimental Techniques and Procedures

This Chapter details the experimental methodologies and procedures utilized to characterize the thermo-mechanical behavior of 410 stainless steel. To capture the material's response across various environments from static to dynamic loadings. This matrix includes three loading modes: tension, compression, and torsion, executed across a spectrum of strain rates from 10^{-4} s^{-1} to 5000 s^{-1} . Furthermore, the influence of temperature and anisotropy was investigated to provide a robust dataset for constitutive modeling. The following sections describe the specimen designs, the instrumentation of the Split-Hopkinson Pressure Bar (SHPB), and the synchronization of high-speed imaging with infrared thermography to quantify the conversion of plastic work into heat.

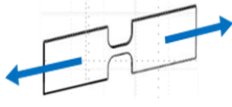
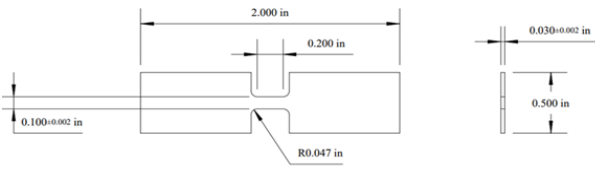
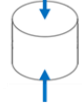
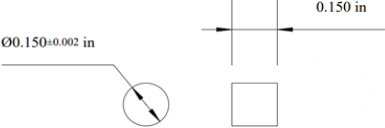
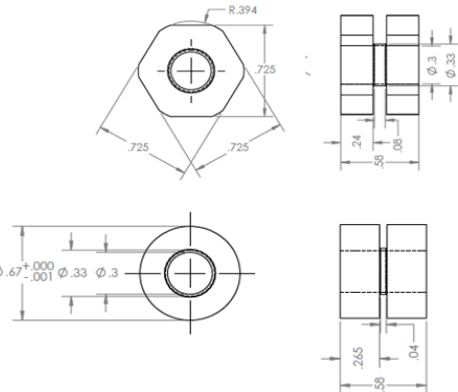
2.1 Material Properties and Test Matrix

The material investigated in this study is a 410 stainless steel plate stock with the thickness of 12.7 mm. The chemical composition and baseline mechanical properties of the alloy are summarized in Table 2-1, confirming a chromium content of 11.95% and a measured yield strength of 326.81 MPa with a significant elongation at fracture of 38%. To characterize the thermo-mechanical response of this material, an extensive experimental test matrix was executed across three loading modes: tension, compression, and torsion. As detailed in Table 2-2, these tests span a wide range of strain rates, from the quasi-static (10^{-4} s^{-1}) to high-strain-rate dynamic loading (500 s^{-1} ~ 5000 s^{-1}). Specific specimen geometries were designed for each mode to ensure data acquisition, including flat dog-bone specimens for tensile testing, cylindrical sample for compression, and specialized hexagonal and circular geometries for torsion.

Table 2-1 Material Properties and Chemical Composition of 410 Stainless steel

Plate Stock		Mechanical Properties				Chemistry										
Material	Thickness (in)	Ultimate Strength (MPa)	Yield Strength (MPa)	Elongation (%)	C	Mn	P	S	Si	Cr	Ni	Al	Mo	Cu	N	Sn
SS 410	0.5	540.55	326.81	38	0.13	0.51	0.025	0.001	0.61	11.95	0.24	0.01	0.05	0.15	0.01	0.01

Table 2-2 Test Matrix

Test mode	Strain Rate (1/s)	Sample
Tension 	1.00E-04 1.00E-03 1.00E-02 1 500 2000	
Compression 	1.00E-03 1.00E-02 1 2000 5000	
Torsion  	1.00E-04 1.00E-02 1 1000 4500	

2.2 Specimen Design

The design of the specimens for this study is driven to maintain stress state and facilitate data acquisition across a wide range of strain rates. The geometries are adapted from the methodologies established by Ressa [38], with modifications to accommodate the specific material thickness of the 0.5-inch 410 SS plate and the requirements of synchronized DIC and IR diagnostics.

2.2.1 Tensile Specimen Design

The tensile specimens are designed as flat dog-bone geometries with a gauge section of 0.2 inches in length and 0.1 inches in width, Figure 2-1. While the length of the grip sections is adjusted to accommodate the specific mounting requirements and stroke limitations of the high-strain-rate Split-Hopkinson Bar (SHB) setup. The gauge section dimensions remained constant across all strain rates. The transition from the grip to the gauge utilizes a 0.047-inch radius fillet to minimize stress concentrations and ensure localization within the gauge section.

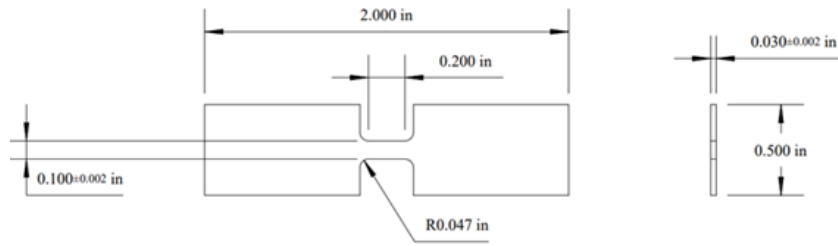


Figure 2-1 Engineering drawing of tension specimen

2.2.2 Compression Specimen Design

Compression specimens are small cylinders with a diameter and height of 0.15 inches ($L/D=1$), Figure 2-2. These dimensions are designed for high-strain-rate testing. It provides a balance between minimizing longitudinal inertia effects and preventing structural buckling. The small length of these specimens allows for rapid stress equilibration during the high-velocity impact pulses of the SHPB. In addition, the small specimen diameter keeps the transmitted force low enough for the SHPB system to achieve the highest targeted strain rates.

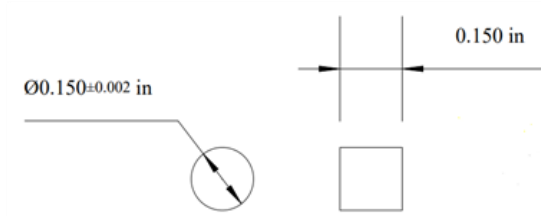


Figure 2-2 Engineering drawing of compression specimen

2.2.3 Torsion Specimen Design

The torsion specimen geometry is designed to accommodate different fixture interfaces while maintaining a consistent gauge wall thickness of 0.03 inches. For low-strain-rate testing, a hexagonal outer geometry is utilized to provide a mechanical interlock with the quasi-static torsion fixture. For high-strain-rate testing, the gauge length is reduced to achieve the targeted higher shear strain rates.

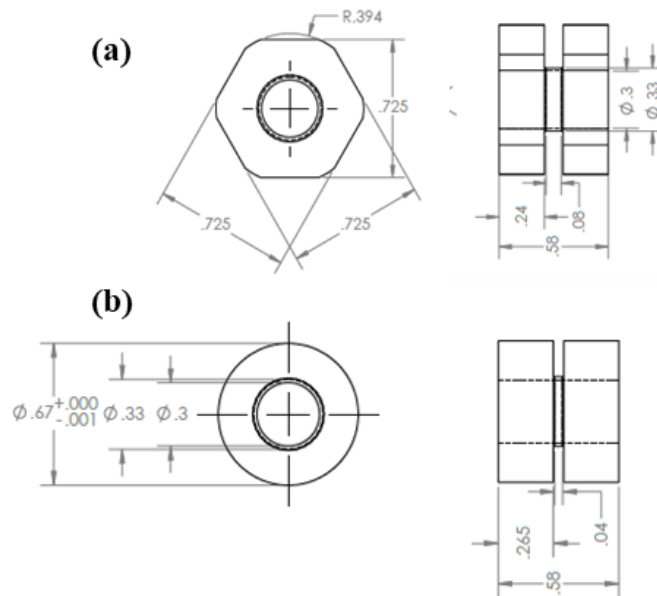


Figure 2-3 Engineering drawing of torsion specimens used for (a) Quasi-static and intermediate tests (b) Dynamic tests

2.3 Experimental Techniques of Tension Tests

The tensile testing program is conducted across six distinct strain-rates, each requiring specific test apparatuses and settings. Digital Image Correlation (DIC) and Infrared (IR) thermography are integrated into the requirements of each test to ensure accurate strain and temperature mapping. The specific data acquisition parameters, including camera types, frame rates, and apparatus for each tensile test, are detailed in Table 2-3.

Table 2-3 Camera parameters and experimental apparatus for tension testing

Test mode	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate	IR Camera	IR Frame Rate
Tension	1.00E-04	Servo-hydraulic load frame	Two Point Grey Gazelles with 35 mm lens	0.2	Telops FAST-IR	0.2
	1.00E-03			2	Telops FAST-IR	2
	1.00E-02			20	Telops FAST-IR	20
	1	Tension SHB	Photron MC2.1 100 with 35 mm lens	2000	Telops FAST-IR	500
	500		Shimadzu HPV-X2 with 180 mm lens	200000	Telops FAST-IR	82000
	2000			500000	Telops FAST-IR	82000

2.3.1 Quasi-static and Intermediate tensile tests

Quasi-static tests (10^{-4} to 10^{-2} s⁻¹) and intermediate tests (1.0 s⁻¹) are performed using an Instron 1321 servo-hydraulic load frame testing machine. The experimental configuration, shown in Figure 2-4, features dual visual (DIC) and IR cameras that are positioned facing each other to simultaneously map the front and back surfaces of the specimen gauge. Typical images of these cameras are shown in Figure 2-5. To ensure synchronization between the mechanical deformation and the thermal response, the frame rates for both the DIC and IR cameras are set to identical values.

Axial deformation in tensile testing is typically measured using extensometer, which determines the engineering strain from the change in gauge length. Because the

extensometer reports an average strain over the gauge length, it cannot accurately capture the local deformation once strain localization develops. Accurate strain measurement is important in this study because the Taylor–Quinney coefficient (TQC) is determined from the plastic work, which depends on the measured plastic strain.

To accurately characterize localized deformation, Digital Image Correlation was employed to obtain full-field surface strain measurements. Among the available strain tensors, the logarithmic (Hencky) strain was selected because it provides an accurate representation of local deformation after the strain field becomes non-uniform. Assuming constant volume during plastic deformation while neglecting the very small elastic volume changes, the axial Hencky strain component (ϵ_{yy}) was adopted as the local true strain because it represents deformation along the loading direction.

The true stress equation (eq.2-1) is derived under the assumption of uniform deformation and is valid only before necking. After strain localization, the engineering strain measured by an extensometer no longer represents the deformation experienced within the necking region. In the present work, the engineering strain in the true stress equation is replaced by the engineering strain converted from the local Hencky strain measured by DIC (eq.2-2).

$$\sigma_t = \sigma_e(1 + \epsilon_e) \quad 2-1$$

$$\epsilon_e = \exp(\epsilon_t) - 1 \quad 2-2$$

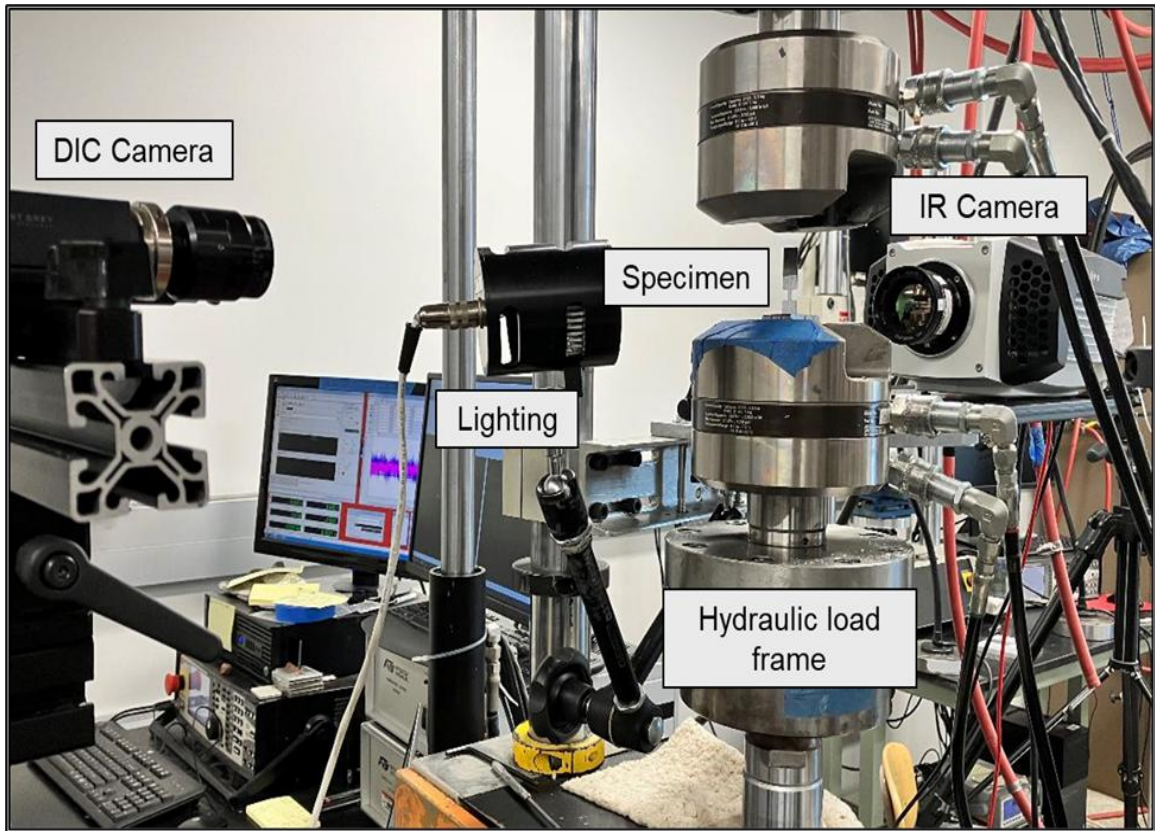


Figure 2-4 Experimental setup for quasi-static and intermediate tensile tests

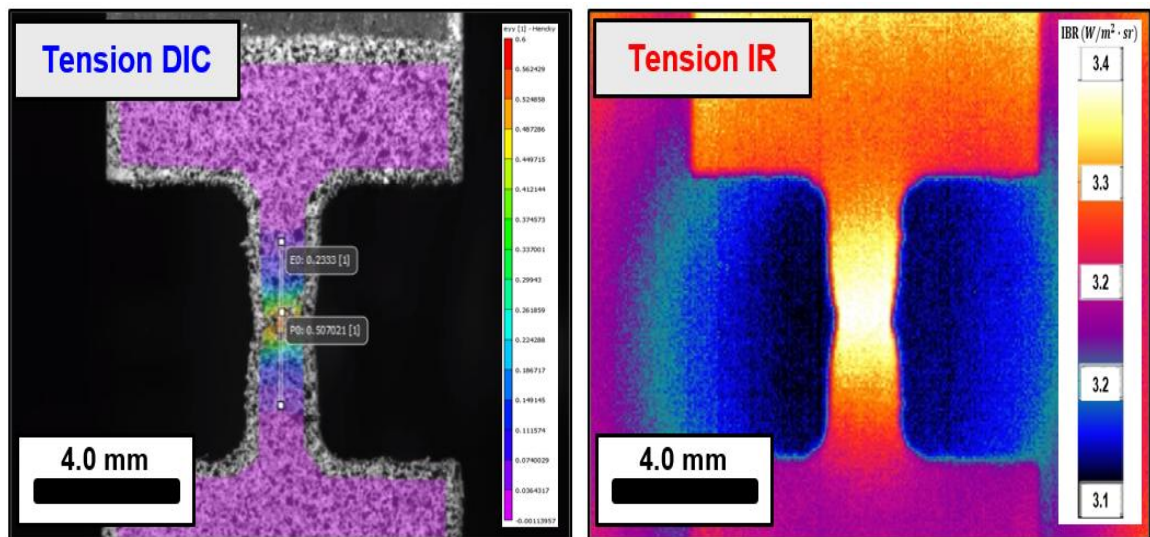


Figure 2-5 Representative DIC and IR images obtained during quasi-static and intermediate tensile tests

2.3.2 Dynamic Tensile Tests

High-strain-rate tensile tests (500 to 2000 s^{-1}) are conducted using a Direct Tension Split-Hopkinson Bar (SHB) apparatus. As illustrated in Figure 2-6, the system consists of two 7075-T6 aluminum bars with a diameter of 0.5 inches: an incident bar (12 ft. long) and a transmitter bar (5.5 ft. long). The loading wave is generated by the sudden release of a static tensile force initially clamped into the incident bar. Upon release, a tensile pulse propagates toward the specimen, which is fastened between the bars using a high-strength, two-part epoxy adhesive to ensure rigid load transfer. The waves are measured at two locations on the incident bar (Gages *A* and *B*) and one on the transmitter bar (Gage *C*) using 1000 Ohm strain gages in full Wheatstone bridge circuits to isolate the axial load signals. The detail of SHB technique for high strain rate tension is described in Staab and Gilat [39]. Similar to the quasi-static setup, the DIC and IR cameras are positioned facing each other (Figure 2-7) to capture simultaneous strain and temperature. The triggering of these high-speed cameras is established following the methodology proposed by Smith [36] to ensure precise alignment with the mechanical wave arrival. The resulting synchronized images are shown in Figure 2-8.

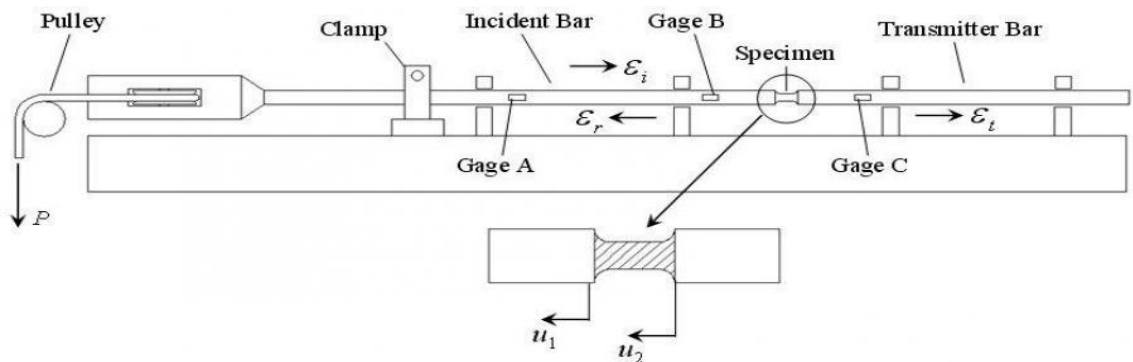


Figure 2-6 Schematic of the tensile Split-Hopkinson Bar (SHB) apparatus

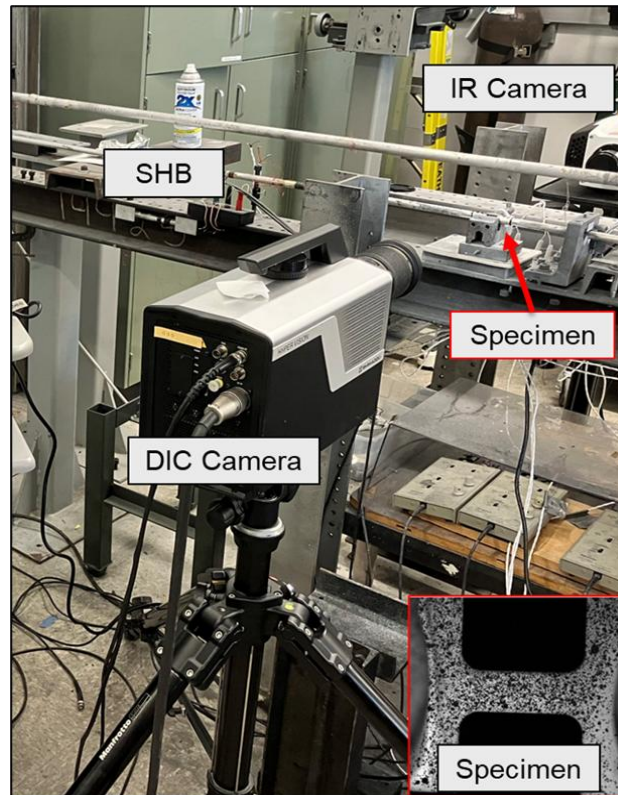


Figure 2-7 Experimental setup for dynamic tensile tests

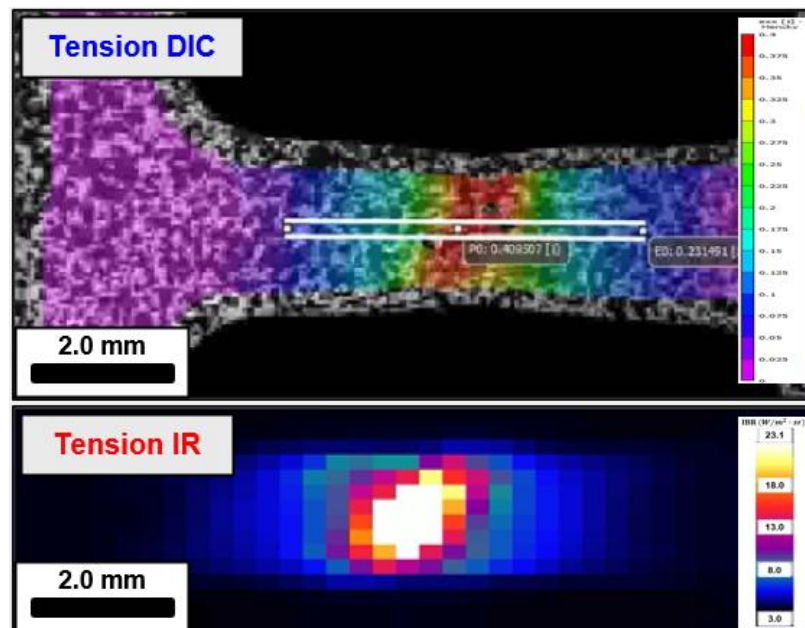


Figure 2-8 Representative DIC and IR images obtained during dynamic tensile tests

2.3.3 Elevated Temperature Tests

To evaluate temperature dependency and thermal softening, specimens are heated using a furnace to target temperatures of 200°C, 400°C, 600°C, and 800°C. The setup (Figure 2-9) includes a cooling system to protect the load frame and maintain the accuracy of the load cell measurements by preventing heat transfer from the furnace to the load cell. To ensure accurate thermal control, a K-type thermocouple is cemented to the back of the center of the gauge section and connected to a temperature logger for real-time monitoring (Figure 2-10). The camera parameters and apparatus for these tests are listed in Table 2-4. Unlike the room temperature tests, the elevated temperature testing does not utilize an IR setup, as the furnace environment and specimen radiation interfere with thermal imaging. Stress-strain curves are calculated following the same procedures as the quasi-static tests.

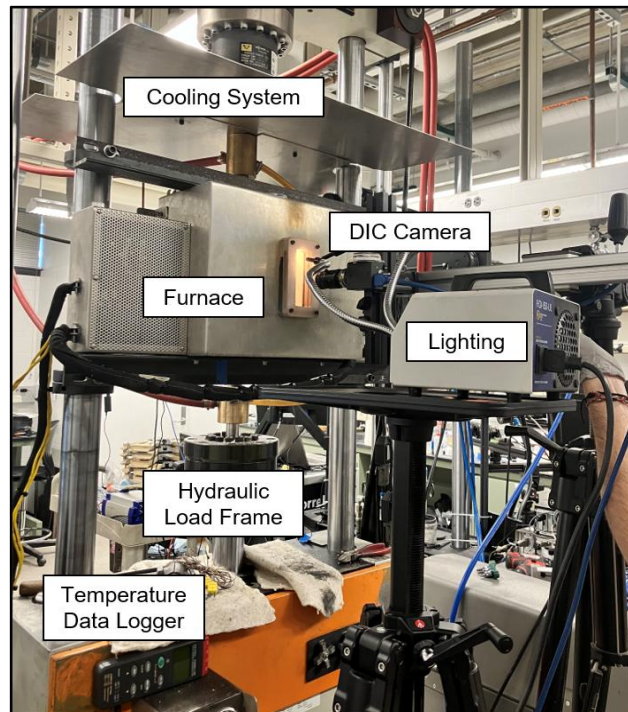


Figure 2-9 Experimental setup for elevated-temperature tensile tests at a strain rate of 10^{-3} s^{-1}

Table 2-4 Camera parameters and experimental apparatus for elevated-temperature tensile tests

Test mode	Temperature (°C)	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate
Tension	RT	1.00E-03	Servo-hydraulic load frame with furnace	Two Point Grey Gazelles with 35 mm lens	2
	200				
	400			Point Grey Gazelles with 35 mm lens	
	600				
	800				

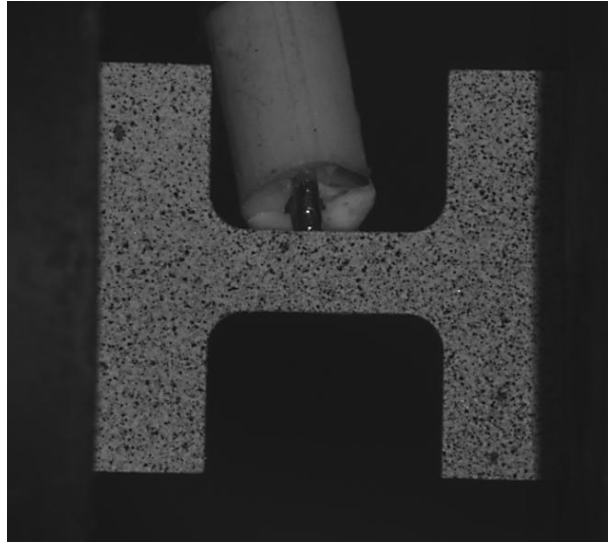


Figure 2-10 Thermocouple installation for elevated-temperature tensile tests

2.3.4 Anisotropy Tests

The anisotropy tests are conducted to identify the directional dependence of the 410 SS plate properties. The experimental setup and diagnostic parameters remain identical to the quasi-static tensile tests, as summarized in Table 2-5. Specimens are extracted at 0°, 45°, and 90° orientations relative to the rolling direction. Unlike the conventional thin-sheet configuration used for sheet-metal anisotropy tests, the specimens in this study were machined from the plate thickness. Figure 2-11 compares the conventional specimen configuration (thin-sheet) with the thick-plate configuration adopted in this study.

Plastic anisotropy is quantified by the Lankford coefficient (R-value), which represents the ratio of transverse plastic strain to thickness plastic strain ($R = \epsilon_w^p / \epsilon_t^p$), where ϵ_w^p is the transverse strain and ϵ_t^p is the thickness strain. This mathematical definition remains unchanged in the present study because the R-value is defined with respect to the specimen coordinate system. However, due to the thick-plate specimen configuration shown in Figure 2-11, the physical meaning of the measured strains differs from that of the conventional thin-sheet configuration. In the present study, the specimen width direction coincides with the plate through-thickness direction, whereas the specimen thickness direction corresponds to different in-plane material directions depending on the specimen orientation. The calculated R-values characterize the relationship between the plate through-thickness plastic strain and the plastic strain along different in-plane material directions. For example, the 0° specimens relate the through-thickness strain to the transverse direction, the 90° specimens relate the through-thickness strain to the rolling direction, and the $\pm 45^\circ$ specimens relate the through-thickness strain to the corresponding $\pm 45^\circ$ directions of the plate.

Because the DIC system measures only the specimen surface, the thickness strain cannot be measured directly. Assuming constant plastic volume, the plastic deformation satisfies $\epsilon_1^p + \epsilon_2^p + \epsilon_3^p = 0$, where ϵ_1^p , ϵ_2^p , and ϵ_3^p are the principal Hencky (logarithmic) strains. The DIC system directly measures the in-plane principal strains, where ϵ_1^p corresponds to the longitudinal (loading) direction and ϵ_2^p corresponds to the transverse direction. The thickness plastic strain is then calculated $\epsilon_3^p = -(\epsilon_1^p + \epsilon_2^p)$. Substituting the calculated thickness strain into the R-value definition gives:

$$R = \frac{\epsilon_2^p}{-(\epsilon_1^p + \epsilon_2^p)} \quad 2-3$$

Table 2-5 Camera parameters and experimental apparatus for tensile anisotropy tests

Test mode	Specimen Orientation	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate
Tension	0°	1.00E-03	Servo-hydraulic load frame	Point Grey Gazelles with 35 mm lens	2
	90°				
	+45°				
	-45°				

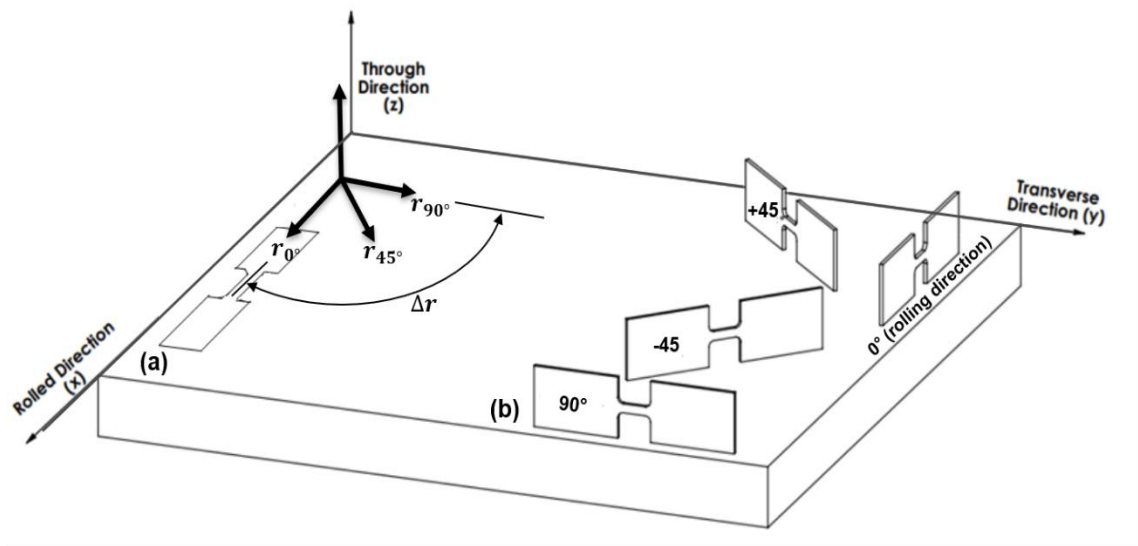


Figure 2-11 Comparison of (a) the conventional thin-sheet specimen configuration (b) the thick-plate specimen configuration used in this study

2.4 Experimental Techniques of Compression Tests

The compression testing program is conducted across three strain-rate regimes, loading frames and configurations. To ensure consistent mapping of the material response, Digital Image Correlation (DIC) and Infrared (IR) thermography are integrated into each experimental setup to capture synchronized surface strain and temperature. Furthermore, to isolate the intrinsic constitutive behavior of the 410 stainless steel from geometric

artifacts, an analytical model for the evaluation of barreling is implemented to correct the measured flow stress. The specific data acquisition parameters, including camera specifications, frame rates, and the testing apparatus for each compression test, are detailed in Table 2-6.

2.4.1 Quasi-static and Intermediate strain rate Compression Tests

Quasi-static (10^{-3} to 10^{-2} s $^{-1}$) and intermediate (1 s $^{-1}$) strain rate compression tests are performed using an Instron servo-hydraulic load frame. To capture the simultaneous mechanical and thermal response, DIC and IR cameras are positioned facing each other and synchronized at identical frame rates. The experimental setup for these tests is illustrated in Figure 2-12, and representative synchronized images from the DIC and IR systems are shown in Figure 2-13. The engineering stress, σ_e , and engineering axial strain, ε_e , are calculated from the recorded load, F , and initial specimen dimensions (A_0 , L_0):

$$\sigma_e = \frac{F}{A_0} \quad 2-4$$

$$\varepsilon_e = \frac{\Delta L}{L_0} \quad 2-5$$

These values are converted to true axial strain, ε_t , and true stress, σ_t , using the following conversion relationships:

$$\sigma_t = \sigma_e(1 + \varepsilon_e) \quad 2-6$$

$$\varepsilon_t = \ln(1 + \varepsilon_e) \quad 2-7$$

The displacement, ΔL , used for the strain calculation is derived from two distinct DIC measurements. One is the relative movement of the loading platens and the other is the localized surface displacement within the specimen gauge section, as illustrated in Figure

2-14. During the initial phase of the test, the surface-measured displacement is compared to the platen-measured displacement to verify system rigidity. A match between these two values confirms the minimal impact of platen compliance, which might cause an overestimation of the strain. While the surface measurement is highly accurate at the start, it eventually underestimates the displacement as specimen barreling causes the surface to move out of the initial plane. Therefore, once the platen data is verified against the surface data to ensure no compliance issues exist, the displacement from the platens is utilized for the stress-strain calculation to ensure the full compressive behavior is accurately captured.

Table 2-6 Camera parameters and experimental apparatus for compression tests

Test mode	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate	IR Camera	IR Frame Rate
Compression	1.00E-03	Servo-hydraulic load frame	Two Point Grey Gazelles with 35 mm lenses	2	Telops MS-IR	2
	1.00E-02			20	Telops MS-IR	20
	1	Compression SHB	Two Photron MC2.1 100 with 50 mm lens	500	Telops MS-IR	500
	2000		Shimadzu HPV-X2 with 180 mm lens	250000	Telops MS-IR	72000
	5000			625000	Telops MS-IR	72000

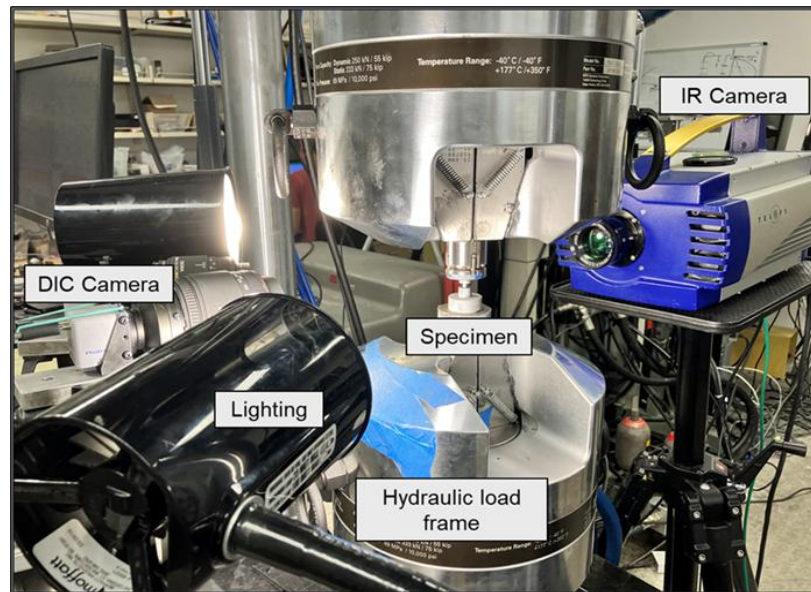


Figure 2-12 Experimental setup for quasi-static and intermediate compression tests

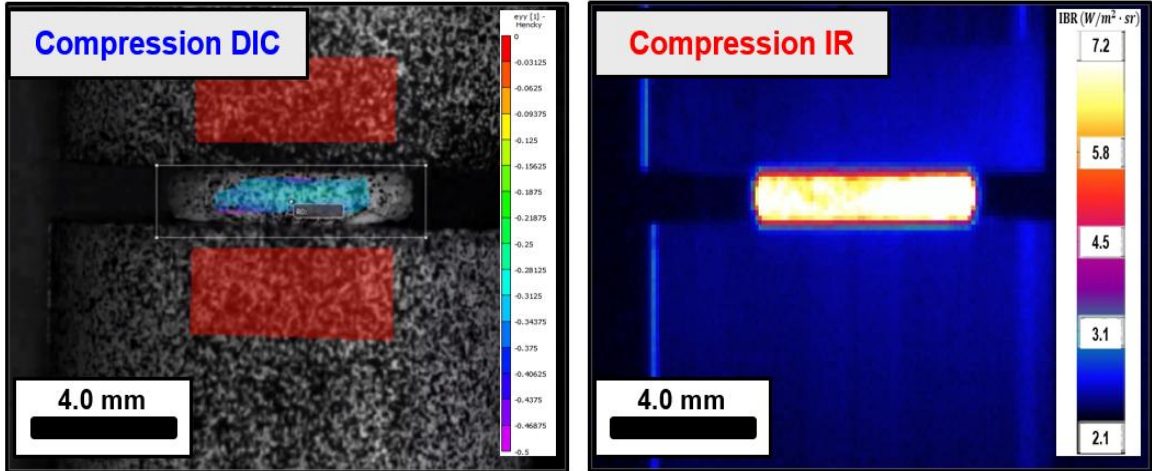


Figure 2-13 Representative DIC and IR images obtained during quasi-static and intermediate compression tests

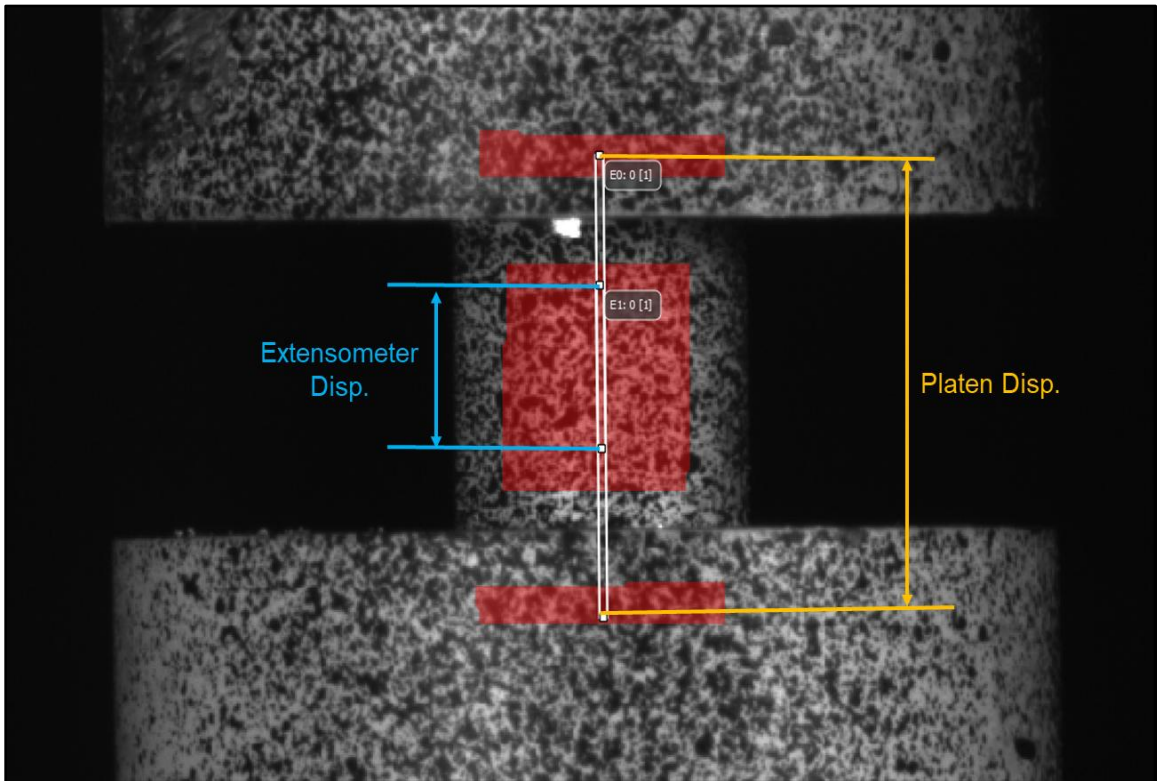


Figure 2-14 Definition of virtual extensometer and platen displacement measurements in quasi-static and intermediate compression tests

2.4.2 Dynamic Compression Tests

High-strain-rate compression tests (strain rates of 2000 to 5000 s⁻¹) are conducted using a Split-Hopkinson Pressure Bar (SHPB) apparatus, as illustrated in the schematic in Figure 2-15. The system is comprised of three 0.5 in. diameter Ti-6Al-4V titanium alloy bars: a striker bar, an incident bar (6 ft. long), and a transmitter bar (6 ft. long). In contrast to the tension setup, the compressive loading wave is generated by firing the striker bar into the end of the incident bar using a high-pressure gas gun. To minimize friction and subsequent "barreling" deformation, the contact surfaces between the cylindrical specimen and the bars are lubricated. The fundamental principles and detailed techniques for classic SHPB testing are described by Gray [40].

The experimental setup is shown in Figure 2-16. Due to laboratory space constraints, the DIC and IR cameras are positioned at a 90 degree angle relative to each other rather than opposite to each other (Figure 2-16). This configuration allowed for simultaneous full-field strain and temperature data collection on the specimen surface while maintaining the necessary focal lengths for both high-speed systems (Figure 2-17). The waves during a test are measured at the center of both the incident bar (Gage *A*) and transmitter bar (Gage *B*) using four 1000 Ohm strain gages connected in a full Wheatstone bridge circuit to isolate bending load signals. The specific data acquisition and camera parameters for the dynamic compression series are detailed in Table 2-6.

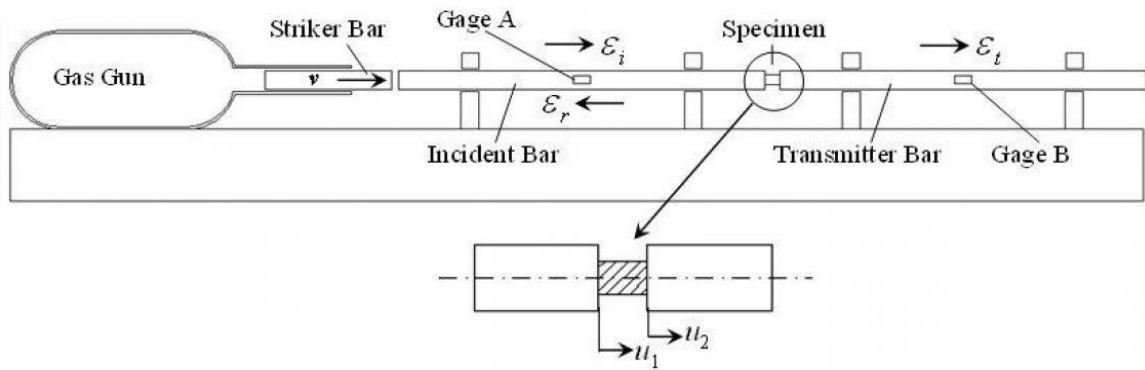


Figure 2-15 Schematic of the compressive Split-Hopkinson Bar (SHB) apparatus

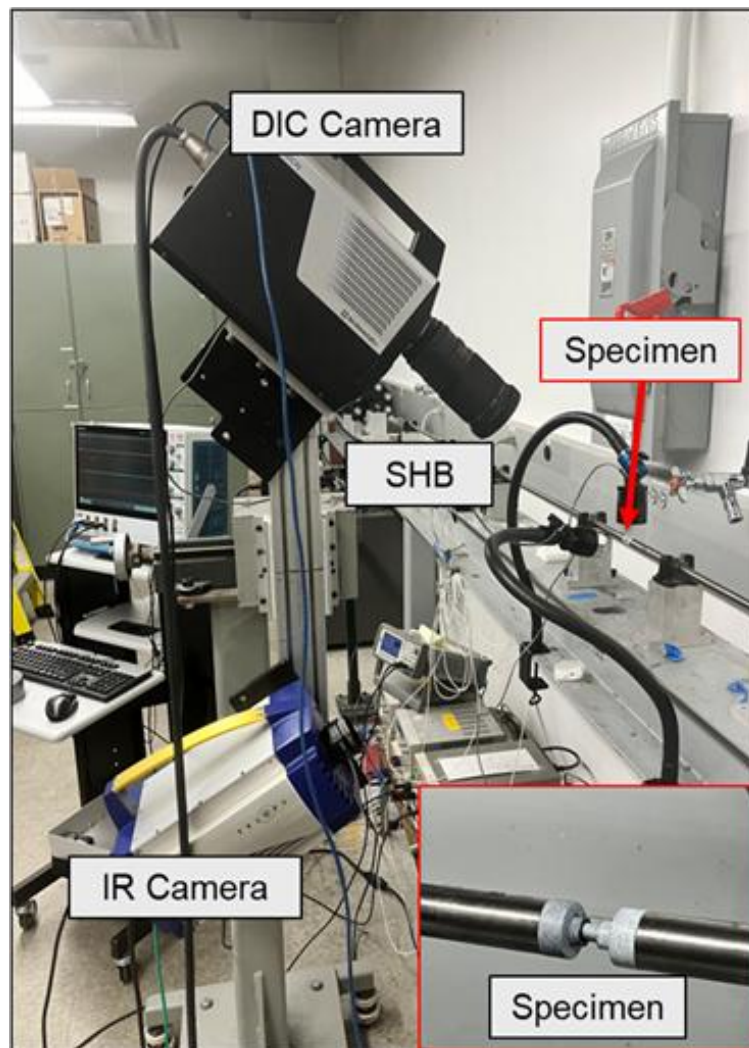


Figure 2-16 Experimental setup for dynamic compression tests

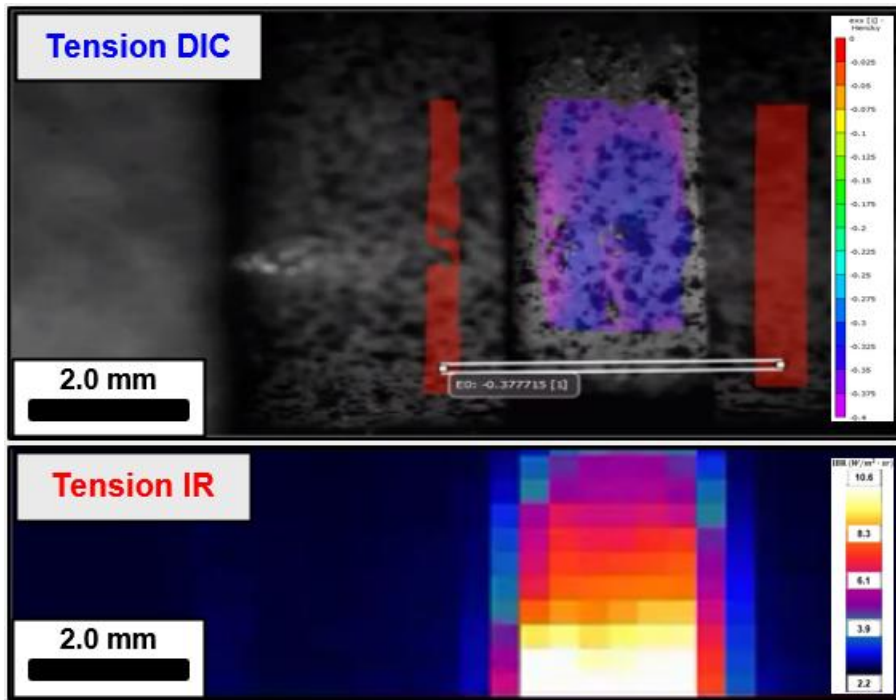


Figure 2-17 Representative DIC and IR images obtained during dynamic compression tests

2.4.3 Elevated Temperature Tests

To evaluate the temperature dependency and thermal softening effects of the 410 stainless steel in compression, specimens are tested at target temperatures of 200°C, 400°C, 600°C, and 800°C. The experimental setup, shown in Figure 2-9, utilized the same furnace and servo-hydraulic load frame configuration as the elevated temperature tension tests. To protect the load cell and maintain machine stability, a water-cooling system is employed to prevent heat transfer from the furnace to the load frame components.

Due to the small dimensions of the compressive specimens, attaching the thermocouples directly to the gauge section is not feasible without interfering with the deformation. Therefore, thermal control is used by cementing K-type thermocouples to

both the top and bottom loading platens, as shown in Figure 2-18. These thermocouples are connected to a temperature data logger for real-time monitoring to ensure the specimen reached thermal equilibrium. The specific camera parameters and apparatus for these tests are detailed in Table 2-7.

As with the tensile experiments, infrared (IR) thermography is not utilized for the elevated temperature compression tests. The furnace environment and the high-intensity radiation emitted by the specimen and platens at these temperatures. The stress-strain curves are calculated using the procedures established for the quasi-static tests in Section 2.4.1.

Table 2-7 Camera parameters and experimental apparatus for elevated-temperature compression tests

Test mode	Temperature (°C)	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate
Compression	RT	1.00E-03	Servo-hydraulic load frame with furnace	Two Point Grey Gazelles with 35 mm lenses	2
	200				
	400			Point Grey Gazelles with Tokina 100 mm lenses	0.5
	600				
	800				

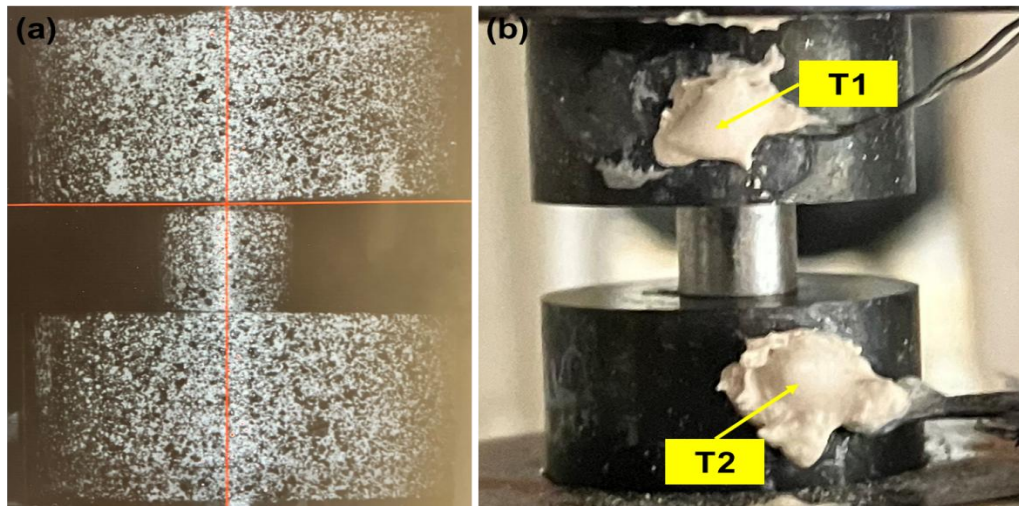


Figure 2-18 Thermocouple installation for elevated-temperature compression tests: (a)Front view (b)Back view

2.4.4 Evaluation of Barreling

During cylindrical compression testing, friction at the specimen-platen interfaces leads to non-uniform deformation, referred to as barreling. Even though lubricants are employed to minimize interface friction, the constraint at the contact surfaces still leads to a complex, triaxial stress state that deviates from the ideal uniaxial condition, potentially resulting in an overestimation of the flow stress. To ensure the accuracy of the constitutive modeling for 410 stainless steel, the barreling effect is quantified and corrected.

For quasi-static and intermediate strain rate tests, an analytical model proposed by Ebrahimi and Najafizadeh [41] is employed. A significant advantage of this method is that it allows for stress correction based solely on the instantaneous change in specimen shape. The geometric barreling factor, b , is determined by comparing the measured maximum radius at the mid-section (R_{max}) to the theoretical average radius (R_{th}) derived from the assumption of volume constancy ($R_{th} = R_0\sqrt{H_0/H_i}$). This relationship is illustrated in Figure 2-19, which defines the parameters for the deformed specimen. The barreling factor is calculated as:

$$b_i = 4 \cdot \frac{R_{max,i} - R_{th,i}}{R_{th,i}} \cdot \frac{H_i}{H_0 - H_i} \quad 2-8$$

Using this factor, the constant friction coefficient, m , is derived based on the instantaneous specimen dimensions and the degree of barreling:

$$m_i = \frac{b_i \cdot \left(\frac{R_{max,i}}{H_i}\right)}{\frac{4}{\sqrt{3}} - \frac{2b_i}{3\sqrt{3}}} \quad 2-9$$

Finally, the corrected uniaxial flow stress, σ_{corr} , is calculated from the experimental stress, σ_{exp} , using the following relationship. The effectiveness of the analytical correction is evaluated by comparing the experimental and corrected stress–strain curves (Figure 3-4):

$$\frac{\sigma_{exp}}{\sigma_{corr}} = 8b_i \frac{R_{max,i}}{H_i} \left\{ \left[\frac{1}{12} + \left(\frac{H_i}{R_{max,i}} \right)^2 \frac{1}{b_i^2} \right]^{3/2} - \left(\frac{H_i}{R_{max,i}} \right)^3 \frac{1}{b_i^3} - \frac{m_i}{24\sqrt{3}} \frac{e^{-b_i/2}}{e^{-b_i/2} - 1} \right\} \quad 2-10$$

In dynamic SHPB testing, 2D-DIC measurements are often insufficient to capture with high-precision the value of R_{max} required for the analytical solution. Consequently, a reverse-engineering approach is implemented. First, the stress solution for a plate forged in plane strain [42] is considered:

$$p = \sigma_0 \exp \left[\frac{2\mu}{h} (a - x) \right] \quad 2-11$$

This solution is modified (see Appendix A) to fit the specific boundary conditions of cylindrical compression:

$$\mu = \left(\frac{\sigma_{exp}}{\sigma_{corr}} - 1 \right) \cdot \frac{3H}{2R} \quad 2-12$$

Equation 2-12 is applied to the 1.0 s^{-1} strain rate tests to calibrate the friction coefficient μ and construct a validated finite element simulation model. The accuracy of this model is ensured by verifying the force-displacement curves and the visual barreling condition.

The validated simulation model is used to replicate the dynamic SHPB compression tests. By simulating these high-rate testing, the R_{max} , which could not be directly measured with high accuracy via 2D-DIC, is extracted from the numerical results. To ensure the accuracy of the dynamic simulation, the force-displacement history, temperature rise, and

visual barreling profiles are cross-verified against experimental observations. Once the simulation is confirmed to accurately represent the experiments, the resulting geometric data are used in the analytical model (Equations 2-9 to 2-11) to generate the final, corrected dynamic stress-strain curves. The overall procedure used to determine the friction condition and apply the analytical stress correction model to both low and high strain rate compression tests is summarized in Figure 2-20.

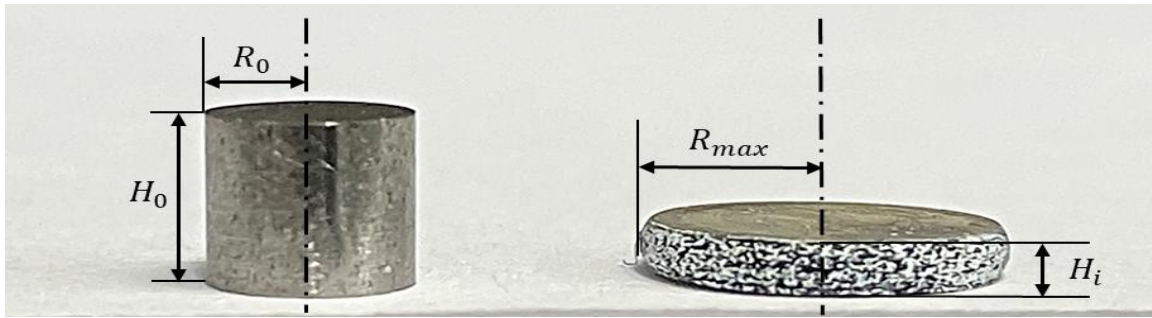


Figure 2-19 Geometric parameters of undeformed and barreled compression specimens used for barreling correction

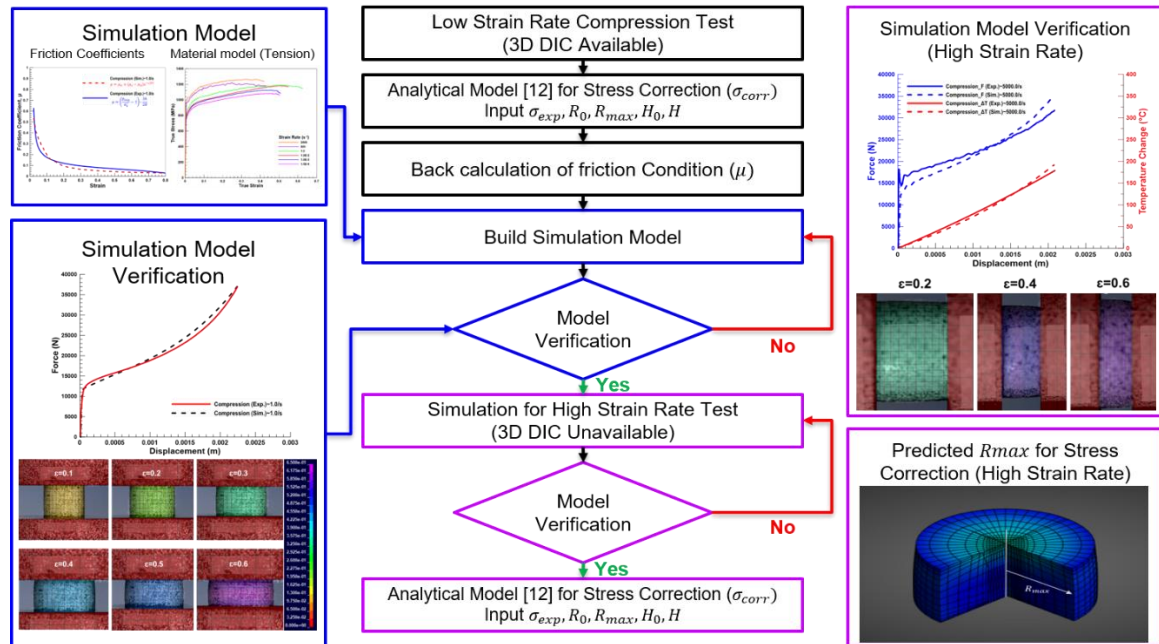


Figure 2-20 Workflow for stress correction in compression tests using the analytical barreling model

2.5 Experimental Techniques of Torsion Tests

The torsion testing program is conducted across multiple strain-rate regimes to evaluate the shear properties behavior of the 410 stainless steel. The integration of Digital Image Correlation (DIC) and Infrared (IR) thermography is maintained throughout the torsion testing to ensure the mapping of the shear strain and thermal evolution. The specific data acquisition parameters, including camera types, frame rates, and the apparatus for each torsion test, are detailed in Table 2-8.

Table 2-8 Camera parameters and experimental apparatus for torsion tests

Test mode	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate	IR Camera	IR Frame Rate
Torsion	1.00E-04	Servo-hydraulic load frame	Two Point Grey Gazelles with 35 mm lenses	0.1	Telops MS-IR	0.1
	1.00E-02			5	Telops MS-IR	5
	1		Two Photron MC2.1 100 with 50 mm lens	500	Telops MS-IR	500
	1000	Torsion SHB	Two Photron SA1.1 with 180 mm lens	150000	Telops MS-IR	82000
	4500			270000	Telops MS-IR	82000

2.5.1 Quasi-static and Intermediate Torsion Tests

Quasi-static tests (10^{-4} to 10^{-2} s⁻¹) and intermediate tests (1.0 s⁻¹) are performed using an Instron servo-hydraulic torsion frame. The general experimental configuration remains identical to the setup used for tensile testing (Figure 2-4); however, a specialized set-screw fixture was integrated to accommodate the rotational requirements of torsion, as shown in Figure 2-21. The torsion specimens feature hexagonal flanges to facilitate secure gripping, though the gauge section dimensions vary by strain rate regime. Specifically, the specimens for quasi-static and intermediate tests utilize a 0.08-inch gauge thickness, while specimens for dynamic tests are reduced to 0.04-inch to achieve the target high strain rates. To simultaneously map the mechanical deformation and thermal response, DIC and IR cameras are positioned facing the specimen gauge section. Typical images are shown in

Figure 2-22. To ensure synchronization between the strain field and the thermal evolution, the frame rates for both the DIC and IR cameras are set to identical values.

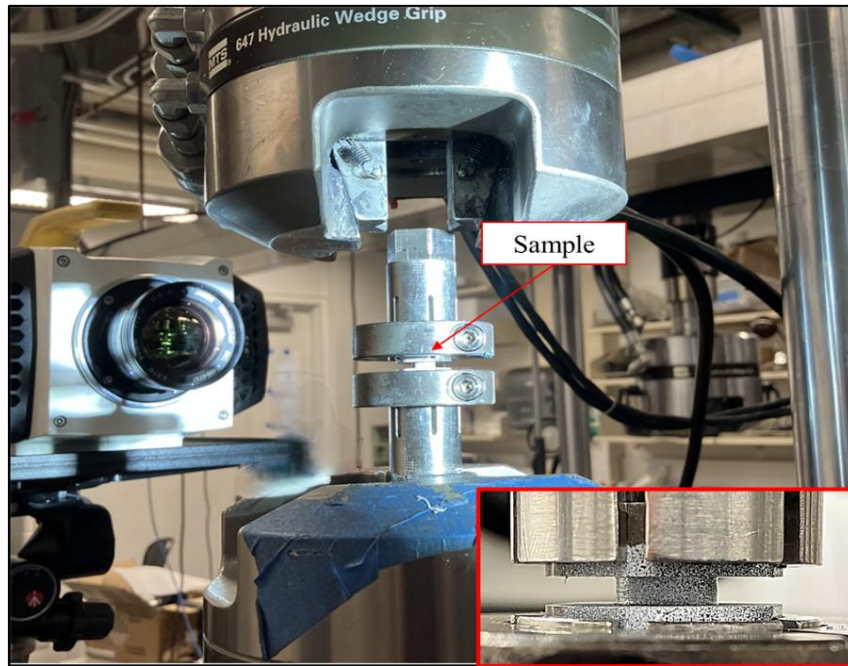


Figure 2-21 Experimental setup for quasi-static and intermediate torsion tests

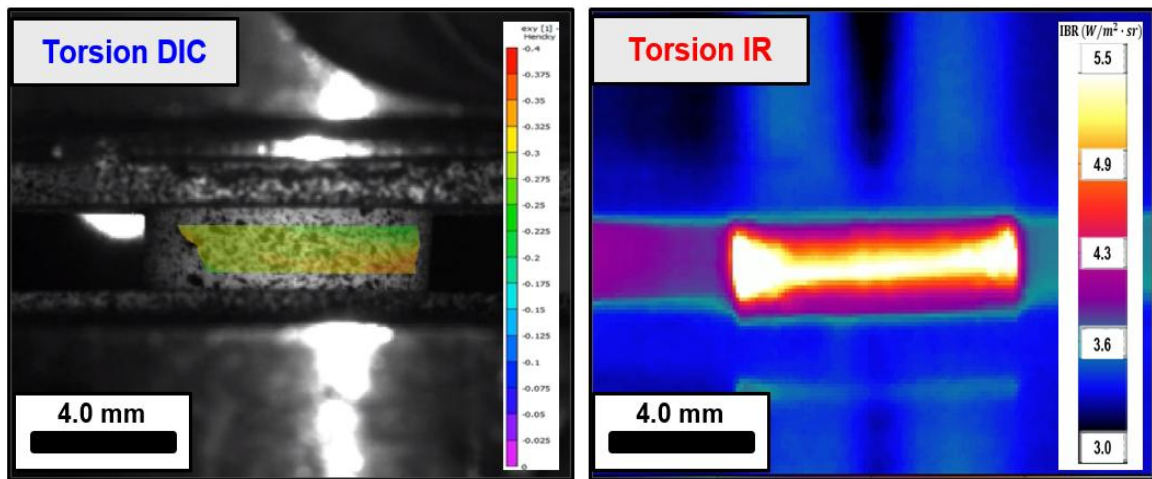


Figure 2-22 Representative DIC and IR images obtained during quasi-static and intermediate torsion tests

The reduction of raw torsion data into stress-strain curves is performed using the thin-wall approximation. The engineering shear stress (τ) is derived from the torque (T)

recorded by the Instron load frame and the specific geometry of each specimen, as measured prior to testing:

$$\tau = \frac{T}{2\pi R_m t} \quad 2-13$$

Where R_m is the mean radius and t is the wall thickness.

To ensure the accuracy of the deformation measurements, the engineering shear strain (γ_s) is calculated using two methods. While the machine-recorded angular displacement (θ_m) provides a secondary reference, the primary global strain is derived via DIC by tracking the motion of the gauge section boundaries of specimen. The rotation angle (θ_{DIC}) is determined by resolving the horizontal (u) and out-of-plane (w) displacement components of the tracked surface points. For a point at a radial distance R_0 , the angular position φ is calculated as:

$$\varphi = \arccos\left(1 - \frac{u^2 + w^2}{2R_0^2}\right) \quad 2-14$$

The total angle of twist, θ , is defined as the relative difference in rotation between the top and bottom boundaries ($\varphi_{top} - \varphi_{bot}$). The global engineering shear strain is then calculated as:

$$\gamma_s = \frac{R_m \cdot \theta}{L_g} \quad 2-15$$

where L_g represents the initial gauge length. Calculating the angle through these displacement components effectively bypasses errors introduced by fixture compliance or grip slippage.

To facilitate a direct comparison between the torsional response and the results obtained from tensile and compressive loading modes, the shear data is transformed into a von Mises equivalent framework. The effective stress ($\bar{\sigma}$) and equivalent strain ($\bar{\epsilon}$) are calculated using the following relationships:

$$\bar{\sigma} = \tau\sqrt{3} \quad 2-16$$

$$\bar{\epsilon} = \frac{\gamma_s}{\sqrt{3}} \quad 2-17$$

By converting the data, the flow behavior of the 410 stainless steel can be evaluated consistently across all three stress states.

2.5.2 Dynamic Torsion Tests

High-strain-rate torsion tests (800 to 4500 s⁻¹) are conducted using a Torsion Split-Hopkinson Bar, as illustrated in the schematic in Figure 2-23. The system is comprised of two 0.875-inch diameter 7075-T6 aluminum bars: an incident bar (12 ft. long) and a transmitter bar (7 ft. long). The torsional loading wave is generated by the release of a pre-stored torque. This torque is initially clamped at a specific location on the incident bar and released rapidly to produce a sharp shear wave that travels toward the specimen. The fundamental principles and detailed analytical techniques for Torsion SHB testing are described by Gilat [43].

The experimental setup is shown in Figure 2-24. Similar to the dynamic tension, the DIC and IR cameras are positioned to allow for simultaneous full-field strain and temperature data on the specimen surface (Figure 2-25). The high-speed camera configuration ensured that the localized thermal and mechanical responses could be

captured within the microsecond timescales of the dynamic test. The torque waves are measured using strain gages at three locations (Gage *A*, *B*, and *C*) using a full Wheatstone bridge to ensure only the torque signals are recorded. These gages are configured in a full Wheatstone bridge circuit to isolate pure torque signals and eliminate bending or axial data.

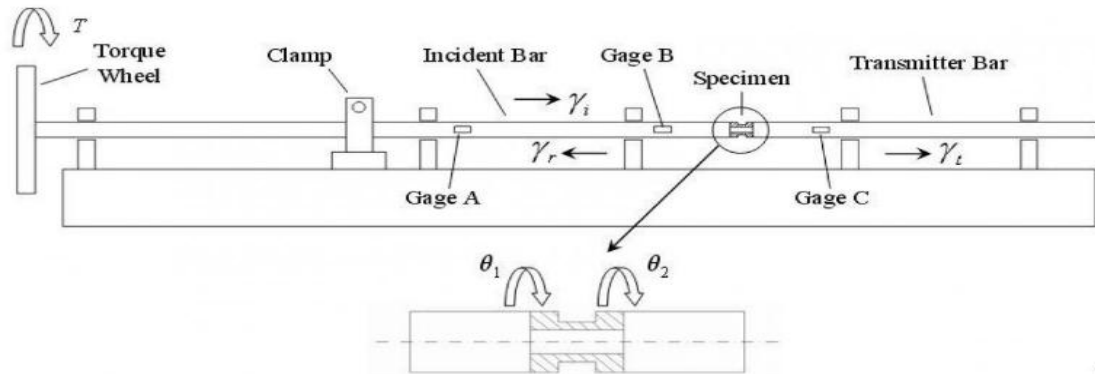


Figure 2-23 Schematic of the torsional Split-Hopkinson Bar (SHB) apparatus

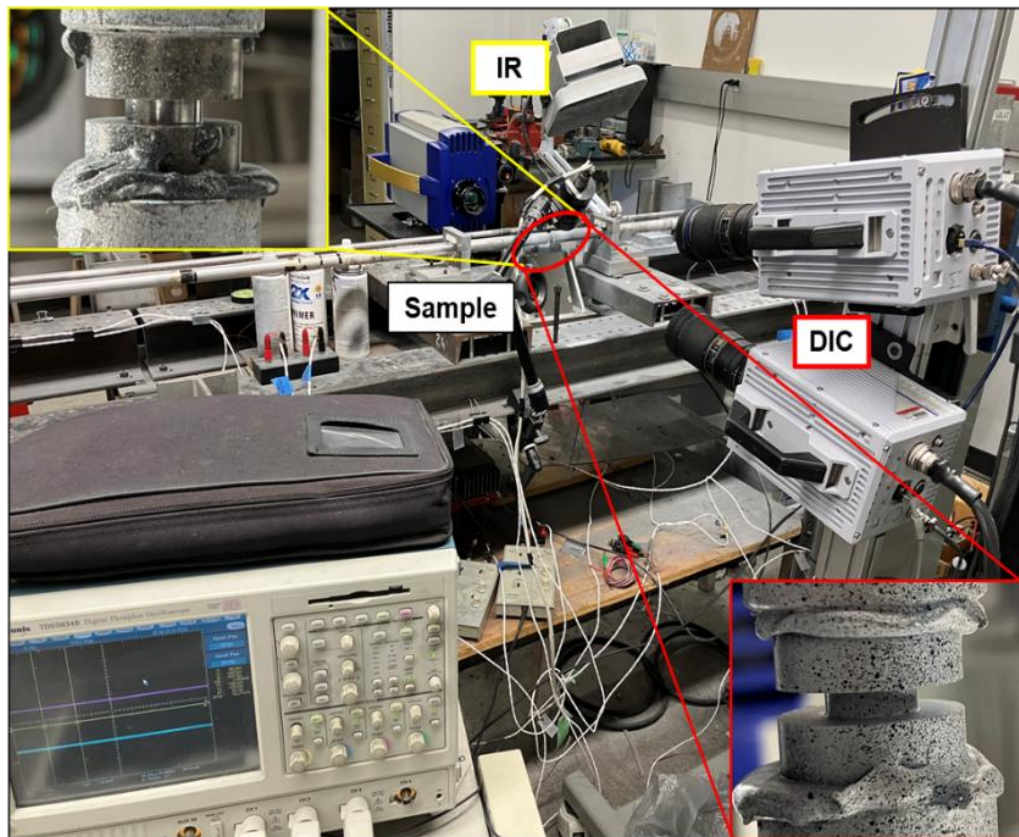


Figure 2-24 Experimental setup for dynamic torsion tests

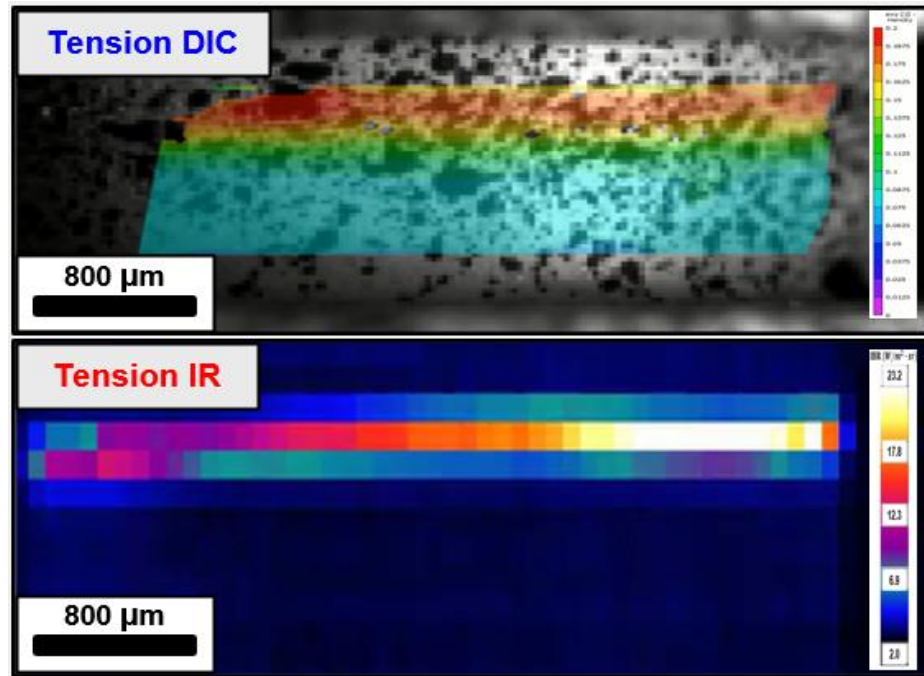


Figure 2-25 Representative DIC and IR images obtained during dynamic torsion tests

The data is processed using the same equations for effective stress and equivalent strain as described in Section 2.5.1. In this dynamic torsion test, the shear strain is calculated using both the bar wave data and the DIC displacement data. Because the wave signals reflect the movement of the entire system while the DIC tracks the specimen directly, comparing these two measurements allows for an evaluation of the difference between the system's recorded motion and the actual material deformation. The specific camera and test settings are listed in Table 2-8.

2.5.3 Elevated Temperature Torsion Tests

To investigate the thermal softening behavior and temperature dependency of 410 stainless steel under shear loading, torsion tests are conducted at target temperatures of 200°C, 400°C, 600°C, and 800°C. The experimental setup utilized the same furnace and

servo-hydraulic load frame configuration as the elevated temperature tensile tests and compression tests. To protect the mechanical components of the Instron frame, Inconel 718 adapter rods are used to connect the load cell and actuator to the specimen inside the furnace, and a water-cooling system is employed to prevent heat transfer to the load frame. As with the tension and compression experiments, IR thermography is not utilized due to high-intensity radiation from the furnace.

Thermal control is achieved by cementing two K-type thermocouples to the top and bottom shoulders of the specimen, as shown in Figure 2-26. Due to the geometry of the torsion specimens and the potential for interfering with the shear deformation, direct attachment to the thin-walled gauge section is avoided. These thermocouples are connected to a temperature logger for real-time monitoring to ensure the specimen reached thermal equilibrium. The measured values from both thermocouples are averaged to determine the test temperature. The specific camera parameters and apparatus for these tests are detailed in Table 2-9.

Due to the limited width of the viewing window and the size of the specimen fixtures, direct DIC measurement of the gauge section is not possible. Following the method established by Ressa [38], rotation is measured instead using L-shaped bent tabs attached to the inside of the upper and lower specimen flanges (Figure 2-26). These tabs extend 15 mm from the gauge section for DIC tracking. By monitoring the rotation of specific tracking points on these rigid tabs, the relative angle of twist is determined. The gauge length (L_g) and mean radius (R_m) are adjusted in the calculations to account for the geometry of the tab-tracking setup.

Table 2-9 Camera parameters and experimental apparatus for elevated-temperature torsion tests

Test mode	Temperature (°C)	Strain Rate (1/s)	Apparatus	DIC Camera	DIC Frame Rate
Torsion	RT	1.00E-03	Servo-hydraulic load frame with furnace	Two Point Grey Gazelles with 35 mm lenses	1
	200				
	400			Two Point Grey Gazelles with Tokina 100 mm lenses	
	600				
	800				

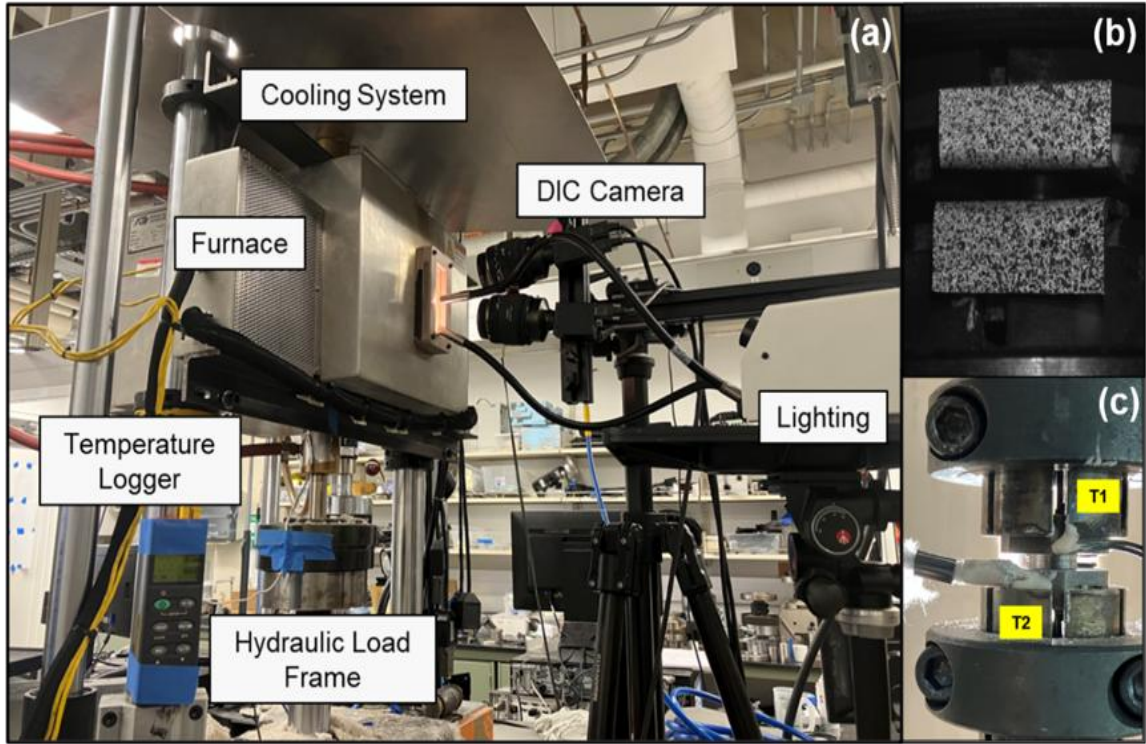


Figure 2-26 (a) Experimental setup for elevated-temperature torsion tests at a strain rate of 10^{-3} s^{-1} and Thermocouple setup (b)Front view (c)Back view

2.6 Taylor-Quinney Coefficient

This section details the methodology used to determine the Taylor-Quinney coefficient (β), which defines the proportion of plastic work converted into thermal energy, while the remaining fraction is termed the stored energy of cold work. This stored energy is fundamentally linked to the evolution of the material's microstructure, specifically the

multiplication of dislocations and the creation of lattice defects. The process is divided into three parts: the mathematical formulation, the numerical synchronization of thermal and mechanical data, and the validation procedures used to ensure the accuracy of the adiabatic assumption.

2.6.1 Governing Equation

To maintain consistency across loading modes, the equivalent plastic strain ($\bar{\varepsilon}_p$) is derived by subtracting the elastic strain from the total equivalent strain ($\bar{\varepsilon}$):

$$\bar{\varepsilon}_p = \bar{\varepsilon} - \frac{\bar{\sigma}}{E} \quad 2-18$$

The integral form of the coefficient, β_{int} , is then determined by the ratio of the generated heat energy to the cumulative plastic work:

$$\beta_{int} = \frac{\rho C_p \Delta T}{\int_0^{\bar{\varepsilon}_p} \bar{\sigma} d\bar{\varepsilon}_p} \quad 2-19$$

where ρ is the density and C_p is the specific heat capacity, and ΔT is the adiabatic temperature rise. This formulation provides a direct way to track how much mechanical energy is being dissipated as heat. For 410 stainless steel, the value of E is 200 GPa, ρ is 7800 kg/m³ and C_p is 0.46 kJ/kg · °C .

2.6.2 Data Preparation

A primary challenge in TQC calculation is the sensitivity of the β_{int} value to noise in the experimental signals. While the plastic work is an integral and therefore acts as a natural averaging filter for noise in the stress and strain data, the thermal energy calculation relies on the instantaneous temperature change (ΔT). Consequently, even minor high-frequency

noise in the infrared signal can lead to significant fluctuations in the calculated TQC. To address this, the temperature change is curve-fitted by the lowest-order polynomial function of the equivalent plastic strain that represented the experimental data. Representative fitting results is shown in Figure 2-27.

The analytical window for the TQC calculation is defined by three distinct physical markers, the starting point, the transition point, and the end point. which vary according to the loading mode. The starting point is established immediately following the thermoelastic effect, where the material transitions from initial elastic cooling to plastic heating. This ensures that the energy balance accounts for plastic work. As deformation progresses, a transition point is identified to mark the onset of significant deformation localization. As illustrated in Figure 2-28, Figure 2-29, and Figure 2-30, this divergence represents the transition from uniform deformation to a localized state, such as necking in tension, barreling in compression, or non-uniform shear in torsion. The end point is set at the moment of material fracture or specimen failure, where the thermal-mechanical synchronization is terminated.

In conventional research, the TQC calculation is typically terminated at the transition point because global measurements of stress and strain no longer represent the material behavior once non-uniform deformation begins. However, in this study, the high-resolution local data provided by synchronized DIC and IR thermography allows the calculation to be extended to the end point, which is defined as the moment of material fracture or test ended. Because the local temperature and local strain can be tracked at the site of failure, a valid energy balance is maintained throughout the localization phase.

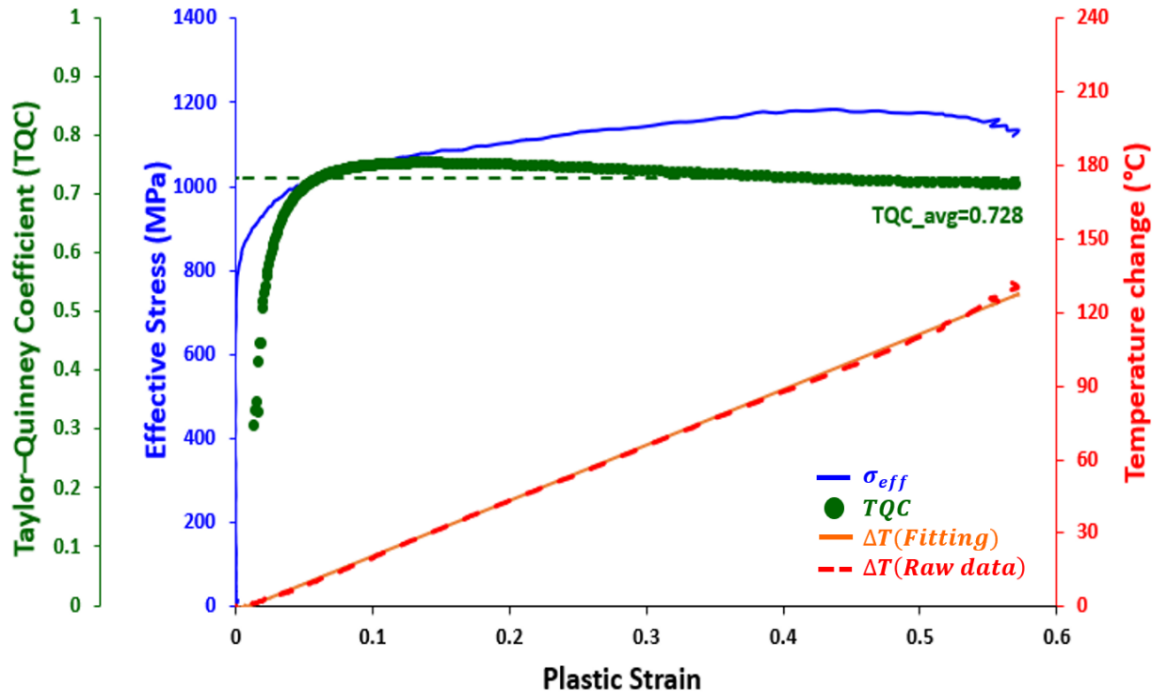


Figure 2-27 Representative polynomial fitting of temperature rise for Taylor–Quinney coefficient calculation

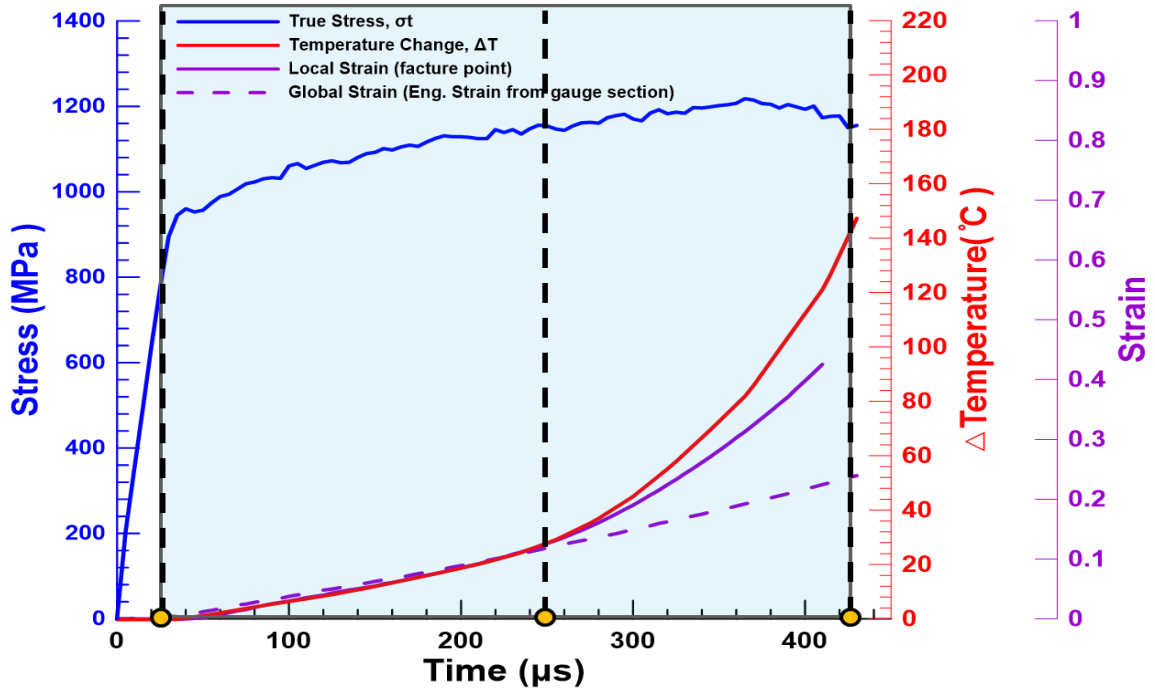


Figure 2-28 Analytical window for Taylor–Quinney coefficient calculation in a dynamic tensile SHPB test at a strain rate of 500 s^{-1}

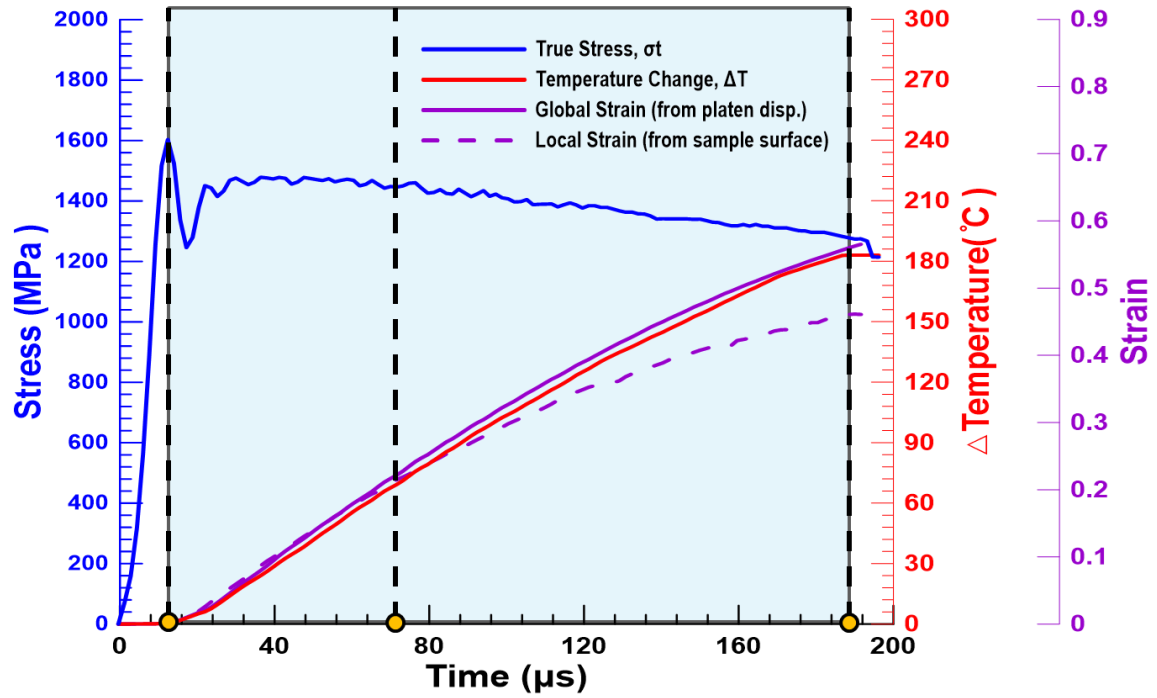


Figure 2-29 Analytical window for Taylor–Quinney coefficient calculation in a dynamic compression SHB test at a strain rate of 5000 s^{-1}

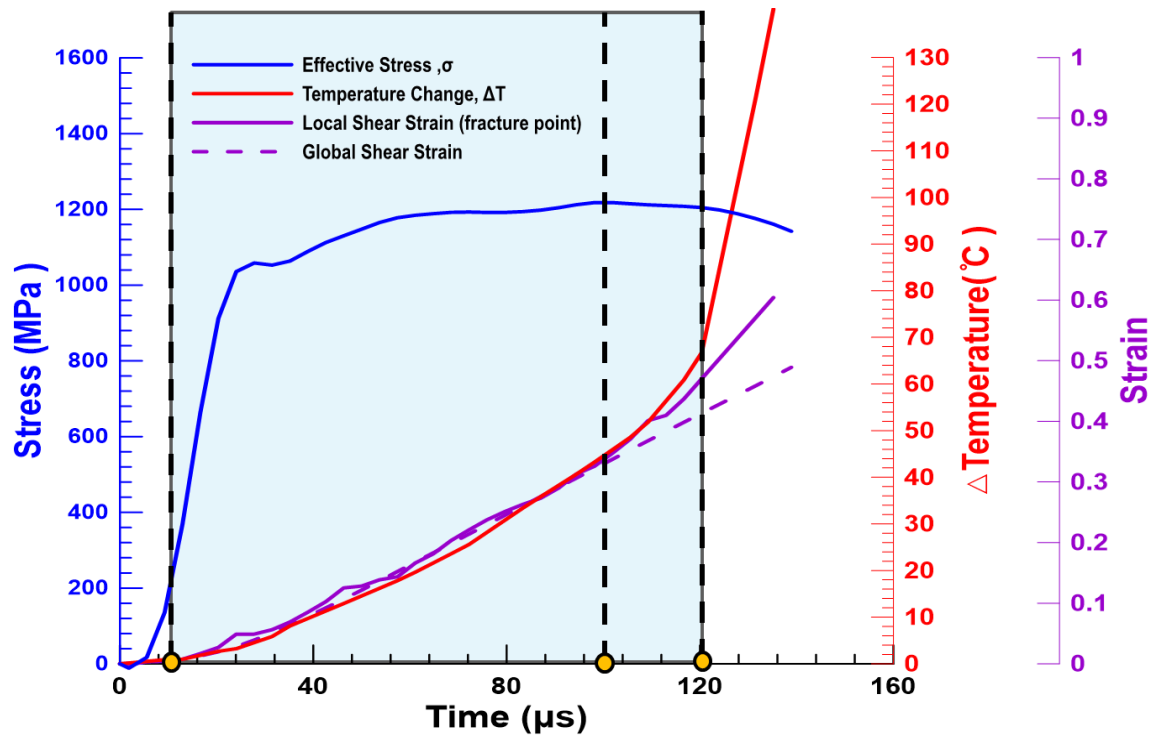


Figure 2-30 Analytical window for Taylor–Quinney coefficient calculation in a dynamic torsion SHB test at a shear strain rate of 4500 s^{-1}

2.6.3 Determination of Valid Analytical Region

A critical requirement for an accurate TQC is that the temperature rise must be adiabatic. Because the coefficient is a ratio of the energy converted into heat versus the total mechanical work performed on the material. If the process is not adiabatic, the thermal data becomes affected by external factors, leading to an incorrect energy balance. In this study, the degree of adiabaticity is quantified using three methods: low-strain rate reference, Fourier number, and temperature-strain distribution. These metrics are used to confirm that the deformation occurs fast enough to be considered adiabatic, ensuring that conductive heat loss to the grips or environment does not artificially lower the measured temperature rise.

The low strain-rate tests in this study are designed to fulfill two primary objectives. First, these tests establish a framework to evaluate the strain-rate dependency of the material's mechanical behavior. Second, because low-rate testing is inherently non-adiabatic, these data sets provide a conductive reference to observe the characteristic of the TQC curve.

To ensure the validity of the TQC calculation, the Fourier number (F_o) is utilized to quantify the thermal diffusion during the test. The Fourier number is a dimensionless ratio of the rate of heat conduction to the rate of thermal energy storage, defined as:

$$F_{o,i} = \frac{\alpha \cdot t_i}{L_i^2} \quad 2-20$$

where α is the thermal diffusivity of the 410 stainless steel, t is the characteristic time of deformation, and L is the characteristic length. The value of α is $7.5 \times 10^{-6} m^2/s$.

For intermediate and high strain-rate tests, a low Fourier number (F_o) serves as the indicator of adiabatic conditions. An F_o significantly greater than 1.0 indicates that there is sufficient time for the specimen to reach a steady state or experience substantial heat loss. When F_o is close to 1.0, the temperature at the boundaries is affected by the heat source at the center, signaling the onset of conductive loss. Conversely, an F_o much smaller than 1.0 implies that the deformation speed is fast enough to confine the temperature change to the heated center. The calculated Fourier numbers for all loading modes and strain rates are presented and discussed in 3.4 Taylor-Quinney Coefficient Dependency.

The temperature-strain distribution analysis is employed as an additional approach to verify the adiabatic state of each test. This analysis provides a direct experimental measurement alongside the Fourier number. By monitoring the relationship between local temperature and local strain across the gauge section, the validity of the TQC can be confirmed; if the temperature-to-strain ratio remains uniform, the test is verified as adiabatic. Conversely, if the local temperature begins to drop relative to the strain, it identifies the exact moment when conductive heat loss begins to interfere with the TQC calculation.

Chapter 3 Results and Discussion

This chapter provides a comprehensive analysis of the mechanical behavior of 410 Stainless Steel, structured to transition from experimental observations to a predictive constitutive model. The discussion begins in Section 3.1 and 3.2 by evaluating the material's sensitivity to strain rate and temperature across tension, compression, and torsion loading modes, including a dedicated barreling analysis to ensure the integrity of compression data. Section 3.3 then examines the influence of anisotropy to determine how directional properties affect these mechanical responses. Following the characterization of stress-strain behavior, Section 3.4 investigates the Taylor-Quinney coefficient to quantify the conversion of plastic work into thermal energy. The chapter concludes in Section 3.5 with the development of a constitutive material model, synthesizing the experimental data into a calibrated mathematical tool capable of simulating plastic deformation across diverse loading rates and temperatures.

3.1 Strain Rate Dependency

This section characterizes the strain rate sensitivity of the material across a wide experimental range from quasi-static conditions at 10^{-4} s^{-1} to high-rate dynamic loading reaching 5000 s^{-1} . As summarized in Table 2-2, this investigation utilizes distinct specimen geometries optimized for each loading mode: thin-sheet dog-bone samples for tension, cylindrical specimens for compression, and circular thin-walled tubes for torsion.

3.1.1 Tension

The tensile response of 410 Stainless Steel across the strain rate range is presented in Figure 3-1, illustrating a clear strain rate sensitivity where flow stress increases with

increasing strain rate. To ensure statistical reliability, each condition is tested on three samples for repeatability. The standard deviations at yield stress and ultimate tensile strength (UTS) shows consistently within 2%. A single representative curve is displayed for each strain rate. Inspection of the specimens after testing (Figure 3-2) reveals a moderately ductile fracture.

The transition from isothermal to adiabatic conditions is captured by the temperature-strain history shown in Figure 3-3. At the quasi-static rate of 10^{-4} s^{-1} , the process is isothermal. At strain rate of 10^{-3} s^{-1} , the material remains nearly isothermal during plastic deformation, with a sharp temperature spike occurring only at the instant of fracture. As the strain rate increases to 10^{-2} s^{-1} , the specimen begins to exhibit a slight, measurable temperature increase during the deformation process, showing the early onset of heat accumulation compared to the lower strain rates.

In the dynamic test at strain rates of 1.0 s^{-1} and 500 s^{-1} , the temperature-strain relationship becomes near-linear throughout the plastic deformation, displaying adiabatic heating where plastic work is converted to thermal energy. At the highest strain rate of 2000 s^{-1} , the temperature-strain curve maintains this near-linear trend; however, the temperature data ends at a strain of 0.32 while the strain data continues to 0.4. This difference is due to the hardware limits of the cameras: the DIC camera records at 500,000 fps, while the IR camera is much slower at 82,000 fps. Because the metal fractured so quickly, the final part of the deformation happened in the gap between two IR frames.

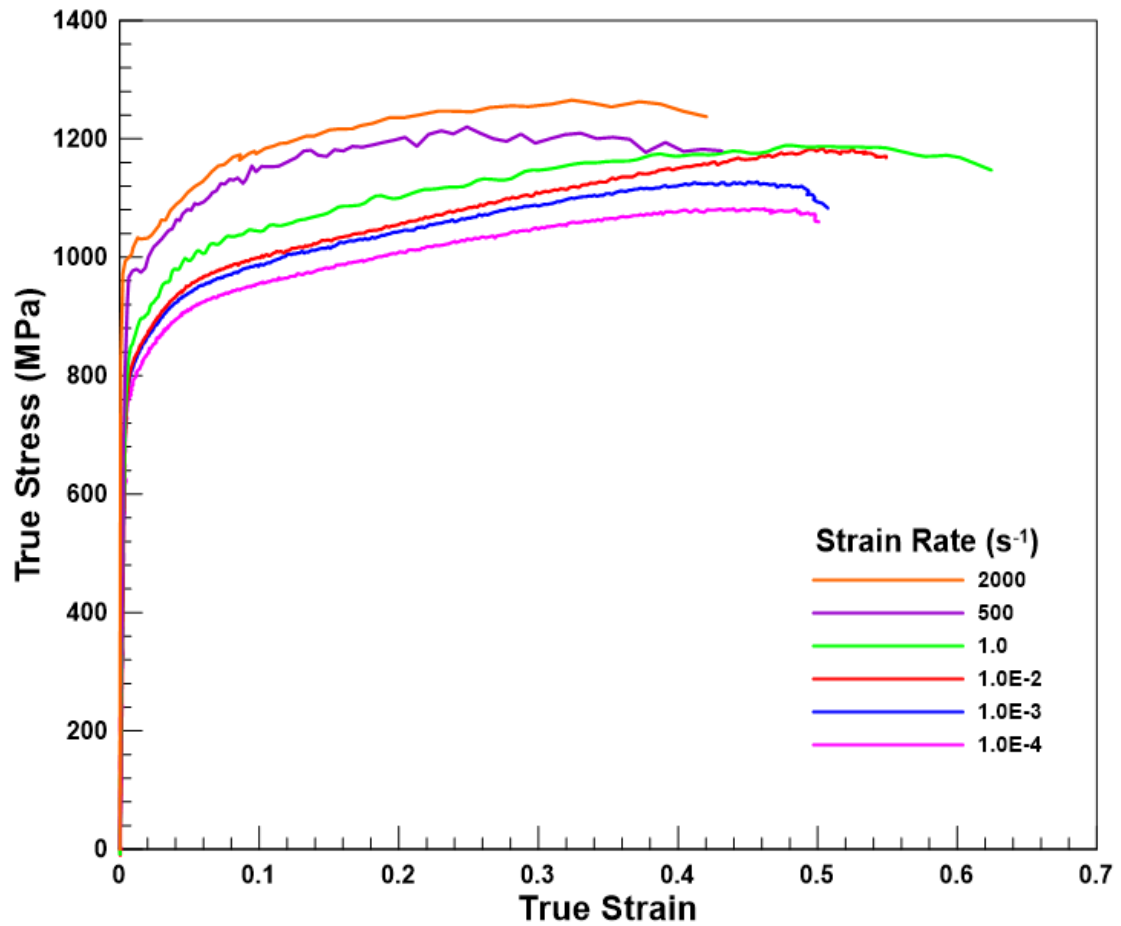


Figure 3-1 True Stress-strain curves from tension tests at various strain rates

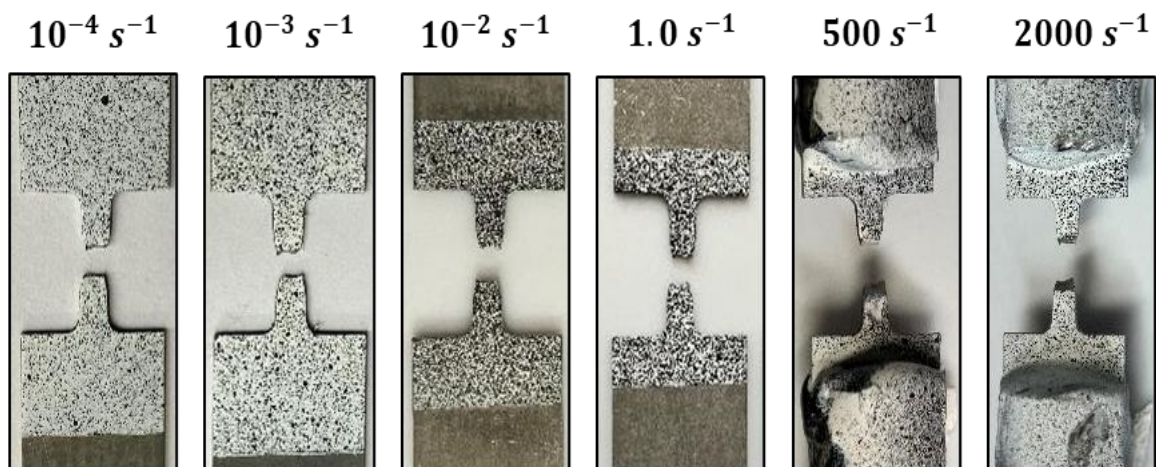


Figure 3-2 Specimens after tensile testing at various strain rates

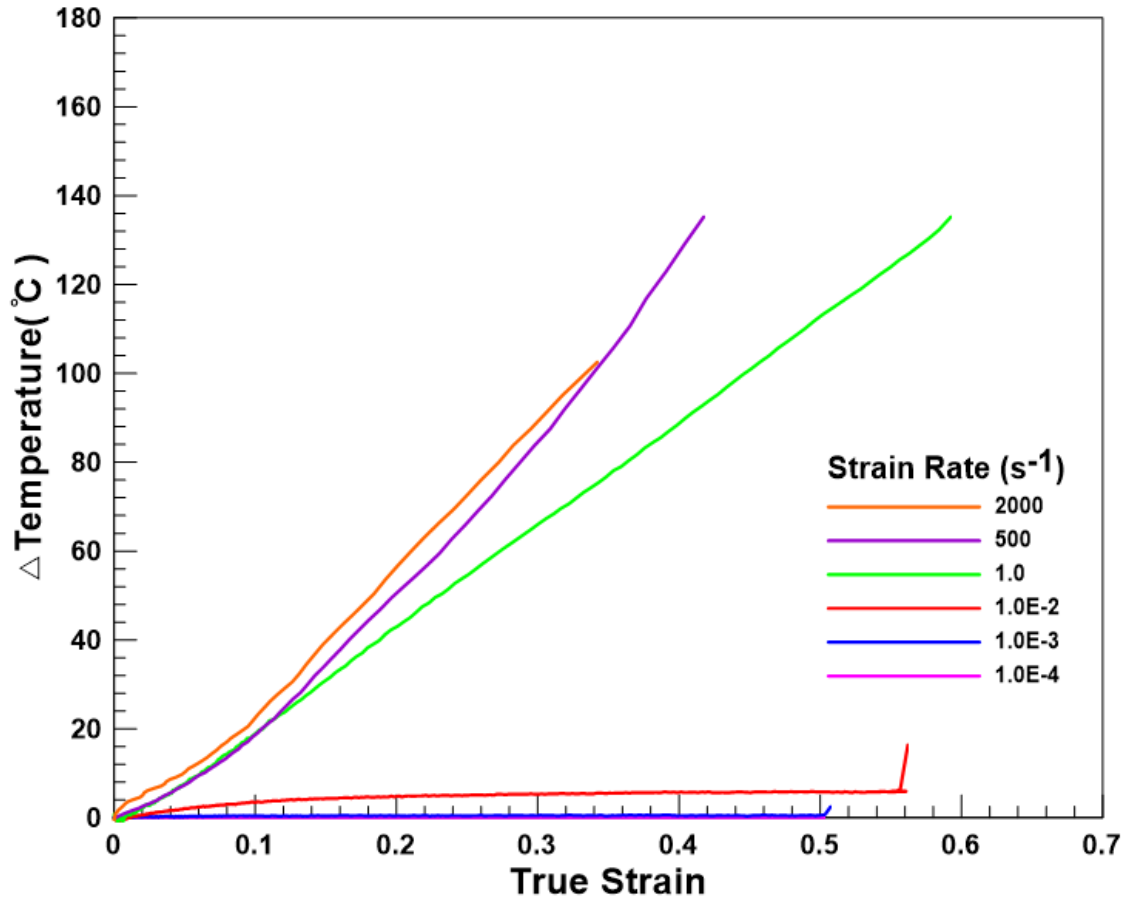


Figure 3-3 Temperature change-true strain curves from tension tests at various strain rates

3.1.2 Compression

The compressive response of 410 Stainless Steel across the tested strain rate range is presented in Figure 3-4. Similar to the tensile results, the material exhibits a clear strain rate sensitivity where flow stress increases along with increasing strain rate. To ensure repeatability, each condition is tested on three samples; with standard deviations at yield and peak stress remaining within 2%. A single representative curve is shown for each rate. Post-test inspection of the specimens (Figure 3-5) reveals significant plastic deformation without visible fracture, with the characteristic barreling that occurs during compression testing.

The thermal response, shown in Figure 3-6 highlights the transition from isothermal to adiabatic conditions. The temperature is averaged from a 1 mm by 1 mm region at the center of the specimen surface. At strain rate of 10^{-3} s^{-1} , the process remains isothermal. For tests at strain rate of 1.0 s^{-1} and above, the temperature-strain curves remain nearly linear, but they exhibit a curvature direction opposite to that observed in tension. This behavior is attributed to the increasing cross-sectional area during compression, which leads to a flatter temperature profile compared to the area reduction (necking) in tension. Furthermore, the curve at strain rate of 1.0 s^{-1} appears flatter than the higher strain rate curves. This is likely due to heat loss to the platens and the specimen itself since the deformation speed at this strain rate is not sufficiently high to achieve fully adiabatic conditions.

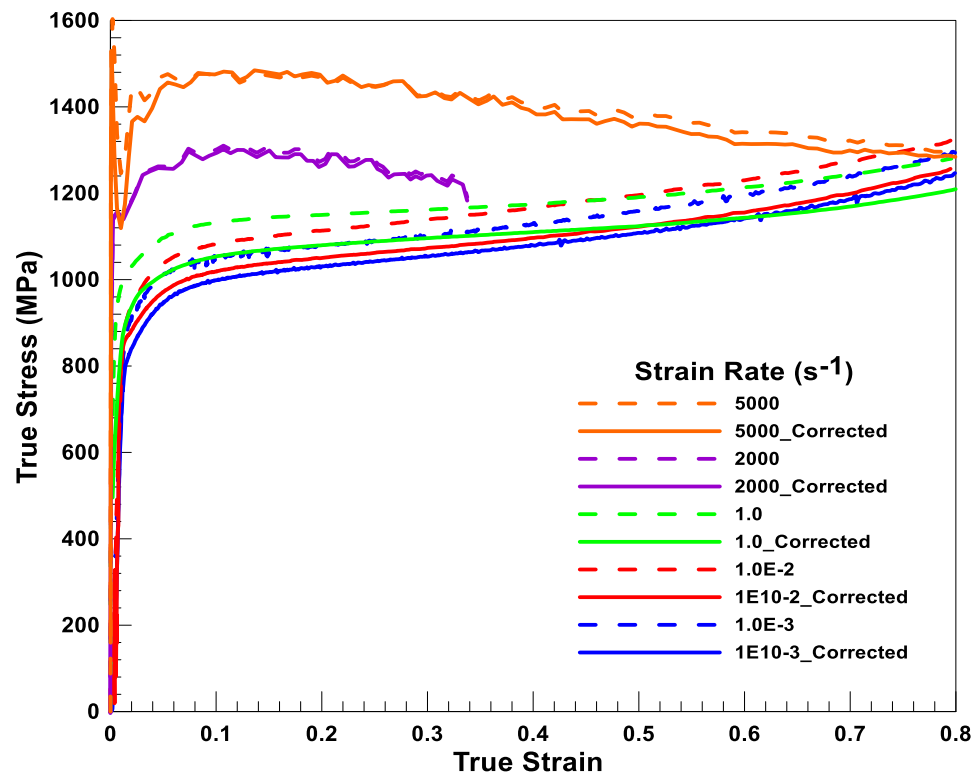


Figure 3-4 True stress-strain curves from compression tests at various strain rates

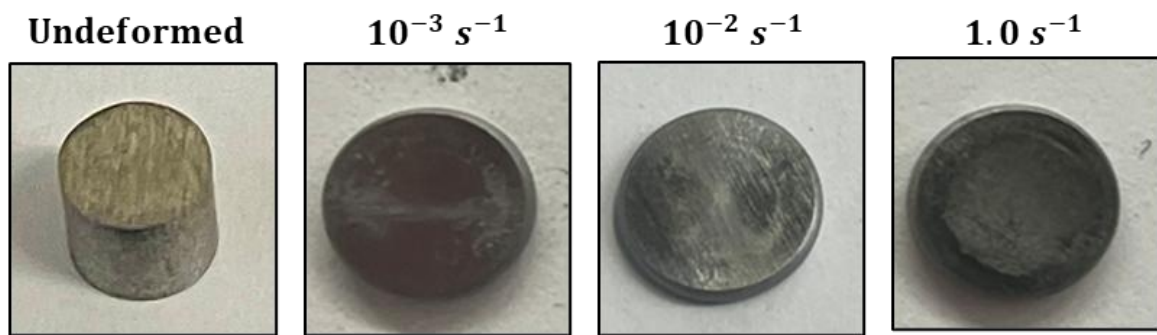


Figure 3-5 Specimens after compression testing at various strain rates

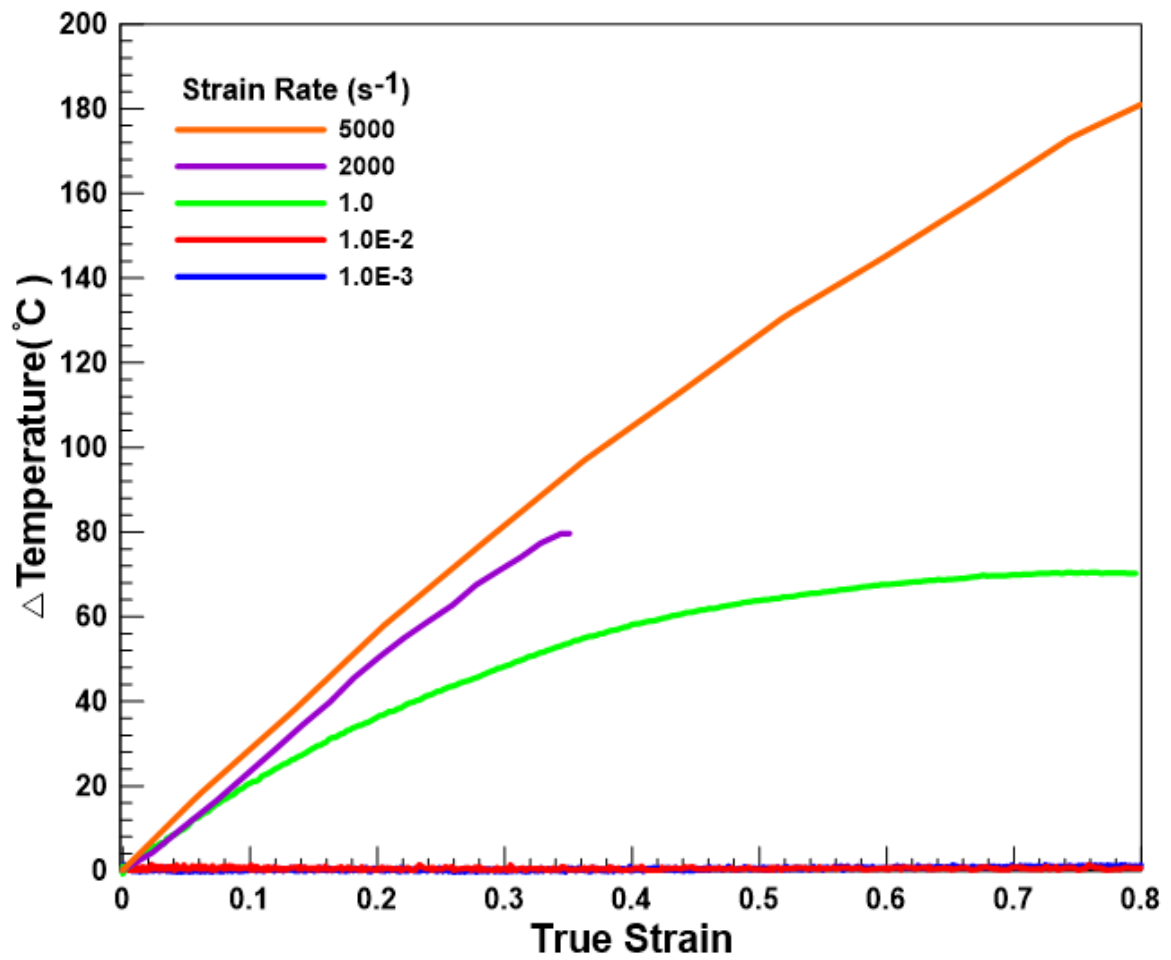


Figure 3-6 Temperature change-true strain curves from compression tests at various strain rates

To account for friction and non-uniform deformation, the barreling effect is analyzed using two methods. For low and intermediate strain rates where 3D DIC data is available, an analytical model (eq.2-8~2-10) is employed. At high strain rates, where direct 3D measurement is not feasible, a validated simulation model is used. As shown in Figure 3-7 and Figure 3-8, the simulation at 1.0 s^{-1} shows an excellent match with experimental force-displacement data and the barreling shape at various strain. By using the friction conditions estimated from the analytical model (eq. 2-12), the simulation is validated as a reliable tool to predict the maximum radius (R_{max}) at high strain rates. This allowed for the calculation of the corrected stress-strain curves shown in Figure 3-4, ensuring the data represents the true material response.

The friction of simulation model requires a transition from the friction coefficient μ obtained analytically to a velocity-dependent numerical framework. As shown in Figure 3-9, the experimental μ value is used as a target to calibrate the simulation's friction model. This is achieved by adjusting the static friction coefficient (μ_s), the dynamic friction coefficient (μ_d), and the decay constant (c). This refinement ensures that the numerical interface mimics the experimental slip behavior.

The simulation results for 2000 s^{-1} and 5000 s^{-1} serve as a verification step to ensure the model, calibrated at 1.0 s^{-1} , maintains its accuracy at higher strain rates. In the absence of 3D DIC data for these high-rate tests, the model's accuracy is confirmed by its ability to replicate the experimental barreling level (Figure 3-10 and Figure 3-12) and the coupled force-displacement and temperature histories (Figure 3-11 and Figure 3-13). This verification justifies the use of the simulation to extract the instantaneous maximum radius

(R_{max}) throughout the deformation process at high strain rate. These simulated R_{max} values are used to calculate the corrected true stress for the high-rate curves in Figure 3-4.

The influence of friction on the measured stress is quantified by comparing the raw experimental data to the corrected results. As shown in Figure 3-4, the uncorrected stress for quasi-static loading (strain rates of 10^{-3} s^{-1} and 10^{-2} s^{-1}) is approximately 8% higher than the actual material strength due to friction. This discrepancy decreases to 6% at an intermediate strain rate of 1.0 s^{-1} , while the dynamic compression tests exhibit almost negligible frictional effects. This reduction in friction is driven by the average slip velocity at the specimen-platen interface, which is calculated by dividing the horizontal displacement of the specimen edge via DIC by the test duration. Figure 3-14 illustrates the inverse relationship between the Average Barreling Factor (b_{avg}) and the calculated slip velocity. The data demonstrates that as the strain rate increases, the higher slip velocity significantly lowers the frictional resistance at the interface. Representative high-speed camera images captured at an axial strain of $\varepsilon = 0.6$ (and $\varepsilon = 0.4$ for the 2000 s^{-1} strain rate test) visually confirm this trend. Specimens subjected to higher slip velocities maintain a more cylindrical profile compared to the barreled quasi-static samples.

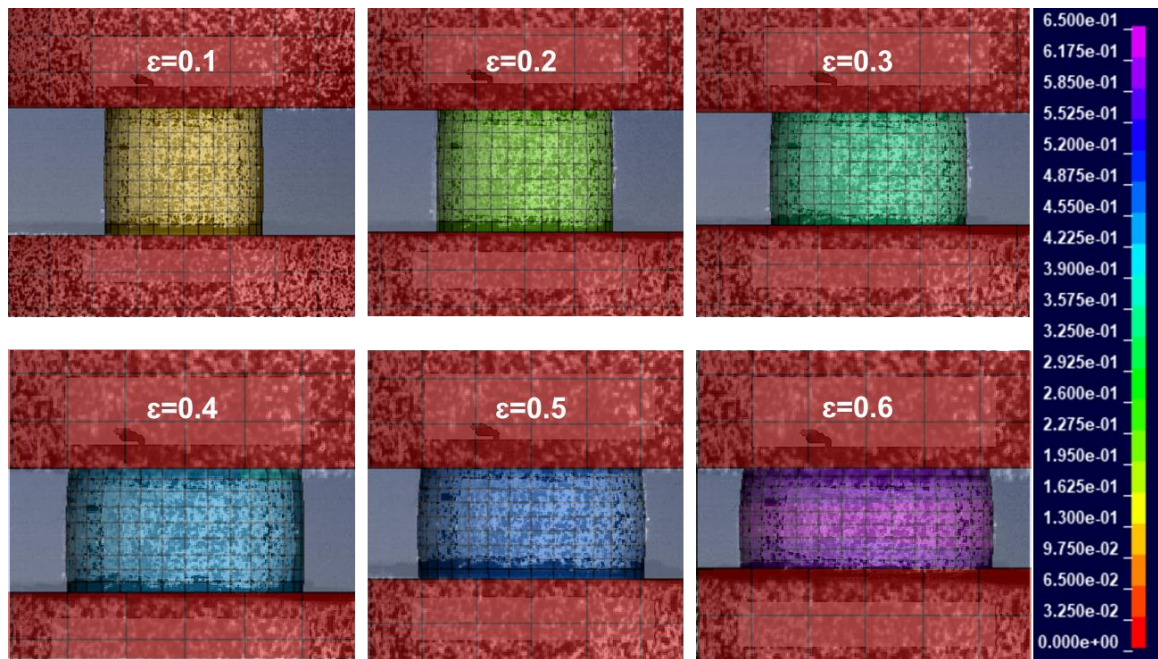


Figure 3-7 Overlay of experimental DIC images and finite element simulation for validation of barreling evolution at 1.0 s^{-1}

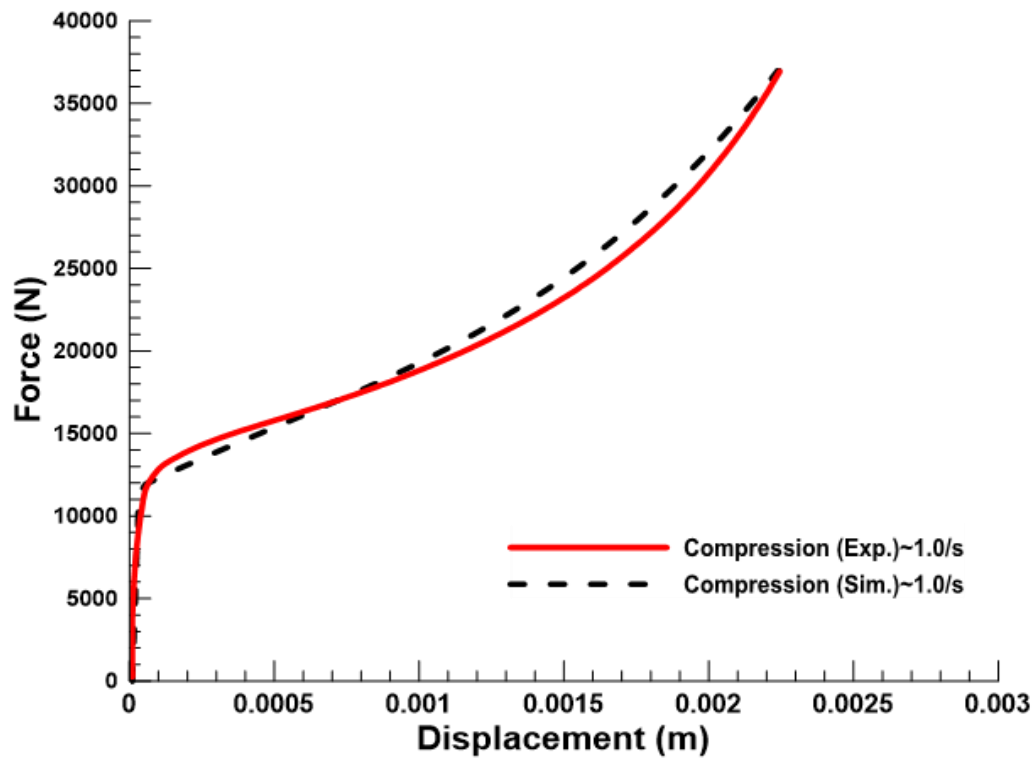


Figure 3-8 Experimental and simulated force-displacement responses for validation of the compression model at 1.0 s^{-1}

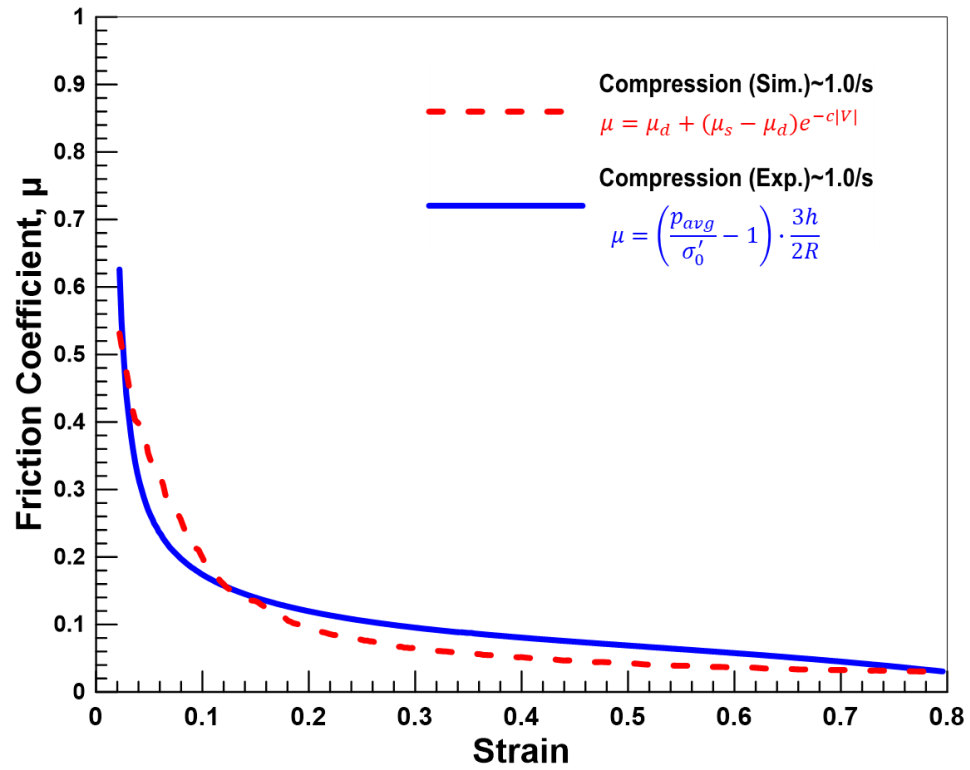


Figure 3-9 Evolution of friction coefficient with strain for compression simulations

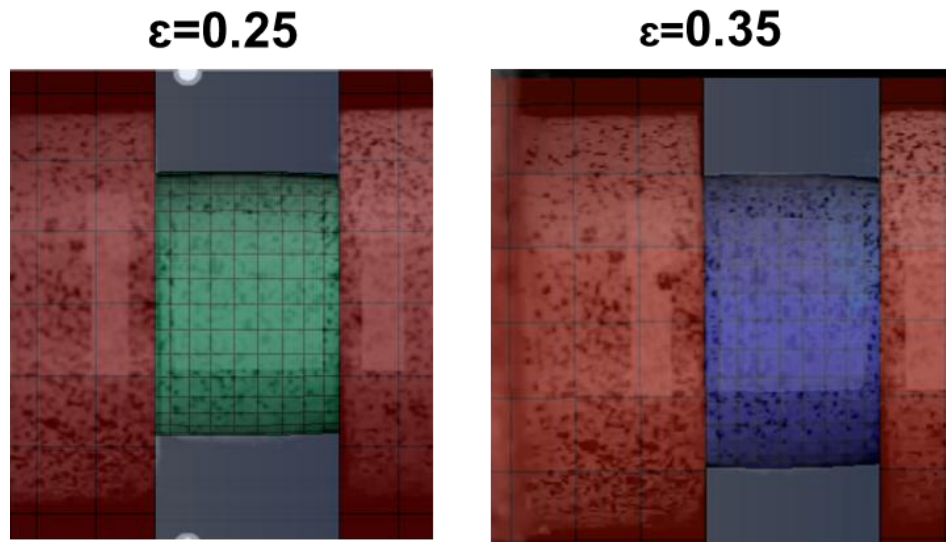


Figure 3-10 Overlay of experimental DIC images and finite element simulation for validation of barreling evolution at 2000.0 s⁻¹

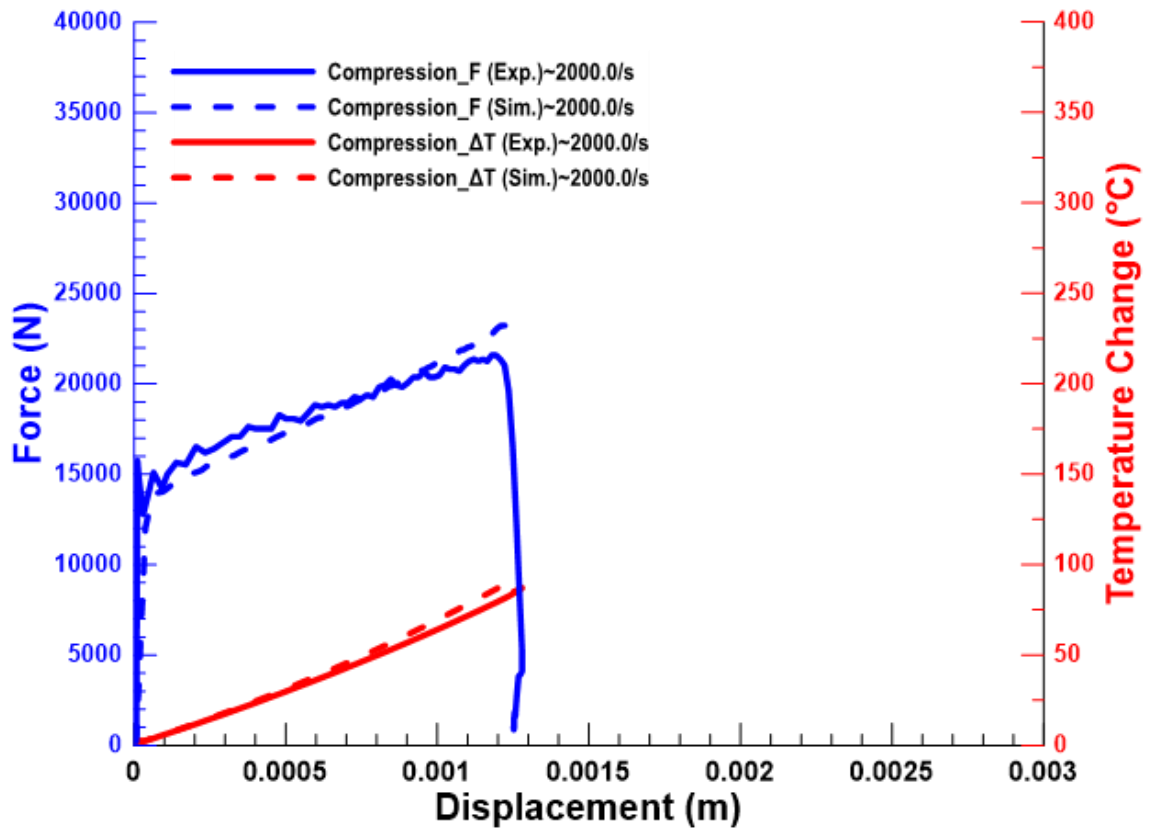


Figure 3-11 Comparison of experimental and simulated force–displacement and temperature histories at 2000 s⁻¹

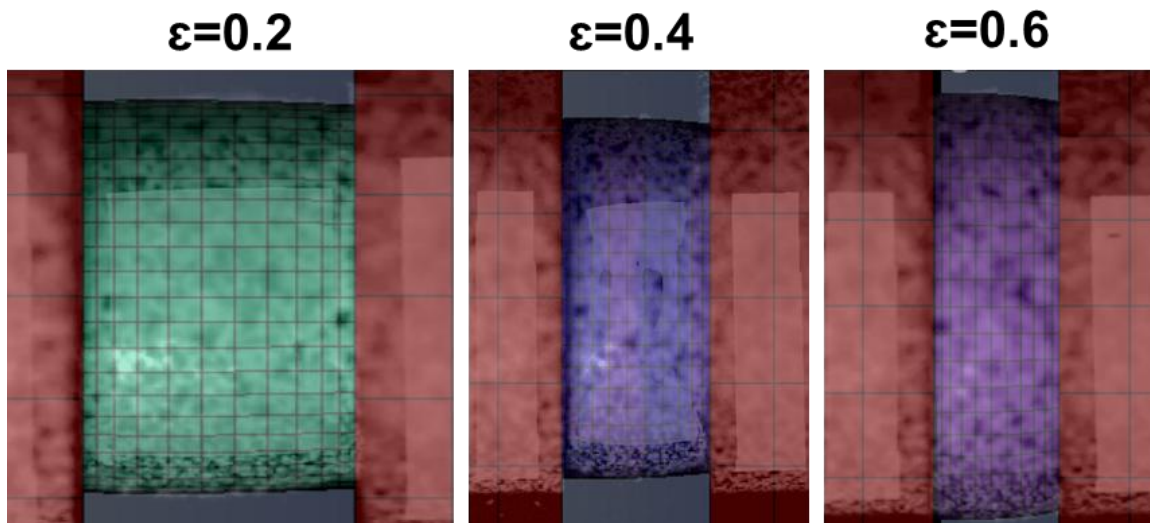


Figure 3-12 Overlay of experimental DIC images and finite element simulation for validation of barreling evolution at 5000.0 s⁻¹

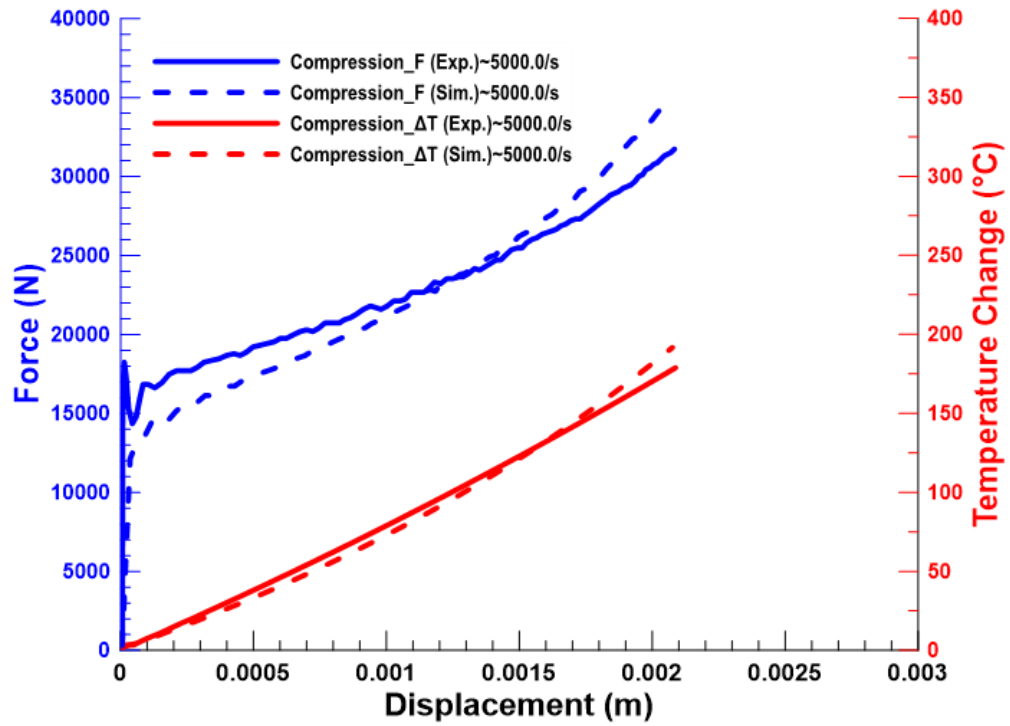


Figure 3-13 Comparison of experimental and simulated force–displacement and temperature histories at 5000 s⁻¹

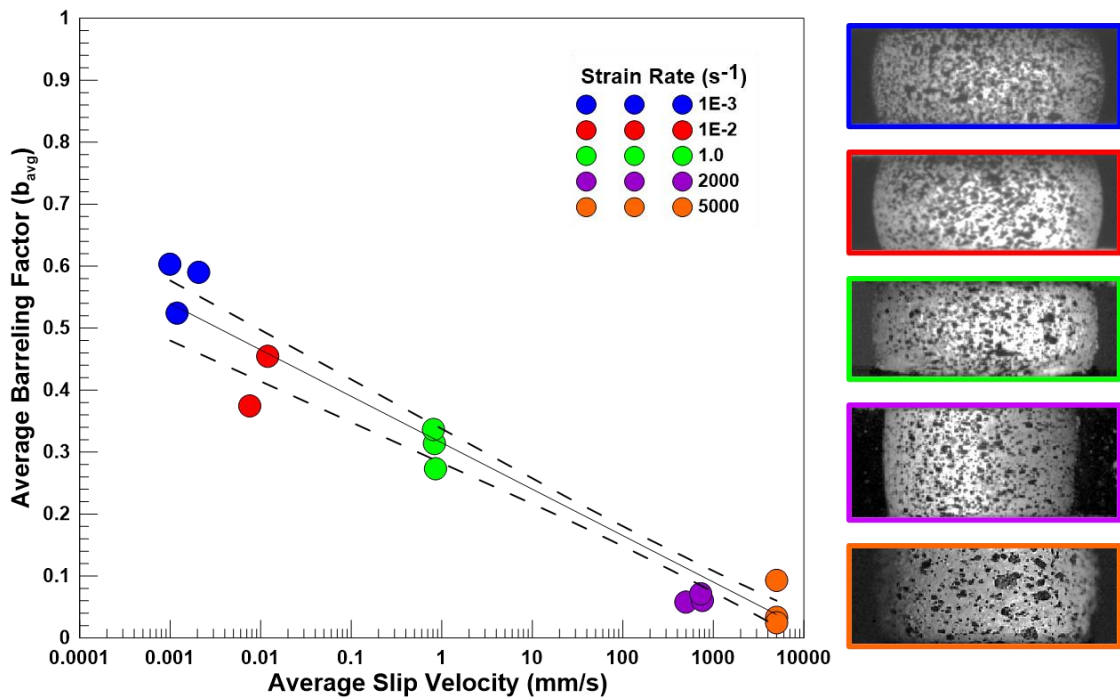


Figure 3-14 Relationship between average barreling factor and average slip velocity in compression tests at various strain rates

3.1.3 Torsion

The torsional response of 410 Stainless Steel across the investigated strain rate range is presented in Figure 3-15. Consistent with the tensile and compressive data, the material displays a positive strain rate sensitivity. However, the strain hardening rate in torsion is lower than that observed in tension and compression. Each condition is tested on three samples to ensure repeatability; since the standard deviations at yield stress and maximum stress remains within 2%, a single representative curve is provided for each rate.

The thermodynamic behavior under torsional loading, shown in Figure 3-16, reveals distinct differences from the other loading modes. At the quasi-static strain rates of 10^{-4} s^{-1} and 10^{-2} s^{-1} , the material remains nearly isothermal throughout the deformation. Unlike the tension tests, these low-rate torsion tests lack a sharp temperature spike immediately before fracture. At a strain rate of 1.0 s^{-1} , the temperature-strain relationship is nearly linear, though the curvature is flatter than that seen in tension, more close to the compression results.

At high dynamic strain rates of 1000 s^{-1} and 4500 s^{-1} , the temperature-strain curves maintain a near-linear trend before the temperature spike. In these torsional cases, the temperature spike occurs at a lower equivalent strain compared to the tension tests. While the temperature continues to rise sharply, the torque-rotation data confirms that the material continues to sustain a mechanical load for a duration before separation occurs.

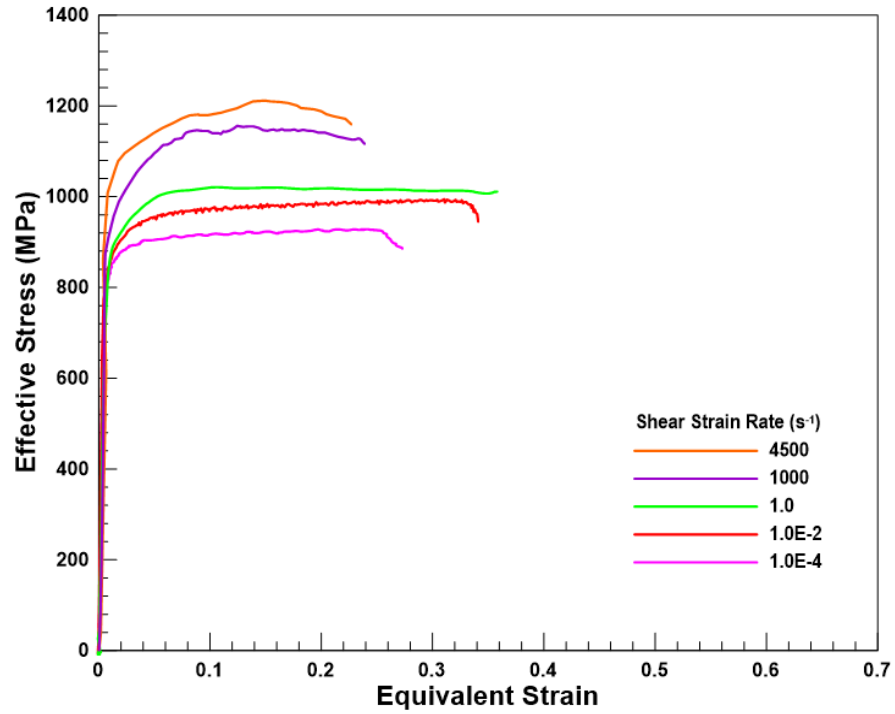


Figure 3-15 Effective Stress-equivalent strain curves from torsion tests at various strain rates

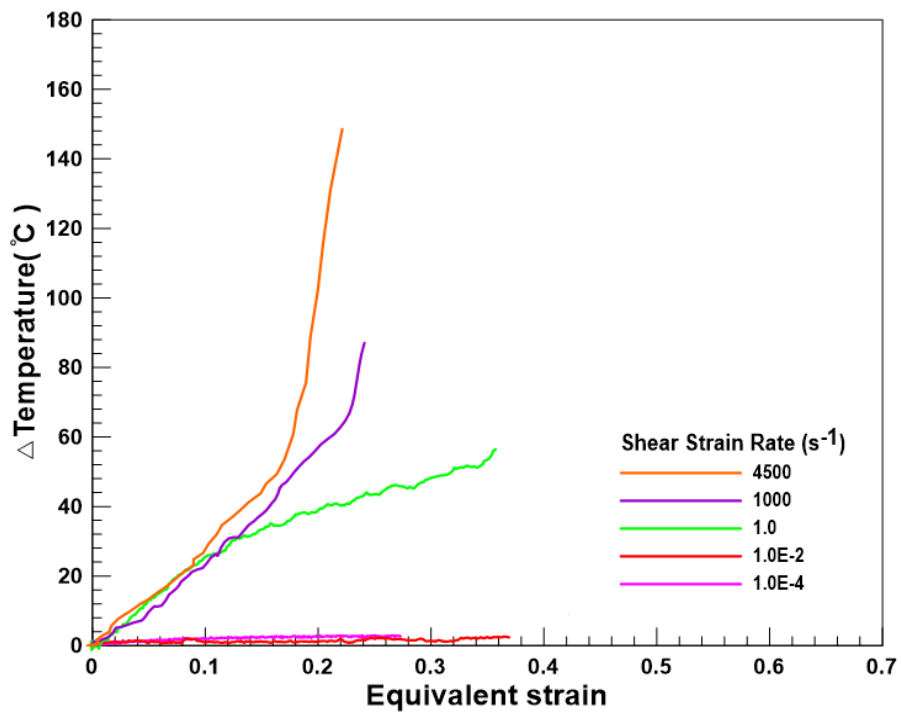


Figure 3-16 Temperature change-equivalent strain curves from torsion tests at various strain rates

3.1.4 Discussion

At the quasi-static strain rate of 10^{-3} s^{-1} (Figure 3-17), the stress-strain response is consistent between tension and compression. However, at strain rates of 10^{-2} s^{-1} (Figure 3-18) and 1.0 s^{-1} (Figure 3-19), while the curves remain similar during elastic deformation and the early stages of plasticity, they diverge after 5% strain. Tension and compression show the higher hardening rates, whereas torsion shows a flatter stress-strain profile with less strain hardening.

The thermal behavior at strain rate of 1.0 s^{-1} (Figure 3-20) further differentiates these loading modes. While the temperature change in tension stays linear, the curves for compression and torsion flatten as deformation progresses. This difference is explained by the temperature-strain distributions shown in Figure 3-21 ~ Figure 3-23. In tension (Figure 3-21), the temperature and strain accumulation are mostly consistent at the same spots. Despite some heat loss, the reduction in cross-section area keeps the temperature increase linear. In contrast, compression data (Figure 3-22) shows that strain is highest at the specimen edges due to friction, yet the temperature does not increase accordingly, revealing heat loss to the platens and the specimen interior. Torsion (Figure 3-23) shows a similar result, with a clear temperature drop relative to the shear strain at the edges of the gauge section.

As the strain rate increases into the dynamic range (500 s^{-1} to 5000 s^{-1}), the thermal responses become more consistent with the accumulated strain, showing a near-linear relationship prior to fracture (Figure 3-24 and Figure 3-25). This behavior occurs because the high deformation speeds significantly reduce the time available for heat to conduct into

the test fixtures or the specimen itself, effectively achieving adiabatic heating conditions. The summary of effective stress across all tested rates (Figure 3-26) reveals that 410 Stainless Steel is most sensitive to strain rate under compressive loading, especially in the dynamic range, where the flow stress increases more sharply compared to the tensile and torsional responses. The accompanying increase in temperature rise further indicates that adiabatic heating becomes more significant at higher strain rates.

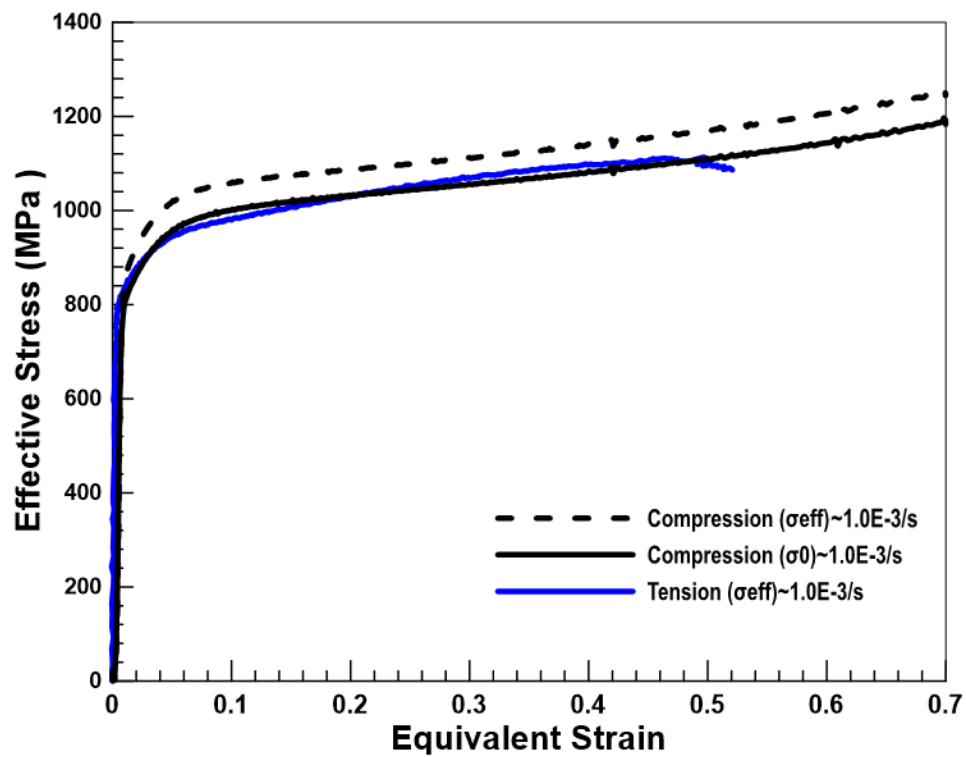


Figure 3-17 Comparison of effective stress–equivalent strain curves under different loading modes at 10^{-3} s^{-1}

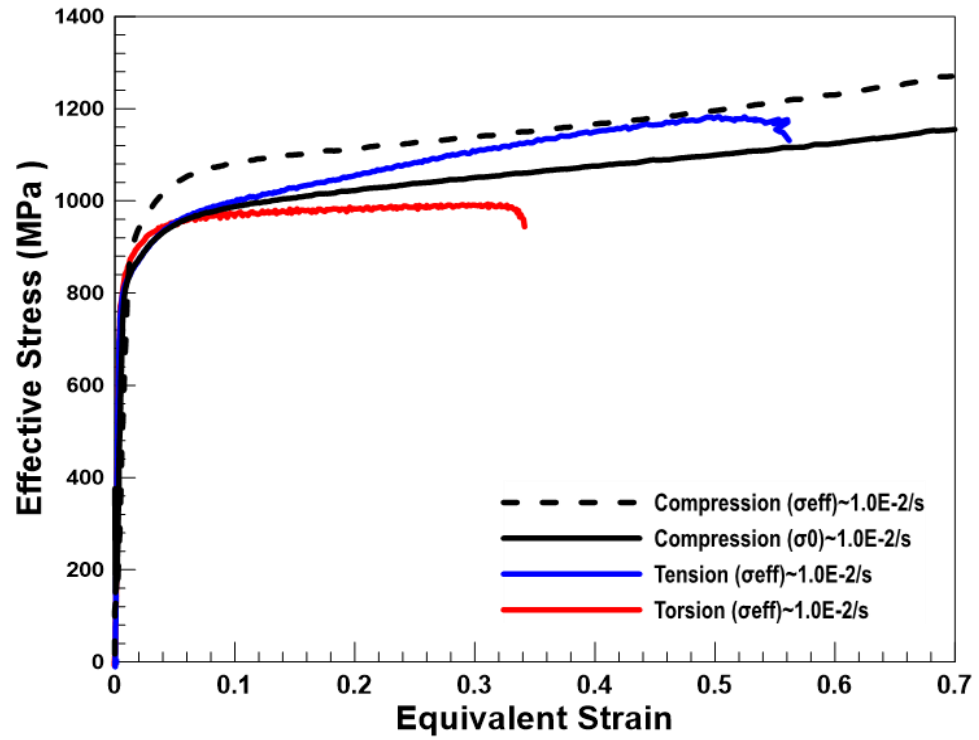


Figure 3-18 Comparison of effective stress–equivalent strain curves under different loading modes at 10^{-2} s^{-1}

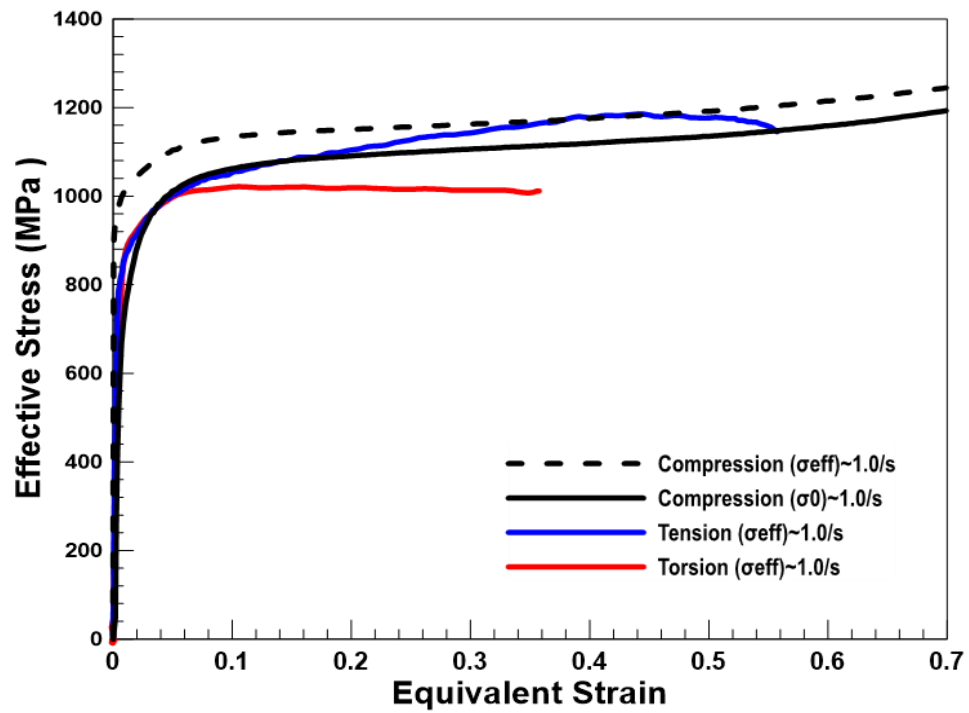


Figure 3-19 Comparison of effective stress–equivalent strain curves under different loading modes at 1.0 s^{-1}

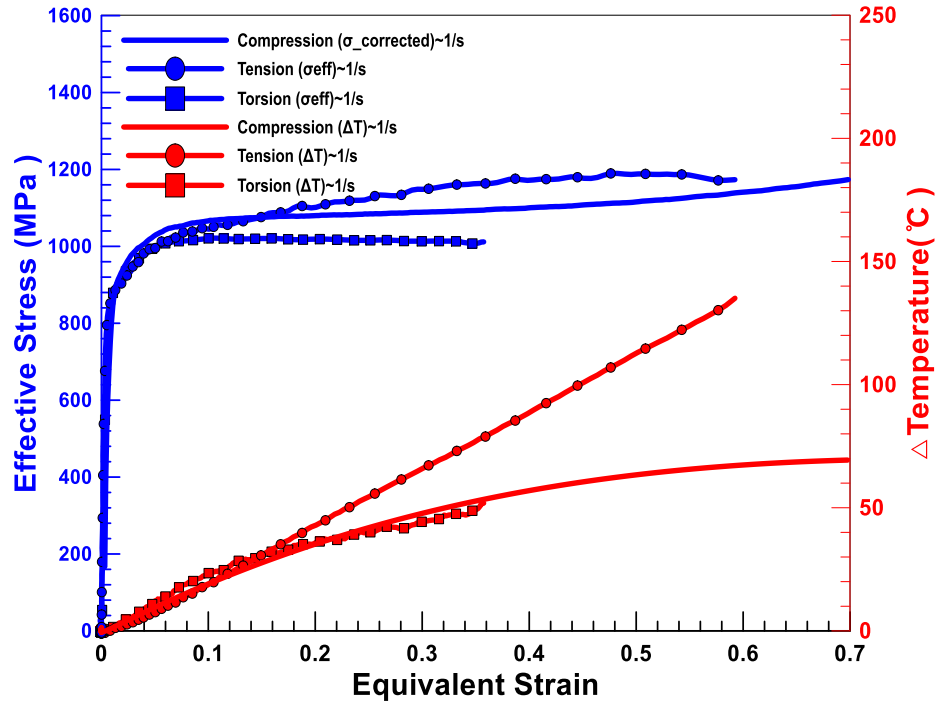


Figure 3-20 Comparison of effective stress, temperature rise, and equivalent strain under different loading modes at 1.0 s^{-1}

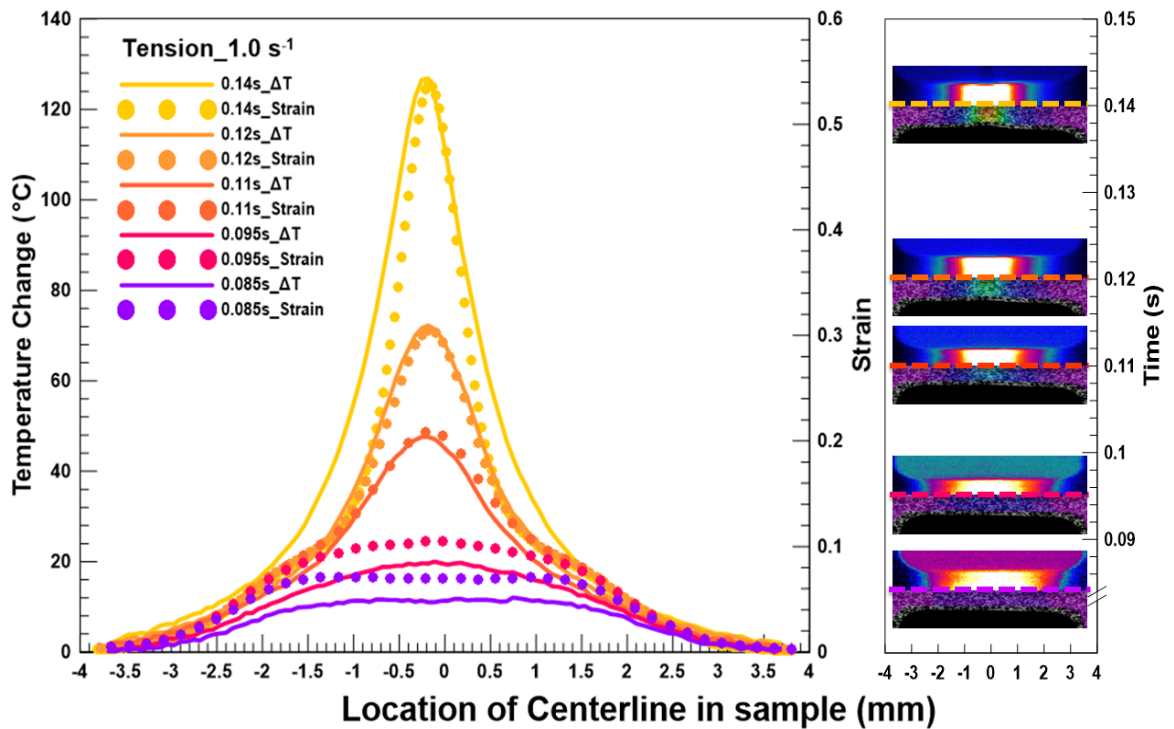


Figure 3-21 Evolution of temperature and strain distributions across the specimen centerline during tension at 1.0 s^{-1}

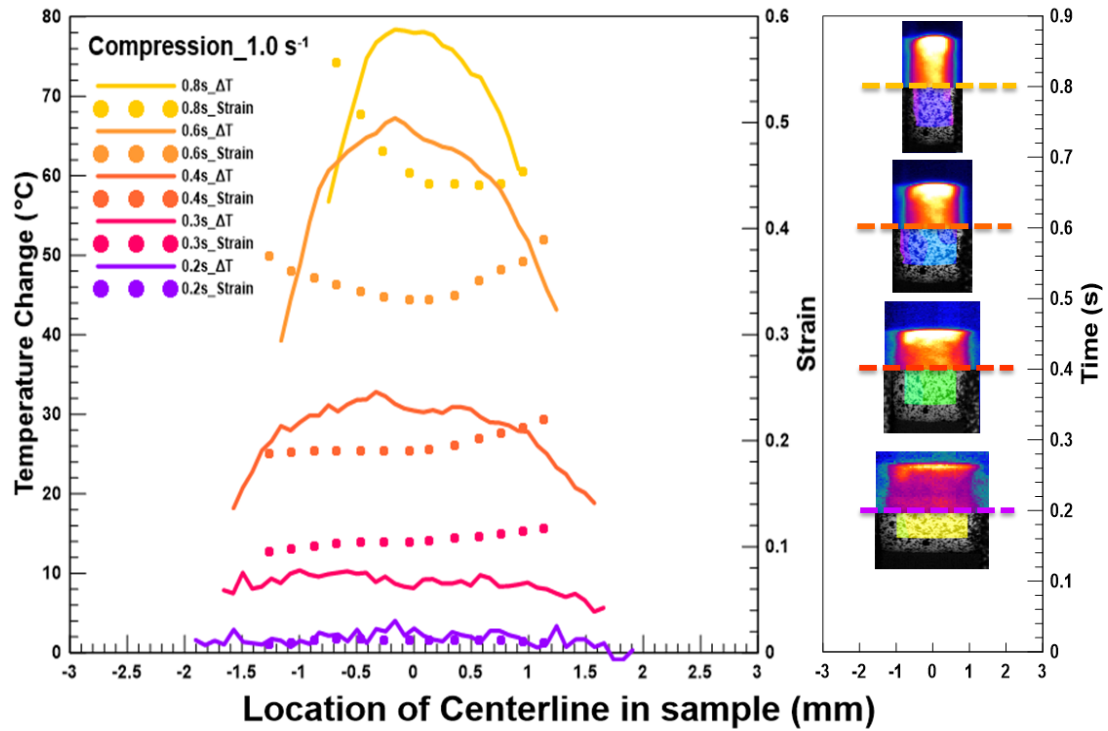


Figure 3-22 Evolution of temperature and strain distributions across the specimen centerline during compression at 1.0 s⁻¹

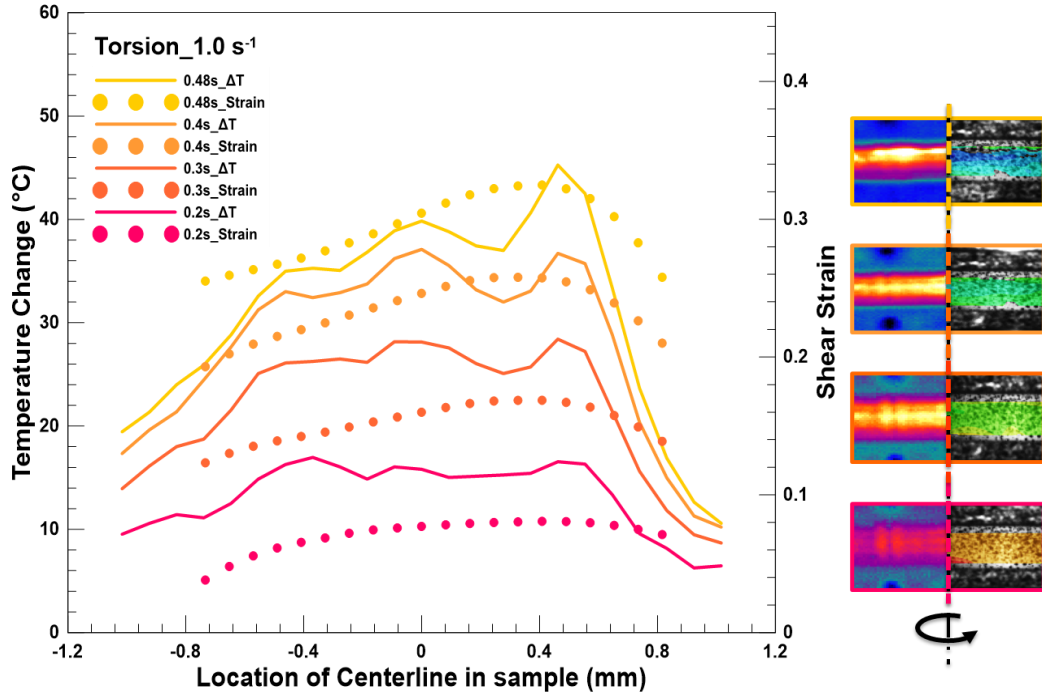


Figure 3-23 Evolution of temperature and strain distributions across the specimen centerline during torsion at 1.0 s⁻¹

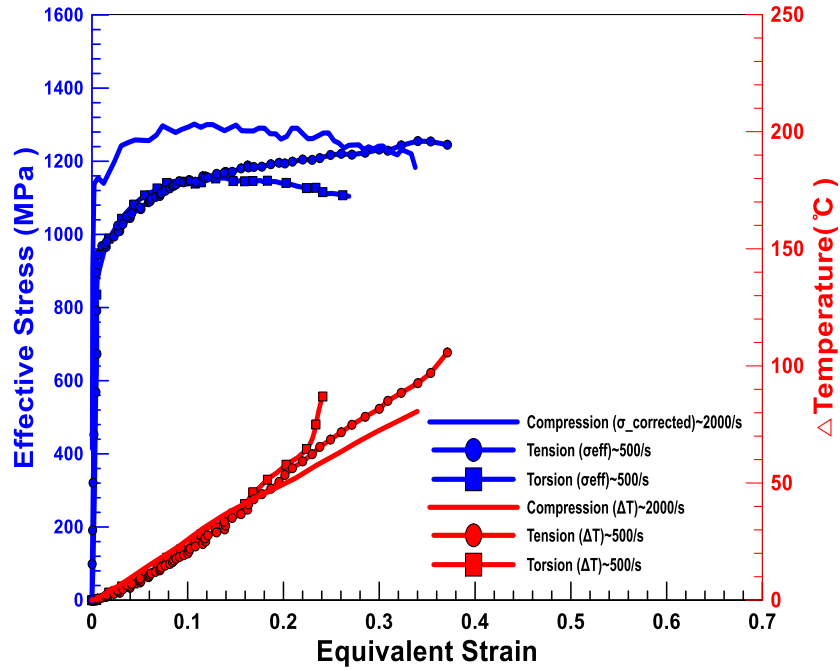


Figure 3-24 Comparison of effective stress, temperature rise, and equivalent strain under different loading modes at dynamic strain rates (500~2000 s⁻¹)

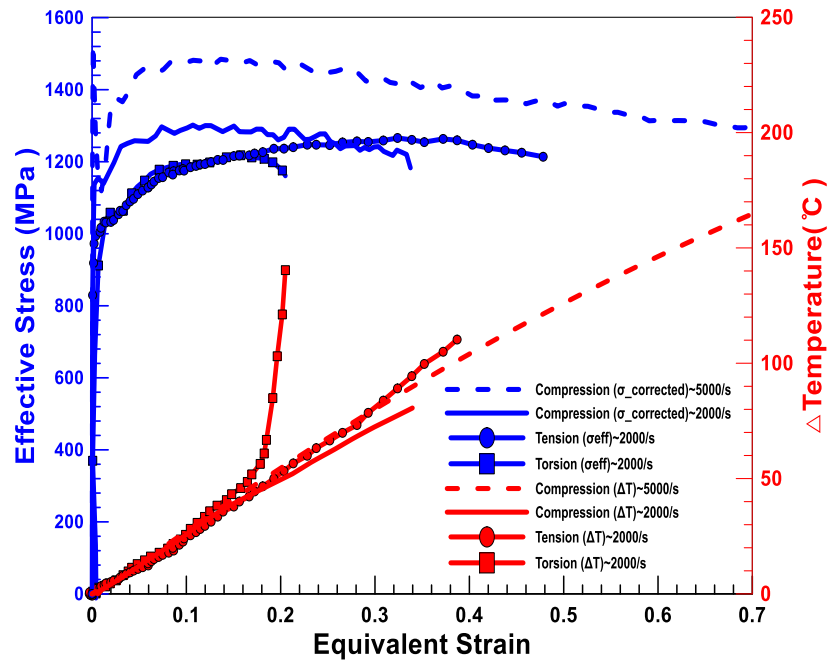


Figure 3-25 Comparison of effective stress, temperature rise, and equivalent strain under different loading modes at high strain rates (2000~5000 s⁻¹)

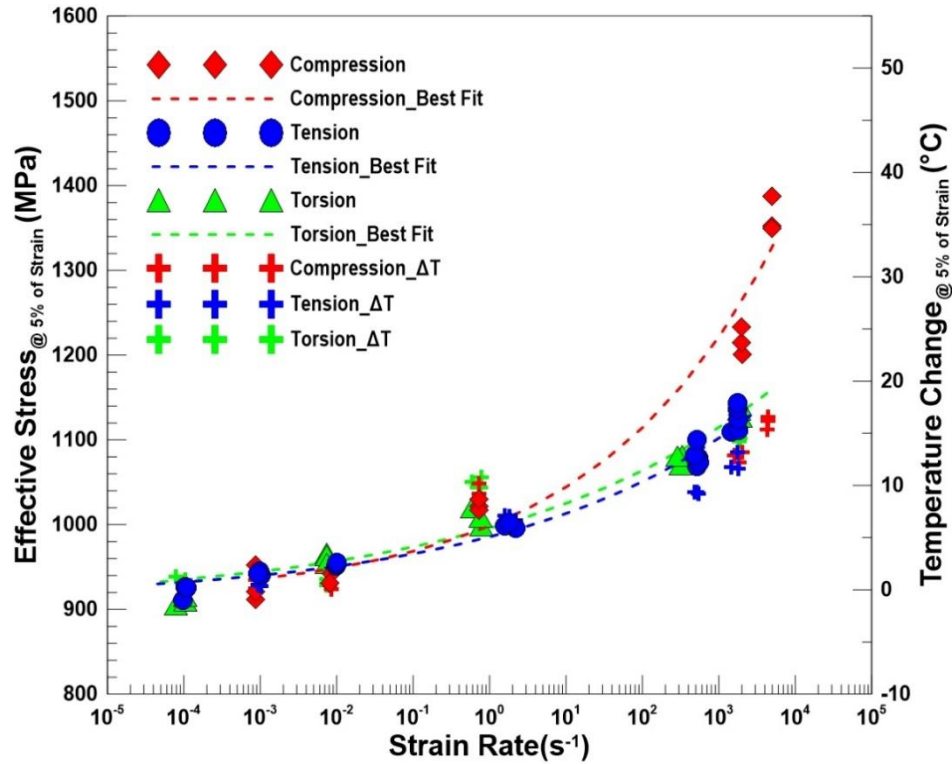


Figure 3-26 Comparison of effective stress and corresponding temperature rise at 5% plastic strain under tension, compression, and torsion over a wide range of strain rates

3.2 Temperature Dependency

This section examines the mechanical response of 410 Stainless Steel under quasi-static loading (strain rate of 10^{-3} s^{-1}) across temperatures ranging from room temperature (RT) to 800°C . As the temperature increases, the material displays significant thermal softening, characterized by a reduction in flow stress and distinct changes in strain hardening behavior depending on the loading mode.

The tensile stress-strain curves at elevated temperatures are presented in Figure 3-27. The strain hardening behavior remains consistent from room temperature up to 400°C ; however, as the temperature reaches 600°C and 800°C , the curves become nearly flat. The

tested specimens (Figure 3-28) reveal a clear increase in ductility with rising temperature, indicated by more pronounced necking before fracture. At 800°C, the flow stress drops to approximately 12% of its room-temperature value.

The compressive response, shown in Figure 3-29, shows moderate strain hardening at room temperature. This hardening disappears as the temperature increases, with the curves becoming flat from 200°C through 800°C. In contrast, torsion shows the most stable hardening response across the entire temperature range (Figure 3-30), with almost no strain hardening observed from room temperature through 800°C. The material reaches a steady-state flow stress almost immediately after yielding under shear loading.

Figure 3-31 summarizes the thermal softening behavior by comparing the effective stress across all loading modes at 10% equivalent strain. All modes show a consistent non-linear softening trend, where the flow stress remains relatively high up to 400°C before dropping sharply as the material approaches 800°C. The effective stress values for tension, compression, and torsion align closely throughout the entire temperature range. The 410 Stainless Steel reaches a nearly identical low-strength state and hardening state across all loading modes at the 600°C threshold.

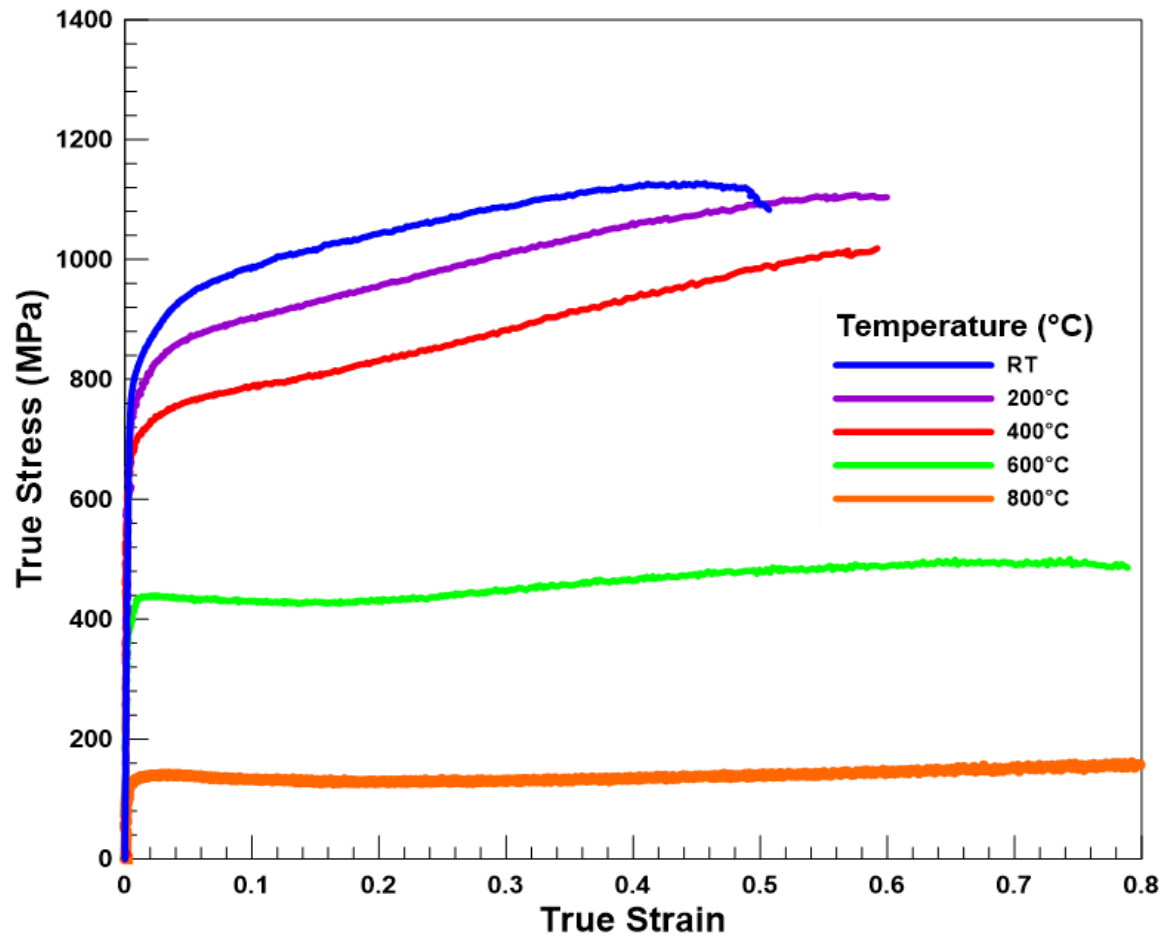


Figure 3-27 The tension of effective stress from various temperatures at 10^{-3} s^{-1}

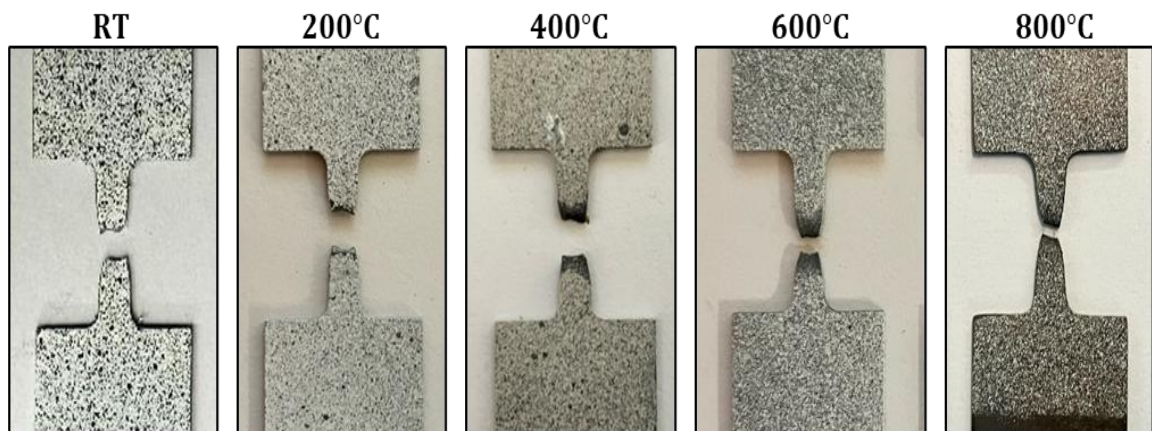


Figure 3-28 Specimens after tension testing at various temperatures

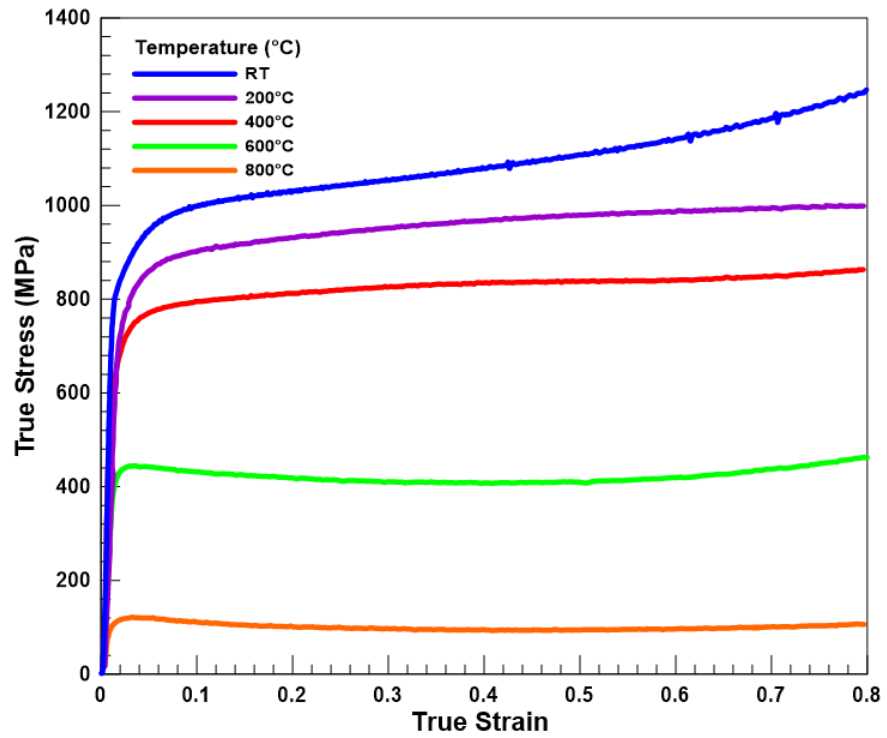


Figure 3-29 The compression of effective stress from various temperatures at 10^{-3} s^{-1}

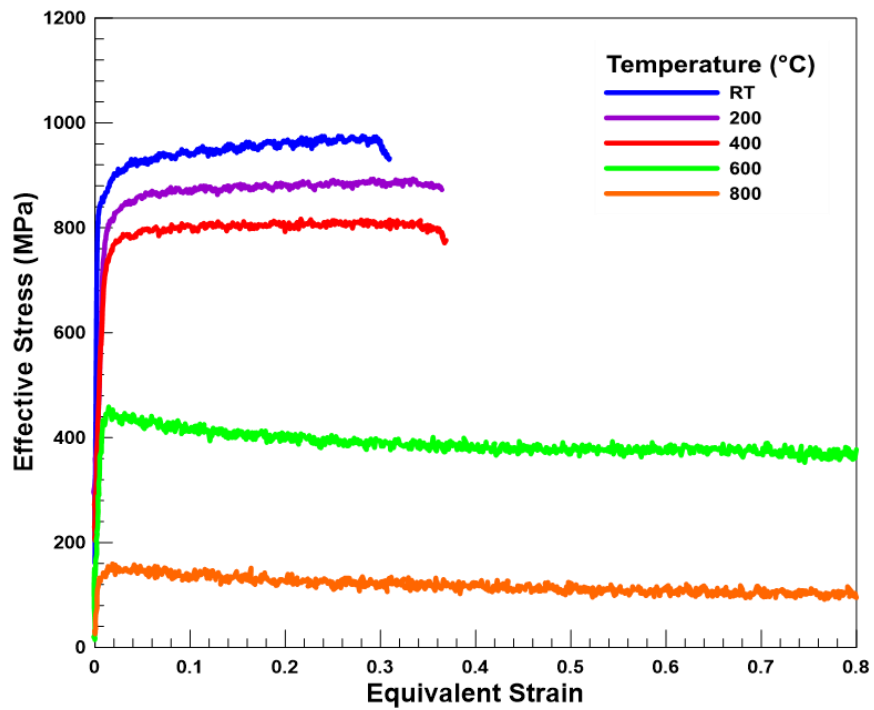


Figure 3-30 The torsion of effective stress from various temperatures at 10^{-3} s^{-1}

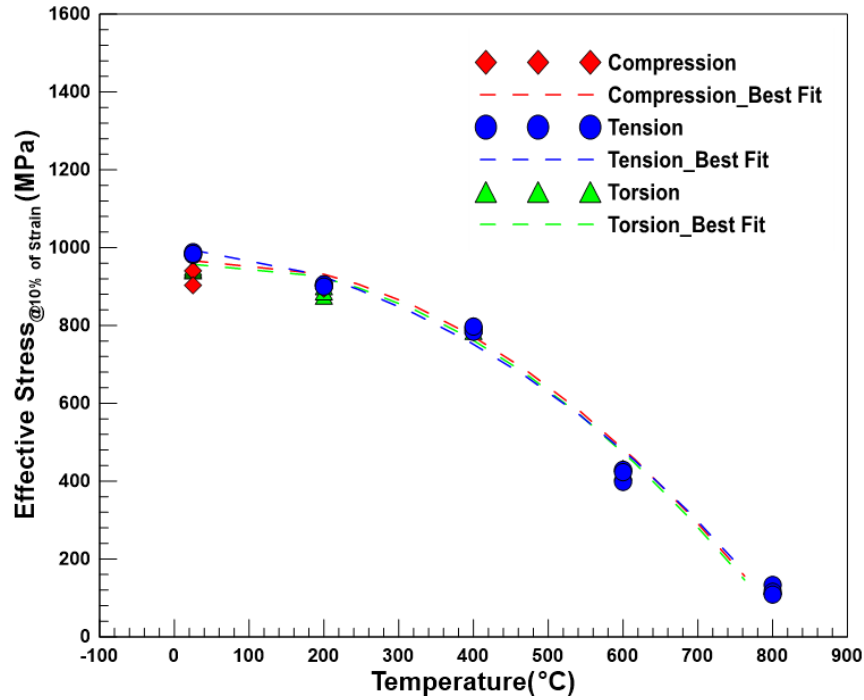


Figure 3-31 The comparison of effective stress from various temperatures at 10^{-3} s^{-1}

3.3 Anisotropy

This section evaluates the directional dependence of the mechanical properties of the 410 SS plate using specimens extracted at 0° , 90° , and $\pm 45^\circ$ relative to the rolling direction. As described in Section 2.3.4, the thick-plate specimen configuration adopted in this study differs from the conventional thin-sheet configuration used for R-value characterization. While the mathematical definition of the Lankford coefficient remains unchanged, the measured R-values represent different in-plane material directions of the plate depending on the specimen orientation. The anisotropy results presented in this section should therefore be interpreted according to the specimen configuration described previously

The tensile response across different orientations is shown in Figure 3-32. The stress strain behavior remains consistent regardless of the orientation. The temperature change

across all specimens remains near zero (Figure 3-33). Similar to the behavior observed in Section 3.1.1, the deformation occurs at a sufficiently slow rate such that any heat generated by plastic work is dissipated to the surroundings.

The tested specimen (Figure 3-34) shows consistent necking and fracture profiles across all orientations. The calculated Lankford coefficients (R-values) for tension, measured at 20% strain, are summarized in Table 3-1. The average R-values are 1.012, 1.047, 0.992, and 1.089 for the 0° , 90° , $+45^\circ$, and -45° orientations, respectively. These values are close to the isotropic value of 1.0.

The combined stress–strain response, temperature evolution, fracture characteristics, and R-values consistently indicate that the 410 SS plate exhibits only weak plastic anisotropy. Therefore, an isotropic constitutive description is considered appropriate for the material investigated in this study.

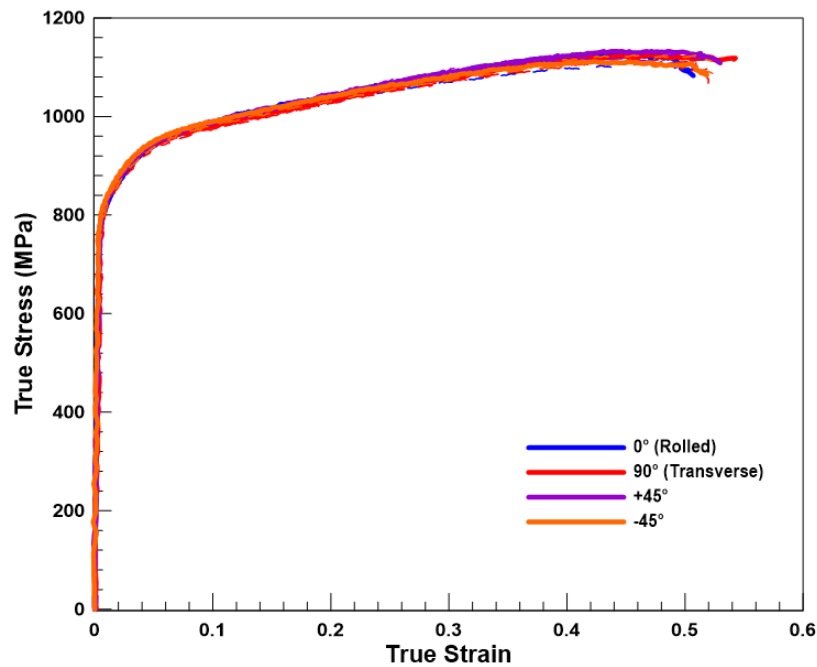


Figure 3-32 True stress-strain curves from tension tests conducted on specimens fabricated in various orientations

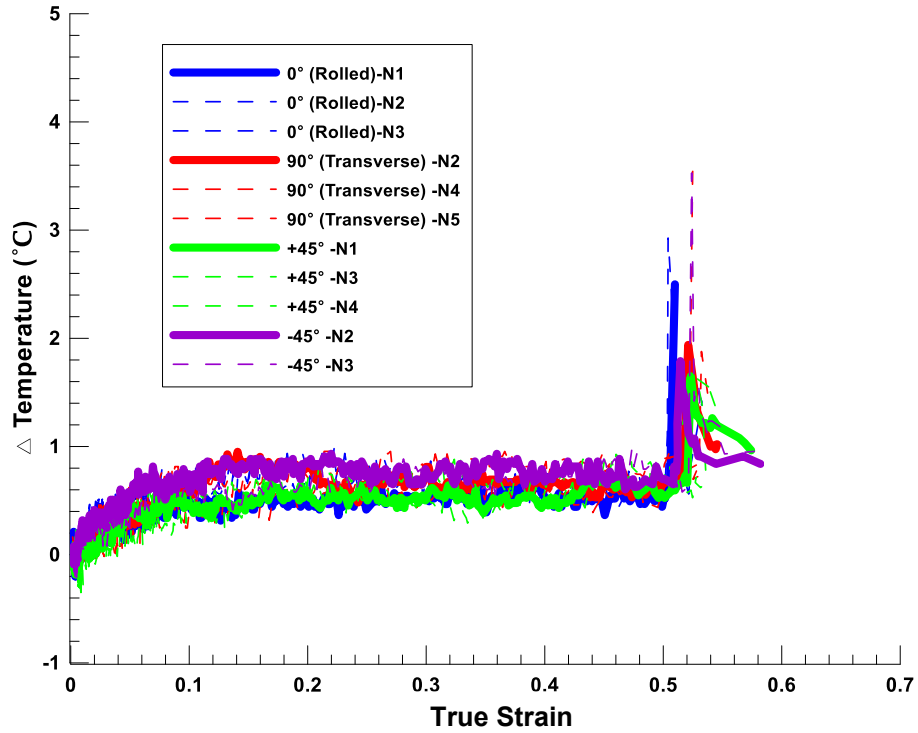


Figure 3-33 Temperature change-strain curves from tension tests conducted on specimens fabricated in various orientations

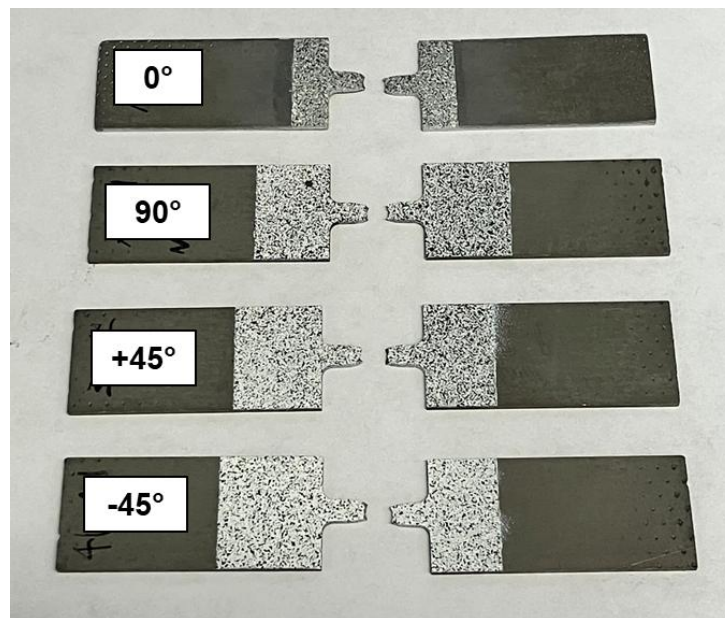


Figure 3-34 Specimens after tension tests conducted on specimens fabricated in various orientations

Table 3-1 Lankford coefficients from tensile anisotropy tests at 10^{-3} s^{-1}

Orientation ($^{\circ}$)	R-value	R_{avg}
0-N1	0.989	1.012
0-N2	1.145	
0-N3	0.903	
90-N2	1.084	1.047
90-N4	1.027	
90-N5	1.029	
+45-N1	0.938	0.992
+45-N3	1.032	
+45-N4	1.007	
-45-N2	1.122	1.089
-45-N3	1.094	
-45-N4	1.052	

3.4 Taylor-Quinney Coefficient Dependency

This section examines the evolution of the Taylor-Quinney Coefficient (TQC) as a function of plastic strain and strain rate. By analyzing the fraction of plastic work converted into heat, the energy dissipation characteristics of 410 SS are characterized across various loading modes. To ensure statistical reliability, all TQC evolution curves presented in this section are calculated from an average of three separate tests per strain rate, with the shadow-filled regions representing the standard error.

3.4.1 Tension

The evolution of the TQC in tension is characterized by three stages: the onset of plastic deformation, deformation progress, and saturation. During the onset of plastic deformation, the TQC exhibits a rapid increase as the material transitions from elastic to plastic deformation. As deformation progresses, the coefficient stabilizes, reflecting a steady state of energy dissipation. Finally, the saturation stage is reached, where the TQC

represents the steady-state conversion efficiency of the material prior to fracture. At high strain rates, the TQC remains stable between 0.75 and 0.9 during the saturation stage. At the intermediate strain rate of 1.0 s^{-1} , the TQC initially stabilizes at 0.74 but exhibits a slight decay toward 0.7 after a strain of 0.3.

At low strain rates ($10^{-3} \text{ s}^{-1} \sim 10^{-2} \text{ s}^{-1}$), Figure 3-36 exhibits a clear temperature change at the borders of the gauge section despite a lack of strain accumulation in those areas. This observation provides direct evidence of heat loss. Consequently, the TQC trend is influenced by thermal dissipation to the environment. In these quasi-static cases, the TQC appears to decay during the saturation stage, as the timescale of the test allows a larger fraction of the generated heat to dissipate, leading to an underestimation of the energy conversion. Conversely, at high strain rates ($500 \text{ s}^{-1} \sim 2000 \text{ s}^{-1}$), the TQC remains stable during saturation. The absence of this decay suggests that these rates are sufficiently fast to maintain adiabatic conditions at the gauge section.

The distribution of temperature and strain further clarifies these thermal conditions. In Figure 3-37 and Figure 3-38, the temperature distribution is sharper than the strain distribution, indicating that heat is trapped at the strain localization spots. While some strain is observed at the borders of the gauge section, there is no corresponding temperature increase in those regions. In Figure 3-39, the temperature rise is more localized than the strain accumulation at strain rates of 500 s^{-1} and 2000 s^{-1} . Specifically, when the strain reaches 0.1, strain is evident at the borders of the gauge section, yet the temperature remains unchanged at those same locations. This localization of heat confirms that thermal diffusion is restricted at higher rates.

The intermediate strain rate (1.0 s^{-1}) represents a transition between isothermal and adiabatic. While the TQC initially suggests adiabatic behavior, a slight decay begins at a strain of 0.2 to 0.3. Figure 3-21 reveals that the distribution of temperature and strain is only consistent until approximately 0.12 s (corresponding to a strain of ≈ 0.3). Beyond this point, the temperature profile begins to diverge from the strain profile, indicating the onset of heat loss even at intermediate rates. Furthermore, Figure 3-40 shows that the Fourier number (F_o) exceeds 0.1 at a strain range of 0.2 to 0.3. These metrics demonstrate the same trend of increasing heat loss, even though Figure 3-39 may not show a clear visual inconsistency between strain and temperature. TQC values derived after a strain of 0.3 in these cases may be considered invalid, as they no longer reflect a purely adiabatic energy conversion.

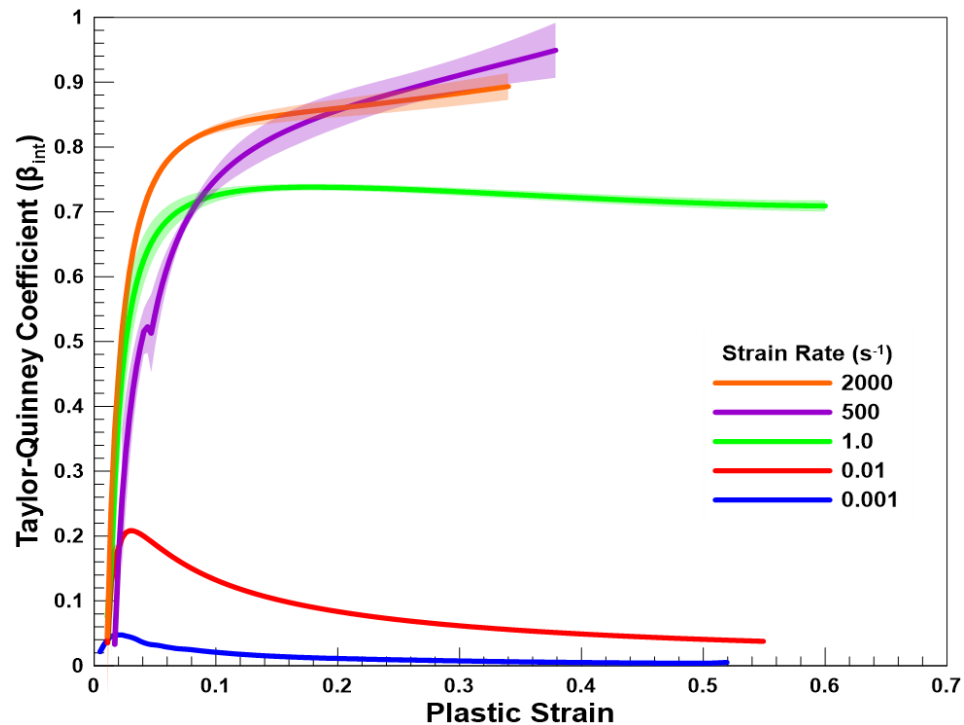


Figure 3-35 Taylor-Quinney Coefficient versus plastic strain curves from tension tests in various strain rates

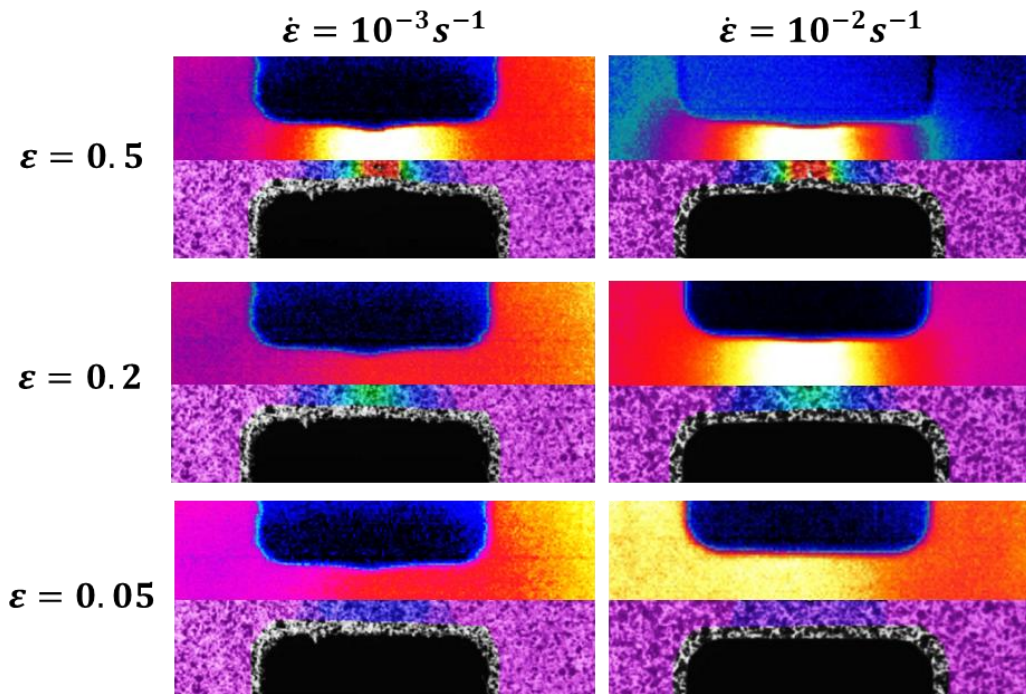


Figure 3-36 Evolution of synchronized strain and temperature distributions in 410 stainless steel at 10^{-3} s^{-1} and 10^{-2} s^{-1}

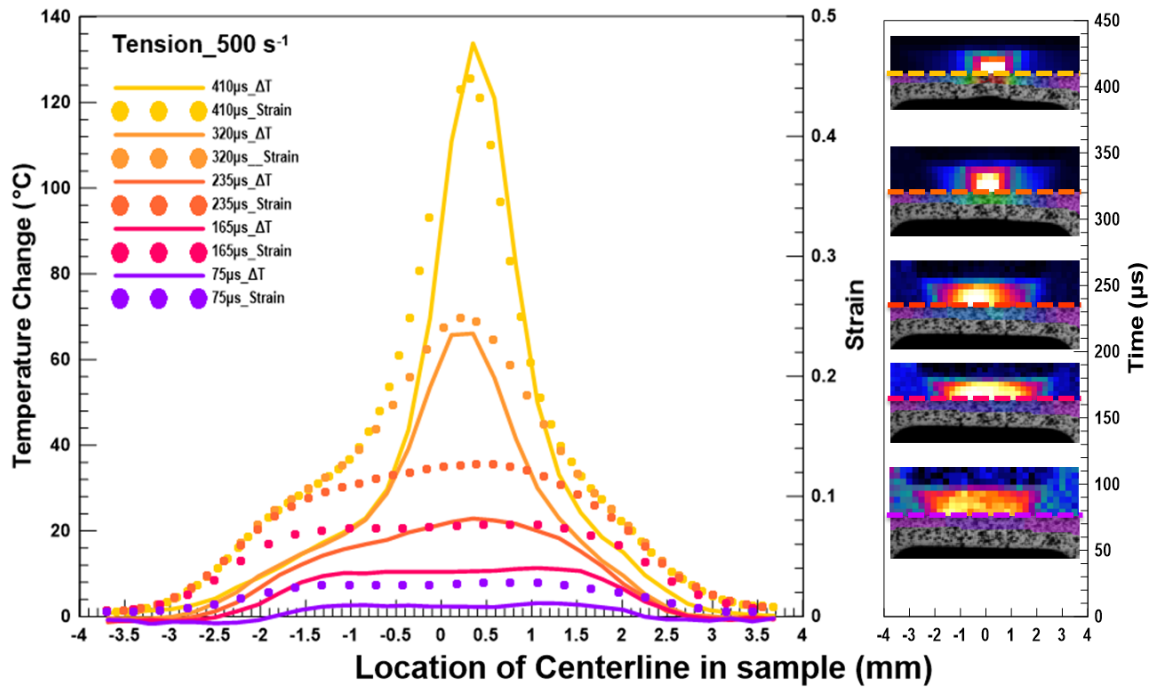


Figure 3-37 The temperature-strain distribution from tension at 500.0 s^{-1}

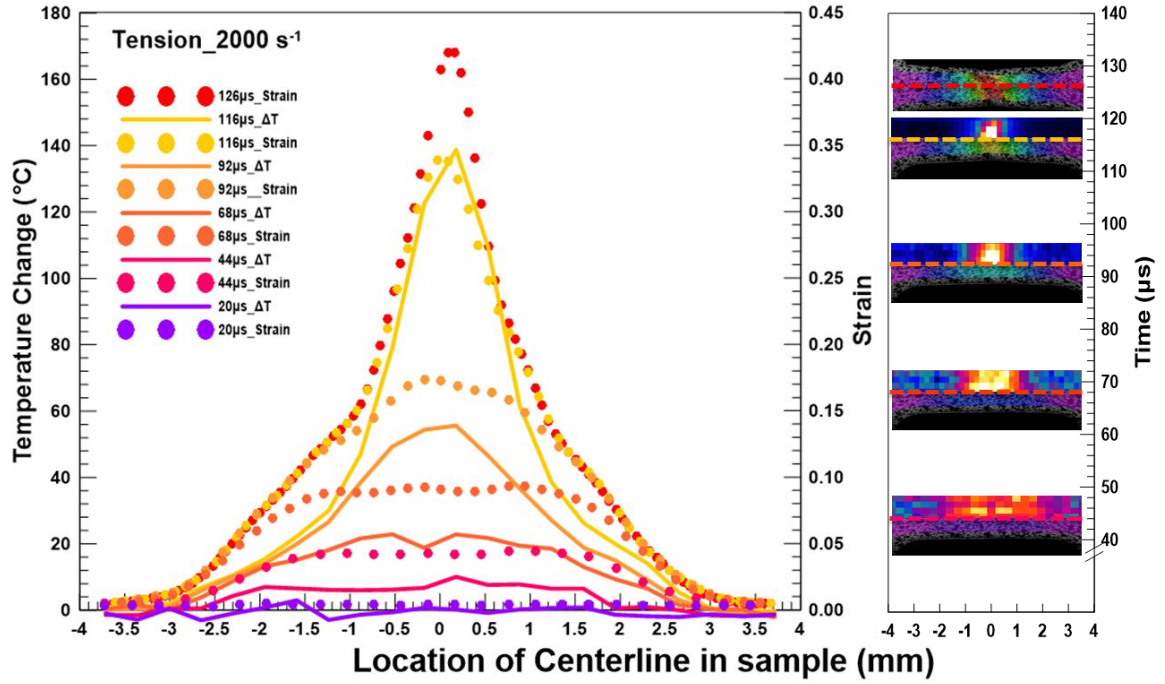


Figure 3-38 The temperature-strain distribution from tension at 2000.0 s⁻¹

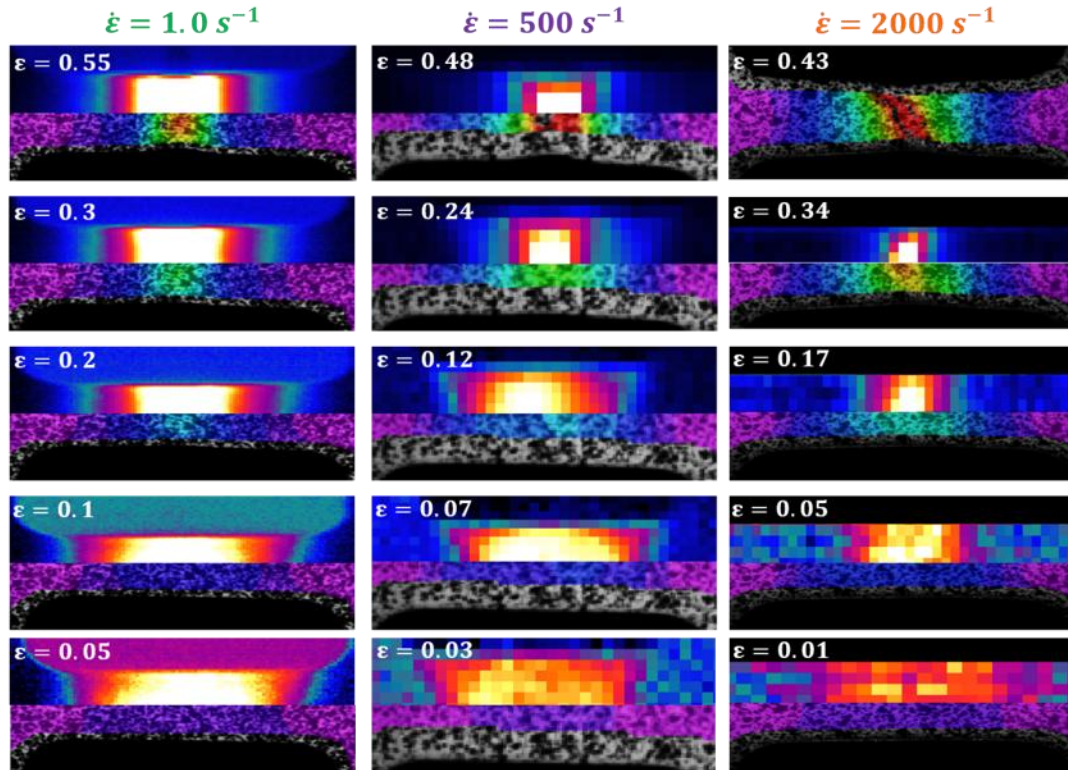


Figure 3-39 Evolution of synchronized strain and temperature distributions in 410 stainless steel at 1.0 s⁻¹~2000 s⁻¹

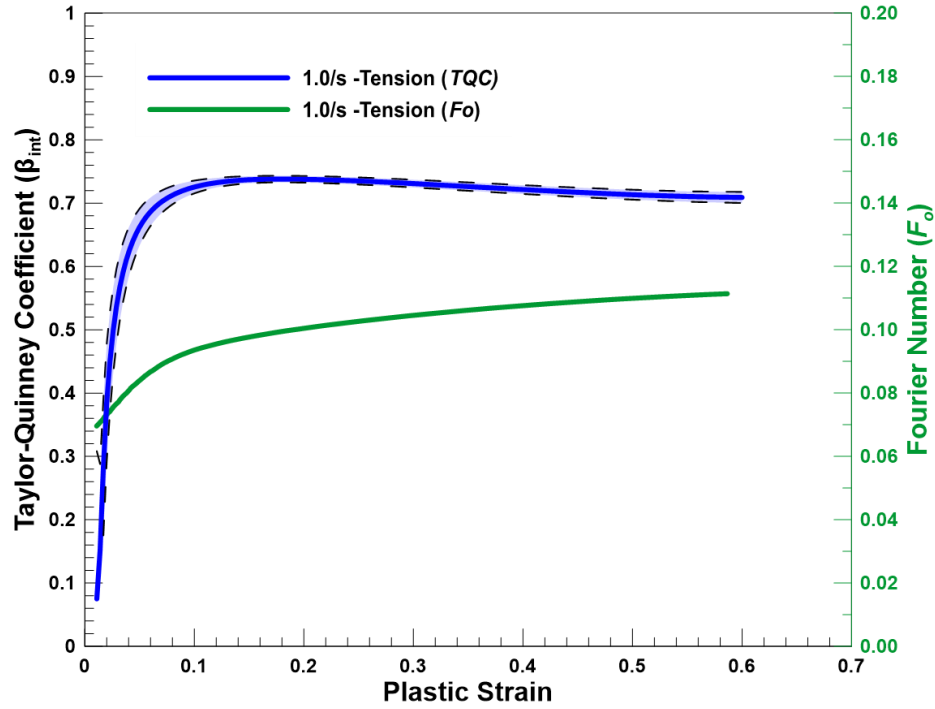


Figure 3-40 Taylor-Quinney Coefficient-Fourier Number versus plastic strain curves from tension test at 1.0 s^{-1}

3.4.2 Compression

The evolution of the Taylor-Quinney Coefficient (TQC) under compressive loading follows three stage progressions similar to the tensile response: the onset of plastic deformation, deformation progress, and saturation. As shown in Figure 3-41, the TQC initially rises during the elastic-to-plastic transition, followed by a stabilizing phase as plastic work is converted into thermal energy. However, the influence of the strain rate on the saturation stage reveals distinct differences compared to tension due to the evolving geometry of the compression specimens. At high strain rates (2000 s^{-1} to 5000 s^{-1}), the TQC remains stable at approximately 0.7 during the saturation stage. In contrast, at the intermediate strain rate of 1.0 s^{-1} , the TQC exhibits a clear decay from 0.7 to 0.3.

At low ($10^{-3} \text{ s}^{-1} \sim 10^{-2} \text{ s}^{-1}$) and intermediate (1.0 s^{-1}) strain rates, the TQC reveals a decay during the saturation stage. This trend is a result of non-adiabatic conditions where heat loss to the specimen body and the platens become dominant. Recalling the thermal-mechanical data in Figure 3-20, Figure 2-24, and Figure 2-25, the temperature increase in compression is flatter than that observed in tension across all strain rates. This flatter temperature profile is more evident at the intermediate rate than in the high-rate tests, providing clear evidence of heat loss. As the specimen height decreases and the cross-sectional area increases, the path for thermal conduction into the platens, and the sample is shortened while the contact area increases. This heat dissipation causes the calculated TQC to drop sharply as deformation progresses.

The validity of the adiabatic assumption for the intermediate strain rate tests (1.0 s^{-1}) is evaluated by comparing the evolution of strain and temperature. As shown in Figure 3-21, the distributions are consistent until a strain of approximately 0.05~0.1, indicating that heat generation dominates over thermal diffusion in this initial stage. At a strain of 0.2 ($t = 0.4 \text{ s}$), the temperature distribution develops a characteristic parabolic profile despite a uniform strain field. In Figure 3-42, while the maximum local strain at strain rate of 1.0 s^{-1} occurs at the bottom edge of the sample, the temperature distribution does not follow the same pattern. This divergence confirms the onset of heat loss to the specimen and platens, making the TQC calculations less reliable beyond this strain threshold. Furthermore, Figure 3-43 shows that the Fourier number (F_o) reaches 0.1 at a strain of 0.05 and increases to 0.17 by a strain of 0.1. These metrics demonstrate the same trend of

increasing heat loss. Consequently, TQC values derived after a strain of 0.05 in these cases may be considered invalid, as they no longer reflect a purely adiabatic energy conversion.

The correlation between mechanical and thermal data at high strain rate is illustrated in Figure 3-39, Figure 3-44 and Figure 3-45. Unlike the tension tests, where heat is concentrated at a localization spot, the compression specimens display a temperature distribution that remains highly consistent with the strain distribution across the whole specimen. At the high strain rate, the TQC remains stable throughout the saturation stage, indicating that the timescale of the deformation is sufficiently short to prevent heat loss, thus maintaining adiabatic conditions.

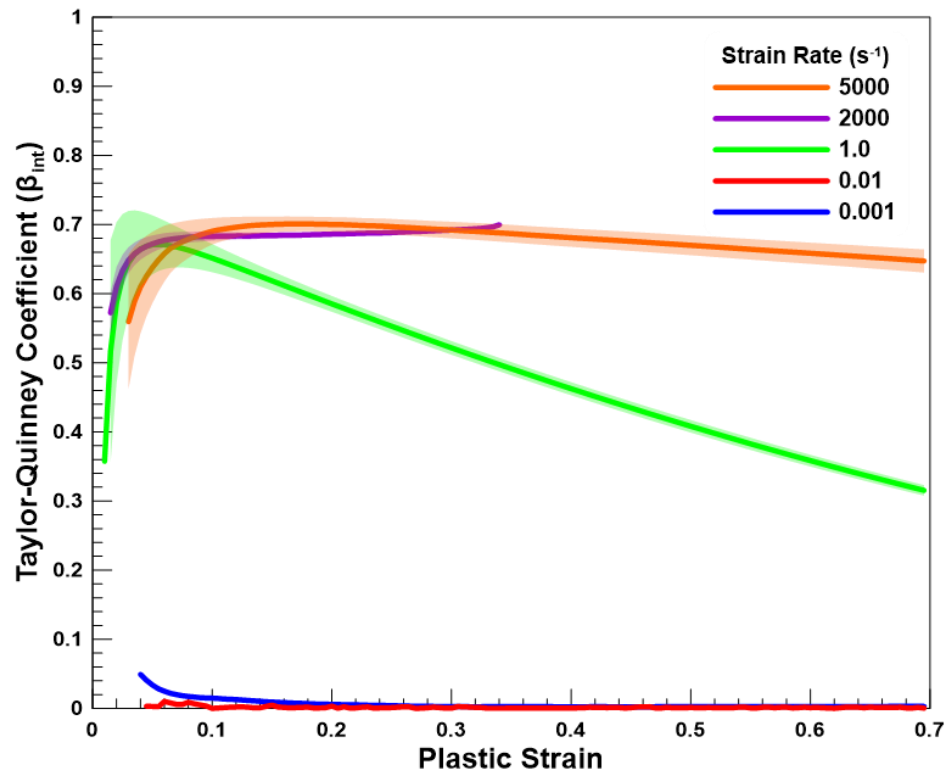


Figure 3-41 Taylor-Quinney Coefficient versus plastic strain curves from compression tests in various strain rates

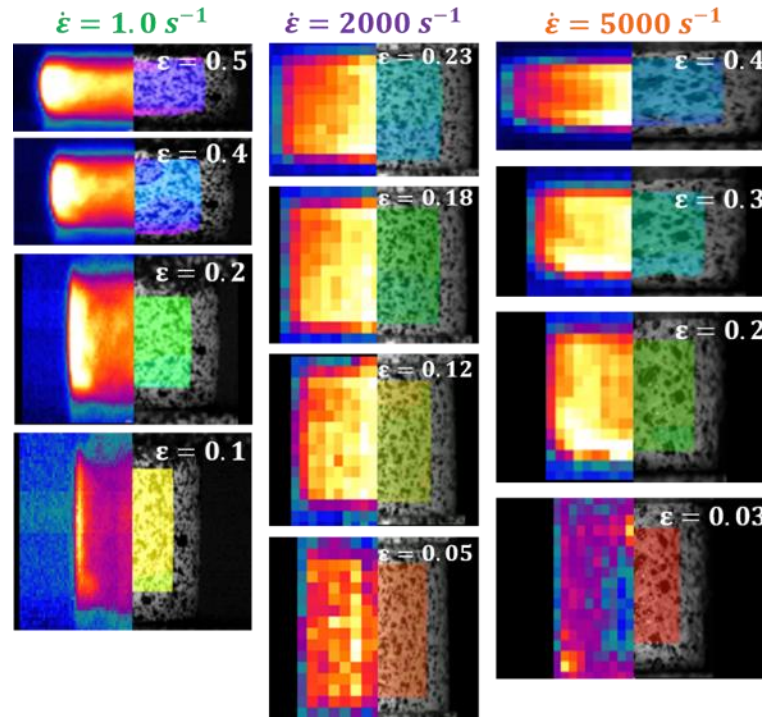


Figure 3-42 Evolution of synchronized strain and temperature distributions in 410 stainless steel at 1.0 s^{-1} ~ 5000 s^{-1} .

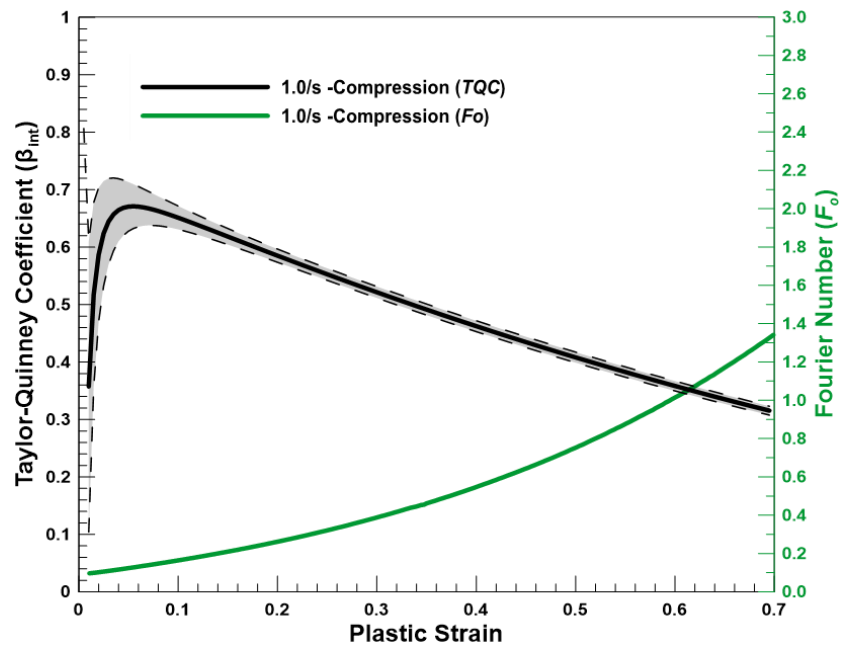


Figure 3-43 Taylor-Quinney Coefficient-Fourier Number versus plastic strain curves from compression test at 1.0 s^{-1}

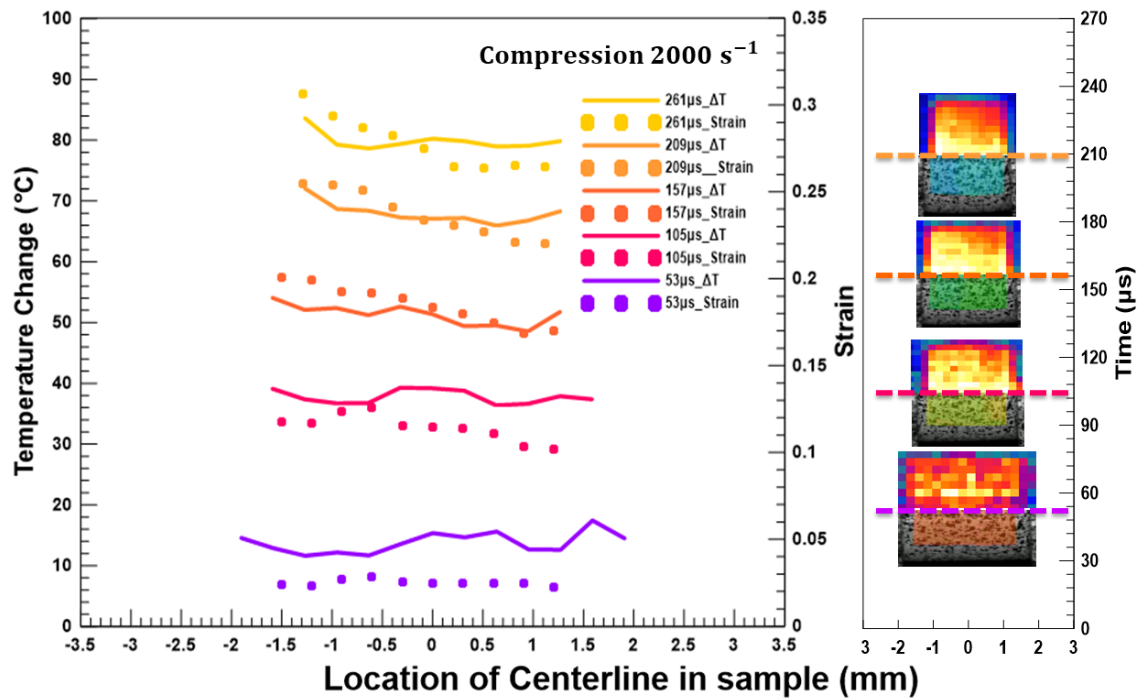


Figure 3-44 The temperature-strain distribution from compression at 2000.0 s⁻¹

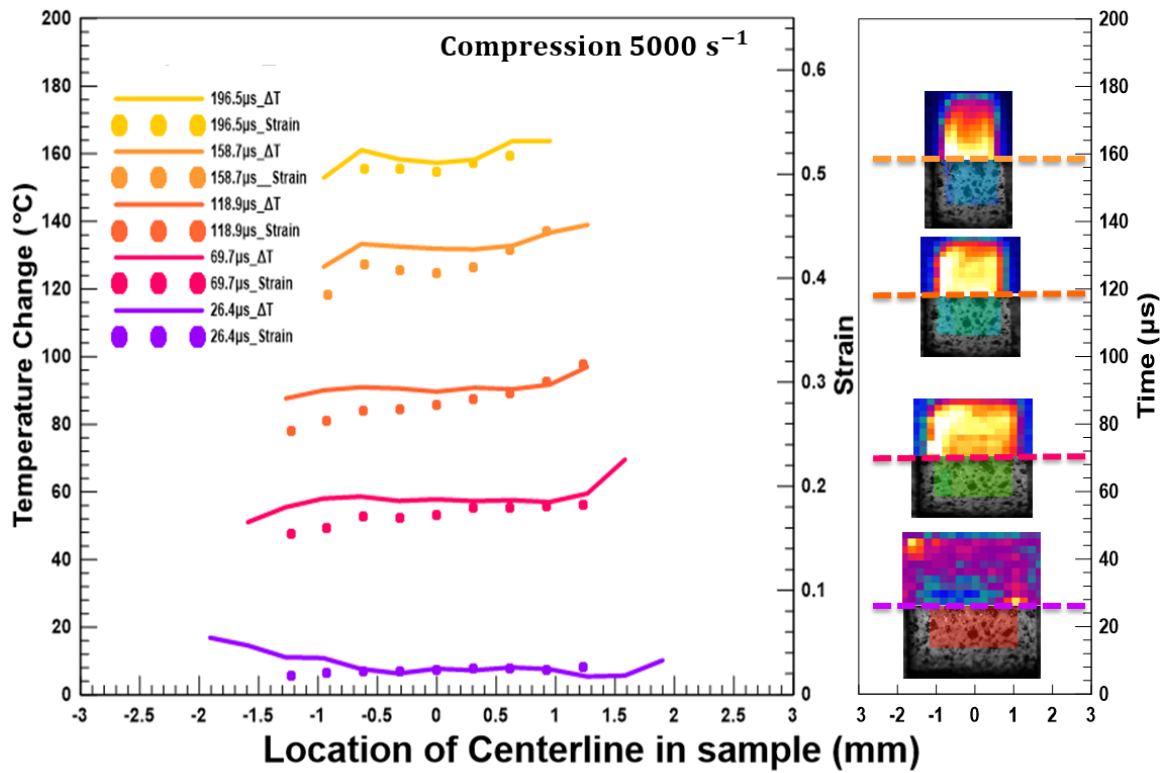


Figure 3-45 The temperature-strain distribution from compression at 5000.0 s⁻¹

3.4.3 Torsion

The evolution of the Taylor-Quinney Coefficient (TQC) under torsional loading follows the characteristic progression observed in tension and compression: the onset of plastic deformation, deformation progress, and saturation. As illustrated in Figure 3-46, the TQC initially rises during the transition to plastic flow before reaching a stable level. At high strain rates, the TQC remains stable between 0.86 and 0.94 during the saturation stage. In contrast, the TQC exhibits a clear decay from 0.88 to 0.6 at the intermediate strain rate of 1.0 s^{-1} .

Similar to the behavior observed in compression, this decay is driven by heat loss to the specimen shoulder. Recalling the thermal-mechanical data in Figure 3-20, the temperature rise in torsion is flatter than that observed in tension. Because the torsional gauge section is thin-walled and has a short gauge length, it possesses a large surface area that leads rapid thermal dissipation into the solid shoulders. At strain rate of 1.0 s^{-1} , Figure 3-23 reveals the temperature increase observed at the gauge section edges despite minimal local strain. Indicating heat conducts away almost as quickly as the plastic work generates.

Figure 3-48 shows that Fourier number (F_o) reaches 1.0 at a strain of 0.03 and increases to 1.5 by a strain of 0.06. This confirms that heat loss occurs from the very beginning of the test. However, the apparent decay of the TQC evolution curve does not manifest until a strain of 0.06. This threshold corresponds to the ultimate shear strength and the beginning of the saturation stage. The peak TQC of 0.88 measured at strain rate of 1.0 s^{-1} represents a combination of the intrinsic material energy conversion and a small amount of heat loss. To quantify the thermal dissipation at the intermediate strain rate (1.0 s^{-1}), the stable TQC

measured from the high-strain-rate tests (0.9 ± 0.04) is established as the baseline for purely adiabatic energy conversion. At a shear strain of 0.06, the peak TQC for the 1.0 s^{-1} strain rate test is measured at 0.88 ± 0.02 . By comparing these two cases, it can be inferred that the magnitude of heat loss ranges from 0% to 8% (a coefficient difference of 0 to 0.08).

In contrast, at high strain rates, the TQC remains stable because the deformation happens too fast for the heat to dissipate. Recalling the stress-temperature change curves in Figure 3-24 and Figure 3-25, the temperature rise in torsion is near-linear as that observed in tension. At a shear strain rate of 1000 s^{-1} (Figure 3-49), the temperature distribution is consistent with the strain distribution. This shows that the heat generation is tracking the plastic work across the gauge section. The shear strain localization is captured by the DIC even though its field of view is narrower than the IR image. At the higher shear strain rate of 4500 s^{-1} (Figure 3-50), the temperature profile becomes sharper than the strain profile. This indicates that the heat is trapped at the localization spot, leading to adiabatic heating that accumulates faster than the strain spreads.

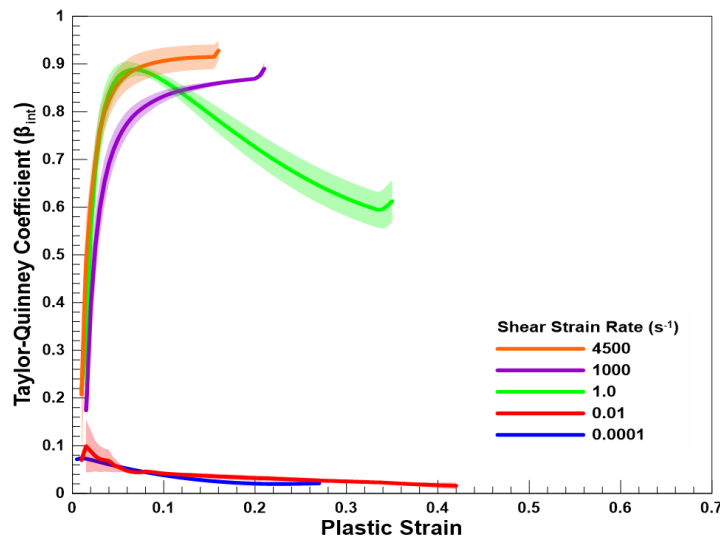


Figure 3-46 Taylor-Quinney Coefficient versus plastic strain curves from torsion tests in various strain rates

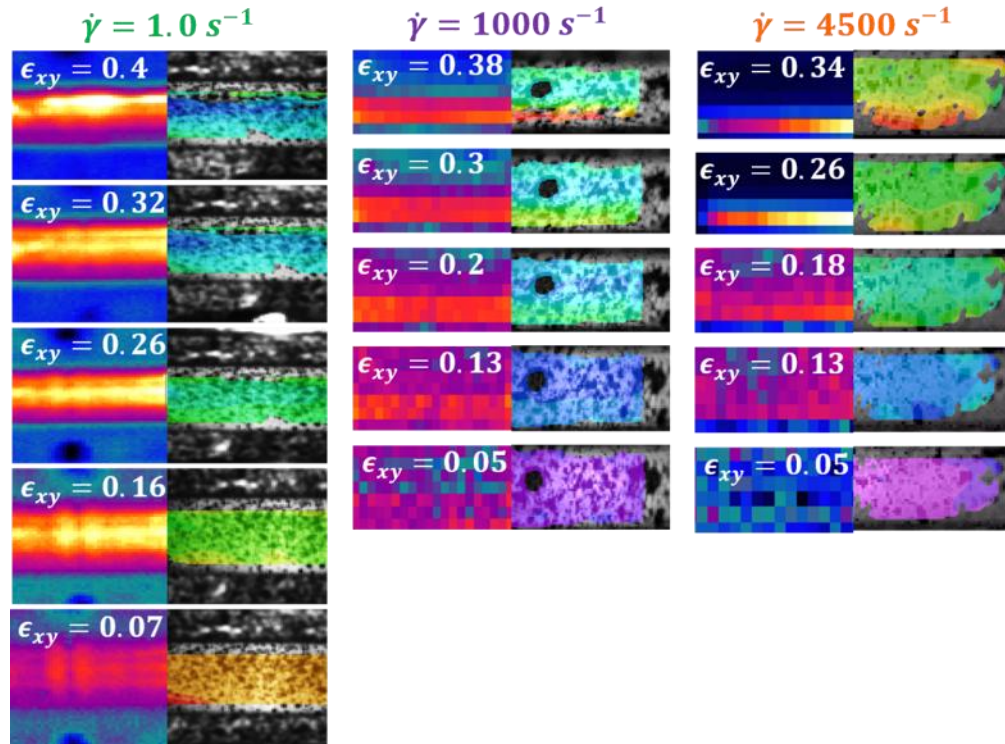


Figure 3-47 Evolution of synchronized strain and temperature distributions in 410 stainless steel at 1.0 s^{-1} ~ 4500 s^{-1} .

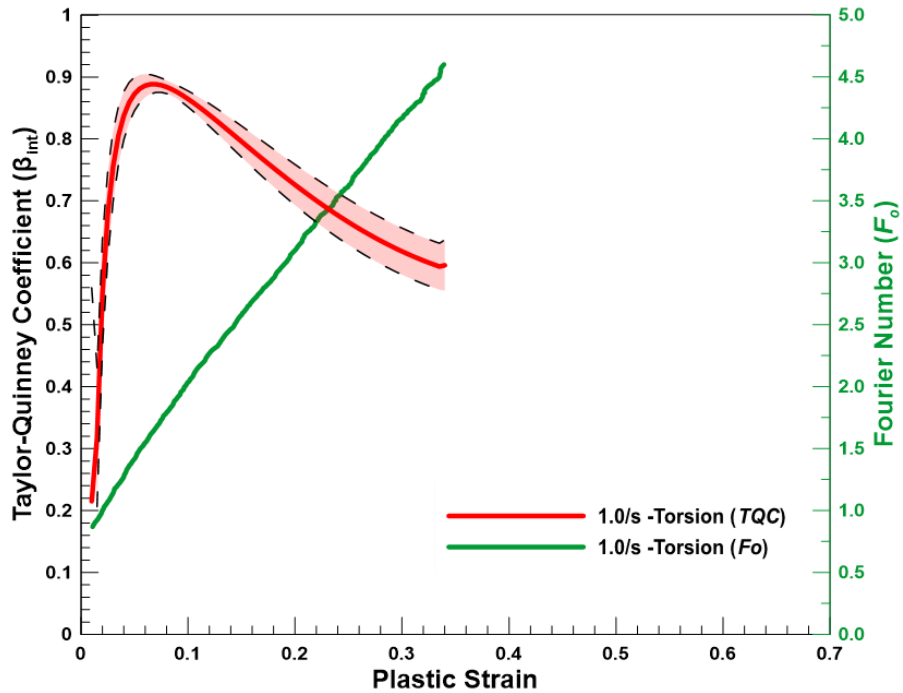


Figure 3-48 Taylor-Quinney Coefficient-Fourier Number versus plastic strain curves from torsion test at 1.0 s^{-1}

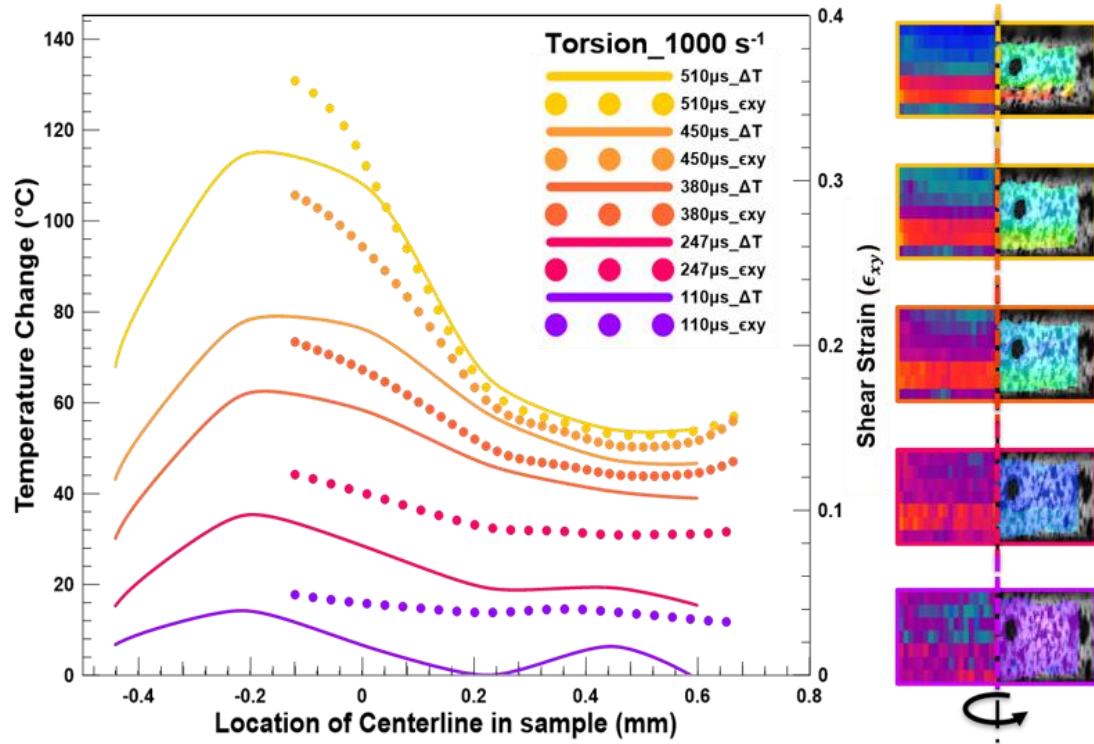


Figure 3-49 The temperature-strain distribution from torsion at 1000.0 s⁻¹

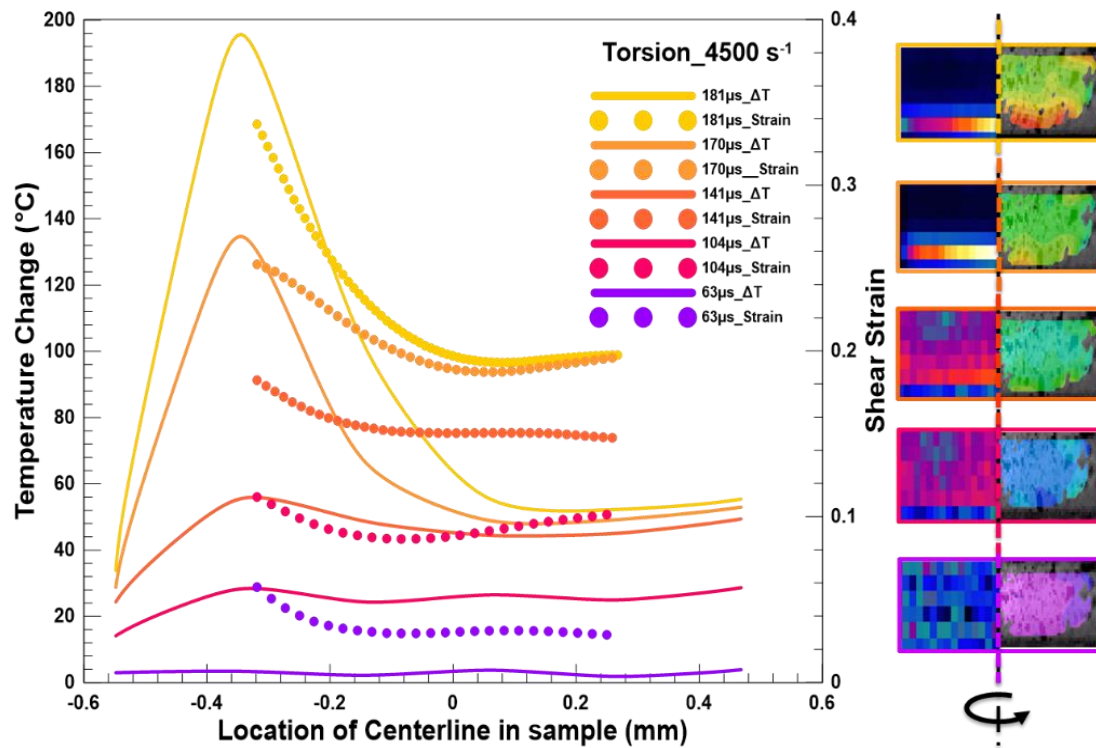


Figure 3-50 The temperature-strain distribution from torsion at 4500.0 s⁻¹

3.4.4 Discussion

Analysis of the high-rate experimental data indicates that there is no clear strain rate dependency of the TQC for 410 SS (Figure 3-35, Figure 3-41, Figure 3-46). However, a loading mode dependency is observed (Figure 3-51~ Figure 3-53). Torsion exhibits the highest energy conversion efficiency, with TQC values ranging between 0.86 and 0.94, followed by tension and compression.

The evolution of the Fourier Number (F_o) serves as a mechanical explanation for the observed heat loss trends across different geometries: In tension, the F_o slope is relatively flat. As deformation progresses, the path for heat loss increases due to specimen elongation, which helps maintain a localized thermal state. In compression, F_o increases with an exponential shape. As the specimen deforms, the thermal path to the platens is shortened while the contact surface area increases, accelerating heat dissipation. In torsion, The F_o slope is linear. Because the gauge section maintains a constant cross-section and thickness, the path for heat loss remains consistent throughout the deformation process.

The critical Fourier numbers corresponding to the onset of TQC invalidity, determined from the strain–temperature distribution analysis, are approximately 0.1 for tension, 0.2 for compression, and 1.7 for torsion. This indicates that the critical Fourier number depends on the specimen geometry and heat-transfer path. It should be noted that the Fourier number is derived from a one-dimensional heat conduction assumption, whereas the present experiments involve three-dimensional heat transfer. The Fourier number remains a useful scaling parameter for evaluating adiabatic condition, but its value is expected to vary with specimen geometry and heat-transfer conditions.

At high strain rates, the adiabatic assumption is verified. As shown in Table 3-2, the calculated Fourier numbers are far smaller than 1.0, indicating that the timescale of deformation is several orders of magnitude faster than the timescale of thermal diffusion. This adiabatic condition is further confirmed by the synchronized strain-temperature distribution maps, which show that heat remains trapped within the plastic zone for high-rate tests.

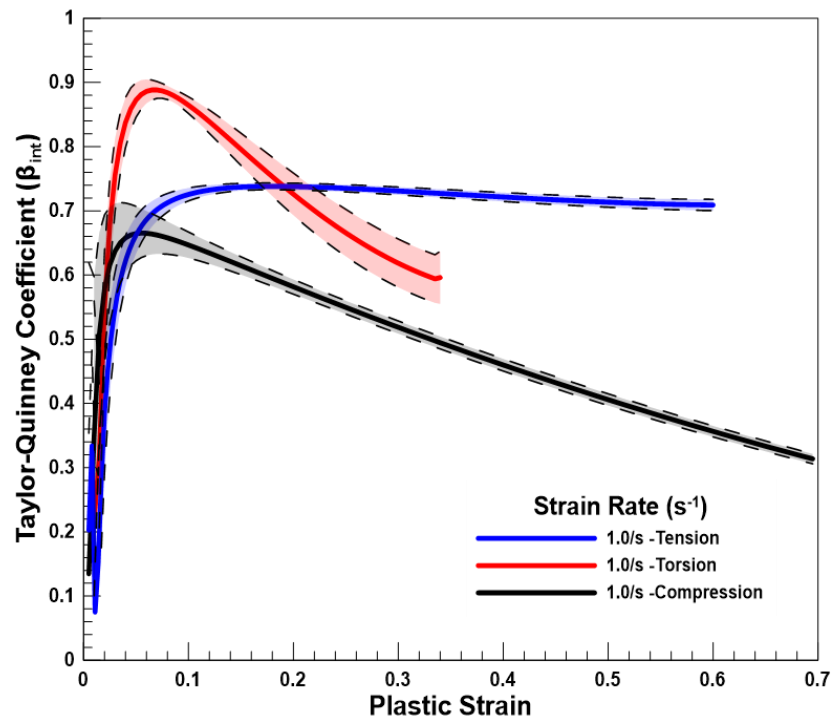


Figure 3-51 Taylor-Quinney Coefficient versus plastic strain curves from various loadings at 1.0 s⁻¹

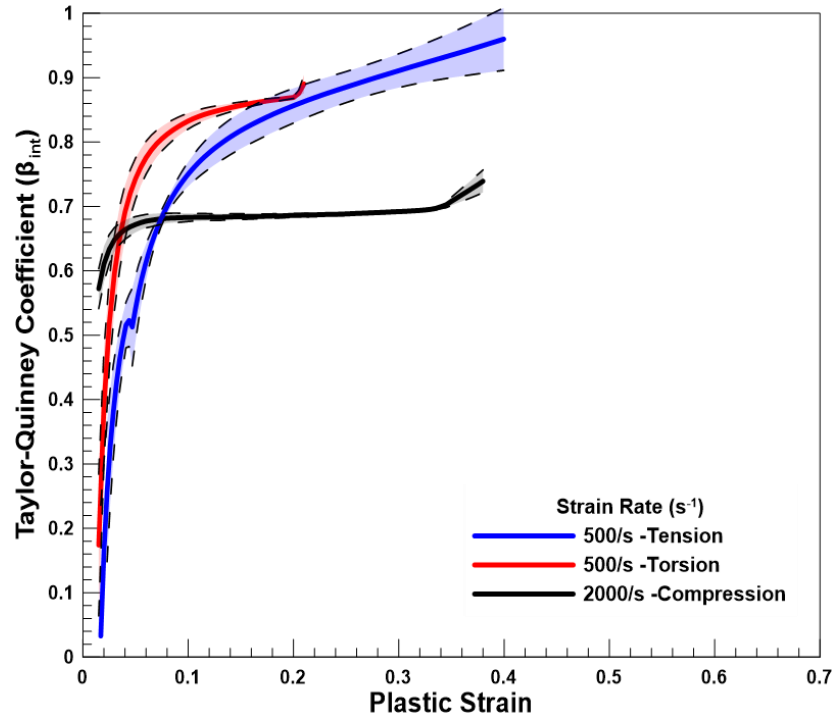


Figure 3-52 Taylor-Quinney Coefficient versus plastic strain curves from various loadings at $500 \text{ s}^{-1} \sim 2000 \text{ s}^{-1}$

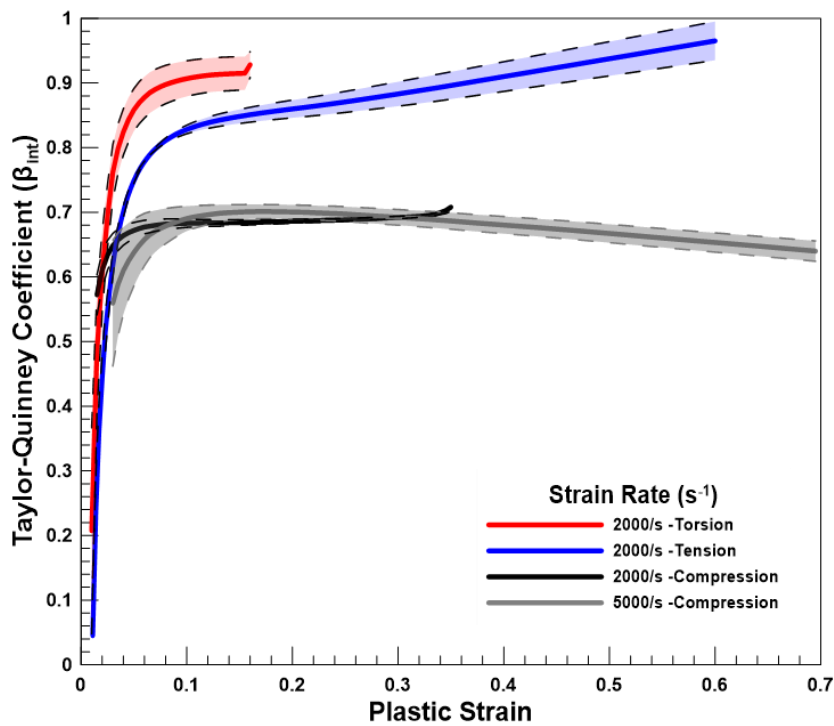
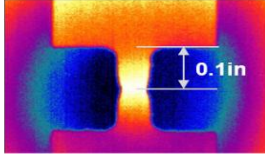
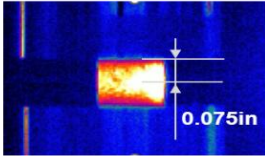
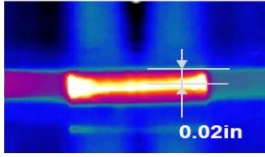


Figure 3-53 Taylor-Quinney Coefficient versus plastic strain curves from various loadings at $2000 \text{ s}^{-1} \sim 5000 \text{ s}^{-1}$

Table 3-2 Fourier Number (Fo) from various loadings at intermediate and high strain rate

Loading Mode	Characteristic Length (L)	Strain Rate (s^{-1})	Plastic Strain at Valid TQC Limit	Fourier Number (Fo)
Tension		1	0.2	0.1
		500	End of test	0.00042
		2000	End of test	0.00013
Compression		1	0.1	0.2
		2000	End of test	0.00063
		5000	End of test	0.00037
Torsion		1	0.08	1.7
		1000	End of test	0.01391
		4500	End of test	0.00483

3.5 Constitutive Material Modeling

The constants of the constitutive equation of Johnson–Cook (J–C) are evaluated for modeling the response of the 410 stainless steel. The traditional J–C formulation, Eq. 3-1, consists of three terms with seven constants. The first term corresponds to strain hardening and its three constants A , B and n are typically determined from a low strain rate, room-temperature, stress strain curve. The second term corresponds to the effect of strain-rate and its two constants C and $\bar{\epsilon}_0$ are determined from stress strain curves from tests at different strain rates. The tests are conducted at room temperature, and it is assumed that the specimens' temperature does not change during the test. The form of the equation corresponds to a straight line in a plot of stress versus log strain rate. The third term corresponds to the effect of temperature. The constant m is determined from stress strain curves of tests (all at the same strain rate) at different temperatures, and T_{melt} is the melting

temperature of the material. The three terms in the Johnson–Cook equation are independent of each other, which means that the strain hardening, and temperature and strain rate effects are not coupled. In other words, strain hardening is the same at all strain rates and all temperatures and tests at all strain rates are isothermal.

$$\bar{\sigma}(\bar{\epsilon}^p, \dot{\bar{\epsilon}}^p, T) = [A + B(\bar{\epsilon}^p)^n] \left(1 + C \ln \frac{\dot{\bar{\epsilon}}^p}{\dot{\bar{\epsilon}}_0} \right) \left[1 - \left(\frac{T - T_{room}}{T_{melt} - T_{room}} \right)^m \right] \quad 3-1$$

Results from preliminary testing show that the effects of strain hardening, strain rate and temperature are coupled. For example, room-temperature tests at different strain rates, displayed in Figure 3-54(a), show that the strain hardening is not the same at all strain rates. Measurements of specimen temperature during the tests show that the specimen's temperature increases during the test, unless the test is conducted at an extremely low strain rate (Figure 3-54(b)). Simulations of the experiments with the classical Johnson–Cook equation, also displayed in Figure 3-54, show significant discrepancies between the testing data and the model at high strain rates.

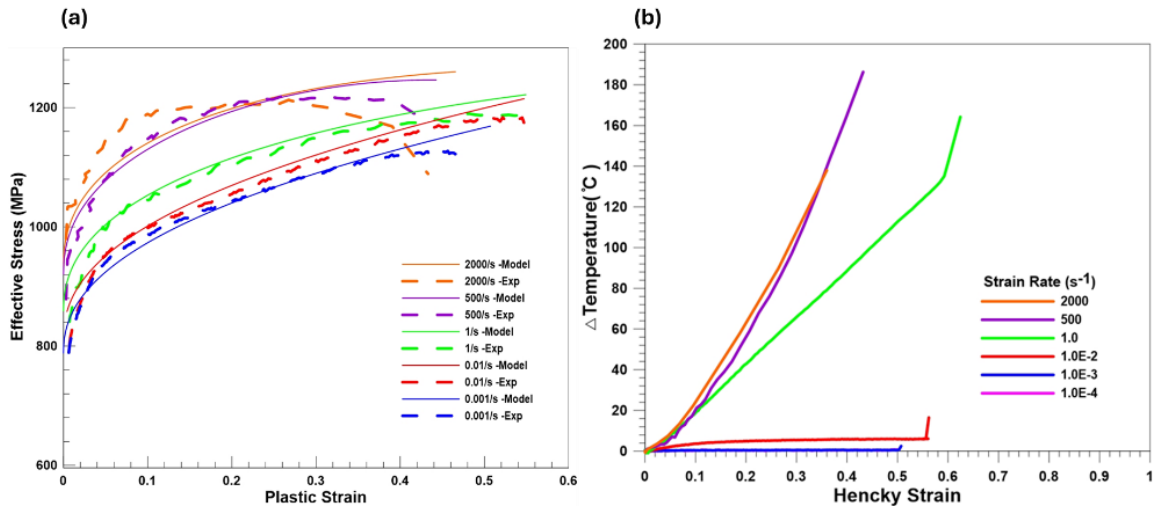


Figure 3-54 Illustration of original JC model concept for strain rate and thermal softening effect. (a) Strain rate from 0.001 to 2000 s⁻¹, (b) Temperature increasing

One objective of the research is to modify the strain rate sensitivity term and temperature term in the classical Johnson–Cook model. The plots of stress versus log strain rate and stress versus temperature, Figure 3-55, shows a linear relationship between the stress and log strain rate at strain rate up to about 1 s^{-1} . At higher strain rates there is an increase in strain rate sensitivity with increasing strain rate. And the relationship between stress and temperature is also non-linear. To address this observed behavior, the traditional J-C model will be modified as shown in Eq.3-2, where $T^* = T - T_{room}/T_{melt} - T_{room}$.

$$\bar{\sigma}(\bar{\epsilon}^p, \dot{\epsilon}^p, T) = [A + B(\bar{\epsilon}^p)^n] \left(1 + C_1 e^{C_2 \ln \frac{\dot{\epsilon}^p}{\dot{\epsilon}_0}} \right) \{1 - [m_1(T^*)^3 + m_2(T^*)^2 + m_3(T^*)]\} \quad 3-2$$

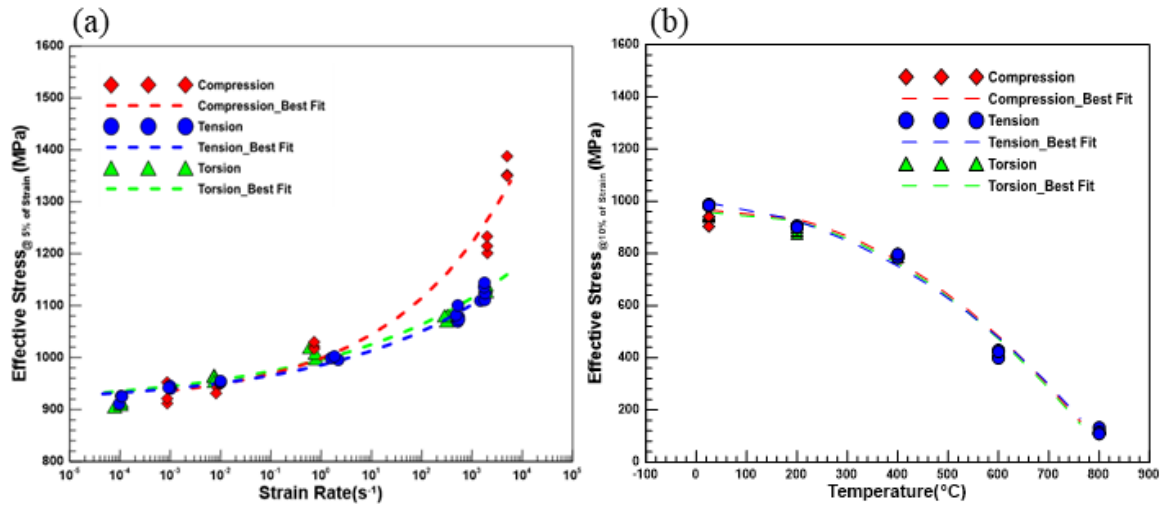


Figure 3-55 Illustration of concept of (a) strain rate sensitivity, (b) Temperature sensitivity

3.5.1 Tension

The modified Johnson–Cook (J–C) model demonstrates high accuracy in replicating the tensile response of 410 stainless steel across the investigated strain rate spectrum. As illustrated in Figure 3-56, the model replicates the sharp increase in flow stress observed

as the rate transitions from quasi-static (strain rate of 10^{-4} s^{-1}) to the dynamic (strain rate of 2000 s^{-1}) by replacing the traditional linear strain rate sensitivity term with a non-linear exponential formulation. The integration of a polynomial temperature term further enhances the model's accuracy by accounting for non-linear adiabatic softening during high-rate deformation. To evaluate the quality of the fit, the Mean Absolute Percentage Error (MAPE) was calculated to be 1.3012%, with an R-squared value of 0.9672 (Figure 3-57).

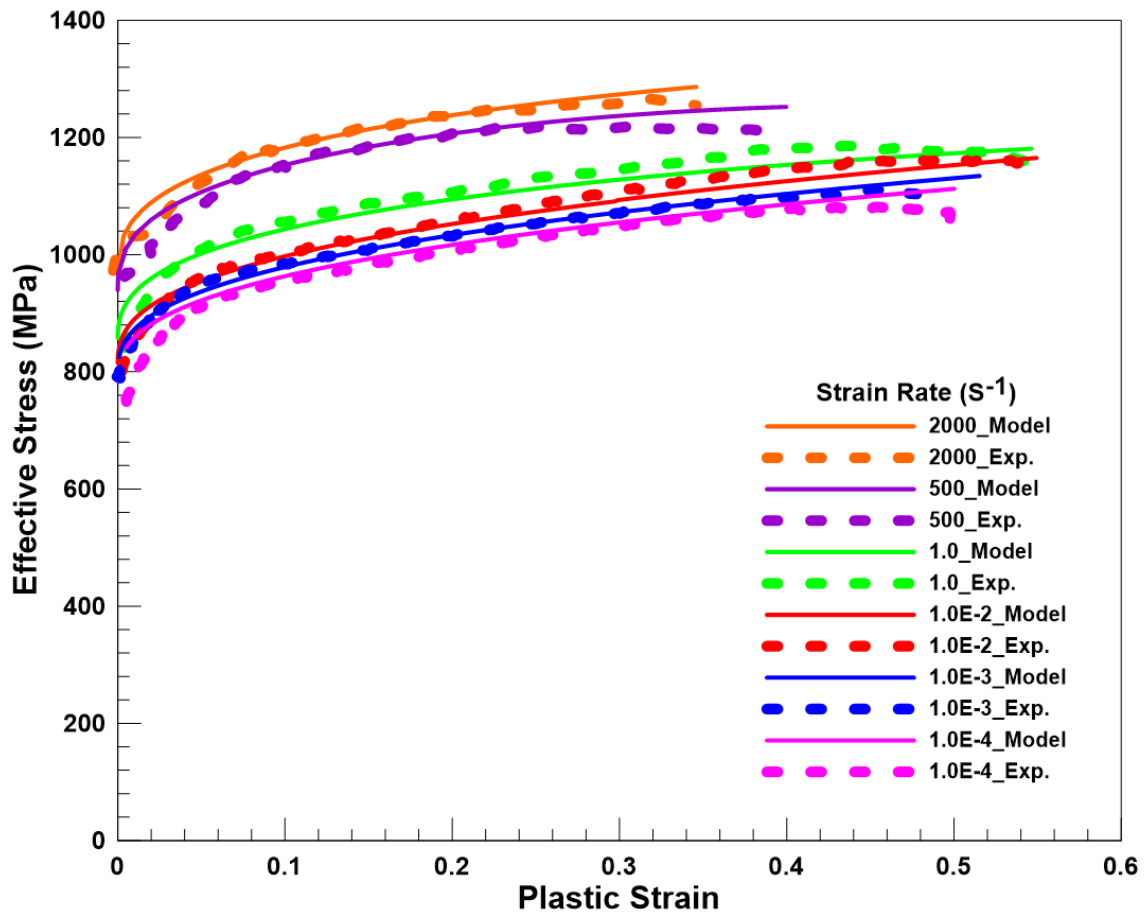


Figure 3-56 Comparison of stress-strain curves between experiment and modified JC model from tension test at various strain rate

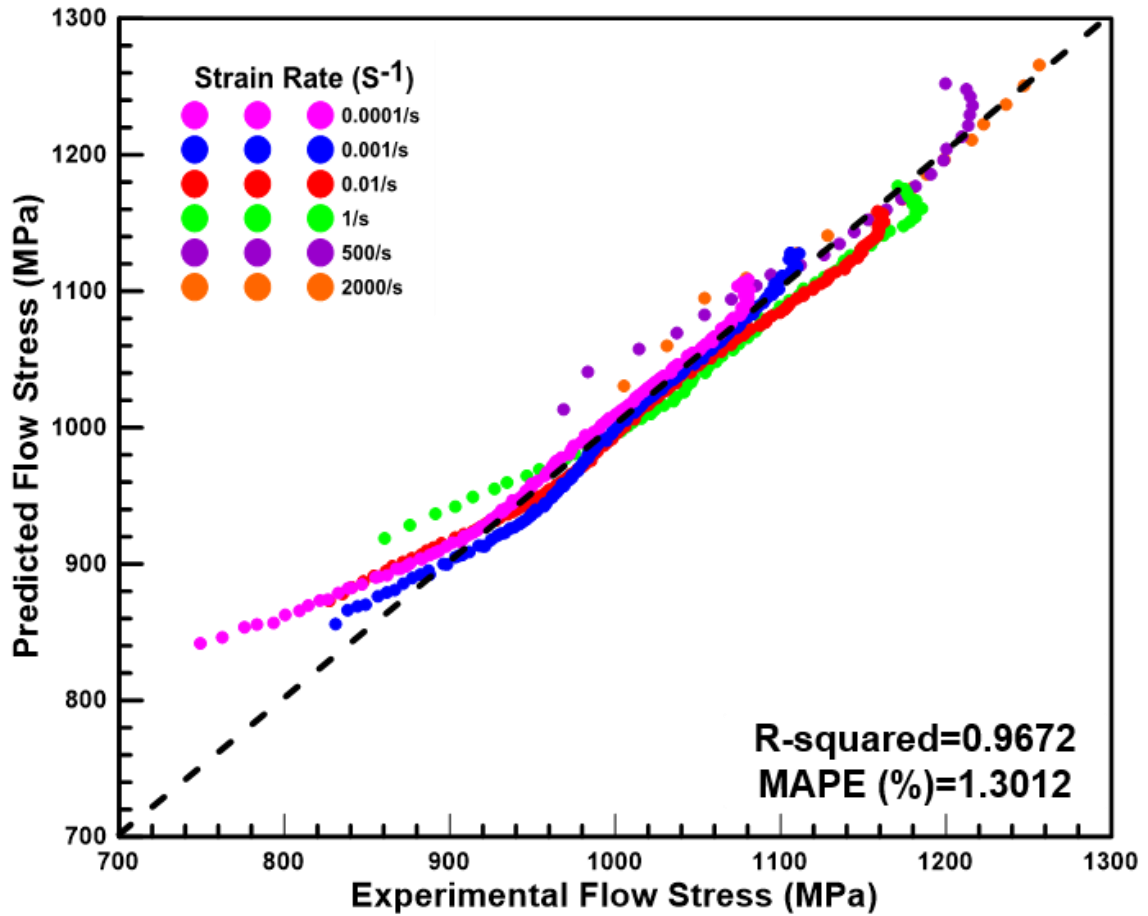


Figure 3-57 Comparison between predicted data and experimental data from tension test at various strain rate

3.5.2 Compression

The modified Johnson–Cook (J–C) model demonstrates high predictive accuracy for the compressive response of 410 stainless steel, consistent with the results observed in tension. As shown in Figure 3-58, the model tracks the flow stress behavior across a wide range of strain rates, from quasi-static conditions (strain rate of 10^{-3} s^{-1}) to high-strain-rate dynamic loading (strain rate of 5000 s^{-1}). The statistical validation in Figure 3-59 confirms this performance, with a calculated R-squared value of 0.9763 and a Mean Absolute Percentage Error (MAPE) of 1.1498%.

The fitting in this section is attributed to the non-linear exponential strain rate term, which captures the material's increased sensitivity at high dynamic rates. In the original J–C model, a linear logarithmic term often fails to represent the rapid rise in flow stress at strain rates exceeding 2000 s^{-1} , but the modified version aligns with the experimental data. Furthermore, the polynomial temperature term accounts for the adiabatic heating and thermal softening that occur during high-rate compression, where the specimen geometry and contact with the platens create a unique thermal environment compared to tension.

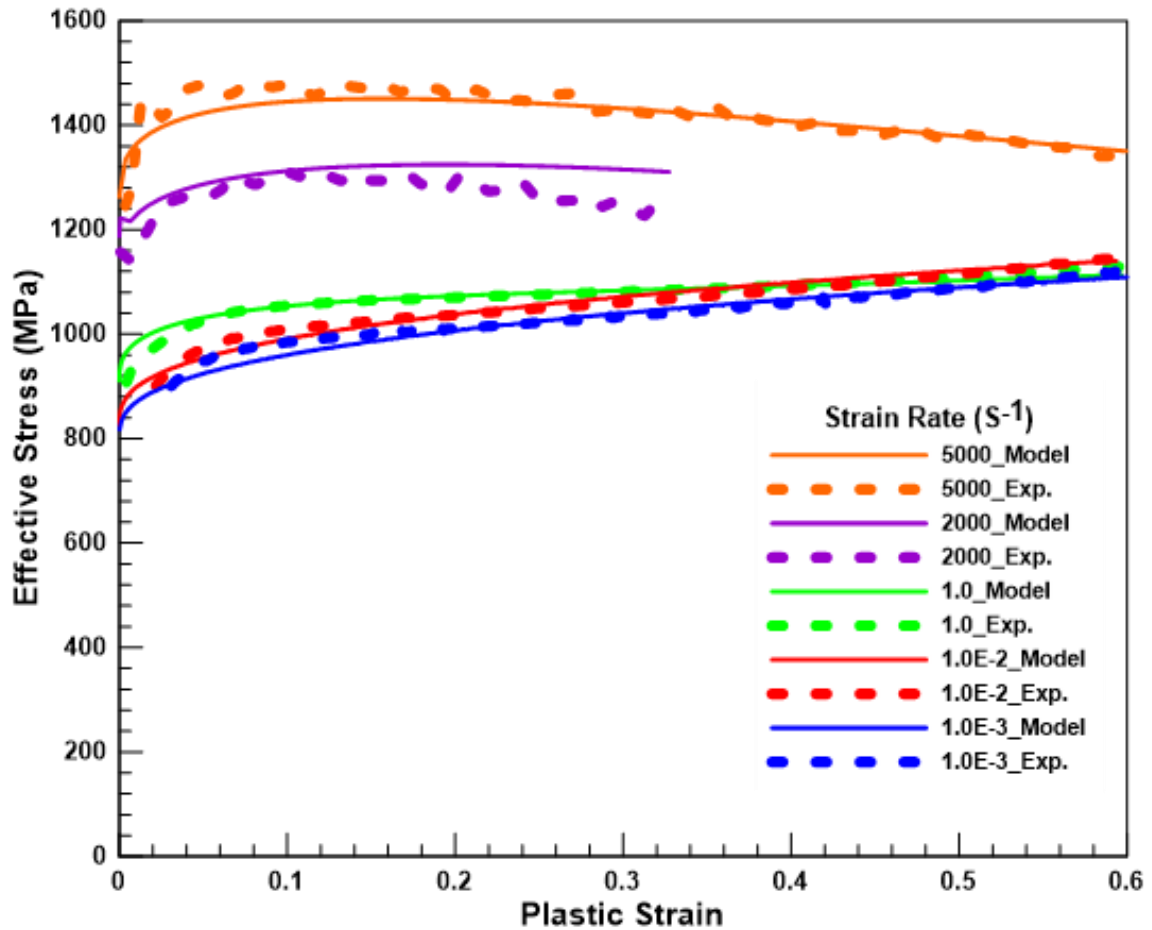


Figure 3-58 Comparison of stress-strain curves between experiment and modified JC model from compression test at various strain rate

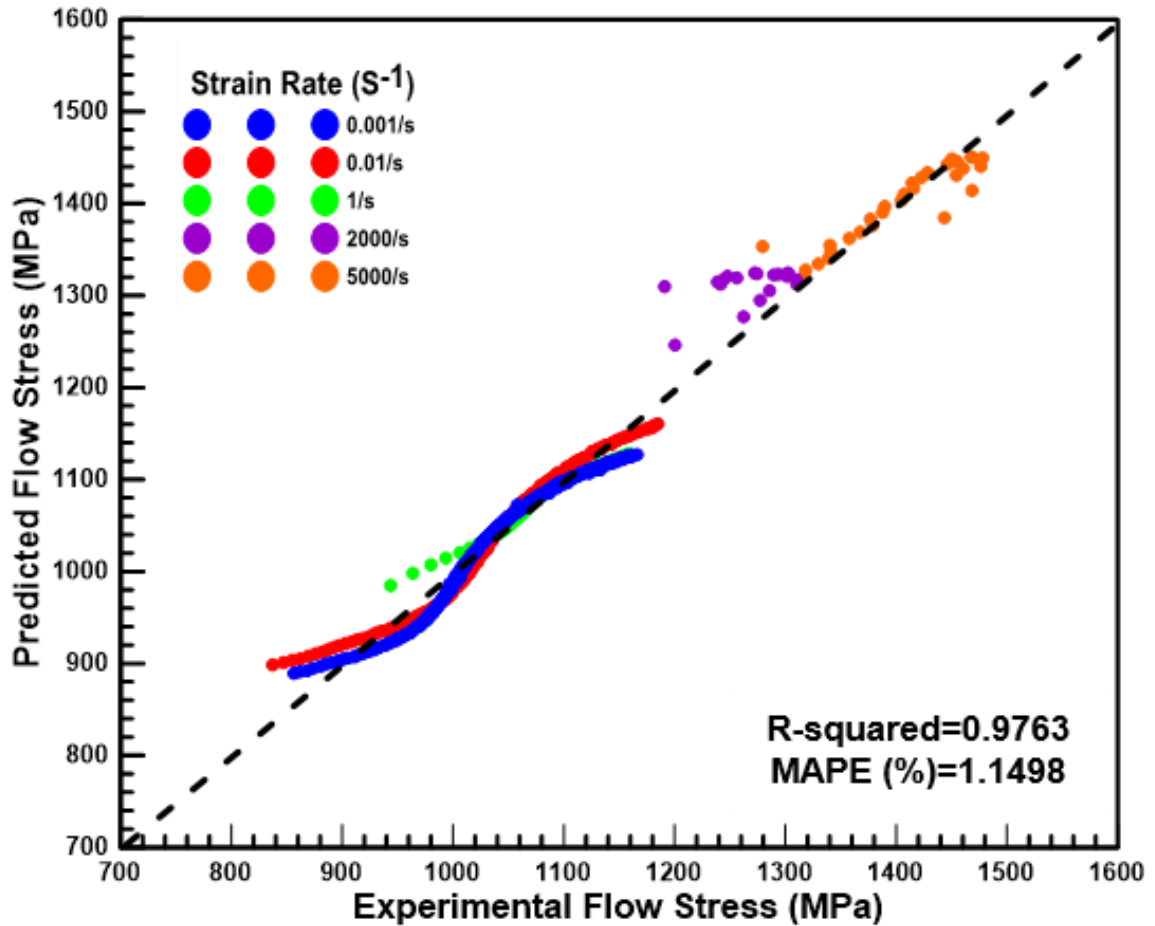


Figure 3-59 Comparison between predicted data and experimental data from compression test at various strain rate

3.5.3 Torsion

The modified Johnson–Cook (J–C) model is further evaluated under torsional loading, as presented in Figure 3-60. The model replicates the shear stress-strain response for 410 stainless steel at both quasi-static (strain rates in the range of $10^{-4} \text{ s}^{-1} \sim 10^{-2} \text{ s}^{-1}$), intermediate (strain rate of 1.0 s^{-1}), and high strain rates (strain rates in the range of $1000 \text{ s}^{-1} \sim 4500 \text{ s}^{-1}$). The correlation between the experimental data and the model predictions is quantified, with an R-squared value of 0.9347 and a Mean Absolute Percentage Error (MAPE) of 1.2528% (Figure 3-61).

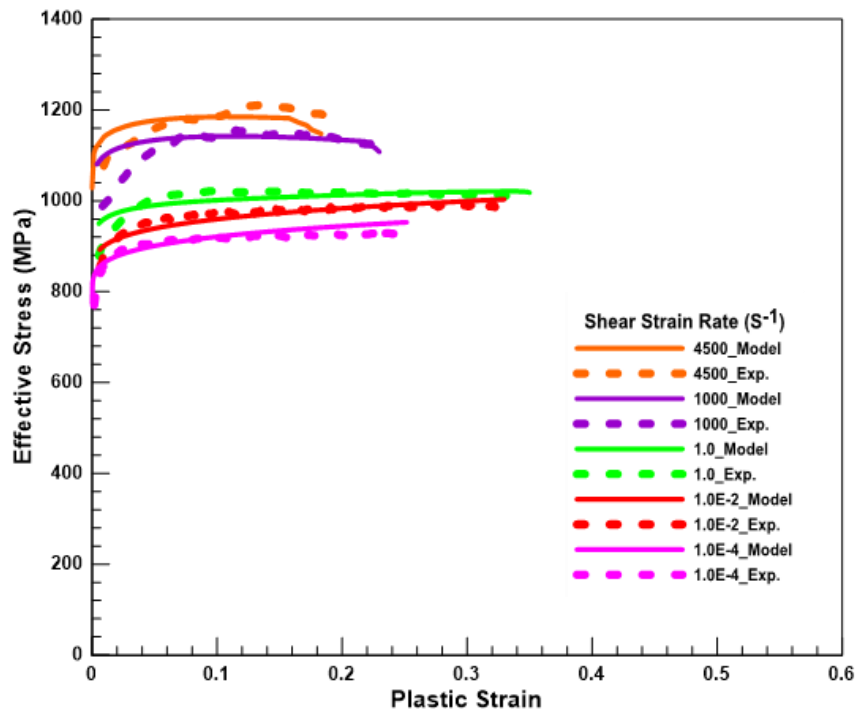


Figure 3-60 Comparison of stress-strain curves between experiment and modified JC model from torsion test at various strain rate

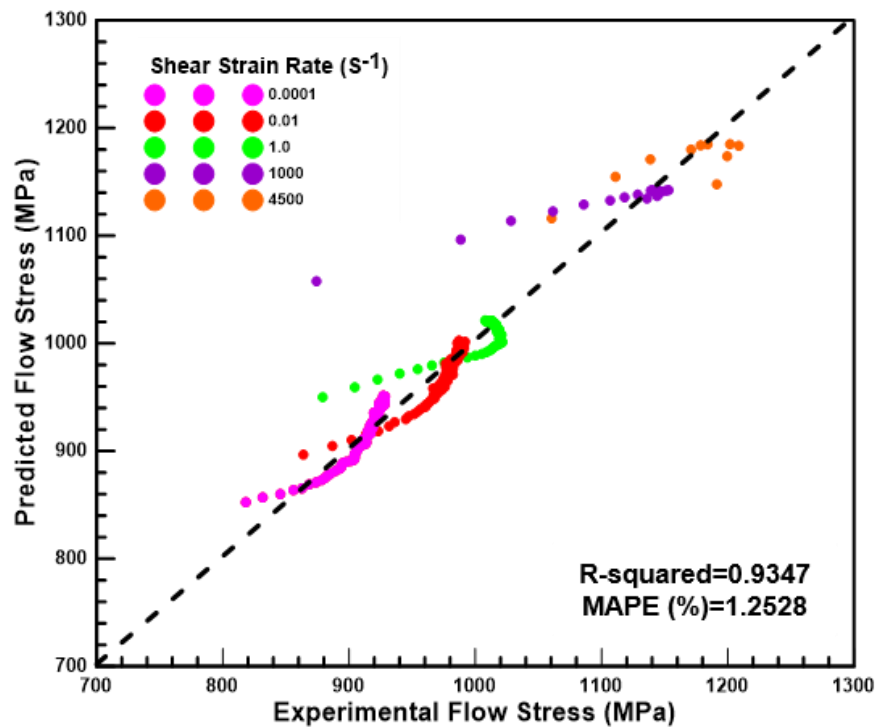


Figure 3-61 Comparison between predicted data and experimental data from torsion test at various strain rate

3.5.4 Discussion

The calibrated parameters for the modified Johnson–Cook model, summarized in Table 3-3, provide a quantitative basis for comparing the thermomechanical behavior of 410 stainless steel across different loading modes. The distinct hardening and rate-sensitivity characteristics of the material under tension, compression, and torsion can be identified by examining the constants values at each loading.

The yield stress parameter (A) indicates that 410 Stainless Steel exhibits nearly identical 0yielding behavior in tension and torsion (720.15MPa), while the yield stress in compression is higher (740.52MPa). Regarding work hardening, the B coefficient reveals that tension undergoes the most significant hardening (395.23MPa), followed by compression (340.23MPa) and torsion (220.35MPa). The hardening exponent (n) follows a similar trend, displaying that the tensile loading mode results in the most pronounced strength increase with strain.

The strain rate sensitivity is captured by the C_1 and C_2 coefficients. Torsion exhibits the lowest C_1 value (0.07811), indicating that the material is relatively rate-stable in the low to intermediate shear strain rate. In contrast, tension and compression show higher logarithmic sensitivity. However, at high dynamic rates, compression displays the largest C_2 coefficient (0.1445). This corresponds to the sharp non-linear upturn in flow stress observed in high-rate compressive tests, where dynamic effects become dominant.

The polynomial temperature terms (m_1 , m_2 , m_3) represent the complex thermal softening during adiabatic heating. While tension and compression share the same m_1 and m_2 values, the higher-order term (m_3) varies significantly. Compression reveals the largest

m_3 value (2.52), followed by torsion (1.41) and tension (0.41). The large m_3 in compression signifies an accelerated rate of thermal softening at large plastic strains. This is likely influenced by the unique thermomechanical complexity of the compressive loading state. Unlike the relatively pure loading observed in tension and torsion, compression involves the non-linear conversion of plastic work into heat across the entire specimen volume. This bulk heating, combined with changes in the contact surface between the sample edges and the platens, creates a complex reduction in flow stress. Such multifaceted factors cannot be captured by a simple two-term temperature formulation.

Table 3-3 The coefficient of Modified JC model

Test mode	A	B	n	C_1	C_2	m_1	m_2	m_3
Tension	720.15	395.23	0.3716	0.08461	0.0831	0.5213	0.2565	0.41
Compression	740.52	340.23	0.3664	0.08486	0.1445	0.5213	0.2565	2.52
Torsion	720.15	220.35	0.2154	0.07811	0.1015	0.6135	0.3249	1.41

3.6 Microstructure Observation

To determine the fundamental deformation mechanics of 410 stainless steel, systematic microstructural analysis is conducted on tested specimens from the 1.0 s^{-1} strain rate experiments. This investigation aims to explain the lower Taylor-Quinney coefficient (TQC) value observed at strain rate of 1.0 s^{-1} compared to the higher TQC value generated under dynamic tensile loading (Split-Hopkinson Bar testing). Electron Backscatter Diffraction (EBSD) phase mapping is first utilized to check the baseline status and confirm phase. Kernel Average Misorientation (KAM) metrics are evaluated to determine the dominant mode of deformation. Crystallographic pole figures are analyzed to characterize texture evolution and directional lattice rotation path. Finally, grain size characterization and boundary analysis are implemented to double-check structural evolution under this

intermediate strain rate. This microstructural assessment provides a physical and thermodynamic explanation for the lower Taylor-Quinney coefficient obtained at intermediate strain rates from tensile testing.

The electron backscatter diffraction (EBSD) characterization in this work adhered to the standardized experimental setup and sample preparation developed by [36]. The target evaluation areas, the undeformed grip baseline and the post-test deformed gage section, are polished to a highly reflective surface using a vibratory polisher with 50 nm colloidal silica, following the multi-step grinding and polishing routines. EBSD data acquisition is executed on the identical hardware platform comprising a Thermo Scientific Quattro ESEM and an EDAX Velocity camera. The scanned area spans a comprehensive field of 1373 μm by 1120 μm captured with a continuous step size of 1 μm , approximately 10^6 total pixels to guarantee statistical reliability across the boundaries. Further crystal orientation analysis is processed via the MTEX toolbox in MATLAB to constrain the low-carbon martensite within its representative body-centered cubic (BCC) group.

Quantitative phase analysis from the EBSD phase maps (Figure 3-62) indicates that the volume fractions of retained austenite before and after deformation are 6.84% and 6.50%, respectively. This variation falls within the statistical margin of error, confirming that phase transformation does not occur. The starting material used in this study is a commercial-grade 410 martensitic stainless steel, supplied in the form of a hot-rolled, annealed, and pickled plate. This initial thermomechanical processing routine is consistent with the observed undeformed microstructure characterized via EBSD. The annealing treatment led to a microstructure dominated by a dislocated lath martensitic matrix while

preserving a minor volume fraction (6.84%) of retained austenite films along the lath boundaries [44]. The small amount of remaining austenite undergoes mechanical stabilization during the tensile test. As the material deforms, the plastic strain creates a high density of dislocations inside the austenite. These dislocations act as strong physical barriers. They block the transformation interfaces from moving, which stops the phase change happening [45]. Because the phase cannot change, the applied strain is handled entirely by the existing martensite.

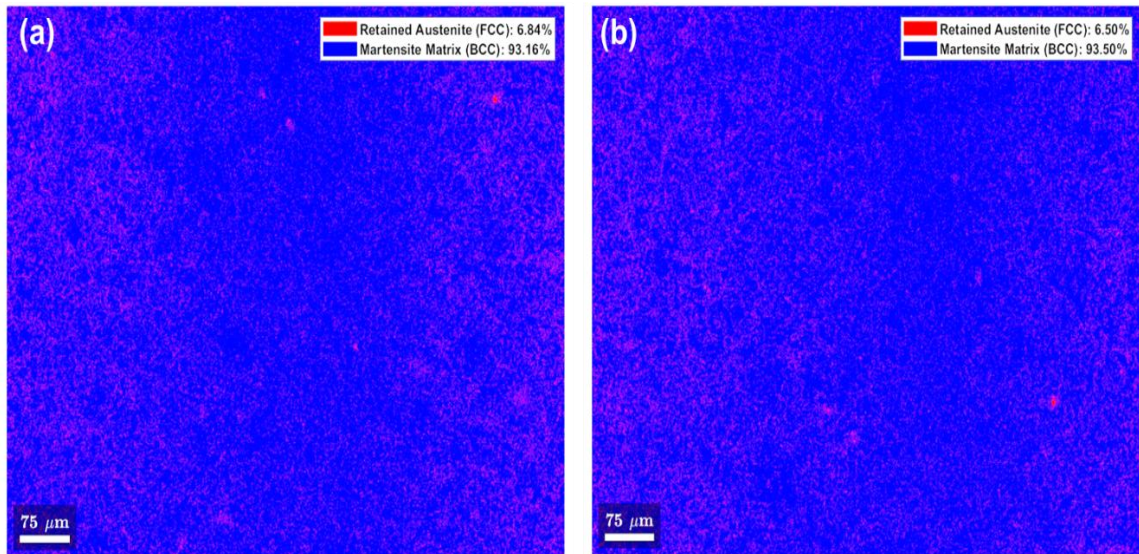


Figure 3-62 Phase map of 410 stainless steel tension sample (a) before and (b) after deformation at a strain rate of 1.0 s⁻¹

The evolution of the Kernel Average Misorientation (KAM) profile, which quantifies local lattice distortions by measuring the average misorientation between a specific point and its nearest neighbors within a grain, is evaluated. Higher KAM values correlate with an increased density of geometrically necessary dislocations (GNDs), while a decrease in KAM is associated with microstructural resetting through dynamic recrystallization (DRX) or twinning. Figure 3-63 exhibits a slight shift toward higher misorientation angles. While

the initial martensite already contains a very high dislocation density, the deformed state exhibits an increase and an extended tail toward higher misorientation angles. The absorbed mechanical energy is primarily stored as severe lattice distortion and the accumulation of geometrically necessary dislocations (GNDs).

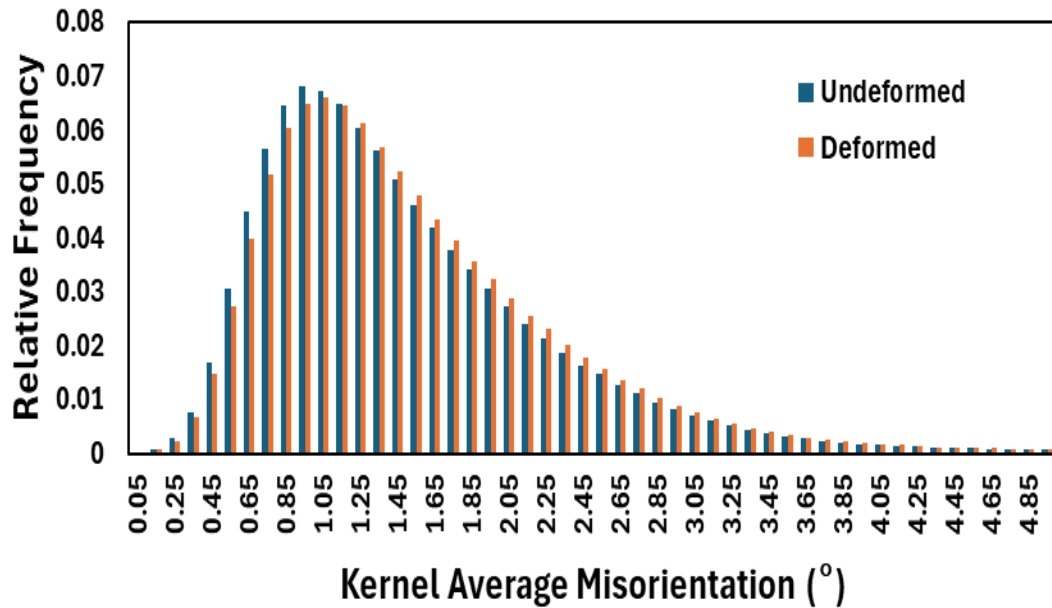


Figure 3-63 Distribution of KAM of 410 stainless steel tension sample before and after deformation at a strain rate of 1.0 s^{-1}

This deformation characteristic is further supported by microstructural metrics. As illustrated in the inverse pole figure (IPF) maps (Figure 3-64) and the statistical grain size parameters (Figure 3-65 and Table 3-4), the average grain size remains stable, shifting from $5.079 \text{ }\mu\text{m}$ to $5.033 \text{ }\mu\text{m}$. The high degree of overlap between the two grain size distributions confirms that the insufficient thermal activation energy suppresses long-range atomic diffusion and dynamic recrystallization (DRX), which would cause grain refinement.

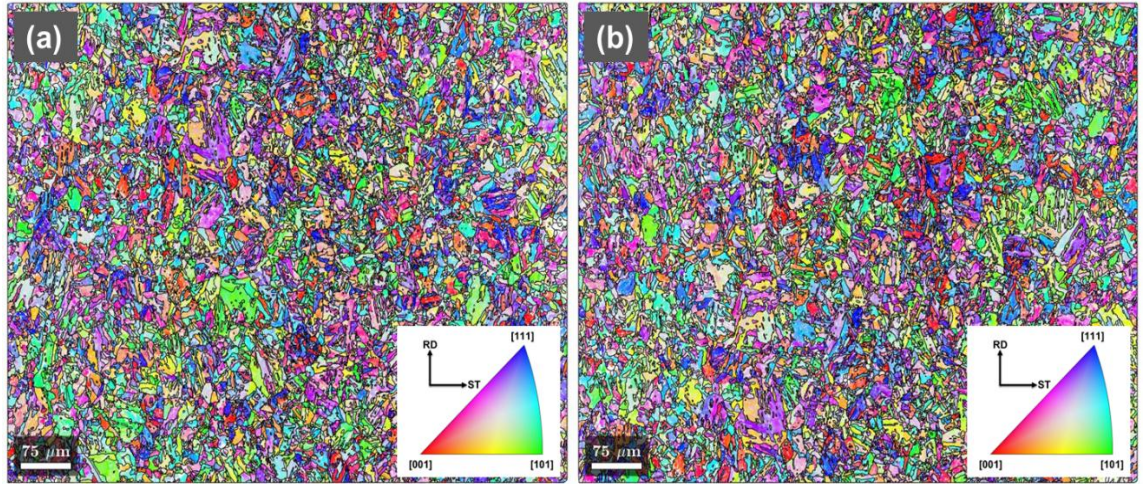


Figure 3-64 Inverse Pole Figure of 410 stainless steel tension sample (a) before and (b) after deformation at a strain rate of 1.0 s^{-1}

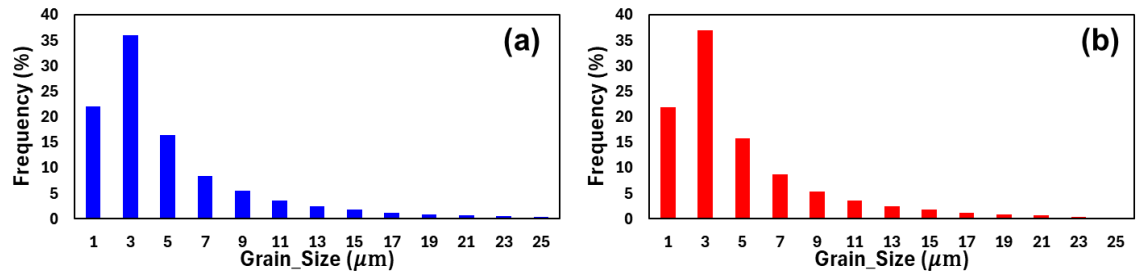


Figure 3-65 Grain Size of 410 stainless steel tension sample (a) before and (b) after deformation at a strain rate of 1.0 s^{-1}

Table 3-4 Grain Size Characterization of 410 Stainless Steel under a 1.0 s^{-1}

	Average Grain Size (μm)	Maximum Grain Size (μm)	Minimum Grain Size (μm)
Undeformed	5.079	62.572	0.988
Deformed	5.033	61.114	0.871

The pole figures (Figure 3-66) provide evidence for the dislocation-driven deformation mechanisms. Because the retained austenite undergoes mechanical stabilization and resists transformation, the martensitic matrix must accommodate the whole of the applied plastic strain. This is achieved through dislocation slip and subsequent

localized lattice rotation. This behavior is illustrated in the $\{110\}$ pole figure, which exhibits increased polar intensity parallel to the tensile axis (RD). The maximum Multiple of Random Distribution (MRD) of the $\{110\}$ pole figure exhibits an increase from 1.9144 to 1.9824 (Table 3-5). It represents the classical orientation of $\{110\}$ plane for BCC metals under uniaxial tension. On the other hand, the maximum MRD of the $\{100\}$ pole figure exhibits decreasing from 1.9682 to 1.6811. This drop reflects the migration of these unstable grain orientations toward the stable $\{110\}$ texture via continuous lattice rotation. While a strong baseline texture is established during the initial martensitic quenching process, this texture development demonstrates that the intermediate strain rate deformation of 410 stainless steel is via lattice rotation and dislocation accumulation, rather than through a phase transformation.

Table 3-5 Measured 2D Pole Figure Peak MRD of 410 Stainless Steel under a 1.0 s^{-1}

	$\{100\}$	$\{110\}$	$\{111\}$
Undeformed	1.9682	1.9144	1.7973
Deformed	1.6811	1.9824	1.8291

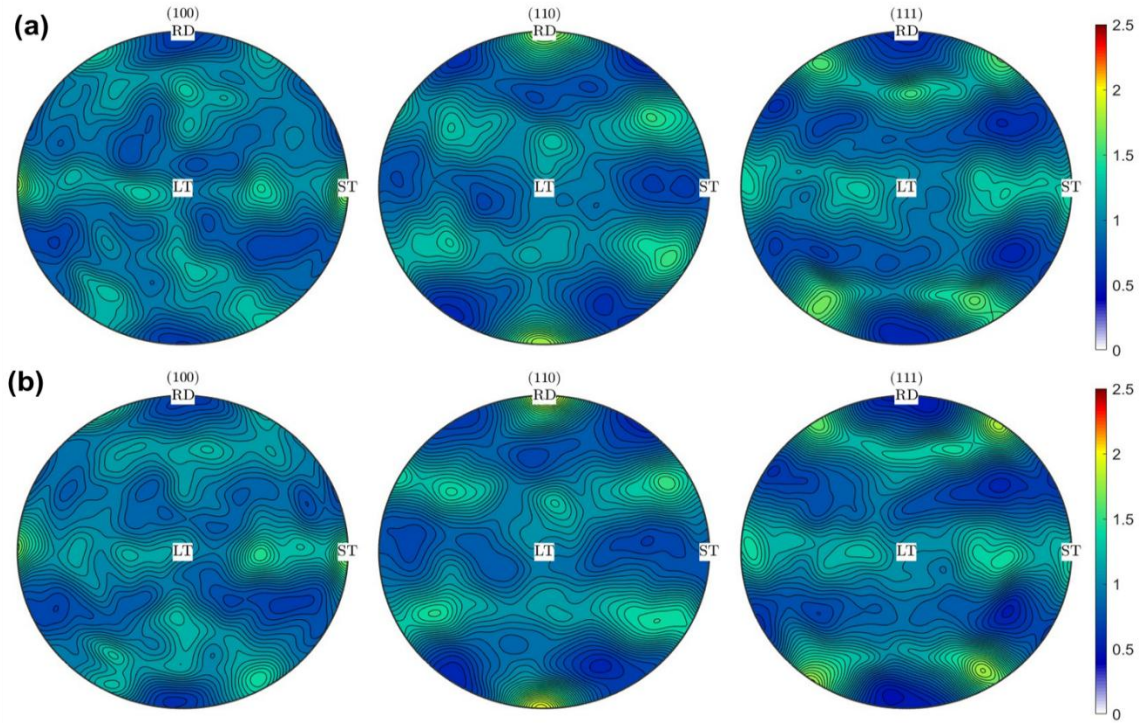


Figure 3-66 Pole Figure of 410 stainless steel tension sample (a) before and (b) after deformation at a strain rate of 1.0 s^{-1}

The observed discrepancy in the Taylor-Quinney coefficient (TQC) between the intermediate (strain rate of 1.0 s^{-1}) and SHB tension can be explained by the competition between dislocation storage efficiency (stored defect energy) and thermal dissipation. At an intermediate strain rate of 1.0 s^{-1} , the deformation timescale allows interaction and organization into stable dislocation. This stable microstructural defect storage locks a higher fraction of the plastic work as stored crystalline energy, reflecting the TQC at a lower plateau of approximately 0.72. On the other hand, under dynamic tension, two possible mechanisms may suppress this crystalline energy storage and elevate TQC. First, the severe time limitation is expected to suppress high velocity dislocations into stable configurations, forcing the plastic work into lattice vibrations (heat) under adiabatic conditions [46]. Second, considering the deformation of body-centered cubic (BCC) metals,

dislocation slips along the close-packed planes is suppressed at high strain rates due to the high Peierls-Nabarro stress barrier. Under these extreme conditions, the dominant plastic deformation mechanism is reported to transition toward mechanical twinning [47][48]. Because coherent twin boundaries have low energy interfaces, the accumulation defect energy of material would diminish. Most of the subsequent plastic work cannot be efficiently converted into stored defect energy and is instead converted into thermal dissipation, driving the TQC toward the value around 0.75 to 0.94.

Chapter 4 Conclusions

This chapter evaluates the factors influencing the material's behavior under various loadings, from geometric considerations to the fundamental energy conversion captured by the Taylor-Quinney coefficient. By integrating these observations, a constitutive framework is established to describe the coupling between plastic work, heat generation, and flow stress across various strain rates and loading modes.

4.1 Thermo-mechanical Behavior

The 410 stainless steel exhibits significant ductility across all loading conditions. In tension, the material reaches true strains of 0.6 at a strain rate of 1.0 s^{-1} , and in torsion, shear strains up to 0.7 are achieved at the same rate. Compressive loading does not result in any visible fracture, even when subjected to large plastic deformations. While the initial effective yield stresses are consistent across all three loading modes, their subsequent strain hardening responses diverge significantly. Tensile loading shows the most pronounced hardening rate, followed by compression, while torsion reveals the least. In terms of rate sensitivity, all loading modes share a similar logarithmic sensitivity from quasi-static to intermediate rates. However, at high strain rates, the sensitivity becomes more pronounced under compressive loading.

Under quasi-static elevated temperature testing, all loading modes demonstrate identical temperature sensitivity. At a very low strain rate of 10^{-4} s^{-1} , the material remains isothermal with negligible heat generation. The intermediate strain rate of 1.0 s^{-1} acts as the transition between isothermal and adiabatic conditions. The heat generation is high enough and heat loss to the surroundings is observable compared to the quasi-static state.

During high strain rate testing, all loading modes converge to a near-linear adiabatic heating profile with no heat loss. Torsion and compression share similar heat evolution when heat loss occurs due to their specific boundary conditions, but their behavior remains consistent with tension once adiabatic conditions are reached.

An objective for future work is to perform a detailed microstructural investigation to identify the physical mechanisms behind these loading-mode-dependent hardening rates in tension, compression, and torsion. Future studies could also investigate physical origins behind the sharp increase in strain rate sensitivity observed under high-rate compressive loading, with a specific focus on determining whether this trend is from a transition in the dominant deformation mechanism, such as the onset of mechanical twinning.

4.2 Taylor-Quinney Coefficient

The evolution of the Taylor-Quinney Coefficient (β) across all investigated loading modes follows a distinct three-stage progression relative to plastic strain. At the onset of plastic deformation, β typically reaches its minimum value, reflecting a phase where the material exhibits its maximal rate of stored energy [49]. As deformation progresses, a competitive mechanism emerges. The creation of new defects and dislocations (which tends to decrease β) acts against deformation-induced heating (which increases β) [50]. This results in a gradual rise in β , though the varying slopes observed across different test data at this stage lead to a higher standard error. As the microstructure evolves toward a saturation of stored energy, the dislocation density reaches a limit where the lattice can no longer store additional cold work. The energy of plastic work is dissipated as heat, causing β to rise and stabilize in the later stages of deformation.

Under high strain rate conditions, the material exhibits adiabatic heating, a behavior supported by the calculation of the Fourier number (Fo) and the observed strain-temperature distributions. While the data does not reveal a clear dependency on the strain rate, a significant loading mode dependency is evident. The TQC for tension ranges from 0.75 to 0.94, while compressive loading has lower values between 0.68 and 0.70. Torsion exhibits the highest average TQC, ranging from 0.86 to 0.94.

At quasi-static strain rates, the TQC progression reflects significant heat loss. In tension, a clear decay in β is observed when the test duration is long enough and the strain rate slow enough to allow for thermal dissipation. In compression, the temperature rise at the onset of plastic deformation is too small to distinguish from background thermal noise. As deformation continues and heat generation increases, the temperature becomes detectable, displaying a rapid decay of TQC due to heat loss. Torsion exhibits a similar phenomenon to compression, which is attributed to similarities in their conductive thermal paths governed by specimen geometries and gripping boundary conditions.

The intermediate strain rate at 1.0 s^{-1} acts as a transition between adiabatic conditions and significant heat loss. In tension, the material sustains near-adiabatic behavior up to a strain of approximately 0.3, beyond which the limited thermal path and increasing timescale leads to a slight decay in β . Conversely, compression and torsion fail to sustain adiabatic conditions for extended duration due to shorter thermal paths and wider conductive contact surface areas. The TQC in torsion displays a steeper decay slope than in compression. Despite the torsion test duration being nearly half that of compression, its shorter characteristic thermal path lends faster cooling. By using the Fourier number and

strain-temperature distributions to identify the threshold of heat loss, the data at 1.0 s^{-1} remains a reliable indicator of the transitional thermomechanical response.

Microstructural analysis may also play a critical role in explaining these observed variations in TQC across different loading modes. For materials that exhibit asymmetric responses, such as commercially pure Titanium, prior literature has linked higher TQC values under shear and compression to profuse twinning rather than tension [8]. For the 410 martensitic stainless steel investigated here, the higher thermal conversion efficiencies observed under torsion and tension relative to compression suggest that a different partition of slip dislocation, or potential high-rate twinning mechanisms may be active. Future microstructural characterization is necessary to determine if these mechanisms govern the loading-mode-dependent thermal conversion seen in this study.

4.3 Constitutive Material Model

The modified Johnson–Cook (J–C) model demonstrates flexibility and predictive accuracy in replicating the thermomechanical response of 410 stainless steel across tensile, compressive, and torsional loading modes. This improved formulation addresses several limitations in the traditional J–C model, which fails to capture the complex coupled behaviors observed in experimental data. Specifically, the modified model replaces the standard linear logarithmic strain rate term with a non-linear exponential formulation to better replicate the sharp increase in flow stress as deformation transitions from quasi-static to high dynamic rates..

The calibrated parameters provide a quantitative basis for correlating the material's behavior under different stress states. The yield stress parameter (A) indicates that 410

stainless steel exhibits nearly identical yielding behavior in different loadings. Work hardening characteristics vary significantly across modes. The B coefficient and hardening exponent (n) shows that tension undergoes the most pronounced hardening, followed by compression and torsion. Regarding dynamic effects, compression displays the largest value of the C_2 coefficient, which corresponds to the sharp non-linear upturn in flow stress observed at high rates, compared to tension and torsion. At the temperature terms, higher-order thermal softening term (m_3) reveals some differences from compressive loading, which features the largest value at 2.52, followed by torsion at 1.41 and tension at 0.41. The high m_3 value in compression signifies an accelerated rate of thermal softening at large plastic strains.

Future work should focus on the characterization of ductile fracture to complete the modified Johnson–Cook model by determining the damage parameters ($D_1 \sim D_5$ constants). This requires conducting tensile tests on different specimen geometries, specifically using diverse notch sizes to change the local stress state. These experiments are necessary to calculate the stress triaxiality and the Lode parameter, which define the relationship between the loading state and the onset of failure. Integrating these fracture criteria will extend the current constitutive model's capability beyond flow stress prediction to dynamic failure modeling.

4.4 Evaluation of Barreling

The study quantified the extra stress induced by friction as approximately 8% under quasi-static conditions and 6% at intermediate strain rates with lubricant. At high strain rates, the extra stress is negligible and stays within 1% until a plastic strain of 0.4. This

phenomenon occurs because the high slip velocity, which exceeds 1000 mm/s, results in average barreling factor (b_{avg}) lower than 0.1. As deformation progresses beyond this strain level (strain $\approx$ 0.4), the lubricant likely breaks down, leading to the gradual reappearance of barreling effects.

The evaluation of specimen barreling further confirmed that 410 stainless steel exhibits a consistent yielding stresses across all investigated loading modes. After adjusting the compressive flow stress using the barreling factor, the equivalent yield stress for tension, compression, and torsion are found to be consistent.

A consideration for the validity of these findings is the spatial and displacement accuracy of the 3D DIC system, since the stress correction relies on the recorded transient geometric profiles. Specifically, the calculation of the maximum radius and the resulting barreling factor are highly sensitive to the measured out-of-plane displacement or the transverse true strain. Technical challenges, such as speckle pattern breakdown at large strains, calibration errors, and resolution limits can cause errors in these measurements. High quality imaging is necessary to ensure these geometric inputs are precise so that the corrected flow curves represent the material response.

A primary area for future work is the comparison of different lubricants using this analytical method. Since friction is much more significant in low-rate tests, this framework provides a reliable way to evaluate lubricant effectiveness. Future research could systematically test various interfaces to identify optimal interfacial conditions required to maintain a uniaxial stress state during quasi-static compression test.

Appendix A. Stress Analysis of a Compressed Cylinder

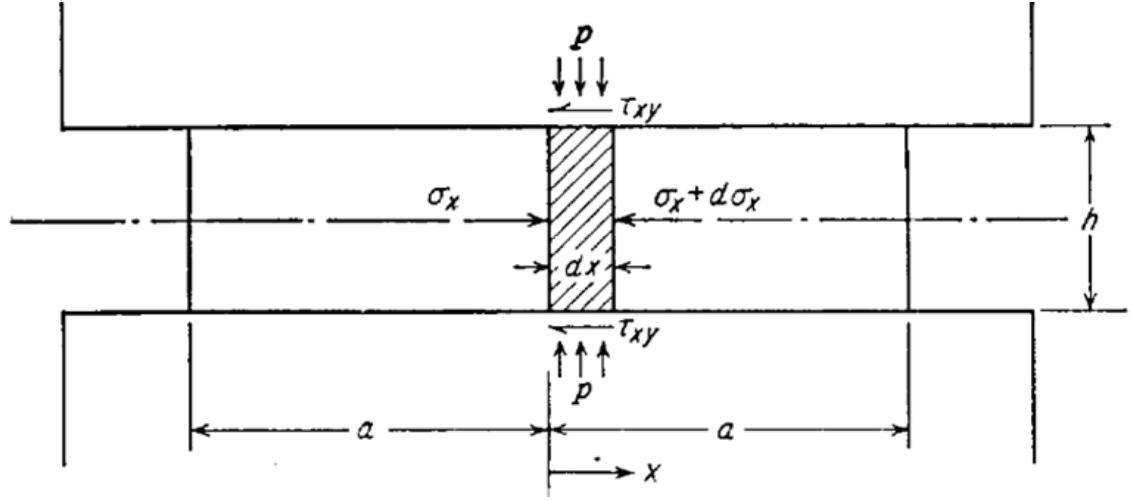


Figure A-1 Stresses acting on a cylinder

Recalling the eq.2-11, the analytical solution for the interface pressure solution for a plate forged in plane strain is given by [42]:

$$p = \sigma_0 \exp \left[\frac{2\mu}{h} (a - x) \right]$$

To adapt this for a cylinder, Replacing the half-width a with the radius R and the coordinate x with the radial distance r . The average pressure, P_{avg} , is then determined by integrating over the contact surface:

$$p_{avg} = \int \frac{dF}{A} = \frac{1}{\pi R^2} \int_0^R 2\pi r \cdot \sigma'_0 \exp \left[\frac{2\mu}{h} (R - r) \right] dr \quad \text{A-1}$$

Evaluating the integral:

$$p_{avg} = \frac{\sigma'_0}{2} \left(\frac{h}{\mu R} \right)^2 \left[\exp \left(\frac{2\mu R}{h} \right) - \frac{2\mu R}{h} - 1 \right] \quad \text{A-2}$$

To simplify the expression, let $x = 2\mu R/h$, Substituting this into eq.A-2 gives:

$$p_{avg} = \sigma'_0 \cdot \frac{2}{x^2} [e^x - x - 1] \quad \text{A-3}$$

Using the Taylor series expansion for e^x :

$$e^x = 1 + x + \frac{x^2}{2!} + \frac{x^3}{3!} + \dots \quad \text{A-4}$$

Substituting the first four terms of the expansion into eq.A-4:

$$p_{avg} \approx \sigma'_0 \cdot \frac{2}{x^2} \left(\frac{x^2}{2} + \frac{x^3}{6} \right) \quad \text{A-5}$$

Rearranging eq.A-5 in term of μ , R , and h

$$p_{avg} = \sigma'_0 \left(1 + \frac{2\mu R}{3h} \right) \quad \text{A-6}$$

Finally, solving for the friction coefficient μ during cylinder compression results in:

$$\mu = \left(\frac{p_{avg}}{\sigma'_0} - 1 \right) \cdot \frac{3h}{2R} \quad \text{A-7}$$

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