

7.0 INDUCTION TEMPERING OF INDUCTION HARDENED SHAFTS

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7.1 Introduction

This project was initiated by the ASPPRC in August 2020 as the basis for a Doctor of Philosophy thesis for Michael Jacobs in the Metallurgical and Materials Engineering department. The expected completion date is May 2024. The purpose of this project is to evaluate the effects of low temperature induction tempering on the performance of induction hardened AISI 1045 shafts as compared to conventional tempering, with an initial focus on torsional fatigue and bending fracture performance.

7.2 Industrial Relevance

Case hardening improves performance by taking advantage of increased strength at the surface while maintaining toughness in the core. Induction hardening is a fast and efficient heat treatment that produces a martensitic microstructure in the case without transforming the core. Martensite has high strength but often needs to be tempered to increase toughness. Conventional tempering for an hour or more is conducted in a furnace and takes up significant manufacturing space. The use of induction heating for rapid, short time tempering presents the possibility of decreasing cycle times and reducing space requirements. Rapid induction tempering for short times has been shown to produce hardness similar to traditional tempering [7.1] with *improved toughness* in some tempering regimes. The differences in microstructure development [7.2] due to the different thermal history could also have an impact on the development of the residual stress distribution [7.3] and influence both cyclic and static mechanical performance [7.4–7.6]. Understanding the effects of rapid induction tempering on mechanical properties and the sensitivity of those mechanical properties to changes in the induction tempering cycle is of great importance for the application of rapid induction tempering in industry.

7.3 Literature Review

This project will utilize commercially produced AISI 1045 bar stock with an as-rolled ferrite-pearlite initial microstructure that will be rapidly austenitized, quenched, and tempered at low temperature to form a martensitic microstructure. Martensite is obtained by suppressing diffusional transformations during quenching. Martensite forms by a diffusionless, athermal transformation that takes place by rapid and cooperative shear displacement [7.7]. During transformation from the face centered cubic (FCC) austenite phase to the body centered phase the equilibrium solubility of carbon in iron decreases significantly [7.8]; as insufficient time is available for diffusion to occur during transformation from austenite to martensite, an excess carbon atoms can become trapped non-randomly in the octahedral sites of the otherwise body-centered cubic (BCC) structure, distorting the lattice to from a body centered tetragonal (BCT) structure [7.7].

Several factors contribute to the high strength of martensite. The lattice distortion resulting from the trapping of carbon atoms at octahedral sites in BCC iron helps to inhibit dislocation movement. In lower carbon martensite, where the martensite start temperature is relatively high, some rearrangement of carbon atoms can take

place during cooling. Carbon atoms segregate to defect sites, pinning dislocations and increasing strength. Martensite grows on specific planes in the parent austenite, called habit planes, which remain undistorted and unrotated during transformation. To accommodate the lack of rotation of the habit planes a lattice invariant deformation is also required along with the lattice deformation, and is accomplished by slip, or by twinning in higher-carbon plate martensite. Martensite is associated with high dislocation densities which also contribute to the high strength of martensite. The shear displacement and shape change associated with the martensitic transformation requires the parent austenite phase to plastically accommodate the growing martensite. Once the parent phase can no longer accommodate further growth of a martensite crystal a new crystal must be nucleated, leading to an increasingly fine, strong microstructure as the ability of the increasingly martensitic microstructure to accommodate further transformation diminishes [7.7]. Some amount of retained austenite may be present, especially in higher carbon steels, even below the apparent martensite finish temperature. This is thought usually to be due to the suppression of further martensite crystal growth due to the elastic strains produced by earlier growth [7.9]. The presence of retained austenite is important because it can have an important influence on mechanical properties, both positively through transformation induced plasticity (the TRIP effect) or negatively, such as by reducing hardness, depending on its volume fraction and distribution [7.5, 7.10]. Changes in retained austenite may also influence component dimensions and distortion during heat treatment or in the end-use application.

7.3.1 Effects of Retained Austenite in Induction Hardened Shafts

It is generally suggested to avoid significant amounts of retained austenite in induction hardened components because the transformation of retained austenite into fresh, untempered martensite in service can reduce toughness [7.11]. However, experimental work has shown that the effects of retained austenite on mechanical properties is more complex and often beneficial. Previous work by Judge showed that rapid tempering produced microstructures with higher amounts of retained austenite having improved impact toughness compared to conventionally tempering to an equivalent hardness in quenched and tempered 4340 steel [7.12]. Earlier work by Boazhu and Krauss compared the impact toughness of tempered martensitic and bainitic microstructures in 4340 steel achieved by quenching and isothermal transformation respectively. These microstructures contained varying amounts of retained austenite, ranging from 0-6.9 vol pct, without exhibiting a difference in impact toughness [7.13].

Retained austenite is considered beneficial to low cycle fatigue properties due to the strain induced transformation to martensite that occurs, increasing the rate of strain hardening and producing compressive stresses that slow fatigue crack initiation and propagation [7.14, 7.15]. At high stress and strain amplitudes the increased ductility provided by retained austenite can improve fatigue life, however reduced strength and hardness may be detrimental to high cycle fatigue properties (reduced endurance limit) which are dominated primarily by the initiation portion of the fatigue crack lifecycle [7.15, 7.16]. While the effects of retained austenite on the monotonic strength appear detrimental to high cycle fatigue properties, the cyclic behavior is more complex. Richman and Landgraf investigated the effects of retained austenite on the fatigue performance of several quenched and tempered 4XXX series steels, including cryogenic treatments to vary the amounts of retained austenite in a given alloy, by comparing cyclic stress-strain responses in strain-controlled fatigue tests. It was found that steels containing retained

austenite cyclically strain hardened rather than strain softening, and ultimately had higher cyclic flow stresses once stabilization had occurred, suggesting the potential for greater high cycle fatigue lives with increased retained austenite [7.17].

7.3.2 Residual Stresses in Induction Hardened Shafts

The transformation of austenite to martensite during quenching occurs with an associated volume expansion. The expansion of the martensitic case is constrained by the non-transformed core in surface hardened components, causing elastic strain in both phases and a residual stress distribution to form. The magnitudes of the residual stresses in induction hardened shafts are strongly dependent on the process parameters (frequency, power density, scan speed, dwell time, etc.) used and the final case depth [7.18]. The compressive residual stresses in the martensitic case, where the in-service stresses are highest, increases the fatigue life of the part by decreasing the total applied stress intensity during crack nucleation and growth [7.19].

Although compressive residual stresses in the surface are beneficial to fatigue properties, they are balanced by tensile stresses in the core (going to zero at the neutral axis). Under the torsional and bending loading conditions induction hardened shafts are generally subjected to, these tensile residual stresses in the core are not of great concern, although the transition region between the case and core can experience the maximum “net” tensile stress and is thus an area of concern for subsurface crack initiation [7.20]. To balance the benefits and risks of the compressive surface tensile subsurface residual stress states, a low temperature temper is often employed to reduce the magnitude of the residual stresses without compromising strength [7.20–7.22].

7.4 Residual Stress Evolution During Tempering

Residual stresses and their associated strain fields create stored mechanical energy within a component. Residual stresses, which are elastic, can be relieved by plastic deformation. During tempering, several processes can take place that relax residual stresses below the room temperature yield stress. In the case of residual stresses approaching the material yield stress, a straightforward mechanism is the reduction of the material’s yield stress with increased temperature [7.8, 7.23]. For residual stresses of lower magnitude other mechanisms must be active.

A study by Brown *et al.* investigated the relief of applied stresses in hardened steel split rings during tempering. A wedge was used to apply a constant elastic strain, corresponding to an elastic stress, during tempering for a wide range of temperatures and times after which the amount of springback was measured. The amount of stress relief was quantified as the extent of incomplete springback. This study found that, for timescales on the order of hours traditionally associated with tempering, stress relief began to plateau at 200 °C [7.3]. Due to the plateau in stress relief occurring at the temperature traditionally associated with the end of transition carbide precipitation, Brown *et al.* suggested that carbide precipitation was associated with stress relief, employing a Greenwood-Johnson model for transformation plasticity [7.3]. Greenwood and Johnson describe a significant loss in strength and an associated increased plastic strain during phase transformations, attributed to volume changes imposing additional strains that must be accommodated by the weaker of the parent and product phases [7.24, 7.25]. This transformation plasticity mechanism has been incorporated into more recent simulation work done for induction hardening and conventional tempering that shows good agreement with analytical models as well as experiments [7.21, 7.22, 7.26].

Modeling done by Dong *et al.* and Eser *et al.* account for strains resulting from creep as well as transformation plasticity. Simulations of residual stresses in hardened steels were carried out, and suggested that creep had a significant effect on residual stresses during tempering above 400 °C.

7.5 Experimental Procedure

The overall objective of this project is to determine the effects of rapid induction tempering on the fatigue and fracture performance of induction hardened shafts. The experimental material for this project is in the form of 39.7 mm bars of AISI 1045 steel with a ferrite-pearlite starting microstructure. The chemical composition of this material is listed in Table 7.1. Rothleutner previously studied the torsional fatigue performance of this same material after induction hardening and *furnace* tempering [7.27].

Induction hardened shafts are tempered at low temperatures to relieve stress and increase fracture resistance without a significant loss in strength. It has been hypothesized that the possibility of tempering mechanisms operating in different orders during rapid tempering [7.28] could result in different microstructure development pathways as compared to conventional tempering. A possible consequence of this difference in microstructure development is that rapid tempering to the same tempering parameter (hardness) obtained by conventional tempering may improve fracture resistance. This hypothesis represents the essence of the project intent, following encouraging prior work of Euser [7.2].

Table 7.1 – Chemical Composition of Research Material (wt pct)

Steel	C	Mn	Si	Ni	Cr	Mo	V	Al	N	S	P	Cu
1045	0.45	0.74	0.23	0.10	0.12	0.03	0.002	0.016	0.0068	0.006	0.010	0.26

Before planning pilot studies using *actual* induction processing, a laboratory heat treatment study is being conducted using the Gleeble® 3500 to perform *simulated* induction hardening, followed either by simulated rapid induction tempering or conventional furnace tempering. The simulations are designed to match a thermal signature associated with the induction process, but using an alternative (resistive) heating method in the laboratory. The thermal signature is applied to the full cross-section of the specimen, so this method emphasizes the tempered microstructure and does not capture the microstructure gradients and residual stress evolution associated with induction hardening. The fracture resistance will be evaluated by a notched three-point bend test, and the hardness and fracture resistance of the different conditions will be used to inform the selection of processing parameters for subsequent induction hardening and tempering of torsional fatigue test specimens to examine the effects of induction tempering on torsional fatigue properties. The specimens will also be used to assess the microstructure changes occurring during induction tempering. Since Gleeble® processing does not provide a correct simulation of residual stress development during induction processing, actual induction hardening of fatigue test specimens will be employed later to characterize performance, including the residual stress evolution for each induction tempered condition. Microstructural characterization will compare the results of induction tempering with those of the laboratory tempered microstructures.

7.5.1 Simulated Induction Hardening

Notched three-point bend fracture blanks, shown in Figure 7.1, have been processed to simulate induction hardening using the Gleeble® 3500 according to the thermal signature computed by electro-thermal finite element modelling of the selected hardening process using parameters detailed in Table 7.2.

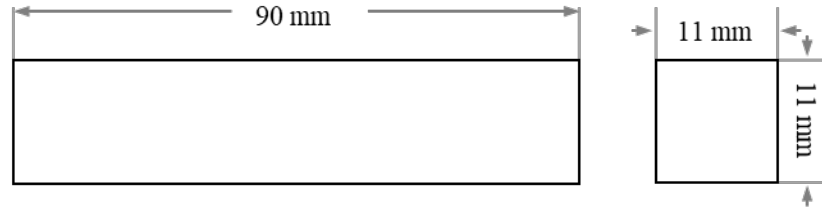


Figure 7.1 Notched three-point bending specimen blank.

This induction hardening process schedule will be used for future *actual* induction hardening of torsional fatigue samples, and was chosen for simulation to for three-point bending to produce a consistent, fine starting microstructure for all mechanical tests. This process schedule was used previously by Rothleutner [7.27] and was chosen because it produces a case depth ratio after induction processing that comparable to that which is used in many applications [7.27, 7.29]. Large case depths are often used in torque transmission applications because increased case depth leads to increased torsional yield strength. Comparable case depths to the one simulated in this study are often used due to reports that increases in torsional yield strength and ultimate torsional strength begin to plateau at higher case depths [7.30]. A single shot hardening process was chosen to maximize repeatability. The computed thermal profile, provided by Fluxtrol Inc, can be seen in Figure 7.2. This computed thermal profile was simulated using a programmed thermal profile with a linear heating rate of $600\text{ }^{\circ}\text{C}\cdot\text{s}^{-1}$ until a peak temperature of $1100\text{ }^{\circ}\text{C}$, followed by an immediate quench. Quenching in the Gleeble® was done using a 15 vol pct solution of Aqua Quench® 3699, a polymer quenchant, in water at a pressure of 60 psi. These quenching conditions were determined by iterative testing to achieve a cooling curve matching the modelled time-temperature profile at the sample surface shown in Figure 7.2. A K-type thermocouple was spot welded at the midspan of the sample to serve as the process variable for the PID (Proportional-Integral-Derivative controller) control loop used by the Gleeble® 3500. Proportional (P), integral (I), and derivative (D) values of 0.03, 0.05, and 0.005 were used. 30 mm of each end of the sample was gripped using water cooled copper grips during Gleeble processing, leaving 30 mm of free span.

Table 7.2 – Induction Hardening Processing Parameters [7.27]

Induction Hardening Method	Frequency (kHz)	Power (kW)	Scan Rate	Polymer Concentration in Quenchant (pct)	Quenchant Flow Rate (L/min)	Nominal Normalized Effective Case Depth, t/r
Single Shot	31	128	N/A	2	173	0.43

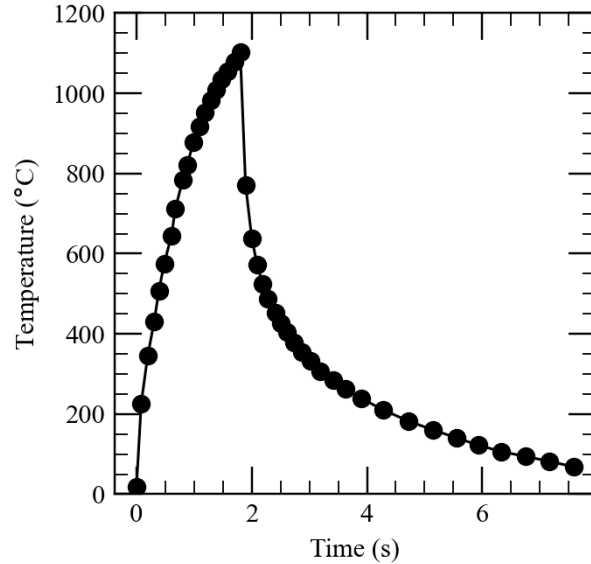


Figure 7.2 Computed thermal profile at the specimen surface using modeled induction hardening process parameters [7.27].

7.5.2 Tempering Matrix

A tempering matrix of conventional one hour and rapid, one second tempering heat treatments was selected as shown below in Table 7.3, with the approximate target Hollomon-Jaffe TP, given by

$$TP = T(c + \log t) \quad (7.1)$$

where T is absolute temperature, c is a material constant, and t is the time in seconds. A c -value of 15 was used [7.31]. For rapid tempering cycles where heating and cooling are larger contributors to the heat treatment, a non-isothermal tempering parameter can be calculated. The non-isothermal tempering parameter equates the average temperature, T_i , over a short time increment, Δt_i , to the tempering parameter of an isothermal heat treatment of the peak temperature, T^* . This allows for the calculation of equivalent time contributions, Δt^* , for each time increment. The tempering parameter calculated using the peak temperature and the summation of equivalent time contributions, Δt^* , has been shown to correlate well with tempered hardness [7.1, 7.2]. A schematic of this calculation is shown in Figure 7.3. Some additional tempering times may be considered depending on results. The proposed range of temperatures covers Stage I (transition carbide precipitation), Stage II (decomposition of retained austenite), and Stage III (replacement of transition carbides by cementite) in conventional tempering [7.7, 7.32]. This temperature range is larger than that commonly used for induction-hardened shafts to achieve the desired high hardness [7.33] and stress relief [7.3] during low temperature tempering.

Table 7.3 – Tempering Temperatures and Tempering Parameters (TP) for Conventional and Rapid Tempering Simulations

Time (s)		Temperature (°C)				
3600	100	150	200	250	300	350
1	188	250	312	373	435	497
TP (approximate)	6,920	7,850	8,780	9,700	10,630	11,560

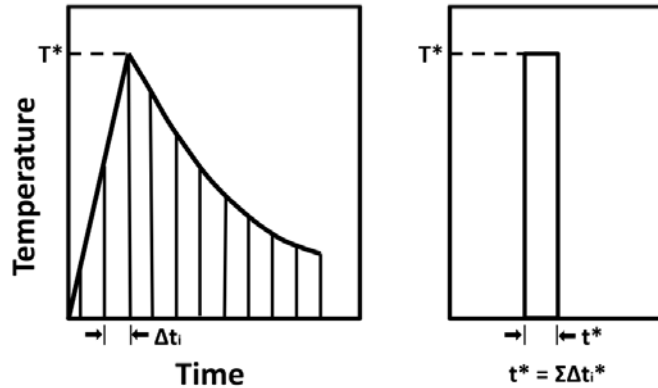


Figure 7.3 Schematic representation of non-isothermal cycle converted to an “equivalent” isothermal cycle, adapted from [7.1].

7.5.3 Development of Rapid Tempering Procedures

After simulated induction hardening of fracture resistance samples, the blanks will undergo either simulated rapid induction tempering using the Gleeble® or conventional furnace tempering. A heating rate of $600\text{ }^\circ\text{C}\cdot\text{s}^{-1}$ was chosen for rapid tempering as the maximum reliably attainable heating rate for the fracture blank geometry, previously shown in Figure 7.1. After a 1 s hold time the sample will be quenched to room temperature. The same quenchant used for simulated induction hardening will be used for simulated induction tempering. Proposed simulated thermal signatures for rapid tempering heat treatments were discussed with InductoHeat® to ensure industrial viability of these simulations for future induction tempering of torsional fatigue samples.

7.6 Results and Discussion

The following sections present the results from testing using simulated induction hardened and conventionally tempered samples as well as the details of the development of rapid tempering procedures.

7.6.1 Confirmation of Relationship between Hardness and Tempering Parameter

Confirmation of the applicability of the non-isothermal tempering parameter is ongoing. Simulated induction hardening was carried out on the notched three-point bend fracture specimen blanks intended for furnace tempering using the Gleeble® as described in Section 7.5.1. Samples were then tempered conventionally in box furnaces for 3600 s according to the tempering matrix given in Table 7.3. The remaining samples will undergo simulated induction hardening followed by simulated induction tempering using the procedures described in Section 7.6.2.

Bulk hardness measurements were taken from the conventionally tempered samples to establish a relationship between the tempering parameter and hardness, shown in Figure 7.4. The measured hardness for the as-quenched and the tempered samples are in good agreement with literature values [7.34].

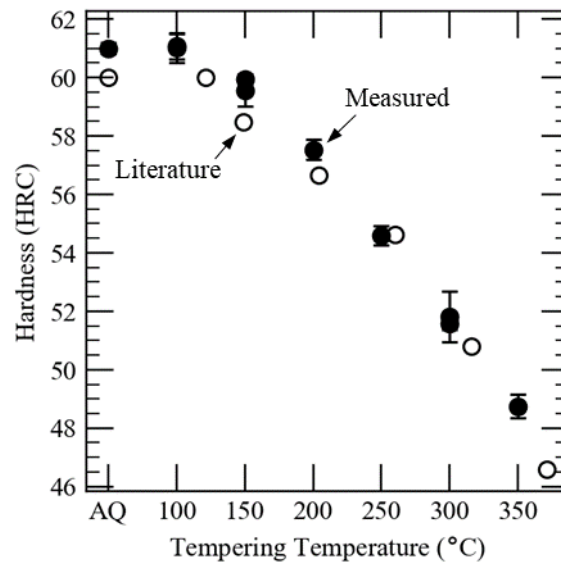


Figure 7.4 Hardness as a function of tempering temperature after furnace tempering for 3600 s compared to literature values [7.34]. Error bars correspond to standard deviation of five measurements for individual samples.

7.6.2 Simulated Induction Tempering

Gleeble® programs were developed to execute the 1 s rapid tempering heat treatments at the temperatures given in Table 7.3 with a heating rate of $600\text{ }^{\circ}\text{C}\cdot\text{s}^{-1}$. A K-type control thermocouple was spot welded at the sample midspan to serve as the process variable for the Gleeble® PID control loop. A P-value of 0.005 was used for tempering. Figure 7.5 shows an example target (programmed) time-temperature profile and the measured time-temperature profile, not including a programmed quench. Figure 7.5(b) shows the heating portion of the time-temperature profile in more detail.

A time delay, Δt , between the programmed start of heating at 0.5 s and heating beginning at 0.6 s can be seen, after which the heating rate is approximately linear until the programmed hold temperature. Once the hold temperature is reached the heating rate decreases in relation to the difference between the programmed and measured temperatures, ΔT . Integral control is not being used to eliminate the steady state error due to increased overshoot observed, however there is a steady state error that is dependent on the hold temperature that needs to be eliminated. This error is eliminated by adjusting the programmed hold temperature to be equal to the target temperature plus the steady state error.

To account for the time delay and the decreasing heating rate up to the hold temperature, the programmed time-temperature profile was modified to extend the heating time by Δt , approximately 0.1 s, to account for the temperature difference, ΔT , and reach the desired hold temperature at the desired rate. After heating concluded, the programmed temperature was set to the desired hold temperature for the 1 s hold time. When accounting for the steady state error in the hold temperature, this approach more accurately produced the desired time-temperature profile. The response time of the quench system will also be accounted for by adjusting the hold time to incorporate

tempering occurring during cooling. A comparison of the target, measured, and programmed time-temperature profile after refining the procedure can be seen in Figure 7.6.

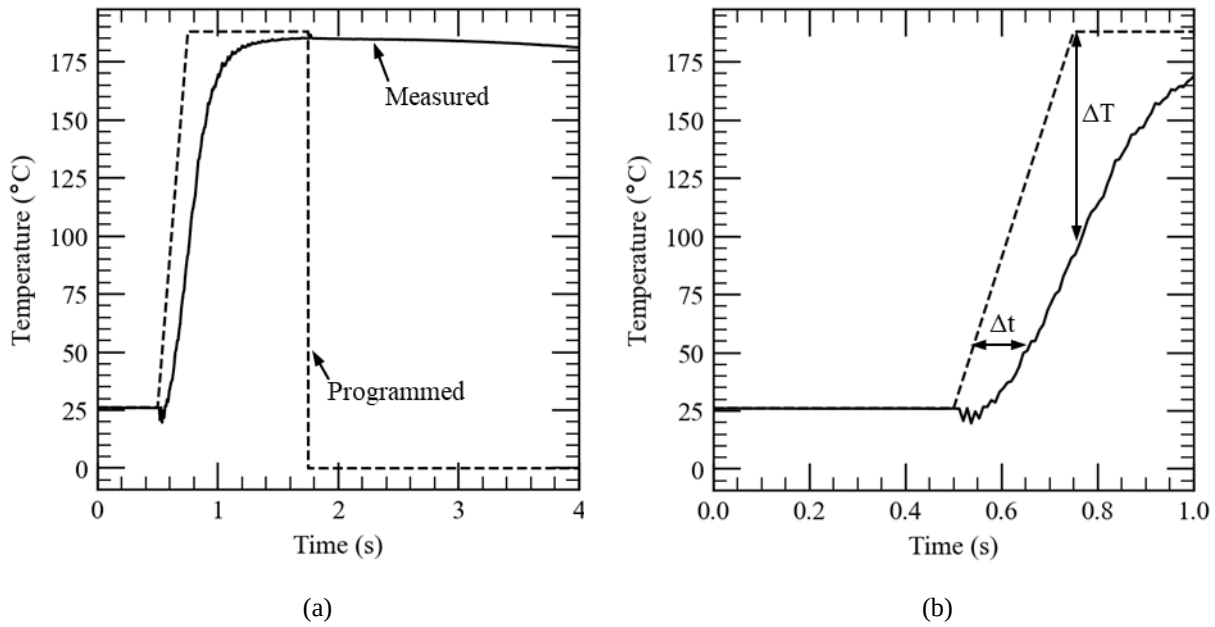


Figure 7.5 Programmed and experimental time-temperature profiles for a 1 s rapid tempering heat treatment at a hold temperature of 188 °C showing (a) the complete heating cycle and (b) magnified view during heating to the hold temperature.

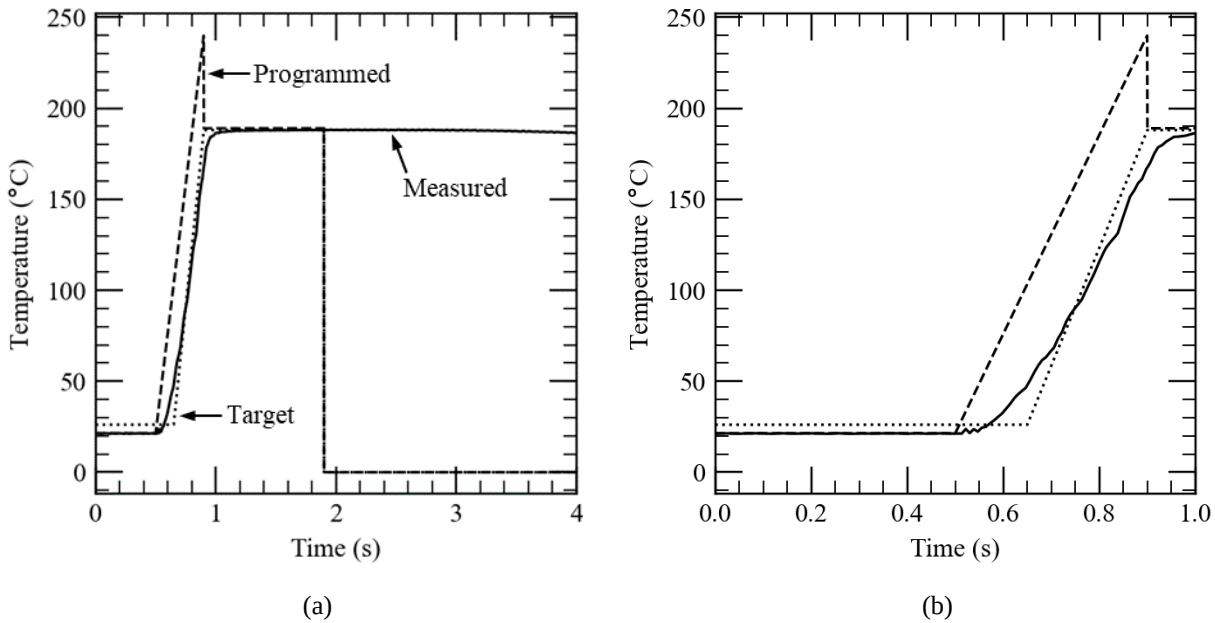


Figure 7.6 Altered programmed, target, and experimental time-temperature profiles for a 1 s rapid tempering heat treatment at a hold temperature of 188 °C showing (a) the complete program and (b) heating to the hold temperature.

7.6.3 Establishing Acceptable Error for Rapid Tempering Conditions

The methods described in Section 7.6.2 allow for more closely matching experimental results to the desired time-temperature profiles. However, there are still some small discrepancies between the experimental and target

profiles that cannot be eliminated. These deviations can be monitored/characterized using the non-isothermal tempering parameter. An acceptable deviation of the non-isothermal tempering parameter from the target equivalent isothermal tempering parameter was estimated, similarly to the work done previously by Euser [7.2], using the bulk hardness data collected from furnace tempered samples shown in Figure 7.4. The linear fit between the hardness and the isothermal tempering parameter was found to be

$$H = 79.36 - 0.0027 \times TP \quad (7.2)$$

where H is the hardness measured in HRC and TP is the isothermal tempering parameter. The average uncertainty of five hardness measurements collected for furnace tempered conditions was found to be 0.40 HRC, which corresponds to a measurable uncertainty of 147 tempering parameter units using Equation 7.2. The Holloman-Jaffe tempering parameter, given previously in Equation 7.1, is not equally sensitive to changes in time and temperature, so iso-tempering curves were calculated to define a region of acceptable equivalent isothermal time and temperature combinations. To maintain the essence of the target heat treatment cycle, the target times and temperatures were used as bounds on the iso-tempering curves. An example of an acceptable area of equivalent isothermal tempering temperatures and times is shown in Figure 7.7. Within this area, the hardness should be within the uncertainty of 0.4 HRC units. Nonetheless, it is emphasized that short times are most preferred to maintain focus on the project objectives.

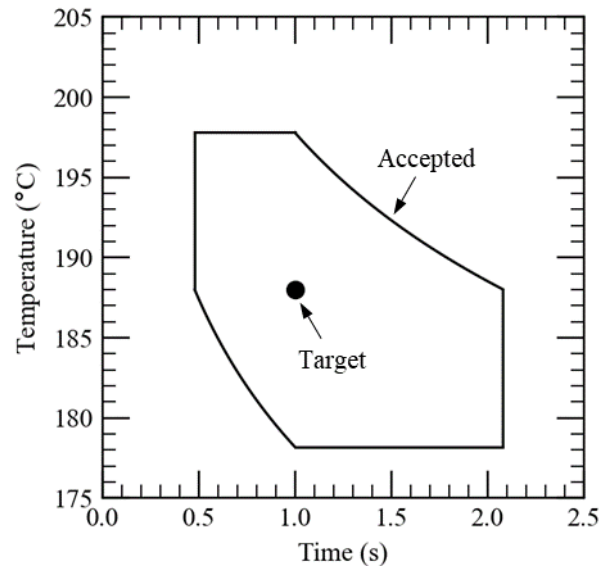


Figure 7.7 Example range of acceptable equivalent isothermal tempering times and temperatures for rapid tempering at 188 °C for 1 s.

7.6.4 Fracture Resistance Testing

After samples underwent simulated induction hardening and furnace tempering, the samples were machined to their final 10 mm square cross section to ensure any surface effects developed during heat treatment (such a decarburization) were eliminated prior to testing. Sample blanks were marked to indicate the original orientation of the samples within the bar stock. After grinding, 2 mm deep notches with a root radius of 0.25 mm were machined on the sample surface corresponding most closely to the original bar surface, at the position corresponding closely to the Gleeble® control thermocouple. The notched samples were cut down to 60 mm in

length to fit in the fracture resistance test setup. The experimental setup is shown in Figure 7.8. Hardened steel cylinders were used to apply the load at the contact points, while the load was applied at the notch position which was centered between the two cylinders. The span between the contact points was 40 mm. Knife edges placed 10 mm apart were superglued to the sample surface to allow for measurement of the crack opening using an extensometer.



Figure 7.8 Three point bending fracture test setup.

The peak bending load before fracture as a function of tempering temperature for samples that underwent simulated induction hardening followed by furnace tempering, following the processing matrix of Table 7.3, is shown in Figure 7.9. Peak load initially increases with tempering temperature up to a maximum at 200 °C, after which there is a downward trend. This temperature regime roughly corresponds with the tempered martensite embrittlement (TME) regime reported for Charpy impact tests [7.7]. Further tests will be carried out on samples that have undergone simulated induction tempering as well as duplicate tests for furnace tempered conditions currently with a single datapoint.

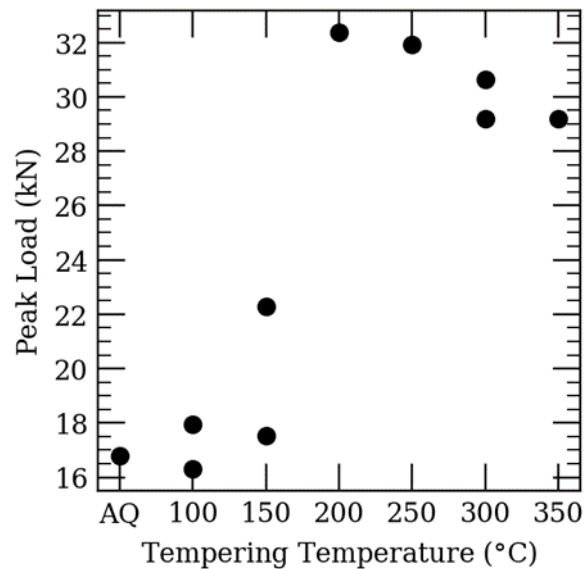


Figure 7.9 Peak load before fracture in three-point bending as a function of tempering temperature for samples that have undergone simulated induction hardening and furnace tempering.

7.6.5 Mössbauer Spectroscopy

After fracture resistance testing, the fractured samples will be sectioned for use in Mössbauer Effect Spectroscopy (MES). Mössbauer spectroscopy will allow for the differences in the phase fractions of retained austenite, cementite, and transition carbides to be observed between samples that undergo conventional furnace tempering and rapid, simulated induction tempering, to begin to understand the differences in details of microstructure evolution. 1 mm thick sections were taken directly adjacent to the fracture surface to minimize the effects of the longitudinal temperature gradient present during both the simulated induction hardening and tempering processes. These sections were then ground down to 80 μm thick and thinned in HF for MES analysis. Mössbauer spectrum collection and analysis were performed by Dr. Don Williamson.

Mössbauer results from a preliminary sample taken from a tested fracture sample are shown in Table 7.4 show good phase fraction resolution. Sample preparation for the remaining furnace tempered conditions is in progress and will continue for simulated induction tempered conditions once heat treatments and fracture testing have been completed.

Table 7.4 – Preliminary Mössbauer Spectroscopy Results

Tempering Temperature (°C)	Tempering Time (s)	Retained Austenite (at pct Fe)	Transition Carbides (at pct Fe)	Cementite (at pct Fe)
150	3600	3.5 ± 0.1	3.1 ± 0.6	0.2 ± 0.2

7.7 Summary

Simulated induction hardening processing using the Gleeble® followed by conventional furnace tempering was carried out, followed by hardness measurement and three-point bend fracture testing. Bulk hardness of furnace tempered samples showed good agreement with values in literature. The peak load in three-point bending increased up to 200 °C, after which tempered martensite embrittlement was observed. Plans to execute rapid, simulated induction tempering were developed as well as a measure of acceptable deviation from the ideal heat treatment using hardness data collected from furnace tempered samples and the non-isothermal tempering parameter.

7.8 Research Plans for Next Reporting Period

Plans for the next reporting period include simulation of induction tempering, marking completion of the currently planned Gleeble® processing, and completion of fracture testing. Torsional fatigue samples are currently being machined and plans for induction processing of these samples with InductoHeat Inc® before the next reporting period will be discussed. Metallography of the as-quenched and the furnace tempered microstructures is underway and will continue with simulated induction tempered microstructures once they have been processed. Similarly, Mössbauer spectroscopy of the as quenched and furnace tempered conditions is underway and will be followed by analysis of the simulated induction tempered conditions.

7.9 References

- 7.1 S. L. Semiatin, D. E. Stutz, and T. G. Byrer, "Induction Tempering of Steel: Part I. Development of an Effective Tempering Parameter," *Journal of Heat Treating*, vol. 4, no. 1, pp. 39–46, 1985.

- 7.2 V. K. Euser, "The Effect of Rapid Tempering on Microstructural Evolution and Toughness Within the Tempered Martensite Embrittlement Regime of 4340 and 300M," 2020.
- 7.3 R. L. Brown, H. J. Rack, and M. Cohen, "Stress Relaxation during the Tempering of Hardened Steel," *Materials Science and Engineering*, vol. 21, no. C, pp. 25–34, 1975.
- 7.4 V. K. Euser, A. J. Clarke, and J. G. Speer, "Rapid Tempering: Opportunities and Challenges," *Journal of Materials Engineering and Performance*, vol. 29, no. 7, pp. 4155–4161, 2020.
- 7.5 V. Rudnev, G. A. Fett, and L. Semiatin, "Tempering of Induction Hardened Steels," *Induction Heating and Heat Treatment*, vol. 4, pp. 130–159, 2018.
- 7.6 J. Underwood, L. Pook, and J. Sharples, "Fatigue-Crack Propagation Through a Measured Residual Stress Field in Alloy Steel," *Flaw Growth and Fracture*, pp. 402–402–14, 1977.
- 7.7 G. Krauss, *Steels : Processing, Structure, and Performance*, 3rd ed. Materials Park, Ohio: ASM International, 2005.
- 7.8 W. C. Leslie, *The Physical Metallurgy of Steels*. Herndon, VA: TechBooks, 1991.
- 7.9 D. A. Porter, K. E. Easterling, and M. Y. Sherif, *Phase Transformations in Metals and Alloys*, 3rd ed. Boca Raton, FL: Taylor & Francis Group, 2009.
- 7.10 P. J. Jacques, E. Girault, A. Mertens, B. Verlinden, J. van Humbeeck, and F. Delannay, "The Developments of Cold-Rolled TRIP-Assisted Multiphase Steels. Al-Alloyed TRIP-Assisted Multiphase Steels.," *ISIJ International*, vol. 41, no. 9, pp. 1068–1074, Sep. 2001.
- 7.11 V. Rudnev, G. E. Totten, ASM International, and ASM Heat Treating Society, *ASM handbook. Volume 4C, Volume 4C.*, 2014.
- 7.12 V. K. Judge, J. G. Speer, K. D. Clarke, K. O. Findley, and A. J. Clarke, "Rapid Thermal Processing to Enhance Steel Toughness," *Scientific Reports*, vol. 8, no. 1, pp. 1–6, 2018.
- 7.13 G. Baozhu and G. Krauss, "The Effect of Low-Temperature Isothermal Heat Treatments on the Fracture of 4340 Steel," *Journal of Heat Treating*, vol. 4, no. 4, pp. 365–372, 1986.
- 7.14 J. L. Pacheco and G. Krauss, "Microstructure and High Bending Fatigue Strength in Carburized Steel," *Journal of Heat Treating*, vol. 7, no. 2, pp. 77–86, 1989.
- 7.15 G. Krauss, "Bending Fatigue of Carburized Steels," 1996.
- 7.16 L. Pook, *Metal fatigue : What it is, why it matters*. Dordrecht; London: Springer, 2007.
- 7.17 R. H. Richman and R. W. Landgraf, "Some Effects of Retained Austenite on the Fatigue Resistance of Carburized Steel," *Metallurgical Transactions A*, vol. 6, no. 5, pp. 955–964, 1975.
- 7.18 H. Kristoffersen and P. Vomacka, "Influence of Process Parameters for Induction Hardening on Residual Stresses," *Materials & Design*, vol. 22, no. 8, pp. 637–644, 2001.
- 7.19 G. E. Dieter and D. J. Bacon, "Mechanical Metallurgy." McGraw-Hill, London; New York, 1988.
- 7.20 K. Weiss, V. Rudnev, R. Cook, D. Loveless, and M. Black, "Induction Tempering of Steel," *Advanced Materials and Processes*, vol. 156, no. 2, 1999.
- 7.21 T. Inoue, K. Haraguchi, and S. Kimura, "Analysis of Stresses Due to Quenching and Tempering Steel," *Transastion ISIJ*, vol. 18, 1978.
- 7.22 D. Tong, J. Gu, and F. Yang, "Numerical Simulation on Induction Heat Treatment Process of a Shaft Part: Involving Induction Hardening and Tempering," *Journal of Materials Processing Technology*, vol. 262, no. January, pp. 277–289, 2018.
- 7.23 A. Eser, C. Broeckmann, and C. Simsir, "Multiscale Modeling of Tempering of AISI H13 Hot-Work Tool Steel - Part 2: Coupling Predicted Mechanical Properties with FEM Simulations," *Computational Materials Science*, vol. 113, pp. 292–300, 2016.
- 7.24 G. W. Greenwood and R. H. Johnson, "The Deformation of Metals under Small Stresses during Phase

- Transformations,” *Proceedings of the Royal Society of London. Series A. Mathematical and Physical Sciences*, vol. 283, no. 1394, pp. 403–422, 1965.
- 7.25 F. D. Fischer, G. Reisner, E. Werner, K. Tanaka, G. Cailletaud, and T. Antretter, “New View on Transformation Induced Plasticity (TRIP),” *International Journal of Plasticity*, vol. 16, no. 7, pp. 723–748, 2000.
 - 7.26 D. Hömberg, Q. Liu, J. Montalvo-Urquiza, D. Nadolski, T. Petzold, A. Schmidt, and A. Schulz, “Simulation of Multi-Frequency-Induction-Hardening Including Phase Transitions and Mechanical Effects,” *Finite Elements in Analysis and Design*, vol. 121, pp. 86–100, 2016.
 - 7.27 L. M. Rothleutner, “Assessment of the Microstructure and Torsional Fatigue Performance of an Induction Hardened Vanadium Microalloyed Medium-Carbon Steel,” Colorado School of Mines, 2015.
 - 7.28 G. A. Thomas, J. G. Speer, D. K. Matlock, G. Krauss, and R. E. Hackenberg, “Time-Temperature Equivalence in Martensite Tempering,” in *Proceedings of the International Conference on Martensite Transformations*, pp. 595–600, 2010.
 - 7.29 R. Cryderman, N. Shamsaei, and A. Fatemi, “Effects of Continuous Cast Section Size on Torsion Deformation and Fatigue of Induction Hardened 1050 Steel Shafts,” *Journal of Materials Processing Technology*, vol. 211, no. 1, pp. 66–77, 2011.
 - 7.30 G. A. Fett, “Importance of Induction Hardened Case Depth in Torsional Applications,” *Heat Treating Progress*, vol. 9, no. 6, pp. 15–19, 2009.
 - 7.31 J. H. Hollomon and L. D. Jaffe, “Time-Temperature Relations in Tempering Steel -- Technical Publication No. 1831,” *Metals Technology*, vol. XI, pp. 1–26, 1945.
 - 7.32 H. Bhadeshia and R. Honeycombe, *Steels: Microstructure and Properties*, 3rd ed. Burlington, MA: Butterworth-Heinemann, 2017.
 - 7.33 J. L. Dossett and G. E. Totten, *ASM handbook : Steel Heat Treating Fundamentals and Processes*. Materials Park, OH: ASM International, 2013.
 - 7.34 R. E. Haimbaugh, *Practical Induction Heat Treating, Second Edition*. ASM International, 2015.