ASSESSMENT OF THE MICROSTRUCTURE AND TORSIONAL FATIGUE PERFORMANCE OF AN INDUCTION HARDENED VANADIUM MICROALLOYED MEDIUM-CARBON STEEL

by

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ABSTRACT

Vanadium microalloying of medium-carbon bar steels is a common practice in industry for a number of hot rolled as well as forged and controlled-cooled components. However, use of vanadium microalloyed steels has expanded into applications beyond their originally designed controlled-cooled processing scheme. Applications such as transmission shafts often require additional heat-treatments such as quench and tempering and/or induction hardening to meet packaging or performance requirements. As a result, there is uncertainty regarding the influence of vanadium on the properties of heat-treated components, specifically the effect of rapid heat-treating such as induction hardening.

In the current study, the microstructural evolution and torsional fatigue behavior of induction hardened 1045 and 10V45 (0.08 wt pct V) steels were examined. Torsional fatigue specimens specifically designed for this research were machined from the as-received, hot rolled bars and induction hardened using both scanning (96 kHz/72 kW) and single-shot (31 kHz/128 kW) methods. Four conditions were evaluated, three scan hardened to 25, 32, and 44 pct nominal effective case depths and one single-shot hardened to 44 pct. Torsional fatigue tests were conducted at a stress ratio of 0.1 and shear stress amplitudes of 550, 600, and 650 MPa. Physical simulations using the thermal profiles from select induction hardened conditions were conducted in the Gleeble® 3500 to augment microstructural analysis of torsional fatigue specimens. Thermal profiles were calculated by a collaborating private company using electro-thermal finite element analysis. Residual stresses were evaluated for all conditions using a strain gage hole drilling technique.

The results showed that vanadium microalloying has an influence on the microstructure in the highest hardness region of the induction-hardened case as well as the total case region. Vanadium microalloyed conditions consistently exhibited a greater amount of non-martensitic transformation products in the induction-hardened case. In the total case region, vanadium reduced the total case depth by inhibiting austenite formation at low austenitizing temperatures; however, the non-martensitic constituents in the case microstructure and the reduced total case depth of the vanadium microalloyed steel did not translate directly to a degradation of torsional fatigue properties. In general, vanadium microalloying was not found to affect torsional fatigue performance significantly with one exception. In the 25 pct effective case depth condition, the 10V45 steel had a ~75 pct increase in fatigue life at all shear stress amplitudes when compared to the 1045 steel. The improved fatigue performance is likely a result of the significantly higher case hardness this condition exhibited compared to all other conditions. The direct influence of vanadium on the improved fatigue life of the 25 pct effective case depth condition is confounded with the slightly higher carbon content of the 10V45 steel. In addition, the 10V45 conditions showed a consistently higher case hardness than the in 1045 conditions. The increased hardness of the 10V45 steel did not increase the compressive residual stresses at the surface. Induction hardening parameters were more closely related to changes in residual stress than vanadium microalloying additions. Torsional fatigue data from the current study as well as from literature were used to develop an empirical multiple linear regression model that accounts for case depth as well as carbon content when predicting torsional fatigue life of induction hardened medium-carbon steels.
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To my family
Microalloying of medium-carbon bar steels is common practice in industry for a number of hot-rolled as well as forged and controlled-cooled components. As a result, the influence of vanadium additions on the microstructure, mechanical properties, and fatigue performance of wrought medium-carbon ferrite-pearlite steels is well documented [1-7]. However, use of vanadium microalloyed steels has expanded into applications beyond their originally designed controlled-cooled processing scheme. Applications such as transmission shafts, having specific packaging or performance requirements, often require additional heat-treatments such as quench and tempering and/or induction hardening. Current literature lacks systematic investigations regarding the role vanadium plays in the performance of heat-treated components. This research intends to contribute to the literature by addressing one specific application of interest to the automotive industry – the torsional fatigue performance of induction hardened vanadium microalloyed medium-carbon steel shafts.

The influence of vanadium microalloying has been investigated on various aspects of induction hardening. Vanadium microalloying additions have been shown to result in shallower effective case depths [8]. Vanadium carbide size distribution remain relatively constant throughout the induction-hardened case although slightly larger than precipitates in the core of both as-forged and normalized starting microstructures [9-11]. Bending fatigue has been conducted on induction hardened vanadium microalloyed steels; however, confounding factors kept the influence of vanadium from being examined directly [12]. A systematic investigation into vanadium’s role in specific aspects of microstructural evolution during induction hardening as well as its effect on the torsional fatigue performance of medium-carbon steels may allow further optimization of processing and alloy design, potentially improving component performance. Below are specific research goals of the present study.

1) Characterize the influence of vanadium microalloying on the induction-hardened case and case/core microstructures.

2) Determine the influence of vanadium microalloying on residual stresses developed during induction hardening.

3) Determine the influence of vanadium microalloying on the torsional fatigue performance of induction hardened shafts.

The following chapters present a two-faceted study used to achieve these research goals. The first being a torsional fatigue study of smooth specimens induction hardened to three case depths. The second being a series of Gleeble® physical simulations utilizing predicted thermal profiles from finite element modeling. All testing was conducted using two low-sulfur medium-carbon ferrite-pearlite steels, one with and one without vanadium, supplied specifically for this research. A suite of methods was used to characterize fracture surfaces and microstructure including macro-photography, light optical microscopy, scanning electron microscopy (SEM), and transmission electron microscopy (TEM). Additional methods include dilatometry for Gleeble® physical simulations, strain-gage hole drilling for residual stress measurement, and Vickers microhardness testing.
CHAPTER 2
BACKGROUND & LITERATURE REVIEW

Microalloying effects can be complex and challenging to identify. Concentration of the microalloying element of interest is generally very small and results in very small volume fractions of nano-sized precipitates. The dynamic nature of the induction hardening process with heating and cooling rates as well as peak temperatures that continuously vary within the component increases the complexity of identifying such alloying effects. This chapter introduces fundamental concepts on microalloying, induction hardening, and torsional fatigue. Each concept is introduced in generalities then discussed in the context of vanadium microalloying utilizing literature relevant to the current study.

2.1 Microalloying of Steels

Microalloying is a commonly used alloying strategy in ferrous metallurgy. Small additions, typically less than 0.1 wt pct, of vanadium, titanium, and/or niobium can dramatically affect a steel’s mechanical properties through grain refinement, microstructure modification, and precipitation strengthening. Each can be achieved simultaneously or independently depending on the solubility of the microalloying element and processing routine utilized. The temperature in which the microalloy elements precipitate from solid solution, or vice versa, is determined by the precipitate’s solubility. The solubility of a microalloy compound is of critical importance during processing because it dictates the size and volume fraction of precipitates, both of which can influence grain size as well as precipitation strengthening. As a result, both alloying and processing must be coordinated and controlled to maximize utilization of the microalloying additions.

2.1.1 Solubility of Microalloy Carbides and Nitrides

Microalloy compounds are rarely stoichiometric due to mutual solubility between compounds. For example, carbides and nitrides often create carbonitrides due to the presence of both carbon and nitrogen in most steels. Although calculations can be made for off-stoichiometric compounds, pure 1:1 stoichiometric compounds are typically used as a first approximation for equilibrium calculations

\[ M + nX \leftrightarrow MX_n \]  \hspace{1cm} (2.1)

where \( M \) is the microalloying element and \( X \) is the interstitial element. The inverse of the equilibrium constant for a compound is the solubility product, \( K_s \), and is expressed as a function of temperature as

\[ \log_{10} K_s = \log_{10} [M][X]^n = A - \frac{B}{T} \]  \hspace{1cm} (2.2)

where \( K_s \) is a product of the microalloying element, \( M \), and the interstitial element, \( X \), in wt pct, \( n \) is an exponent representing the stoichiometry, \( A \) and \( B \) are constants, and \( T \) is the absolute temperature [13]. Figure 2.1 shows microalloy solubility products from literature for carbides and nitrides of primary interest. Aluminum and boron are not typically considered microalloying elements; however, solubility of their nitrides are presented for completeness. Table 2.1 shows the solubility products for vanadium carbide (VC\(_{0.75}\)), vanadium nitride (VN), and aluminum nitride (AlN) in both ferrite and austenite. These specific compounds will be discussion throughout the study. In general,
carbides are more soluble than nitrides in a given phase with ferrite having a lower solubility than austenite for a given compound. All of the compounds exhibit a significant drop in solubility between austenite and ferrite. The sudden decrease in solubility results in a substantial driving force for precipitation during the transformation from austenite to ferrite; although, not all of the compounds presented have been observed to precipitation strengthen (e.g. AlN).

Figure 2.1 Solubility products for microalloy (a) carbides and (b) nitrides as a function of temperature in ferrite and austenite. Solubility products from Turkdogan [14].

<table>
<thead>
<tr>
<th>Compound</th>
<th>Ferrite</th>
<th>Austenite</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>A</td>
<td>B</td>
</tr>
<tr>
<td>VC$_{0.75}$</td>
<td>4.24</td>
<td>7050</td>
</tr>
<tr>
<td>VN</td>
<td>3.90</td>
<td>9720</td>
</tr>
<tr>
<td>AlN</td>
<td>2.05</td>
<td>8790</td>
</tr>
</tbody>
</table>

Vanadium microalloying is commonly used in medium-carbon steels due to its high solubility in medium-carbon austenite. The high solubility provides considerable processing flexibility and ensures the vanadium is in solid solution during processing, allowing maximum precipitation strengthen to occur during direct-cooling after hot rolling or forging [15]. Lower temperature precipitation provides finer carbonitrides, which strengthen to a greater degree than coarse precipitates for a given volume fraction. Many carbides and nitrides have mutual solubility and form complex carbonitrides, with the nitride having the lowest solubility (i.e. precipitates at the highest temperature). As a result, the first precipitate to form during cooling is nitrogen rich and transitions to carbon rich at low temperatures. Consequently, microalloy precipitates will be referred to as carbonitrides for the remainder of this study unless discussing a specific microalloy compound.
2.1.2 Austenite Grain Size Control

Austenite grain refinement in microalloyed steels typically occurs by two mechanisms. The first is by solute atoms introducing a drag force on grain boundaries (i.e. solute drag) impeding boundary movement. The second is by fine precipitates inhibiting grain boundary mobility [16]. Both solute atoms and fine precipitates can also affect phase transformation behavior by influencing hardenability, which will be discussed later in the chapter, as well as recrystallization during processing [13].

Grain growth occurs in materials to minimize grain boundary energy. Microalloy precipitates can inhibit grain boundary mobility by reducing the grain boundary surface area due to the intersection of the precipitates with the grain boundary. When the intersecting surface area is at a maximum, the surface energy is at a minimum, resulting in a pinning force on the grain boundary. Zener first derived an equation describing the pinning force on a grain boundary assuming a shrinking isolated matrix grain

$$ R = \left( \frac{4}{3} \right) \frac{r}{f_v} \tag{2.3} $$

where $R$ is the pinned matrix grain radius, $r$ is the precipitate radius, and $f_v$ is the precipitate volume fraction [16]. Gladman derived a more sophisticated equation incorporating a growing tetrakaidecahedron (i.e. 14-sided polyhedral) grain in a matrix of smaller pinned grains

$$ R_o = \left( 1 - \frac{4}{3Z} \right) \frac{r}{f_v} \tag{2.4} $$

where $R_o$ is the pinned matrix grain radius, $r$ is the precipitate radius, $Z$ is the size advantage factor, and $f_v$ is the precipitate volume fraction. The size advantage factor describes the advantage a growing grain has over other matrix grains and is equal to $R/R_o$, where $R$ is the growing grain radius. Due to the structure of the equation, only polycrystalline grain structures with a size advantage greater than $4/3$ result in a real solution for $R_o$. Experimental observations of the size advantage factor typically range from 1.4 to 2.0 [16]. Both equations indicate that the force required to pin a grain increases as the matrix grain size decreases, requiring a higher volume fraction or smaller precipitates to maintain the finer grain size. For example, a microalloyed steel with 0.08 wt pct vanadium can result in a maximum precipitate volume fraction of ~0.0013. Using Equation 2.3, a grain size of 10 µm (ASTM No. 10) can be maintained with a precipitate size of ~10 nm and a grain size of 5 µm (ASTM No. 12) can be maintained with a precipitate size of ~5 nm.

2.1.3 Precipitation Strengthening

The precipitation strengthening effect of microalloying is a result of fine carbonitrides precipitating within the grains (i.e. intragranularly) as non-shearable or hard particles that impede dislocation slip [17]. Precipitates fine enough to significantly strengthen a steel form either during the austenite-to-ferrite transformation (i.e. interphase precipitation) or randomly within ferrite. Figure 2.2 shows examples of both interphase and random precipitation of vanadium carbonitrides. Interphase precipitates appear as aligned rows, or sheets, parallel to the advancing austenite/ferrite interface while random precipitates have no apparent alignment. Microalloy precipitates have an
NaCl (B1) crystal structure which results in both precipitation behaviors having a Baker-Nutting (B-N) orientation relationship with the ferrite matrix [13].

\[
{\{100\}}_a // {\{100\}}_{MX} \quad (011)_a // (010)_{MX}
\]  

The NaCl structure of the microalloy carbides and nitrides results in very high hardness and elastic modulus mismatch with the ferrite matrix. For example, vanadium carbide (VC) has a Vickers hardness of 27.2 GPa and an elastic modulus of 430 GPa [19] while ferrite has a Vickers hardness of 3.6 GPa and an elastic modulus of 203 GPa [20]. This results in the VC having a 656 pct higher hardness and 112 pct higher elastic modulus than the ferrite. As a result, microalloyed steels are a model system for application of the Ashby-Orowan precipitation strengthening model.

![Figure 2.2 Bright field TEM micrographs of (a) interphase precipitation and (b) randomly precipitated vanadium carbonitrides in ferrite [18].](image)

The Ashby-Orowan model was derived for small volume fractions of fine, hard precipitates randomly intersecting a slip plane. As a dislocation glides across a slip plane, it interacts with the precipitates resulting in an increase in dislocation line tension. The degree of dislocation line bowing around a precipitate is then related to applied stress. After taking into account the edge-to-edge precipitate spacing the following equation is developed

\[
\Delta\sigma_y = 0.538 \left( \frac{G_m b f^{0.5}}{X} \right) \ln \left( \frac{X}{6.125 \times 10^{-4}} \right)
\]  

where \(\Delta\sigma_y\) is the change in yield strength in MPa, 0.538 is a pure constant, \(G_m\) is the matrix shear modulus in MPa (80,600 for ferrite), \(b\) is the Burger’s vector (2.5 \(\times\) 10\(^{-4}\) \(\mu\)m), \(f\) is the precipitate volume fraction, \(X\) is the precipitate diameter in \(\mu\)m, and 6.125 \(\times\) 10\(^{-4}\) is a constant with units of \(\mu\)m [13]. Figure 2.3 shows an application of the Ashby-Orowan equation for alloy design using microalloying additions of 0.08 wt pct vanadium. The maximum yield strength increase due to precipitation strengthening that can be achieved is approximately 200 MPa with a precipitate diameter of 3 nm. However, if the precipitation process is only 70 pct efficient, increases in yield strength between approximately 160 and 125 MPa can still be achieved if precipitate diameter is maintained between 3 and 6 nm.
2.1.4 Structure-Property Relationships in Ferrite-Pearlite Steels

Structure-property relationships have been developed for a wide variety of steel compositions and microstructures. The relationships quantify the independent contributions of chemistry and microstructure to mechanical properties such as yield strength, ultimate tensile strength, and fatigue limit [1, 2, 21, 22]. Structure-property relationships for microalloyed steels are challenging due to the varied impact of each element on multiple strengthening mechanisms, such as grain refinement and precipitation strengthening. A detailed study by Gladman et al. [1] provides some insight on the influence of microalloying on the yield and ultimate tensile strength of medium and high-carbon (0.40 to 0.80 wt pct C) ferrite-pearlite steels.

Gladman et al. developed multiple linear regression models for yield strength and ultimate tensile strength for 40 different plain-carbon and microalloyed steels in the normalized, air cooled from 1100 °C, and control-rolled conditions. A modified law of mixtures approach was used to account for the non-linear behavior of yield and ultimate tensile strength as pearlite fraction increases. The model developed for yield strength in MPa is

$$\sigma_y = f^{1/3} \left( 35 + 58.5 \cdot Mn + 17.4 \cdot d^{-1/2} \right) + \left( 1 - f^{1/3} \right) \left( 178 + 3.85 \cdot s^{-1/2} \right) + 63.1 \cdot Si + 425 \cdot \sqrt{N_{free}}$$

(2.7)

where all of the elements have units of wt pct, $f$ is the volume fraction of ferrite, $d$ is the ferrite grain size in mm, and $s$ is the interlamellar pearlite spacing in mm. The relationship accounts for 93.7 pct of the variation observed in yield strengths. In this model, manganese ($Mn$) was found to significantly influence the ferrite fraction while silicon ($Si$) and free (or dissolved) nitrogen ($N_{free}$) only provided a solid solution strengthening component. The equation determined for ultimate tensile strength in MPa is
where all variables were previously defined with Equation 2.7. The relationship accounts for 95.5 pct of the variation observed in tensile strengths. In this model, free nitrogen is found to significantly influence ferrite content while Si only provides solid solution strengthening. Figure 2.4 shows a plot of calculated versus observed yield and ultimate tensile strength for the steels used to develop the empirical structure-property relationships. The vanadium microalloyed steels typically result in a significantly higher observed yield strength (Figure 2.4a), which is believed to be the strength contribution due to precipitation strengthening, which is not contained in the model [1]. In many cases vanadium microalloyed steels exhibit higher ultimate tensile strengths than the other steels (Figure 2.4b), although not as significantly as yield strength. The study by Gladman et al. suggests V is the only microalloy element that can significantly influence the yield strength of medium-carbon steels under the conditions examined. This result is particularly important for the present project because microalloyed medium-carbon steels are typically induction hardened in the as-forged or as-hot rolled conditions that are very similar to the conditions examined by Gladman et al.

Microalloying has also been shown to result in fatigue performance equivalent to quenched and tempered (Q&T) alloy steels at the same hardness [23]. This observation has two important implications. First, desired fatigue performance can be achieved from direct cooling a microalloyed steel versus the costly Q&T processing of more expensive alloy steel. Second, the fatigue performance of steels with very different microstructures, precipitation-strengthened ferrite-pearlite and high temperature tempered martensite, is nearly the same as long as they have the same hardness.
The factors that influence the fatigue behavior of microalloyed low and medium-carbon ferrite-pearlite steels were quantified in a study by Abe et al. [2]. Abe et al. utilized a model that summed static strengthening mechanisms to quantify their independent contribution to the overall fatigue limit of the steels examined. The general form of their equation for the fatigue limit in MPa is

\[ \sigma_w = \sigma_{wo} + A \cdot \sigma_{ss} + B \cdot \sigma_{ppt} + C \cdot \sigma_{prlt} + D \cdot \sigma_{dis} + E \cdot \sigma_{gr} \]

(2.9)

where \( \sigma_{wo} \) is the friction stress, \( \sigma_{ss} \) is the solid solution strengthening component, \( \sigma_{ppt} \) is the precipitation strengthening component, \( \sigma_{prlt} \) is the pearlite strengthening component, \( \sigma_{dis} \) is the dislocation strengthening component, \( \sigma_{gr} \) is the ferrite grain size component, and A through E are coefficients. The ferrite grain size component was further defined to be a Hall-Petch type relationship

\[ \sigma_{gr} = K \cdot d^{-1/2} \]

(2.10)

where \( d \) is the ferrite grain size in mm and \( K \) is the strengthening coefficient. The equation developed by Abe et al. as a result of the study is

\[ \sigma_w = 8.4 + 0.92 \cdot \sigma_{ss} + 0.70 \cdot \sigma_{ppt} + 0.53 \cdot \sigma_{prlt} + 0.43 \cdot \sigma_{gr} + 0.23 \cdot \sigma_{dis} \]

(2.11)

This equation indicates that under the conditions analyzed by Abe et al., solid solution and precipitation strengthening influence the fatigue limit the most for low and medium-carbon ferrite-pearlite microalloyed steels.
The above relationships have convincingly shown the positive influence fine microalloy precipitates can have on yield strength, ultimate tensile strength, and fatigue limit. However, additional heat treatment of microalloy steels after hot rolling or forging can have a negative effect on properties. Figure 2.6 shows that both yield strength and fatigue limit are affected by normalizing hot rolled plain carbon and microalloyed ferrite-pearlite steels [2]. While the plain carbon steels exhibit both an increase in yield strength and fatigue limit as a result of the normalizing heat treatment, performance of the vanadium microalloyed steel degrades significantly. Grain refinement due to AlN precipitation is likely the contributing factor for the improved properties of the plain carbon steels after normalizing, while precipitate coarsening likely reduced the properties of the vanadium microalloyed steels.

![Figure 2.6 Influence of normalizing heat treatment on the fatigue limit and yield strength of hot-rolled plain carbon and vanadium microalloyed steels. Adapted from Abe et al. [2].](image)

### 2.1.5 Hardenability

Although microalloying additions undoubtedly increase properties such as strength and fatigue performance, they can also adversely influence phase transformation behavior through effecting hardenability. A smaller austenite grain size has been shown to markedly, and negatively, influence hardenability (i.e. Jominy hardenability) in plain carbon steels [24]. However, the influence of microalloy elements, such as vanadium and niobium, is not as straightforward as a simplistic austenite grain refinement mechanism. Depending on the alloy composition and the austenitizing temperature, microalloying elements may be in solution, precipitating from austenite, or both which may all impact hardenability differently.

Figure 2.7 shows the influence of vanadium on hardenability as a function of reheat temperature from Grossmann [24] as well as data currently used in the ASTM International standard for determining the hardenability of steels [25]. Although the current ASTM standard shows no dependence of reheat temperature on vanadium’s hardenability effect, Grossmann showed a clear effect in data from 1952. At low vanadium levels, VC can readily go
into solid solution, even at low reheat temperatures, and can have a very large impact on hardenability. As vanadium
content increases, higher reheat temperatures are required to get the vanadium into solid solution [24].

The mechanism for the influence of microalloying additions on hardenability has been examined further
since Grossmann [26-30]. With vanadium microalloying additions, a grain boundary pinning mechanism was
introduced by Garbarz and Pickering [26, 27] and later supported by Adrian [28] as well as Adrian and Staško [29].
Garbarza and Pickering found if austenite grain boundaries are not inhibited, the boundary moves too quickly for
segregation to occur, decreasing hardenability. However, hardenability is enhanced if austenite grain boundaries are
inhibited by undissolved carbonitrides, allowing V to segregate to the grain boundaries, decreasing boundary surface
energy and inhibiting the diffusional transformation of austenite. Nucleation of non-martensitic transformation
products may be favorable when carbonitrides coarsen and are no longer effective at inhibiting grain growth. Work
by Fossaert et al. [30] suggests the opposite effect is occurring with niobium microalloy additions. Niobium in
solution was shown to increase hardenability by possible segregation to the austenite grain boundaries, but the effect
decreases when the niobium is precipitated as carbonitrides.

![Figure 2.7 Multiplying factors for calculating the effect of vanadium on the hardenability of steel. Adapted from Grossman [24] and ASTM-A255 [25].](image)

2.2 Induction Hardening of Steels

Induction hardening is a rapid heat-treating processes used to selectively surface harden regions of a
component. High-frequency alternating current is passed through a specially designed copper coil positioned near
the work piece. The magnetic field produced by the high-frequency alternating current induces eddy currents in the
work piece that, in turn, heats the work piece via Joule heating. The depth at which the magnetic field penetrates in
to a work piece is inversely proportional to the frequency and is known as the “skin effect.” Coil geometry and setup
selection are dictated by component design, required productivity, and/or equipment availability. Shafts are either
induction hardened using a scanning (progressive) or single-shot (stationary) coil. In either case, the shaft is rotated
during hardening to ensure even heating. Once the shaft is heated to the required surface temperature and radial
depth, it is quenched using water with or without a polymer quenchant additive, typically. Process parameters such as coil design, coil/part separation distance, scan speed, frequency, and power as well as quenchant concentration, flow rate, and initial temperature can influence case depth, microstructure, hardness, distortion, and residual stresses.

Figure 2.8 shows the power density and temperature evolution in a cylindrical bar of steel. At very short times, all of the power is focused near the surface. After the surface reaches the Curie point (770 °C) the efficiency of Joule heating decreases and the power density drops significantly at the surface, reducing the heating rate at the surface. However, if frequency and power are sufficiently high, surface heating rate slows very little. In medium-carbon steel shafts, the frequency and power is controlled to heat the surface to an approximate peak temperature of 950 °C (typically >50 °C higher for microalloy steels) as quickly as possible before being quenched [31]. Keeping the heating rate high reduces the amount of time in the austenite phase field, limiting austenite grain growth and assisting in control of heat penetration depth [32]. The result of this surface hardening process is a very high strength martensitic case around a ferrite-pearlite core that often provides significantly better mechanical properties over conventional heat-treatments [31].

Figure 2.8  (a) Power density and (b) temperature as a function of cylindrical bar radius and elapsed time. Adapted from Haimbaugh [32].

### 2.2.1 Microstructure

Three distinct microstructural regions are seen in the cross-section of an induction-hardened shaft: case, core, and case/core transition. Figure 2.9 shows an example macrograph of an induction-hardened shaft indicating the case, core, and case/core transition regions. The case region is fully martensitic (ideally); however, retained austenite on the order of a couple percent can be expected in the 0.40 to 0.60 wt pct carbon range [23]. The microstructure of the case/core transition region can consist of a broad range of constituents depending on the processing parameters, exact radial location, core (i.e. prior or starting) microstructure, and alloy content. For a ferrite-pearlite core microstructure, typical of induction hardened vanadium microalloyed medium-carbon steels [8],
the transition region can be a mix of martensite, bainite, partially transformed pearlite, and retained ferrite. Improper selection of processing parameters can result in undissolved carbides and retained ferrite in the case, which can lead to detrimental properties.

Induction hardening generates large internal thermal gradients that are extremely difficult to measure. As a result, thermal profiles are often modeled using computer simulations instead of measured directly using imbedded thermocouples. Figure 2.10 shows predicted induction hardening thermal profiles at different depths in a 35 mm diameter bar of steel. Three regions can be identified in the near surface thermal profiles: heating, dwell (often very short), and quench. After heating to a target peak temperature, a dwell period can be used to provide additional austenitizing time to dissolve alloy carbides and nitrides as well as allow heat to diffuse into the bar, increasing case depth [31]. Figure 2.10 shows a decrease in temperature during the dwell stage, which is due to radiative heat loss at the surface as well as conductive heat loss into the shaft. As the distance from the surface increases, changes in heating rate, peak temperature, and time above critical temperatures (Ac$_1$ and Ac$_3$) vary greatly and result in a significant microstructural gradient upon quenching. Figure 2.11 shows the influence of heating rate on the homogeneity of austenite from a ferrite-pearlite Ck45 steel. The Ac$_3$ temperature increases more than Ac$_1$ as heating rate increases. Approximate combinations of heating rate and peak temperature for the case and case/core transition regions are indicated. In general, induction hardening results in either inhomogeneous austenite or a mixture of ferrite, pearlite, and austenite before quenching. Inhomogeneous austenite consists of significant compositional gradients as well as undissolved alloy carbides, both of which can result in significant differences in local hardenability. For example, if just carbon compositional gradients from 0.10 to 0.80 wt pct carbon are considered for a plain medium-carbon steel, the critical cooling rate from Ac$_3$ to create a fully martensitic microstructure varies from 411 to 16 °C/s, respectively [35]. Variation in carbon content of the martensite in the case, along with very fine austenite grain size, is one explanation for a phenomenon sometimes seen in induction-hardened components called superhardness [36]. Superhardness is simply an increase in surface hardness of 2-4 HRC above the expected martensite hardness levels observed during furnace hardening. Other studies argue the superhardness phenomenon is
not a result of inhomogeneous martensite; instead, a result of the high cooling rate of the thin induction heated layers and is independent of heating rate [37]. Refining the austenite grain size from ASTM 7 to 14 was found to enhance the proposed superhardness effect by 2 HRC [38].

Figure 2.10  Predicted induction hardening thermal gradients generated using electro-thermal modeling. Plot adapted from Li et al. [33].

Figure 2.11  Austenite transformation behavior of a ferrite-pearlite Ck45 steel as a function of heating rate. Adapted from the Orlich et al. [34].
2.2.2 Residual Stresses

The microstructural gradient from the martensitic case to the ferrite-pearlite core results in a large hardness gradient within the induction-hardened shaft. Lattice expansion due to the martensite formation results in large compressive residual stresses at the surface transitioning to tensile residual stresses in the core. Figure 2.12 shows the resultant hardness and residual stress profiles produced during induction hardening. In the case, the hardness as well as the compressive axial and circumferential (i.e. hoop) residual stresses are very high. The hardness rapidly decreases to the core hardness and the residual stresses begin to transition to tensile further into the shaft. The radial residual stress can only be measured using bulk analysis techniques (e.g. neutron diffraction) which are rarely used in industry. However, the radial component can be calculated from biaxial residual stress measurements made incrementally as function of depth in to the component using X-ray diffraction or strain gage drilling methods [40, 41]. The residual stress profile in Figure 2.12 shows the radial component increases from zero at the surface towards the shaft center, although relatively low in magnitude. The induction hardening processing parameters affect the residual stresses. Shallower case depths (i.e. high frequencies) tend to result in higher surface compressive residual stress [42]. Kristoffersen and Vomacka conducted a systematic study varying starting microstructure, frequency, power, heating time, and peak temperature of an AISI 4140 steel confirming this observation [43]. Residual stresses are completely elastic and balanced, between compressive and tensile, within the shaft. As such, residual stresses can greatly influence high-cycle fatigue properties because fatigue damage incurs well below the macroscopic yield strength of a material [44].

![Figure 2.12](image-url)  
Figure 2.12  Vickers hardness and residual stress profiles for an induction hardened medium-carbon steels. Adapted from Yonetani and Isoda [39].

2.2.3 Induction Hardenability

The phrase *induction hardenability* has been used in literature to describe the influence of alloying additions on case depth achieved during induction hardening [45, 46]. Studies by both Fett and Held [45] as well as
by Urita and Namiki [46] found that vanadium additions resulted in a shallower case depth for a given induction hardening routine. Fett and Held also found that with the right thermal treatments (i.e. forging) vanadium steels can meet induction hardened case depths; however, an “inferior” case structure was observed. Ferrite and bainite was present at the grain boundaries instead of martensite, which greatly reduced torsional fatigue performance. Fett and Held emphasized that this decrease may be a sulfur effect due to the use of resulfurized steels in some of the conditions examined.

The phrase induction hardenability is misleading because hardenability can be defined many different ways [23], all of which refer to the formation of a high hardness constituent upon cooling. The added implication in all of these definitions is that cooling occurs from a fully austenitic microstructure. The study by Fett and Held [45] as well as by Urita and Namiki [46] compared the case depth at a specified hardness – 40 HRC for Fett and Held, 450 HV for Urita and Namiki – against the composition of each alloy. The specified depth of hardness for both studies is significantly below the expected hardness for a fully martensitic microstructure. As a result, an aggregate consisting of as-quenched martensite and bainite as well as the starting microstructure is being measured. These mixed microstructures suggest the behavior actually being examined is the effect of alloying additions on austenite nucleation and growth kinetics at low austenitizing temperatures and not that of either martensite or bainite formation upon cooling from austenite.

2.2.4 Precipitate Dissolution

Depending on the specific alloy composition, heating rate, peak temperature, time at peak temperature, and cooling rate, a microalloying element can be in a variety of different forms simultaneously. Whether the microalloying element is residing in solid solution, precipitating, coarsening, or dissolving/reverting, it can directly impact subsequent phase transformations as well as precipitation strengthening response. Consequently, characterizing and understanding precipitate dissolution during induction hardening is essential to understanding the effect of a specific microalloying element.

Dissolution kinetics of precipitates was first described by Aaron [47] for the one-dimensional diffusion controlled dissolution of planar precipitates. Whelan [48] extended this analysis to the three-dimensional dissolution of spherical precipitates, which is more applicable to microalloyed steels. Figure 2.13 shows schematically the concentration profiles of the rate limiting species as a function of time for a dissolving precipitate. The mathematical model assumes a stationary interface between the precipitate and the matrix; therefore, concentration $C_m$ represents a local equilibrium condition that can be approximated using a solubility product. The concentration profiles can be related by

$$k = 2 \left( \frac{C_m - C_o}{C_p - C_m} \right)$$

(2.12)

where $k$ is the super saturation parameter, $C_m$ is the equilibrium concentration of the rate limiting species at the interface, $C_o$ is the matrix concentration far away from the precipitate, and $C_p$ is the precipitate concentration. The general form of the equation derived by Whelan is
\[
\frac{dR}{dt} = -\frac{kD}{2R} - \frac{k}{2} \sqrt{\frac{D}{\pi t}}
\]  

(2.13)

where \( R \) is the radius of the dissolving precipitate in meters, \( t \) is time in seconds, \( D \) is the diffusion coefficient of the rate limiting species in m\(^2\)/s, and \( k \) is the supersaturation parameter, which is unitless. The \( 1/R \) term arises from the steady-state section of the diffusion field while the \( \sqrt{t} \) term arises from the transient part. The diffusion coefficient follows an Arrhenius-type relationship

\[
D = D_o \cdot \exp \left( \frac{-E_A}{R_o T} \right)
\]

(2.14)

where \( D_o \) is the frequency factor in m\(^2\)/s, \( E_A \) is the activation energy in kJ, \( R_o \) is the universal gas constant \((8.314 \times 10^{-3} \text{kJ/mol} \cdot \text{K})\), and \( T \) is the absolute temperature. At short times, an approximate solution for Equation 2.12 is

\[
R = R_o - \frac{kD}{2R_o} - \frac{k}{\pi} \sqrt{Dt}
\]

(2.15)

where \( R_o \) is the initial precipitate radius in meters, \( \pi \) is the constant 3.1416, and all other variables have been previously identified.

Figure 2.13  Schematic of solute concentration profiles of a dissolving precipitate as a function of time. Adapted from Whelan [48].

Aaron and Kotler extended the relationship developed by Whelan to account of curvature effects on both spherical and planar precipitates [49]. A modified Gibbs-Thompson equation was used to describe the composition of the matrix as well as the precipitate/matrix interface as a function of precipitate radius. Figure 2.14 schematically shows the effect of curvature on precipitate dissolution rate. The dissolution rate of planar and spherical precipitates is identical and very high during the early stages of dissolution. The rate decreases markedly due to the change in the concentration gradient from an initial “infinite” value. The planar and spherical precipitate dissolution rates diverge.
after some time, with the dissolution rate of the planar precipitate continuing to decrease while the spherical precipitate increases significantly due to the Gibbs-Thompson effect. Dissolution of precipitates in microalloyed steels, having a fine precipitate size, is likely highly influenced by curvature effects.

![Figure 2.14 Schematic of dissolution velocity for spherical and planar precipitates as a function of size and elapsed time. Adapted from Aaron and Kotler [50].](image)

In a study by Rivas et al. [9-11] a vanadium microalloyed medium-carbon steel (0.11 wt pct vanadium and 0.30 wt pct carbon) was forged at 1315 °C, cooled to room temperature, and induction hardened. Figure 2.15 shows the results of carbon extraction replicas taken at different depths in the component. Even though case depth was not reported, the two measurements closest to the surface were indicated as being within the case region and the sampling depth furthest into the forging was the core. The average particle diameter was observed to increase from 7 nm near the core to 12.6 nm near the surface. Even though the overall increase in the average particle diameter from core to surface is not statistically significant, Rivas et al. indicated that there was evidence of dissolution and reprecipitation of the finer carbides. Law and Edmonds found that fine VC can dissolve below their equilibrium solubility temperature by as much as 220 °C [51]. The same study showed 2 nm thick VC\textsubscript{0.75} dissolved in only 15 s at 900 °C. With diffusional processes being much more sensitive to temperature than time at temperature, this has significant implications during induction hardening at peak temperatures in excess of 1000 °C. The mechanism for the dissolution of VC\textsubscript{0.75} at much lower temperatures and times than anticipated is due to very specific series of events. First, the VC\textsubscript{0.75} originally precipitated in ferrite. Upon heating, the ferrite transforms to austenite and a sudden increase in interfacial energy occurs due to the precipitate being in contact with a face-centered cubic structure instead of a body-centered cubic structure [52]. The increase in interfacial energy accelerates the reversion kinetics of the VC\textsubscript{0.75}. 
2.2.5 Mechanical Properties

Induction hardening of medium-carbon steel shafts has been shown to significantly increase static torsional [53, 54] and bending [55] strength as well as fatigue life [56, 57] and wear resistance [31, 58]. Increased surface hardness greatly improves wear resistance, while compressive residual stresses inhibit crack nucleation, which positively influences both fatigue life and wear resistance [58, 59]. Two values are typically specified during component certification because they have a direct influence on mechanical properties: effective and total case depth [31]. The effective case depth is the hardened depth to a specific hardness that is a function of the steel’s carbon composition. For steels with a carbon content of 0.43-0.53 wt pct, the effective case depth is the depth to a hardness of 450 HV (45 HRC) [60, 61]. The total case depth is the depth of the case determined optically or the depth at which the hardness is equal to the core hardness [61].

Fett [62] indicated in a paper on the influence of case depth on torsional performance that the effective case depth is closely associated to torsional strength and total case depth is a reasonable predictor for fatigue performance. Ochi and Koyasu [53] showed that torsional strength is closely related to the equivalent hardness. An equation for equivalent hardness developed by Ochi and Koyasu relating the shear yield stress distribution in the shaft cross-section to hardness is

\[ H_{eq} = \frac{3}{R^3} \int_0^R HV(r) r^2 dr \]  

(2.16)

where \( R \) is the shaft radius in mm and \( HV \) is the Vicker’s hardness as a function of radial distance, \( r \), in mm.

Figure 2.16 shows the relationship between equivalent hardness and torsional strength for studies by Ochi and Koyasu as well as Cunningham et al. [54]. Ochi and Koyasu examined the torsional strength of steels varying in carbon content (0.41-0.71 wt pct) as well as effective case depth (21-100 pct of radius). A linear trend was observed with a coefficient of determination of 0.91 for 34 specimens. However, the study by Cunningham et al. did not show...
as strong of a trend. Cunningham et al. studied the effect of cold work, induction hardening method, and case depth on the torsional strength of a vanadium microalloyed steel. Increasing the case depth significantly increased the torsional strength, as seen by Ochi and Koyasu. However, increasing the percent cold work only slightly increased the torsional strength and changing processing method was observed to have no effect.

![Figure 2.16](image)

**Figure 2.16** Torsional strength as a function of equivalent hardness for various induction hardened steels. Data from Ochi and Koyasu [53] and Cunningham et al. [54].

### 2.3 Torsional Fatigue

Fatigue damage is accumulated in a component due repeated cyclic loading. Torsional fatigue and bending fatigue are common in shaft-type applications; however, torsional fatigue is the more aggressive of the two loading routes [63]. Figure 2.17 shows a sinusoidal torsional loading scheme. Although a large number of descriptors can be used to define the fatigue loading process, only those relevant to the present study will be discussed. The minimum and maximum shear stress are self-explanatory, shown as \( \tau_{\text{min}} \) and \( \tau_{\text{max}} \) respectively. The shear stress range is defined as

\[
\tau_r = \tau_{\text{max}} - \tau_{\text{min}}
\]

and is rarely seen in literature except to calculate the shear stress amplitude

\[
\tau_a = \frac{\tau_r}{2} = \frac{\tau_{\text{max}} - \tau_{\text{min}}}{2}
\]

which is one of the primary descriptors used in presenting torsional fatigue data. The mean shear stress is defined as

\[
\tau_m = \frac{\tau_{\text{max}} + \tau_{\text{min}}}{2}
\]

and is the topic of some controversy within the fatigue literature, to be discussed later in this section. Often values for mean shear stress are not reported, instead the stress (or load) ratio, commonly referred to as \( R\)-ratio,

\[
R = \frac{\tau_{\text{min}}}{\tau_{\text{max}}}
\]
is substituted. Fatigue testing can be conducted at any R-ratio; however, laboratory tested smooth or notched specimens are commonly tested at logical values such as -1, 0, and 0.1. An R-ratio of -1 is referred to as reversed stress or fully reversed fatigue testing and is the result of a mean stress of zero. As the mean stress is increased, the R-ratio becomes positive and the specimen no longer travels through a time where it is at zero stress, as seen in Figure 2.17.

![Figure 2.17 Schematic of repeated stress torsional fatigue loading. Adapted from Dieter [44].](image)

Fatigue testing can be conducted in either low-cycle or high-cycle regimes, the demarcation typically being around $10^4$ cycles. Low-cycle fatigue testing is typically used to understand fatigue damage due to large strains because of cyclic thermal stresses or specific microstructural effects. However, most engineering data are developed in the high-cycle regime because damage is accumulated at stresses well below the macroscopic yield strength of the material, making it of critical importance in engineering design [44]. During a fatigue test, a steel is considered to have reached its fatigue limit once it has been tested to $10^7$-$10^8$ without failure. Fatigue limit has been shown to correlate well to hardness (or ultimate tensile strength) in steels below 400 HV and to defect populations above approximately 400 HV [64].

Fatigue fractures typically exhibit three separate stages: Stage I – microcrack initiation, Stage II – microcrack propagation, and Stage III – final overload [65]. Stage I is defined as the formation of a microcrack from fatigue damage and extension of that microcrack without change in direction. Stage II occurs when the Stage I microcrack has reached a critical length and changes direction such that the crack propagates normal to the maximum tensile stress. Stage III is therefore when the Stage II microcrack propagates to a critical length and the remainder of the specimen or component fails within a single loading cycle. Microcracks propagate normal to the maximum tensile stress in Stage II, is referred to as Mode I crack extension. Figure 2.18 shows schematic explaining opening (Mode I), sliding (Mode II), and tearing (Mode III) crack extension modes. Mode I is when the crack is extended due to a tensile stress acting normal to the plane of the crack. Mode II is when the crack is extended due to a shear stress acting parallel to the plane of the crack and perpendicular to the crack front (i.e. in-plane shear). Mode III is when the crack is extended due to a shear stress acting parallel to the crack plane and parallel to the crack front (i.e. out-of-plane shear).

Mean stress effects have been well defined in uniaxial and bending fatigue [44]; however, mean shear stress effects are not clear and subject to some debate in literature. Hurd and Irving [66] showed that R-ratio does
not have a significant role in torsional fatigue life. Figure 2.19 shows two strain-life curves for a quench and tempered (300 °C) medium-carbon steel tested at stress ratios of 0 and -1. Findley [67] observed that the effect of mean shear stress is conditional, and no effect was observed as long as the maximum shear stress did not exceed the yield strength. Wang and Miller [68-70] conducted torsional fatigue tests at different mean stresses on a single quenched and high temperature tempered NiCrMo steel (280 HV). Tests ranged in shear stress amplitude from 277 to 330 MPa and R-ratios from -0.42 to -1. As a result, Wang and Miller developed a relationship for shear stress amplitude in MPa as a function of cycles to failure, $N_f$, and mean shear stress in MPa which is

$$\tau_a = 675 \cdot \exp\left(\frac{-|\tau_m|}{797}\right) \cdot N_f^{-0.063} \quad (2.21)$$

This relationship indicates that increasing the mean shear stress decreases the cycles to failure for a given shear stress. However, a study by Davoli et al. [71] showed that the effect of mean shear stress on torsional fatigue limit is not statistically significant, examining data from literature as well as a material and fatigue testing parameters nearly identical to Wang and Miller.

Figure 2.18 Crack extension mode for (a) opening – tensile stress normal to the plane of the crack, (b) sliding – shear stress parallel to the plane of the crack and perpendicular to the crack front, and (c) tearing – shear stress parallel to the crack plane and parallel to the crack front. Adapted from Shigley et al. [63].

Figure 2.19 Plot of shear strain amplitude versus reversals to failure for torsional fatigue specimens tested at two stress ratios. Adapted from Hurd and Irving [66].
2.3.1 Induction Hardened Shafts

Medium-carbon steel shafts are induction hardened primarily to improve static strength and fatigue strength. Induction hardening increases both the hardness/strength at the surface of the component, which is exposed to the highest stresses, as well as induces a favorable residual stress profile that retards crack nucleation. Residual stress components in both axial and hoop directions in induction hardened shafts are typically compressive at the surface and decrease in magnitude towards the center, eventually transitioning to tensile. The transition from compressive to tensile residual stresses generally occurs in the same region as the microstructural and hardness transitions. The region in which these transitions cross is susceptible to initiating sub-surface cracks and consequently sensitive to internal stress concentrators such as inclusions [72]. Fatigue failures because of sub-surface crack nucleation typically occur in low stress amplitude, high cycle (i.e. long fatigue life) situations. In high stress amplitude situations, the residual stresses at the surface relax and fatigue failure is shifted to the surface [59, 73, 74]. Figure 2.20 shows schematically the relationship between applied shear stress, hardness/strength distribution, and failure origin in an induction hardened medium-carbon steel at a single case depth. Although residual stresses play a critical role in the fatigue performance of induction-hardened components, the diagram in Figure 2.20 provides a reasonable first approximation for use in component design.

![Schematics showing the effect of shear stress amplitude on fracture origin for (a) high stress – low cycles to failure and (b) low stress – high cycles. Adapted from Ochi et al. [74].](image)

Three torsional fatigue studies of induction hardened medium-carbon steels were found in literature that provided insight for the current study. Hurd [75] studied both smooth and splined 20 mm diameter induction hardened SAE 1050H shafts at case depths of 30, 40, and 70 pct of radius. Hurd found that fatigue failures, for all case depths investigated, were surface initiated at high shear stress amplitudes. In these surface initiated specimens, initiation and early crack growth were observed in Mode III on the order of 1 mm in length and occurred along longitudinal shear planes which were followed by a Mode I crack propagation to failure. The longitudinal crack growth is believed to be a result of linking micro-cracks initiated at inclusions in the crack plane. At low shear stress amplitudes, Hurd observed the lowest case depth condition initiated failure sub-surface at the case-core transition.
then propagated in a variety of complex growth modes. Ochi et al. [74, 76] studied three plain-carbon steels (0.35, 0.41, and 0.54 wt pct C) and of each alloy, three case depths were examined ranging from 21 to 68 pct of radius in a 20 mm diameter test specimen. Ochi et al. observed similar results to those found by Hurd in that high shear stress amplitudes resulted in surface initiated failure while low amplitudes initiated failure sub-surface. However, Ochi et al. observed a shift in initiation from surface to sub-surface that was not only a function of case depth but also a function of carbon content. Figure 2.21 shows fracture origin and how it relates to case depth for the 0.41 and 0.54 wt pc C steels investigated by Ochi et al. As case depth and shear stress amplitude decreased for a given carbon content, failure initiation shifted from surface to sub-surface. These results are in agreement with the observations made by Hurd. Cryderman et al. [73] looked at the influence of continuous cast section size on the torsional fatigue performance of induction hardened SAE 1050. Four slight variations in carbon content were tested (0.52 to 0.56 wt pct C) at a nominal case depth of 34 pct of radius in a 12.8 mm diameter test specimen. Although negligible differences between the torsional fatigue performances of the steels were observed, differences in initiation mode were noted. At high strains, or high shear stress amplitudes, initiation occurred in Mode III along longitudinal shear planes, while at low strains, or low shear stress amplitudes, initiation occurred in Mode I due to normal stresses.

Table 2.2 shows a summary of the carbon contents, effective case depths, and torsional fatigue testing parameters examined in the three studies.

![Figure 2.21](image.png)

(a) Plot of fracture origin as a function of shear stress amplitude and normalized case depth and for two carbon levels, (a) 0.41 wt pct C and (b) 0.54 wt pct C. Adapted from Ochi et al. [74].

<table>
<thead>
<tr>
<th>Reference</th>
<th>Carbon Content (wt pct)</th>
<th>Effective Case Depth (mm)</th>
<th>Normalized Effective Case Depth, t/r</th>
<th>Control Type</th>
<th>Stress Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hurd [75]</td>
<td>0.50</td>
<td>3 / 4 / 7</td>
<td>0.3 / 0.4 / 0.7</td>
<td>Stress</td>
<td>NR</td>
</tr>
<tr>
<td></td>
<td>0.35</td>
<td>2.1 / 3.4 / 4.8</td>
<td>0.21 / 0.34 / 0.48</td>
<td>Stress</td>
<td>NR</td>
</tr>
<tr>
<td>Ochi et al. [74, 76]</td>
<td>0.41</td>
<td>2.5 / 4.2 / 6.2</td>
<td>0.25 / 0.42 / 0.62</td>
<td>Stress</td>
<td>NR</td>
</tr>
<tr>
<td></td>
<td>0.54</td>
<td>3.7 / 5.2 / 6.8</td>
<td>0.37 / 0.52 / 0.68</td>
<td>Stress</td>
<td>NR</td>
</tr>
<tr>
<td>Cryderman et al. [73]</td>
<td>0.52-0.54</td>
<td>4.2-4.5</td>
<td>0.33-0.35</td>
<td>Strain</td>
<td>-1</td>
</tr>
</tbody>
</table>
Experimental materials and testing methods were designed in close consultation with industrial mentors, ensuring the goals of this study were met while maintaining a strong industrial relevance. Figure 3.1 shows a simplified flowchart outlining the research. Two medium-carbon steels, one with vanadium and one without, were machined into torsional fatigue specimens designed specifically for this study. Specimens were induction hardened using a variety of different processing routes at Inductoheat® Inc. and fatigue tested at CSM. Fatigue failures were characterized to identify failure initiation location as well as fracture mode. Fluxtrol® Inc. utilized electro-thermal modeling along with the induction hardening recipes from Inductoheat® Inc., torsional fatigue specimen geometry, and failure initiation data from the fatigue testing to produce simulated thermal profiles for the maximum shear stress region of the induction hardened torsional fatigue specimens. The simulated thermal profiles were subsequently used to conduct physical simulations on the Gleeble®; after which, specimens were characterized using a variety of techniques. Supplemental Gleeble® studies were also conducted to provide additional insight.

Figure 3.1  Graphical overview of experimental plan. Items outlined indicate sections completed with the support of companies external to the ASPPRC.

3.1  Experimental Materials

Two medium-carbon steels were obtained for this research via an industrial sponsor, one with aluminum additions (1045) and one with vanadium additions (10V45). Table 3.3 shows the chemical composition for both alloys. Subtle differences in chemistry, other than aluminum, vanadium, and nitrogen, are observed between the 1045 and 10V45, with the 10V45 having slightly higher carbon, manganese, and silicon levels. Additional testing was conducted on a random sample of each steel to assess the difference between the carbon content of the two steels with greater certainty. The carbon content of each steel in Table 3.3 indicates plus or minus one standard deviation of five random tests for carbon content of each material. Calculated hardenability of the 1045 and 10V45 steels differ as a result of the accumulated differences in chemical composition. Calculated ideal diameter (DI) for
the 10V45 steel is 45.5 mm, 21 pct higher than the 1045 steel [25]. In addition, sulfur levels are much lower than most commercially produced heats of 1045/10V45 to reduce crack nucleation at inclusions during fatigue testing. Both alloys were continuous cast as separate industrial heats and hot-rolled to a final diameter of 39.7 mm (1.563 in), resulting in a reduction ratio of 18.8 to 1. Finishing temperatures were consistent with typical hot-rolling practices (∼1000 °C).

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Al</th>
<th>N</th>
<th>S</th>
<th>P</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>1045</td>
<td>0.45±0.01</td>
<td>0.74</td>
<td>0.23</td>
<td>0.10</td>
<td>0.12</td>
<td>0.03</td>
<td>0.002</td>
<td>0.016</td>
<td>0.0068</td>
<td>0.006</td>
<td>0.010</td>
<td>0.26</td>
</tr>
<tr>
<td>10V45</td>
<td>0.47±0.01</td>
<td>0.82</td>
<td>0.28</td>
<td>0.09</td>
<td>0.12</td>
<td>0.03</td>
<td>0.080</td>
<td>0.007</td>
<td>0.0100</td>
<td>0.009</td>
<td>0.007</td>
<td>0.22</td>
</tr>
</tbody>
</table>

Understanding the solubility of microalloy carbides and nitrides are critically important in this research. Figure 3.2 graphically shows the solubility products for VN and AlN along with the compositions for the two steels. An arrow and dashed line indicates the chemical composition evolution of each alloy under equilibrium precipitation conditions in austenite. The 1045 is slightly hyperstoichiometric with aluminum, when considering AlN precipitation. The 10V45 is hyperstoichiometric with vanadium when considering VN precipitation.

![Figure 3.2](image_url) Solubility products of (a) VN and (b) AlN in austenite. Compositions of the experimental materials are indicated with arrows point in the direction of composition change of the material during independent equilibrium precipitation of each compound. Solubility products from Turkdogan [14].

### 3.2 Torsional Fatigue

Torsional fatigue tests were conducted using a universal fatigue tester at CSM. Test fixturing was procured and adapted for this study. Considerable care was given during specimen design as well as processing design to ensure consistent and repeatable results were achieved.
3.2.1 Universal Fatigue Testing Machine

All fatigue tests were conducted on a single Satec SF-1U universal fatigue tester (SN: 1202) at a frequency of 30 Hz. The SF-1U is a load amplitude controlled test frame with a maximum alternating load (dynamic load) of 4450 N, maximum preload (static load) of 4450 N, and maximum displacement amplitude of 12.7 mm. Calibration of both the dynamic and static load were conducted using a strain gage rosette installed on a torsional fatigue specimen. Strain gage data were acquired on a Vishay System 6000 digital acquisition system. Figure 3.3 shows the loading assembly of the SF-1U universal fatigue tester. An electric motor rotates an eccentric mass, which is adjusted to apply the desired load amplitude. Compensator springs oppose the inertial forces of the oscillating frame and ensure only the centrifugal force is applied to the test piece, as long as the angular velocity of the eccentric mass is constant. Flexure plates ensure the oscillator frame only moves vertically and are prone to fatigue failure if testing at high displacements. Tuning weights are adjusted during calibration to ensure the dynamic load is accurate and were adjusted to the theoretical shear stress calculation of the torsional fatigue specimen used in this research. Specimens are preloaded to adjust the stress ratio using an electric motor on the base of the loading assembly and a dial gage measuring the displacement of the oscillator frame relative to the rigid machine frame. The fatigue tester is turned off automatically by activating the travel limit switches after a specimen failure.

![Loading assembly for a Satec SF-1U universal fatigue tester.](image)

3.2.2 Torsion Fixture

The torsion test fixture was rebuilt prior to use. All fasteners and bearings were replaced. Figure 3.4 shows the torsion fixture mounted to the SF-1U universal fatigue tester. Torque is applied to the specimen through the reciprocating platen that is attached to the oscillating end of the fixture through a lever arm. Two lever arms were provided with the fixture, 158.75 and 387.35 mm. The shorter lever arm was used in all the testing for this study. Specimens are held in place between the oscillating and fixed ends of the fixture using double taper XZ collets from Universal Engineering. Each collar around the collets is fastened to the fixture using eight bolts that provide clamping force to the specimen. Dow Corning Molykote® Z Moly-powder was applied between the collets and the
collars to prevent galling. Bearings were rotated and lubricated frequently to prolong life. Anti-size lubricant was applied to all fasteners. All fasteners were torqued to manufacture’s specifications.

Figure 3.4  Torsion fixture for Satec SF-1U universal fatigue tester.

Two modifications were made to the fixture to ensure consistent test-to-test specimen placement: a specimen depth stop and a specimen span stop. The specimen depth stop is a precision ground rod with a hardened tip inserted in to the oscillating end of the fixture that limits how far the specimen can be inserted into the oscillating end collet. The rod is pinned to a bushing that allows length of the stop to be adjusted and fixed using a locking nut. Once the length of the stop is adjusted appropriately, the pin is used to remove the rod from the fixture before testing. The span stop is a plate placed between the oscillating and fixed end of the fixture that limits how far the specimen is inserted into the fixed end collet. The plate is rigidly mounted to the machine and is machined to a specific length determined by the specimen geometry.

3.2.3 Specimen Design

The specimen was designed to be induction hardened without difficulty while also being able to fail at stress levels and displacement amplitudes within the designed limits of the SF-1U universal fatigue tester. Figure 3.5 shows the specimen design that met the aforementioned criteria. The smooth test specimen has a shallow U-shaped groove (or hourglass) in the middle that increases the shear stress at the center, through reduction of diameter, while maintaining a stress concentration factor of less than 0.5 pct. In addition, the shallow U-shaped groove provides a region in the center of the specimen approximately 15 mm wide that is within 95 pct of the maximum shear stress. The transition from the reduced section to the grip ends was kept small to facilitate both scan and single-shot induction hardening. In addition, the ends were center drilled for ease of induction hardening and chamfered for ease of inserting into the test fixture. Appendix A provides detail regarding the specimen design criteria, design validation, and uncertainty analysis of applied shear stress.
3.2.4 Specimen Metrology Prior to Induction Hardening

Careful consideration was given to the geometrical tolerances of the as-machined torsional fatigue specimens to ensure consistent induction hardening as well as specimen alignment during fatigue testing. A random sample of specimens was checked for conformance to geometrical tolerances specified in the technical drawing. Minimum diameter, maximum diameter, total runout, length of reduced section, and surface roughness of reduced section were checked. Although all measured specimens were within tolerances, and as a result, tabulated values of measurements are not presented, the methodology is presented in detail. Table 3.4 shows a summary of the instruments used in dimensional measurements along with each instrument’s resolution. Standard good practices were used in measurement of maximum diameter, minimum diameter, and length of reduced section. Total runout was measured in a lathe in which the specimens were held between centers. Figure 3.6 shows the setup used to measure total runout. An electronic indicator was zeroed at three different locations along the length of the specimen – 12.7 mm from both ends and at the middle – and slowly rotated by hand with the lathe gearbox in neutral. Total indicator reading was recorded at each location and the orientations of the maximum as well as minimum readings were marked on the specimen. Alignment of the maximum and/or minimum indicator readings at the three locations is an indication of the centers being drilled off-axis, which was not observed in the random sample.

<table>
<thead>
<tr>
<th>Dimension</th>
<th>Instrument</th>
<th>Manufacturer</th>
<th>Model</th>
<th>Resolution</th>
</tr>
</thead>
<tbody>
<tr>
<td>Maximum Diameter</td>
<td>Electronic Micrometer</td>
<td>Starrett®</td>
<td>799</td>
<td>0.001 mm (0.00005 in)</td>
</tr>
<tr>
<td>Minimum Diameter</td>
<td>Electronic Caliper</td>
<td>Starrett®</td>
<td>796</td>
<td>0.01 mm (0.00005 in)</td>
</tr>
<tr>
<td>Reduced Section Length</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Total Runout</td>
<td>Electronic Indicator</td>
<td>Mitutoyo®</td>
<td>IDU25E</td>
<td>0.01 mm (0.00005 in)</td>
</tr>
<tr>
<td>Surface Roughness at</td>
<td>Portable Surface Roughness Gage</td>
<td>Federal®</td>
<td>EAS-2632-W3</td>
<td>0.03 µm (1 µin)</td>
</tr>
<tr>
<td>Minimum Diameter</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Surface roughness was measured using a Federal® PocketSurf® portable surface roughness gage (Model: EAS-2632-W3) using a general-purpose probe which conical diamond stylus with a 0.03 µm resolution. Figure 3.7
shows the setup used to measure specimen surface roughness. As with measurement of total runout, specimens were
held in a lathe between centers and rotated by hand. The surface roughness measurement must be made on stationary
surfaces, therefore; multiple measurements were made on each specimen, rotating the specimen to a different
orientation between each measurement. A scan length of 5 mm was used for each measurement.

Figure 3.6  Photograph showing the setup used in measuring the total runout for a random sample of specimens
prior to induction hardening.

Figure 3.7  Photograph showing the setup used in measuring the surface roughness of the reduced area for a
random sample of specimens prior to induction hardening.
3.2.5 Induction Hardening

Specimens were induction hardened at Inductoheat® Inc. in Auburn Hills, Michigan. Four induction hardening processing routines were developed, three scan hardened and one single-shot hardened. The three scan hardened conditions had progressively deeper nominal effective case depths of 25, 32, and 44 pct of the radius, identified as Low, Med, and High, respectively. The single-shot condition is identified as High-SS since it had the same 44 pct nominal effective case depth as the High condition. Figure 3.8 shows the cross-sections for the three separate processing setups used for the four induction hardened conditions. The processing setup in Figure 3.8a was used for the Low and High conditions and shows a single-turn scan coil with an attached quench ring and flux concentrator. Figure 3.8b shows the configuration used to induction harden the Med condition. The same coil was used for the Med condition as was used for the Low and High conditions, but a different quench ring and spacer were used. The High-SS condition used the process setup in Figure 3.8c where a single coil, with flux concentrator, extends the length of the specimen and a quench plate covers the entire coil and specimen.

Table 3.5 shows the induction hardening processing parameters used for each condition. Frequency and power were constant for the scan-hardened conditions at 196 kHz and 72 kW, respectively. All other processing parameters between the Low and High conditions were identical except for a change in coil scan speed. Parameters for the Med condition were changed significantly from all other scan hardened conditions. The changes in processing were to achieve a shallow sloped case/core transition region. The High-SS condition was processed at a lower frequency and higher power than the scan-hardened conditions. A higher intensity quench was utilized to extract the heat adequately fast since entire specimen is being heated at once versus heating only a small section in scan hardening. The machine recipes used at Inductoheat® Inc. are provide in Appendix B. All specimens were tempered at 177 °C for 90 min within only a few minutes after being induction hardening.

<table>
<thead>
<tr>
<th>Condition</th>
<th>Induction Hardening Method</th>
<th>Frequency (kHz)</th>
<th>Power (kW)</th>
<th>Scan Rate (mm/s)</th>
<th>Polymer Concentration in Quenchant (pct)</th>
<th>Quenchant Flow Rate (L/min)</th>
<th>Alloy</th>
<th>Nominal Normalized Effective Case Depth, t/r</th>
</tr>
</thead>
<tbody>
<tr>
<td>Low</td>
<td>Scan</td>
<td>196</td>
<td>72</td>
<td>22.9</td>
<td>6</td>
<td>75</td>
<td>1045</td>
<td>0.25</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>10V45</td>
<td>0.25</td>
</tr>
<tr>
<td>Med</td>
<td>Scan</td>
<td>196</td>
<td>72</td>
<td>36.8</td>
<td>12</td>
<td>75</td>
<td>1045</td>
<td>0.31</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>10V45</td>
<td>0.32</td>
</tr>
<tr>
<td>High</td>
<td>Scan</td>
<td>196</td>
<td>72</td>
<td>17.3</td>
<td>6</td>
<td>75</td>
<td>1045</td>
<td>0.44</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>10V45</td>
<td>0.45</td>
</tr>
<tr>
<td>High-SS</td>
<td>Single-shot</td>
<td>31</td>
<td>128</td>
<td>N/A</td>
<td>2</td>
<td>173</td>
<td>1045</td>
<td>0.43</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>10V45</td>
<td>0.45</td>
</tr>
</tbody>
</table>

*Condition was preheated at a low power then induction hardened with a quench delay.*
Figure 3.8  Schematics of induction coils used in this study to scan harden conditions (a) Low and High as well as (b) Med and (c) single-shot harden condition High-SS. Specimen was located between the “center” and the “ball end mill” for processing. Specimens that were scan hardened were done so from bottom to top.

3.2.6 Specimen Identification

All specimens were assigned serial numbers, in sequential order, as they were being removed from the induction-hardening machine. Initially, numbers were placed on the samples using paint pen, which remained on the specimens through tempering. After tempering, the serial numbers were stamped into both ends of the specimens by hand using metal stamps and a ballpeen hammer. Table 3.4 shows the serial numbers for each combination of material and induction hardened condition.
Table 3.4 – Serial Numbers for Induction Hardened Specimens

<table>
<thead>
<tr>
<th>Condition</th>
<th>Alloy</th>
<th>Serial Numbers</th>
</tr>
</thead>
<tbody>
<tr>
<td>Low</td>
<td>1045</td>
<td>1-25</td>
</tr>
<tr>
<td></td>
<td>10V45</td>
<td>26-50</td>
</tr>
<tr>
<td>Med</td>
<td>1045</td>
<td>106-130</td>
</tr>
<tr>
<td></td>
<td>10V45</td>
<td>131-160</td>
</tr>
<tr>
<td>High</td>
<td>1045</td>
<td>51-75</td>
</tr>
<tr>
<td></td>
<td>10V45</td>
<td>76-105</td>
</tr>
<tr>
<td>High-SS</td>
<td>1045</td>
<td>161-175</td>
</tr>
<tr>
<td></td>
<td>10V45</td>
<td>176-208</td>
</tr>
</tbody>
</table>

3.2.7 Surface Preparation Post Induction Hardening

Limited surface preparation was conducted on the specimens after induction hardening and before fatigue testing. Surface oxidization was only removed from the specimen grip areas, leaving the reduced section in the as-induction hardened condition for testing. The grip areas were sanded to ensure no slippage in the fixture collets during testing. Specimens were held in a metal lathe between centers, as previously shown in Figure 3.6 and Figure 3.7. The transition from the grip area to the reduced section was masked with blue painter’s tape to ensure only the grip area was sanded. The specimen was rotated at a high speed (>1000 RPM) and all oxide was removed by dry sanding with 120 grit alumina abrasive. After all oxide was removed, the grip areas were dry sanded to a smooth finish using 180, 240, and 320 grit alumina abrasive. Specimens were cleaned using isopropanol and a cloth prior to testing.

3.2.8 Testing Stress

A multiple linear regression model from literature data of torsional fatigue tested induction hardened plain-carbon steels was used to determine appropriate test stresses as a function of cycles to failure, carbon content, and normalized effective case depth. Figure 3.9a shows the 159 data points collected from three resources used to develop the following multiple linear regression model

\[ \text{Shear Stress Amplitude (MPa)} = 1115.379 - 169.815 \times \log_{10} N_f + 467.975 \times C \text{ (wt pct)} + 445.416 \times \frac{t}{r} \]  

(3.1)

where \( N_f \) is the number of cycles to failure between \( 10^2 \) and \( 3 \times 10^6 \) cycles, \( C \) is the carbon content of the steel in wt pct between 0.35 and 0.56 wt pct, and \( t/r \) is the normalized effective case depth between 0.21 and 0.68. All variables were found to be significant to the regression above the 99 pct level. Standardized residuals as well as the predicted output are normally distributed and the adjusted coefficient of determination is 0.8344. Figure 3.9b shows the model graphically with upper and lower 95 pct confidence limits for the 1045 and 10V45 steels. The upper bound was calculated using the highest carbon content (0.48 wt pct) and normalized effective case depth (0.45) while the lower bound uses the lowest carbon content (0.43 wt pct) and normalized effective case depth (0.25). The nominal fatigue specimen induction hardened case depths, nominal carbon contents of the two steels, and targeted
cycles to failure of $10^4$ to $10^6$ cycles were used to determine the shear stress amplitudes used in this study: 550, 600, and 650 MPa.

Figure 3.9  Plot of shear stress amplitude versus cycles to failure for (a) data collected from literature [73, 75, 76] and (b) the fitted multiple linear regression model.

3.2.9  Testing Procedures

In general, specimens were installed in the torsion fixture and tested in the SF-1U universal fatigue tester in accordance with the Sonntag manual [77] and, as a result, these details will not be repeated in this section. The one deviation from the manual was with regard to preload, since the fatigue tests were all run at a positive R-ratio of 0.1. Prior to clamping the specimen into the torsion fixture collets the reciprocating platen was lowered the required amount of displacement for a given shear stress amplitude. Correlation between reciprocating platen displacement and load was determined from calibration with a strain-gaged specimen. The specimen was then clamped in the collets and the reciprocating platen was moved up to zero displacement applying the preload. This procedure ensures the platen always oscillates around zero. If the platen oscillates around a slightly positive or negative displacement then additional forces from the flexure plates are introduced.

3.3  Gleeble® Physical Simulations

Three studies were conducted utilizing the Gleeble® 3500 at CSM: a selected microstructure study, a continuous heating transformation study, and a quench rate study. These studies were developed to investigate the influence of vanadium microalloying on microstructure, hardness, and transformation behavior. The continuous heating transformation study takes an idealized approach for the induction hardening process. Specimens were heated at a range of heating rates relevant to induction hardening processes to a single temperature and gas quenched. The selected microstructure study specifically examines the induction-hardened case and total case depth locations in select induction hardened conditions in the torsional fatigue specimens. This processing was done with
the aid of electro-thermal modeling. The quench rate study examines a single thermal profile from the selected microstructure study, the modeled surface thermal profile the low case depth condition, at a range of cooling rates.

### 3.3.1 Specimen and Fixture Design

The ISO-Q® specimen design was selected for this study due to its versatility in achievable heating and cooling rates while allowing dilatometry data to be collected. Figure 3.10 shows a technical drawing of the ISO-Q® specimen. A solid center section is held in the Gleeble® low-force fixturing using copper hot-grips that are designed to reduce physical contact with the specimen and flatten thermal gradients, to be discussed in detail later in this chapter. Nozzles are inserted into the tubular ends to quench the specimen internally using any fluid (e.g. nitrogen gas, water, etc.). Because the quenchant is contained within the specimen, dilatometric measurements are not disturbed during quenching and vacuum is maintained for the duration of the entire test.

![Figure 3.10 Technical drawing of ISO-Q® specimen used in Gleeble® physical simulations.](image)

### 3.3.2 Specimen Metrology Prior to Processing

Specific geometric features of each ISO-Q® specimen were scrutinized before being processed in the Gleeble®. Specimens were first ultrasonically cleaned in methanol to remove all oils and ensure the tubular ends are free of machining chips. The reduced section diameter and length as well as the depth of the tubular ends of each specimen were measured to ensure adherence with specified tolerances using an electronic caliper with a 0.013 mm resolution. All of these dimensions are important for heating and cooling rate consistency as well as needed for dilatometric measurements.

### 3.3.3 Thermal Gradients

Thermal gradients occur in all Gleeble® specimens; however, they can be minimized through careful specimen and grip design as well as proportional-integral-derivative (PID) controller settings. Figure 3.11a shows the surface thermal gradients at different distances from the ISO-Q® specimen center at 1050 °C for heating rates of 100 and 1000 °C/s. At lower heating rates the surface thermal gradient decreases and a larger volume of material is representative of the intended heat treatment. However, at high heating rates the thermal gradients are very sharp, emphasizing the need for great care during specimen analysis. Figure 3.11b shows the deviation of the measured temperature at the center of an ISO-Q® specimen from the programmed thermal cycle for heating rates of 100 and
1000 °C/s. Although the 1000 °C/s heating rate shows significant deviation from the programmed cycle PID settings were selected primarily to minimize overshoot. PID controller settings were iteratively determined for heating rates from 100 to 1000 °C/s to be P=2.0 and I=1.0.

![Graph](image1)

Figure 3.11  Quantification of (a) lengthwise thermal gradients in an ISO-Q® specimen as a function of heating rate at 1050 °C and (b) temperature difference of specimen center from the programmed thermal profile at 100 to 1000 °C/s.

### 3.3.4 Continuous Heating Transformation Study

The Gleeble® 3500 was used to heat ISO-Q® specimens to 1050 °C at constant heating rates of 100, 250, 500, and 1000 °C/s. A hold time of 0.3 s at 1050 °C was used for all heating rates before quenching to room temperature using an internal 550 kPa nitrogen gas quench. One additional condition was tested at the 100 °C/s heating rate with a 3.0 s hold at 1050 °C. Figure 3.12 shows the recorded quench for one of the tests overlaid on calculated CCT diagrams for the 1045 and 10V45 steels. Cooling rate indicates the expected microstructure of homogenous 1045 and 10V45 steels to be martensitic.

![Graph](image2)

Figure 3.12  Calculated continuous cooling transformation (CCT) curves for the 1045 and 10V45 materials used in this research [78]. The cooling curve for a processed ISO-Q® specimen is overlaid.
3.3.5 Selected Microstructure Study

Specific locations within the induction hardened torsional fatigue specimens were modeled and physically simulated using ISO-Q® specimens in the Gleeble®. Table 3.7 shows the locations in the induction hardened torsional fatigue specimens for each condition modeled as well as the software package used. The surface and total case depth locations for the Low, High, and High-SS conditions were modeled by Fluxtrol® Inc. using one- and two-dimensional electro-thermal modeling software packages, ELTA and Flux 2D respectively. ELTA was used in the analysis of the High-SS condition while Flux 2D was used for the Low and High conditions. The Med condition was not simulated due to the complexity of the processing as well as anticipated usefulness of the data. The induction hardening parameters, coil design, and specimen geometry were all incorporated into the modeling. Both software packages use the finite element method for solving coupled electromagnetic and thermal phenomena.

<table>
<thead>
<tr>
<th>Condition</th>
<th>Software</th>
<th>Location Description</th>
<th>Distance from Surface (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Low</td>
<td>Flux 2D</td>
<td>Surface</td>
<td>0.00</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Total Case Depth</td>
<td>2.10</td>
</tr>
<tr>
<td>High</td>
<td>Flux 2D</td>
<td>Surface</td>
<td>0.00</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Total Case Depth</td>
<td>3.80</td>
</tr>
<tr>
<td>High-SS</td>
<td>ELTA</td>
<td>Surface</td>
<td>0.00</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Total Case Depth</td>
<td>3.80</td>
</tr>
</tbody>
</table>

Figure 3.13 shows the surface and total case thermal profiles from the modelling for the Low, High, and High-SS conditions. Surface profiles show the heating rate is very similar between the conditions, but peak temperature and cooling behavior are different. The High condition spends a significantly longer amount of time in the austenite phase field than any other condition. Total case profiles show that the three conditions have nearly identical peak temperatures but time at peak temperature and cooling behavior are different.

![Figure 3.13](image)

Figure 3.13 Simulated thermal profiles for (a) the surface and (b) the total case depth for the Low, High, and High-SS induction hardened conditions.
3.3.6 Quench Rate Study

The modeled thermal profile for the surface of the Low induction hardening condition was used to conduct a series of quench rate experiments. Figure 3.14 shows the control thermocouple data for all the conditions in the quench rate study. Specimens were heated following the temperature profile from the modelling results with no hold or a 30 s hold at the peak temperature and quenched at 350, 1000, and 3000 °C/s to room temperature. The 350 °C/s cooling rate was achieved using an 550 kPa water quench of the standard ISO-Q® specimen. The faster cooling rates were achieved using the same quenching media and pressure with modified specimens. ISO-Q® specimens were made tubular by drilling 2 and 4 mm holes, axially through the specimen center, allowing heat to be extracted more quickly.

![Figure 3.14](image)

Figure 3.14 Control thermocouple data for the Gleeble® physical simulations conducted during the quench rate study.

3.4 Characterization

Various methods were used to characterize the as-received material, torsional fatigue specimens, and Gleeble® physical simulations. This section describes the methodology and procedures for all quantitative metallography, macro-photography, microscopy, and mechanical testing conducted during this study.

3.4.1 Metallographic Specimen Preparation

Specimens were sectioned using a water-cooled high-speed abrasive saw. Torsional fatigue specimens were sectioned to provide a number of dedicated specimens for characterization utilizing a variety of methods. Figure 3.15 shows the sectioning plan for an untested torsional fatigue specimen. Specimens A and B were used for all microscopy methods at the maximum and 95 pct of maximum shear stress, respectively. Specimen C was used strictly for determining the prior austenite grain size of the induction hardened case region. Specimens E and F were used for specimens that were macroetched for total case depth measurement. Gleeble® specimens were sectioned adjacent to the control thermocouple. After mounting, the specimens were ground until the thermocouple weldments were exposed prior to polishing. Specimens that were mounted were done so in a mineral filled thermosetting epoxy
powder using a hot press with a 32 mm diameter mold. The mounting press cures the epoxy by applying a pressure of 27.6 MPa at a temperature of 138-160°C for 10-15 min. All specimens were polished to a 0.05 µm diamond finish prior to etching for either light or scanning electron microscopy. Table 3.8 provides a detailed polishing procedure for all metallographic specimens. Table 3.9 shows the etchants and associated procedures used in this research. In general, 2 pct nital was used for all light and scanning electron microscopy. Macro-etched specimens were etched with 2 pct nital until heavy contrast was achieved then immersed in 4 pct hydrochloric acid in methanol for 4-8 s to remove staining left by the nital etchant then rinsed with methanol. The 4 pct picral etchant was used on specific specimens with varied success to reveal carbides. The prior austenite grain size etchant worked very well, but lost effectiveness after 2-4 specimens.

A Metallographic Specimen
B Metallographic Specimen
C Metallographic Specimen
D Not Used
E Macroetch Specimen
F Macroetch Specimen
G Not Used

Figure 3.15 Sectioning plan for untested torsional fatigue specimen.

3.4.2 Macro-photography

A Nikon D70 digital single-lens reflex (DSLR) camera with an AF Micro-Nikkor 60 mm f/2.8D macro lens and a SB-700 AF speed light flash was used to take all photographs of macro-etched cross-sections as well as fracture surfaces. Optimal camera settings were found to be a shutter speed of 1/60 s, focal ratio of f/30, and light sensitivity of ISO-200 with a flash setting of 1/32. Clay and magnets were often used to hold specimens in position for photography while the camera was rigidly mounted to a tripod.

3.4.3 Light Microscopy

Standard bright-field light microscopy was conducted utilizing an Olympus PMG-3 inverted microscope with a 3 MP digital camera. Dark-field was only used as a method to determine the quality of polish on critical specimens. Magnifications from 50 to 500 times were utilized. Light intensity and apertures were first adjusted at 500X magnification and neutral density filters were used to adjust the light intensity for lower magnifications. Oblique lighting was used in specific circumstances to accentuate grain boundary etching (i.e. prior austenite grain size etching).
### Table 3.6 – Polishing Procedure for Metallographic Specimens

<table>
<thead>
<tr>
<th>Step</th>
<th>Disc</th>
<th>Abrasive &amp; Fluid</th>
<th>Time (mm:ss)</th>
<th>Head Force (N)</th>
<th>Head/Wheel Speed (RPM)</th>
<th>Head/Wheel Direction</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Platinum 2 CAMEO® (LECO®)</td>
<td>Water</td>
<td>Until Plane</td>
<td>110</td>
<td>75/150</td>
<td>CW/CCW</td>
</tr>
<tr>
<td>2</td>
<td>Platinum 4 CAMEO® (LECO®)</td>
<td>Water</td>
<td>5:00</td>
<td>110</td>
<td>75/150</td>
<td>CW/CCW</td>
</tr>
<tr>
<td>3</td>
<td>Ultra Silk (LECO®)</td>
<td>9 µm Diamond / Ultralap® Extender (LECO®)</td>
<td>3:00</td>
<td>150</td>
<td>100/200</td>
<td>CW/CCW</td>
</tr>
<tr>
<td>4</td>
<td>Ultra Silk (LECO®)</td>
<td>3 µm Diamond/ Ultralap® Extender (LECO®)</td>
<td>3:00</td>
<td>150</td>
<td>100/200</td>
<td>CW/CCW</td>
</tr>
<tr>
<td>5</td>
<td>Lecloth® (LECO®)</td>
<td>1 µm Diamond/ Ultralap® Extender (LECO®)</td>
<td>3:00</td>
<td>70</td>
<td>100/200</td>
<td>CW/CCW</td>
</tr>
<tr>
<td>6</td>
<td>Imperial (LECO®)</td>
<td>0.05 µm Diamond/ Ultralap® Extender (LECO®)</td>
<td>3:00</td>
<td>50</td>
<td>150/250</td>
<td>CW/CW</td>
</tr>
</tbody>
</table>

### Table 3.7 – Etchants and Etching Procedures

<table>
<thead>
<tr>
<th>Etchant</th>
<th>Procedure</th>
</tr>
</thead>
<tbody>
<tr>
<td>2 pct Nital</td>
<td>1.) Make new 2 pct nital solution (68 pct assay nitric acid in methanol).&lt;br&gt;2.) Swab specimen with saturated cotton ball for 5 s.&lt;br&gt;3.) Rinse with methanol.&lt;br&gt;4.) Dry with clean compressed nitrogen.&lt;br&gt;5.) Inspect with light microscope.&lt;br&gt;6.) Repeat steps 2-5 until desired contrast is achieved.</td>
</tr>
<tr>
<td>4 pct Picral</td>
<td>1.) Make new 4 pct picral solution (4 g picric acid in 100 ml methanol).&lt;br&gt;2.) Immerse specimen for 4-8 s.&lt;br&gt;3.) Rinse with methanol.&lt;br&gt;4.) Dry with clean compressed nitrogen.&lt;br&gt;5.) Inspect with light microscope.&lt;br&gt;6.) Repeat steps 2-5 until desired contrast is achieved.</td>
</tr>
<tr>
<td>Prior Austenite Grain Size Etchant</td>
<td>1.) Make new solution consisting of 5.5 g of picric acid, 8.4 ml of Teepol, and 280 ml of deionized water (add 2 ml hydrochloric acid to solution just before etching).&lt;br&gt;2.) Heat to 65 °C (150 °F) on hot plate while magnetically stirring.&lt;br&gt;3.) Preheat specimen on hot plate.&lt;br&gt;4.) Immerse specimen with tongs in solution for 15 s.&lt;br&gt;5.) Rinse thoroughly with deionized water then with methanol.&lt;br&gt;6.) Dry with clean compressed nitrogen.&lt;br&gt;7.) Back-polish by hand with 1 µm diamond on LECO® Lecloth® pad as needed to remove corrosion product martensite etching.&lt;br&gt;8.) Rinse with methanol.&lt;br&gt;9.) Inspect with light microscope.&lt;br&gt;10.) Repeat steps 3-9 until desired contrast is achieved.</td>
</tr>
</tbody>
</table>
3.4.4 Scanning Electron Microscopy

Both characterization of microstructure and fractography were conducted using a field emission scanning electron microscopy (FESEM). Only secondary electron images (SEI) were collected. Microstructure was characterized at 5 kV, a working distance of 10 mm, and a probe current of three in the fine range. When the FESEM is well aligned, these settings work well for magnifications up to 50,000 times. Fractography was conducted at 15 kV, a working distance of 25 mm, and a probe current of nine on the coarse setting and middle on the fine scale. These settings provide excellent imaging up to 5000 times magnification of specimens with a large amount of topography. Fracture surfaces were ultrasonically cleaning in methanol for 10 minutes, dried with clean compressed nitrogen, and stored in a desiccator until examination. All specimens were grounded to the FESEM specimen holder using either double-sided carbon tape or single-sided copper tape.

3.4.5 Transmission Electron Microscopy

Characterization via transmission electron microscopy (TEM) was limited to the as-received 10V45 material to characterize the V(C,N) precipitation. A bulk specimen of 10V45, near the center of the as-received bar, was sectioned to a thickness of 300-400 µm using a low-speed saw with a high concentration cubic boron nitride (CBN) blade then bonded to an aluminum stub using double-sided tape. The bulk specimen was dry ground by hand to a uniform thickness of 80-90 µm using 600 grit silicon carbide and punched into several 3 mm discs using an SPI® Supplies Disc Punch (Item No. 17001-AB). Punched foils were deburred using 800 grit silicon carbide and stored in a TEM grid holder until jet polishing. Foils were polished with 1200 grit silicon carbide and cleaned with methanol directly prior to jet polishing. Jet polishing was done using a solution of 30 pct nitric acid (68 pct assay) in methanol, cooled to -30 to -34 °C using liquid nitrogen, and electro-polished using 30 mA under current control. Flow rate for the two jets was adjusted so they slightly impinge on one another without the specimen present. Polishing time was typically between 270 and 290 s.

The process to highlight the V(C,N) precipitates took several steps. The Baker-Nutting orientation relationship along with diffraction patterns from literature were used to highlight the precipitates [79]. This method assumes the V(C,N) precipitation occurred in ferrite. First, a ferritic region was tilted to a 011 zone and a diffraction pattern was collected. Figure 3.16 shows an example 011 zone diffraction pattern in ferrite from this study as well as a 011 zone diffraction pattern in ferrite from Edington [79] showing V(C,N) and Fe₃O₄ diffraction. Diffraction from V(C,N) is often very faint and under most occasions not discernible due to precipitate size. This faint pattern emphasizes the need for either simulated of literature diffraction to know the locations of V(C,N) precipitation. The specimen was then tilted along 200 Kikutchi band to a systematic row of diffraction spots that contain the 200 ferrite spot and the 020 V(C,N) spot. A two-beam condition was then created by tilting to make the 200 ferrite spot bright. Figure 3.17 is the resulting two-beam diffraction pattern. An objective aperture was then centered over the direct beam of the diffraction pattern to collect a bright field image of the ferrite. The dark field beam controls were then used to move the 020 V(C,N) spot to the optic axis. An objective aperture was then centered over the 020 V(C,N) spot and optic axis to generate a centered dark field image which highlighted the V(C,N) precipitates. The diameters of the precipitates were measured in ImageJ, an open source image processing and analysis software.
3.4.6 Quantitative Metallography

The quantitative metallography conducted in the current study was limited to constituent/phase area fractions, grain sizes, and pearlite interlamellar spacing. Constituent area fraction was determined from point counting of light micrographs using randomly placed segmented concentric circles. More than 1000 point-counts were conducted per condition. Ferrite grain size in the as-received condition for each material was determined from light micrographs using ImageJ. Individual ferrite grains were outlined to quantify the minimum Feret diameter and circularity. More than 500 grains were measured per condition. Mean lineal intercept length of prior austenite grains was determined via randomly placed concentric circles on light micrographs. More than 500 intercepts were measured per condition. The mean random interlamellar spacing was measured using the same concentric circle method described for measuring the mean lineal intercept length of prior austenite grains, only using FESEM micrographs. True pearlite interlamellar spacing was determined by dividing the mean random interlamellar spacing in half. This method was employed because it was found to be in good agreement with analysis of TEM foils [80].
3.4.7 Hardness Testing

Both macro- and micro-hardness testing were used in this research. Rockwell macro-hardness testing, either B (HRB) or C (HRC) scale, was used to determine the bulk hardness of specimens after machining and thermal processing. Vickers micro-hardness (HV), primarily with a 500 gmf load, was used to quantify the bulk hardness as well as hardness gradients of mounted specimens in the as-polished condition, 1 µm finish if not finer. Verification of instrument calibration before each series of measurements was done with appropriate hardness standard blocks. Indentations were placed inward from the edge of a specimen as well as spaced apart more than three times the indentation width. ASTM standard hardness conversion tables were used to make any necessary conversions between the Rockwell and Vickers hardness scales [81].

All bulk measurements of hardness were reported as the average of multiple tests from replicates, anywhere from 5 to 15 measurements per specimen depending on the specimen size. Hardness profiles were conducted only on transvers cross-sections of the induction hardened torsional fatigue specimens. Hardness measurements were taken in a straight line at regular intervals from the surface to the center of the specimen, always maintaining a minimum spacing of three indentation widths. Figure 3.18 shows a representative photograph of a single hardness profile. Three hardness line profiles were measured per specimen, at 0°, 90°, and 180°. Figure 3.19 shows a schematic of the three hardness line profiles. The first indent of each line profile was inset from the surface at a slightly different distance allowing the three line profiles to be used together to make a single hardness profile per condition without duplicate measurements at a given distance from the surface. The effective case depth was determined by the distance from the surface to a hardness of 45 HRC [82].

Figure 3.18 Photograph of a single Vickers micro-hardness line profile conducted on a transverse cross-section of an induction hardened torsional fatigue specimen.

Figure 3.19 Schematic showing the initial offset of the first indent in the Vickers micro-hardness line profiles.
3.4.8  Uniaxial Tensile Testing

All tensile tests were conducted at room temperature on an MTS® Systems Alliance RT/100 electro-mechanical tensile testing frame with a quasi-static crosshead displacement rate of 0.51 mm/min. Figure 3.20 shows the tensile specimen geometry used in the testing. The ends were held in the tensile testing frame using wedge grips specifically designed to hold round specimens. The initial and final dimensions of each specimen were measured using an electronic caliper with a 0.013 mm resolution. Load was determined using a 100 kN MTS® Systems load cell (SN: 104393). Strain was determined using a 25 mm Shepíc extensometer capable of 50 pct extension. Yield strength was determined either by the 0.2 pct plastic strain offset method or by observation of a distinct yield point. Ultimate tensile strength was determined from the maximum load. Uniform elongation was determined using Considère’s construction. In some instances, the true fracture strain and true fracture strength were determined. True fracture strain is the natural log of initial cross-sectional area divided by the cross-sectional area at fracture. True fracture strength is the load at fracture divided by the cross-sectional area at fracture [44].

![Figure 3.20](image)

Figure 3.20  Technical drawing of tensile specimen used in this research.

3.4.9  Dilatometry

Volume expansion due to phase transformations was measured in the Gleeble® using a quartz push-rod linear variable differential transformer (LVDT) dilatometer. A chisel-point quartz rod was centered on the control thermocouple to measure diametrical changes during heating and cooling. Critical temperatures were determined graphically by the deviation from linearity of the diametrical strain versus temperature plot [83].

3.4.10  Residual Stress Hole Drilling

Residual stresses were determined using an RS-200 Milling Guide from Vishay Micro-Measurements. This technique uses a tool steel, carbide, or diamond end mill to incrementally remove material from the center of a strain gage rosette. As each layer of material is removed, the strain from the relaxing residual stresses is recorded and converted to stress. Figure 3.21 shows the RS-200 at CSM mounted to a custom-built universal clamping base. After a strain gage rosette was installed on a specimen, using standard techniques [84], the specimen is held rigidly to the base using a variety of clamps. Figure 3.22 shows a specimen clamped to the base using a step block, strap clamp, and a pair of V-blocks. Clamping was always done while collecting data from the strain gage rosette to ensure no external forces were applied to the specimen that would affect the residual stress measurement. The milling guide was aligned normal to the specimen using the Z adjustments and locked in place. The microscope tube was inserted into the milling guide and focused manually by changing the distance of the tube to the specimen. Once focused, the
locking collar was installed on to the microscope tube. The X-Y adjustments were then used to align the crosshairs in the microscope tube eyepiece to the center of the strain gage rosette and locked in place. The microscope tube with locking collar was then removed from the RS-200 and replaced with an anti-rotation collar, micrometer head, and air turbine assembly. The micrometer head was locked at zero and the air turbine assembly was lowered until in contact with the strain gage rosette. A locking collar was installed on the air turbine assembly, which allowed depth to be controlled strictly using the micrometer head. The center of the strain gage rosette was then milled away until just touching the specimen using an inverted cone diamond end mill from Brasseler USA (#805.31.016). This position was determined to be the zero point because at all depths below this point residual stresses were relieved. Material was then removed at 0.25 mm (0.001 in) increments, rotating the air turbine by hand at each depth to ensure a uniform hole. Figure 3.23 shows a torsional fatigue specimen after drilling. Replicates of each steel were evaluated, for each induction-hardened condition, to a depth of 1 mm (0.040 in). The RS-200 Milling Guide Instruction Manual provides additional detail regarding procedures [85].

![Diagram of microscope setup](figure3.21.png)

*Figure 3.21  Vishay Micro-Measurements RS-200 residual stress milling guide with universal clamping base developed for the instrument at CSM.*

Two aspects of this technique were found to be critically important: determining the zero point while hole drilling and determining the concentricity of the end mill. Determining an accurate zero-point requires good experimental technique and practice; however, the concentricity of the end mill can be determined relatively easily with a few tools. Figure 3.24 shows the setup and tools used to adjust the concentricity of the end mill. The concentricity of the end mill was adjusted so that a 2 mm (0.080 in) hole was drilled using a 1.6 mm (0.063 in) end mill. The orbiting action created by offsetting the end mill from center reduces cutting forces and ultimately reduces wear on the end mill. The air turbine was placed on a granite flat-surface plate using a matching pair of precision ground V-blocks. A dial test indicator with a 0.025 mm (0.0001 in) resolution was placed on the shank of the diamond cutter inserted into the air turbine. The air turbine was rotated by hand and adjustments were made to the
end mill alignment set screws systematically until the desired offset was achieved. Test holes were drilled in a piece of scrap steel and measured optically to ensure correct offset.

![Figure 3.22 Clamping setup for torsional fatigue specimens.](image1)

![Figure 3.23 Strain gaged and milled torsional fatigue specimen.](image2)

Strain data were analyzed using H-DRILL, a commercial software package. The software allows the user to use a variety of different analysis techniques depending on the application. The residual stress profile was determined to be non-uniform in all of the specimens examined and therefore was analyzed using the integral method. The author of the software provides additional information regarding the analysis technique in the literature [86, 87]. Figure 3.25 shows a representative set of strain data as a function of depth with the corresponding residual stress profile calculated using the integral method in H-DRILL. Strain gage one is the axial direction and gage three is the hoop direction. The implications of the residual stress profile components on torsional fatigue performance were determined by summing the residual and applies stress states.
Figure 3.24  Setup for adjusting the concentricity of the end mill.

Figure 3.25  Strain gage drilling data for a 1045-Low specimen. Plot (a) shows the raw strain data as a function of depth from the surface and (b) shows the raw data converted to axial, hoop, and shear stress components using the H-DRILL software.
CHAPTER 4
RESULTS

This chapter presents the results of the current project. Characterization of the as-received material includes quantification of microstructural features and precipitate size as well as hardness and quasi-static tensile testing. Induction hardened torsional fatigue specimens were characterized at the region of maximum shear stress and 95 pct of maximum shear stress in the transverse direction as well as longitudinally to the sample shoulder. Hardness as a function of depth from the surface was used to quantify peak case hardness, effective case depth, and total case depth along the region of maximum shear stress and 95 pct of maximum shear stress. All cross-sections were macro-etched to determine total case depth visually. The microstructure of the induction-hardened case was qualitatively examined for retained ferrite, “ghost” pearlite, and other non-martensitic transformation products. Residual stresses were quantified for all conditions to a depth of 1 mm from the surface. Torsional fatigue stress-life data are presented at 550, 600, and 650 MPa for each condition as well as macro- and SEM fractography of representative specimens. Lastly, results of microstructure, hardness, and dilatometric data from Gleeble® physical simulations of the induction hardening process are presented.

4.1 As-received Material Characterization

Comparison studies require detailed characterization of the starting condition prior to processing so that the post processing condition can understood. In this section, data are presented showing quantified differences between the two as-received, hot-rolled materials regarding ferrite grain size, ferrite grain morphology, pearlite interlamellar spacing, vanadium carbonitride precipitate size, hardness, and uniaxial tensile properties. Unless stipulated, all data in this section are from the bar mid-radius taken transverse to the rolling direction.

4.1.1 Ferrite Fraction and Grain Size

Ferrite grain characteristics were quantified using the public domain image-processing program ImageJ. Over 1000 point counts from multiple images were used to determine the area fraction of ferrite in both steels. In determining the ferrite grain size and circularity, multiple images were analyzed per condition, which resulted in a total of 352 grains being measured in the 1045 material and 388 grains in the 10V45 material. Overall, the 1045 steel was found to have a smaller ferrite fraction, larger ferrite grain size, and more elongated ferrite grains than the 10V45. Figure 4.1 shows SEM secondary electron images of both the 1045 and 10V45 steels. In both steels, the ferrite grains outline prior-austenite grain boundaries, with the 10V45 exhibiting some intergranular ferrite. Although the steels have a very low sulfur content, intergranular ferrite was observed to have predominantly nucleated from inclusions. The 10V45 steel was found to have a lower fraction of ferrite than the 1045, 12.0-15.0 pct for the 10V45 and 14.9-18.9 pct for the 1045.

Figure 4.2 shows both the histogram and cumulative fraction plot for the ferrite grain minimum feret diameter for the hot rolled materials. Minimum feret diameter is the shortest dimension of the ferrite grain and is used here because it represents twice the minimum distance carbon would have to diffuse to transform the ferrite grain to austenite via carbon diffusion control. A schematic is inset in Figure 4.2a describing minimum feret
diameter. Both steels have ferrite grain minimum feret diameter distributions which are approximately lognormal, with the 10V45 exhibiting a much finer and narrower distribution than the 1045. Figure 4.3 shows the histogram and cumulative fraction plot of ferrite grain circularity. Both steels exhibit skewed distributions that range from 0.5 to 1.0, or highly elongated to completely circular. A schematic is inset in Figure 4.3b shows ideal geometries exhibiting circularity of 0.5 and 1.0. Ferrite grains in the 10V45 steel have a higher circularity than in the 1045. Table 4.10 shows the tabulated results of ferrite fraction, ferrite grain size, and ferrite grain circularity.

Figure 4.1  Representative SEM secondary electron images for (a) 1045 and (b) 10V45 steels taken from the hot-rolled bar mid-radius. Images show distinct differences in both ferrite grain size and area fraction. Micrographs taken at 1000 times magnification with a 2 pct nital etch.

Figure 4.2  Ferrite grain minimum feret diameter (a) histogram and (b) cumulative fraction plots for the 1045 and 10V45 steels in the hot-rolled condition.
Figure 4.3  Ferrite grain circularity (a) histogram and (b) cumulative fraction plots for the 1045 and the 10V45 steels in the hot-rolled condition.

Table 4.1 – Summary of Ferrite Fraction, Grain Size, and Circularity

<table>
<thead>
<tr>
<th>Characteristic</th>
<th>1045</th>
<th>10V45</th>
</tr>
</thead>
<tbody>
<tr>
<td>Area Fraction (pct)</td>
<td>14.9 – 18.9</td>
<td>12.0 – 15.0</td>
</tr>
<tr>
<td>Minimum Feret Diameter (µm)</td>
<td>5.0 – 5.7</td>
<td>2.7 – 3.1</td>
</tr>
<tr>
<td>Circularity</td>
<td>0.76 – 0.79</td>
<td>0.83 – 0.85</td>
</tr>
</tbody>
</table>

Data presented as 95 pct confidence limits of the sample distribution mean.

4.1.2  Pearlite Interlamellar Spacing

True pearlite interlamellar spacing was estimated for both steels in the transverse direction by analyzing the mean random intercept method using images from multiple specimens of each steel. Figure 4.4 shows representative SEM secondary electron images of the 1045 and 10V45 steels. Pearlite regions of both steels consistently show both continuous and interrupted cementite lamella. Table 4.11 shows the pearlite area fraction and true mean interlamellar spacing of the two steels. Pearlite area fraction is simply the remainder of the microstructure not quantified as ferrite and is presented for completion only. The 10V45 steel showed slightly finer interlamellar spacing than the 1045.

Table 4.2 – Summary of Pearlite Fraction and Mean True Interlamellar Spacing

<table>
<thead>
<tr>
<th>Characteristic</th>
<th>1045</th>
<th>10V45</th>
</tr>
</thead>
<tbody>
<tr>
<td>Area Fraction (pct)</td>
<td>81.1 – 85.1</td>
<td>85.0 – 88.0</td>
</tr>
<tr>
<td>True Mean Interlamellar Spacing (µm)</td>
<td>0.216 – 0.229</td>
<td>0.198 – 0.212</td>
</tr>
</tbody>
</table>

Data presented as 95 pct confidence limits of the sample distribution mean.
4.1.3 Vanadium Carbonitride Precipitation

Vanadium carbonitride precipitation in the 10V45 steel was quantified using centered dark field TEM imaging. Figure 4.5 and 4.6 show bright field / dark field image pairs highlighting V(C,N) precipitation in both proeutectoid ferrite and pearlitic ferrite, respectively. Approximately 100 V(C,N) precipitates were measured in each region. The 95 pct confidence limits for the V(C,N) mean diameter is 4.2-4.8 nm in the proeutectoid ferrite and 3.0-3.6 nm in the pearlitic ferrite. The intent of this analysis was not to develop precipitate size distributions for each region, but to estimate of the average size. In general, the V(C,N) precipitates appear to be randomly distributed; however, Figure 4.7 shows possible evidence of aligned precipitates in pearlitic ferrite.

Figure 4.5 TEM micrographs showing (a) bright field and (b) dark field images of proeutectoid ferrite near the bar center in the as-received, hot-rolled 10V45 steel. V(C,N) precipitates are seen as white spots in the centered dark field image. Micrographs were taken at 125,000 times magnification.
Figure 4.6 TEM micrographs showing (a) bright field and (b) dark field images of pearlite near the bar center in the as-received, hot-rolled 10V45 steel. V(C,N) precipitates are seen in the pearlitic ferrite as white spots in the centered dark field image. Micrographs were taken at 125,000 times magnification.

Figure 4.7 Dark field TEM micrograph highlighting V(C,N) precipitates in the pearlitic ferrite indicating possible alignment. Micrographs was taken at 125,000 times magnification.

4.1.4 Mechanical Properties

Hardness and uniaxial tensile properties were quantified for both steels. Figure 4.8 shows representative engineering stress-strain curves for both 1045 and 10V45. The criteria for selecting representative engineering stress-strain curves was the location at which the tensile specimen failed within the extensometer. The 10V45 steel showed significantly higher yield strength, higher ultimate tensile strength, and lower total elongation than the 1045 steel. Table 4.12 shows a summary of the hardness and tensile testing results along with the percent difference in the properties relative to the 10V45 steel. The 10V45 steel exhibits approximately a 23 pct higher hardness, 33 pct
higher yield strength, and 20 pct higher tensile strength than the 1045 steel while also showing a 32 pct lower uniform elongation, 33 pct lower total elongation, and 10 pct lower reduction in area.

![Figure 4.8 Representative engineering stress-strain curves for as-received, hot-rolled 1045 and 10V45 steels.](image)

Table 4.3 – Mechanical Properties of Hot Rolled 1045 and 10V45 Steels

<table>
<thead>
<tr>
<th>Property</th>
<th>1045</th>
<th>10V45</th>
<th>Difference (pct)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Vickers Microhardness (HV₅₀) ²</td>
<td>216.9 ± 4.9</td>
<td>280.7 ± 8.7</td>
<td>+22.7</td>
</tr>
<tr>
<td>Yield Strength (MPa)</td>
<td>375 ± 6.6</td>
<td>562 ± 2.6</td>
<td>+33.3</td>
</tr>
<tr>
<td>Ultimate Tensile Strength (MPa)</td>
<td>714 ± 1.7</td>
<td>887 ± 2.2</td>
<td>+19.5</td>
</tr>
<tr>
<td>Uniform Elongation (pct)</td>
<td>14.3 ± 0.1</td>
<td>10.8 ± 0.1</td>
<td>-32.4</td>
</tr>
<tr>
<td>Total Elongation (pct)</td>
<td>27.3 ± 0.2</td>
<td>20.5 ± 0.1</td>
<td>-33.2</td>
</tr>
<tr>
<td>Reduction in Area (pct)</td>
<td>46.3 ± 0.2</td>
<td>42.3 ± 0.4</td>
<td>-9.5</td>
</tr>
</tbody>
</table>

² Uncertainty given as 95 pct confidence interval of the mean.

4.2 Induction Hardened Specimen Characterization

Microstructural gradients in induction-hardened components can be challenging to characterize given the thermal gradients induced during the heat treatment. Heating rates, peak temperatures, and cooling rates can significantly change as a function of depth from the surface. As a result, this section presents results that have been reported in literature as being factors affecting fatigue performance such as hardness, case microstructure, and residual stresses.

4.2.1 Radial Hardness Profiles

Vickers microhardness testing was used to characterize the hardness from the surface to the center of the induction hardened torsional fatigue specimens. Radial traverses were conducted at the plane of maximum shear stress (i.e. smallest diameter, 15.88 mm) as well as 95 pct of maximum shear stress (16.16 mm) to determine peak case hardness, effective case depth, and total case depth as well as equivalent hardness (see Section 2.2.5 for details)
for each steel in each induction-hardened condition. Figures 4.9 and 4.10 show radial hardness profile trend lines for the plane of maximum shear stress and 95 pct of maximum shear stress, respectively, for all four induction hardened conditions. Hardness data for each traverse are provided in Appendix C. In general, the two steels exhibit similar profile characteristics and the same nominal effective case depth – normalized for specimen radius – of 0.25, 0.32, 0.44, and 0.44 for the Low, Med, High, and High-SS conditions, respectively. The nominal normalized effective case depth is used for the remainder of this study. A nominal value is used because the scatter of the hardness data in the case/core transition region increases variability in measured effective case depth (see Appendix C).

![Graphs showing radial hardness profiles for Low, Med, High, and High-SS conditions at the maximum shear stress region.](image)

Figure 4.9  Radial hardness profiles for (a) Low, (b) Med, (c) High, and (d) High-SS conditions at the maximum shear stress region. Only trend lines are shown. Data used to develop trend lines are provided in Appendix C. A secondary axis converting Vickers to Rockwell C hardness is provided.
The *Med* condition was intentionally processed to achieve a diffuse case/core transition while all other conditions were intentionally processed to produce the sharpest transition possible. The *Low* condition achieved a very sharp case/core transition. The transition region in the *10V45-High* and *10V45-High-SS* conditions – at maximum shear stress – have two distinct zones, a diffuse transition zone from peak hardness to the effective case depth followed by a sharp transition zone to the core hardness.

![Normalized Depth from Surface](image1)

![Normalized Depth from Surface](image2)

![Normalized Depth from Surface](image3)

![Normalized Depth from Surface](image4)

Figure 4.10 Radial hardness profiles for (a) *Low*, (b) *Med*, (c) *High*, and (d) *High-SS* conditions at the 95 pct maximum shear stress region. Only trend lines are shown. Data used to develop trend lines are provided in Appendix C. A secondary axis converting Vickers to Rockwell C hardness is provided.

Figure 4.11 shows the average case hardness for each condition at both locations examined. In all but the *Med* condition, the 10V45 has a higher average case hardness than the 1045. This observation is consistent at both
maximum shear stress and 95 pct of maximum shear stress. The 10V45-Low condition has a higher average case hardness than any other condition, between 45 and 55 HV (about 2 HRC in this hardness range) higher than 1045-Low. The 10V45-High-SS condition shows the next higher case hardness, between 22 and 35 HV (again about 2 HRC) higher than 1045-High-SS. Figure 4.12 shows the average core hardness for each condition at both locations examined. The 1045 steel exhibited the same core hardness for all conditions. The 10V45 steel exhibited nearly the same core hardness for all conditions except High-SS, which indicates possible softening.

**Figure 4.11** Average case hardness for all induction hardened conditions of both steels at maximum and 95 pct of maximum shear stress. Error bars represent 95 pct confidence limits. A secondary axis converting Vickers to Rockwell C hardness is provided.

**Figure 4.12** Average core hardness for all induction hardened conditions of both steels at maximum and 95 pct of maximum shear stress. Error bars represent 95 pct confidence limits. A secondary axis converting Vickers to Rockwell C hardness is provided.

Figure 4.13 shows the equivalent hardness calculated for all conditions. Increasing case depth increases the equivalent hardness, as anticipated, since the equivalent hardness is effectively the integrated hardness over the transverse cross-section of an induction-hardened specimen. In general, the 1045 steel has lower equivalent hardnesses than 10V45, likely corresponding to both the lower case and core hardness for a given condition. High and High-SS conditions exhibit large differences between the maximum and 95 pct of maximum shear stress locations. The Med condition has similar hardness profiles for both steels with only a difference in core hardness.
This difference results in 10V45-Med only having a slightly higher equivalent hardness than 1045-Med. Table 4.13 provides a summary of data collected from the radial hardness profiles.

![Graph showing hardness values for different conditions](image)

Figure 4.13 Calculated equivalent hardness for all induction hardened conditions of both steels at maximum and 95 pct of maximum shear stress.

### Table 4.4 – Summary of Data Determined from Vickers Microhardness Traverses of All Induction Hardened Conditions

<table>
<thead>
<tr>
<th>Condition</th>
<th>Steel</th>
<th>Section From Cut Plan</th>
<th>Fraction of Shear Stress</th>
<th>Equivalent Hardness, HVeq (HV)</th>
<th>Average Hardness (HV)</th>
<th>Case/Core Hardness Ratio</th>
<th>Normalized Case Depth, t/r</th>
</tr>
</thead>
<tbody>
<tr>
<td>Low</td>
<td>1045</td>
<td>A</td>
<td>1.00</td>
<td>453</td>
<td>667 ± 6.8</td>
<td>225 ± 4.1</td>
<td>2.96</td>
</tr>
<tr>
<td></td>
<td></td>
<td>B</td>
<td>0.95</td>
<td>466</td>
<td>678 ± 10.2</td>
<td>229 ± 5.3</td>
<td>2.96</td>
</tr>
<tr>
<td>10V45</td>
<td>A</td>
<td>1.00</td>
<td>511</td>
<td>723 ± 5.0</td>
<td>292 ± 6.6</td>
<td>2.48</td>
<td>0.25</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>0.95</td>
<td>513</td>
<td>721 ± 7.8</td>
<td>282 ± 5.0</td>
<td>2.56</td>
<td>0.25</td>
</tr>
<tr>
<td>Med</td>
<td>1045</td>
<td>A</td>
<td>1.00</td>
<td>509</td>
<td>675 ± 6.2</td>
<td>227 ± 3.3</td>
<td>2.97</td>
</tr>
<tr>
<td></td>
<td></td>
<td>B</td>
<td>0.95</td>
<td>507</td>
<td>687 ± 11.2</td>
<td>225 ± 4.3</td>
<td>3.05</td>
</tr>
<tr>
<td>10V45</td>
<td>A</td>
<td>1.00</td>
<td>527</td>
<td>680 ± 5.0</td>
<td>282 ± 3.5</td>
<td>2.41</td>
<td>0.32</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>0.95</td>
<td>517</td>
<td>689 ± 8.2</td>
<td>282 ± 6.1</td>
<td>2.44</td>
<td>0.31</td>
</tr>
<tr>
<td>High</td>
<td>1045</td>
<td>A</td>
<td>1.00</td>
<td>559</td>
<td>657 ± 6.9</td>
<td>228 ± 4.7</td>
<td>2.88</td>
</tr>
<tr>
<td></td>
<td></td>
<td>B</td>
<td>0.95</td>
<td>553</td>
<td>674 ± 5.0</td>
<td>232 ± 9.7</td>
<td>2.91</td>
</tr>
<tr>
<td>10V45</td>
<td>A</td>
<td>1.00</td>
<td>582</td>
<td>681 ± 4.2</td>
<td>290 ± 5.3</td>
<td>2.35</td>
<td>0.45</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>0.95</td>
<td>605</td>
<td>687 ± 3.5</td>
<td>280 ± 5.1</td>
<td>2.45</td>
<td>0.43</td>
</tr>
<tr>
<td>High-SS</td>
<td>1045</td>
<td>A</td>
<td>1.00</td>
<td>551</td>
<td>662 ± 7.5</td>
<td>224 ± 2.7</td>
<td>2.96</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>0.95</td>
<td>550</td>
<td>655 ± 7.0</td>
<td>223 ± 3.8</td>
<td>2.94</td>
<td>0.42</td>
</tr>
<tr>
<td>10V45</td>
<td>A</td>
<td>1.00</td>
<td>592</td>
<td>697 ± 3.4</td>
<td>273 ± 5.3</td>
<td>2.55</td>
<td>0.45</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>0.95</td>
<td>566</td>
<td>677 ± 3.5</td>
<td>266 ± 5.9</td>
<td>2.55</td>
<td>0.43</td>
</tr>
</tbody>
</table>

Uncertainty given as the 95 pct confidence limits of the mean.

### 4.2.2 Macroetched Cross-sections

Macroetched cross-sections of the 1045 and 10V45 steels were prepared along the shoulder region of the torsional fatigue specimen (sections E and F from sectioning plan) as well as maximum and 95 pct of maximum shear stress (sections A and B from sectioning plan, respectively) for all conditions. The shoulder sections were
examined to ensure adequate induction hardening around stress concentrators as well as to ensure the grip section of the specimens were adequately hardened to prevent fretting fatigue during testing. Macroetched shoulder sections indicated that adequate hardening, between 20-30 pct of radius, was achieved. The maximum and 95 pct of maximum shear stress sections were examined to corroborate total case measurements from hardness data. Appendix D contains images of all of the macroetched cross-sections as well as tabulated visual total case depths for all conditions.

Figure 4.14 shows the visual total case depth measurements along the reduced section of torsional fatigue specimens from each condition along with scan direction for the Low, Med, and High conditions. Close examination revealed the visual total case depth is higher in the 1045 than the 10V45 steel in the scan-hardened conditions with only slight differences observed in the single-shot condition. In addition, scan hardening caused one of the specimen shoulder regions to have a higher visual total case depth in the Med and High conditions.

Figure 4.14  Normalized total case depth as a function of distance from the centerline of the induction hardened torsional fatigue specimens. Scan hardening direction indicated for Low, Med, and High conditions.

4.2.3  Microstructural Gradients

Light optical micrographs of the maximum shear stress cross-section were digitally stitched together to generate a single micrograph extending from the surface to the 4.5 mm into the torsional fatigue specimen. Figures 4.15-18 show microstructural gradient micrographs for each condition. Important observations from the Vickers microhardness profiles were converted either to Rockwell C or B Scale and incorporated into each micrographs along with a length scale. All conditions exhibit similar characteristics. The highest hardness region of the case etches uniformly, from a low magnification perspective. The zone between the high, constant hardness case and the effective case depth of 45 HRC exhibits a mixed microstructure with clear indications of large retained ferrite fractions. The transition between the effective case depth and the total case depth (either 26 HRC for the 10V45 or 96 HRB for the 1045) appears to be immediate in several conditions and gradual in others.
Figure 4.15  Microstructural gradient of Low induction hardened condition from the surface (left) to the core (right) for (a) 1045 and (b) 10V45 steels. Critical hardesses are indicated for each.

Figure 4.16  Microstructural gradient of Med induction hardened condition from the surface (left) to the core (right) for (a) 1045 and (b) 10V45 steels. Critical hardesses are indicated for each.
Figure 4.17  Microstructural gradient of High induction hardened condition from the surface (left) to the core (right) for (a) 1045 and (b) 10V45 steels. Critical hardnesses are indicated for each.

Figure 4.18  Microstructural gradient of High-SS induction hardened condition from the surface (left) to the core (right) for (a) 1045 and (b) 10V45 steels. Critical hardnesses are indicated for each.
Figure 4.15 shows the 1045 and 10V45-Low conditions. These conditions exhibit nearly identical microstructural gradients with a sharp transition between the effective and total case depth. All other conditions show clear differences between the 1045 and 10V45 steels in this zone; however, the High and High-SS conditions exhibit the largest differences. Figure 4.17 shows the microstructural gradient for the High condition. The effective case depth is very similar between the 1045 and 10V45; however, the core hardness for the 1045 steel is much further into the specimen, by almost 0.5 mm. Figure 4.18 shows the microstructural gradient for the High-SS condition. The High-SS condition exhibits an abrupt transition between the effective and total case depth in the 10V45 steel and a gradual transition in the 1045 steel, which extends to a deeper total case depth.

4.2.4 Case Microstructures

In general, all conditions showed a martensitic case with no clear evidence of ghost pearlite in the peak hardness region; however, some conditions showed a small fraction (<1 pct) of either blocky ferrite and/or another non-martensitic constituent. Figure 4.19 shows examples of micrographs highlighting non-martensitic transformation products. Blocky ferrite was observed to contain no carbides while other ferritic constituents vary in morphology, including Widmanstätten ferrite, and can contain carbides. Figure 4.20 shows SEM micrographs adjacent to the surface for each induction-hardened condition. Only the 10V45-Med and 10V45-High-SS conditions were observed to have blocky ferrite near the surface. Figure 4.21 shows SEM micrographs 0.5 mm below the surface for each induction-hardened condition. All of the 10V45 conditions were observed to have both blocky and other ferritic constituents at the depth of 0.5 mm. Only the Med and High-SS conditions in the 1045 steel exhibited either blocky ferrite and/or other ferritic constituents at the same depth. Table 4.14 summarizes the presence of ferritic constituents in the induction-hardened case of each condition.

Figure 4.19 Example SEM secondary electron micrographs showing non-martensitic transformation products in the induction hardened case. Taken at 10,000 times magnification with a 2 pct nital etch.
Figure 4.20  Representative SEM secondary electron micrographs showing the near surface induction hardened case microstructure. Alloy is constant by column and induction hardened condition is constant by row. Micrographs taken at 10,000 times magnification with a 2 pct nital etch.
Figure 4.21 Representative SEM secondary electron micrographs showing the induction hardened case microstructure 0.5 mm from the specimen surface. Alloy is constant by column and induction hardened condition is constant by row. Micrographs taken at 10,000 times magnification with a 2 pct nital etch.
### Table 4.5 – Observations of Non-martensitic Transformation Products (NMTP) in the Induction Hardened Case

<table>
<thead>
<tr>
<th>Location</th>
<th>Low</th>
<th>Med</th>
<th>High</th>
<th>High-SS</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>1045</td>
<td>10V45</td>
<td>1045</td>
<td>10V45</td>
</tr>
<tr>
<td>Surface</td>
<td>-</td>
<td>-</td>
<td>NMTP</td>
<td>-</td>
</tr>
<tr>
<td>0.5 mm</td>
<td>-</td>
<td>NMTP</td>
<td>NMTP</td>
<td>NMTP</td>
</tr>
</tbody>
</table>

Appendix E contains micrographs of the induction hardened case prior austenite grain structure, at the surface, for all conditions in the maximum shear stress region of the torsional fatigue specimens. No additional tempering was conducted on the specimens to obtain the presented results. Figure 4.22 shows the quantified mean lineal intercept of the austenite grain size as a function of condition for each alloy. Clear differences in austenite grain size are observed between the conditions, with the Low and Med conditions having much finer grain size than both High and High-SS conditions. The 10V45-Low condition had the overall smallest austenite grain size (9.9 ± 0.4 µm / ASTM No. 10) while the 1045-High condition had the overall highest grain size (27.0 ± 1.9 µm / ASTM No. 7). No statistical significance was found between the 1045 and 10V45 in the High and High-SS conditions; however, significant differences were found between the 1045 and 10V45 in the Low and Med conditions at α = 0.05 level. Significance was determined for each condition using two-sample t-tests with unequal variances.

![Austenite Grain Size Graph](image)

Figure 4.22  Near surface induction hardened case average austenite grain size for both steels as a function of condition. Uncertainty is shown as 95 pct confidence limits of the mean

### 4.2.5 Residual Stresses

Near surface residual stresses were quantified for replicates of each induction hardened condition by hole drilling with strain gages. Appendix F contains the residual stress profiles for all specimens tested. Differences in residual stress profiles between the 1045 and 10V45 steels for a given condition were not appreciable; as a result,
profiles were averaged for each processing condition. Figure 4.23 shows residual stress profiles for all conditions, which show similarities in profile shape. The highest residual stresses exist sub-surface, approximately 0.2-0.4 mm. In general, the axial direction exhibits the highest compressive residual stresses and the shear stresses in all conditions are measured around zero. The Low and Med conditions have a residual stress profile in which the axial stress is distinctly more compressive than the hoop stress, while the High and High-SS conditions exhibit a near-balanced biaxial profile. The Low and High-SS conditions have the most compressive residual stress profiles.

Figure 4.23  Near surface residual stresses profiles for (a) Low, (b) Med, (c) High, and (d) High-SS induction hardened conditions. Uncertainty is standard error of the mean for two specimens from each steel (i.e. four specimens total per condition).
4.3  Torsional Fatigue

Torsional fatigue testing was used to evaluate the influence of vanadium on the performance of induction-hardened shafts. This section presents stress-life data, macro-fractography, and SEM fractography for all of the induction-hardened conditions.

4.3.1  Fatigue Life

The torsional fatigue performance for all induction-hardened conditions was evaluated at shear stress amplitudes of 550, 600, and 650 MPa with a constant R-ratio of 0.1. All specimens tested were tested to failure. Figure 4.24 shows all 116 fatigue tests conducted in this study on a single stress-life plot. Surface initiated failures are shown as solid symbols while sub-surface initiations are open symbols. Initiation location was determined by visual inspection of the fracture surfaces, which will be discussed in more detail later in the chapter. Surface initiated failures were predominately observed (95 of 116 specimens, or 82 pct). Specimens that exhibited the longest fatigue lives for a given condition consistently resulted in sub-surface initiated failures. Tests ranged in life from 8,000 to almost 2,000,000 cycles depending on the test stress and induction hardened condition. Appendix G shows the stress-life data and initiation behavior graphically for each condition individually.

![Figure 4.24 Fatigue life as a function of shear stress amplitude for all 116 specimens tested. Surface and sub-surface initiations are shown as solid and open symbols, respectively.](image)

Figure 4.25 shows the mean stress life behavior for each condition. Error bars indicate the standard error of the mean. The 10V45-Low condition exhibited significantly higher cycles to failure at all stress levels than the 1045-Low condition; however, the difference between the two steels in all other conditions is almost indistinguishable. In addition, the 10V45-Low condition exhibits higher cycles to failure than both steels in the Med condition at all stress levels, which has a 28 pct higher effective case depth. Although the High and High-SS conditions have the same nominal effective case depth of 0.44, the conditions exhibit very different fatigue life behavior. The High-SS condition shows the highest fatigue life at the two highest test stresses for all conditions while in the High condition shows the highest fatigue life at the lowest test stress.
4.3.2 Fractography

Fractography was conducted using both low (<5x) and high (>100x) magnification techniques to characterize both macro- and micro-scale fatigue fracture features. Appendix G shows macro-photographs of all torsional fatigue specimens tested along with serial number, steel, stress amplitude, induction hardened condition, and cycles to failure. The general initiation location was determined via visual inspection. Select specimens were examined via SEM to confirm initiation location as well as characterized initiation and crack propagation behavior. Figure 4.26 shows macro-fractography of two representative fracture surfaces. Arrows indicate the initiation
locations in Figures 4.26a and 4.26b, exhibiting surface and sub-surface (case/core transition region in this case) fracture initiation, respectively. Both specimens are from the 10V45-High condition tested at a shear stress amplitude of 550 MPa. Table 4.15 provides the fraction of sub-surface initiations observed by condition. In general, specimens were observed to be more susceptible to sub-surface initiation with decreasing shear stress amplitude. Both the 10V45-Low and High-SS conditions resulted in a greater amount of sub-surface initiation than the 1045 steel in the same conditions. Both surface and sub-surface initiations exhibit distinct Stage II (stable crack propagation) regions. In surface initiated specimens a semi-elliptical stable crack growth region (i.e. thumbnail) was observed. In sub-surface initiated specimens, the Stage II region extends to the surface as Mode I and into the core as Mode III. The Mode I propagation into the case exhibits a distinctive pattern shown in Figure 4.26b.

![Figure 4.26](image)

Figure 4.26  Representative torsional fatigue fracture surfaces exhibiting (a) surface and (b) sub-surface (case/core transition region) initiation. Images are from the 10V45-High condition tested at 550 MPa. Arrows indicate fracture initiation sites.

<table>
<thead>
<tr>
<th>Shear Stress Amplitude (MPa)</th>
<th>Low</th>
<th>Med</th>
<th>High</th>
<th>High-SS</th>
</tr>
</thead>
<tbody>
<tr>
<td>1045</td>
<td>10V45</td>
<td>10V45</td>
<td>10V45</td>
<td>10V45</td>
</tr>
<tr>
<td>650</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>600</td>
<td>-</td>
<td>1/5</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>550</td>
<td>-</td>
<td>4/5</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

Fracture surfaces exhibit changes in macroscopic failure mode depending on the induction hardened condition and steel. Figure 4.27 shows representative fracture surfaces exhibiting the two primary fracture appearances observed in this study. Figure 4.27a shows a macroscopic Mode I failure initiation region followed by Mode I propagation around the case to final overload. Figure 4.27b shows a macroscopic Mode I initiation region followed by crack propagation around the case, which begins as Mode II (sliding shear) and transitions to Mode I.
Table 4.16 provides a summary of the observed failure modes for all conditions. Distinct differences are observed between the 1045 and 10V45 steels in the Low, High, and High-SS conditions. The 1045-Low condition exhibited greater ductility than any other condition tested with failure initiating and propagating in Mode II, while the 10V45-High and 10V45-High-SS conditions initiated and propagated exclusively in Mode I, a brittle-type fracture appearance in torsion. The Med condition exhibited multiple initiation at the two highest stress levels for both steels.

Figure 4.27 Representative torsional fatigue fracture surfaces showing macroscopic failure (a) Mode I initiation followed by Mode I propagation and (b) Mode I initiation followed by Mode II propagation which transitions to Mode I.

<table>
<thead>
<tr>
<th>Shear Stress Amplitude (MPa)</th>
<th>Low</th>
<th>Med</th>
<th>High</th>
<th>High-SS</th>
</tr>
</thead>
<tbody>
<tr>
<td>1045</td>
<td>I</td>
<td>P</td>
<td>I</td>
<td>P</td>
</tr>
<tr>
<td>10V45</td>
<td>I</td>
<td>P</td>
<td>I</td>
<td>P</td>
</tr>
<tr>
<td>650</td>
<td>2</td>
<td>2/1</td>
<td>1</td>
<td>2/1</td>
</tr>
<tr>
<td>600</td>
<td>2</td>
<td>2/1</td>
<td>1</td>
<td>1</td>
</tr>
<tr>
<td>550</td>
<td>1</td>
<td>2/1</td>
<td>1</td>
<td>1</td>
</tr>
</tbody>
</table>

Three criteria were used to determine which torsional fatigue specimens were examined via SEM: sub-surface initiation, lowest cycles to failure for a given test condition, or highest cycles to failure for a given test condition. These criteria are most likely to reveal differences between the steels, however, also increase the likeliness for a Type II error (false positive). As a result, SEM characterization was conducted on 62 torsional fatigue specimens (53 pct of all specimens tested). Figure 4.28 shows example SEM secondary electron images indicating commonly observed features at the initiation sites of all the torsional fatigue specimens examined.
Figure 4.28a shows an inclusion initiated failure. The rounded morphology of the inclusion suggests it is an oxide that formed during steelmaking. Of the specimens examined, 20 pct of surface initiations (8 of 41) and 14 pct of sub-surface initiations (3 of 21) were the result of inclusions. Inclusions ranged in size from 13 to 73 µm (\(\sqrt{\text{area}}\)) in surface initiations and 65 to 108 µm in sub-surface initiations. Figure 4.28b shows intergranular fracture at the failure initiation. Of the specimens examined, 20 pct of surface initiated failures (8 of 41) exhibited intergranular fracture at the initiation site, with the majority (6 of 8) being from the Med condition for both steels. The other two intergranular fractures were observed in the 10V45-High and High-SS conditions at the lowest cycles to failure at the highest stress amplitude (650 MPa). Figure 4.28c and 4.28d show indeterminate features at the initiation sites for surface and sub-surface initiated failures, respectively. Indeterminate features represent 69 pct (43 of 62) of the initiation sites examined.

Figure 4.28  Representative SEM secondary electron micrographs showing the four commonly observed initiation sites in all conditions of 1045 and 10V45 steels showing (a) an oxide inclusion, (b) intergranular fracture, (c) an indeterminate surface feature, and (d) an indeterminate sub-surface feature.
Figure 4.29 shows commonly observed fracture features in the Stage II stable crack propagation region. Intergranular regions surrounded by transcrystalline fracture dominate the Stage II region in all conditions. The intergranular regions are almost completely flat in the plane of the fracture. In addition, secondary cracking is observed sporadically in this region. Figure 4.30 shows a representative SEM secondary electron image indicating the demarcation of Stage II and Stage III. The flat, smooth intergranular fracture regions surrounded by transcrystalline fracture is seen in the upper-right portion of the micrograph (Stage II) while a mixture of intergranular and ductile fracture can be seen in the lower-left portion (Stage III).

Figure 4.29 Representative SEM secondary electron micrographs showing commonly observed Stage II crack propagation features, intergranular regions surrounded by transcrystalline fracture, in torsional fatigue tests of induction hardened 1045 and 10V45 steels.
Figure 4.30  Representative SEM secondary electron micrographs showing the Stage II (lower-left) to Stage III (upper-right) transition observed in torsional fatigue testing of induction hardened 1045 and 10V45 steels.

Figure 4.31-4.34 show the Stage III fracture behavior in the high hardness region of the case for the Low, Med, High, and High-SS, respectively. The Low and Med conditions exhibit very similar fracture features, consisting of both intergranular fracture and ductile fracture features, irrespective of steel. The amount of intergranular fracture appears to increase with increasing case depth. The High-SS condition exhibits primarily intergranular fracture with 10V45 having more than 1045.

Figure 4.31  Representative SEM secondary electron micrographs showing Stage III fast fracture in the high hardness region of the induction hardened case in (a) 1045-Low and (b) 10V45-Low conditions.
Figure 4.32  Representative SEM secondary electron micrographs showing Stage III fast fracture in the high hardness region of the induction hardened case in (a) 1045-Med and (b) 10V45-Med conditions.

Figure 4.33  Representative SEM secondary electron micrographs showing Stage III fast fracture in the high hardness region of the induction hardened case in (a) 1045-High and (b) 10V45-High conditions.
4.4 Gleeble® Physical Simulations

Three separate studies were conducted using Gleeble® simulations: a continuous heating study, an induction hardening simulation study, and a quench rate study. Results from microstructural analysis, hardness testing, and dilatometry are presented.

4.4.1 Continuous Heating Study

This study was devised to investigate vanadium effects under laboratory conditions similar to those seen during induction hardening, thus an industrially relevant peak temperature (1050 °C) and range of heating rates (100-1000 °C/s) were investigate. The microstructure, transformation behavior, and hardness were examined for each condition. Figure 4.35 shows representative light optical micrographs for the extremes of the parameters investigated for both steels. Microstructure of all conditions consisted primarily of martensite with regions of non-martensitic transformation products. Figure 4.36 shows an example of the non-martensitic transformation products observed in all of the conditions. The feature on the right is ferritic constituent without clear evidence of carbides while the feature on the left is a transformed ferritic constituent consisting of Widmanstätten ferrite, which grew from a grain boundary constituent consisting of ferrite and carbides. Qualitatively, the 10V45 steel appears to have a greater amount of non-martensitic transformation products than the 1045 steel.

Figure 4.37 shows representative dilatometry curves for the four heating rates of each material. As heating rate is increased, the $\text{Ac}_1$ (austenite transformation start during heating) appears to increase slightly while the $\text{Ac}_3$ (austenite transformation finish during heating) increases dramatically. The curves in Figure 4.37 can be somewhat misleading because under close examination the curves do not become linear until relatively high temperatures, above 900 °C in some instances. Even though the 1045 has a measurably larger ferrite fraction and grain size in the as-rolled microstructure, the critical temperatures are not significantly different from the 10V45.
Figure 4.35 Representative light optical micrographs of 1000 °C/s - 0.3 s, 100 °C/s - 0.3 s, and 100 °C/s - 3.0 s heat treatments for 1045 and 10V45. Alloy is constant by column and heat treatment is constant by row. Micrographs taken at 500 times magnification with a 2 pct nital etch.
Figure 4.36  Representative SEM secondary electron image showing non-martensitic transformation products observed in all test conditions in the continuous heating study for both steels. Micrographs taken at 15,000 times magnification with a 2 pct nital etch.

Figure 4.37  Representative dilatometry curves showing austenite formation during heating for the (a) 1045 and (b) 10V45 steels for all four heating rates used in this study, 100, 250, 500, and 1000 °C/s.

Figure 4.38 shows the prior austenite grain structure for the extremes of the parameters examined. Austenite grain size increased with increasing heating rate in both steels. Therefore, the finest grain size was achieved at the 1000 °C/s heating rate, 16.5 μm in 10V45 and 17.6 μm in 1045 (approximately ASTM No 8.5). The prior austenite grain size for a given heating rate was found to not vary significantly between the 1045 and 10V45 steels.
Figure 4.38  Representative light optical micrographs showing the prior austenite grain size of 1000 °C/s - 0.3 s, 100 °C/s - 0.3 s, and 100 °C/s - 3.0 s heat treatments for both 1045 and 10V45 steels. Alloy is constant by column and heat treatment is constant by row. Micrographs were taken at 200 times magnification.
Figure 4.39 shows Vickers microhardness results for replicates of each condition in this study. Although the 1045 material has a relatively constant hardness for the range of heating rates examined, the 10V45 shows an increasing hardness from 250 °C/s to 100 °C/s for 3.0 s. Table 4.17 provides a summary of all the dilatometric, cooling curve, and hardness data for this study.

![Vickers microhardness results for 1045 and 10V45 steels. Heat treatment is indicated on the x-axis as heating rate in °C/s followed by hold time at 1050 °C in seconds. Uncertainty is the 95 pct confidence interval of the mean.](image)

<table>
<thead>
<tr>
<th>Steel</th>
<th>Heating Rate (°C/s)</th>
<th>(\text{Ac}_1) (°C)(^a)</th>
<th>(\text{Ac}_3) (°C)(^a)</th>
<th>Holding Time (s) at 1050 °C</th>
<th>800-500 °C Quench Time (s)(^a)</th>
<th>(\text{Ms}) (°C)(^a)</th>
<th>Vickers Hardness (HV(_{0.5}))(^b)</th>
<th>Prior Austenite Grain Size (µm)(^b)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1045</td>
<td>1000</td>
<td>774 ± 1</td>
<td>895 ± 9</td>
<td>0.3</td>
<td>2.03 ± 0.02</td>
<td>360 ± 1</td>
<td>710.9 ± 12.2</td>
<td>17.6 ± 1.3</td>
</tr>
<tr>
<td></td>
<td>500</td>
<td>772 ± 2</td>
<td>891 ± 4</td>
<td>0.3</td>
<td>2.21 ± 0.02</td>
<td>357 ± 2</td>
<td>704.3 ± 9.9</td>
<td>17.5 ± 1.2</td>
</tr>
<tr>
<td></td>
<td>250</td>
<td>776 ± 3</td>
<td>887 ± 8</td>
<td>0.3</td>
<td>2.33 ± 0.04</td>
<td>357 ± 6</td>
<td>710.5 ± 8.7</td>
<td>19.8 ± 2.0</td>
</tr>
<tr>
<td></td>
<td>100</td>
<td>764 ± 3</td>
<td>875 ± 3</td>
<td>0.3</td>
<td>2.61 ± 0.04</td>
<td>354 ± 3</td>
<td>707.4 ± 9.9</td>
<td>25.1 ± 1.1</td>
</tr>
<tr>
<td></td>
<td>100</td>
<td>764 ± 3</td>
<td>875 ± 3</td>
<td>3.0</td>
<td>2.70 ± 0.03</td>
<td>357 ± 2</td>
<td>705.0 ± 3.9</td>
<td>32.3 ± 1.8</td>
</tr>
<tr>
<td>10V45</td>
<td>1000</td>
<td>777 ± 3</td>
<td>894 ± 2</td>
<td>0.3</td>
<td>2.12 ± 0.06</td>
<td>328 ± 5</td>
<td>722.1 ± 9.6</td>
<td>16.5 ± 1.0</td>
</tr>
<tr>
<td></td>
<td>500</td>
<td>778 ± 1</td>
<td>892 ± 6</td>
<td>0.3</td>
<td>2.30 ± 0.04</td>
<td>320 ± 1</td>
<td>720.3 ± 12.1</td>
<td>16.8 ± 1.0</td>
</tr>
<tr>
<td></td>
<td>250</td>
<td>771 ± 2</td>
<td>886 ± 3</td>
<td>0.3</td>
<td>2.45 ± 0.04</td>
<td>323 ± 3</td>
<td>705.6 ± 10.2</td>
<td>20.6 ± 1.3</td>
</tr>
<tr>
<td></td>
<td>100</td>
<td>770 ± 2</td>
<td>873 ± 2</td>
<td>0.3</td>
<td>2.69 ± 0.07</td>
<td>326 ± 4</td>
<td>734.5 ± 11.2</td>
<td>24.4 ± 1.4</td>
</tr>
<tr>
<td></td>
<td>100</td>
<td>770 ± 2</td>
<td>873 ± 2</td>
<td>3.0</td>
<td>2.81 ± 0.05</td>
<td>332 ± 1</td>
<td>754.9 ± 6.9</td>
<td>35.0 ± 2.0</td>
</tr>
</tbody>
</table>

\(^a\) Uncertainty given as ± one standard deviation.

\(^b\) Uncertainty given as 95 pct confidence interval of the mean.

### 4.4.2 Induction Hardening Simulation Study

Gleeble® simulations of the induction hardening process were conducted utilizing modeling data provided through industrial support. The surface and total case depth of the Low, High, and High-SS conditions were simulated using an electro-thermal finite element analysis software and then physically simulated for analysis via
dilatometry, hardness, and microscopy methods. Figure 4.40 shows the hardness results for all conditions of both steels. In general, all conditions exhibited consistent hardness levels in both the surface and total case depth simulations for a given steel, except the total case depth simulation in the High condition, which had much higher hardness than its 10V45 counterpart. Figure 4.41 shows SEM images of the 1045 and 10V45 steels in the total case depth simulation of the High induction hardened condition. Regions of martensite are highlighted in the 1045 with an "M" and circled in the 10V45 microstructures. The microstructures were quantified to have 13.5 ± 0.6 pct martensite in the 1045 and 1.1 ± 0.5 pct in the 10V45.

![Vickers microhardness results](image)

**Figure 4.40** Vickers microhardness results for 1045 and 10V45 steels from the simulated (a) surface and (b) total case depth for Low, High, and High-SS conditions. Error bars indicate the 95 pct confidence interval of the mean hardness.

![Representative SEM secondary electron images](image)

**Figure 4.41** Representative SEM secondary electron images of simulated total case depth in High condition for (a) 1045 and (b) 10V45 steels. Regions on martensite are emphasized by either a circle or an “M.” Micrographs were taken at 2000 times magnification with a 2 pct nital etch.
Figure 4.42 shows higher magnification SEM images of the regions quantified in the microstructure to be martensitic. Figure 4.42a shows retained carbides are present in the martensitic regions as well as evidence of newly precipitated carbides around the periphery. Figure 4.42b shows a region of martensite that is primarily surrounded by ferrite. The observed difference in microstructure was corroborated with dilatometric measurements. Figure 4.43 shows dilatometry curves for the total case depth simulation in the High condition for 1045 and 10V45. Distinct differences between the 1045 and 10V45 steels are observed at high temperature indicating more austenite formed in the 1045, which transformed to martensite upon cooling. Table 4.18 provides a summary of all the dilatometric and hardness data for this study.

![Figure 4.42 Representative SEM secondary electron images of simulated total case depth in High condition for (a) 1045 and (b) 10V45 steels at higher magnification. Regions on martensite are emphasized by either a circle or an “M.” Micrographs were taken at 20,000 times magnification with a 2 pct nital etch.](image)

![Figure 4.43 Representative dilatometry curves of simulated total case depth in the High condition for (a) 1045 and (b) 10V45 steels.](image)
Table 4.9 – Summary of Induction Simulation Study Data

<table>
<thead>
<tr>
<th>Steel</th>
<th>Condition</th>
<th>Location</th>
<th>Depth (mm)</th>
<th>Ac1 (°C)a</th>
<th>Ac3 (°C)a</th>
<th>Ms (°C)a</th>
<th>Vickers Hardness (HV0.5)b</th>
</tr>
</thead>
<tbody>
<tr>
<td>1045</td>
<td>Low</td>
<td>Surface</td>
<td>0.0</td>
<td>776 ± 6</td>
<td>901 ± 9</td>
<td>339 ± 2</td>
<td>684.6 ± 5.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Total Case</td>
<td>2.1</td>
<td>---</td>
<td>---</td>
<td>---</td>
<td>214.8 ± 2.9</td>
</tr>
<tr>
<td></td>
<td>High</td>
<td>Surface</td>
<td>0.0</td>
<td>780 ± 2</td>
<td>890 ± 4</td>
<td>366 ± 1</td>
<td>688.5 ± 5.3</td>
</tr>
<tr>
<td></td>
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<td>752 ± 1</td>
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<td>254.5 ± 10.8</td>
</tr>
<tr>
<td></td>
<td>High-SS</td>
<td>Surface</td>
<td>0.0</td>
<td>779 ± 2</td>
<td>899 ± 2</td>
<td>361 ± 8</td>
<td>684.5 ± 6.6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Total Case</td>
<td>3.8</td>
<td>---</td>
<td>---</td>
<td>---</td>
<td>214.6 ± 2.3</td>
</tr>
<tr>
<td>10V45</td>
<td>Low</td>
<td>Surface</td>
<td>0.0</td>
<td>791 ± 8</td>
<td>905 ± 2</td>
<td>342 ± 5</td>
<td>721.1 ± 5.2</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Total Case</td>
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<td>---</td>
<td>---</td>
<td>---</td>
<td>268.3 ± 2.8</td>
</tr>
<tr>
<td></td>
<td>High</td>
<td>Surface</td>
<td>0.0</td>
<td>784 ± 6</td>
<td>899 ± 9</td>
<td>360 ± 1</td>
<td>715.9 ± 4.2</td>
</tr>
<tr>
<td></td>
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<td>Total Case</td>
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<td>752 ± 0</td>
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<td>2701 ± 3.4</td>
</tr>
<tr>
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<td>High-SS</td>
<td>Surface</td>
<td>0.0</td>
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<td>902 ± 9</td>
<td>336 ± 2</td>
<td>716.3 ± 4.9</td>
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<tr>
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<td>---</td>
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<td>267.3 ± 4.0</td>
</tr>
</tbody>
</table>

a Uncertainty given as ± one standard deviation.
b Uncertainty given as 95 pct confidence interval of the mean.

4.4.3 Quench Rate Study

The simulated surface thermal profile for the Low induction hardened condition was used along with modified Gleeble® ISO-Q® specimens to determine the quench rate dependence of 1045 and 10V45 hardness after induction hardening. Three quench rates ranges were achieved: a low quench rate (LQR) of 300-400 °C/s, a medium quench rate (MQR) of 800-1000 °C/s, and a high quench rate (HQR) of 2500-3250 °C/s. A hold at 1070 °C for either 0 s (i.e no hold) or 30 s was incorporated as well. The microstructure of select specimens was examined via light optical microscope and was determined to be completely martensitic. Appendix I shows representative light optical micrographs of the LQR and HQR specimens for both steels.

![Figure 4.44 Average hardness for low (LQR), medium (MQR), and high quench rates (HQR) specimens of 1045 and 10V45 steels with and without a 30 s hold at 1070 °C prior to quenching. Error bars represent 95 pct confidence limits.](image-url)
Figure 4.44 shows the hardness data from the quench rate study. The 10V45 has the highest hardness at all quench rates. The 1045 and 10V45 conditions—without a hold—exhibited a slight increase in hardness with increasing quench rate. With a 30 s hold, the 1045 had a lower hardness at all quench rates than it had without a hold. The 10V45, with a 30 s hold, had a consistent hardness that is independent of quench rate and is similar to the LQR 10V45 with no hold. Table 4.19 provides a summary of hardness, dilatometric, and cooling curve data for this study. The medium and high cooling rate specimens were tubes; therefore, diametrical dilatometry was not conducted.

Table 4.10 – Summary of Quench Rate Study Data

<table>
<thead>
<tr>
<th>Steel</th>
<th>Simulated Thermal Cycle</th>
<th>Holding Time (s) at 1070 °C</th>
<th>Ac₁ (°C)a</th>
<th>Ac₃ (°C)a</th>
<th>800-500 °C Quench Time (s)a</th>
<th>Nominal Cooling Rate (°C/s)</th>
<th>Mₛ (°C)a</th>
<th>Vickers Hardness (HV₀.₅)b</th>
</tr>
</thead>
<tbody>
<tr>
<td>1045</td>
<td>Low Surface (0.0 mm)</td>
<td>0</td>
<td>776 ± 6</td>
<td>901 ± 9</td>
<td>0.728 ± 0.007</td>
<td>410</td>
<td>339 ± 2</td>
<td>684.6 ± 5.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>30</td>
<td>---</td>
<td>---</td>
<td>0.828 ± 0.016</td>
<td>360</td>
<td>---</td>
<td>675.5 ± 3.1</td>
</tr>
<tr>
<td></td>
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<td>0</td>
<td>---</td>
<td>---</td>
<td>0.282 ± 0.006</td>
<td>1060</td>
<td>---</td>
<td>706.7 ± 5.5</td>
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<tr>
<td></td>
<td></td>
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<td>---</td>
<td>0.370 ± 0.054</td>
<td>810</td>
<td>---</td>
<td>697.3 ± 3.5</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0</td>
<td>---</td>
<td>---</td>
<td>0.099 ± 0.001</td>
<td>3030</td>
<td>---</td>
<td>712.8 ± 5.9</td>
</tr>
<tr>
<td></td>
<td></td>
<td>30</td>
<td>---</td>
<td>---</td>
<td>0.117 ± 0.003</td>
<td>2560</td>
<td>---</td>
<td>701.4 ± 4.8</td>
</tr>
<tr>
<td>10V45</td>
<td>Low Surface (0.0 mm)</td>
<td>0</td>
<td>791 ± 8</td>
<td>905 ± 2</td>
<td>0.740 ± 0.010</td>
<td>400</td>
<td>342 ± 5</td>
<td>721.1 ± 5.2</td>
</tr>
<tr>
<td></td>
<td></td>
<td>30</td>
<td>---</td>
<td>---</td>
<td>0.887 ± 0.000</td>
<td>330</td>
<td>---</td>
<td>716.8 ± 5.9</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0</td>
<td>---</td>
<td>---</td>
<td>0.288 ± 0.021</td>
<td>1040</td>
<td>---</td>
<td>738.5 ± 4.6</td>
</tr>
<tr>
<td></td>
<td></td>
<td>30</td>
<td>---</td>
<td>---</td>
<td>0.353 ± 0.000</td>
<td>840</td>
<td>---</td>
<td>718.7 ± 4.0</td>
</tr>
<tr>
<td></td>
<td></td>
<td>0</td>
<td>---</td>
<td>---</td>
<td>0.092 ± 0.005</td>
<td>3240</td>
<td>---</td>
<td>746.2 ± 3.7</td>
</tr>
<tr>
<td></td>
<td></td>
<td>30</td>
<td>---</td>
<td>---</td>
<td>0.118 ± 0.001</td>
<td>2530</td>
<td>---</td>
<td>721.3 ± 3.8</td>
</tr>
</tbody>
</table>

a Uncertainty given as ± one standard deviation.

b Uncertainty given as 95 pct confidence interval of the mean.
The working hypothesis for this research was that vanadium precipitation strengthens the core ferrite-pearlite microstructure of induction-hardened shafts, improving torsional strength and therefore torsional fatigue performance. Figure 5.1 shows the measured and calculated yield and ultimate tensile strength for the 1045 and 10V45 steels in the as-received, hot rolled condition. Properties were calculated using microstructural data for the two steels along with the established structure-property relationships presented in Sections 2.1.3 and 2.1.4. The measured yield and ultimate tensile strength are approximately 5 pct lower than calculated for the 1045 steel. In contrast, the 10V45 steel exhibited properties significantly higher than calculated, 20 pct for yield and 8 pct for ultimate tensile strength. The estimated maximum yield strength increase due to precipitation strengthening was 30 pct, assuming a precipitate diameter of 3 nm and volume fraction of 0.0013. The observed yield strength in the ferrite-pearlite 10V45 falls between the calculated values assuming no precipitation strengthen and maximum. Although, precipitation strengthening primarily influences yield strength, tensile strength increases as well due the strain-hardening rate being controlled predominately by interlamellar spacing at these high pearlite fractions [88].

Figure 5.1 Comparison of calculated and measured tensile data for the as-received, hot-rolled 1045 and 10V45 steels. Yield strength (YS) and ultimate tensile strength (UTS) were calculated using equations from Gladman et al. [1]. Estimated maximum precipitation strengthening contribution to the 10V45 yield strength is from Gladman [13].

Although the 10V45 steel has substantially greater tensile properties in the as-receive, hot rolled condition than the 1045 steel, little is known about how it responds to induction hardening and ultimately influences torsional fatigue performance. This chapter discusses the evolution of the as-received, hot-rolled 1045 and 10V45 microstructures though the induction hardening process and its subsequent effect on torsional fatigue performance.
5.1 Microstructure Evolution during Induction Hardening

The microstructure of an induction-hardened component dictates characteristics linked to mechanical performance such as hardness, strength, and fatigue life (see Sections 2.2.5 and 2.3.1). Differences in microstructure were observed between the 1045 and 10V45 steels both pre- and post-induction hardening. The following sections discuss the previously presented data regarding differences in as-received condition, case region, and case/core transition region microstructures.

5.1.1 Effect of Starting Microstructure

The 1045 and 10V45 steels show distinct differences in as-received, hot rolled microstructure. The 10V45 steel has smaller ferrite fraction and finer ferrite grain size as well as finer pearlite interlamellar spacing. The difference in microstructure can be explained by the difference in carbon content between the two steels, with the 10V45 steel having approximately 0.03 wt pct more carbon than the 1045 steel. Figure 5.2 shows calculated continuous cooling transformation (CCT) diagrams for the 1045 and 10V45 steels as well as their approximate cooling rate after hot rolling. The higher carbon content results in a suppression of the ferrite and pearlite transformation temperatures for a given cooling rate. Transformation at lower temperatures results in an overall finer microstructure, reducing ferrite fraction and grain size as well as producing finer pearlite. The finer microstructure of the 10V45 steel is advantageous for induction hardening. Microstructures with shorter carbon diffusion distances as well as finer carbide size distributions are more readily austenitized, resulting in a more uniformly induction hardened component [31].

![Figure 5.2 Calculated continuous cooling transformation (CCT) curves for the 1045 and 10V45 steels with the approximate cooling rate of the bars after hot rolling overlaid. CCT diagrams calculated from Edison Welding Institute web tool [1]. Cooling rate data is from Cryderman [89].](image)

Finer microstructures typically exhibit a downward shift in critical temperatures for a given steel composition and heating rate [31]. Although the 10V45 steel has a finer microstructure than the 1045, a downward
shift in critical temperatures during continuous heating experiments was not observed. Figure 5.3 shows a time temperature austenitizing (TTA) diagram for ferrite-pearlite Ck45 steel with data from the current study superimposed. Ck45 is a German (DIN) steel specification for a wrought medium-carbon steel with a chemical composition range that encompasses both the 1045 and 10V45 steels (without microalloying). Differences in critical temperatures between the 1045 and 10V45 steels were indistinguishable for the heating rates examined, 100-1000 °C/s. The Ac1 nominally occurred at 775 °C while the Ac3 occurred approximately 100 °C higher, between 875 and 895 °C, with the higher heating rates resulting in a higher Ac3. The 1045 and 10V45 data only deviate slightly from the Ck45 data from literature. The Ck45 steel exhibited a more profound change in Ac3 over the range of heating rates examined. The degree of homogeneity of the austenite formed during these experiments was not quantified.

In addition to a finer microstructure, the 10V45 is precipitation strengthened. V(C,N) precipitates were previously shown via dark field TEM to range in size from 3 to 5 nm in diameter. Dissolution kinetics of the V(C,N) precipitates were calculated for conditions similar to those experienced during an induction hardening cycle using the analysis by Whelan [48] (see Section 2.2.4). Figure 5.4 shows the isothermal dissolution kinetics for 3 to 6 nm diameter precipitates of both VC0.75 and VN at 1000 and 1070 °C. Both VC0.75 and VN dissolution kinetics were calculated because they represent the lower and upper bound of chemical composition for vanadium precipitates present in the 10V45 steel, respectively. The dissolution temperature of 1000 °C corresponds to the minimum recommended induction hardening temperature for a 0.45 wt pct C steel containing strong carbide formers (i.e. vanadium) [31]. The dissolution temperature of 1070 °C corresponds to the calculated peak temperature for the Low induction hardened condition in the current study. At 1000 °C, VC0.75 precipitates dissolve rapidly, going in to solution in 2 to 6 s, while VN precipitates are more sluggish and take between 8 and 33 s. At 1070 °C, VC0.75
precipitates dissolve almost instantly (<1 s), while VN precipitates can take as long as 4-5 s to dissolve. Given the extremely fine size of the precipitates, the dissolution time may be even less due to the Gibbs-Thompson effect (see Section 2.2.4). A study by Rivas et al. [9-11] examined the evolution of VC precipitate size distributions, of approximately the same initial size, during induction hardening. Rivas et al. suggest VC precipitate size actually increases in the induction hardened case region; however, the statistical significance of this observation was inconclusive. The coarsening of VC precipitates was attributed due to the relatively slow heating cycle used in the study to the peak temperature, 22 s to 1040 °C. In the current study, the longest calculated heating time to peak temperature was 1.8 s.

![Isothermal dissolution kinetics of (a) VC_{0.75} and (b) VN at 1000 and 1070 °C for both 3 and 6 nm precipitates. Calculations made using Equations 2.2, 2.12, 2.14, and 2.15 as well as data from Table 2.1.](image)

**Figure 5.4** Isothermal dissolution kinetics of (a) VC_{0.75} and (b) VN at 1000 and 1070 °C for both 3 and 6 nm precipitates. Calculations made using Equations 2.2, 2.12, 2.14, and 2.15 as well as data from Table 2.1.

### 5.1.2 Case Region Non-Martensitic Transformation Products

In the current study, the induction hardening cycles were very rapid. Calculated peak surface temperatures of 1070, 1089, and 1100 °C were achieved in 1.0, 1.5, and 1.8 s for the Low, High, and High-SS induction hardened conditions, respectively (see Section 3.3.5). Although the calculated surface temperatures were very high, the 10V45 steel consistently produced a case microstructure that is not fully martensitic. Figure 5.5 shows an SEM micrograph at the surface of an induction hardened torsional fatigue specimen of 10V45 in the High-SS condition with a region of non-martensitic transformation product. Interestingly, the case hardness was not affected by the existence of the non-martensitic transformation products. This is likely due to the very small fraction of non-martensitic transformation products present in the microstructure not being resolvable with microhardness testing, and simply contribution to the variability of hardness data in the case region. The observation of non-martensitic transformation products at the induction-hardened surface suggest either the modeling is inaccurate or there is a vanadium effect on hardenability. Although, the induction hardening simulations were not validated for the current study by induction hardening instrumented specimens, visual confirmation of the surface temperatures during specimen processing was
determined to be reasonable by experienced metallurgists. Fett and Held [45] reported vanadium microalloyed steels have been observed by to exhibit “inferior” induction hardened case microstructures when compared to plain carbon steels; however, no mechanism was proposed. Vanadium precipitates along austenite grain boundaries have been shown to promote ferrite nucleation in low-carbon steels [90]. This mechanism may provide evidence for the formation of non-martensitic transformation products in medium-carbon steels as well. Indicating vanadium reduces the hardenability of medium-carbon steels by promoting ferrite formation.

![Figure 5.5](image1.png)

**Figure 5.5**  Representative SEM secondary electron image of High-SS induction hardened condition in the 10V45 steel showing regions of non-martensitic transformation products in the case region adjacent to the surface. Micrograph taken at 10,000 times magnification with a 2 pct nital etch.

### 5.1.3 Case/Core Transition Region Austenitization Behavior

In the case/core transition region, the heating rates and peak temperatures are significantly lower compared to the surface. Figure 5.6 shows the calculated thermal profile for the total case depth (3.8 mm from the surface) in the High induction hardened condition. A reasonably high heating rate of approximately 450 °C/s is achieved; however, the peak temperature is within the intercritical regime, which is sustained for approximately 1.2 s before quenching. In the current study, this thermal cycle resulted in clear differences in austenitization behavior. The 1045 steel transformed to approximately 10 pct austenite while the 10V45 steel transformed to approximately 1 pct.

![Figure 5.7](image2.png)

**Figure 5.7**  Shows representative SEM micrographs of the 1045 steel processed according to the thermal profile in Figure 5.6 using the Gleeble®. Regions that were transformed to austenite are highlighted with arrows. Although there are possible stereological affects, austenite nucleation sites appear to be predominately regions around the proeutectoid ferrite such as pearlite/proeutectoid ferrite boundaries and not within the pearlite colonies. This observation was consistent with the 10V45 steel. Figure 5.8 shows representative SEM micrographs of the 10V45 steel processed identically to those presented in Figure 5.7 for the 1045 steel. Higher magnification micrographs are shown for the 10V45 specimens due to the lower fraction of austenite that was formed during processing. Regions of austenite that formed during the short thermal simulations transformed upon cooling to primarily martensite with retained carbides. However, the periphery of many martensitic regions exhibit very fine carbides between the cementite lamella. These regions likely transformed to austenite but had insufficient cooling rate to form martensite. As a result, austenite fractions measured by only quantifying the martensite fraction are likely a lower limit of the true amount of austenite that formed during the thermal cycle. The difference in austenitizing behavior between the two steels was repeated and microstructural observations were corroborated with dilatometry.
The reduced austenite formation in the 10V45 steel at the simulated total case depth in the High condition is consistent with measurements and observations made on induction hardened torsional fatigue specimens in the same condition. Fett and Held [45] made a similar observation in a series of medium-carbon steels, including both vanadium and niobium microalloyed steels. Fett and Held found that both vanadium and niobium microalloying reduced the depth of 40 HRC, with niobium having the greatest effect. Suggesting the mechanism reducing austenite formation is closely related to microalloy precipitates either reducing austenite nucleation rate at low austenitizing temperatures or inhibiting austenite grain growth.

![Figure 5.6](image)

Figure 5.6 Calculated total case depth (3.8 mm) thermal profile for the High induction hardened condition. Critical temperatures calculated for the 1045 steel using equations from Andrews [91].

![Figure 5.7](image)

Figure 5.7 Representative SEM secondary electron images of 1045 total case depth Gleeble ® simulation from High induction hardened condition. Arrows indicated regions of that transformed to austenite during the physical simulation. Micrographs taken at 2000 times magnification with a 2 pct nital etch.
Figure 5.8  Representative SEM secondary electron images of 10V45 total case depth Gleeble ® simulation from *High* induction hardened condition. Arrows indicated regions of that transformed to austenite during the physical simulation. Micrographs taken at (a) 5000 times and (b-d) 10,000 times magnification with a 2 pct nital etch.

### 5.2 Torsional Fatigue Performance

Torsional fatigue performance for each of the four induction hardened conditions was assessed at shear stress amplitudes of 550, 600, and 650 MPa. A positive stress ratio (R=0.1) was selected to preserve fracture surfaces, therefore the maximum shear stress at the specimen surface was very high – 1222, 1333, and 1444 MPa for the 550, 600, and 650 MPa shear stress amplitudes, respectively. Torsional yield and ultimate strength for induction-hardened steel shafts with a similar chemical composition and range of case depths were reported in literature to be 550-850 MPa and 1050-1450 MPa, respectively [62]. The high shear stress amplitude may have implications regarding mean stress effects (see Section 2.3); however, the high amplitude was necessary to ensure all specimens were tested to failure at the same three stress levels to facilitate direct comparison.
The following sections will discuss the torsional fatigue performance for the conditions examined in the present study. Differences in fatigue life and fracture behavior will be discussed and related to factors including hardness, residual stress, and induction hardening parameters. Finally, the empirical model developed from literature as part of experimental design (see Section 3.2.8) was compared to data from the present study and subsequently incorporated into a new revision of the model.

5.2.1 Fatigue Life – Effects of Hardness and Residual Stress Distribution

It has been proposed in literature that effective case depth is most closely related to torsional strength and total case depth to torsional fatigue performance [62]. However, the present study suggests other attributes are possibly more important to torsional fatigue performance than either effective or total case depth. Figure 5.9 shows the radial hardness profiles for the 1045 and 10V45 steels in all induction hardened conditions. Hardness profiles for each condition comparing the two steels were previously presented in Section 4.2.1. The “as-received” horizontal lines on the plots are the as-received, hot rolled hardness determined from independent specimens and the line width are the upper and lower 95 pct confidence limits. Overall, the 10V45 conditions exhibited both higher case and core hardnesses than the 1045 steel. The 1045 induction hardened conditions exhibited very similar case hardness with a smooth transition region and a core hardness that is consistent with the as-received, hot rolled steel. The 10V45 induction hardened conditions exhibited very different behavior. The 10V45-Low condition has a distinctly higher case hardness than all other conditions (~1 HRC). Higher variation in core hardness was also observed; however, still relatively close to the as-received, hot rolled steel. The case/core transition region in the High and High-SS conditions has two distinct zones; a near linear decrease from the case hardness to approximately the effective case depth followed by an abrupt drop to the core hardness. The depth from the surface where the sudden drop in hardness occurs is approximately the visually measured total case depth. This result supports the earlier observation that showed the 10V45 steel formed less martensite in the region around the total case depth than the 1045 steel. The case hardness in both 1045 and 10V45 were nearly identical in the Med condition, with 1045-Med having the highest case hardness of the 1045 conditions and 10V45-Med having the lowest case hardness of the 10V45 conditions. The High and High-SS conditions exhibited nearly identical hardness profiles, with only slight differences in total case depth in the 1045 steel as well as case and core hardness in the 10V45 steel.

Case hardness explains the occurrence of sub-surface initiated fatigue failures in all conditions of the present study. The higher the case hardness, the more likely initiation will occur sub-surface for a given shear stress amplitude and case depth. As shear stress amplitude is decreased, the likeliness of approaching the fatigue limit of the case microstructure increases. This is the reason the 10V45 steel appears to be more susceptible to sub-surface initiation than the 1045 steel for a given induction hardened condition. For example, the 10V45-Low steel condition exhibited a very high case hardness. This high case hardness resulted in sub-surface initiations being observed at the two lowest stress levels while no sub-surface initiations were observed in the 1045 steel at any stress level. In addition, the number of sub-surface initiated failures increased with decreasing shear stress amplitude. The case hardness of both steels in the Med condition was very similar; as a result, no sub-surface initiations were observed. In addition, sub-surface initiated fatigue failures exhibited the longest fatigue lives for a given induction hardened condition and shear stress amplitude. Fatigue failure initiation data were presented previously in Section 4.3.2.
Compressive residual stress formed at the surface during induction hardening can be very beneficial in high-cycle fatigue by inhibiting fatigue crack initiation; however, the effect is diminished at high stress amplitudes due to stress relaxation [92]. In the current study, residual stresses were quantified in replicates of each steel for each induction hardened condition by incremental strain gage hole drilling. Due to the limitations of the hole drilling method, only the near surface residual stresses were measured (<1 mm). The magnitude of the compressive residual stresses were found to be primarily a function of the processing and not a function of steel (see Section 4.2.5). Therefore, the specimens were analyzed by condition and not by steel.

It is generally accepted that residual stresses influence the mean stress amplitude during fatigue [92]; however, in torsional fatigue the influence of mean stress is debatable (see Section 2.3). Therefore, the maximum influence of near surface compressive residual stresses were determined by subtracting them from the maximum principal stress. Figure 5.10 shows schematically the summation of the maximum compressive residual stress state and applied stress state. From Mohr’s circle representations of each state it is clear that the applied stress dominates the net stress state. Figure 5.11 shows the reduction of the maximum applied principal stress due to compressive residual stresses for each induction hardened condition at an applied shear stress amplitude of 550 and 650 MPa. The Low and High-SS conditions reduced the applied principal stress the most, ~45 pct at 550 MPa and ~39 pct at 650 MPa. The Med and High conditions reduced the applied principal stress to a lesser extent, ~34 pct at 550 MPa and ~29 pct at 650 MPa.

Figure 5.12 shows the shear stress-life data for both 1045 and 10V45 steels. Data are presented as the mean of all specimens at a single stress level – both surface and sub-surface initiated – and standard error of the mean with trend lines for visual reference only. Prior austenite grain size for each condition in shown as well. Fatigue life data for each condition, comparing the two steels, were previously presented in Section 4.3.1.
Figure 5.10  Schematic showing the summation of the residual and applied stress states to create a net stress state that can be transformed to principal stresses. Mohr’s circle is shown along the bottom for a visual aide.

Figure 5.11  Reduction of maximum applied principal stress due to residual stresses in all induction hardened conditions at (a) 550 MPa and (b) 650 MPa shear stress amplitude.

In general, vanadium microalloying was observed to negligibly influence the torsional fatigue life of induction-hardened shafts for the parameters examined in this study. The one exception is the Low induction hardened condition, in which case the 10V45 steel showed an approximate 75 pct improvement in fatigue life over the 1045 steel at all stress levels. The improved fatigue performance of the 10V45-Low condition is likely a combination of higher case and core hardness as compared to the 1045 steel in the same condition. The effect of residual stresses are constant in this comparison as well are the effects of prior austenite grain size due to the debatable difference between a 10 and 13 µm grain size. Vanadium precipitation strengthening undoubtedly increased the core hardness; however, it is unclear if vanadium played a role in the increased case hardness due to the increased carbon content of the 10V45 steel. Besides the 10V45-Low condition, the most significant differences
in fatigue behavior were observed between induction hardening conditions. The fatigue life of the induction hardened conditions in the 1045 steel increase with increasing effective case depth, for a given shear stress amplitude. This behavior is in agreement with previously reviewed literature (see Sections 2.2.5 and 2.3.1). However, the 10V45-Low condition has higher fatigue life than the Med condition (7 pct higher effective case) and a comparable fatigue life to the High condition (19 pct higher effective case) at the two highest stress levels. The three factors that may be attributed to this observation are case hardness, residual stresses, and prior austenite grain size. Both 10V45-Med and High conditions had less favorable residual stress profiles and significantly lower case hardness than the 10V45-Low condition, while the 10V45-High condition had a considerable larger prior austenite grain size. Larger grain sized materials tend to have a lower strength and endurance limit; however, they are more tolerant to defect initiated failure [92].

The High and High-SS conditions differ significantly despite having the same effective case depths. At the two high stress levels, the High-SS condition exhibits almost three times higher fatigue life (200-300 pct) than the High condition. At the lowest stress level, the opposite is observed, with the High condition having higher fatigue life than the High-SS condition. The sudden improvement in mean fatigue life along with increased variance of the High condition data at the lowest stress level suggests the condition is approaching its fatigue limit [92]. Differences in both prior austenite grain size and residual stresses were observed between the two conditions with the High condition having a higher prior austenite grain size and a less compressive residual stresses near the surface. Residual stress effects should be emphasized at lower shear stress amplitudes; however, the High condition has higher mean cycles to failure at the low stress level and a less compressive residual stress profile than the High-SS condition. These results suggest residual stresses are not a predominant factor in the fatigue life behavior between the two conditions. The higher case prior austenite grain size of the High condition may result in a lower fatigue life

Figure 5.12  Torsional fatigue life results for all induction hardened conditions of (a) 1045 and (b) 10V45 steels. Symbols indicate the mean life of five tests. Uncertainty is shown as standard error of the mean.
at high stress levels. As shear stress amplitude is decreased to below a critical shear stress amplitude, the surface is no longer sensitive to features initiating failure, driving initiation subsurface.

Case hardness has been shown to influence initiation behavior as well as likely improving the fatigue life in the 10V45-Low condition. However, the role of vanadium on the case hardness is unclear due to the higher carbon content of the 10V45 steel. Figure 5.13 shows hardness as a function of quench rate for Gleeble® specimens rapidly austenitized to a peak temperature of 1070 °C and held for either zero or 30 s before quenching. The heating cycle simulated the calculated surface thermal profile for the Low induction hardened condition (see Sections 3.3.5 and 3.3.6). Calculate upper and lower bounds for the hardness of both steels are superimposed [23]. The 1045 steel hardness shows a quench rate dependence for both hold times at the peak temperature. Increasing the quench rate by an order of magnitude, from ~350 to ~3500 °C/s, increases the hardness by ~25 HV (~1 HRC) and ~15 HV (~0.5 HRC) above literature values. The shape of the quench rate dependency appears to be relatively consistent between the two hold times, with the longer hold time shifting to slightly lower hardnesses, likely due to a larger austenite grain size. Hardness of the 10V45 steel with no hold shows the same functional dependency with quench rate as the 1045 steel, except shifted to significantly higher hardnesses. In this condition, an increase of ~25 HV (1 HRC) was observed over the same range of quench rates as the 1045 steel. At the highest quench rate, an increase in hardness of ~25 HV (1 HRC) over literature values was observed. The hardness of the 10V45 steel with a 30 s hold at 1070 °C becomes nearly constant at ~720 HV (~61 HRC) for all quench rates. The time at 1070 °C was calculated to be sufficient to dissolve all vanadium precipitates (see Figure 5.4) suggesting vanadium in solid solution is influencing the quench rate dependence of the martensite hardness. These results also provide evidence that not all vanadium precipitates are going into solution in the case during induction hardening, regardless of quench rate.

![Figure 5.13](image-url) Vickers hardness as a function of cooling rate for Gleeble® specimens of both steels processed through a simulated surface thermal profile of the Low induction hardened condition. Specimens were held at the peak temperature (1070 °C) for either zero (0) or 30 s then quenched at rates from 330 to 3500 °C/s to room temperature. Specimen were not tempered before hardness testing.
5.2.2 Fracture Behavior – Effects of Hardness and Induction Hardening Parameters

Fracture behavior was found to be primarily dependent on case hardness and the induction hardening condition. Figure 5.14 shows the typical fracture behavior of a surface initiated failure. In both 1045 and 10V45 specimens, the Stage II stable crack growth region (lower left) is primarily transcrystalline with flat, smooth areas approximately the same size as the measured prior austenite grain size. Stage III unstable crack growth (upper right) progressed around the case from the area of initiation. The high hardness region of the case (>650 58 HRC) fractured predominantly in an intergranular mode with small areas of transcrystalline and ductile void coalescence, observed as the semi-reflective outer case region in the macro-photograph (lower center). The lower hardness region of the case (<650 58 HRC) and the case/core transition region fractured in a ductile manner during Stage III, perceived as the dull region between the outer case and the core. No apparent relationship was observed for Stage I micro-crack initiation or Stage II stable crack growth between the two steels.

Figure 5.14 Progression of crack growth in a surface initiated torsional fatigue failure. Micrographs of the fracture surface are SEM secondary electron images.
Depending on the induction-hardened condition, clear differences in crack propagation mode were observed. Figure 5.15 shows macro-photographs from specimens tested at the highest stress level that are representative of the fracture behavior for both steels in each condition. Arrows indicate the point of fatigue failure initiation. In the Low and Med conditions, failures initiate in either Mode I or Mode II (1045-Low only), propagate during Stage II in Mode I, and transition to Mode II for a short distance in Stage III propagation before continuing around the entire case in Mode I. In the High and High-SS conditions, the transition between Mode I in Stage II and Mode II in Stage III occurs only in the 1045 specimens while the 10V45 specimens fail completely in Mode I.

Figure 5.15  Representative macro-photographs of fracture surfaces from (a) Low, (b) Med, (c) High, and (d) High-SS induction hardened conditions tested at 650 MPa shear stress amplitude. Arrows indicate location of surface initiation when identifiable.
In the *Low* and *Med* conditions, the induction hardening is very rapid, resulting in the finest austenite grain size of all the conditions. The short time at temperature likely results in an inhomogeneous austenite that maintains a very high hardness but has improved ductility due to regions of low carbon martensite. The higher ductility case increases the likeliness of Mode II crack propagation occurring during final failure. This observation is supported by higher magnification analysis of the Stage II. Figure 5.16a and 5.16b show representative SEM micrographs of Stage III crack propagation in the near surface case region of specimens from both steels in the *Low* condition. Although the 10V45 has a considerably higher case hardness than all other conditions, the case still shows considerable ductility in the form of ductile voids around regions of intergranular fracture.

![Representative SEM secondary electron micrographs showing Stage III fast fracture in the high hardness region of the induction hardened case in (a) 1045-Low, (b) 10V45-Low, (c) 1045-High-SS, and (d) 10V45-High-SS conditions.](image)

In the *High* and *High-SS* conditions, the 1045 specimens were austenitized either at a higher temperature or for longer time to produce significantly larger austenite grain sizes. These process conditions produce a more
homogenous austenite, which transforms to a more consistently high hardness martensite upon quenching, with the 1045 having a lower hardness than the 10V45 steel due to slight differences in carbon content. Figure 5.16c and 5.16d show representative SEM micrographs of Stage III crack propagation in the near surface case region of specimens from both steels in the High-SS condition. The 10V45 steel in this condition appears to exhibit a higher amount of intergranular fracture than the 1045; however, it still has some small regions of ductile voids.

5.2.3 Empirical Model

Shear stress amplitudes used in the present study were determined using an empirical model developed from literature (see Section 3.2.8). Figure 5.17 shows the literature torsional fatigue data and fitted linear regression model with the data for the present study overlaid. Both 95 pct confidence and prediction intervals are shown for the upper and lower bounds of the carbon content (0.43 and 0.48 wt pct C) and normalized effective case depths (0.25 and 0.45 r/t) from the present study. All of the data from the present study fall within the prediction interval from the model; however, the model appears to overpredict the cycles to failure for a given shear stress amplitude. This overprediction may be an indication of a mean stress effect since the torsional fatigue data from literature were collected using a fully reversed stress ratio (R=-1). The high shear stresses used in the present study were near, or above, the estimated torsional yield strength. These high stresses may be the source of possible mean stress effects suggested in literature (see Section 2.3).

![Figure 5.17 Empirical model developed from literature to predict torsional fatigue performance as a function of carbon content and normalized effective case depth for induction hardened plain carbon steels.](image-url)
The model was revised to incorporate the data from the present study. Figure 5.18 shows the revised model with the present data overlaid along with 95 pct confidence and prediction intervals. Data from 274 torsional fatigue tests resulted in the following multiple linear regression model

\[
\text{Shear Stress Amplitude (MPa)} = 956.295 - 177.240 \times \log_{10} N_f + 741.483 \times C \ (\text{wt ppt}) + 532.482 \times \frac{t}{r} \tag{3.1}
\]

where \(N_f\) is the number of cycles to failure between \(10^2\) and \(3 \times 10^6\) cycles, \(C\) is the carbon content of the steel in wt ppt between 0.35 and 0.56 wt ppt, and \(t/r\) is the normalized effective case depth between 0.21 and 0.68. All variables were found to be significant to the regression above the 99 pct level. Standardized residuals as well as the predicted output are normally distributed and the adjusted coefficient of determination is 0.8405. All of the coefficients of the revised model changed. The regression model constant and cycles to failure coefficient decreased by approximately 250 and 8 respectively, while the coefficients for the carbon composition and normalized effective case depth increased by 275 and 87 respectively. These large increases in the coefficient indicate a greater sensitivity to these factors as compared to the original model.

Fig. 5.18 Revised empirical model relating stress amplitude, cycles to failure, normalized effective case depth, and carbon content for plain carbon steels. Model incorporates data from literature as well as the present study. Confidence (CI) and prediction (PI) limits were calculated using the carbon content and case depth ranges from this study, 0.43-0.48 wt ppt C and 0.25-0.45 t/r respectively.
The purpose of this project was to evaluate the influence of vanadium microalloying on the microstructure and torsional fatigue performance of induction hardened medium-carbon steel in the ferrite-perlite starting condition. In particular, the objectives were to assess vanadium effects on case and case/core transition microstructure, residual stresses, and torsional fatigue behavior.

Two industrial heats of 1045 – one with vanadium microalloying (0.08 wt pct) and one without – were produced for this study with low sulfur levels (0.006-0.009 wt pct) to reduce inclusion content. Torsional fatigue specimens designed specifically for this research were machined from the as-received, hot rolled bars. Torsional fatigue specimens were induction hardened using both scanning (96 kHz/72 kW) and single-shot (31 kHz/128 kW) techniques in collaboration with Inductoheat® Inc. Three nominal effective case depths of 25, 32, and 44 pct were produced using the scanning coil setup, with the 44 pct nominal effective case depth also being produced using the single-shot coil setup. Induction hardened specimens were fatigue tested in torsion at shear stress amplitudes of 550, 600, and 650 MPa and a stress ratio of 0.1. Thermal profiles of the surface and total case depths from the 25 and 44 pct effective case depth conditions were simulated in collaboration with Fluxtrol® Inc. using electro-thermal finite element analysis. Computer simulated thermal profiles were physically simulated using the Gleeble® 3500 at CSM. Physical simulations were extended to include idealized induction heating cycles (i.e. continuous heating) examining heating rate dependence as well as modified induction heating cycles to examine quench rate dependence. Light optical, scanning electron, and transmission electron microscopy were used to characterize microstructure. Vickers microhardness and Rockwell macrohardness testing was used to characterize hardness. Torsional fatigue testing was conducted under load control on an average of five specimens at each shear stress amplitude for each condition. Fractography was conducted using macro-photography and scanning electron microscopy. Dilatometry was used to characterize critical temperatures and phase transformation behavior during Gleeble® simulations. Residual stresses were quantified via incremental hole drilling of strain gaged torsional fatigue specimens. Major conclusions developed from this research are:

1) The vanadium microalloyed steel in this research exhibited significant finer ferrite grain size, smaller ferrite area fraction, and pearlite interlamellar spacing in the as-received, hot rolled condition than the steel without vanadium. The difference in microstructure is attributed to the slight difference in chemical composition, other than strictly vanadium, between the two steels (i.e. carbon and manganese). Average vanadium precipitate diameter was quantified to be 4.2-4.8 nm in proeutectoid ferrite and 3.0-3.6 nm in pearlitic ferrite.

2) Tensile properties of the two materials in the as-received, hot rolled composition showed the vanadium steel had a 33 pct (187 MPa) higher yield and 20 pct (173 MPa) higher ultimate tensile strength than the steel without vanadium. The maximum precipitation strengthening potential was calculated to result in a
30 pct (200 MPa) increase in yield strength; however, only a 20 pct (114 MPa) increase due to precipitation strengthening was observed.

3) Continuous heating experiments conducted over the heating rate range of 100-1000°C/s showed the Ac₁ and Ac₃ critical temperatures increase significantly over calculated values in literature, ~50 and ~100 °C respectively. Differences in the as-received, hot rolled microstructure did not significantly affect the Ac₁ and Ac₃ critical temperatures.

4) Vanadium microalloying was observed to result in an induction hardened case microstructure with more observed non-martensitic transformation products; however, it was not observed to influence hardness significantly. Case hardness of the vanadium microalloying steel was significantly higher than without the vanadium.

5) Martensite hardness was found to be quench rate dependent at cooling rates of 350-3500 °C/s for the steel without vanadium microalloying as well as the vanadium microalloyed steel without all of the vanadium precipitates in solution. The nominal increase in hardness with increasing heating rate was 1 HRC. With all of the vanadium in solution, the hardness of the martensite was not quench rate dependent.

6) Vanadium microalloying reduced the total case depth significantly at the higher case depth examined. Gleeble® simulations of the calculated thermal profile for the total case depth showed ~1 pct austenite in the vanadium microalloyed steel while ~13 pct austenite formed in the steel without vanadium.

7) Residual stresses in the induction-hardened case were primarily influenced by processing route and not significantly affected by vanadium microalloying. The scan hardened 25 pct effective case depth condition and the single-shot hardened 44 pct effective case depth condition exhibited the highest compressive residual stresses resulting in a 39-46 pct reduction of the maximum principal stress during fatigue testing. The residual stresses in the other two conditions were much lower, reducing the maximum principal stress during fatigue testing by 28-34 pct.

8) Vanadium microalloying did not significantly influence the torsional fatigue life of conditions with effective case depths greater than 25 pct. The 25 pct effective case depth condition with vanadium microalloying showed ~75 pct increase in fatigue life at all shear stress amplitudes when compared to the same case depth without vanadium. The improved fatigue performance is likely due to the significantly higher case hardness than any other condition.

9) Crack initiation and propagation behavior was not significantly influenced with vanadium microalloying. Unstable crack propagation behavior was a function of case hardness and austenite grain size. Fracture
surfaces exhibited increased ductility in the unstable crack propagation region of the case for all hardnesses if the austenite grain size was relatively fine (~10 µm). The amount of intergranular fracture in the case increased with increasing austenite grain size. Above an austenite grain size of ~20 µm, the macroscopic unstable crack propagation mode transitioned from Mode III to Mode I with increasing case hardness.

10) An empirical multiple linear regression model was developed from the current study as well as data from literature relating shear stress amplitude to nominal effective case depth, carbon content, and cycles to failure. Data from 274 torsional fatigue tests were used to develop the model with an adjusted coefficient of determination 0.8405. Data used in the model range in carbon content between 0.35 and 0.56 wt pct, nominal effective case depth between 0.21 and 0.68 t/r (case depth/total radius), and cycles to failure between 100 and 3,000,000.
CHAPTER 7
FUTURE WORK

In order to fully assess the influence of vanadium microalloying on the microstructure and properties of induction hardened medium-carbon steels additional studies should be conducted. Listed below are the areas of potential interest based on data and observations from the current study.

1) In induction-hardened shafts, the majority of the load is carried by the case. Therefore, understanding the influence of vanadium on the strength and hardness of martensite is of critical importance. The current study has shown a cooling rate dependence on the hardness of martensite that is not fully understood. Carefully designed specimens and Gleeble® simulations coupled with characterization techniques such as electron backscatter diffraction (EBSD) and static torsion testing may provide additional insight.

2) Extension of the study proposed in (1) to torsional fatigue performance of specimens produced via Gleeble® simulations could be very insightful. This work would be a fundamental study into the mechanisms of fatigue damage in high hardness martensite with and without vanadium precipitates present.

3) The current study has identified two specific circumstances for which vanadium has an observed influence on microstructure. Vanadium microalloying appears to increase the amount of non-martensitic transformation constituents in the case while there is less austenite formation at the total case depth. One hypothesis is that vanadium precipitates are affecting both the nucleation of austenite during heating as well as the formation of non-martensitic transformation products during cooling. Although the present study did not show these phenomena to have a significant impact on torsional fatigue performance, a characterization study into the mechanisms causing these observed differences in microstructure would provide useful results from a fundamental perspective.
REFERENCES


[78] Edison Welding Institute, “TTT and CCT Prediction,” calculations.ewi.org/vjp/secure/TTTCCTPlots.asp


Careful consideration has been given to the design of the torsional fatigue specimen used in this research. The following sections discuss specimen design criteria, specimen design validation, as well as uncertainty analysis.

A.1 Meeting Specimen Design Criteria

An optimal design was developed utilizing specific design criteria in conjunction with machine and fixture limitations of the SF-1U universal fatigue tester. The three design criteria were as follows:

1) Specimen failure location should be material controlled not geometry controlled.
2) Specimen should be capable of achieving > 839 MPa (121.7 ksi) in shear.
3) Specimen should be as long as possible to allow the greatest surface area possible to be within 95 pct of maximum shear stress.

The first and third criteria are closely related and were established so that failures due to stress concentrations could be eliminated as well as allow the largest amount of material possible to be tested. The second criterion was developed from data published by Fett [62] for the maximum torsional yield strength of a hot-rolled 1050 (modified) bar with 18.4 pct of the diameter induction hardened to 40 Rockwell C (HRC). For the design calculations conducted as part of this research, 850 MPa (123 ksi) was used.

Machine and fixture limitations, with no factor of safety [77], are as follows:

1) Maximum applied load is 4448 N (1000 lbf) at a stress ratio of -1 (R = -1, fully reversed).
2) Maximum amplitude of loading mechanism is 12.7 mm (0.5 in).
3) Lever arm lengths are limited to either 158.75 mm (6.25 in) or 387.35 mm (15.25 in).
4) Specimen must be < 38.1 mm (1.5 in) diameter with cylindrical ends.

Since the machine and fixture limits cannot be relaxed, a factor of safety was incorporated into the design calculations limiting the maximum applied load to 4226 N (950 lbf) and the maximum amplitude to 11.4 mm (0.45 in). Figure A.1 shows the torsional fatigue testing fixture mounted to an SF-1U universal fatigue tester. The fixture holds one end of the specimen stationary while the opposite end is free to rotate when loaded.

The first criterion, specifying material controlled failure, was satisfied by evaluating four candidate torsional fatigue specimen geometries. Additional geometries were hypothesized but were not evaluated due to their complexity. Figure A.2a shows a “simple fillet” design which has a constant cross-sectional area in the reduced section and a relatively small radius fillet that joins the reduced section to the grip section of the specimen. The simple fillet design is commonly used in static torsional overload testing as well as fatigue testing that is not limited by the specimen’s angle of twist (i.e. hydraulic torsional fatigue frames). Figure A.2b shows a “compound fillet” design which has a constant cross-sectional area for the majority of the reduced section as well; however, two different radii fillets are used to join the reduced section to the grip section. The compound fillet design is used in torsional loaded components that have length restrictions [93, 94]. Figure A.2c shows a shallow U-shaped groove with fillet design which has a constantly varying cross-sectional area in the reduced section. The U-shaped groove has a radius typically an order of magnitude larger than the minimum diameter of the component [94] and connects...
to the grip section via a relatively small radii fillet. Figure A.2d shows the simple shallow U-shaped groove design which has a constant varying cross-sectional area from the midline, throughout the reduced section, to the grip section of the specimen. This design is common in fatigue testing because the geometry of the specimen can be modified such that the stress increase due to the groove can be reduced to less than 1 pct (i.e. the stress intensity factor approaches unity) by meeting two criteria [94]:

1) The ratio of the U-shaped groove radius over the minimum diameter is greater than 10 ($R/D_{min} > 10$).
2) The ratio of the maximum diameter over the minimum diameter is greater than 1.10 ($D_{max}/D_{min} > 1.10$).

Figure A.2 Schematics of four candidate geometries for torsional fatigue specimens: a) simple filet, b) compound filet, c) shallow U-shaped groove with fillet, and d) shallow U-shaped groove.

Linear elastic isotropic finite element analysis (LEI-FEA) using SolidWorks® Simulation was conducted to compare the four selected specimen geometries. Yield strength, 1270 MPa (184.2 ksi), and Young’s modulus, 207 GPa (30,023 ksi), data from Nissan [95] for a simulated induction hardened SAE 1045 case microstructure were
used in conjunction with an assumed Poisson’s ratio of 0.29 [63] for material property data. Applied torque and minimum diameter were held constant to facilitate an accurate comparison and all fillet radii were incrementally modified to reduce stress concentration effects while maintaining a constant reduced section length of 63.5 mm (2.5 in). Figure A.3 shows the results of the LEI-FEA for the four candidate specimen geometries. The stress increase due to the fillets in the simple and compound fillet designs eliminate them as viable geometries for the current study according to the first design criterion. The fillet in the shallow U-shaped groove geometry does prove to be a significant stress concentrator; however, the introduction of the fillet offers little to no benefit over the simple shallow U-shaped groove design. Both shallow U-shaped groove designs decrease in stress from the midline of the specimen with similar slopes while the design without the fillet would be easier to machine. Therefore, the shallow U-shaped groove design was determined to be the geometry of choice for this research.

Figure A.3 Comparison of linear elastic isotropic finite element modeling data (SolidWorks® Simulation) for maximum shear stress as a function of distance from the midline for the four candidate specimen geometries.

Since a specimen geometry was determined, the second design criterion, minimum shear stress of 839 MPa (121.7 ksi), is determined analytically. The equation for maximum shear stress is

\[
\tau_{\text{max}} = \frac{T d}{J} \tag{A.1}
\]

where \(\tau_{\text{max}}\) is the maximum shear stress in MPa, \(T\) is the applied torque in N·mm, \(d\) is the minimum diameter of the specimen in mm, and \(J\) is the polar moment of inertia for a circular cross-section in mm⁴. Torque is simply the cross product of applied load, \(P\), in N and lever arm length, \(Y\), in mm

\[
T = P \times Y \tag{A.2}
\]
When the applied load and lever arm are oriented 90° from each other, the sine of the angle between them is unity and the applied load and lever arm length can be multiplied to give a value for torque. Combining Equations A.1 and A.2 with the polar moment of inertia

$$j = \frac{\pi d^4}{32}$$

(A.3)

results in an equation for maximum shear stress

$$\tau_{max} = \frac{16PY}{\pi d^3}$$

(A.4)

that can be rearranged to solve for the minimum specimen diameter as a function of three constants: maximum shear stress, applied load, and lever arm length [63].

$$d = \sqrt[3]{\frac{16PY}{\pi \tau_{max}}}$$

(A.5)

Figure A.4 shows specimen diameter as a function of applied load (Equation A.5) for a given maximum shear stress, 850 MPa (123 ksi), and given lever arm length. At an applied load of 4226 N (950 lbf) the maximum specimen diameter that will achieve the second design criterion is 15.90 mm (0.626 in) for the 158.75 mm (6.25 in) lever arm and 21.40 mm (0.843 in) for the 387.35 mm (15.25 in) lever arm. Although larger diameters can be used with the longer lever arm, and hence more material, both the final design criterion as well as the machine and fixture limitations still need to be fully taken into account.

![Figure A.4 Specimen diameter as a function of applied load for both the 158.75 mm (6.25 in) and 387.35 mm (15.25 in) lever arms at a constant maximum shear stress of 850 MPa (123 ksi).](image-url)
Two critical aspects of this specimen design, maximum specimen length and maximum amplitude of the loading mechanism, can be determined by solving for the angle of twist the specimen will undergo at maximum applied load. Figure A.5 shows the longitudinal and transverse cross-sections of the torsional fatigue specimen along with the reference frame used in the angle of twist calculation. Defined within the schematic are the radial direction \( r \), the axial direction \( x \), the minimum radial distance \( r_{\text{min}} \), the maximum radial distance \( r_{\text{max}} \), the depth of the shallow U-shaped groove \( h \), the width of the shallow U-shaped groove \( c \), the radius of curvature \( R \), the arc of curvature \( \theta \), the lever arm length \( Y \) in mm, specimen angle of twist \( \phi \) in radians, and amplitude of loading mechanism within the SF-1U fatigue machine \( a \) in mm. All variables are in units of mm except the arc of curvature and angle of twist, which are in radians. The following equations are the mathematical form of the aforementioned variables.

![Schematic indicating the reference frame for the torsional fatigue specimen design calculations where (a) is the longitudinal cross-section and (b) is the transverse cross-section of the specimen.](image)

The two-dimensional polar coordinate system is defined by

\[
x = R \cos \theta + \frac{c}{2}
\]  
(A.6)

\[
r = R \sin \theta + R + r_{\text{min}}
\]  
(A.7)

As seen in the schematic above, the cylindrical coordinate system has been shifted to the central axis of the specimen as well as to one side of the reduced section. The reasoning for this will become apparent later on when angle of twist is calculated. The depth of the shallow U-shaped groove is

\[
h = r_{\text{max}} - r_{\text{min}}
\]  
(A.8)

with the width of the shallow U-shaped groove
\[ c = 2\sqrt{h(2R - h)} \]  

(A.9)

and the arc of curvature

\[ \theta = \cos^{-1}\left(1 - \frac{h}{R}\right) \]  

(A.10)

Figure A.6 shows a plot of Equations A.6 and A.7 where the range in \(x\) and \(r\) are limited by fixing the radius of curvature, maximum radius, and minimum radius at 215.90 mm (8.50 in), 10.32 mm (0.4063 in), and 7.94 mm (0.3125 in) respectively.

The angle of twist, \(\varphi\), is calculated from

\[ \varphi = \frac{Tc}{GJ} \]  

(A.11)

where the \(G\) is the shear modulus of the material, 80.2 GPa (11637 ksi). In this specific case the geometry is such that the polar moment of inertia varies as a function of axial distance \(J_x\)

\[ \varphi = \frac{T}{G} \int_0^c dx \]  

(A.12)

and must be integrated from zero to \(c\) [63]. First, an expression for the polar moment of inertia must be determined as a function of axial distance by combining Equations A.3 and A.7 to give

\[ J_x = \frac{\pi}{2}(R\sin\theta + R + r_{\text{min}})^4 \]  

(A.13)
The derivative of Equation A.6 with respect to the arc of curvature results in

\[ dx = (-R \cos \theta) d\theta \]  

(A.14)

Combining Equations A.12, A.13, and A.14 results in an equation that defines the angle of twist as a function of the radius of curvature, maximum radius, and minimum radius.

\[ \varphi = \frac{2T}{\pi G} \int_{\cos^{-1}(\frac{\epsilon}{2R})}^{\cos^{-1}(\frac{\epsilon}{2R})} \left( \frac{-R \cos \theta}{(R \sin \theta + R + r_{\min})^4} \right) d\theta \]  

(A.15)

The limits of integration are converted to the minimum and maximum arc of curvature. From the schematic provided in [63] the angle of twist can be used to calculate the amplitude of the loading mechanism.

\[ a = Y \sin \varphi \]  

(A.16)

Mathematica® was utilized to iteratively solve the series of equations presented above to develop optimal specimen geometries for both lever arm lengths that satisfy all of the machine and fixture limitations. Figure A.7 shows the two specimens optimized for the present research. Taking into account the third design criterion, the specimen for the 158.75 mm (6.25 in) lever arm, Figure A.7a, was selected as the geometry to be used in this research. The specimen in Figure A.7a has both a reasonably large minimum diameter as well as a reduced section that is twice the length of the specimen in Figure A.7b.

![Figure A.7 Torsional fatigue specimen geometries designed for (a) the 158.75 mm (6.25 in) lever arm and (b) the 387.35 mm (15.25 in) lever arm.](image)

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Figure A.8 shows LEI-FEA of the torsional fatigue specimen optimized for the 158.75 mm (6.25 in) lever arm. Highlighted in the plot are regions that are greater than 90 and 95 pct of the maximum shear stress. The 95 pct region is approximately the same width as the specimen diameter. Figure A.9 shows the final technical drawing of the specimen that was used in this research.

Figure A.8  Linear elastic isotropic finite element modeling data (SolidWorks® Simulation) for maximum shear stress as a function of distance from the midline of the torsional fatigue specimen designed for the 158.75 mm (6.25 in) lever arm.

Figure A.9  Detailed technical drawing of the torsional fatigue specimen designed for the current study.
A.2 Specimen Design Validation

Specimens were machined from hot-rolled SAE 1045, hardened, polished, and fatigue tested to validate the specimen geometry calculations as well at the three design criteria. Figure A.10 shows photographs of two trial specimens. Material was taken in the as hot-rolled condition and machined according to the drawing provided in Figure A.9. The specimens were then austenitized at 845 °C (1553 °F) for 45 min, water quenched to room temperature, and tempered at 177 °C (350 °F) for 90 min. Hardness was measured on both specimens in the grip section of the specimen to be approximately 58 ± 1 HRC. Specimens were polished on a lathe to a surface finish of R\textsubscript{a} < 0.1 µm (< 4 µin), which is difficult for a one-armed man to accomplish [96-98], and tested at a fully reversed stress amplitude (R=−1) until failure. Temperature of each specimen was monitored during testing. Heating was only observed during the highest stress amplitude test, 650 MPa (94.3 ksi), and it was easily controlled to 26 °C (79 °F) by compressed air. Both specimens were tested at relatively high stress amplitudes to better evaluate specimen performance. Figure A.10a shows the specimen that failed at 600 MPa (87.0 ksi) after 338,000 cycles while Figure A.10b shows the specimen that failed at 650 MPa (94.3 ksi) after 22,000 cycles. A tensile fracture mode (i.e. Mode I) was observed in both specimens resulting in characteristic helical type fractures in torsional loading.

Figures A.11a and A.11b show macroscopic photographs of the two through-hardened SAE 1045 trial specimens tested at 600 MPa (87.0 ksi) for 338,000 cycles and 650 MPa (94.3 ksi) for 22,000 cycles, respectively. Arrows in the two photographs indicate both possible locations of crack initiation as well as direction of crack propagation. Scanning electron microscopy (SEM) was attempted with little success due to smearing of the fracture surfaces. Although the 600 MPa (87.0 ksi) specimen shows beach marks that clearly indicate surface initiated failure, the 650 MPa (94.3 ksi) specimen is too damaged to clearly identify the initiation site. Both specimens show a significant amount of shearing from the two fracture surfaces impacting each other while the fatigue machine is turning off after a failure is detected (i.e. engagement of displacement limit switches). This issue can be avoided by testing at a positive a stress ratio rather than fully reversed.

Figure A.10 Macroscopic view of two through-hardened SAE 1045 torsional fatigue specimens exhibiting Mode I fracture after (a) 338,000 cycles at 600 MPa (87.0 ksi) and (b) 22,000 cycles at 650 MPa (94.3 ksi).
A.3 Uncertainty in Applied Shear Stress

Uncertainty in the applied shear stress is a function of machine calibration, resolution of the applied load, accuracy of the measured lever arm length, and accuracy of the measured specimen minimum diameter. Machine calibration was limited to the low applied loads and extrapolated to the higher applied loads, due to strain gage limitations. Since the machine calibration must be physically changed by adding or removing weights it does not changed during the course of testing and is therefore assumed to be constant for a given load. As a result, uncertainty in machine calibration is omitted from this analysis. The other three primary contributors to the uncertainty in applied shear stress are related by Equation A.4. The propagation of uncertainty can be assessed by calculating the Pythagorean summation of the discrete uncertainties [99] of load, lever arm length, and specimen minimum diameter using the equation

\[ u_{\tau_{\text{max}}} = \sqrt{\left( u_p \frac{\partial \tau_{\text{max}}}{\partial P} \right)^2 + \left( u_y \frac{\partial \tau_{\text{max}}}{\partial Y} \right)^2 + \left( u_{d_{\text{min}}} \frac{\partial \tau_{\text{max}}}{\partial d_{\text{min}}} \right)^2} \]  

(A.1)
where \( u_{\tau_{\text{max}}} \) is the uncertainty in maximum applied shear stress, which is calculated. The uncertainty in applied load, \( u_p \), is 4.448 N (1 lbf) which is one-half the minimum load increment. The uncertainty in lever arm, \( u_Y \), is given a reasonable approximation 0.10 mm (0.004 in) due to the difficulty in making a precise measurement. The uncertainty in the specimen minimum diameter, \( u_d_{\text{min}} \), is 0.025 mm (0.001 in) which is one-half the minimum increment of the calipers used to make the measurement. All other variables were previously defined with Equation A.4. The three partial derivatives of the maximum applied shear stress for use in Equation A.17 are:

\[
\frac{\partial \tau_{\text{max}}}{\partial P} = \left( \frac{16}{\pi} \right) \left( \frac{Y}{d_{\text{min}}^{3/2}} \right) \tag{A.18}
\]

which is with respect to load,

\[
\frac{\partial \tau_{\text{max}}}{\partial Y} = \left( \frac{16}{\pi} \right) \left( \frac{P}{d_{\text{min}}^{3/4}} \right) \tag{A.19}
\]

which is with respect to lever arm length, and

\[
\frac{\partial \tau_{\text{max}}}{\partial d_{\text{min}}} = \left( \frac{16}{\pi} \right) \left( -\frac{3PY}{d_{\text{min}}^{7/4}} \right) \tag{A.20}
\]

which is with respect to specimen minimum diameter. Figure A.12 shows the uncertainty in maximum applied shear stress as a function of applied shear stress for both lever arms at a specimen minimum diameter of 15.88 mm (0.625 in). The uncertainty is relatively low at all maximum shear stresses achievable. In both lever arm lengths the uncertainty is highest at low applied shear stresses and becomes nearly asymptotic to 0.5 pct at high applied shear stresses.

![Figure A.12](image-url)  

Figure A.12  Uncertainty in maximum shear stress for a 15.88 mm (0.625 in) specimen minimum diameter for 158.75 mm (6.25 in) and 387.35 mm (15.25 in) lever arm lengths as a function of maximum shear stress.
Figure A.13 illustrates the different contributions of applied load, measurement of lever arm length, and measurement of specimen minimum diameter to the overall uncertainty in maximum shear stress. The plot was generated using the specimen minimum diameter of 15.88 mm (0.625 in) and lever arm length of 387.35 mm (15.25 in). At low maximum shear stress, the overall uncertainty is predominantly due to the uncertainty in applied load while at high maximum shear stress, the major contributor shifts to measurement of the specimen minimum diameter.

Figure A.13 Contributions of individual components of uncertainty in maximum shear stress as a function of maximum shear stress for a specimen minimum diameter of 15.88 mm (0.625 in) and a lever arm length of 158.75 mm (6.25 in).
APPENDIX B
INDUCTION HARDENING PROGRAMS

This appendix provides the induction hardening programs used at Inductoheat® Inc. to produce the conditions tested in the current study.

Table B.1 – Induction Hardening Recipe for Low (0.25 t/r) Condition Using a Power Supply with a Maximum Power of 100 kW and Frequency of 200 kHz with a UCON A Quench of 6 pct at 76 L/min

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<th>Position (mm)</th>
<th>Power (pct)</th>
<th>Dwell (s)</th>
<th>Speed (mm/s)</th>
<th>Rotation (pct)</th>
<th>Quench</th>
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Table B.2 – Induction Hardening Recipe for Med (0.32 t/r) Condition Using a Power Supply with a Maximum Power of 100 kW and Frequency of 200 kHz with a UCON A Quench of 12 pct at 76 L/min

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<th>Speed (mm/s)</th>
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Table B.3 – Induction Hardening Recipe for High (0.44 t/r) Condition Using a Power Supply with a Maximum Power of 100 kW and Frequency of 200 kHz with a UCON A Quench of 6 pct at 76 L/min

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<th>Rotation (pct)</th>
<th>Quench</th>
<th>Aux</th>
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Table B.4 – Induction Hardening Recipe for High-SS (0.44 t/r) Condition Using a Power Supply with a Maximum Power of 150 kW and Frequency of 30 kHz with a UCON A Quench of 2 pct at 144 L/min

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</table>
This appendix provides Vickers microhardness traverse data for the plane of maximum shear stress and 95 pct of maximum shear stress for all conditions. Each figure shows the data and an approximate spline fit along with the effective case depth hardness of 450 HV (45 HRC).

Figure C.1 Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the Low condition at the plane of maximum shear stress.

Figure C.2 Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the Low condition at the plane of 95 pct maximum shear stress.
Figure C.3  Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the Med condition at the plane of maximum shear stress.

Figure C.4  Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the Med condition at the plane of 95 pct maximum shear stress.
Figure C.5  Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the High condition at the plane of maximum shear stress.

Figure C.6  Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the High condition at the plane of 95 pct maximum shear stress.
Figure C.7  Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the High-SS condition at the plane of maximum shear stress.

Figure C.8  Radial Vickers microhardness profile for (a) 1045 and (b) 10V45 in the High-SS condition at the plane of 95 pct maximum shear stress.
Figure D.1 Macroetched transverse cross-sections at plane of maximum shear stress (Specimen A of sectioning plan) of induction hardened torsional fatigue specimens. Alloy is constant by column and induction hardened condition is constant by row. Etched with 4 pct nital.
Figure D.2 Macroetched transvers cross-sections at plane of 95 pct maximum shear stress (Specimen B of sectioning plan) of induction hardened torsional fatigue specimens. Alloy is constant by row and induction hardened condition is constant by column. Etched with 4 pct nital.
Figure D.3 Macroetched longitudinal cross-sections of induction hardened torsional fatigue specimens showing only the shoulder region (Specimens E and F of sectioning plan). Alloy is constant by column and induction hardened condition is constant by row. Etched with 4 pct nital.
Table D.1 – Summary of Total (Visual) Case Depth Data for All Induction Hardened Conditions Determined from Macroetched Cross-sections

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<th>Steel</th>
<th>Section From Cut Plan</th>
<th>Fraction of Maximum Shear Stress</th>
<th>Normalized Total (Visual) Case Depth, t/r</th>
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APPENDIX E

PRIOR AUSTENITE GRAIN SIZE MICROGRAPHS OF INDUCTION HARDENED CASE NEAR SURFACE

Figure E.1  Representative light optical micrographs showing the prior austenite grain structure of all conditions for both steels. Alloy is constant by column and induction hardened condition is constant by row. Micrographs taken at 200 times magnification with a picral based etchant.
This appendix provides residual stress profiles for each induction hardened condition separated by stress direction (axial, hoop, or shear) comparing the 1045 and 10V45 steels.

Figure F.1 Near surface residual stress profiles for Low induction hardened condition comparing 1045 to 10V45 in the (a) axial, (b) hoop, and (c) shear directions. All directions show differences between the steels; however, these differences are likely not significant. Uncertainty is standard error of the mean for two specimens.
Figure F.2  Near surface residual stress profiles for Med induction hardened condition comparing 1045 to 10V45 in the (a) axial, (b) hoop, and (c) shear directions. No difference is observed between the steels. Uncertainty is standard error of the mean for two specimens.
Figure F.3  Near surface residual stress profiles for *High* induction hardened condition comparing 1045 to 10V45 in the (a) axial, (b) hoop, and (c) shear directions. Axial and hoop directions show differences between the steels; however, these differences are likely not significant. Uncertainty is standard error of the mean for two specimens.
Figure F.4  Near surface residual stress profiles for High-SS induction hardened condition comparing 1045 to 10V45 in the (a) axial, (b) hoop, and (c) shear directions. Axial and hoop directions show differences between the steels; however, these differences are likely not significant. Uncertainty is standard error of the mean for two specimens.
This appendix provides torsional fatigue life data for all conditions tested. Cycles to failure and general initiation behavior are shown for each steel and induction hardened condition.

Figure G.1  Fatigue life as a function of shear stress amplitude for (a) 1045 and (b) 10V45 in the Low induction hardened condition. Surface and sub-surface failure initiation is indicated for each specimen tested.

Figure G.2  Fatigue life as a function of shear stress amplitude for (a) 1045 and (b) 10V45 in the Med induction hardened condition. Surface and sub-surface failure initiation is indicated for each specimen tested.
Figure G.3 Fatigue life as a function of shear stress amplitude for (a) 1045 and (b) 10V45 in the High induction hardened condition. Surface and sub-surface failure initiation is indicated for each specimen tested.

Figure G.4 Fatigue life as a function of shear stress amplitude for (a) 1045 and (b) 10V45 in the High-SS induction hardened condition. Surface and sub-surface failure initiation is indicated for each specimen tested.
This appendix is a repository of macro-fractographs for all torsional fatigue tests conducted during this study. Steel, induction hardened condition, and test shear stress amplitude are indicated above each group of images. Within each group, images are ordered from left-to-right then top-to-bottom from lowest to highest cycles to failure.

Figure H.1 Macro-fractographs of the 1045-Low condition tested at 650 MPa.
Figure H.2  Macro-fractographs of the 1045-Low condition tested at 600 MPa.
Figure H.3  Macro-fractographs of the 1045-Low condition tested at 550 MPa.
Figure H.4  Macro-fractographs of the 10V45-Low condition tested at 650 MPa.
Figure H.5  Macro-fractographs of the 10V45-Low condition tested at 600 MPa.
Figure H.6  Macro-fractographs of the 10V45-Low condition tested at 650 MPa.
Figure H.7  Macro-fractographs of the 1045-Med condition tested at 650 MPa.
Figure H.8  Macro-fractographs of the 1045-Med condition tested at 600 MPa.
Figure H.9  Macro-fractographs of the 1045-Med condition tested at 550 MPa.
Figure H.10 Macro-fractographs of the 10V45-Med condition tested at 650 MPa.
Figure H.11 Macro-fractographs of the 10V45-Med condition tested at 600 MPa.
Figure H.12  Macro-fractographs of the 10V45-Med condition tested at 550 MPa.
Figure H.13 Macro-fractographs of the 1045-High condition tested at 650 MPa.
Serial Number: 70
Cycles to Failure: 47,000

Serial Number: 61
Cycles to Failure: 52,000

Serial Number: 68
Cycles to Failure: 54,000

Serial Number: 63
Cycles to Failure: 63,400

Serial Number: 54
Cycles to Failure: 71,500

Figure H.14 Macro-fractographs of the 1045-High condition tested at 600 MPa.
Figure H.15 Macro-fractographs of the 1045-High condition tested at 550 MPa.
Figure H.16  Macro-fractographs of the 10V45-High condition tested at 650 MPa.
Figure H.17 Macro-fractographs of the 10V45-\textit{High} condition tested at 600 MPa.
Figure H.18 Macro-fractographs of the 10V45-High condition tested at 550 MPa.
Figure H.19 Macro-fractographs of the 1045-High-SS condition tested at 650 MPa.
Figure H.20 Macro-fractographs of the 1045-High-SS condition tested at 600 MPa.
Figure H.21 Macro-fractographs of the 1045-High-SS condition tested at 550 MPa.
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<th>Serial Number</th>
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</table>

Figure H.22  Macro-fractographs of the 10V45-High-SS condition tested at 650 MPa.
Figure H.23  Macro-fractographs of the 10V45-High-SS condition tested at 600 MPa.
Figure H.24 Macro-fractographs of the 10V45-High-SS condition tested at 550 MPa.
This appendix provides light optical micrographs of the low and high quench rate conditions, with and without a 30 s hold at 1070 °C, for the 1045 and 10V45 steels.

Figure I.1  Representative light optical micrographs showing martensitic microstructure in 1045 for (a) low quench rate (LQR) with no hold, (b) low quench rate with a 30 s hold, (c) high quench rate (HQR) with no hold, and (d) high quench rate (HQR) with a 30 s hold. Micrographs taken at 500 times magnification with a 2 pct nital etch.
Figure I.2  Representative light optical micrographs showing martensitic microstructure in 10V45 for (a) low quench rate (LQR) with no hold, (b) low quench rate with a 30 s hold, (c) high quench rate (HQR) with no hold, and (d) high quench rate (HQR) with a 30 s hold. Micrographs taken at 500 times magnification with a 2 pct nital etch.