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ABSTRACT

A series of simulations were performed on a 25.4 mm (1 in) diameter 254 mm (10 in) long cylindrical bar. These simulations included three carburization levels: non-carburized, carburized to 0.8 wt pct C and 1.0 wt pct C at the surface utilizing a plain carbon steel (1020) and three alloy steels (4120, 4320, and 8620) representing a range of hardenabilities. Both industrially standard oil quenching as well as high intensity quenching which has a heat transfer rate of 20 kW/(m² °C) were simulated.

After quenching, the non-carburized and oil quenched bars were predicted to have tensile residual hoop stresses at the surface while the carburized bars were predicted to have compressive residual hoop stresses. All carburization levels of 1020 were predicted to have compressive residual hoop stresses after quenching. After high intensity quenching, all four alloys at all three carburization levels were predicted to have compressive residual stresses at the surface.

It was shown that the high intensity quenching compressive residual hoop stresses at the surface were a result of the high heat transfer rate decreasing the temperature within fractions of a second resulting in a martensitic shell forming around a high temperature austenitic core. As the core cooled and thermally contracted, the shell was pulled inward to maintain coherency between the shell and the core. When the core austenite transformed, the volume expansion was insufficient to overcome the thermal contraction resulting in large compressive stresses at the surface and core, and a large tensile stress at the mid-radius. This profile was not found in literature. 1020 was found to transform to a mixture of ferrite and pearlite. As the core contracted and transformed, the volume expansion initially resulted in tensile hoop stresses near the surface. These tensile hoop stresses were decreased and became compressive due to the thermal contraction after transformation.

A critical heat transfer rate for each of the alloys was determined, above which the tensile residual hoop stresses were reversed to compression. This critical heat transfer rate was 3.2 kW/(m² °C), 9.0 kW/(m² °C), 8.9 kW/(m² °C), and 9.0 kW/(m² °C) for 1020, 4120, 4320, and 8620 respectively. This was generalized to a Biot number with a minimum of 2.5 needed to create compressive hoop stresses at the surface.
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CHAPTER 1 LITERATURE REVIEW

1.1 Introduction

Intensive quenching has been promoted widely to various industrial sectors as a way to improve parts. The claims are that the process, utilizing high velocity water as a quenchant, can create highly compressive residual stresses, reduce distortion, and lower alloy content or decrease carburization while maintaining surface hardness [1]. These advantages and the limitations of the process will be discussed further below.

In order to better understand the development of microstructure, component dimensions and internal stresses during the quenching process, heat treatment simulations will be used. From these simulations, the interaction between temperature, phase transformation, and mechanical response will be investigated in order to better understand what is occurring during the quench. These simulations of intensive quenching will be compared to an industrially standard oil quench. This thesis will explore the differences in material response (microstructure, distortion, and residual stress) between oil and intensive quenching through computer simulations for several common steel grades.

1.2 Heat Treatment and Quenching Methods

Heat treatment can be considered to be any elevation of the temperature of a part above ambient temperature. By this definition, there are many different types of heat heat treatment. Of main concern when considering distortion are those heat treatments that involve a phase transformation [2]. In steel, transformation from the room temperature stable body centered cubic (BCC) crystal structure to austenite which has a face centered cubic (FCC) crystal structure at high temperature involves a volume change. This transformation is reversible and upon cooling austenite transforms back to a microstructure containing an iron-rich BCC structure under equilibrium conditions [2].

The severity of cooling affects the kinetics of this transformation. As the cooling rate increases, metastable phases such as martensite are possible [2]. Each of the different
transformations is associated with a different volume change and results in variations in distortion. Figure 1.1 shows this interrelation between cooling rate, phase transformation and volume changes and stress schematically.

![Schematic representation of the interrelation between cooling rate, phase transformation, and volume changes and stress and the overall effects on the final mechanical properties of the part](image)

Figure 1.1 Schematic representation of the interrelation between cooling rate, phase transformation, and volume changes and stress and the overall effects on the final mechanical properties of the part [3].

The interrelationship can be made more complex through carburization. Carburization is a process where carbon is added to the surface of the part and allowed to diffuse into the part. This leads to a region near the surface of changing carbon level with the highest level at the surface and steadily decreasing to the bulk composition over a region called the case. The varying carbon level also means the transformation behavior varies. For example, if cooling is fast enough to form martensite, the martensite start temperature varies through the case.

The varying martensite start temperature means there are several possibilities for how the phase transformation occurs. Ebert investigated how the residual stress develops upon quenching of carburized bars [4]. Figure 1.2 shows the development of the microstructure and residual stresses when the case transforms prior to the core transformation during cooling. As the case transforms, the soft core is deformed to accommodate the volume change in the case. This places the core in tension and the case in compression. As the core then transforms, the
associated volume change is restricted by the case which is now at low temperature and thus very strong. This means that the core is in compression while the case is in tension when cooling is completed [4]. This situation usually leads to a part with lower than expected fatigue properties due to the tensile residual stresses that are retained near the surface of the bar.

![Figure 1.2](image)

Figure 1.2 Development of the microstructure and residual stress during continuous cooling (left) when the case transforms before the core of the bar. (a) Before cooling showing the carbon profile. (b) After case transformation. (c) Final results [4].

Figure 1.3 shows the opposite situation, in which the core transforms before the case. When the core transforms, the volume change due to the transformation leads to plastic deformation in the austenitic case. As the austenite is strained to accommodate the volume change in the core, tensile stresses are formed in the case. Upon the case transformation, a large volume change occurs that is constrained by the now stronger core. This leads to the core being in tension while the case is in compression [4]. This effect can lead to better fatigue properties.

During heat treatment, quenching can be the final step of the process. Typically, the final microstructure, distortion, and mechanical properties are all controlled by the quench. The microstructure and residual stresses can be modified somewhat if the parts are tempered after the quench. Currently, there are several different quenching methods that are used in industry. Though there are other media that could also be included, the focus here will be on liquid
quenchants such as water, oil, and intensive quenching. Each of these liquids has a different heat transfer rate that will affect the final microstructure, part distortion, and the residual stress of a part.

1.2.1 Water Quenching

Water quenching has been used for many years due to the low cost [5]. Depending on the temperature, water can be one of the most severe quench media available [5]. There are three stages during quenching a steel part into a liquid quenchant. During the first stage, a vapor blanket is formed and acts as an insulation to heat transfer leading to radiation heat transfer being the predominant form of heat transfer [5]. As the part cools, the vapor blanket surrounding the part breaks down leading to nucleate boiling on the surface and the second stage of the quench [5]. Under nucleate boiling, the part is cooling at such a rate that boiling occurs at the surface. The heat transfer rate increases dramatically compared to the first stage. During the final stage of the quench, there is not enough heat available in the part to sustain boiling in the quenchant
and heat is transferred via conduction and convection [5]. The length of the various stages can be affected by the starting temperature of the quenchant as well as the agitation level in the bath. If the bath is highly agitated, liquid flow will reduce the vapor blanket and the heat transfer rate will increase in the other two stages of cooling. As the temperature of the bath is increased, the heat transfer rate is slowed due to a lower temperature difference between the part and the quenchant [5].

The cooling properties of the water are greatly affected by the level of salts found in the water as shown in Figure 1.4. As the amount of salts is increased, as would be found in tap water compared to distilled water, the vapor blanket layer is reduced, thereby increasing the cooling rate [5].

![Figure 1.4](image)

Figure 1.4 Effect of hard water salts on the cooling rate of water. Cooling curves were obtained with a silver sphere. (a) Hard water, (b) Distilled water [5].

Figure 1.5 shows the effect of water temperature on maximum cooling rate, temperature at maximum cooling rate, cooling time, and cooling rate at 343 °C (650 °F). As can be seen, an increase in the water temperature decreased the cooling rate, the temperature at the maximum cooling rate, and cooling rate at 343 °C (650 °F) while the cooling time increased.

There are various additions that can be made to water that affect the cooling rate. Salt and various polymers can be added to the water. Polymers act to increase the viscosity of water and reduce the quench severity to between that of oil and pure water [5]. It was also shown that
salt additions can reduce the variability in the quench by reducing the vapor blanket breakdown variations that can increase distortion [5].

Figure 1.5  Analysis of cooling curve data obtained with a 12.5 x 100 mm (0.5 x 4 in) Inconel 600 probe. (a) Temperature at maximum cooling rate, (b) maximum cooling rate, (c) cooling time, (d) cooling rate at 343 °C (650 °F) [5].

1.2.2 Oil Quenching

One problem with water quenching is that the instantaneous stresses in the part can overcome the tensile strength of the steel during quenching, causing quench cracking. With the lower heat transfer rate and a lower quench severity and therefore lower probability of quench cracking, oil quenching has become an industry standard. Unlike water, oil can be heated to higher temperatures thereby reducing the quench severity and the potential for distortion.

Quench oils can come from a number of sources including petroleum, certain plants such as soy beans, or various animal fats such as from fish [5]. Each of these sources produces oils with different heat transfer characteristics so selection of the oil for the application is very important. There are also numerous types of oils that can be specified for various processes or different types of steels such as marquenching or accelerated oils [5]. These oils have typical bath temperatures higher than conventional oils. Where conventional oil is used up to 65 °C
(150 °F), accelerated oils are used up to 120 °C (250 °F) and marquenching oils up to 200 °C (400 °F) [5]. Figure 1.6 shows cooling curves for several oils along with cooling curves for fused salt and molten metal.

Figure 1.6 Comparison of cooling curves for various quench media [5].

Figure 1.7 shows a cooling curve comparison between conventional and accelerated quench oils and water. As can be seen, the water quench cools the part much quicker with a maximum cooling rate of 54.3 °C/s (97.7 °F/s) while the conventional oil had a maximum cooling rate of 31.6 °C/s (56.9 °F/s) and 34.6 °C/s (62.3 °F/s) for the accelerated oil.

One of the drawbacks of oils as a quenching media is their ability to burn. If an oil bath is overheated, such as when a very large part in quenched in a small volume of oil without cooling of the oil, the flash point of the oil can be exceeded leading to a fire. For conventional oils, the flash point is around 170 °C (340 °F) while accelerated oils have a flash point of 180 °C (355 °F) and marquenching oils have a much higher flash point at 300 °C (570 °F) [5].

As would be expected, the viscosity of the oil can also affect heat transfer. Higher viscosity oils decrease heat transfer rates due to slower bath movement leading to warmed oil remaining close to the part [5]. Viscosity also affects the amount of oil that is retained on the parts, often called dragout. As the viscosity increases, so does dragout, leading to higher oil losses and higher costs [5].
Another issue that needs to be controlled is oil degradation. Degradation can manifest itself in several ways. As the oil is used over and over for different batches, the chemistry of the oil gradually changes. This change in chemistry will usually result in a change in the viscosity of the oil, thereby changing the cooling characteristics [5]. Figure 1.8 shows how the viscosity of a martempering oil changes over time. These changing cooling characteristics can mean that as-quenched hardness does not meet the specification or an increase in distortion or number of cracked parts on quenching is seen [5]. As the oil is used, a sludge also typically forms. If large amounts of sludge are noticed, the oil is degrading. The life of the oil can be extended by refining the used oil. Figure 1.9 shows cooling rate profiles of new, used, and refined quenching oil. The refining of the used quenching oil returned the cooling properties to nearly those found in new quenching oil.

1.2.3 Intensive Quenching

As stated in the introduction, intensive quenching has been promoted as a way to improve mechanical properties for a wide range of part geometries. The patent for the process says it optimizes the cooling conditions and prevents film and nucleate boiling during quenching, relying on convection to cool the part, resulting in an optimum hardened layer and an increased strength level [6].
Kobasko identifies three different types of intensive quenching [1, 7-9]. The three types of intensive quenching are designated IQ-1, IQ-2 and IQ-3. IQ-1 utilizes a two-step cooling process for medium and high alloy steels. In the first step, the part is cooled to the Ms temperature slowly via hot oil or other slow cooling medium, resulting in a negligible temperature gradient through the part [7]. The second stage is then intensive quenching with high velocity water starting from the Ms temperature. As a result of the intensive quenching, the
residual stresses are compressive at the surface [7-9]. Kobasko notes several drawbacks to the IQ-1 process. The dual quench tanks needed for the initial oil quench and the second intensive quench add complexity and cost [7-9].

IQ-2 is a three-step process where the first step in the quench utilizes a water and salt mix to eliminate film boiling while allowing nucleate boiling to occur as the primary heat transfer mode. At the end of nucleate boiling, the parts are removed from the quench and allowed to air cool until the temperature equalizes in the part. This air cool allows the core to cool while effectively heating the surface martensite that has formed, thereby “self-tempering” the martensite [7]. The final step is to place the part back into the quench to finish cooling the part to the quenchant temperature and finalize the transformation. In order to control the process, the initial quench step can be adjusted by changing the salt concentration in the bath or by changing the pressure above the quench tank, thereby modifying the boiling point of the quenchant. This adjustment allows the user to control the amount of martensite that is formed at the surface during the quench. For instance, if the boiling temperature of the quenchant is the same temperature as that needed to form 50% martensite from the transformation diagram for the steel, then in the first stage, the surface will become 50% martensite with the balance austenite. The 50% surface martensite level is needed to prevent quench cracking as the austenite can deform much more easily to overcome tensile stresses than the martensite [9]. As the core is cooled, the surface martensite will be tempered as the core heat is transferred to the surface resulting in an increase in surface temperature. During the final quench, the austenite near the surface will transform to new un-tempered martensite [7]. According to Kobasko, the ideal quenchant boiling temperature would be the martensite finish temperature \(M_f\) as this would lead to a fully martensitic layer at the surface and higher compressive residual stresses at the surface [7].

The third type of intensive quenching, IQ-3, is a two-step quench where all film and nucleate boiling is avoided completely during intensive quenching followed by air cooling. This leads to very high heat transfer rates and what Kobasko calls the most intensive of the three methods. The quench gives uniform cooling to the entire surface of the part. The part is quenched until the compressive residual stresses reach their peak and are produced to an optimal depth [7]. This quench time is specified via a proprietary model that incorporates the various quench variables such as flow rate and part geometry [1]. There are several limitations on the IQ-3 process. The first is that the needed temperature gradients to produce the high compressive
residual stresses are difficult to obtain in small cross-sections; parts over 19 mm (0.75 in) in section are reportedly preferred. Next, complex geometries make it difficult to achieve a consistent heat transfer rate at the surface of the part. Finally, the process is typically done one part at a time as the quench uniformity decreases when multiple parts are quenched in a batch [9].

Kobasko showed that as the cooling rate increases, the probability of quench cracking increases with a peak around 375 °C/s (700 °F/s) and then falls again to very little probability of cracking at cooling rates of 600 °C/s (1100 °F/s) [10]. Liščić provided the tensile strength of AISI 5140 steel as a function of cooling rates over the same range [11]. As seen in Figure 1.10, the tensile strength decreases to a minimum near the same cooling rates that create the highest quench cracking probability. At higher cooling rates, the tensile strength is shown to increase. This change in tensile strength is attributed to changing residual stresses with compressive stresses at the surface and tensile stresses in the core [11]. As the cooling rate increases, the residual stresses also become more tensile. At cooling rates above 375 °C/s (700 °F/s), the residual tensile stresses decrease and become compressive [11].

![Figure 1.10](image)

Figure 1.10 Probability of producing quench cracking in parts as a function of cooling rate in the martensite formation temperature range (solid line). Also shown is the relationship for the room temperature tensile strength of AISI 5140 as a function of cooling rate in the same temperature range (dashed line) [11].
Kobasko further compared SAE 1045 steel cylinders that were quenched intensively and in oil [12]. A 6 mm (0.25 in) diameter bar was intensively quenched to room temperature. During the quench, the surface stresses in the bar were calculated to become very compressive and then decrease so that the final stresses were near zero (Figure 1.11a). The details related to the calculation procedure were not provided. The oil quench bar was 10 mm (0.40 in) in diameter and predicted to show a decrease in the residual stresses to compressive during the quench and then a reversal of the stresses back to tension and a final residual tensile stress around 200 MPa (29 ksi) (Figure 1.11b) [12]. Kobasko claims the residual stresses can be maintained at the highly compressive levels seen if the quench is interrupted when the surface stresses reach the peak and the part temperature is held at the martensite start ($M_S$) temperature [12]. This hold would stop the martensite transformation, prevent the core from transforming to martensite and expand, and thereby “lock in” the compressive stresses. Kobasko does claim that the actual residual stresses would be slightly less compressive than the peak due to stress relaxation during the hold but does not comment on the ensuing austenite phase transformation that will occur during further cooling of the core.

![Figure 1.11](image_url)  
Figure 1.11 Calculated instantaneous stresses of SAE 1045 quenched (a) intensively or (b) in oil. The intensively quenched bar had a diameter of 6 mm (0.25 in) while the oil quench bar had a diameter of 10 mm (0.40 in) [12].

Kobasko and others [1, 13-20] have investigated the use of intensive quenching for a number of steel parts of various sizes and shapes. Kobasko showed that truck half axles made
from 1045 steel and intensively quenched had better fatigue life than 4340 steel axles that were oil quenched [13]. This increase was attributed to the compressive residual stresses at the surface [13]. Aronov et al. investigated using intensive quenching on a five tooth sprocket used in mining equipment. This sprocket made from 86B30 steel was normally quenched in oil and was required to have a hardness of 46-50 HRC after tempering. The parts that were intensively quenched averaged 52 HRC [14]. Aronov further studied AISI S5 tool steel punches that were normally oil quenched and tempered. After oil quenching, the punches had a hardness of 62-63 HRC while the intensively quenched punches were 63-64 HRC. After tempering, both quenching processes produced a hardness of 60-61 HRC with the oil quenched samples having a lower standard deviation of 0.3 HRC versus a deviation of 0.6 HRC for the intensively quenched samples. Aronov also compared the residual stresses resulting from the two processes and showed that the intensively quenched punches had compressive axial and circumferential residual stresses while the oil quenched punches were found to have predominately tensile axial and circumferential residual stresses (Figure 1.12) [15]. Aronov further showed the measured distortion in the punches quenched intensively was in opposite directions compared to the oil quench (Figure 1.13) [15].

![Figure 1.12](image-url)  
(Mid-length (a) axial and (b) circumferential residual stresses as measured by x-ray diffraction with tensile stress being positive of AISI S5 tool steel punches, 38.1 mm (1.5 in) in diameter and 55.9 mm (2.2 in) long. [15].)
Figure 1.13  Distortion maps of AISI S5 tool steel quenched via (a) oil or (b) intensive quenching. Line length, arrowhead size and direction indicate the magnitude and direction of the measured distortion [15].
Aronov et al. further showed that in the core of intensively quenched 50 mm (2 in) diameter steel bars, yield and tensile strength, impact strength and hardness were increased, while elongation and reduction in area results were mixed for 5140, 4130, and 4118 steels [16]. Further they showed that core hardness and tensile strength increased in 4140 or 1340 steel bolts when intensively quenched compared to oil quenching. They also showed that the hardness and bend test strength in a 4140 king pin were greater and more consistent across the diameter for intensive quenching compared to induction hardening [16]. The increase in overall hardness is expected due to the change in processing. Whereas the induction hardening process only modifies the hardness at the surface, the intensive quenching effectively through-hardened the parts.

As a follow up to a previous study, Aronov et al. showed that the variation in hardness through an 86B30 sprocket tooth 25.4 mm (1 in) thick greatly decreased when intensively quenched compared to oil quenching [17]. The hardness in the oil quenched part ranged from 50 HRC at the surface to 35 HRC in the core while the intensive quenched part had hardness levels between 50 and 53 HRC (Figure 1.14) [17]. Further publications by the same group show similar results [18-20].

![Hardness profiles (HRC) of 86B30 steel sprockets. The maximum and minimum hardness after intensive quenching along with the historical hardness data are shown [17].](image)
Canale et al. [21] compared intensive quenching with a reduced carburization cycle time to a conventional carburization schedule and oil quenching for both a simple rectangular bar and automotive “crosses” shown in Figure 1.15 made from 8620 steel. The hardness profiles for the cross and the bars are shown in Figure 1.16. They reported that the carburization cycle time could successfully be reduced by 25% while maintaining hardness and increasing compressive residual stress by using intensive quenching. They also showed it was not possible to eliminate carburization for 8620 in this application due to the low carbon content and the surface hardness not meeting the minimum requirements.

Figure 1.15  Photograph of the automotive cross (left) and sketch of the square bar (right) used to investigate the effect of intensive quenching on the ability to reduce the carburization cycle while maintaining surface hardness and effective case depth. Dimensions for the crosses were not provided and the bars are dimensioned in mm [21].
In order to explain how the residual stresses are developed and maintained during the quenching process, Intensive Quenching Technologies provided the following explanation (emphasis maintained from original text) along with Figure 1.17 [1]:

Imagine a cylindrical steel part. To simplify a mechanism of the stress formation in the part, let us assume that the part consists of only two sections: a “surface layer” and a “core.” (It would be more accurate to consider the part as a series of concentric layers, like layers of an onion, where the heat and the phase transformation are “transferring” from layer to layer.) Now assume that the part’s “surface layer” consists of a set of “segments” joined together by “springs” to form an elastic “ring” (Figure 1.17). When the whole steel part is heated above the Ac3 temperature before quenching there is no tension in the “springs” and there are no stresses between the “segments” (σ=0, see Figure 1.17a). During quenching, the surface layer cools rapidly resulting in the contraction of the “elements.” To compensate for the contraction of the segments in the surface
layer during cooling, the “springs” expand simulating the development of tangential (hoop) tensile thermal stresses (see Figure 1.17b).

When the surface layer reaches the martensite formation start temperature, Ms, the austenite in the surface “segments” transforms into martensite (see Figure 1.17c). The martensite specific volume is greater than the austenite’s. This results in the expansion (swelling) of the surface layer “segments”, causing the “springs” to contract. The contraction of the springs illustrates the development of surface compressive hoop stresses.

It is important to note that during intensive quenching, the part surface layer reaches the martensite start temperature Ms so quickly that the part core is still very hot (practically at the initial austenitizing temperature). (This is in contrast to conventional quenching, for example marquenching, when the part core temperature may be just above the Ms temperature at this period of time.)

While the martensitic structure is forming in the part surface layer, the part’s austenitic core continues to cool down to the Ms temperature, shrinking in size as it cools (Figure 1.17d). We call this core thermal contraction “pre-phase transformation shrinkage.” As the core shrinks, the strong martensitic shell maintains the part’s initial size with low distortion – almost as though a “die” has been built on the outer shell of the part. The shrinking (cooling) austenitic core draws the martensitic surface shell toward the part center increasing the surface hoop compressive stresses (with the “springs” between the surface layer “segments” contracting). Note that in a real quench the material does not “break” between the shrinking austenitic core and the fixed martensitic “shell” (as shown in Figure 1.17d). This is because the hot austenite is in a “super-plastic” state; when stresses between the “surface” and “core” sections of the part exceed the austenite yield strength, the austenite deforms to maintain part integrity within the shell.

If intensive quenching continues further, then within a short time (in a matter of seconds), the martensite starts forming in the part “core,” resulting in core swelling (see Figure 1.17e). The expanded part core pushes the part surface layer back from the part center resulting in diminution, but not elimination of the
high surface compressive stresses. (Put another way, the distance between the surface layer “segments” increase, resulting in the expansion of the “springs” and the lowering of the compression in the surface shell.) The surface residual stresses are still compressive even in a through hardened part because the size of the expanded, martensitic core is actually smaller than the size of the initial, hot austenitic core. In other words, the steel’s pre-phase transformation shrinkage (of the cooling austenitic core) offsets the following phase transformation expansion in the final martensitic core.

At some point in time, the surface compressive stresses reach their maximum value. It happens just before martensite starts forming in the core. The key element of the IntensiQuenchSM technique is to “interrupt” the rapid, uniform cooling of the part’s “shell” when compressive stresses in the part’s surface are at their maximum. The “interruption” is done by simply removing the part from the intensive quench. As the cooling rate of the part “shell” slows, the part “core” will also begin cooling more slowly and the martensite phase transformation advance may slow or cease entirely if the part is thick enough (over approximately one inch). If the martensite formation ceases, the remaining austenite in the core transforms into intermediate phases, such as bainite, ferrite, pearlite, etc. (See Figure 1.17f). Since this mixed “core” structure has less specific volume than a “pure” martensite core (as discussed above), the quench results in a higher level of surface residual compressive stresses compared to the through hardened version (see Figure 1.17e). The precise time for interruption is predicted by the IQ Technologies computer software model. Usually there is a window of several seconds to move from each stage of the intensive quench process; the thicker the part, the “bigger the window.”

Although this analysis provides a hypothesized explanation of how residual stresses may form during intensive quenching, no proof of this sequence of events is given either through experiments or simulation. Can this hypothesis be validated using heat treatment simulation techniques?
Figure 1.17 Surface stress conditions during intensive quenching [1]. (a) Uniform temperature before quenching begins with austenite above the $\text{Ac}_3$ temperature, (b) surface cooled with austenite above the $\text{M}_s$ temperature, (c) martensite formed on surface while core is still austenite above the $\text{Ac}_3$ temperature, (d) core has cooled to below the $\text{Ac}_3$, (e) core has transformed to martensite, (f) core has transformed to non-martensitic constituents [1].
1.3 **Heat Treatment Modeling**

As computer technology has advanced, increasing numbers of researchers are taking advantage of these advances to develop computer simulations of different processes. A number of researchers [22-34] around the world have worked to develop simulation codes that can simulate the microstructure, residual stress, and distortion response in steels during heat treatment. Generally, these researchers developed the simulation codes to investigate a single steel grade in a set part shape heat treated under a single quench method.

Şimşir and Gür showed that there are a number of combined influences between the thermal field, the phase transformations, and the stress/strain evolution which must considered to accurately simulate the heat treatment process as shown in Figure 1.18 [35-36]. Several sets of computer codes have been developed to simulate the heat treatment process such as DANTE [37], HEARTS [38], DEFORM™-HT [39], SYSWELD [40], FORGE, ANSYS, ABAQUS, and MSC.MARC have been commercialized and are available for license. None of these packages simulate the fluid flow and thermophysical events shown in Figure 1.18. ANSYS, ABAQUS, and MSC.MARC are finite element software packages that allow the user to program subroutines in order to simulate the heat treating process and do not have preprogrammed modules for heat treatment simulation [35-36]. Şimşir and Gür showed that HEARTS and DANTE covered the full range of applications that includes tempering, induction hardening and carburizing along with using both elasto-plastic and elasto-viscoplastic models for the mechanical properties. However, HEARTS only couples the thermal field to the phase transformation and stress field while not accounting for the stress field - phase transformation interactions while DANTE couples all three [35-36]. The other simulation packages either do not use both elasto-plastic and elasto-viscoplastic mechanical models (SYSWELD) or are not capable to do the special applications such as tempering, induction hardening and carburizing (DEFORM-HT, FORGE) [35-36]. Due to the fully coupled system along with the ability to simulate carburization and other special processes and its availability at CSM for this project, DANTE will be reviewed further in the next section.
1.3.1 DANTE® Development and Data

A group of companies and research organizations that included Ford, General Motors, Torrington, Sandia National Laboratories, Los Alamos National Laboratory, Lawrence Livermore National Laboratory, and Oak Ridge National Laboratory formed a collaborative group under the direction of the National Center for Manufacturing Sciences (NCMS). The goal of this group was “to develop a modeling methodology and software tool to simulate and predict the results of heat treatment, especially distortion” [37]. The modeling methodology produced a computational simulation model known as DANTE. Other packages (HEARTS, TRAST, SYSWELD) had reportedly attempted to do this while focusing on the mechanics and largely ignored distortion [37]; these were apparently hard to use and not accurate enough to be useful to industry or government [37].
The program development group started with a small gear made of 51xx series steel that was known to have distortion after heat treatment. A gear blank was used to simplify the simulation and reduce run times during development. The group also narrowed the focus to the heating, carburization cycle, quench and air cool steps of the heat treat. The incoming material condition was not a factor in the simulation [37]. Data for other steel grades has been obtained since the initial development of DANTE [41].

In order to simulate heat treatment processes, the program development group had to develop the various equations that govern multiphase transformations from various single phase transformation models. They included transformation plasticity due to the volume difference between crystal structures as a source of microscopic plastic flow. In the presence of a deviatoric stress field, the strain from volume misfit has a deviatoric part that is proportional to the surrounding stress [42]. This results in a relationship for transformation plastic strain rate described by Leblond, et al. as:

\[
\dot{\epsilon}^{TR} = \frac{K}{\Sigma_y} \left( \frac{\Delta V}{V} \right) \Psi(\Phi) \dot{\Phi} \sigma'
\]

where the transformational plastic strain rate tensor, \( \dot{\epsilon}^{TR} \), is proportional to the applied stress deviator tensor, \( \sigma' \), the rate of change of volume fraction, \( \dot{\Phi} \), the volume misfit, \( \Delta V/V \), inversely proportional to the yield strength of the weaker phase, \( \Sigma_y \), and proportional to the transformation-induced stress as a function of transformation, \( \Psi(\Phi) \), while \( K \) is a constant [42]. Additionally, this function was considered to resemble a phenomenological form of flow laws for state variable descriptions of viscoplasticity [42].

To generalize the state variable model to multiphase transformation kinetics, it is assumed that all points in the continuum can hold all the phases at the same time and each phase is described by a separate single phase state variable that is dependent on the kinematic hardening tensor and the isotropic hardening scalar. The deviatoric Cauchy stress in each phase is represented by \( \sigma''(\Phi) \) and assuming a volume fraction weighted rule of mixtures results in the deviatoric Cauchy stress as:

\[
\sigma' = \sum_i \Phi_i \sigma''(\Phi_i) ; \quad \sum_i \Phi_i = 1
\]

where \( \sigma' \) is the deviatoric Cauchy stress and \( \Phi_i \) is the volume fraction of each phase subject to the total volume fraction of all phases summing to 1 [42].
The multiphase transformation kinetics were modeled using a fundamental balance principle with Avrami-type kinetics and the Koistinen-Marburger equation shown to be the result of special classes of energy and mobility functions. For transformation to a single phase (constituent), a simplified energy function was shown to predict TTT and CCT behavior and these functions were expanded to competitive multiphase (i.e. multi-constituent) transformations [42]. The volume fractions of ferrite, pearlite, bainite, martensite and austenite were represented as $\Phi_f$, $\Phi_p$, $\Phi_b$, $\Phi_m$, $\Phi_a$, respectively. A set of global balance postulates and constitutive restrictions were applied and resulted in the following expressions that govern phase evolution with time:

$$
\frac{d\Phi_f}{dt} = v_f(C, \theta)\Phi_f^{\alpha_f} \Phi_a^{\beta_f} \{\Phi_f, f_{inal}(C, \theta) - \Phi_f\}, \Phi_f(0) = 0.0001 \\
\frac{d\Phi_p}{dt} = v_p(C, \theta)\Phi_p^{\alpha_p} \Phi_a^{\beta_p}, \quad \Phi_p(0) = 0.0001 \\
\frac{d\Phi_b}{dt} = v_b(C, \theta)\Phi_b^{\alpha_b} \Phi_a^{\beta_b}, \quad \Phi_b(0) = 0.0001 \\
\frac{d\Phi_m}{d\theta} = \begin{cases} 
0, & \Phi_m(0) = 0.0001, \quad \theta > M_s(C) \\
-v_m(C, \theta)\Phi_m^{\alpha_m} \Phi_a^{\beta_m}, & \Phi_m(0) = 0.0001, \quad \theta < M_s(C) 
\end{cases} \\
\Phi_a = (1 - \Phi_f - \Phi_p - \Phi_b - \Phi_m)
$$

where $C$ is the carbon concentration, $\theta$ is the temperature, and functions $v_f(C, \theta)$, $v_p(C, \theta)$, and $v_m(C, \theta)$, and $\alpha_f$, $\alpha_p$, $\alpha_m$, $\beta_f$, $\beta_b$, $\beta_p$, and $\beta_m$ are material constants that are based on TTT data [42]. In the equations, the subscripts a, f, p, b, and m stand for austenite, ferrite, pearlite, bainite, and martensite, respectively.

The DANTE simulation was built for use in a finite element solver, Abaqus®. All model creation and meshing is completed using Abaqus [37]. DANTE is a set of user subroutines that are called by Abaqus to define the microstructure and mechanical properties of the part at any given point (spatial and temporal) in the process. In order to fully model a carburized part, three individual simulations are completed. The first is the carbon diffusion model. This results in the predicted carbon profile in the part after carburization. Next, a thermal model is computed. Abaqus then uses the DANTE subroutines to calculate the phase transformations during heating and cooling. This model gives the temperature and phases at each node for each time step for the
entire process. The carbon profile from the first model is inserted once the part is at the carburization temperature.

Finally, a mechanical model is completed. In this model, the carbon profile is ramped up to the final carburized profile over several calculation steps to account for the mechanical effects and the distortion of the lattice caused by the increased carbon content near the surface. Dimensional changes associated with thermal expansion (contraction), phase transformations and elastic/plastic accommodation are incorporated. The coupled models provide predictions of the transformation products, the distortion in the final part, and the residual stress profile.

As mentioned previously, DANTE is a set of user subroutines for Abaqus. Once the model is created, the input files for Abaqus must be created. These can either be accomplished using the Abaqus pre-processing model to specify all the steps, or a previously created input file can be modified manually to change the steps in the simulation. The carbon profile can be viewed after the carburization model has been completed. The thermal profile can be viewed after the thermal model has been completed. The final microstructure, residual stress, strain, as well as the carbon and thermal profiles can be viewed at any region of the part and at any time step in the process from the results of the mechanical model.

The steel grades that have been analyzed by dilatometry to define appropriate transformation kinetics and volume changes are listed in Table 1.1 below. Where a carbon level is given, that specific alloy was analyzed independently. Where the carbon level is not specified, data were obtained from several sets of steels containing varying carbon levels. Table 1.2 shows the heating/cooling processes that have been analyzed in detail for temperature dependent heat transfer coefficients included in DANTE. Other steels can be modeled with data obtained from continuous cooling/isothermal transformation diagrams or Jominy tests for the kinetics data if the more accurate dilatometry data are not available. Standard mechanical tests at varying strain rates and temperatures may be used to develop the necessary mechanical data.

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>10XX</th>
<th>1038M</th>
<th>15B37</th>
<th>1541</th>
<th>1547</th>
<th>40XX</th>
<th>41XX</th>
<th>43XX</th>
</tr>
</thead>
<tbody>
<tr>
<td>46XX</td>
<td>48XX</td>
<td>51XX</td>
<td>51XXE</td>
<td>86XX</td>
<td>93XX</td>
<td>94BXX</td>
<td>52100</td>
<td></td>
</tr>
<tr>
<td>325CR</td>
<td>CRSI</td>
<td>HP94</td>
<td>M249</td>
<td>M256</td>
<td>PYRO</td>
<td>WDH11</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
1.3.2 DANTE® Comparisons

Li et al. [43] investigated the use of DANTE® in an optimization program conducted using a sensitivity analysis of the induction heat treatment for a small spur gear. The DANTE software was specifically used to optimize the water spray quench variables in order to produce a favorable residual stress profile. The results of one model were then used to create some modifications to the input for the next model leading to an optimized quench. This stress profile was then included in a calculation of the performance of the gear in bending fatigue. When the residual stress profile is not accounted for, the bending stress is significantly lower than when the residual stress profile is included.

Freborg et al. [44] investigated, through simulation and experimentation, the use of a simple multiple notch sample (Figure 1.19) to assess the sensitivity to residual stress and distortion response during quenching of Pyrowear® 53, a carburizing steel designed for use at temperatures above those where AISI 9310 is usable [45]. For the study both oil and intensive quenching (quenching with high velocity water thus preventing the film boiling stage of quenching) were examined. They showed that the bar distorted in opposite directions for the two quench methods while producing approximately equal hardnesses. They further showed that the distortion in the large notch was greater for the intensive quench compared to the oil quench. For the three samples of each quench type, the average surface residual compressive stresses between notches for intensive quenching was 515 MPa (74.8 ksi) while for oil quenching the average surface residual compressive stress was 170 MPa (24.6 ksi). However the variation in the residual stress between parts processed via intensive quenching was approximately three times that of oil quenching as seen in Figure 1.20. DANTE was also employed to show how the interaction between transformation and thermal changes affect the final distortion and residual stress profile. Freborg showed through DANTE simulations that the variation in the sequence of

<table>
<thead>
<tr>
<th>Furnace Heating</th>
<th>Oil Quench</th>
<th>Air Cooling</th>
<th>Nitrate-Nitrite Salt at 232C</th>
<th>Water Wash</th>
</tr>
</thead>
</table>

Table 1.2 – Heat Transfer Data Included with the Dante License [41]
transformation strains relative to the thermal gradient changes caused the differences in residual stress and distortion between the two quench methods.

Figure 1.19  Schematic of test specimen used to determine the effects of the heat treatment on distortion using both oil quench and intensive quenching. The specimen is 14mm square [45].

Ferguson et al. [46] used DANTE to compare a conventional oil quench heat treatment to an intensive quench. The goal was to improve the fatigue life by 25% of a spur gear of Pyrowear® 53 used in a military helicopter transmission. A comparison of the final residual stress profiles as predicted and as measured by x-ray diffraction showed the predicted oil quench residual stresses to be slightly more compressive than the measured values while the predicted stresses for the intensively quenched part were in line with the measured values (Figure 1.21).
This change in the residual hoop stresses resulted in an increase in the bending fatigue strength in steels that had very similar microstructures and hardesses.

Banka et al. [47] showed that the response of a part during quenching, specifically intensive quenching, is very dependent on the flow of the quench media over the part. A simulation of the quench medium flowing over a part was performed using FLUENT®. The resulting heat transfer distribution over the part was then used as an input to a DANTE simulation to predict the part response. It was concluded that the results of the DANTE simulation are significantly affected by the heat transfer gradient as compared to a single heat transfer value over the entire part surface, as is normally assumed.

![Figure 1.21 Predicted and measured hoop stress profiles at the center of the root for oil and intensively quenched Pyrowear® 53 gears as predicted by DANTE and calculated from x-ray diffraction measurements [46].](image)

Ferguson et al. simulated AISI 52100 steel bearing rings using DANTE in an axisymmetric model and compared the results after oil and intensive quenching to experiments. The simulated microstructures and the surface stress states were compared to the experiments and showed good agreement between both intensive and oil quenching [48].

Ferguson et al. further compared oil and intensive quenching of 9260 coil springs via simulations. They concluded that intensive quenching involves heat transfer coefficients between 20 and 40 kW/(m² °C) while the oil quench heat transfer coefficients peaked around 5 kW/(m² °C) with an average less than 1 kW/(m² °C) [49]. In the simulations, the only variables that changed were the heat transfer coefficients and the quench time for intensive quenching. The simulations showed that the intensively quenched coil had a very high
compressive residual stress at the surface while the oil quenched coil had essentially no residual stresses while the microstructures were similar [49].

Freborg and Ferguson compared simulations to experiments of a 1045 steel shaft with a keyway quenched via oil and intensive quenching. They applied a uniform heat transfer coefficient of 1200 W/(m² °C) for the oil and 28.2 kW/(m² °C) for the intensive quenching [50]. The microstructure was predicted to be approximately 83 % martensite and 90 % martensite for oil and intensive quenching respectively. These levels compared favorably to the experimental samples. They also showed the distortion levels and hardness levels were similar between the simulations and the experiments [50].

Freborg et al. compared standard and intensive quenching experiments to simulations using a DANTE axisymmetric model for pressure vessels with a wall thickness of 114.3 mm (4.5 in) made from 4340 steel. The standard water quench was simulated with a 700 W/(m² °C) heat transfer coefficient while the intensive quenching was simulated with 7 kW/(m² °C) on the outside and 20 kW/(m² °C) on the inside of the part [51]. They showed the residual stress profile through the section was more compressive at the surface for the intensive quench than the standard quench. With the increase in the surface compressive stresses, the core shows a corresponding increase in tensile residual stress (Figure 1.22) [51]. Hardness was also increased through the entire cross section after intensive quenching compared to the standard quench (Figure 1.23) [51].

![Figure 1.22](image)

**Figure 1.22** Comparative stress profiles through the vessel section for standard water and intensive quenching as predicted using DANTE [51].
Hubbard et al. used the Oak Ridge National Lab’s High Flux Isotope Reactor to investigate the development of residual stresses during oil and intensive quenching of 25 mm (1 in) diameter 5160 steel rods and then compared the results to DANTE simulations. For the oil quench, the simulations and the measured residual stresses were considered to be in good agreement (Figure 1.24). For the intensive quench, the simulations showed the same general trends as the measurements with some slight differences in the levels of the stress where the measurements showed a higher surface compressive stress than the simulations (Figure 1.25) [52].
Freborg et al. compared residual stresses after oil quenching with intensive quenching via simulation and experimentation of Pyrowear® 53. The experimental study was conducted in two parts; the first evaluated three point bending of a modified “V” notch sample while the second dealt with the full gear from a helicopter transmission [53]. Using simulations of the “V” notch samples in either the carburized and oil quenched and tempered or intensively quenched and tempered condition, they showed that the surface residual stress would become more compressive when the sample was intensively quenched. Based on the simulation work, the samples were experimentally heat treated and examined. This comparison showed that the surface residual stresses did follow the simulations in that the compressive stresses at the surface increased when intensively quenched compared to the oil quench (Figure 1.26). The samples were then fatigue tested in bending and again showed that intensive quenching improved the bending fatigue resistance, primarily due to the higher compressive stresses. In the second phase of the testing, the simulations showed that the residual stresses at the gear tooth root were more compressive after intensive quenching while at mid-tooth the residual stresses were similar for the two quenching methods [53]. In a follow-up publication, Freborg et al. showed that the stresses in the gear were decreased for the same load for the intensive quench condition compared to the oil quench [54].

Li et al. coupled the DANTE model with an induction heating model, Flux2D®, to investigate the effects of induction hardening and subsequent spray quenching on the residual stress and distortion in a 1541 axle shaft. The Flux2D model provided a power density map that was then imported to the DANTE model to control the thermal profile in the shaft. Changes to the spray quenching heat transfer rate were made and the resultant models compared. They showed that as the heat transfer rate increased, the surface residual stresses became more compressive with a corresponding increase in the tensile stresses at the case-core interface in both the axial and hoop directions [55]. Ferguson et al. reported additional results for additional 1040 and 4140 steels in comparison to the 1541 of the original work. They reported little variation in the level of surface compressive residual stresses between the three alloys [56].
Figure 1.25  (a) Neutron and x-ray diffraction measured and (b) DANTE predicted residual stress results for 5160 steel rods oil quenched and tempered at 350 °C (660 °F). (c) Neutron and x-ray diffraction measured and (d) DANTE predicted residual stress results for 5160 steel rods oil quenched and tempered at 700 °C (1300 °F) [52].

Figure 1.26  Comparison between DANTE predicted and x-ray diffraction calculated residual stress profiles for Pyrowear® 53 steel notch samples [53].
Banka et al. coupled a computational fluid dynamics (CFD) model (Azore®) for fluid flow around three 4340 steel rings of varying wall thicknesses to the DANTE model under two different fluid flow directions. The CFD model showed a variation in the gas velocity around the ring resulting in variation of the heat transfer coefficients over the surface of the ring as shown in Figure 1.27. The distortion and residual stresses were then compared between the wall thicknesses with maximum radial displacement predicted at the bottom inside corner and peak residual tensile axial stress on the ID of the rings [57].

![Figure 1.27](image)

Figure 1.27  Top and bottom perspective views of an Azore® CFD model used to predict heat transfer coefficients for a fluid quench media flowing vertically up around 4340 steel rings of varying wall thicknesses. These heat transfer coefficients were then mapped onto the rings for the DANTE simulations of the heat treatment [57].

1.4  Focus of the Thesis

DANTE simulations have been shown to approximately predict the microstructure, residual stress state, and distortion that results from the heat treating process. This includes heating, carburizing, and quenching. With the availability of DANTE at CSM, DANTE heat treatment simulations will be performed in order to investigate the following questions:

1. How does the progression of the phase transformation compare between oil quenching and intensive quenching?

2. Is the expected response to intensive quenching confirmed by modeling, and does the development of the residual stresses conform to the hypothesis promoted by IQT in the literature?
CHAPTER 2: DANTE HEAT TREATMENT SIMULATION PROCEDURES

2.1 Creating Input Files for Simulations

In order to simulate heat treatment utilizing DANTE, the part must be created in ABAQUS. The part is drawn in ABAQUS similar to other computer aided drawing programs such as AutoCAD or Solidworks. Once the part has been drawn, it is typically partitioned so that a fine mesh size can be created near the surface and a larger mesh size can be created in the interior of the part. After partitioning, the various boundaries that make up the different volumes of the part are seeded with the initial node locations. After all the edges have been seeded, the element type that is desired is chosen. The part can then be meshed. If the mesh does not conform to size requirements, the individual mesh areas can be deleted, the edges reseeded, and the area remeshed until an acceptable mesh is created. An acceptable mesh varies based on element type and model requirements. Typically, the mesh needs to be finer near the surface to account for the carburization process as well as any large thermal gradients that may result during quenching [41].

Next the material properties need to be defined and assigned to the part. The material properties control the density and the carbon diffusivity. This can be done completely in ABAQUS or it can be started in ABAQUS and then be finalized in the input file that is created. The majority of the material properties, such as elastic and plastic mechanical properties, are contained within the DANTE data files and are imported into ABAQUS automatically when needed. A section is the mechanism ABAQUS utilizes to link the material information created to the part in the model. The section contains the material properties along with other information pertaining to the type of part created (solid, shell, beam). The section is then assigned to the part and thereby the material the part is made of is assigned. The section definition becomes more important in multi-part assemblies that can be analyzed using ABAQUS.

The steps used in the simulation can then be defined. Each different step through the heat treatment process is a separate step in ABAQUS. The steps in each of the simulations needed to fully simulate a heat treatment vary by what is required in the simulation. For the carburization
simulations, the only step needed is the carburization step. Here, the part is assumed to be at the carburizing temperature at the start of the simulation. The carbon potential is then applied to the surface nodes and ABAQUS calculates the diffusion of the carbon into the part at the nodal level. For a thermal model, only steps in the heat treatment where the temperature is changing need to be considered. The mechanical model has a step for each step in the heat treatment process. Similar to the material properties, this can be done completely within ABAQUS or a blank step can be created for input file creation. After the input file is completed, the simulation steps can be copied from another simulation input file and edited manually in the input file. This can be done easily in Microsoft® Notepad or another text editor. The output from each step is also defined here and can be easily copied from other input files. Sample input files can be found in Appendix A, B, or C for the carburization, thermal, and mechanical models respectively.

Interactions are also defined. Interactions control how carbon is diffused into the part in the carburization model or heat is transferred to and from the part in the thermal model. Several sets of nodes also need to be defined. The first node set is the set of surface nodes where heat and carbon are introduced to the part. Depending on the type of simulation (2D or 3D), various fixed planes, lines and individual node sets will need to be created also as required to fix the part in space and define any symmetry boundaries.

2.2 Running Simulations

After the input file has been created, it can be modified to account for the various simulations that must be completed for a single heat treatment. As stated previously, for carburized parts, three individual simulations are completed for a single heat treatment condition. The first is the carburization model. This model predicts the carbon distribution as a function of time during the carburization step in the heat treatment. The nodal carbon level is exported and utilized in the thermal and mechanical models. Next the thermal model is completed to calculate the temperature and phase fraction changes at the nodal level as a function of time. Finally, the stress model is completed to calculate the stress level and distortion at the elemental level through the heat treatment process. Before the mechanical model can be started, the results from the thermal model are imported.
2.2.1 Carburization Model

This model takes the initial material (usually of uniform carbon content) at the carburizing temperature and applies a constant carbon potential to the surface nodes. The increased carbon on the surface is then allowed to diffuse into the bar using node level diffusion. The carbon diffusion variations with temperature and concentration are included in the input file as shown in Appendix A. The goal of the carburization model is to calculate the final nodal carbon level throughout the part. The carbon profile data are contained in the .dat file that is created by ABAQUS during the model run. The data are copied from the .dat file and placed into a spreadsheet to format the data correctly. Once the data are formatted, a .nod file is created from the formatted data. This file is then used in subsequent simulations to input the final carbon level at the appropriate time in the simulation. This is a pure diffusion simulation and does not account for any microstructure or dimensional changes that may result from the carburization process.

When a multistep carburization is simulated such as for vacuum carburizing which has several boost and diffuse cycles, a .nod file is created for each boost or diffuse cycle in the process. For a carburization that has five boost and diffuse cycles, 10 .nod files are needed to fully capture the carbon level, one file at the end of each boost or diffuse time.

2.2.2 Thermal Model

After the carburization model has been completed and the .nod file(s) have been created, the thermal model can then be run to incorporate the response to heating and cooling. The instruction set is modified to include the film properties for the various heat transfer steps. The number of steps is also increased so that there is a simulation step for every step during the heat treatment for which the temperature is expected to change. Any constant temperature steps in the heat treatment process are not simulated as the temperature does not change. For example, a heat treatment consists of heating to austenitizing temperature, carburizing at the temperature, transfer of the part through air at ambient temperature, immersion of the part in the quenching bath, quenching of the part, and air cooling to room temperature. In this example, the temperature of the part is expected to change in five of the steps, all except the carburization step. The thermal model would then contain five steps to represent these five steps in the actual
heat treatment. A typical heat up and quench simulation may contain at least two thermal steps and any number more than that.

The first step in the present work is the heating. Next, the transfer from the furnace to the quench tank is modeled. Depending on the speed of immersion of the part and the quench media, an immersion step is then modeled. For the immersion step, a small text file is used to instruct ABAQUS as to part orientation, the velocity at which the part is submerged, the heat transfer rate and temperature of the atmosphere above the quenchant and the heat transfer rate and temperature of the quenchant. For gas quenching, the immersion step is not used and the gas is assumed to fully envelop the part at the start of the quenching step, resulting in a more uniform quench.

Once the part is fully immersed, the quench is allowed to progress for the desired time. The part is typically equilibrated to the quenchant temperature by the end of the quench step. Finally, the part is “removed” from the quenchant and allowed to cool to room temperature in air. There is no reverse immersion accounted for as the temperature difference between the quenchant and the air is quite small compared to the hot part and the quenchant temperature.

2.2.3 Mechanical Model

The last model run for each heat treatment simulation is the mechanical model. This model imports the data from the thermal model for the temperature and phase transformations at the nodal level and averages the data across each element. It then uses this data to calculate the thermal and phase transformation stresses and distortion through the heat treatment. If there is a preexisting residual stress profile from prior part manufacturing steps, it can be incorporated at the beginning of the mechanical model.

Each step in the heat treatment receives a simulation step. During the carburization step, the carbon profile is ramped to the final carbon profile as previously calculated in the carburization model. For vacuum carburization, this ramp is done for each boost or diffuse step in the carburization process. The times used in the previous models should also match those used for each step in the mechanical model.

The type of element must be changed from a diffusion/thermal type to a mechanical property type of element. This is typically done while editing the input files but can be done in ABAQUS if so desired. Also, to reduce calculation time, ABAQUS is instructed to use a single
calculation point for each element. The temperature, carbon profile (if applicable), and the phase fractions are averaged from the nodes at the corners of the elements into the centroid of the element. The mechanical properties and distortion are then calculated at the centroid of the element. The temperature, carbon profile, and phase fractions are also then stored at the element level thereby reducing the amount of data in the output database.

2.3 Post Processing

Once the mechanical model has been completed, the output database can be viewed in ABAQUS. Here the phase fraction, temperature, carbon profile, stress components, and distortion can be viewed throughout the part. If required, the thermal model can be employed to obtain carbon profile, phase transformation, or temperature data from the nodes at each time step in the process. The time steps start at a defined time as specified in the input file. If the simulation does not converge at that length of time, ABAQUS automatically reduces the time in each step until the solution converges. If the equations converge without any issues, ABAQUS automatically increases the time covered by each step to accelerate the calculation time. The maximum time for each time step is also specified in the input files.

2.4 DANTE Data

Due to the large number of varying steel alloys that are available, only a limited number of these materials have been characterized to the extent needed for DANTE to simulate their heat treatment. The characterization of a steel consists of thermodynamic and kinetic data such as transformation behavior similar to that contained in both the time-temperature-transformation (TTT) diagram as well as the continuous cooling transformation (CCT) diagram. This can be done utilizing dilatometry or by reading the data from the TTT and CCT diagrams [41]. These data include the various transformation temperatures for both on-heating and on-cooling for each of the possible microstructural constituents. Also needed are the thermal conductivity and specific heat by phase. Mechanical property data required include elastic properties as a function of temperature, the coefficient of thermal expansion for each phase (to calculate distortion), and hardness data by phase and carbon level [41].
Also, there are a large number of processes that can be used during a heat treatment. This section will present the materials and processes that have been analyzed and the resulting data included in the DANTE license.

2.4.1 Material Data

DANTE includes data for a number of materials as was shown in Table 1.1. If other steels are modeled, the data for these steels would need to be collected and formatted properly for DANTE. The data include the needed thermodynamic and kinetic data as a function of carbon content such as transformation behavior similar to that contained in both the time-temperature-transformation (TTT) diagram as well as the continuous cooling transformation (CCT) diagram along with mechanical property data as a function of temperature. When additional data are required between carbon contents listed in the material data files in DANTE or temperatures where values are given, a linear interpolation is automatically performed by ABAQUS in the calculations for each time step.

2.4.2 Process Data

As was shown in Table 1.2, a limited number of processes have been analyzed to obtain the heat transfer rates during the process. Table 2.1 list the heat transfer rates in W/(m² °C) for furnace heating, oil quenching and air cooling [41]. These data are typical of what is found of these steps and oils. Other heat transfer rates may be found in the literature such as for intensive quenching (fixed 20,000 W/(m² °C) [51]) or gas quenching. These heat transfer rates can then be used in the thermal model input file as a replacement for other data that is not needed for that model.

2.5 Part Model

Due to the high degree of interdependence between thermal gradient, mechanical properties and phase transformations, a cylindrical bar one inch (25.4 mm) in diameter and 10 inches (254 mm) long (see Figure 2.1) was selected for the heat treatment simulations, to simplify the geometry and focus on understanding better the fundamental thermal/transformation/mechanical interactions. This bar shape reduces influence of shape on the responses of temperature and
phase transformation. Also, due to the inherent symmetry of the bar, the part used can be reduced from the full bar to a reduced section. Two different simulations were used for the bar. The first was a three dimensional half bar simulation that allows the immersion step to be modeled more accurately. The second was a two dimensional axi-symmetrical simulation in order to quickly run entire simulation sets with minor heat transfer changes between them. Each of these will be discussed further in the following subsections.

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>h</th>
<th>T (°C)</th>
<th>h</th>
<th>T (°C)</th>
<th>h</th>
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<td>800</td>
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<tr>
<td>120</td>
<td>162</td>
<td>93</td>
<td>143</td>
<td>300</td>
<td>1500</td>
</tr>
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<td>148</td>
<td>178</td>
<td>176</td>
<td>197</td>
<td>400</td>
<td>2500</td>
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<td>533</td>
<td>565</td>
<td>616</td>
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<td>4750</td>
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<tr>
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<td>708</td>
<td>676</td>
<td>811</td>
<td>650</td>
<td>3000</td>
</tr>
<tr>
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<td></td>
<td>1000</td>
<td>800</td>
</tr>
</tbody>
</table>

Table 2.1 – Heat Transfer Coefficients (h) in W/(m² °C) for Furnace Heating, Air Cooling, and Oil Quenching as a Function of Temperature [41]

**Figure 2.1** Schematic of modeled part used in simulations of various heat treatments.

2.5.1 3D Part Model

During the training sessions at Deformation Control Technologies (DCT), the 3D part model was created. This model was meshed following the suggestions of Dr. Charlie Li. In the
previous work done at DCT, it was found that using a fine mesh to a depth of 1.5 mm (0.060 in) produced accurate results for carbon distribution, phase transformation near the surface under high heat transfer rates, and residual stress [41]. The mesh can be seen in Figure 2.2. ABAQUS restricts the size of the model to less than 10,000 nodes at CSM due to the nature of the license. In creating the input file while at DCT, the ABAQUS restrictions that are in place at CSM were avoided. Due to this restriction, modification of the node spacing created during training was not possible.

![Figure 2.2](image)

**Figure 2.2** Schematic of element layout for the 3D 25.4 mm (1 in) diameter cylindrical bar (top) and a close up of the left end (bottom) used in the simulations. The fine elements extend to a depth of 1.5 mm from all exterior surfaces. Data were collected from approximately the center of the length of the bar. The centerline surface is a mirror plane and not an exterior surface in the simulations.

All the microstructure, plastic strain and residual stress data were taken from elements at the mid-length of the bar. Due to the averaging done during the stress model in ABAQUS, the data come from the centroid of each element. This means the data are not exactly on the centerline but is offset by 0.47 mm (0.019 in). Also, this means that the data do not extend to the surface but are displaced inward 0.15 mm (0.006 in). Table 2.2 shows the distance from the top
surface to the centroids of the elements used to collect the data. This group of elements is located approximately equidistant from the bar ends in order to avoid end effects. Center line distortion data were collected from the entire length of the bar. When looking at the cross-section of the bar, the entire centerline surface was fixed such that it could not move out of that plane. In order to further locate the bar in space, the very bottom node on the left end was fixed so it could not move parallel to the centerline of the bar but can expand and contract radially. The resulting centerline distortion was measured according to a convention where upward distortion is positive.

Table 2.2 – Distance in mm from Top Surface of Element Centroid by Element

<table>
<thead>
<tr>
<th>Element</th>
<th>Distance to Centroid</th>
<th>Element</th>
<th>Distance to Centroid</th>
<th>Element</th>
<th>Distance to Centroid</th>
</tr>
</thead>
<tbody>
<tr>
<td>5376</td>
<td>0.15</td>
<td>11777</td>
<td>8.12</td>
<td>18934</td>
<td>19.32</td>
</tr>
<tr>
<td>5377</td>
<td>0.45</td>
<td>11756</td>
<td>9.14</td>
<td>18969</td>
<td>20.34</td>
</tr>
<tr>
<td>5378</td>
<td>0.75</td>
<td>11805</td>
<td>10.15</td>
<td>18950</td>
<td>21.35</td>
</tr>
<tr>
<td>5379</td>
<td>1.05</td>
<td>11789</td>
<td>11.17</td>
<td>18980</td>
<td>22.37</td>
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<td>1.35</td>
<td>11774</td>
<td>12.19</td>
<td>18947</td>
<td>23.39</td>
</tr>
<tr>
<td>11766</td>
<td>2.01</td>
<td>18937</td>
<td>13.21</td>
<td>725</td>
<td>24.05</td>
</tr>
<tr>
<td>11828</td>
<td>3.03</td>
<td>18955</td>
<td>14.23</td>
<td>1475</td>
<td>24.35</td>
</tr>
<tr>
<td>11781</td>
<td>4.05</td>
<td>18987</td>
<td>15.25</td>
<td>2225</td>
<td>24.65</td>
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<tr>
<td>11788</td>
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<td>18938</td>
<td>16.26</td>
<td>2975</td>
<td>24.95</td>
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<td>11803</td>
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<td>18954</td>
<td>17.28</td>
<td>3725</td>
<td>25.25</td>
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<td>11769</td>
<td>7.10</td>
<td>18973</td>
<td>18.30</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

2.5.2 2D Part Model

A 2D part model was created for simulations of a variety of heat transfer rates to understand the heat transfer rate at which surface tensile stresses become compressive. In the literature, intensive quenching is modeled without an immersion step [46, 48-51]. This is a result of the process employed by IQ Technologies and reduces model complexity. However, it was decided to use the 3D model for the majority of this work in order to also account for some immersion effects during the intensive quenching simulations. The 2D model was used to quickly simulate a number of heat transfer rates to develop surface residual stress data as a function of heat transfer rate, as explained in the next chapter. The mesh for the 2D model was slightly different than that of the 3D model. It is shown in Figure 2.3. As can be seen, the node
Spacing increases with increasing depth to 1.5 mm (0.060 in) followed by a large node space from 1.5 mm (0.060 in) to the centerline. The 2D model was created as an axi-symmetric model. This means ABAQUS only allows heat transfer and distortion to happen radially over most of the part and axially along the ends of the bar. Again, a fine mesh was used along the external surfaces in order to account for the thermal gradient seen due to high heat transfer rates.

(a)

(b)

Figure 2.3 Schematic of element layout of the 2D 25.4 mm (1 in) diameter bar. (a) entire bar, (b) close up of left end. Smaller node spacing was used to a depth of 1.5 mm from the surface to account for a carbon profile and variations in phase transformation that are expected near the surface due to heat transfer.

2.5.3 Materials Simulated

As noted earlier, the material data included with DANTE are somewhat limited. The focus of this work was not to develop data for more materials but use data already included to develop insight and further understanding related to interactions between heat transfer, phase transformation, distortion, and residual stress. This, along with the desire to use materials widely used in industry and a need to consider a range of hardenability, the materials chosen were: 41XX, 43XX, 86XX, and 10XX. The 10XX alloys were included due to claims of comparable surface hardn esses compared to the other alloys after intensive quenching. The chemistry of each alloy is shown below in Table 2.3. Each of the alloy families was analyzed at several carbon levels in order to account for carburization in the thermal and mechanical data.
ABAQUS uses linear interpolation for data at carbon levels between those for which data are included in DANTE.

Table 2.3 – Steel Chemical Compositions for Alloys Used to Develop Data Employed in Simulations [41]

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Mn</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>41XX</td>
<td>0.67</td>
<td>0.23</td>
<td>0.20</td>
<td>1.01</td>
<td>0.23</td>
</tr>
<tr>
<td>43XX</td>
<td>0.76</td>
<td>0.3</td>
<td>1.58</td>
<td>0.62</td>
<td>0.25</td>
</tr>
<tr>
<td>86XX</td>
<td>0.84</td>
<td>0.22</td>
<td>0.44</td>
<td>0.54</td>
<td>0.22</td>
</tr>
<tr>
<td>10XX</td>
<td>0.78</td>
<td>0.19</td>
<td>0.06</td>
<td>0.23</td>
<td>0.02</td>
</tr>
</tbody>
</table>

2.5.4 Heat Treatments Simulated

The heat treatment schedules are shown in Table 2.4 for oil and high intensity quenching. As can be seen, the differences between the oil quench and the high intensity quenching start at the immersion step. The heating rate and “carburization” time are consistent across the simulations. Each of the high intensity quench conditions was simulated with different hold times in the quench with the goal of finding an “ideal” quench time leading to large residual compressive stresses at the part surface.

Table 2.4 – Heat Treatment Schedule Times in Seconds Used in Simulations

<table>
<thead>
<tr>
<th>Heat Treatment Step</th>
<th>Oil</th>
<th>HIQ</th>
</tr>
</thead>
<tbody>
<tr>
<td>Heat Up to 960 °C</td>
<td>1000</td>
<td>1000</td>
</tr>
<tr>
<td>“Carburization”</td>
<td>21600</td>
<td>21600</td>
</tr>
<tr>
<td>Transfer</td>
<td>12</td>
<td>12</td>
</tr>
<tr>
<td>Immersion</td>
<td>5</td>
<td>0.5</td>
</tr>
<tr>
<td>Quench</td>
<td>1000</td>
<td>5, 10, 15, or 20</td>
</tr>
<tr>
<td>Air Cool</td>
<td>1500</td>
<td>3000</td>
</tr>
</tbody>
</table>
CHAPTER 3: RESULTS

3.1 Overview

The 3D bar was used to simulate the heat treatment schedules shown in Table 2.4 using the base carbon level along with two carburization levels: carburized to 0.80 wt pct C and 1.0 wt pct C at the surface. A base carbon of 0.20 wt pct C was used for all materials and conditions. The bar was simulated in the horizontal position to increase predicted distortion. Each oil quenched condition utilized the heat transfer rates for the oil quench included in DANTE while the high intensity quenching utilized a set heat transfer rate of 20,000 W/m²°C. Whereas the heat transfer rate for oil quenching varies as a function of temperature, the heat transfer rate is held constant for high intensity quenching. This chapter presents the results of the simulations by material, carburization level, and quenching method.

3.2 Oil Quench

This section will present the results for oil quenching. Results from the non-carburized condition are in Section 3.2.1 while carburized to 0.80 wt pct C and 1.0 wt pct C are in Sections 3.2.2 and 3.2.3 respectively.

3.2.1 Non-carburized

Figure 3.1 shows the final predicted microstructure of 1020, 4120, 4320, and 8620 steels for the non-carburized condition. As can be seen in Figure 3.1a and as expected, there is no retained austenite predicted. Figure 3.1b shows that only 1020 contains any appreciable ferrite content with 39% in the core where the cooling rate is the slowest. Figure 3.1c shows the pearlite fraction for the various steels. Similar to ferrite, only 1020 is predicted to contain pearlite with peaks of 77% 1.25 mm (0.05 in) below the surface. Whereas ferrite peaked in the middle, pearlite peaks below the surface due to lower temperature transformation products dominating near the surface. Figure 3.1d shows that phase transformation in 1020 was dominated by higher temperature products (ferrite and pearlite) in the core leaving very little to
transform at lower temperatures into bainite or martensite (Figure 3.1e). For the other steels, 8620 shows the highest bainite fraction with 89%, followed by 4120 and 4320 with 77% and 44% respectively. This shows the differences in hardenability in these three steels with the highest hardenability (4320) having the lowest fraction of bainite. Figure 3.1e shows the final predicted martensite fraction. As expected, 4320 has the highest martensite fraction with 68% near the surface followed by 4120 and 8620 with 58% and 55% respectively. 1020 was only able to transform near the surface to martensite with a peak of 2% as the entire bar transformed at higher temperatures to ferrite, pearlite or bainite as was seen previously.

Figure 3.2 shows the predicted final effective plastic strain, distortion, and residual stress. In Figure 3.2a, the 1020 bar is shown to have an overall lower effective plastic strain while 4120, 4320 and 8620 all have similar strain levels. The distortion of the 1020 bar is predicted to be upward with a deflection of 0.35 mm (0.014 in). The other three alloys show a downward deflection with 4320 having the smallest deflection of 0.33 mm (0.013 in). 4120 and 8620 show deflections of 0.52 mm (0.02 in) and 0.54 mm (0.021 in) respectively. Figure 3.2c shows that the residual stress is predicted to be tensile at the surface for 4120, 4320, and 8620 with hoop stresses of 390 MPa (56.5 ksi), 475 MPa (69 ksi), and 185 MPa (27 ksi) respectively. 1020 is predicted to be compressive at the surface (225 MPa (33 ksi)) and core (210 MPa (30 ksi) and tensile (190 MPa (28 ksi) in the mid-radius region. A slight increase in stress is seen for 8620 from the surface to about 2 mm below the surface and a stress of 220 MPa (32 ksi) then decreasing to the core compressive stress. Core compressive stress levels are 215 MPa (31 ksi), 275 MPa (40 ksi), and 75 MPa (11 ksi) for 4120, 4320 and 8620 respectively.

3.2.2 Carburized to 0.80 Wt Pct C

Figure 3.3 shows the predicted carbon profile for the carburized condition for a carbon surface potential of 0.8 wt pct for six hours at 960 °C. The profile appears jagged due to the node spacing near the case-core interface. Figure 3.4 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 0.8 wt pct C at the surface. As would be expected, the carbon profile changes the phase transformation response upon cooling. Figure 3.4a shows that all four alloys are predicted to have some retained austenite at the surface due to the increased carbon from carburization. 1020 has the lowest predicted retained austenite
Figure 3.1 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched in 65 °C oil for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.2  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched in 65 °C oil.
at approximately 1% followed by 4120 with approximately 3%. 8620 is predicted to have approximately 9% and 4320 is predicted to have just under 10% retained austenite. Due to the immersion process and the additional cooling that the bottom of the bar receives, the microstructure profiles are not symmetrical. This is more pronounced for 1020 as seen in the retained austenite profile. Figure 3.4b shows the ferrite fraction is essentially zero for 4120, 4320, and 8620, while 1020 contains approximately 40% ferrite in the core. Similar to ferrite, Figure 3.4c shows the pearlite fraction is essentially zero for 4120, 4320, and 8620 while 1020 contains approximately 60% pearlite in the core increasing to approximately 90% at the case-core interface and falling to zero at the surface. Bainite fractions are shown in Figure 3.4d. 8620 is predicted to have approximately 85% bainite while 4120 contains approximately 75% and 4320 contains approximately 40%, all in the core. Near the surface, the increased carbon level in the case delayed transformation and produced more martensite in the case than the non-carburized condition. In contrast, 1020 only contained approximately 55% bainite in the mid-case region with the high carbon level at the surface allowing martensite to form as seen in Figure 3.4e. The added carbon from carburization has delayed the transformation in the case allowing a large fraction of martensite to form.

Figure 3.5 shows the final predicted residual stress and effective plastic strain in the bar for the four alloys. Due to the carburization, the effective plastic strain in 1020 has increased to levels similar to but still below the other three alloys. The strain has also increased in 4120, 4320, and 8620 compared to the non-carburized level. Figure 3.5b shows that the distortion for 1020 is opposite to the other three alloys. The magnitude of the distortion is greatest for 8620, followed by 4120, 1020 with 4320 having the smallest predicted distortion. The residual stress profiles are all similar as shown in Figure 3.5c. All four alloys are predicted to have compressive residual stress at the surface ranging from 370 MPa (53.7 ksi) for 4120 to 300 MPa (43.5 ksi) for 1020. 1020 shows a more gradual change from increasing tensile stress to decreasing tensile stress around the case-core interface compared to the other three alloys that all show sharp changes in residual stress profiles at the case-core interface. 8620 shows the lowest tensile peak of approximately 125 MPa (18.1 ksi) at the case-core interface while 4320 shows the highest peak of approximately 260 MPa (37.7 ksi) with 4120 and 1020 intermediate. Due to the low peak in stress of 8620, in order for the bar to be in equilibrium, the core remains in tension while the other three alloys have compressive stresses in the core.
Figure 3.3  Predicted carbon profile of the carburization to 0.8 wt pct C heat treatment. This profile is independent of alloy and quenching conditions.
Figure 3.4 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched in 65 °C oil for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.5  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched in 65 °C oil.
3.2.3 Carburized to 1.0 Wt Pct C

Figure 3.6 shows the predicted carbon profile when the bar is carburized with a carbon surface potential of 1.0 wt pct for six hours at 960 °C. Again, due to node spacing the carbon profile appears jagged. The additional carbon potential leads to a higher surface carbon level compared to the 0.8 wt pct C case. Since the carburization time is unchanged, the profile has a similar effective case depth as the prior case, as desired.

![Figure 3.6 Predicted carbon profile of the carburization to 1.0 wt pct C heat treatment. This profile is independent of alloy and quenching conditions.](image)

Figure 3.7 shows the predicted final microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 1.0 wt pct C at the surface. Figure 3.7a shows that the retained austenite fraction has increased compared to the last 0.8 wt pct C case. Retained austenite has only increased slightly for 4120, while the other alloys show a much greater increase in retained austenite. This increase in retained austenite comes from the added stability of austenite due to the increased carbon content and corresponding decrease in martensite start and finish temperatures. The increased austenite at room temperature also reduces the amount of
martensite that forms as seen in Figure 3.7e. As can be seen in Figures 3.7b-c, only 1020 contains any appreciable amount of ferrite or pearlite with the peak ferrite in the center and the peak pearlite near the case-core interface. Bainite is predicted to be similar to the previous case as seen in Figure 3.7d. The martensite results are also similar as can be seen in Figure 3.7e with the increased retained austenite in 8620 and 4320 leading to a lowered fraction at the surface.

Figure 3.8 shows the final predicted residual stress and effective plastic strain in the bar for the three alloys. The effective plastic strain near the surface has increased for all four alloys compared to the previous case as shown in Figure 3.8a. Figure 3.8b shows that the distortion for 1020 has decreased in magnitude while the other three alloys have increased centerline bending. Predicted residual hoop stresses are shown in Figure 3.8c. Compressive stresses at the surface have increased approximately 50 MPa (7.3 ksi) while the peak tensile stresses at the case-core interface are similar. This leaves the core stresses to be more tensile in order to maintain equilibrium.

3.3 High Intensity Quench

This section will present the results for high intensity quenching. The heat treatments were similar to the oil quenched conditions consisting of heating, carburizing (or hold for the non-carburized conditions), a short immersion followed by high intensity quenching for a set time and finally air-cooling. Four different quench hold times were simulated for the high intensity quenching to investigate sensitivity to hold times and determine the preferred quench time. Results from the five second hold in the non-carburized condition are in Section 3.3.1 while carburized to 0.80 wt pct C and 1.0 wt pct C are in Sections 3.3.2 and 3.3.3 respectively. Results from the 10 second hold in the non-carburized condition are in Section 3.3.4 while carburized to 0.80 wt pct C and 1.0 wt pct C are in Sections 3.3.5 and 3.3.6 respectively. Results from the 15 second hold in the non-carburized condition are in Section 3.3.7 while carburized to 0.80 wt pct C and 1.0 wt pct C are in Sections 3.3.8 and 3.3.9 respectively. Results from the 20 second hold in the non-carburized condition are in Section 3.3.10 while carburized to 0.80 wt pct C and 1.0 wt pct C are in Sections 3.3.11 and 3.3.12 respectively.
Figure 3.7  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched in 65 °C oil for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.8 Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched in 65 °C oil.

3.3.1 Non-carburized Five Second Quench Hold Time

Figure 3.9 shows the final predicted microstructure of 10XX, 41XX, 43XX, and 86XX steels for the non-carburized condition. From Figure 3.9a it can be seen that all but 1020 show
some retained austenite in the non-carburized condition. 8620 is predicted to have the most retained austenite at approximately 0.5% followed by 4320 and 4120. Ferrite fractions are shown in Figure 3.9b. As with the oil quenching, only 1020 contains ferrite up to 23% in the core. 1020 also is the only alloy predicted to have any pearlite (Figure 3.9c) in the final microstructure at up to approximately 65% in the core trailing down to zero at the surface similar to the carburized conditions previously. Unlike the previous results, 1020 is predicted to contain up to 20% bainite in the microstructure (Figure 3.9d). The other three alloys also have bainite in the core region, from approximately 40% for 4320 up to approximately 50% for 4120 and 8620. The balance of the microstructure is martensite as shown in Figure 3.9e. With a large portion of the microstructure of 1020 transformed before the temperature reaches the martensite start temperature, there is little martensite predicted in the core as might be expected.

Figure 3.10 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. In Figure 3.10a, the effective plastic strains are similar near the surface for all for alloys. As the distance from the surface increases, 1020 falls to a lower predicted strain of approximately 0.005 while 4120, 4320, and 8620 all have similar strain levels of approximately 0.012. All the bars have upward curves to their centerlines as shown in Figure 3.10b. 1020 is predicted to distort the least at about 0.3 mm (0.12 in) while 4120 and 8620 have similar predicted deflections around 1 mm (0.039 in) and 4320 is slightly less. In Figure 3.10c, the residual hoop stresses are similar near the surface for 4120, 4320, and 8620 at around 200 MPa (29 ksi) in compression. 1020 shows a much more compressive residual stress level at the surface of 450 MPa (65.3 ksi). As a result of the higher compressive stress at the surface the core hoop stresses are much higher than the other three alloys in order to maintain equilibrium.

3.3.2 Carburized to 0.80 Wt Pct C Five Second Quench Hold Time

The simulated carburization process for the intensive quenching was the same as for the oil quenching, therefore the carbon profiles are the same as previously presented in Figure 3.3. Figure 3.11 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 0.8 wt pct C at the surface. Similar to what was seen in the oil quench, the retained austenite (Figure 3.11a) has increased for all four alloys compared to the non-carburized condition with 4320 containing approximately 6.5% at the surface followed by 8620 and 4120 at approximately 4.7%. 1020 has the least amount of retained austenite at just less than 2%. As
Figures 3.11b and 3.11c show, only 1020 contains ferrite at about 25% and pearlite at about 70%. Bainite is shown in Figure 3.11d. With the large amount of ferrite and pearlite already formed, 1020 contains a peak bainite fraction of 20% approximately 2.5 mm (0.098 in) below the surface. The other three alloys show bainite fractions in the core ranging from 40% for 4320 to 55% for 4120 up to approximately 60% for 8620. Due to fast cooling from the high heat transfer rate of high intensity quenching, the surface of the bar cools very quickly and avoids higher temperature transformations leading to high fractions of martensite as shown in Figure 3.11e. As was seen in Figure 3.11d, 1020 contains some bainite at the surface thus reducing the amount of martensite that can form to a peak of 90% just below the surface. The bainite and retained austenite levels result in a decreasing martensite fraction closer to the surface. This effect is also seen in 4320 and 8620 but is primarily a result of the higher retained austenite levels due to carburization. The core martensite levels are set by the cooling of the bar and once the bar is removed from the quench after 5 s, the cooling slows which allows the core to transform to higher temperature transformation products.

Figure 3.12 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. 1020 shows similar surface effective plastic strain levels compared to the other three alloys at around 0.03 as shown in Figure 3.12a. However, the core plastic strain levels are much lower in 1020 at around 0.005 while the other three alloys have a strain of approximately 0.012. In the carburized condition, the bars are all predicted to bend downward with 4320 showing the lowest deflection of 0.2 mm (0.008 in) as seen in Figure 3.12b. The other three alloys show similar deflections of approximately 0.5 mm (0.020 in) between the two ends of the bar. Figure 3.12b also shows that 1020 is predicted to deflect upward from the left end then deflect downward about 100 mm (4 in) from the left end. This means 1020 has a greater overall distortion compared to the other alloys that only deflect downward the entire length. All four alloys show similar compressive residual hoop stress levels at the surface of the bar at about 300 MPa (43.5 ksi) as shown in Figure 3.12c. However, the of 1020 differs from that point on, maintaining and even increasing the compressive stress to about 2.5 mm (0.098 in) then becoming more tensile and staying tensile in the core. The peak tensile stress in 1020 is about 200 MPa (29.0 ksi) at a depth of approximately 6 mm (0.25 in). The other three alloys become more tensile starting at the surface and increasing to approximately the same maximum tensile
stress of 1020 at 200 MPa (29.0 ksi) at a depth of about 2.5 mm (0.01 in). The stress then
decreases and becomes slightly compressive in the core.

3.3.3 Carburized to 1.0 Wt Pct C Five Second Quench Hold Time

The simulated carburization process for the intensive quenching was the same as for the oil
quenching, therefore the carbon profiles are the same as previously presented in Figure 3.6
above. Figure 3.13 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX
steels when carburized to 1.0 wt pct C at the surface. The increased carbon content near the
surface has further increased the retained austenite levels to about 12% for 4320, 7% for 1020
and 8620 and 2% for 4120 as shown in Figure 3.13a. The core ferrite and pearlite levels are
similar to the previous case at up to 25% ferrite and 68% pearlite in 1020 and the other three
alloys have essentially no ferrite or pearlite as seen in Figure 3.13b and c. Bainite levels are also
similar as seen in Figure 3.13d. 8620 contains approximately 65% bainite followed by 4120
with about 50% and 4320 with about 45%. 1020 shows a peak about 2.5 mm (0.01 in) below the
surface of 20% and decreasing both to the surface and the core levels of about 10% for each.
The predicted martensite fraction is shown in Figure 3.13e and is similar to the last case. Again,
1020 contains some bainite at the surface thus reducing the amount of martensite that can form to
a peak of 85% just below the surface. The bainite and retained austenite levels result in a
decreasing martensite fraction closer to the surface. This effect is also seen in 4320 and 8620 but
is primarily a result of the higher retained austenite levels due to carburization and not higher
temperature transformation products.

Figure 3.14 shows the predicted effective plastic strain, centerline distortion and final
residual hoop stress. Similar to the last case, the effective plastic strain for all four alloys is
similar near the surface with 1020 having less strain in the core than the other three alloys as
shown in Figure 3.14a. All four alloys also predominately bend downward the entire length with
very little upward deflection shown (Figure 3.14b). 1020 shows the most deflection at about
0.8 mm (0.031 in) with the other three alloys showing similar deflection levels of 0.35 mm
(0.014 in). Figure 3.14c shows the surface residual stresses are similar for all four alloys at about
Figure 3.9 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched for 5 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.10  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 5 s under high intensity quenching.
Figure 3.11  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched for 5 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.12  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 5 s under high intensity quenching.
375 MPa (54.4 ksi) in compression. Again, 1020 differs from the other three alloys starting approximately 1.5 mm (0.059 in) below the surface. At that point, the compressive residual stress starts increasing to about 2.5 mm (0.098 in) below the surface then becoming more tensile and staying tensile in the core. Again, the peak tensile stress in 1020 is about 200 MPa (29.0 ksi) at a depth of approximately 7.5 mm (0.3 in). The other three alloys become more tensile starting at the surface and increasing to approximately 150 MPa (21.8 ksi) for 8620 at a depth of 4 mm (0.157 in). For 4120, the peak is approximately 200 MPa (29.0 ksi) at 6 mm (0.25 in), while 4320 peaks approximately at 250 MPa (36.3 ksi) at a depth of 6 mm (0.25 in). The stress then decreases for all three alloys and becomes slightly compressive in the core.

3.3.4 Non-carburized 10 Second Quench Hold Time

Figure 3.15 shows the final predicted microstructure of 10XX, 41XX, 43XX, and 86XX steels for the non-carburized condition under high intensity quenching with a hold time of 10 s. Similar to the 5 s hold time, Figure 3.15a shows that up to 1% austenite is retained in 4120, 4320 and 8620 near the surface. Figures 3.15b and c show that only 1020 contains any ferrite (23%) or pearlite (65%). Figure 3.15d show the bainite fraction for all alloys is less than 20%. For 1020, there are peaks of bainite at 20% about 2.5 mm (0.1 in) below the surface with about 7.5% bainite in the core. The other three alloys all have their peak bainite fractions in the bar centers. 8620 and 4120 peak around 10% while 4320 peaks around 15%. Due to the high heat transfer rates, the temperature can reach the martensite start temperature quickly, therefore leading to high martensite fractions as shown in 3.15e. Martensite in 1020 peaks around 90% at the surface then quickly falls due to the low martensite start temperature. The other alloys contain 99% martensite at the surface falling to 85% in the core of 4320 and 90% in the core of 4120 and 8620.
Figure 3.13 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched for 5 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.14  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 5 s under high intensity quenching.
Figure 3.16 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. Overall, 1020 shows a lower effective plastic strain as shown in Figure 3.16a. The majority of the bar is at 0.005 strain while the surface effects result in strain levels similar to that of the other alloys. 4120, 4320, and 8620 all show similar strain levels throughout the bars, with a strain of 0.02 at the top surface, 0.012 in the core and 0.017 at the bottom surface. Under this heat treatment, all four alloys have an upward centerline deflection. 1020 deflects the least at just under 0.2 mm (0.008 in). 8620 deflects the most at 0.9 mm (0.035 in), followed by 4120 with a deflection of 0.8 mm (0.031 in) and 4320 with a deflection of 0.7 mm (0.028 in). 1020 is predicted to have a very high compressive residual hoop stress at the surface as can be seen in Figure 3.16c. The hoop stress is predicted to be approximately 1050 MPa (150 ksi) and remains compressive to a depth of 8 mm (0.315 in). The high compressive stress near the surface results in the core of the bar being in tension of approximately 300 MPa (43.5 ksi) for the bar to be in equilibrium. The other three alloys show similar stress profiles starting at the surface in compression of approximately 600 MPa (87.0 ksi) going to a peak tensile stress of approximately 500 MPa (72.5 ksi) at a depth of 7.5 mm (0.3 in) then continuing to compression in the center of 400 MPa (58.0 ksi).

3.3.5 Carburized to 0.80 Wt Pct C 10 Second Quench Hold Time

Figure 3.17 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 0.8 wt pct C at the surface. Again, the carbon profile is the same as previously presented in Figure 3.3. Figure 3.17a shows the retained austenite levels are between 6 and 7% for 1020, 4320, and 8620 while 4120 is predicted to have approximately 2% retained austenite at the surface. Only 1020 is predicted to contain ferrite or pearlite with Figure 3.17b showing ferrite is predicted to be 25% and Figure 3.17c showing pearlite at a peak of 70%. Less than 20% bainite is predicted in all four alloys as shown in Figure 3.17d with 1020 showing the highest of 20% at a depth of about 2 mm (0.08 in). The core contains approximately 7.5% bainite. 8620 and 4120 show a peak of 10% while 4320 shows a peak of 15%, all in the core of the bars. Carburization affects all alloys as seen in Figure 3.17e. The added carbon delayed the phase transformation in 1020 so that martensite formed with the previously mentioned 7% retained austenite. The other alloys show similar levels with surface martensite at 94% for
Figure 3.15  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched 10 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.16  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 10 s under high intensity quenching.
8620 and 4320 while 4120 is at 98%. Core martensite levels are at 90% for 4120 and 8620 with 4320 at 85%.

Figure 3.18 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. The surface effective plastic strain is similar in all four alloys as shown in Figure 3.18a. However, 1020 has a much lower strain in the core compared to the other three alloys. Figure 3.18b shows a range of centerline deflection responses. 1020 starts bending up and eventually ends up with a downward deflection to 0.4 mm (0.016 in). 8620 shows a downward deflection of about 0.6 mm (0.024 in) with a curvature similar to 1020. 4120 has an overall downward deflection of 0.35 mm (0.014 in) with a slight concave upward shape. 4320 has a more pronounced concave up shape but also finishes with the smallest deflection at 0.17 mm (0.007 in). Residual stress levels for 1020 are predicted to be more compressive than the other alloys as shown in Figure 3.18c. The compressive hoop stress starts at 550 MPa (80.0 ksi) at the surface and becomes more compressive to approximately 725 MPa (105 ksi) at a depth of about 1.2 mm (0.050 in) then becomes more tensile to a peak of approximately 300 MPa (43.5 ksi) at a depth of 7.5 mm (0.30 in). 4120 has a surface compressive stress of 450 MPa (65 ksi) while 8620 and 4320 start closer to 350 MPa (51 ksi) then moves to tensile with a peak of 450 MPa (65 ksi) for 8620 up to 500 MPa (72.5 ksi) for 4320 with all three alloys reaching their maximum tensile hoop stress at a depth of 4 mm (0.16 in). Beyond this depth, the stresses decrease and became compressive in the core with a compressive stress of 450 MPa (65 ksi).

3.3.6 Carburized to 1.0 Wt Pct C 10 Second Quench Hold Time

Figure 3.19 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 1.0 wt pct C at the surface. Again, the carbon profile is the same as previously presented in Figure 3.6. With the increased carbon in near the surface, the retained austenite level has increased as shown in Figure 3.19a. 4320 contains approximately 11% austenite at the surface, 1020 contains approximately 8.5%, 8620 contains approximately 7% and 4120 contains the lowest amount of retained austenite with just over 2%. Also it can be seen that the decrease of retained austenite is much more gradual for 4120, 4320, and 8620 compared to 1020. As with the previous case, only 1020 is predicted to contain ferrite or pearlite. Ferrite is
Figure 3.17  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched 10 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.18  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 10 s under high intensity quenching.
approximately 25% while pearlite is approximately 68% at the peak as shown in Figures 3.19 b and c respectively. Bainite levels are shown in Figure 3.19d with 4120 and 8620 containing similar fractions peaking at 11% in the core. 4320 peaks in the core as well at 15% while 1020 contains two peaks near the case-core interfaces of approximately 20% falling to zero at the surfaces and to 6% in the core. The balance of the microstructure is martensite as shown in Figure 3.19e. 4120 and 8620 are very similar in the core with martensite approximately at 90%. Near the surface, the similarities disappear with 8620 containing more retained austenite and less martensite while 4120 contains 98% martensite and 2% retained austenite as previously mentioned. 4320 contains 85% martensite in the core and increases to the case core interface then falls with increasing retained austenite near the surface. 1020 also contains a peak martensite near the case-core interface and decreasing both toward the surface due to retained austenite and toward the core due to higher temperature transformation products.

Figure 3.20 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. The surface effective plastic strain is similar in all four alloys as shown in Figure 3.20a. Again, 1020 has a much lower strain in the core compared to the other three alloys. Downward centerline deflection is very small in 4320 (0.06 mm (0.002 in)) and 8620 (0.1 mm (0.004 in)). 4120 has more deflection at 0.5 mm (0.02 in) while 1020 shows the highest deflection at 1.0 mm (0.04 in). From Figure 3.20c, it can be seen that the residual hoop stress profiles for 4120, 4320, and 8620 are similar with surface compressive stresses of 400 MPa (58 ksi) for 4120 and 8620 and 525 MPa (76 ksi) in 4320. All three become less compressive and have tensile peaks of 465 MPa (67 ksi) in 8620, 475 MPa (69 ksi) in 4120 and 505 MPa (73 ksi) in 4320 all at a depth of 4 mm (0.16 in). Beyond this point, the tensile stress decreased and the core is again in compression. 1020 has similar surface compressive hoop stresses of 535 MPa (78 ksi, however the stress drops slightly at the next data point then become more compressive to 740 MPa (107 ksi) and then become more tensile and stay tensile in the core.
Figure 3.19  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched 10 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.20  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 10 s under high intensity quenching.
3.3.7 Non-carburized 15 Second Quench Hold Time

Figure 3.21 shows the final predicted microstructure of 10XX, 41XX, 43XX, and 86XX steels for the non-carburized condition under high intensity quenching with a hold time of 15 s. Again 4120, 4320 and 8620 all show a small fraction (1%) retained austenite near the surface as shown in Figure 3.21a. Figure 3.21b shows that only 1020 is predicted to contain any ferrite with about 23% in the core. Pearlite is also only predicted to be found in 1020 with the peaks at 65%. Predicted bainite levels are shown in Figure 3.21c. 1020 shows bainite peaks at 18% at a depth of 2 mm (0.08 in) falling to 3% at the surface and 7% in the core. 4120, 4320, and 8620 all have their peak bainite in the core with peaks of 10%, 15% and 10% respectively. With so little of the microstructure transformed into higher temperature products, the microstructures are predominately martensite for 4120, 4320, and 8620 as shown in Figure 3.21e. 4120 and 8620 contain 90% martensite in the core while 4320 contains 85% martensite in the core with all three alloys increasing to 99% at the surface. Due to the other transformation products, 1020 contains 91% martensite at the surface falling to 5-7% in the core.

Figure 3.22 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. Effective plastic strain at the surface is similar in all four alloys while 1020 has a lower strain level in the core as seen in Figure 3.22a. All four alloys are predicted to deflect upward as a result of the quench as seen in Figure 3.22b. 1020 is predicted to deflect the least at 0.2 mm (0.01 in) followed by 4320 with a deflection of 0.7 mm (0.027 in), 4120 with a deflection of 0.8 mm (0.03 in) and 8620 with the highest deflection of 0.9 mm (0.037 in). Residual stresses are shown in Figure 3.22c. As seen, 1020 has a very large compressive residual hoop stress at the surface of 1070 MPa (155 ksi) which becomes tensile approximately 4.5 mm (0.18 in) below the surface and peaks in tension at 350 MPa (51 ksi). The other three alloys have similar stress levels at 590 MPa (86 ksi), 625 MPa (91 ksi), and 600 MPa (87 ksi) for 4120, 4320, and 8620 respectively at the surface. Each then becomes tensile approximately 1.25 mm (0.05 in) below the surface and peaks in tension at 4 mm (0.16 in) below the surface at 475 MPa (69 ksi) for 4120 and 8620 or 500 MPa (73 ksi) for 4320. The stress then decreases and becomes tensile again in the core.
Figure 3.21 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched for 15 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.22  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 15 s under high intensity quenching.
3.3.8 Carburized to 0.80 Wt Pct C 15 Second Quench Hold Time

Figure 3.23 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 0.8 wt pct C at the surface under high intensity quenching with a hold time of 15 s. As is expected, carburization has increased austenite stability near the surface due to the increased carbon content as seen in Figure 3.23a. 1020 is predicted to retain 9% austenite while 4120, 4320 and 8620 are predicted to contain 3%, 8% and 7% respectively at the surface. As seen in Figures 3.23b and c, only 1020 is predicted to contain ferrite or pearlite with a peak ferrite content of 25% in the core and a peak pearlite of 68% 5 mm (0.20 in) below the surface. Predicted bainite is shown in Figure 3.23d. Bainite is predicted to be 10% for 4120 and 8620 while 4320 is predicted to be 15% in the core. 1020 has bainite peaks of 20% 2 mm (0.08 in) below the surface and core content of 6%. Conversely to bainite, martensite is a minimum in the core at 90% for 4120 and 8620 and 85% for 4320 as seen in Figure 3.23e. Surface martensite levels are also lowered with the balance being the retained austenite seen in Figure 3.23a. 1020 has a peak in martensite at 96% and a depth of 0.75 mm (0.03 in) due to retained austenite nearer the surface and higher temperature transformations deeper in the bar.

Figure 3.24 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. All four alloys are predicted to have similar effective plastic strains near the surface with 1020 having lower strain in the core compared to the others as shown in Figure 3.24a. 1020 shows a slight upward deflection near the left end followed by a large downward deflection at the right end of 0.7 mm (0.28 in) leading to a curved shape as seen in Figure 3.24b. The other three alloys are predicted to have much less curvature with final downward deflections of 0.5 mm (0.02 in), 0.4 mm (0.016 in), and 0.5 mm (0.02 in) for 4120, 4320, and 8620 respectively. Residual hoop stresses are shown in Figure 3.24c. 4120, 4320, and 8620 show similar profiles with compressive surface residual hoop stresses of 500 MPa (73 ksi), 380 MPa (55 ksi), and 350 MPa (51 ksi) respectively. They each have a tensile peak at a depth of 4 mm (0.16 in) and tensile stresses of 460 MPa (67 ksi), 485 MPa (70 ksi), and 485 MPa (70 ksi) for 4120, 4320, and 8620 respectively. From these peaks, the residual hoop stress level falls and becomes compressive in the core. 1020 shows a higher surface compressive hoop stress of 690 MPa (100 ksi) followed by an increase in compression to 750 MPa (109 ksi) at a depth of 1 mm (0.04 in). Beyond this point, the stresses decrease and become tensile with a peak of 300 MPa (44 ksi) at 7 mm (0.28 in) below the surface.
Figure 3.23  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched for 15 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.24  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 15 s under high intensity quenching.
3.3.9 Carburized to 1.0 Wt Pct C 15 Second Quench Hold Time

Figure 3.25 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 1.0 wt pct C at the surface under high intensity quenching with a hold time of 15 s. With the increased carbon content from carburization, there is more retained austenite compared to the last case as seen in Figure 3.25a. 1020 is predicted to have 11% austenite at the surface while 4120, 4320, and 8620 contain 3%, 13%, and 9% respectively. 1020 also is the only alloy to have ferrite or pearlite predicted in the microstructure as shown in Figures 3.25b and c. Ferrite peaks in the core at 25% while pearlite peaks at 68% 6 mm (0.25 in) below the surface. Bainite is similar to the previous case with predicted core levels of 11% for 4120 and 8620 while 4320 contains 15%. 1020 peaks 20% at a depth of 2 mm (0.08 in) and a core level of 6%. Martensite levels peak around the case-core interface around 98% for 4120, 4320, and 8620 as shown in Figure 3.25e. Core levels drop to 89% for 4120 and 8620 while 4320 contains 85% martensite. 1020 shows a peak of 95% 1 mm (0.04 in) below the surface and falling sharply to approximately 5% in the core. Surface levels also drop from the peaks due to retained austenite as shown.

Figure 3.26 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. Again, all four alloys are predicted to have similar effective plastic strains near the surface with 1020 having lower strain in the core compared to the others as shown in Figure 3.26a. Centerline deflection is shown in Figure 3.26b with 1020 showing the largest deflection at 1.1 mm (0.04 in) with a curvature as well. The other three alloys are fairly straight with smaller deflections of 0.45 mm (0.18 in), 0.05 mm (0.002 in), and 0.08 mm (0.003 in) for 4120, 4320, and 8620 respectively. Residual stress levels for 1020 are again significantly different than the other three alloys as shown in Figure 3.26c. Compressive hoop stresses are predicted at the surface of 680 MPa (99 ksi) followed by a dip then increase to 740 MPa (107 ksi) at a depth of 1.4 mm (0.06 in). After this point, the hoop stress decrease and become tensile in nature peaking around 325 MPa (47 ksi) at a depth of 7 mm (0.28 in) and decreasing somewhat in the core. Residual hoop stresses are compressive at the surface for 4120, 4320, and 8620, measuring 530 MPa (77 ksi), 450 MPa (65 ksi), and 420 MPa (61 ksi) respectively. The stresses decrease and become tensile with a peak at 3 mm (0.12 in) below the surface reaching approximately 500 MPa (73 ksi) for all three alloys. Below this depth, the tensile stresses decrease and become compressive in the core.
Figure 3.25  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched for 15 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.26  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 15 s under high intensity quenching.
3.3.10 Non-carburized 20 Second Quench Hold Time

Figure 3.27 shows the final predicted microstructure of 10XX, 41XX, 43XX, and 86XX steels for the non-carburized condition under high intensity quenching with a hold time of 20 s. Similar to the shorter hold times, 4120, 4320, and 8620 all are predicted to have retained austenite near the surface as shown in Figure 3.27a. Also similar to the short quench times, only 1020 is predicted to have any significant ferrite or pearlite as shown in Figures 3.27b and c with ferrite peaking at 23% in the core and pearlite peaking at 66% at a depth of 8 mm (0.3 in). Bainite peaks in the core at 11%, 15%, and 10% for 4120, 4320, and 8620 respectively as shown in Figure 3.27d. For 1020, bainite is predicted to peak at 17% 2 mm (0.08 in) below the surface while the core contains 7%. Martensite makes up a large fraction of the microstructure for 4120, 4320, and 8620 with 89%, 85%, and 90% respectively in the core while surface levels are 99% for all three alloys as shown in Figure 3.27e. 1020 peaks with 91% martensite near the surface falling to a low of 5% in the core.

Figure 3.28 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. All four alloys are predicted to have similar effective plastic strains near the surface with 1020 having lower strain in the core compared to the others as shown in Figure 3.28a. Figure 3.28b shows that all four alloys deflect upwards with 1020 showing the least amount of deflection at 0.2 mm (0.008 in). The deflections for 4120, 4320, and 8620 are 0.8 mm (0.03 in), 0.7 mm (0.028 in) and 0.9 mm (0.036 in) respectively. The residual hoop stresses are shown in Figure 3.28c. 1020 has the most compressive stresses at the surface at 1100 MPa (160 ksi) which then decreases and has a tensile peak of 325 MPa (47 ksi) at a depth of 8 mm (0.31 in) and stays tensile in the core of the bar. The other three alloys have compressive residual hoop stress of 600 MPa (87 ksi) for 4120 and 8620 while 4320 is predicted to be at 625 MPa (91 ksi). The stresses decrease and have a tensile peak at a depth of 4 mm (0.16 in) and stress levels of 475 MPa (69 ksi), 500 MPa (73 ksi), and 450 MPa (65 ksi) for 4120, 4320, and 8620 respectively. After this point, the stresses decrease and are compressive in the core.
Figure 3.27  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched for 20 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.28 Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 20 s under high intensity quenching.
3.3.11 Carburized to 0.80 Wt Pct C 20 Second Quench Hold Time

Figure 3.29 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 0.8 wt pct C at the surface under high intensity quenching with a hold time of 20 s. Figure 3.29a shows that 1020 is predicted to contain 10% retained austenite while 4120, 4320, and 8620 contain 3%, 9%, and 9% respectively. The peaks are generally at the surface except for 4120 which is approximately constant to a depth of 0.75 mm (0.03 in) then drops with the other three alloys. Only 1020 is predicted to contain ferrite or pearlite as shown in Figures 3.29b and c with ferrite peaking in the core at 25% and pearlite peaking 6 mm (0.24 in) below the surface at a level of 68%. Bainite content is shown in Figure 3.29d. Bainite peaks in the core for 4120, 4320, and 8620 at 11%, 15%, and 10% respectively. 1020 shows peaks at a depth of 2 mm (0.08 in) and 20% with 6% in the core. With the increased carbon content and retained austenite, the martensite fraction at the surface is 90%, 97%, 91%, and 91% for 1020, 4120, 4320, and 8620 respectively. Levels increase to peaks of 98% near the case-core interface at a depth of 1.35 mm (0.05 in) for 4120, 4320, and 8620 while 1020 peaks at 96% martensite at a depth of 0.75 mm (0.03 in). Core levels are lower with 89%, 85%, and 90% for 4120, 4320, and 8620 respectively while 1020 contains approximately 5% in the core.

Figure 3.30 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. All four alloys are predicted to have similar effective plastic strains near the surface with 1020 having lower strain in the core compared to the others as shown in Figure 3.30a. Figure 3.30b shows that all four alloys deflect downwards with 1020 showing the highest amount of deflection at 7 mm (0.28 in). The deflections for 4120, 4320, and 8620 are 0.5 mm (0.02 in), 0.4 mm (0.016 in) and 0.5 mm (0.02 in) respectively. 1020 also is shown to have a large amount of curvature while the other alloys are seen to have little if any curvature. The residual hoop stresses are shown in Figure 3.30c. 1020 has the most compressive stresses at the surface at 750 MPa (109 ksi) which stays approximately constant to a depth of 1 mm (0.04 in) and then decreases and has a tensile peak of 305 MPa (44 ksi) at a depth of 7 mm (0.28 in) and stays tensile in the core of the bar. The other three alloys have compressive residual hoop stress of 500 MPa (73 ksi), 400 MPa (58 ksi), and 350 MPa (51 ksi) for 4120, 4320, and 8620 respectively. The stresses decrease and have a tensile peak at a depth of 4 mm (0.16 in) and stress levels of 460 MPa (67 ksi), 480 MPa (70 ksi), and 450 MPa (65 ksi) for 4120, 4320, and 8620 respectively. After this point, the stresses decrease and are compressive in the core.
Figure 3.29 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched for 20 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.30  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 20 s under high intensity quenching.
3.3.12 Carburized to 1.0 Wt Pct C 20 Second Quench Hold Time

Figure 3.31 shows the final predicted microstructure of 10XX, 41XX, 43XX and 86XX steels when carburized to 1.0 wt pct C at the surface under high intensity quenching with a hold time of 20 s. Figure 3.31a shows that 1020 is predicted to contain 12% retained austenite while 4120, 4320, and 8620 contain 3%, 15%, and 10% respectively. The peaks are generally at the surface except for 4120 which is approximately constant to 1 mm (0.04 in) then drops with the other three alloys. Only 1020 is predicted to contain ferrite or pearlite as shown in Figures 3.31b and c with ferrite peaking in the core at 25% and pearlite peaking 6 mm (0.24 in) below the surface at a level of 68%. Bainite content is shown in Figure 3.31d. Bainite peaks in the core for 4120, 4320, and 8620 at 11%, 15%, and 11% respectively. 1020 shows peaks at 20% and a depth of 2 mm (0.08 in) with 6% in the core. With the increased carbon content and retained austenite, the martensite fraction at the surface is 88%, 97%, 85%, and 90% for 1020, 4120, 4320, and 8620 respectively. Levels increase to peaks of 98% near the case-core interface at a depth of 1.35 mm (0.05 in) for 4120 and 4320, 2 mm (0.08 in) for 8620 while 1020 peaks at 95% martensite at a depth of 1 mm (0.04 in). Core levels are lower with 89%, 85%, and 89% for 4120, 4320, and 8620 respectively while 1020 contains approximately 4% in the core.

Figure 3.32 shows the predicted effective plastic strain, centerline distortion and final residual hoop stress. All four alloys are predicted to have similar effective plastic strains near the surface with 1020 having lower strain in the core compared to the others as shown in Figure 3.32a. Figure 3.32b shows that all four alloys deflect downwards with 1020 showing the highest amount of deflection at 1.1 mm (0.04 in). The deflections for 4120, 4320, and 8620 are 0.45 mm (0.018 in), 0.05 mm (0.002 in) and 0.08 mm (0.003 in) respectively. 1020 also is shown to have a large amount of curvature while the other alloys are seen to have little if any curvature. The residual hoop stresses are shown in Figure 3.32c. 1020 has the most compressive stresses at the surface at 750 MPa (109 ksi) which decreases somewhat then returns to approximately 750 MPa (109 ksi) at a depth of 1.35 mm (0.05 in) and then decreases and has a tensile peak of 325 MPa (47 ksi) at a depth of 7 mm (0.28 in) and stays tensile in the core of the bar. The other three alloys have compressive residual hoop stresses of 550 MPa (80 ksi), 475 MPa (69 ksi), and 425 MPa (62 ksi) for 4120, 4320, and 8620 respectively. The stresses decrease and have a tensile peak at a depth of 4 mm (0.16 in) and stress levels of 500 MPa (73 ksi) for all three alloys. After this point, the stresses decrease and are compressive in the core.
Figure 3.31  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched for 20 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite. Distances are from the top of the bar.
Figure 3.32  Predicted results for (a) effective plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 20 s under high intensity quenching.
3.4 Selection of High Intensity Quench Hold Time

In order to approach the requirements of the intensive quenching process which was the inspiration for the high intensity quenching simulations along with the need to reduce the volume of data examined, an “ideal quench hold time” needs to be selected. Based on the literature as shown in Chapter 1, this “ideal quench hold time” should be the quench hold time that produces the most compressive near-surface residual stresses at the end of the heat treatment. The residual stress profiles of the various steel grades were shown earlier for the various hold times during high intensity quenching and are consolidated here by material in Figures 3.33-3.34. Figure 3.33a shows the predicted residual stress levels for non-carburized 1020 while Figure 3.33b and 3.33c show the residual stress profiles for carburized to 0.8 wt pct C and 1.0 wt pct C respectively. This data indicates that the residual stress profiles for hold times of 10 s, 15 s, and 20 s are quite similar.

Figure 3.34a shows the predicted residual hoop stress for non-carburized 4120, while Figures 3.34b and 3.34c are carburized to 0.8 wt pct C or 1.0 wt pct C respectively. Similar to 1020, the results for 4120 also show that the residual stress profiles of quench hold times of 10 s, 15 s, and 20 s are quite similar. 4320 and 8620 also behave similarly and the simulation results are shown in Appendix D.

Due to these similarities in residual stress profiles for the three longer quench times, an “ideal quench time” cannot be selected based on what was shown. Further analysis of the data is required. Figures 3.35-3.36 show differences between residual stress profiles for 15 and 20 second quench times as compared to the 10 second quench time for 1020 and 4120 while the data for 4320 and 8620 are contained in Appendix D. To create these graphs, the residual stress profile for 10 second quench was subtracted from either the 15 or 20 second quench residual stress profile.

Figure 3.35 shows the variation in residual stress profiles for 1020. In Figure 3.35a, the variation between 10 s and 20 s quench times is minimal. For non-carburized 1020, the variation is greatest with the 15 s quench time. When 1020 is carburized to 0.8 wt pct C (Figure 3.35b), differences between residual hoop stresses for 15 s and 20 s quench times are greatly reduced. There is still a large variation from 10 s quench time that the “ideal quench time” can be considered to be 15 s. Similar results are shown in Figure 3.35c for carburization to 1.0 wt pct C.
Figure 3.36a shows that for 4120 steel, the variation for the non-carburized bar from the increased quench time is small and that the results for the 15 and 20 second holds almost overlap. For the 0.8 wt pct C carburized 4120 bar, Figure 3.36b shows more variation from the 10 second quench for the 15 second quench while variation between the 15 and 20 second quench times is again small. Similarly for the 1.0 wt. pct. C carburized 4120 bar, Figure 3.36c shows more variation from the 10 second quench for the 15 second quench while variation between the 15 and 20 second quench times is small. 4320 and 8620 show similar results and can be seen in Appendix D. From these results, it was concluded that the 15 second quench would be the appropriate quench time to use for comparisons to other quench methods.

Figure 3.33  Comparison of the residual hoop stress for the intensive quench simulations for 1020 in three carburized conditions. (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface. The times indicated are the quench time in the high intensity quenching process.
Figure 3.34  Comparison of the residual hoop stress for the intensive quench simulations for 4120 in three carburized conditions.  (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface.  The times indicated are the quench time in the high intensity quenching process.
Figure 3.35  Deviation of the residual stress of 15 and 20 second hold times from the 10 second hold time for high intensity quenching of 1020 in three carburized conditions.  (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface.  The times indicated in the legend are the quench time in the intensive quenching process.
Figure 3.36 Deviation of the residual stress of 15 and 20 second hold times from the 10 second hold time for high intensity quenching of 4120 in three carburized conditions. (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface. The times indicated in the legend are the quench time in the intensive quenching process.
3.5  **Time Development of 1020**

This section will discuss the time dependent thermal profiles, microstructural changes, residual stress, centerline deflection, and effective plastic strain changes for 1020. Oil quenching will be considered first for the three carburization levels followed by the 15 s high intensity quench simulations.

3.5.1  **Oil Quenched Non-Carburized 1020**

Figure 3.37a shows the temperature changes as a function of time. The time is from the start of the immersion step in the heat treatment. As can be seen in Figure 3.37b, the ferrite has started to form at the bottom surface of the bar before the immersion has finished at 4.3 seconds into the immersion. This is the only microstructural constituent that has started forming. As the bar cools further and the immersion step ends and the quench step begins, ferrite continues to transform near the bottom surface of the bar (Figure 3.37b). Austenite has also started to decompose into pearlite near the bottom surface. As the ferrite and pearlite start to transform, the hoop stress at the surface increases between the 4.3 s and 5.0 s time steps. The centerline deflection also reaches a peak at 5.0 s (Figure 3.37g) while the effective plastic strain increases throughout the bar (Figure 3.37h).

At 5.8 s into the immersion/quench, bainite starts to form at the lower surface while ferrite and pearlite continue to increase at the lower surface. At this point in the quench, the residual hoop stress at the lower surface reverses and becomes compressive and the centerline deflection has decreased from its maximum. Austenite at the top surface starts to decompose into both ferrite and pearlite at 7.1 s into the quench. Bainite transformation continues slightly at the lower surface. Residual stresses at both the top and bottom surfaces become compressive with tensile peaks subsurface. Centerline deflection also decreases further so that total deflection is less than 1 mm (0.04 in). Due to the progressing transformations, the effective plastic strain increases near the surface with a fairly large change 5 mm (0.20 in) from the bottom surface. This depth corresponds to the leading edge of the ferrite transformation and is also near the peak in tensile hoop stresses for this time step.

At 10 s, the ferrite and pearlite progress further into the bar while the surface levels have remained essentially constant. Bainite has formed to higher fractions at the lower surface as well
as forming near the top surface. The hoop stresses have become compressive to a greater depth with the core in tension and the effective plastic strain has increased in the core and near the top surface. At 14.6 s, the temperature has almost reached the martensite start ($M_s$) temperature. Ferrite has formed to almost the final fraction in the core while pearlite has started transforming in the core with approximately 40% of the core microstructure as austenite. Bainite has essentially reached its final content with just over 20% near the surface and decreasing away from the surface. With the large change in microstructure in the core along with the resulting volume change, the hoop stress near the surface has become tensile with peaks in the hoop stress subsurface and a core in compression. The centerline deflection is almost zero at this point while the effective plastic strain has increased to slightly more in the core to approximately the final level.

At 23.9 s, the temperature has dropped below the $M_s$ temperature for the entire bar however there is very little austenite left to transform. Ferrite and pearlite complete their transformation in the core and a very small fraction martensite forms with approximately 2% at the top and 1% at the bottom of the bar. This difference is a result of the immersion process affecting the temperature at the bottom of the bar first and allowing the ferrite and pearlite transformation to progress further at the bottom than at the top of the bar. The residual stress has again become compressive at the surface while the tension peaks have decreased but remain. The core has also remained in compression.

The temperature in the entire bar is almost uniform at just over 100 °C (212 °F) at 114.0 s into the quench. The microstructure is essentially unchanged with only slight (less than 1%) changes to bainite and martensite. Due to the continued cooling and thermal contraction, the residual hoop stress near the surface becomes more compressive while the tensile peaks decrease and the core becomes less compressive. Centerline deflection predicted to be 0.36 mm (0.014 in) upward.
Figure 3.37  Time dependent response of non-carburized 1020, quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.37  (continued) Time dependent response of non-carburized 1020, quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.37  (continued) Time dependent response of non-carburized 1020, quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
3.5.2 Oil Quenched 1020 Carburized to 0.8 Wt Pct C

Unlike the non-carburized condition where the \( M_S \) temperature is a constant, the \( M_S \) temperature varies with carbon content near the surface as seen in Figure 3.38a. The predicted volume fractions of ferrite, pearlite, bainite, and martensite are shown in Figures 3.38b-e respectively. Residual hoop stress, centerline deflection and effective plastic strain are shown in Figures 3.38f-h respectively.

Due to the carburization, austenite decomposition at the surface is delayed and transformation starts subsurface. The immersion step is complete and the quench starts at 5.0 s. At this time, ferrite has just started to transform with a peak of approximately 1% at a depth of 1.35 mm (0.05 in). There is no other microstructure transformation occurring, the residual stress is tensile at the surface while the core is in compression. The centerline deflection is at the peak due to the immersion step; the effective plastic strain is higher at the surface at about 0.011 which is higher than the final point of oil quenched non-carburized 1020. At 5.6 s, ferrite and pearlite have continued to transform near the bottom of the bar. Bainite has also started to transform near the lower surface with a peak approximately 1.35 mm (0.05 in) below the surface of 15%. The hoop stress at the lower surface is approximately 200 MPa (29.0 ksi) tension then quickly decreases to 135 MPa (19.6 ksi) in compression at a depth of 0.8 mm (0.03 in). The stress then reverses again to tension. This variation is due to the phase transformation occurring. The centerline deflection is decreased while the core effective plastic strain has increased along with the strain near the top surface.

Ferrite, pearlite, and bainite have started transformation near the top surface as well as continued to transform near the bottom of the bar at 9.6 s. The core has not started to transform. Again, due to the timing of the austenite decomposition, the surface residual hoop stress is tensile while just below the surface, the hoop stress just below the surface is compression. This compressive zone has expanded compared to 6.7 s while the core is in tension. The centerline deflection has reversed and the bar is slightly bent upward. The effective plastic strain is essentially unchanged at the surface and has increased slightly in the core. At 14.2 s, the core of the bar has started to transform to ferrite and pearlite. The residual hoop stress is tensile near the surface and compressive at the core. This change is due to the core phase transformation. It should be noted that the hoop stress is approximately 200 MPa (29.0 ksi) at the surface then decreases to 15 MPa (2.0 ksi) at a depth of 0.45 mm (0.02 in) and then increases again at a depth
of 3 mm (0.12 in). The centerline deflection has decreased slightly from the previous time step due to a higher fraction of the austenite transforming in the top half of the bar compared to the bottom half.

By 22.6 s, the ferrite and pearlite have essentially completed transforming in the core of the bar. While not shown for clarity, the bainite fraction is approaching the fraction seen in the 47.2 s time step. The temperature profile finally reaches the $M_S$ temperature in the mid-radius area. However, the mid-radius volume of the bar that has reached the $M_S$ has already transformed into mostly ferrite and pearlite. The residual hoop stresses are tensile at the surface then quickly become compressive at approximately 150 MPa (21.8 ksi) at a depth of 0.45 mm (0.02 in). From this point the stresses become tensile and peak at a depth of 3-4 mm (0.12-0.16 in). The core is maintained in compression. This profile is the result of the timing of the phase transformation where the previously transformed material constrains the austenite about to transform leading to compressive stress. As the austenite incrementally decomposes, the resultant volume change places the previously transformed material in tension. The austenite at the surface is under tension due to the volume of previously transformed austenite. The centerline deflection has increased upward to above the deflection seen at 9.6 s into the quench.

At 47.2, the entire bar except the surface has reached the $M_S$ temperature. The carbon introduced through the carburization has delayed the final transformation to favor martensite at and just below the surface. At this step, martensite is seen forming subsurface. This causes the remaining austenite at the surface to be placed into higher tension while the material subsurface (0.45 mm (0.02 in)) is compressed more with the martensite transformation. This added compression has lessened the tensile forces at the mid-radius slightly as well as decreasing the core compressive stresses. The final time step shown is at 259.0 s into the quench. At this point, the temperature profile is essentially flat and approaching the 65 °C (149 °F) of the quench oil. The entire bar is well below the $M_S$ temperature leading to a martensite fraction near the surface of 69% and 55% for the top and bottom surfaces respectively. This large variation in the martensite fractions is due to the immersion step. This step in the quench allowed the top of the bar to stay at a high temperature while the bottom of the bar was brought into the temperature range where more pearlite and bainite could form. If the immersion step were eliminated, the microstructural development would be the same when viewed from either the top or the bottom.
leading to symmetrical residual hoop stress and effective plastic strain data. This would also decrease the centerline deflection seen during the quench.

![Graphs showing temperature and calculated fractions of ferrite, pearlite, bainite, martensite, residual hoop stress, centerline deflection, and effective plastic strain.]

Figure 3.38 Time dependent response of 1020 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.38 (continued) Time dependent response of 1020 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.38 (continued) Time dependent response of 1020 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
3.5.3 15 s High Intensity Quench Hold Time Non-Carburized 1020

When the quench method is changed from oil quenching to high intensity quenching, the response in the material changes with the increased heat transfer rate out of the bar. Figure 3.39 shows the time development of the temperature, microstructure, residual hoop stress, centerline deflection and effective plastic strain for non-carburized and 15 s high intensity quenching of 1020. Figure 3.39a shows the temperature development is drastically different compared to the oil quenching. With the increased heat transfer rates, the temperature drops off rapidly at the surface that is in contact with the quench. As the immersion step progresses, the bottom of the bar reaches the $M_s$ temperature before the bar is fully immersed. The core temperature has remained almost unchanged and has a visible temperature change 1.8 s into the immersion/quench.

At 0.1 s into the immersion, both ferrite and pearlite have started to transform at the lower surface. The residual hoop stress has started to increase at the bottom surface while the rest of the bar is essentially stress free. The bar has started to curve downward due to the thermal contraction due to cooling. With the thermal contraction and the small fraction of austenite decomposition, the effective plastic strain is 0.012 at the bottom surface and decreases to zero at a depth of 2 mm (0.08 in). The top surface of the bar has started to cool considerably at 0.4 s into the immersion/quench. There is no predicted austenite decomposition though. The austenite near the bottom surface of the bar continues to transform to ferrite and pearlite while bainite has also started to transform at the lower surface. The hoop stress has increased and became more tensile at both the upper and lower surfaces of the bar. With the additional cooling of the bar, the centerline deflection increased to a maximum toward the bottom of the bar or opposite to the direction of immersion. The cooling at the upper surface has increased the effective plastic strain to 0.009 while the strain at the bottom of the bar has increased in magnitude and depth.

The temperature has decreased below the $M_s$ at 0.8 s into the immersion/quench at both the top and the bottom surfaces. Ferrite, pearlite, and bainite continue to grow towards the center of the bar from the bottom as well as starting to transform near the top surface. With the temperature now below the $M_s$, martensite has also started to transform at the surface of the bar with the lower surface containing 80% and the upper surface at 31% martensite. These changes in the microstructure give rise to compressive residual hoop stresses near the surfaces with
tensile peaks approximately 0.45 mm (0.018 in) and 0.75 mm (0.03 in) below the top and bottom surfaces respectively. The increased depth from the bottom surface is a result of the higher martensite fraction seen at the lower surface. Also with the martensite formation at the lower surface, the associated volume expansion has decreased the centerline deflection. The effective plastic strain is approaching the peak levels near the surface while the core strain has increased.

The core temperature has started to decrease at 1.8 s into the immersion/quench. Ferrite, pearlite and bainite continue to grow towards the center of the bar while martensite fractions near the surface have increased both in magnitude and depth. With the continued increase in martensite, the surface residual hoop stress has become more compressive at the surface and to a greater depth compared to 0.8 s. The tensile peaks have also decreased in value while moving towards the center. This leads to a greater volume in tension to stay in equilibrium. The centerline deflection has reduced further with both thermal contraction and martensite transformation playing a role. The effective plastic strain has continued to increase overall with a larger increase in the vicinity of the ongoing phase transformation.

The austenite decomposition into ferrite, pearlite, bainite, and martensite continues at 4.6 s. The core temperature has fallen to approximately 700 °C (1300 °F) which has allowed some ferrite to start forming. With the new martensite that has transformed subsurface, the surface compressive hoop stress has decreased somewhat while the volume in tension has been pushed to the center of the bar. With the start of the core transformation, the effective plastic strain has increased in the core. As the cooling continues, more of the bar reaches the Ms, which allows martensite to form at greater depths at 7.7 s into the quench. Ferrite, pearlite, and bainite have all transformed in the core while martensite is excluded due to temperature. The compressive stress at the surface has increased with peak tensile hoop stress closer to the core and compressive stresses are found in the core. The increase in compressive hoop stresses at the surface could partially be the result of the thermal contraction further below the surface. This contraction would effectively pull the outer layer of martensite inward, increasing the compressive hoop stress at the surface. The centerline deflection reverses direction and becomes slightly upward while the effective plastic strain in the core increases slightly to essentially the final value.

The quench time is almost finished at 14.1 s. The temperature of the entire bar is below 100 °C (212 °F). At this point, the microstructure is almost set with only martensite formation
occurring in the core with slight increases closer to the surfaces. The residual hoop stress near the surface has become more compressive while the core has become more tensile. With the martensitic transformation in the core, this change in the residual hoop stress is counterintuitive. For the moment, ignore the martensitic transformation. If the microstructure were unchanged from the previous time step, the hard martensitic shell would be holding the bar with very minimal dimensional changes as the cooling continued. This cooling results in a contraction in the core, which had been at a much higher temperature than the surface. The thermal contraction would then try to pull inward against the stiff martensitic shell resulting in tensile forces in the core and compression in the case. Now if we add the martensitic transformation back, the austenite would expand on transformation thereby reducing the tensile stresses in the core while also reducing the forces exerted by the core on the case thereby lessening the overall magnitude of the residual stresses the part sees. The thermal cooling continues to the final time step shown of 23.1 s. This time step is well within the air-cooling regime of the heat treatment. This much lower cooling rate at the surface has allowed the warm core to cool while the surface temperature increased. This final cooling further contracts the core of the bar resulting in higher tension in the core and a higher compressive hoop stress at the surface.

3.5.4 15 s High Intensity Quench Hold Time 1020 Carburized to 0.8 Wt Pct C

The time development for 1020 carburized to 0.8 wt pct C and quenched for 15 s using high intensity quenching is similar to the non-carburized condition. The time development of the temperature, ferrite, pearlite, bainite, martensite, residual hoop stress, centerline deflection, and effective plastic strain is shown in Figures 3.40a-h respectively. As was seen with the oil quenching, the carburization has changed the $M_s$ temperature profile as expected. However, the high intensity quenching creates such a high temperature gradient in the bar that the carburized layer all transforms to martensite at nearly the same time.

The first time step into the immersion/quench, 0.1 s, shows that the lower surface of the bar has cooled to below 500 °C (925 °F). The incubation time of ferrite and pearlite has not been reached and therefore no phase transformation has occurred. The residual hoop stress is essentially flat with tensile stresses at the bottom surface. The bar is also predicted to be curved downward due to the cooling and thermal contraction along the bottom surface. At 0.4 s into the immersion/quench, the upper surface of the bar has also started cooling was slightly above
600 °C (1100 °F) while the lower surface has reached a temperature of approximately 325 °C (615 °F). This reduced temperature and increased time has allowed a very small fraction (<0.5 %) ferrite.

Figure 3.39 Time dependent response of non-carburized 1020, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.39  (continued) Time dependent response of non-carburized 1020, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.39 (continued) Time dependent response of non-carburized 1020, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
and pearlite to form. The residual stress has become tensile at the top surface as well as increasing in magnitude at the lower surface to account for the constraint of the core on the contracting surfaces. The centerline deflection has reached the peak deflection, which is due to the immersion process cooling the bottom thus causing the material to contract along the bottom and bend the bar downward toward the cooler surface. With the cooling at the surfaces, the effective plastic strain has increased at the surfaces while remaining essentially zero in the core.

At the 0.7 s into the quench, the surfaces would have reached the $M_s$ temperature for the non-carburized condition. As mentioned earlier, the $M_s$ has been reduced at the surfaces due to the carburization leading to a delay in the phase transformation. Austenite has continued to decompose into ferrite and pearlite while also forming bainite with all transformations occurring predominately at the lower surface. With the small fraction of phase transformation occurring along with the large thermal gradient, the residual hoop stress levels continue to become more tensile at the surfaces with the core in compression from the thermal contraction of the case. The effective plastic strain has increased across the entire bar with the upper surface having the highest strain of approximately 0.025.

The bottom surface of the bar has just reached the $M_s$ temperature 1.8 s into the quench while the upper surface is approximately 10 °C (18 °F) above the $M_s$ temperature. These temperatures allow the ferrite, pearlite, and bainite to continue to form to greater fractions and depths near lower surface of the bar while also starting to transform near the upper surface. Martensite is not shown to be predicted at the surface yet, most likely due to the microstructure data being collected at the elemental level while the thermal data was collected at the nodal level. As was discussed earlier, this leads to the microstructure data from the nodes being averaged to give elemental data. If data were available from the surface nodes, it is expected a small fraction of martensite at the surface would be predicted.

3.6 Time Development of 4120

This section will show the time dependent thermal profiles, microstructural changes, distortion, and residual stress changes for 4120. Oil quenching will be considered first for the three carburization levels followed by the 15 s high intensity quench results.
Figure 3.40 Time dependent response of 1020 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.40  (continued) Time dependent response of 1020 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
Figure 3.40  (continued) Time dependent response of 1020 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) effective plastic strain.
3.6.1 Oil Quenched Non-Carburized 4120

Figure 3.41a shows the temperature change through the bar as a function of time. As would be expected with the higher alloy content, the phase transformation is delayed and essentially no ferrite or pearlite is formed during the quench. Figure 3.41b-c show the bainite and martensite development respectively. A small fraction of bainite starts to form at 6.7 s, which is the time step prior to the lower surface of the bar reaching the $M_S$ temperature. From Figure 3.41d, the residual hoop stress is tensile at the surfaces and compressive in the center. Also, at 6.7 s into the quench, the tensile hoop stress at the lower surface has a peak subsurface similar to what was seen with oil quenched 1020. At 6.7 s in the quench, the bar has reached the maximum centerline deflection as seen in Figure 3.41e. The effective plastic strain has increased in most of the bar while staying essentially the same at the bottom surface (Figure 3.41f). The bottom of the bar reaches the $M_S$ temperature at 10.1 seconds into the quench and bainite is seen up to approximately 20% at the bottom and approximately 5% at the top of the bar. Due to averaging across the element, there is no martensite predicted at this time step though there should be a small fraction at the surface. The hoop stress has become compressive at the surfaces and core while the mid-radius regions are in tension. The phase transformation and accompanied volume expansion near the lower surface of the bar has reduced the centerline deflection to approximately 0.75 mm (0.03 in).

At 17.3 s into the quench, both surfaces are below the $M_S$ temperature. The further cooling has resulted in a large increase in bainite fraction at the surfaces to above 50% at the lower surface and approximately 40% at the top surface. Martensite fraction has also increased to approximately 37.5% and 42.5% for the bottom and top surfaces respectively. The deeper transformation has caused compressive residual stresses to also be found to a deeper depth while the core is fully in tension. The austenite has been more fully transformed at the bottom of the bar with approximately 90% bainite plus martensite while the top of the bar is approximately 80% bainite plus martensite. This more complete transformation near the bottom surface resulted in an overall centerline deflection that is upward.

The temperature in the entire bar has dropped below the $M_S$ temperature 30.5 s into the quench. At this time step, the center of the bar has transformed and contains approximately 70% bainite and 20% martensite. The bainite near the surfaces has not increased greatly given the
martensitic transformation that has occurred. With the core transformation, the residual hoop stresses are tensile at the surface and compressive in the core.

The final time step shown is 105.1 s into the quench. At this time, the temperature across the bar is essentially uniform. Bainite has increased slightly in the core while martensite has increased approximately 10% to 30% in the core. The surface levels of bainite have not changed and a slight increase was seen in the martensite levels. With the final phase transformation and the time at elevated temperature, the residual hoop stresses have moderated somewhat so that the surface levels are under 400 MPa (58.0 ksi) in tension and the core is approximately 200 MPa (29.0 ksi) in compression. The martensitic transformation and resulting volume expansion near the top surface has also moved the centerline deflection back to downward while the effective plastic strain has increased through the entire bar.

3.6.2 Oil Quenched 4120 Carburized to 0.8 Wt Pct C

Similar to the changes shown earlier with 1020 when carburized, the Ms temperature profile changes when 4120 is carburized both as expected and shown in Figure 3.42a. The temperature profile (Figure 3.42a) at 5 and 6.7 s into the quench is the same as for the non-carburized condition. Beyond these times, the time steps in the simulation changed due to changes in the convergence of the calculations. For both 5 and 6.7 s into the quench, no bainite is predicted to have formed (Figure 3.42b) and the temperature is still above the Ms temperature precluding martensite formation (Figure 3.42c). At 5 s into the quench, the residual hoop stress is tensile at the surfaces and compressive in the core (Figure 3.42d). The centerline deflection is also predicted to be approaching 5 mm (0.20 in) due to the thermal contraction differences caused by the immersion (Figure 3.42e). At 6.7 s into the quench, the residual hoop stresses have become more tensile at the surfaces and show the subsurface peak near the lower surface typical of impending phase transformation that has been seen in the previous conditions. The centerline deflection has also increased past 5 mm (0.20 in) and the effective plastic strain has increased compared to 5 s (Figure 3.42f).
Figure 3.41  Time dependent response of non-carburized 4120, quenched in 65 °C oil for
(a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress,
(e) centerline deflection, and (f) effective plastic strain.
A very small fraction of bainite (less than 2%) is predicted to have formed 10.1 s into the quench while the temperature in the bar is still above the Ms. This small fraction along with the even smaller fraction of ferrite and pearlite (less than 0.05%) that has formed and the thermal contraction has pushed the surface hoop stress to compression and move the tensile peak inward.
With the more even temperature difference, the centerline deflection has decreased to approximately 1.5 mm (0.06 in).

The temperature profile intersects the $M_s$ temperature line at 17.2 s into the quench. The bainite fraction has increased subsurface along with the start of the martensitic transformation approximately 2.5 mm (0.10 in) below the surface. The compressive hoop stress increased in depth while the core is placed in tension. The bar has reversed the centerline deflection and is bent slightly upward while the effective plastic strain is increased across the bar.

A majority of the bar has cooled below the $M_s$ temperature 27.6 s into the quench, though the core is just slightly below the $M_s$. The surface of the bar has not reached the $M_s$ due to the added carbon decreasing the $M_s$ temperature. The bainite fraction has increased in the core up to approximately 65% in the core while martensite has continued to transform with some martensite in the core (approximately 5%) and peaks of 60% located 1.35 mm (0.05 in) below the surface. The subsurface transformation created tensile hoop stresses at the surface, compressive stresses at 1.35 mm (0.05 in) below the surface, followed by tensile stresses at greater depths and compression in the core. The compressive hoop stress near the surface coincides with the high martensite fraction. The centerline deflection reversed again to being deflected slightly downward.

At 44.0 s into the quench, only the surface remains above the $M_s$ temperature. The bainite has increased in the core to approximately 77% while closer to the surface has remained unchanged. Martensite has increased throughout the bar with the core increasing to 23%. There is also some predicted martensite near the surface. This again is due to the averaging across the element as opposed to being at the nodal level. The surface will not have martensite at this time since the temperature has not dropped below the $M_s$ temperature. With the surface transforming to martensite, the residual hoop stress is predicted to be compressive with a tensile peak near the case-core interface. The core is also still in compression. The centerline deflection has increased slightly in the downward direction. The effective plastic strain has increased slightly across the entire bar with a larger increase near the surfaces due to the martensite transformation.

The last time step shown is 335.2 s into the quench. At this time, the temperature is almost uniform across the bar. The bainite has not increased while the only change in the martensite fraction is near the surfaces. With the final martensite increase, the hoop stresses at the surface have become more compressive while the peak tensile stress has decreased slightly.
and the core has become less compressive. The centerline deflection and the effective plastic strain changed only slightly and the curves were omitted from the graphs for clarity.

Figure 3.42  Time dependent response of 4120 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) effective plastic strain.
Figure 3.42 (continued) Time dependent response of 4120 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) effective plastic strain.
3.6.3 15 s High Intensity Quench Hold Time Non-Carburized 4120

Similar to what was seen with intensively quenched 1020, 4120 also transforms to lower temperature products. Whereas 1020 contained less ferrite and pearlite with a corresponding increase in bainite and martensite, 4120 contains less bainite and more martensite. The temperature profiles for 4120 are similar to those for 1020 as shown in Figure 3.43a. At 0.1 s into the quench, the bottom surface of the bar as cooled to approximately 525 °C (977 °F). At this point no phase transformation has occurred (Figures 3.43b-c), the hoop stresses at the lower surface have increased to approximately 140 MPa (20 ksi) (Figure 3.43d), the centerline has deflected downward 1.25 mm (0.05 in), and the effective plastic strain at the lower surface is approximately 0.012. The hoop stresses at this time are due to the thermal contraction in the material as a result of the very fast cooling.

At 0.4 s into the quench, the lower surface of the bar has reached the Ms temperature. This occurred while the top of the bar was still above 600 °C (1115 °F) and the core was still essentially at the heat treating temperature. The martensite fraction at the bottom of the bar was approximately 12%. This has resulted in a peak tensile hoop stress that was subsurface though still relatively small near the lower surface. The thermal contraction in the lower part of the bar is still great enough to overcome the increased strength in the thin layer of martensite and the centerline deflection is at the maximum downward deflection of 3.4 mm (0.13 in). The effective plastic strain has also increased at both surfaces and a slight increase is seen in the lower half of the core of the bar.

The martensite fraction at the lower surface has increased to approximately 75% 0.6 s into the quench. The top surface has just reached the Ms temperature though no martensite was shown due to averaging over the element. With the higher martensite fraction and associated volume expansion, the centerline deflection decreased. Martensite was predicted to have formed on both surfaces 1.1 s into the quench with the lower surface showing both an increased fraction as well as a deeper layer. At this point, the surface residual stresses are compressive with a tensile peak subsurface and compressive hoop stresses in the core. These surface compressive stresses are approximately 250 MPa (36 ksi) in magnitude. The centerline deflection has again decreased with the increased martensite fraction and depth while the effective plastic strain has increased throughout the bar.
The core temperature shows the first decrease 2.1 s into the quench while both surfaces have cooled to below 200 °C (392 °F). At this point, a very small fraction of bainite forms just below the surface while the martensite fraction increased to approximately 95% near the surface. The martensite layer is still thin, with a depth of less than 2.5 mm (0.10 in). The hoop stresses near the surface have become more compressive and remain compressive to a greater depth compared to 1.1 s with the surface stresses of approximately 500 MPa (72.5 ksi). The centerline deflection is upward due to the differences in martensite depth while the thermal contractions from this point on should be the same between the top and bottom surfaces since the temperature profile is essentially symmetric about the centerline.

The bainite fraction has increased to approximately 3% at a depth of 2.5 mm (0.10 in) while the martensite fraction has increased slightly in magnitude near the surface and is at a greater depth 4.1 s into the quench. At this time step, the residual stresses have reached their peak compressive level near the surface of approximately 750 MPa (109 ksi). The stresses remain compressive to a depth of 7 mm (0.25 in) and are tensile in the core. The centerline has reached the maximum deflection upward while the effective plastic strain has continued to increase across the bar.

The core of the bar is approximately 500 °C (925 °F) 7.2 s into the quench. The bainite has increased both magnitude and depth while martensite has continued to transform to greater depths. With the increased depth of martensite, the surface compressive stresses have relaxed somewhat though the hoop stress is compressive to a greater depth. At 12.3 s into the quench, the temperature in the entire bar has dropped below the Ms temperature leading to a large increase in martensite in the core to approximately 80%. Bainite has also increased in the core to approximately 10%. With the increase in martensite in the core, the compressive hoop stress near the surface has decreased in magnitude, while the mid radius region has a large tensile hoop stress peak and the core was also placed into compression.

The last time step of the quench is 15.5 s. After this time, the bar is removed from the quench and allowed to air cool to room temperature. At this point in the quench, the martensite fraction has increased further in the core to approximately 90%. The hoop stresses have relaxed somewhat and the centerline deflection has decreased slightly though is still upward. The effective plastic strain has increased in the core due to the martensite transformation.
Figure 3.43  Time dependent response of non-carburized 4120, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) effective plastic strain.
Figure 3.43  (continued) Time dependent response of non-carburized 4120, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) effective plastic strain.
3.6.4 15 s High Intensity Quench Hold Time 4120 Carburized to 0.8 Wt Pct C

As would be expected and was shown with 1020, carburized and intensively quenched 4120 responds slightly differently than non-carburized 4120 though there are similarities as was seen in 1020. As was seen in the non-carburized 4120 condition, the temperature of the bottom of the bar has dropped significantly 0.1 s into the quench (Figure 3.44a). Although no phase transformation has occurred, the bar has started to deflect downward approximately 1.5 mm (0.06 in) and the effective plastic strain at the surface has increased to 0.017 and 0.006 for the bottom and top surfaces respectively. The top surface has cooled to approximately 600 °C (1115 °F) 0.4 s into the quench. Again, this temperature is still above the Ms temperature and no bainite has formed. Due to the carburization, the temperature at the surface has not reached the Ms temperature 0.7 s into the quench as it had in the non-carburized condition. At this time step, the residual hoop stress is predicted to be tensile at the surface with a magnitude of 300 MPa (44 ksi) and compressive in the core. The downward centerline deflection has reached the peak at this point.

The temperature profile at 1.4 s into the quench shows that the bar has not reached the Ms temperature. However, martensite is predicted to have formed at both the top and bottom surface. This indicates that the Ms temperature line shown is not the same line as ABAQUS used in the prediction. The Ms line as shown was calculated based on the Ms temperature given in the data files and linear interpolation between data points to the predicted carbon levels. The depth of martensite formation is 0.15 mm (0.006 in) and 0.45 mm (0.018 in) for the top and bottom surfaces respectively. This indicates that the linear interpolation/extrapolation used to calculate the Ms temperature only differed for the two elements closest to the surface. This difference is only seen for carburized 4120 and intensively quenched. With the start of the martensite transformation, the surface hoop stresses are compressive with peak tensile stresses where the martensite fraction returns to zero. With the start of the transformation, the downward centerline deflection has decreased. The effective plastic strain has increased overall in the bar from the previous time step.

Unlike the carburized and oil quenched condition, the carburized layer of the bar was cooled below the Ms temperature at 2.7 s. At this time, the core temperature has also dropped from the initial temperature. At this time step, a small fraction of bainite has formed near the case-core interface in addition to the martensite fraction growing in both magnitude and depth.
With the further phase transformation, the residual hoop stresses at the surface have become more compressive and reach to a greater depth than the previous time step. The greater depth of the martensitic transformation along the bottom surface of the bar has pushed the centerline deflection to less than 0.5 mm (0.02 in) downward.

At 5.1 s into the quench, the bainite fraction has increased slightly and the peak has increased in depth similar to the non-carburized condition. Martensite has increased near the surface as well as increased in depth to approximately 5 mm (0.20 in). This increase in transformed volume has further increased the depth of compressive hoop stresses as well as increasing their magnitude at the surface.

The core or the bar has just reached the $M_S$ temperature 9.0 s into the quench. Bainite has started forming in the core with a peak level around 9% at a depth of 7.5 mm (0.30 in). Martensite fraction has increased to a depth of 10 mm (0.40 in) while the surface is approximately 90% martensite. The compressive hoop stress has decreased near the surface due to the subsurface martensite transformation leading to a tensile peak approximately 6 mm (0.25 in) below the surface. The greater fraction of martensite transforming in the upper half of the bar has increased the downward centerline deflection.

At the end of the quench (15.5 s), the entire bar is well below the $M_S$ temperature leading to the final transformations occurring. Bainite in the core peaks around 11% while the rest of the core is martensite. The surface martensite level is predicted to be approximately 96%. The surface compressive hoop stress is further decreased due to the core transformation while the mid-radius tensile peak increases and the core is under high compressive stresses. The centerline deflection has decreased somewhat while the core effective plastic strain increased.

The last time step shown of 19.9 s is after the quench has been completed and the bar is cooling the rest of the way to room temperature. The surface temperature has increased compared to the end of the quench due to core cooling and a much slower heat transfer rate away from the bar. The hoop stress profile becomes slightly more compressive near the surface while in the rest of the bar the profile moderates.
Figure 3.44  Time dependent response of 4120 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) effective plastic strain.
Figure 3.44  (continued) Time dependent response of 4120 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) effective plastic strain.
3.7 Heat Transfer Rate Coefficient Effect on Residual Stresses

This final section of results deals with the effects of the heat transfer rate on the surface residual hoop stress after quenching. While the effects of the intensive quenching were shown through examples in the literature to modify the residual stresses at the surface, the heat transfer rate that represented the transition from tensile to compressive residual stresses was not defined. A series of simulations were performed on a two dimensional axi-symmetric model. In these simulations, the heat transfer rate was varied while the quench time was held constant at 15 s. Each set of non-carburized simulations were also completed for all four alloys investigated (1020, 4120, 4320, and 8620). From these simulations, the predicted residual hoop stresses were gathered and the surface element data collected. Figure 3.45 shows that as the heat transfer rate increases, the various alloys used here transition from tensile to compressive hoop stresses. This transition heat transfer rate varies based on the alloy as would be expected. The transition for 1020 is approximately 3.2 kW/m$^2$ °C while for 4120, 4320, and 8620 the transition occurred at approximately 9 kW/m$^2$ °C. Table 3.1 presents the transition heat transfer rates numerically as found through linear interpolation between the data points on either side of zero surface hoop stress.

Quench time effects were also investigated utilizing the 8620 simulations. For these simulations, the hold time was varied between 15 s and 30 s and the series of simulations was repeated. Figure 3.46 shows very little variation in the predicted hoop stresses of the three quench hold times. Longer quench hold times were seen as redundant as the core temperatures for the smallest heat transfer rate (5 kW/m$^2$ °C) were approximately 93 °C for the 30 s quench hold time.
Figure 3.45  Predicted residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched with a varying heat transfer rate utilizing the 2D axi-symmetric model. The specified heat transfer rate was applied for 15 s then allowed to air cool.

Table 3.1 – Transition Heat Transfer Rate (kW/m² °C) Needed to Form Compressive Residual Hoop Stresses at the Surface of a 25.4 mm (1 in) Diameter Bar

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Heat Transfer Rate (kW/m² °C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1020</td>
<td>3.17</td>
</tr>
<tr>
<td>4120</td>
<td>8.96</td>
</tr>
<tr>
<td>4320</td>
<td>8.85</td>
</tr>
<tr>
<td>8620</td>
<td>9.01</td>
</tr>
</tbody>
</table>
Figure 3.46   Predicted residual hoop stress of non-carburized 8620 quenched with a varying heat transfer rate. The specified heat transfer rate was applied for 15, 20, or 30 s followed by air-cooling to room temperature.
4.1 **Overview**

This chapter will further delve into the new results from the previous chapter. The focus here will be on further understanding the effects of the carburization and heat transfer rate on the microstructure, distortion and residual stress through insights provided by modeling.

4.2 **Microstructure Predictions**

The microstructures predicted for each alloy differ from equilibrium microstructures taken from alloy specific phase diagrams. By definition, the phase diagram shows the equilibrium phases at various temperature and carbon levels. They are based on essentially infinite time at each temperature over the composition range given. Under these conditions, alloying elements have time to diffuse through the matrix and equilibrate between phases leading to the lowest free energy state in the system.

At industrially relevant cooling rates, steel does not have time for substitutional alloying elements to diffuse. Austenite decomposition to daughter products can be shown on a continuous cooling- transformation (CCT) plot for each alloy. The CCT plot for 4140 is shown in Figure 4.1 for example. As shown on the CCT diagram, as the cooling rate increases, the amount of ferrite found in the final microstructure decreases with a corresponding increase in lower temperature transformation products of pearlite, bainite, and martensite.

4.3 **Quenching Effects on Residual Stress**

As was shown in the previous chapter, 4120, 4320, and 8620 are all predicted to have tensile residual hoop stresses at the surface after oil quenching. In contrast, after high intensity quenching, the same alloys are predicted to have compressive residual hoop stresses at the surface. The development of the microstructure during both quenching processes was shown in Figure 3.28 for oil quenching and in Figure 3.30 for high intensity quenching. In order to better explain the development of the residual stresses, another view of the simulation results is needed.
Figure 4.1 Continuous cooling transformation curve for 4140 steel. [65]

Figure 4.2 contains the time development of temperature, martensite, radial expansion, and hoop stress (from Figure 3.28d) for non-carburized and oil quenched 4120. Figure 4.2a shows a 300 °C (560 °F) temperature difference between the core and the surface 10.1 s into the quench. At this time, there is no martensite predicted (Figure 4.2b) while the surface has thermally contracted from the start of the quench (Figure 4.2c). Figure 4.2d shows that at 10.1 s into the quench, the surface is under compression with tensile peaks at the mid-radius. This profile is a result of the bainite forming at the surface prior to the bar reaching the Ms temperature. Figure 4.2a shows that at 17.3 s the surface temperature of the bar has dropped below the Ms temperature, allowing martensite to form. With the martensite formation, the volume expansion on transformation greatly offsets the thermal contraction near the surface while the core, which has not reached the Ms temperature, has continued to thermally contract (Figure 4.2c). Again referring to Figure 4.2d, the compressive zone has increased in depth while the core is placed into tension due to the martensitic transformation. The temperature continues to cool and the entire bar has dropped below the Ms temperature at 30.5 s into the quench. The core austenite transforms into martensite (and bainite) resulting in a large increase in radial
displacement throughout the bar caused by the volume expansion due to transformation. With the transformation in the core, Figure 4.2d shows that the compressive residual stresses at the surface have become tensile with compressive stresses in the core. As the bar cools further, the bar continues to transform in the center as well as thermally contract. As was shown, the core transformation causes the surface to be under tension whereas at this time step, the residual stresses moderate. This would indicate the thermal contraction in the core overcame the expansion on transformation and relieves a portion of the surface tensile stress as seen in Figure 4.2d.

To further help explain the progression of transformation, consider Figure 4.3. The bar can be considered schematically using two distinct volumes, the case and the core. For illustrative purposes, the case extends approximately 2 mm (0.08 in) below the surface. Figure 4.3a shows the bar at the end of the austenitizing heat treatment where the temperature is uniform. Here, both the case and the core are high temperature austenite. After the start of the oil quench and before the Ms temperature is reached, both the case and the core have cooled and thermally contracted. The original outer diameter of the case is shown as a dashed line in Figure 4.3b. As the bar cools further, the case transforms to martensite and undergoes the corresponding volume expansion. The core has also cooled, resulting in further thermal contraction. If the core were unconstrained, the result would look similar to Figure 4.3c where there is a gap between the martensitic case and the cooling austenitic core. After the core has cooled further and transforms into martensite and bainite, the core expands eliminating and expanding beyond the unconstrained differences between the case and core as shown in Figure 4.3d. As the core expands on transformation, the expansion results in a reversal of the residual hoop stresses from compression to tensile as was seen in Figure 4.2d.

Figure 4.4 shows the temperature, martensite fraction, radial expansion, and residual hoop stresses for 4120 after high intensity quenching. At 0.6 s into the quench, the surface temperature decreased below the Ms temperature (Figure 4.4a) with a small fraction of martensite predicted to have formed (Figure 4.4b). This small amount of martensite slightly decreases the thermal contraction associated with the large thermal gradient (Figure 4.4c). Figure 4.4d shows the residual hoop stress is becoming compressive near the surface of the bar.
Figure 4.2  Time dependent response of non-carburized 4120, quenched in 65 °C oil for (a) temperature, (b) martensite, (c) radial expansion from the start of the quench, and (d) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.2 (continued) Time dependent response of non-carburized 4120, quenched in 65 °C oil for (a) temperature, (b) martensite, (c) radial expansion from the start of the quench, and (d) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.3  Volume changes with respect to temperature through the oil quench of non-carburized steel.  (a) Temperature is austenitizing temperature (or carburizing temperature if carburized).  (b) After the start of the quench and prior to reaching the $M_s$ temperature, the “case” has contracted and in constrained by the “core” with the original diameter marked by the dashed line.  (c) After the “case” has transformed to martensite and the “core” has cooled.  When the “case” and “core” are unconstrained by the other, a gap would form between the two sections.  (d) Final condition where the core has transformed into martensite/bainite, thereby decreasing the differences between the unconstrained volumes.

At 1.1 s into the quench, the martensite fraction at the surface has greatly increased while the thermal contraction at the surface of the bar has been further offset by the volume expansion related to the martensitic transformation.  As the bar continues to cool, martensite continues to extend deeper while the untransformed volume thermally contracts.  This combination results in a slight radial growth very near the surface.  Figure 4.4d shows that the compressive residual hoop stresses are increasing in magnitude as a result of the combined thermal contraction and martensite formation.  As the transformation and cooling progress, the core continues to contract pulling the martensite inward while the martensite formation causes the hoop stresses to become more compressive.  At 7.2 s into the quench, the very center of the core remains above the $M_s$ temperature.  The martensite growth “inward” has offset the thermal contraction when the radial displacement is compared to 4.1 s into the quench.  This increase in martensite fraction places a greater fraction of the bar into compression with a corresponding decrease in surface compressive stresses.  As the core finally transforms, the surface correspondingly expands resulting in a decrease, but not elimination, in the compressive hoop stress at the surface.  This final core transformation is constrained by the prior martensite resulting in compressive stresses in the core of the bar.  With the final cooling, the entire bar thermally contracts leading to a decrease in the magnitude of the residual stresses.
Again, to further help explain the order of transformation, consider Figure 4.5. The bar can be considered schematically using two distinct volumes, the case and the core. For illustrative purposes, the case extends approximately 2 mm (0.08 in) below the surface. Figure 4.5a shows the bar at the end of the heat treatment where the temperature is uniform. Here, both the case and the core are high temperature austenite. After the start of the high intensity quench and before the \( M_s \) temperature is reached, the case has cooled while the core remains essentially at the heat treatment temperature. Due to the cooling the case austenite has thermally contracted while being constrained by the hot core. The original outer diameter of the case is shown as a dashed line in Figure 4.5b. As the bar cools further, the case transforms to martensite and undergoes the corresponding volume expansion. The core has also cooled slightly resulting in thermal contraction. If the core were unconstrained, the result would look similar to Figure 4.5c where there is a gap between the martensitic case and the cooling austenitic core. After the core has finished cooling and transforms into martensite and bainite, the core expands reducing the unconstrained differences between the case and core as shown in Figure 4.5d. Since the case and the core are connected, the shrinking core pulls the case inward and increases the compressive hoop stresses. When the core transforms and the constraints are reduced, the residual stresses are decreased somewhat as was seen in Figure 4.4d.

Figure 4.6 shows the time dependent development of 4120 carburized to 0.8 wt pct C and oil quenched. As shown in Figure 4.6a, the temperature first decreases below the \( M_s \) temperature near the case-core interface resulting in the transformation starting there as opposed to the transformation starting at the surface in the non-carburized condition. As the transformation starts, the compressive stress extends to a greater depth (Figure 4.6d). At 27.6 s into the quench, the martensite (and bainite) fractions (Figure 4.6b) are large enough to effect the radial contraction resulting in a large expansion from the previous time step (Figure 4.6c). This expansion pushes the surface into tension while the core is compressed. As the core austenite finishes transforming and the surface has formed an appreciable fraction of martensite, the volume expansion increases the radial expansion while also creating compressive stresses at both the surface and core. The martensite formation at the surface and corresponding volume expansion pulls the case-core region into tension. As the temperature in the bar cools further, the core contracts while the surface continues to transform into martensite. These combined effects cause the final increase in the residual compressive hoop stress at the surface.
Figure 4.4  Time dependent response of non-carburized 4120 after high intensity quenching for (a) temperature, (b) martensite, (c) radial expansion from the start of the quench, and (d) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.4  (continued) Time dependent response of non-carburized 4120 after high intensity quenching for (a) temperature, (b) martensite, (c) radial expansion from the start of the quench, and (d) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.5 Volume changes with respect to temperature through the high intensity quench. (a) Temperature is austenitizing temperature (or carburizing temperature if carburized). (b) After the start of the quench and prior to reaching the $M_S$ temperature, the “case” has contracted and is constrained by the “core” with the original diameter marked by the dashed line. (c) After the “case” has transformed to martensite and the “core” has cooled. When the “case” and “core” are unconstrained by the other, a gap would form between the two sections. (d) Final condition where the core has transformed into martensite/bainite, thereby decreasing the differences between the unconstrained volumes.

4.4 High Intensity Quenching Effects

There are several high intensity quenching findings from the previous chapter that need further discussion. These will be discussed in the following sections.

4.4.1 High Intensity Quench Hold Time

From the literature, intensive quenching is typically performed in a step fashion where the part is cooled for a set amount of time in high velocity water and then allowed to air cool to room temperature. This was called IQ3 by Kobasko [7]. It was also claimed that the surface residual stresses would be the most compressive if the quench was interrupted when the core temperature of the part was at the $M_S$ temperature [1]. Figure 4.7 shows that as the quench hold time increases from 10 s, the core temperature decreased (as expected). However, the drop in the core temperature does not correspond to a decrease in the compressive hoop stresses at the surface in the bar. The decrease in core temperature actually results in a slight increase in compressive stresses near the surface from 575 MPa (83 ksi) for the 10 s quench to 600 MPa (87 ksi) for the 20 s quench. This was the same for the other alloys analyzed.
Figure 4.6  Time dependent response of 4120 carburized to 0.8 wt pct C, quenched in 65 °C oil for (a) temperature, (b) martensite, (c) radial expansion from the start of the quench, and (d) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.6 (continued) Time dependent response of 4120 carburized to 0.8 wt pct C, quenched in 65 °C oil for (a) temperature, (b) martensite, (c) radial expansion from the start of the quench, and (d) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.7  (a) Temperature distribution in the bar at the end of the quench for 4120 in three carburized conditions. (b) Predicted residual hoop stresses at the end of the heat treatment. The times indicated in the legend are the quench time in the intensive quenching process.
4.4.2 **High Intensity Quench Residual Stress Profiles**

The residual hoop stress profiles that were predicted for the non-carburized and intensively quenched condition were found to be opposite of the oil quench. Also, the overall profile characteristics were not reported previously in literature; typically, only the surface and near surface stresses, both measured and simulated, were reported [14-20, 48-52]. The entire residual stress profile is shown in Figure 4.8a for non-carburized 4120, 4320, and 8620 after oil quenching and after 15 s high intensity quenching. For the part shape and dimensions used here, the carburized and oil quenched simulations showed a similar pattern in the residual hoop stress as shown in Figure 4.8b. After carburization, the oil quench produces compressive stresses at the surface as was shown earlier. One interesting result is that high intensity quenching also produces compressive residual hoop stresses at the surface in the non-carburized condition. Secondly, the subsurface peak tensile residual stress could play a role in the life expectancy of parts. The sequence of events was previously discussed in Section 4.3.

4.4.3 **Surface Residual Hoop Stress Tensile to Compression Transition Heat Transfer Rate**

High intensity quenching relies on a shift from tension to compressive surface residual stresses. This transition from tensile to compressive residual stress had been undefined. As was shown in Figure 3.31, the transition for 4120, 4320, and 8620 was very similar while 1020 had a much lower critical heat transfer rate. The critical heat transfer rate where the residual stress goes from tensile to compression of each alloy was determined by performing a series of two-dimensional simulations utilizing an axi-symmetric model where the heat transfer rate was varied. While this model does not allow for varied quenching angles as the half bar model does, the analysis did not call for that level of specificity. For these simulations, the quench time of 15 s was selected based on the results of the three dimensional high intensity quenching results. Shorter times were not used, as they were not expected to produce residual stresses as compressive at the surface as the 15 s quench did.
Figure 4.8  Final predicted residual hoop stress for 4120, 4320, and 8620 in the (a) non-carburized or (b) carburized to 0.8 wt pct C condition.
It was noted that the three alloys, 4120, 4320, and 8620, all were predicted to have similar transition heat transfer rates even though they have different ideal diameters and varying alloy content. This similarity comes from the steel properties contained within the .mec file used in DANTE. This file contains the internal state variable data for each phase along with Young’s modulus, Poisson’s ratio, the latent heat of transformation from austenite to the various product phases, the specific heat and thermal conductivity of each phase as a function of temperature, the coefficient of thermal expansion and transformation strains for austenite and the decomposition of austenite to the various product phases as a function of temperature and carbon level, and the hardness levels for the various phases as a function of carbon content. The four alloys investigated used the same data for these despite the differences in alloy content. Only the data related to the kinetics of transformation, such as $M_S$ temperature, temperature ranges for transformation, and other kinetic data of the various transformations were changed between alloys. The assumption that these values are the same could have an effect on the predicted results. This essentially means that the differences predicted between the alloys are only a result of the variations in the transformation temperatures used while some alloying effects would also be expected.

Figure 4.9 shows the time development of temperature, ferrite, pearlite, radial expansion, and residual stresses similar to what is shown for 4120 in Figure 4.1. As can be seen, as the ferrite and pearlite form 7.1 s into the quench, the radial contraction is reversed due to the transformation. The transformation causes the surface to be placed in compression. As the cooling continues, the core continues to contract while near the surface the austenite continues to transform into ferrite and pearlite offsetting the contraction. The further transformation increases the depth of the compressive stress. When the core of the bar transforms 14.6 s into the quench, the volume near the surface is pushed outward slightly. The core transformation causes the surface to be placed into tension. As the core cools further, the entire bar contracts resulting in a decrease in the tensile stresses large enough to become compressive. The transformation progression is similar to that of 4120 during high intensity quenching where the transformation progresses from the surface. The austenite decomposition causes the compressive stresses at the surface due to the transformation volume expansion. Whereas in intensively quenched 4120 the residual stresses due to martensite formation were so large they survived the core transformation, when the 1020 core transformed, the stresses became tensile. As the bar further cooled, the
thermal contraction in the core paired with the strength in the case to resist the contraction forced the residual stresses to become compressive at the surface.

4.5 Limitations of the Simulations

There are several parts of the simulations that put a limitation on the accuracy of the results. First, it was assumed that the heat transfer rate of the high intensity quench was constant throughout the quench. This assumption has been made elsewhere in literature with reasonable comparisons between simulations and experimental works [48-51]. However, the heat transfer rate could vary somewhat through the quench. Detailed experimentation verification of the heat transfer rates would be required to either justify or disprove this assumption.

Next, it was noted that the carbon diffusion data came from the work by Tibbetts [41]. This work does not take into account the variation that the various alloying elements can contribute to the diffusion of carbon in these steels. Newer carbon diffusion equations could lead to changes in the predicted carbon profiles.

The results of the simulations of carburized and intensively quenched 4120 showed the martensite formation at the elemental level started before the temperature profile at the nodal level reached the $M_S$ temperature. This cannot occur in the real world and is a result of the averaging between the nodal and elemental levels that occurs in ABAQUS when the stress and distortion model is completed. This effect, only seen in 4120, could be avoided by utilizing a finer mesh at the surface. A slightly greater depth of the fine mesh size would also help in adding to the accuracy in the carbon profile at greater depth. This would also help to better define the phase fractions forming at the case-core interface of carburized parts during oil quenching.

Finally, the data used within DANTE could be further refined. It was mentioned that the data contained in the mechanical data file is identical between the four alloys investigated here. This is an assumption on the part of DCT that must be made more clear and accessible to the users of the program. It is suspected that while this reuse of data between steel grades is acceptable under normal quenching conditions, there could be some repercussions in the results, specifically in the critical heat transfer rate to form compressive residual stresses at the surface. It is expected that the critical heat transfer rates would vary slightly more than shown.
Figure 4.9  Time dependent response of non-carburized 1020, quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) radial expansion from the start of the quench, and (e) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
Figure 4.9 (continued) Time dependent response of non-carburized 1020, quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) radial expansion from the start of the quench, and (e) residual hoop stress. The radial position is based on the initial part, prior to heat treatment with the centerline and the surface at 0 mm and 12.7 mm respectively.
A series of heat treatments has been simulated utilizing DANTE®, a commercially available heat treatment simulation software that consists of a set of user subroutines for ABAQUS, a finite element software package. The simulations include three carburization levels (non-carburized, and 0.8 or 1.0 wt. pct carbon at the surface) and two different quenching methods (oil quenching and “intensive” quenching) for four common grades of bar steel (1020, 4120, 8620 and 4320). Horizontal quenching of a cylindrical 25.4 mm (1 in) diameter bar, 254 mm (10 in) long, was simulated in order to reduce complexity due to part geometry or alternate quench configurations. The goals of the simulations were to assess potential benefits of the intensive quenching phenomena, wherein large compressive residual stresses are reported at the surface of parts after quenching, in comparison to the surface tensile stresses expected during traditional oil quenching. The results considered here were for a selected set of parameter inputs and are based on constant immersion rates of 5 mm/s for oil quenching and 50 mm/s for intensive quenching. Selection of alternate input parameters, as discussed in the sensitivity discussion in Appendix J could lead to slight modifications in the predicted results. However, with selection of reasonable parameters, it is anticipated that the major conclusions discussed below would be maintained.

The modeling results provided information related to thermal history and microstructure development, as well as distortion and evolution of residual stresses through the cross-section and along the length. The model results confirm that high intensity quenching generated compressive residual hoop stresses at the surface for all conditions after heat treatment. Oil quenching was predicted to result in tensile residual hoop stresses at the surface in the non-carburized conditions. The temperature difference between the core and the surface exceeded 700 °C (1275 °F) during the initial stages of high intensity quenching. Detailed analysis of the model outputs suggest that this temperature difference means the surface austenite decomposes (to martensite) resulting in circumferential compressive stresses due to the constrained volume expansion of martensite while the core is at a much higher temperature than observed at the same point in the process history for conventional quenching, and the specimen radius is therefore
much greater. As the core cools and thermally contracts, radial tensile stresses develop between the case and core regions resulting in further increases in the compressive residual hoop stresses at the surface. With the large initial diameter of the martensitic “shell”, the core transformation and corresponding volume expansion is not large enough to expand beyond the martensitic shell. This behavior results in a surface and core under high compressive stresses while the mid-radius of the bar is under large tensile stresses. While intensive quenching is predicted to generate consistent and high levels of surface compression, it should also be noted that high tensile residual stresses are predicted to occur subsurface. This aspect of the intensive quenching response may not be beneficial, and has not been reported in prior studies. The combination of carburizing and high intensity quenching was not found to significantly enhance the magnitude or depth of the compressive zone.

The simulations indicate that all quenching methods studied result in compressive residual hoop stresses in a 1020 steel for all conditions. Further, it was noted that the magnitude and depth of the compressive residual stresses in the 1020 alloy were found to be equivalent or greater compared to the other alloys after high intensity quenching, although the microstructure of the 1020 steel may be less desired, containing ferrite and pearlite even after intensive quenching.

The heat transfer rate required to create compressive stresses at the surface of the bar was also investigated for each of the alloys. Table 5.1 summarizes the heat transfer rate needed to form compressive residual stress at the surface of the bar. Any rate lower than those listed will result in tensile stresses at the surface. The critical heat transfer rate for 1020 is well within the heat transfer rates expected during oil quenching. Further analysis in Appendix H indicated that this critical heat transfer rate could be generalized in terms of Biot numbers, where a Biot number greater than 2.5 would be expected to produce compressive residual stresses after quenching.

Some additional insights were provided through modeling efforts reported in Appendices I and J. In Appendix I, a vacuum gas carburized simulation was completed on a bearing cup shape with varying quench parameters (from oil to gas to high intensity quenching) and carburization levels, for comparison with actual dimensional measurements on heat treated parts. The simulations were set up to mimic as closely as possible the heat treatments applied for a single part, although some proprietary heat treating conditions were not available, assumptions
Table 5.1 – Transition Heat Transfer Rate (kW/m² °C) Needed to Form Compressive Residual Hoop Stresses at the Surface of a 25.4 mm (1 in) Diameter Bar

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<th>Alloy</th>
<th>Heat Transfer Rate (kW/m² °C)</th>
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<tr>
<td>4120</td>
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were required in relation to the gas quenching heat transfer rates, gas flow and thermal mass effects related to other parts and part racks used during the experimental portion were ignored, and the latent heat of transformation in the gas quenching conditions had to be artificially reduced to allow mathematical convergence of the models. The simulated microstructures were in good agreement with the experiments although the predicted distortions of the cup were opposite in sign to the dimensional measurements on the heat treated parts. In Appendix J, a sensitivity analysis was conducted to assess the influence of tunable material parameters in DANTE on the predicted results, including specific heat, latent heat, and thermal conductivity of the steels. When modifying these variables, the trends behaved as would be expected, and the magnitudes could be quite substantial. Decreasing the latent heat and specific heat resulted in higher fractions of martensite after the simulation while increasing thermal conductivity increased final martensite fractions.
CHAPTER 6: FUTURE WORK

There are a number of areas relating to this work that deserve further consideration. First is the simulation of heat treatments. This work utilized DANTE, a commercially available simulation tool. While the data contained is adequate for the simulations, newer data utilizing new techniques could refine and improve the simulation accuracy based on comparisons between the simulation results and experimentation found in the literature for the microstructure predictions, distortion results, and the residual stress development. The residual stress levels in particular could be further refined using new measurement techniques as well as measured in-situ during a heat treatment.

As was noted, the carbon diffusivity utilized in DANTE is independent of the alloy. The predicted carbon profiles could be further improved if these interactions were accounted for in the simulation. Also, the thermodynamic properties such as latent heat and specific heat of some of the alloys (41XX, 43XX, and 86XX) are identical while there should be some small differences due to the changes in chemistry between the alloys. Increasing individuality of the data to better match that of the alloys will improve the accuracy of the simulations.

In this work, the heat transfer rate of high intensity quenching was held constant during the quench. This heat transfer rate most likely is a function of temperature similar to oil quenching. Further work to characterize this variation would be useful to further refine the heat transfer coefficients used in the simulations thereby improving the accuracy of the simulations.

Here, a single size and shape was used in all the simulations. Both size and shape will have a large effect on the predicted results. Further work looking at size and shape effects would be suggested.

This work also relied solely on simulations. Taking the results from the simulations already completed, experimental verification would be very useful. This verification would be most useful in examining the predicted critical heat transfer rates. Further experimentation could be used to validate the opposite behavior of 1020 compared to the other alloys in terms of residual stresses and centerline deflection after oil quenching.
REFERENCES


[41] Dante (Version 3.3) [Computer Software]. Cleveland, OH: Deformation Control Technology.


[58] B.L. Ferguson <lynn.ferguson@deformationcontrol.com>, “Re: Question about IQ,” August 14, 2007, Personal email.


** MATERIALS **

*Material, name=S86XX
*Density
7.83e-06,
*Diffusivity
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1.64519e-06, 0.0025, 788.
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*Initial Conditions, TYPE=CONCENTRATION
all, 0.002
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  carburization
*Mass Diffusion, end=period, dcmx=0.001
  10., 21600., 0.1062, 500.
  **
  ** BOUNDARY CONDITIONS
  **
  ** Name: BC-1 Type: Mass concentration
  *Boundary
  outside, 11, 11, 0.008
  **
  ** OUTPUT REQUESTS
  **
  *Restart, write, frequency=1000
  *Monitor, dof=11, node=Monitor, frequency=1
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  *Output, field, frequency=10
  *Node Output
  NNC,NT
  *Element Output
  CONC,
  *El Print, freq=999999
  *Node Print, freq=999999
  NNC
  *End Step
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*Material, name=S86XX
*Density
 7.83e-06,
*Depvar
 100,
*User Material, constants=8, type=THERMAL
 7.83e-06, 0., 0.5, 0.5, 0., 0., 0.
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** INTERACTION PROPERTIES
**
*Film Property, name=HOT-AIR
 0.000237, 231.
 0.000283, 287.
 0.000335, 343.
 0.0001, 400.
 0.0001, 500.
 9e-05, 565.
 8e-05, 620.
 8e-05, 850.
*Film Property, name=air-cool
 0.00011, 37.
 0.000129, 65.
 0.000143, 93.
 0.000162, 120.
 0.000178, 148.
 0.000197, 176.
 0.000237, 231.
 0.000283, 287.
 0.000335, 343.
 0.000393, 398.
 0.000458, 454.
 0.000533, 509.
 0.000616, 565.
 0.000708, 620.
 0.000811, 676.
*Film Property, name=furnace
 0.0002, 20.
 0.0008, 1000.
*Film Property, name=oil-quench
 0.0001, 20.
 0.0005, 150.
0.0015, 300.
0.002, 400.
0.00375, 450.
0.005, 500.
0.005, 550.
0.00475, 600.
0.003, 650.
0.002, 700.
0.0015, 750.
0.0013, 800.
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** FIELDS
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*Initial Conditions, TYPE=field, Var=1
all, 0.002
*Initial Conditions, TYPE=field, Var=2
all, 2
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**
*Step, name="Heating up", AMP=STEP, inc=500
furnace heating up
*Heat Transfer, end=PERIOD, deltmx=30.
1., 1000., 0.01, 50.,
**
*CONTROLS, PARAMETERS=LINE SEARCH
6,
*CONTROLS, PARAMETERS=TIME INCREMENTATION
20, 30
*CONTROLS, FIELD=TEMPERATURE, PARAMETERS=FIELD
0.05, 0.05, , ,
** INTERACTIONS
**
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surface_nodes, F, 960., furnace
**
** OUTPUT REQUESTS
**
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*Monitor, dof=11, node=Monitor, frequency=1
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** FIELD OUTPUT: F-Output-1
**
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*Node Output
NT,
*Element Output
SDV,
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*Node File, freq=1
NT,
*El Print, freq=999999
*Node Print, freq=999999
*End Step
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**
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**
*Step, name="air transfer", AMP=STEP, inc=1000
air transfer from furnace to tank
*Heat Transfer, end=PERIOD, deltmax=30.
0.01, 12., 1e-06, 0.5,
**
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*FIELD, OP=NEW, VAR=2
all, -2.0
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**
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surface_nodes, F, 400., HOT-AIR
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**
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** FIELD OUTPUT: F-Output-1
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NT,
*Element Output
SDV,
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
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NT,
*End Step
** ----------------------------------------------------------------
**
** STEP: immerse quenching
**
*Step, name="immerse quenching", AMP=STEP, inc=2000
immerse quenching
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0.001, 5., 1e-06, 5.,
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** Interaction: Int-1
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surface_nodes, FNU
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**
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*Node Output
NT,
*Element Output
SDV,
**
** HISTORY OUTPUT: H-Output-1
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*Output, history
*Node File, freq=1
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*End Step
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**
** STEP: oil quenching
**
*Step, name="oil quenching", AMP=STEP, inc=2000
oil quenching process
*Heat Transfer, end=PERIOD, deltmx=30.
0.0004, 1000., 0.0001, 50.,
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**
** Interaction: Int-1
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surface_nodes, F, 65., oil-quench
**
** OUTPUT REQUESTS
**
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**
** FIELD OUTPUT: F-Output-1
**
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*Node Output
NT,
*Element Output
SDV,
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*Node File, freq=1
NT,
*End Step
**

** STEP: air cool
**

*Step, name="air cool", AMP=STEP, inc=1000
air cool to room temperature
*Heat Transfer, end=PERIOD, deltmx=30.
1., 1500., 0.01, 100.,
**
** INTERACTIONS
**
** Interaction: Int-1
*Sfilm, OP=NEW
surface_nodes, F, 20., air-cool
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** OUTPUT REQUESTS
**
*Restart, write, frequency=1000
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** FIELD OUTPUT: F-Output-1
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*Node Output
NT,
*Element Output
SDV,
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** HISTORY OUTPUT: H-Output-1
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*End Step
** MATERIALS **

*Material, name=S86XX
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*Depvar
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*Initial Conditions, TYPE=field, Var=1 all, 0.002
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*static
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*CONTROLS, PARAMETERS=TIME INCREMENTATION 20, 30
*CONTROLS, FIELD=displacement, PARAMETERS=FIELD 0.05, 0.05, . .

** INTERACTIONS **

** Interaction: Int-1
*temperature, file=3d_cyl_therm, bstep=1, estep=1

**
** OUTPUT REQUESTS

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**
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*Node Output
NT, U
*Element Output
SDV,s
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*El Print, freq=999999
*Node Print, freq=999999
*End Step
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** STEP: Carbon diffusion
**
*Step, name="Carbon Diffusion", AMP=Ramp, inc=1000
Carbon Diffusion into case
*static
75, 21600., 1e-06, 1000,
**
*FIELD, OP=MOD, VAR=1, INPUT=cc.nod
*FIELD, OP=NEW, VAR=2
all, 0
** INTERACTIONS
**
** Interaction: Int-1
**
** OUTPUT REQUESTS
**
*Restart, write, frequency=1000
**
** FIELD OUTPUT: F-Output-1
**
*Output, field, frequency=10
*Node Output
NT,u
*Element Output
SDV,s
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*Node File, freq=0
*End Step
**------------------------------------------------------------
**
** STEP: air transfer
**
*Step, name="air transfer", AMP=STEP, inc=1000
air transfer from furnace to tank
*static
0.01, 12., 1e-06, 0.5,
**
*FIELD, OP=NEW, VAR=2
all, -2.0
** INTERACTIONS
**
** Interaction: Int-1
*temperature, file=3d_cyl_therm, bstep=2, estep=2
**
** OUTPUT REQUESTS
**
*Restart, write, frequency=1000
**
** FIELD OUTPUT: F-Output-1
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*Node Output
NT,u
*Element Output
SDV,s
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*Node File, freq=0
*End Step
**------------------------------------------------------------
**
** STEP: immerse quenching
**
*Step, name="immerse quenching", AMP=STEP, inc=2000
immerse quenching
*static
0.001, 5., 1e-06, 5.,
**
** INTERACTIONS
**
** Interaction: Int-1
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*Restart, write, frequency=1000
**
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*Node Output
NT, u
*Element Output
SDV, s
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*Node File, freq=0
*End Step
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**
** STEP: oil quenching
**
*Step, name="oil quenching", AMP=STEP, inc=2000
oil quenching process
*static
0.0004, 1000., 0.0001, 50.,
**
** INTERACTIONS
**
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*Restart, write, frequency=1000
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** FIELD OUTPUT: F-Output-1
**
*Output, field, frequency=10
*Node Output
NT, u
*Element Output
SDV, s
**
** HISTORY OUTPUT: H-Output-1
**
*Output, history
*Node File, freq=0
*End Step
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**
** STEP: air cool
**
*Step, name="air cool", AMP=STEP, inc=1000
air cool to room temperature
*static
1., 1500., 0.01, 100.,
**
** INTERACTIONS
**
** Interaction: Int-1
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**
** OUTPUT REQUESTS
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** FIELD OUTPUT: F-Output-1
**
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*Node Output
NT,u
*Element Output
SDV,s
**
** HISTORY OUTPUT: H-Output-1
**
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*Node File, freq=0
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Figure D.1  Comparison of the residual hoop stress for the intensive quench simulations for 4320 in three carburized conditions.  (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface.  The times indicated are the quench time in the high intensity quenching process.
Figure D.2 Comparison of the residual hoop stress for the intensive quench simulations for 8620 in three carburized conditions. (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface. The times indicated are the quench time in the high intensity quenching process.
Figure D.3 Deviation of the residual stress of 15 and 20 second hold times relative to the 10 second hold time for high intensity quenching of 4320 in three carburized conditions. (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface. The times indicated in the legend are the quench time in the intensive quenching process.
Figure D.4  Deviation of the residual stress of 15 and 20 second hold times relative to the 10 second hold time for high intensity quenching of 8620 in three carburized conditions. (a) Non-carburized, (b) 0.8 wt pct C at the surface, (c) 1.0 wt pct C at the surface. The times indicated in the legend are the quench time in the intensive quenching process.
APPENDIX E: PREDICTED FINAL MICROSTRUCTURE AFTER HIGH INTENSITY QUENCHING WITH QUENCH TIMES OF 5, 10, AND 20 s
Figure E.1  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched for 5 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.2 Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 5 s under high intensity quenching.
Figure E.3 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched for 5 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.4  Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 5 s under high intensity quenching.
Figure E.5 Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched for 5 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.6  Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 5 s under high intensity quenching.
Figure E.7  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched 10 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.8  Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 10 s under high intensity quenching.
Figure E.9  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched 10 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.10  Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 10 s under high intensity quenching.
Figure E.11  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched 10 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.12  Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 10 s under high intensity quenching.
Figure E.13  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 non-carburized and quenched for 20 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.14 Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of non-carburized 1020, 4120, 4320, and 8620 quenched for 20 s under high intensity quenching.
Figure E.15  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C and quenched for 20 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.16 Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 0.8 wt pct C quenched for 20 s under high intensity quenching.
Figure E.17  Predicted final microstructure phase fractions of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C and quenched for 20 s under high intensity quenching for (a) retained austenite, (b) ferrite, (c) pearlite, (d) bainite, and (e) martensite.
Figure E.18  Predicted results for (a) accumulated plastic strain, (b) centerline distortion, and (c) residual hoop stress of 1020, 4120, 4320, and 8620 carburized to 1.0 wt pct C quenched for 20 s under high intensity quenching.
Figure F.1  Time dependent response of 1020 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) accumulated plastic strain.
Figure F.1 (continued) Time dependent response of 1020 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) accumulated plastic strain.
Figure F.1 (continued) Time dependent response of 1020 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) accumulated plastic strain.
Figure F.2 Time dependent response of 1020 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) accumulated plastic strain.
Figure F.2 (continued) Time dependent response of 1020 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) accumulated plastic strain.
Figure F.2 (continued) Time dependent response of 1020 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) ferrite, (c) pearlite, (d) bainite, (e) martensite, (f) residual hoop stress, (g) centerline deflection, and (h) accumulated plastic strain.
Figure F.3  Time dependent response of 4120 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure F.3  (continued) Time dependent response of 4120 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure F.4 Time dependent response of 4120 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure F.4 (continued) Time dependent response of 4120 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.1  Time dependent response of non-carburized 4320, quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.1 (continued) Time dependent response of non-carburized 4320, quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.2  Time dependent response of 4320 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.2  (continued) Time dependent response of 4320 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.3  Time dependent response of 4320 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.3  (continued) Time dependent response of 4320 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.4  Time dependent response of non-carburized 4320, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.4 (continued) Time dependent response of non-carburized 4320, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.5  Time dependent response of 4320 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.5 (continued) Time dependent response of 4320 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.6  Time dependent response of 4320 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.6  (continued) Time dependent response of 4320 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.7  Time dependent response of non-carburized 8620, quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.7  (continued) Time dependent response of non-carburized 8620, quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.8  Time dependent response of 8620 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.8 (continued) Time dependent response of 8620 carburized to 0.8 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.9  Time dependent response of 8620 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.9 (continued) Time dependent response of 8620 carburized to 1.0 wt pct C at the surface and quenched in 65 °C oil for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.10  Time dependent response of non-carburized 8620, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.10  (continued) Time dependent response of non-carburized 8620, quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.11  Time dependent response of 8620 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.11  (continued) Time dependent response of 8620 carburized to 0.8 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.12  Time dependent response of 8620 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
Figure G.12  (continued) Time dependent response of 8620 carburized to 1.0 wt pct C at the surface and quenched via high intensity quenching and held for 15 s in the quench for (a) temperature, (b) bainite, (c) martensite, (d) residual hoop stress, (e) centerline deflection, and (f) accumulated plastic strain.
APPENDIX H: DETERMINATION OF CRITICAL HEAT TRANSFER RATES

An analysis of the results showed a dependence of residual stress on the heat transfer rate during quenching. In order to better understand this transition from tensile to compressive residual stresses at the surface of the bar, an analysis was done to find “critical heat transfer rate” at which the surface stresses transition from tension to compression. This “critical heat transfer rate” was then generalized to various part geometries based on the Biot number and characteristic length. This work was published in Quenching Control and Distortion 2012: Proceedings of the 6th International Quenching and Control of Distortion Conference Including the 4th International Distortion Engineering Conference, D.S. MacKenzie, B. Liscic, T. Lubben, and H.W. Zoch (ed).
Relationship Between Heat Transfer and Residual Stresses In Three Carburizing Steels

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Keywords: Simulation, Heat Transfer, Residual Stress

Abstract

Heat treatment simulations have grown in importance since they can help to develop processes and reduce the need for trial and error experiments. Simulations can also provide insight into the metallurgical phenomena which occur during heat treatment that are not easily observed using standard laboratory methods. In the present study DANTE® subroutines in conjunction with ABAQUS® were used to simulate heat treatments of three carburizing grade steels, 4120, 4320, and 8620, for a bar with various diameters and a length of 254 mm. The results of these simulations provide relationships between quenching method, microstructure development, and residual hoop stresses. In a non-carburized condition after oil quenching with a low heat transfer rate, a tensile hoop stress component was predicted at and near the surface of the bar. After high intensity quench with a high heat transfer rate, a compressive hoop stress component was predicted at the surface. The Biot number is commonly used to characterize the importance of heat flow at the surface relative to heat flow via conduction through the solid component during heating or cooling for a component of given volume and surface area. The Biot number where the residual hoop stress component on the surface switches from tension to compression was defined in this study as the “critical” Biot number. For a Biot number greater than approximately 2.5 the hoop stress component on the surface of the part was compressive. The value of the critical Biot number was not dependent on alloy content over the range examined.
Introduction

Until recently, the heat treatment industry has often relied on trial and error for developing heat treatment schedules for new parts. Simulations have become more widespread and accepted as a way to reduce the costs of developing new heat treatment procedures. Simulations are also being used to further the basic understanding of phenomena in the internal regions of components.

DANTE®, a heat treatment simulation package, was introduced at the second Quenching and Control of Distortion conference in 1996 [1-4]. Since that time, the material data sets available within DANTE have increased to encompass a number of different alloys as well as additional quench methods [5-6]. Heat transfer coefficients for gas quenching [7-10] and high intensity quenching [11-12] are available in the literature to expand the breadth of simulations that can be performed.

High intensity quenching can produce a residual compressive hoop stress component on the surface of a non-carburized steel part due to the high heat transfer rates during the quench. The compressive stress components near the surface following high intensity quenching are reported to be similar to those found after carburizing [13-16]. Heat transfer coefficients required for high intensity quenching are between 20 and 40 kW/(m² K) [11-12]. By comparison, oil quenching has a peak heat transfer coefficient of approximately 5 kW/(m² K) [6]. The minimum heat transfer coefficient to form compressive stress components on the surface after quenching has not yet been identified. This work investigates the minimum heat transfer coefficient to obtain compressive stress components on the surface for three common bar steels.

Simulation Parameters

DANTE simulations were conducted utilizing a cylindrical bar to minimize geometry effects. The bar had a diameter of 25.4 mm and length of 254 mm. The simulations were run as a 3-dimensional half bar to allow horizontal quenching. Figure H.1 shows the fine mesh spacing used radially near the surface to account for high heat transfer rate effects. The nodes were radially spaced 0.3 mm (0.012 in) apart to a depth of 1.5 mm (0.059 in) below the surface. Beyond this depth, radial node spacing was increased to 1.02 mm (.040 in) in the core region.
Nodes were also spaced 0.3 mm (0.012 in) apart to a depth of 1.5 mm (0.059 in) from each end of the bar. For the remaining length, the nodes were spaced 5.02 mm (0.198 in) apart.

Figure H.1  Schematic of element layout for the 3-dimensional 25.4 mm diameter cylindrical bar (top) and a close up of the left end (bottom) used in the simulations. The fine elements extend to a depth of 1.5 mm from all exterior surfaces. Data were collected from approximately the center of the length of the bar. The centerline surface is a mirror plane and not an exterior surface in the simulations.

Simulations were performed utilizing the material data sets in DANTE for 4120, 4320, and 8620. The simulations were conducted using two different quench methods: a 65 °C oil quench and a 20 °C water high intensity quench (HIQ), which is presumed to be similar to one variant of the IntensiQuench® process [14-16]. Table H.1 shows the heat treatment steps, in seconds, for both quench methods; the differences between the two heat treatments start at the immersion step. The immersion step is where the bar is horizontally submerged into the quenchant. During the oil quench, the entire part reaches the oil temperature of 65° C and is then allowed to air cool to room temperature. During high intensity quenching, the bar was held in the quenchant for 15 s (i.e. quench hold time) then removed from the highly agitated water and allowed to air cool to room temperature. (Quench hold times of 5, 10, 15, and 20 s were simulated for HIQ, with 15 s providing the most favorable residual hoop stress component profile.)
Table H.1: Heat Treatment Schedule in Seconds

<table>
<thead>
<tr>
<th></th>
<th>Oil</th>
<th>HIQ</th>
</tr>
</thead>
<tbody>
<tr>
<td>Heat Up</td>
<td>1000</td>
<td>1000</td>
</tr>
<tr>
<td>“Carburization”</td>
<td>21600</td>
<td>21600</td>
</tr>
<tr>
<td>Transfer</td>
<td>12</td>
<td>12</td>
</tr>
<tr>
<td>Immersion</td>
<td>5</td>
<td>0.5</td>
</tr>
<tr>
<td>Quench</td>
<td>1000</td>
<td>15</td>
</tr>
<tr>
<td>Air Cool</td>
<td>1500</td>
<td>3000</td>
</tr>
</tbody>
</table>

The quench was simulated with the heat transfer coefficient as a function of temperature. The oil quenching heat transfer coefficient as a function of temperature is embedded within DANTE. The heat transfer coefficient peaks between 500 °C and 550 °C at 5 kW/(m² K) [6]. Heat transfer coefficients were found in literature for intensive quenching that range from 20 to 40 kW/(m² K) [11-12]. For the simulations presented in this study, the heat transfer coefficient for high intensity quenching was set at 20 kW/(m² K) and held constant throughout the quench hold time.

Results: 3-Dimensional Study

Figure H.2 shows the predicted martensite fraction and residual hoop stress profiles for the two quenching conditions for the three steels. The data shown in Figure H.2 were taken from a cross section near the center of the bar and were unaffected by heat transfer from the ends. These results indicate a significant difference in the residual hoop stress components within the part for the two quenching methods. For oil quenching the fraction that is not martensite is mostly bainite. After high intensity quenching, all three steels have similar martensite fractions. All three steels possess a tensile hoop stress component at the surface when oil quenched and compressive component when processed with high intensity quenching. With a compressive hoop stress component on the surface the initiation of axial surface cracks would be inhibited. The high intensity quenching also changes the character of the residual hoop stress profile through the presence of a subsurface tensile peak between the compressive values at the surface and in the center of the bar.

The transition from tensile to compressive in the residual hoop stress component near the surface has been reported in other simulation studies [11-15]. The minimum heat transfer coefficient required to achieve this transition has not been investigated. The heat transfer
coefficient required to produce a zero residual hoop stress at the surface after quenching can be found through a set of simulations that systematically vary the heat transfer coefficient. The heat transfer coefficient that produces a zero residual hoop stress on the surface is defined here as the “critical” heat transfer coefficient. To determine the transition conditions for components with different dimensions the "critical" heat transfer coefficient can be used in the Biot number allowing a “critical” Biot number to be estimated.

Results: 2-Dimensional Parametric Study

The 3-dimensional simulations confirmed and extended the results of other studies but did not identify the “critical” heat transfer coefficient. In order to investigate a large number of parameter variations, further simulations were conducted using a 2-dimensional axi-symmetric model for a bar with the same 25.4 mm diameter and length of 254 mm. The mesh was refined near the surface to account for the high heat transfer rates. To ensure consistency between the 2-dimensional and 3-dimensional models, the results of the two models were compared. Figure H.3 shows that the results from the 2-dimensional model are consistent with the 3-dimensional model for the three steel grades after oil quenching. Similar consistency was seen for the high intensity quenching simulations.
Figure H.3  Comparison of the predicted (a, c, e) microstructure (martensite and bainite), and (b, d, f) residual hoop stress for 4120, 4320, and 8620 respectively from the 3-dimensional and 2-dimensional models after oil quenching. The results from the 2-dimensional model are similar to the results from the 3-dimensional model for the three steels. Please note the different scales for each material.
Since the results from the 2-dimensional and 3-dimensional simulations are consistent, the 2-dimensional model with its reduced computational time was used to establish the "critical" heat transfer coefficients. A series of simulations for each material was conducted varying the heat transfer coefficient between 5 and 15 kW/(m² K) in steps of 2.5 kW/(m² K). For each of these simulations, the high intensity simulation was used as the starting point with a 15 s quench time. The residual hoop stress profile was obtained from these simulations. Figure H.4 shows the residual hoop stress component at the surface as a function of the heat transfer coefficient. Once heat transfer coefficients that bracketed the zero residual hoop stress on the surface were found, further simulations were run in increments of 500 W/(m² °C) to better determine the “critical” heat transfer coefficients.

From these data, the “critical” heat transfer coefficients were determined by linear interpolation for each steel grade. Table H.2 shows the predicted “critical” heat transfer coefficients for the three steels. All three steels have similar critical heat transfer coefficients of approximately 9 kW/(m² K).

The effects of varying quench hold time was also investigated for 8620. Figure H.5 shows that as the quench hold time increased, the “critical” heat transfer coefficient varied only slightly and remained at approximately 9 kW/(m² K). Quench hold times longer than 30 s were
**Table H.2**: Heat transfer coefficients calculated via linear interpolation predicted to create zero residual hoop stress at the surface. Heat transfer coefficients above the critical value result in surface compression while lower coefficients result in surface tensile hoop stresses.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Critical Heat Transfer Coefficient [kW/(m² K)]</th>
</tr>
</thead>
<tbody>
<tr>
<td>4120</td>
<td>8.95</td>
</tr>
<tr>
<td>4320</td>
<td>8.85</td>
</tr>
<tr>
<td>8620</td>
<td>8.99</td>
</tr>
</tbody>
</table>

**Figure H.4**  Predicted residual hoop stress as a function of heat transfer coefficient for non-carburized heat treatment of 4120, 4320, and 8620 steels.

**Figure H.5**  Predicted residual hoop stress on the surface as a function of heat transfer coefficient for non-carburized heat treatment of 8620 steels with a varied quench hold times.
not investigated as the bar was predicted to be fully quenched to 20 °C within 30 s at heat transfer coefficients of 5 kW/(m² K) and higher.

Generalization to Other Geometries

The applicability of the simulation results utilizing a 25.4 mm diameter bar can be extended to other bar sizes and other component geometries through determination of “critical” Biot number. The Biot number is given by

\[ Bi = \frac{h \times L_C}{k} \]  

(1)

where \( Bi \) is the Biot number, \( h \) is the surface heat transfer coefficient, \( L_C \) is a characteristic length, and \( k \) is the thermal conductivity of the material. The characteristic length is the ratio of the volume of the component to the surface area from which heat is being extracted. For a cylinder, the characteristic length is given by

\[ L_C = \frac{Volume}{Surface Area} = \frac{r \times l}{2 \times (r + l)} \]  

(2)

where \( r \) and \( l \) are the radius and length of the cylinder respectively.

In order to generalize the results, the 25.4 mm diameter from the original simulation sets was scaled to 6.35 mm, 12.7 mm, 19.05 mm, and 31.75 mm. Simulations defining the critical heat transfer coefficient for each bar size were run using the 2-dimensional axisymmetric model. Figure H.6 shows that all three steels follow the same pattern – as the diameter of the bar decreases, the “critical” heat transfer coefficient increases. All three alloys show similar values for the heat transfer coefficient at the transition from tensile to compressive for the residual hoop stress component on the surface. Figure H.7 summarizes the increase in “critical” heat transfer coefficient as a function of diameter.
Figure H.6 Residual hoop stress at the surface as a function of heat transfer coefficient over a range of bar diameters for non-carburized heat treatment of (a) 4120, (b) 4320, and (c) 8620 steels. The bar diameter variation is shown as individual series with the crossover from tensile to compressive residual hoop stress occurring at higher heat transfer coefficients as the diameter decreases.

The diameter in Figure H.7 can be converted to characteristic length using Equation (2). Figure H.8 shows the critical heat transfer coefficient as a function of the characteristic length. The results can be applied to bars of other sizes by determining the characteristic length for the new size and obtaining the critical heat transfer coefficient from Figure H.8. When a bar has a small characteristic length (low volume to surface area ratio), a high heat transfer coefficient is
required to create a zero or compressive residual hoop stress component on the surface after quenching. This relationship has been suggested in the “intensive quenching” literature [14-16].

To generalize the critical heat transfer coefficient to other shapes, the Biot number can be used. The thermal conductivity of steel, which varies due to the microstructural constituents present and the temperature, needs to be specified. To simplify, a single value of the thermal conductivity for each steel was used in the calculations. The thermal conductivity used was for an austenitic microstructure at 600 °C. Based on this assumption, Figure H.9 shows the “critical” Biot number as a function of characteristic length. The three steels have approximately the same “critical” Biot number over the range investigated. The critical Biot number is approximately constant at a value of 2.5. Hence, any new part being considered should be expected to have compressive residual hoop stress components after quench if the Biot number is greater than 2.5. Conversely, lower Biot numbers would be associated with tensile residual stresses.

![Figure H.7](image)

Figure H.7 Variation in critical heat transfer coefficient as a function of bar diameter for the three steel alloys.
Figure H.8 Variation in critical heat transfer coefficient as a function of characteristic length for the three steel alloys.

Figure H.9 Variation in critical Biot number as a function of characteristic length for non-carburized 4120, 4320, and 8620.

Conclusions

The response to varied quenching methods of bar steels used in common heat treated components was investigated via DANTE simulation which couples thermal, mechanical, and
transformation behavior. After oil quenching, the microstructure was predicted to be comprised of bainite and martensite of varying fractions. The residual hoop stress components were predicted to be tensile near the surface and compressive near the core of the part for an oil quench. When the heat transfer coefficient was increased, the martensite fraction increased while the residual hoop stress components were predicted to be compressive near the surface.

The heat transfer rate required to reverse the residual hoop stress profile from tensile to compressive was investigated. The residual hoop stresses reversed from tensile to compressive at a heat transfer coefficient of approximately 9 kW/(m² K) when quenching a 25.4 mm diameter bar. As the diameter of the bar decreases, the “critical” heat transfer coefficient increases to as much as 40 kW/(m² K) for a 6.35 mm diameter bar. To generalize to other part geometries and steel grades, the “critical” heat transfer coefficient was converted to a “critical” Biot number. For Biot numbers greater than 2.5, it is expected the part will have compressive residual stress components near the surface after quenching. If the specified quench, which sets the heat transfer from the part, the part geometry, which sets the characteristic length, and the material, which sets the thermal conductivity, lead to a Biot number greater than 2.5, compressive residual stress components at the surface are expected.

Acknowledgments

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References


APPENDIX I: COMPARING SIMULATIONS TO EXPERIMENTAL WORK

A bearing cup shape made from 5120 hot rolled bar steel was selected to compare experimental results to a DANTE simulation of the process. The ring is shown in Figure I.1. Initial microstructures were obtained from a raw cup provided by Timken as shown in Figure I.2. Image analysis of the microstructure across multiple fields of view found the microstructure to average 60% ferrite with 40% pearlite. This microstructure was used as the starting point for the simulations.

Figure I.1  Part for experimental verification work. Ring has an outer diameter of approximately 50.8 mm (2 in), an inner diameter of approximately 40.64 mm (1.6 in), and a height of approximately 10.16 mm (0.4 in).

Figure I.2  Initial microstructure of 5120 ring used in experimental verification trials. Analysis of multiple areas showed an average of 60 percent ferrite and 40 percent pearlite. Sample orientation is unknown. Etched with two percent Nital.
Six heat treatments and associated simulations were completed. The first was vacuum carburized and quenched in oil. The second was vacuum carburized and quenched using 1 MPa (10 bar) nitrogen gas. The third and fourth runs were vacuum carburized and stepped nitrogen gas quenched using two different quench gas parameters. The fifth was non-carburized and intensively quenched while the sixth was carburized and intensively quenched. Table I.1 contains a summary of the quench methods for each batch.

<table>
<thead>
<tr>
<th>Batch</th>
<th>Quench Method</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>55 °C (203 °F) for 20 min</td>
</tr>
<tr>
<td>2</td>
<td>Nitrogen Gas 1 MPa (10 bar) for 10 min</td>
</tr>
<tr>
<td>3</td>
<td>Stepped Nitrogen Gas 1.2 MPa (12 bar) for 12 s, no pressure for 7 s, 581 s of 1 MPa (10 bar)</td>
</tr>
<tr>
<td>4</td>
<td>Stepped Nitrogen Gas 1.2 MPa (12 bar) for 18 s, no pressure for 15 s, 567 s of 1 MPa (10 bar)</td>
</tr>
<tr>
<td>5</td>
<td>Intensive Quench</td>
</tr>
<tr>
<td>6</td>
<td>Carburized and Intensive Quench</td>
</tr>
</tbody>
</table>

Initial heat treatments were carried out by ECM for Batches 1-4 and to carburize the rings for Batch 6. Bearing cups were loaded onto racks that when stacked measured 24” by 39” by 24” tall (609.6 mm by 990.6 mm by 609.6 mm tall) as shown in Figure I.3. Each load was comprised of 9 racks stacked to provide a full load. The experimental cups were placed on layer 4 with scrap cups placed on the other layers. ECM provided the heat treatment schedule as: heat to 700 °C (1292 °F) in 20 min; ramp the temperature from 700 °C (1292 °F) to 955 °C (1751 °F) over one hour; hold for 40 min at 955 °C (1751 °F); provide carbon boost for 953 s (15 min 53 s) followed by a diffuse time of 1 hour 54 min 7 s for a total time of 2 hrs and 10 min; decrease the temperature to 844 °C (1551 °F) over 35 min; hold at 844 °C (1551 °F) for 15 min; and quench. Due to confidentiality concerns, ECM did not provide the individual boost-diffuse cycles.

The oil quench consisted of an oil bath at 55 °C (203 °F) for 20 min. The straight nitrogen quench was a pressure of 1 MPa (10 bar) for 10 min. The stepped nitrogen quench consisted of 1.2 MPa (12 bar) for 12 s, no pressure for 7 s, 581 s of 1 MPa (10 bar) for the third group while the fourth group was quenched using 1.2 MPa (12 bar) for 18 s, no pressure for 15 s, 567 s of 1 MPa (10 bar). The second set of stepped nitrogen quench parameters were the result of a number of simulations utilizing various combinations of gas parameters to determine an
optimal combination to improve residual stress levels and decrease distortion (diameter changes) [66].

Figure I.3  Image of rack set up used in the heat treatment of the bearing cups.

The carburized and intensively quenched samples were carburized by ECM along with the straight nitrogen gas quenched samples and shipped with the raw cups to Intensive Quench Technologies (IQT) for heat treating and quenching. The experimental heat treatment at IQT consisted of heating to 857 °C (1575 °F) over 35 minutes, an air transfer to the quench over 8 seconds, followed by intensively quenching for 8 seconds. The simulations assumed a heat transfer rate of 20 kW/(m² °C). This heat treatment was the only simulation completed for the non-carburized parts while these additional steps were added to the straight nitrogen gas quench simulation for the carburized cups.

Due to privacy concerns of ECM, the exact boost and diffuse cycles used for the carburization cycles were not provided. As a result, numerous carburization simulations were
completed to estimate a boost-diffuse cycle that could result in a similar carbon profile to the one obtained from the completed parts. Actual cooling curve data for multiple locations inside two different stepped gas quenching loads were provided by ECM. One set of the data provided was the same as the heat treatment for Batch 3. The other step-gas quenching utilized 1.8 MPa (18 bar) nitrogen with the recirculating fans set at 30% for 6 s then the fans were increased to 100% for 6 s. Next the gas was stopped for 7 s followed by 581 s of 1 MPa (10 bar) nitrogen. The cooling curves were taken from the center of the loads and re-plotted. Figure I.4 shows the provided cooling curve for the 1.2 MPa (12 bar) step-gas quenching process. The two vertical lines indicate the 12 s for the initial quench and the 7 s hold time that follows. A series of simulations was completed with different heat transfer rates. The predicted and actual cooling curves were compared after each simulation to refine the simulated cooling curve until the simulated cooling curve was similar to the actual cooling curves.

![Figure I.4](image)

**Figure I.4** Cooling curve for the center of the load as quenched by ECM for 12 s using 1.2 MPa (12 bar) nitrogen, held for 7 s with no gas circulation, and final quench using 1.0 MPa (10 bar) nitrogen. Vertical lines indicate the initial quench and the hold.

A two dimensional axi-symmetric model was created in ABAQUS for the cup. DANTE simulations were then completed initially for Batches 1, 2, 3. As mentioned, additional simulations not shown were completed by the author utilizing varying stepped nitrogen gas parameters to allow comparisons and to design optimized quench gas parameters for reduced
distortion and improved residual stress in the experiments. The stepped gas recipe used by ECM for Batch 4 was the result of this optimization. Finally, simulations of intensive quenching were completed including the full simulation of the carburization and straight nitrogen quench for the carburized and intensively quenched condition. For a simulation to complete successfully, the equation results must converge at each time step. If a time step was found to not converge, ABAQUS reduced the time in the step and recalculated the results until convergence was found. If the minimum time for each step was reached without convergence, the simulation was aborted. Unfortunately, the simulations using cooling rates approximating gas quenching did not converge. In order to complete the gas quench simulations, the latent heat of transformation on cooling values used by DANTE were reduced to half of the full value. This deviation allowed the simulations to complete successfully, but clearly may affect the accuracy of the calculated results. Further, the simulations only considered a single bearing cup isolated from any effects of fixture mass and assumed an even heat transfer rate across the entire surface of the cup. These assumptions resulted in variations of the simulation compared to the actual heat treatments. (In order to better simulate the actual heat treatments, the full rack set up shown in Figure I.3 could be simulated using computational fluid dynamics to obtain the gas flow through the rack followed by the DANTE simulations of the cup located on a rack using heat transfer rates as modified by the gas flow through the rack.)

Microstructure predictions are shown for each of the six conditions in Figure I.5 through Figure I.9. For the oil quench and the carburized plus intensively quenched conditions, only retained austenite and martensite are predicted with the peak retained austenite in the case, and the lower carbon core comprised essentially of 100% martensite. For the three gas quenched conditions, the cup is predicted to contain retained austenite and martensite in the case with the peak martensite fraction near the case-core interface while the core is comprised of ferrite and bainite. The decreased martensite fraction in the gas quenched simulations was expected due to the lower cooling rates. The peculiar appearance of the Batch 5 results is an artifact of the scale used in the figure where the difference between the maximum and minimum martensite fractions is only 0.0032 volume fraction.

One cup from each batch was provided for microstructural analysis. The light optical micrographs of the case and core regions for each condition are shown in Figure I.11 through Figure I.16. For Batch 1, the case was found to contain predominately martensite with retained
austenite while the core appeared to contain martensite (Figure I.11). The Batch 2 case appears
to be predominately martensite and retained austenite while the core appears to be a mixed
microstructure (Figure I.12). Figure I.13 and Figure I.14 show similar microstructures to that of
Batch 2 for Batch 3 and Batch 4, respectively. The main difference appears to be in the core
microstructure with Batch 4 appearing to contain less ferrite than the other two gas quench
conditions. The non-carburized and intensively quenched condition did not show a
microstructural difference between the “case” and “core” regions. This was expected as there
was no additional carbon in the “case” and resulted in a consistent microstructure appearing to be
martensite with a small fraction of retained austenite (Figure I.15). The carburized and
intensively quenched condition fully hardened throughout the microstructure with a similar
change in martensite morphology between case and core as was noted in the oil quenched
condition. In all cases, the predicted microstructures appeared to agree well the observed
microstructures.

![Figure I.5](image)

Figure I.5 DANTE predicted (a) retained austenite, and (b) martensite in the
microstructure for Batch 1 which was carburized and oil quenched. The scales
are independent and shown in microstructure fractions. Note the other potential
microstructural constituents were predicted to be less than 0.01 percent.
Figure I.6 DANTE predicted (a) retained austenite, (b) ferrite, (c) upper bainite, (d) lower bainite, and (e) martensite in the microstructure for Batch 2 which was carburized and gas quenched. The scales are independent and shown in microstructure fractions. Note pearlite content was predicted to be less than 0.01 percent.
Figure I.7 DANTE predicted (a) retained austenite, (b) ferrite, (c) upper bainite, (d) lower bainite, and (e) martensite in the microstructure for Batch 3 which was carburized and step gas quenched. The scales are independent and shown in microstructure fractions. Note pearlite content was predicted to be less than 0.01 percent.
Figure I.8  DANTE predicted (a) retained austenite, (b) ferrite, (c) upper bainite, (d) lower bainite, and (e) martensite in the microstructure for Batch 4 which was carburized and the “optimized” step gas quenched. The scales are independent and shown in microstructure fractions. Note pearlitic content was predicted to be less than 0.01 percent.
Figure I.9  DANTE predicted (a) retained austenite, and (b) martensite in the microstructure for Batch 5 which was non-carburized and intensively quenched. The scales are independent and shown in microstructure fractions. The predicted microstructures appear different due to the limited range of the scale used. Note the other potential microstructural constituents were predicted to be less than 0.01 percent.

Figure I.10  DANTE predicted (a) retained austenite, and (b) martensite in the microstructure for Batch 6 which was carburized and intensively quenched. The scales are independent and shown in microstructure fractions. Note the other potential microstructural constituents were predicted to be less than 0.01 percent.
Figure I.11 Light optical micrographs of the (a) case, and (b) core microstructure for Batch 1 which was carburized and oil quenched. Etched with 3% Nital.
Figure I.12 Light optical micrographs of the (a) case, and (b) core microstructure for Batch 2 which was carburized and gas quenched. Etched with 3% Nital.
Figure I.13  Light optical micrographs of the (a) case, and (b) core microstructure for Batch 3 which was carburized and step gas quenched. Etched with 3% Nital.
Figure I.14 Light optical micrographs of the (a) case, and (b) core microstructure for Batch 4 which was carburized and the “optimized” step gas quenched. Etched with 3% Nital.
Figure I.15  Light optical micrographs of the (a) case, and (b) core microstructure for Batch 5 which was intensively quenched. Etched with 3% Nital.
Figure I.16 Light optical micrographs of the (a) case, and (b) core microstructure for Batch 6 which was carburized and intensively quenched. Etched with 3% Nital.
The predicted dimensional changes for each condition are shown in Figure I.17. The changes were magnified 10x and are compared to the original shape. The carburized cups were all predicted to grow in height as well as in diameter. The intensively quenched sample was predicted to be essentially unchanged in height but shrink radially. Timken provided dimensional data from the experimentally heat treated cups as shown in Table I.2 with standard deviation in the measurements in Table I.3. The actual cups were found to grow in height but shrink radially for the oil and gas quenched conditions. The intensively quenched cups were also found to have grown in height and grew in diameter. A Tukey-Kramer means comparison test was completed on the experimentally measured dimensions. This analysis resulted in a list of statistically similar groups for each of the dimensions measured and is shown in Table I.4. A schematic indicating the dimensional changes for Batch 1 and Batch 5 are shown in Figure I.18 and I.19, respectively. Unfortunately, the measured radial distortion was found to be opposite the predicted distortion.

Predicted residual hoop stress data for each condition are shown in Figure I.20. The carburized conditions all were predicted to have compressive residual hoop stresses at the surfaces. The non-carburized intensively quenched condition was predicted to have a varied stress state around the surface of the cup. Experimental residual stress data were not available. Vickers microhardness data were collected through the case of each sample using a 500gf load as shown in Figure I.21. The core hardness of the three gas quenched batches was found to be softer than the other three conditions indicating the microstructure in the core is not martensitic, matching the observed microstructures. The non-carburized and intensively quenched batch was found to have a consistent hardness through the depth and corroborates the fully martensitic microstructure noted. The data for carburized then oil quenched or intensively quenched conditions showed that, as expected, the carburized case had an increased hardness due to the higher carbon content in the martensite.
Figure I.17 DANTE predicted distortion with original part outline for (a) Batch 1, (b) Batch 2, (c) Batch 3, (d) Batch 4, (e) Batch 5, and (f) Batch 6. Note the scales are specific to each batch. Distortion was magnified 10x for visual clarity.
Table I.2 – Average Measured Growth in Outer and Inner Diameters and Height as Provided by Timken where a Negative Value Represents Shrinkage (mm)

<table>
<thead>
<tr>
<th>Batch</th>
<th>Height</th>
<th>Upper OD</th>
<th>Lower OD</th>
<th>Upper ID</th>
<th>Lower ID</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.0117</td>
<td>-0.0264</td>
<td>-0.0287</td>
<td>-0.0340</td>
<td>-0.0620</td>
</tr>
<tr>
<td>2</td>
<td>0.0056</td>
<td>-0.0287</td>
<td>-0.0318</td>
<td>-0.0610</td>
<td>-0.0887</td>
</tr>
<tr>
<td>3</td>
<td>0.0056</td>
<td>-0.0122</td>
<td>-0.0206</td>
<td>-0.0493</td>
<td>-0.0838</td>
</tr>
<tr>
<td>4</td>
<td>0.0030</td>
<td>-0.0259</td>
<td>-0.0302</td>
<td>-0.0721</td>
<td>-0.1026</td>
</tr>
<tr>
<td>5</td>
<td>0.0127</td>
<td>0.1600</td>
<td>0.1646</td>
<td>0.2073</td>
<td>0.2474</td>
</tr>
<tr>
<td>6</td>
<td>0.0279</td>
<td>0.0630</td>
<td>0.0584</td>
<td>-0.0429</td>
<td>0.0048</td>
</tr>
</tbody>
</table>

Table I.3 – Standard Deviation of Measured Growth in Outer and Inner Diameters and Height Provided by Timken (mm)

<table>
<thead>
<tr>
<th>Batch</th>
<th>Height</th>
<th>Upper OD</th>
<th>Lower OD</th>
<th>Upper ID</th>
<th>Lower ID</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.005334</td>
<td>0.006858</td>
<td>0.007112</td>
<td>0.009144</td>
<td>0.008636</td>
</tr>
<tr>
<td>2</td>
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<td>0.008128</td>
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<td>0.017018</td>
</tr>
<tr>
<td>3</td>
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<td>0.012954</td>
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</tr>
<tr>
<td>4</td>
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<tr>
<td>5</td>
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<td>0.021336</td>
</tr>
<tr>
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<td>0.013970</td>
<td>0.028702</td>
<td>0.033782</td>
<td>0.019558</td>
</tr>
</tbody>
</table>

Figure I.18 Comparison of the predicted and measured distortion for Batch 1, the carburized and oil quenched condition.
Figure I.19 Comparison of the predicted and measured distortion for Batch 5, the non-carburized and high intensity quench condition.

Table I.4 – Results of a Tukey-Kramer Means Comparison showing Statistically Similar Heat Treatment Batches based on the Experimentally Measured Dimensional Changes

<table>
<thead>
<tr>
<th>Height</th>
<th>Inner Diameter</th>
<th>Outer Diameter</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Large End</td>
<td>Small End</td>
</tr>
<tr>
<td>1, 5</td>
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<td>1, 3, 6</td>
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<tr>
<td>2, 3, 4</td>
<td></td>
<td>2, 3, 6</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Figure I.20  DANTE predicted residual hoop stress for (a) Batch 1, (b) Batch 2, (c) Batch 3, (d) Batch 4, (e) Batch 5, and (f) Batch 6.  Note the scales are specific to each batch with compressive stresses in blue and tensile stresses in red.
Figure I.21  Measured Vickers hardness using a 500 gf load (HV0.5) as a function of depth for each of the heat treatments completed. Note the increase in Vickers hardness at depths above 2.5 mm is the result of the case on the other side of the cup.

In this study, the 2D axis-symmetric simulations assumed constant heat transfer rates across the entire surface of the bearing cup and effects of racking were also ignored. The change in gas flow through the rack could change the flow and resulting in a change in heat transfer rates over the surface of the cup or from one cup to another within the furnace charge. Further, for gas quenched samples, the latent heat of transformation was reduced to allow the model calculations of the thermal model to converge. The microstructures predicted were found to agree with experimental microstructures well while the measured radial growth (shrinkage) of the cups was found to be opposite of the predictions. Ignoring the issues that may be associated with the material properties internal to DANTE, better understanding of the detailed thermal processing parameters applicable to the experimental heat treatments would be needed to better assess the origin of differences between the experimental and calculated results. The substantial importance of several thermal processing variables was illustrated in Appendix J.
APPENDIX J: INFLUENCE OF USER ACCESSIBLE MATERIAL PROPERTIES ON SIMULATION RESULTS

Due to the commercial nature of DANTE®, very few material parameters are available to the user for fine-tuning of the models. The variables specific heat, latent heat, and thermal conductivity of the steels can be modified by the user. Specific heat and thermal conductivity were defined to have a linear temperature dependence within DANTE while latent heat was defined by individual values for austenite decomposition into ferrite, pearlite, bainite, and martensite. The latent heat values for austenite decomposing into ferrite, pearlite, and bainite were the same with a separate value used for the martensitic transformation.

In order to assess the influence of these parameters on the model results, 2D axisymmetric simulations were completed with modifications to the preset values in the values for these variables, along with austenite temperature prior to quenching and the quenchant temperature. Both the slope and intercept values for the linear temperature dependence of specific heat and thermal conductivity for all phases were modified equally for the comparison. Similarly, the latent heat values for all potential austenite decomposition products were adjusted proportionally. The baseline heat treatment applied was the same as the oil quench condition used in the main work without the immersion step, and utilized 4120 steel. These calculations allow an assessment of the sensitivity of the results to the variations of these property inputs.

A decrease in latent heat or specific heat is expected to result in an increase in martensite fraction while a decrease in thermal conductivity is expected to decrease martensite fraction. The simulation results are shown in Figure J.1, Figure J.2, Figure J.3 for latent heat, specific heat, and thermal conductivity, respectively. These results agree with the expected behaviors.

Increases in austenite temperature prior to quenching are expected to lead to a more severe quench, resulting in an increase in martensite. Figure J.4 shows very little difference in the final microstructures for the various temperatures selected. Finally, increasing the quenchant temperature is expected to lead to a less severe quench and a decrease in martensite in the final microstructure. Figure J.5 shows that DANTE does predict lower martensite fractions with increasing quenchant temperature.
The sensitivity of the results to these material properties can be substantial. The results indicate that accurate understanding of thermal characteristics will be needed to generate accurate predictions using DANTE. Uncertainty in the details of the thermal treatments associated with the bearing cup work in Appendix I may have contributed substantially to the differences between measured and calculated results.

Figure J.1 (a) Final martensite fraction and (b) predicted residual hoop stresses when the latent heat of transformation for austenite decomposition were modified from the original value to the percentage shown.
Figure J.2  (a) Final martensite fraction and (b) predicted residual hoop stresses when the temperature dependent specific heat for each phase were modified from the original value.

Figure J.3  (a) Final martensite fraction and (b) predicted residual hoop stresses when all the temperature dependent thermal conductivity for each phase was modified from the original value.
Figure J.4  (a) Final martensite fraction and (b) predicted residual hoop stresses when the starting austenite temperature was modified.

Figure J.5  (a) Final martensite fraction and (b) predicted residual hoop stresses when the quenchant temperature was modified.