AN EXPERIMENTALLY GENERATED CONSTITUTIVE MODEL FOR PEAK
STRESS (σ_{peak}) IN COMPRESSION SAMPLES

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ABSTRACT

The hot working behavior of AISI 1018 steel was studied by hot-compression deformation tests on the Gleeble 1500 thermo-mechanical simulator at true strain values of -0.143 and -0.405, true strain rate values of 0.01 and 0.1, and working temperatures of 900°C and 1000°C. The tests show that a lower working temperature and lower true strain value results in a greater maximum compressive force. The apparent activation energy Q_{app} was calculated by using the Zener-Hollomon parameter combined with the low

stress law. Q_{app} was calculated to be $311 \frac{kJ}{mol K}$.

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Keywords: compression test; Gleeble; recrystallization; static recrystallization; dynamic recrystallization; metadynamic recrystallization; low-carbon steel; Zener-Hollomon; simulation of hot-working; austenite.

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CHAPTER 1

Introduction

In the realm of material science, understanding the characteristics of a material entails knowing how the specimen performs and behaves under external forces such as stress, strain, strain rate, temperature, etc. In the field of metallurgy this has a special meaning because hot deformation of metals has been used to form metals into useful shapes and to improve material properties for centuries. The understanding of metals and alloy behavior during hot deformation is of special importance to designers of metal forming processes (forging, hot rolling, and extrusion) because of the role it has on the kinetics of metallurgical transformation [1]. The technique of forging has been around for a very long time, and in that span of time, it was commonly understood that metal is most easily shaped when it is hot. Only recently has the metallurgical community begun to understand the behavior of the crystals and atoms during hot-working. This relatively new knowledge of metallurgic behavior has led to the development of constitutive equations of materials from experimentally obtained data. These equations are able to quantitatively describe the relationship between flow stress and strain, strain rate, and temperature in industrial hot-working applications [2].

Iron-carbon alloys, such as steel, are among the prominent structural engineering materials. The versatility of steels is evidenced by the various types that are manufactured. There are very soft steels used for deep-drawing applications like car fenders and stove panels, and on the other end, there are extremely hard and tough steels used for gears and bulldozer blades. Figure 1 is a Fe-Fe₃C phase diagram [3].

Fig. 1. Fe-Fe₃C phase diagram.

As the phase diagram shows there are three main types of iron-carbon: Ferrite (α -iron), Austenite (γ -iron), δ -iron, and Cementite. Ferrite, or α -iron, is the structural modification of pure iron at room temperature; ferrite is relatively soft and ductile. The very low solubility of carbon (< 1 carbon/1000 iron atoms) is attributed to the body-centered cubic structure of ferrite. The spaces among the atoms are too small to readily accommodate even a small carbon atom. Austenite, or γ -iron, is the face-centered modification of iron; it is the most stable form of pure iron at temperature between 912°C and 1394°C. Within this range of temperatures, austenite is soft and ductile and well suited for fabrication processes. Despite being a face-centered cubic structure with larger interatomic spacings than ferrite, the holes are barely big enough to allow carbon into the

interstices. This crowding of carbon atoms introduces strain into the structure. Above 1394°C, austenite turns into δ -iron; the crystal structure reverts back to a body-centered-cubic phase. δ -iron is the same as α -iron except that it operates at an elevated temperature range. Lastly, the region of phase diagram labeled cementite is where the carbon is in excess of the solubility limit causing the formation of a second phase. The crystal lattice of cementite contains a 3:1 of iron to carbon atoms, thus the chemical composition is Fe_3C . In comparison to austenite and ferrite, cementite is very hard. Pure iron carbide is nonductile and cannot adjust to stress concentrations, thus it is relatively weak by itself.

This study will be performing hot-compression tests at 900°C and 1000°C using AISI 1018 Steel; this steel is a plain-carbon steel with an approximate carbon content of 0.18% carbon, and according to the diagram, at 0.18% carbon at 900°C and 1000°C, 1018 Steel is in the austenite or γ -iron region [4].

Recrystallization (RX)

The material properties of steel and its alloys are ultimately governed by their crystalline structure. A metal's crystalline structure is dependent on a number of factors; the percent alloy composition of a metal, the type and number of processes the metal is subjected to, and the parameters of the metal working processes (temperature, ϵ , and $\dot{\epsilon}$). This paper will look in depth at the concept of Recrystallization (RX), but more specifically Dynamic RX (DRX). Recrystallization is defined as the process of growing new crystals from previously deformed crystals [5]. For steels, it is known as the process that causes the formation of new grains by heat treating cold-worked steel [6]. Recrystallization is also subdivided into three sub-RX behaviors: Static RX (SRX),

Dynamic RX (DRX), and Metadynamic RX (MDRX). Static RX is the RX as a metal, in this case steel, is deformed via rolling, torsion, tension, or compression, and is left alone to cool and recrystallize without any further work performed on the metal. Dynamic RX is when nucleation and growth occur during deformation as opposed to afterwards.

Metadynamic RX is defined as the post-dynamic softening that occurs after DRX [7].

Recrystallization nucleation is governed by a number of design parameters. These parameters include but are not limited to: working temperature, metal composition, strain, strain rate, and type of mechanical testing. However, RX is also governed by the movement of dislocations within the microstructure of a metal.

A dislocation is the most common type of line defect that exists within a crystal structure. There are two main types of dislocations: Screw and Edge dislocations, Figure 2 [8]. Since both types of dislocations move in a direction relative to the slip vector, it is necessary to define a slip vector. A slip vector is defined as the displacement distance for atoms around the dislocation, also known as the Burgers vector; this vector is at right angles to the edge of the dislocation line. An edge dislocation is defined as a linear defect at the edge of an extra crystal plane, where the slip vector is perpendicular to the defect line [5]. Zones of compression and tension accompany an edge dislocation so that there is a net increase in energy along the dislocation. A screw dislocation is defined as a linear defect with a slip vector parallel to the defect line. Screw dislocations are usually associated with shear stresses, thus extra energy is involved with screw dislocations than

with edge dislocations. Figure 3 provided by McQueen and Tegart, provides a visual example of the motion of an edge and screw dislocation [2].

Fig. 2. Visual representation of edge and screw dislocations. An edge dislocation occurs at the edge or extra row of atoms; the slip vector is the resulting displacement. A screw dislocation is a spiral ramp with an imperfection line down its axis.

Fig 3. A schematic of the dislocation movement within a crystallite relative to the “Slip Plane”

As seen above in the image, dislocations can either move in parallel or perpendicular to the shear vector. When deformation occurs within a crystal structure, half of the crystal shears along the other half; this produces slip. A slip plane occurs in parallel with the shear vector. An edge dislocation can glide along the slip plane or it can climb perpendicularly to a parallel slip plane if enough vacancies are present to support the climbing process. A screw dislocation can move along any slip plane that contains its Burgers vector and line vector and can glide from one plane to another. The concept of dislocation annihilation comes from dislocations that produce opposing shears that cancel each other out; annihilating each other.

When a metal is deformed, dislocations will array themselves in a manner that is characteristic of the metal being worked. Dislocations will array themselves in patterns

that range between two extremes. In one extreme, dislocations are arranged in planes; dislocations pile up behind one another on their slip planes. Within the crystal structure there are numerous slip planes that intersect each other causing knots and kinks. Increasing the strain causes the dislocations to be more closely packed and thus causing them to interact more strongly. At the other extreme, dislocations are tangled in dense wall-like structures. The walls divide the crystal structure into cells, and within each cell there exists a network of dislocations. The cellular structure originates when dislocation interactions drive and enable dislocation cross-gliding [2].

The formation of the different types of grain boundaries stems from the concept of misorientation. Misorientation is defined as the angle between two crystallites in a polycrystalline structure [5]. Misorientation is increased as dislocations are attracted into arrays of similar dislocations. Continual increase of misorientation, especially in regions of highly dense tangles, causes sub-boundaries to become grain boundaries capable of migration. The surrounding region of the grain boundary is a nucleation site for RX. Figure 4 provides a visual representation of the nucleation of RX [2].

Fig. 4. A visual representation of the different means of nucleus formation for RX. Images progress from left to right.

A nucleus for RX can form in a number of ways. One way is a result of recovery (Figure 4 - top row). The dislocations between two cells (A and B) are attracted to the surrounding sub-boundaries. While some of the dislocations annihilate each other, most of the dislocations enter a neighboring sub-boundary and thereby increasing the degree of misorientation. With further recovery, neighboring cells (A, B, and C) coalesce into a single crystal structure surrounded by a boundary of high misorientation. When the newly form boundary migrates, it is a nucleus for RX. A second means of nucleus formation is in a metal where deformation creates a high density of dislocations in a small region (Panel b). The dislocations combine at the beginning of annealing into a boundary of high

misorientation. Once again, as the boundary migrates it becomes a nucleus for RX. A third means of nucleus formation occurs when an original boundary bulges out (Panel c). When there is a large cell one side of a grain boundary with its ends anchored at neighboring grain boundaries and the density of dislocations is higher on “cell-side” of the grain boundary, the original grain boundary will bulge in that direction [2].

Static Recrystallization (SRX)

The “static” component from SRX derives from the fact that the metal is at rest as RX occurs. Static RX occurs after deformation has been performed on the metal. This form of RX can be advantageous if the metal is going to be subjected to further deformation processes because SRX restores ductility and decreases the flow stress within the metal by decreasing the dislocation density. In SRX, a grain boundary will migrate into the deformed metal and leave behind a new crystal that lacks dislocations. Continuous deformation creates a dislocation substructure within the nucleus [2]. In a paper analyzing the RX characteristics of a 9SmN28 free cutting steel, Karjalainen was able to observe through stress relaxation experiments that SRX was strongly dependent on applied strain of the metal, dependent on the Austenite grain size, and that there was a strong correlation between temperature and SRX kinetics [9]. Karjalainen, in a separate work involving the RX behavior of Ti-Microalloyed Steels, reaffirms the prior results of SRX while experimentally showing that the SRX rate, the speed at which SRX occurs, is influenced, to a small extent the strain history of the specimen in question during the transient stage [10]. The SRX rate is also strongly connected to temperature [2]. Karjalainen was also able to show that the SRX rate is directly correlated to the average

strain, but is ultimately governed by the final strain exhibited by the material. Static RX has been commonly considered to be the dominant phenomenon that removes deformation from a material [11]. When comparing the grain sizes of static recrystallization to dynamic recrystallization, the grain sizes resulting from SRX are larger than those formed from DRX. This is because the nuclei that would be created by subjecting the metal to further increments of strain in DRX do not happen at SRX [2].

Dynamic Recrystallization (DRX)

Dynamic RX owes its name to the fact that RX occurs simultaneously with metal sample deformation. Limiting the cross-glide and climb of dislocations causes an insufficient amount of annihilation and a build-up dislocation tangles during the cooling process. The lack of dislocation annihilation leads to the formation of grain boundaries. This formation of grain boundaries nucleates DRX [2]. In order for DRX to occur, the strain of the material must reach a specific critical value that will initiate the DRX process [11]. The initiation of DRX for most steels occurs at between 50% and 80% of the strain to the peak in the stress-strain curve [12]. This critical strain occurs when the dislocation density reaches a critical level and replaces the elongated grains with new ones that in turn are deformed [13]. In a paper involving the characteristics of SRX and MDRX of hot-deformed Austenite, Karjalainen was able to experimentally display the three characteristic stages typical of the conditions where DRX takes place: the initial work hardening of the material, followed by the softening of the material, and the final stage of steady state [11]. When looking at flow stress graphs, DRX is characterized by the fact that it gives rise to a peak and softening to a plateau [13].

Metadynamic Recrystallization (MDRX)

As mentioned earlier, metadynamic RX is defined as the post-dynamic softening that occurs in a material after being deformed. As opposed to SRX rate, the MDRX rate of a material is greatly affected by the deformation rate of a material and while the strain history of a material does not possess much of an influence on MDRX rate, the final strain rate is a determining factor of the MDRX rate [10]. A number of MDRX characteristics are opposite to that of SRX. During testing of hot-deformed Austenite via stress relaxation method, Karjalainen was able to produce results that showcase a few of the characteristics of MDRX: MDRX is independent of strain, strongly connected to strain rate, and that MDRX is weakly dependent on the temperature of the material [7].

Deformation processes

Recrystallization of a material does not occur without deformation. In metallurgy, there exist two types of metal working processes that are dependent on the RX temperature. While the distinction between hot-working and cold-working processes can be defined by the temperature used in each process, the relationship of the processing temperature to the RX temperature is a more accurate description of the difference between the two processes. Hot-working involves shaping processes that occur above the RX temperature, while cold-working refers to the shaping processes that occur below the RX temperature. A material below its RX temperature is harder and less ductile, which in turn requires more deformation during processing. More deformation means more power and energy must be input into the material, however increased input runs the risk of cracking the material. Above the RX temperature, a material remains soft, relatively

ductile and will even anneal itself upon being subjected to mechanical working processing [14].

Hot Working

A second definition of hot working is based on the homologous temperature of a material. Homologous temperature is defined as the temperature of a material as a fraction of its melting point using the absolute temperature scale. McQueen et. al. define a hot working process when it occurs at a temperature greater than half of the material's melting temperature ($>0.5T_M$). The importance of hot working to the metal industry is undeniable, hot working is an integral part of more than 85% of all metal products [15]. Hot working is generally used in order to lower the flow stress and improve the ductility of a material, and the main objective of hot working is to produce large changes in the dimensions of the base material. The basic method of hot working involves three stages: preheating of the material, deformation via rolling, extrusion, or forging, and finally cooling. The multistage passing method, where the material is subjected to many stages of deformation without reheating, has also been used as another form of hot working. Each of the aforementioned hot working methods possesses a common structural aim: a more uniform and refined grain size through SRX between stages. This causes a continuous improvement in ductility so that the final pass will have a reduction much larger than would have been possible given the initial structure and size. Figure 5 shows the effect of hot working on the substructure of a metal [2].

Fig. 5. The progression of the substructure changes that occur during hot-working. Images progress from left to right

As shown above, hot-working begins with the original and unstrained substructure (Panel a). By increasing the strain the subgrains increase in size, eventually leading to the formation of sub-boundaries (Panels b-d). The sub-boundaries created appear more regularly in hot-working than in the deformed substructure of cold-working.

Hot working involves low yield stress and a low work hardening rate. The maximum flow stress in hot working increases as strain rate increases, and decreases as temperature T increases. The ductility of metal subject in hot working is very high, but is controlled by the range of T and $\dot{\epsilon}$; usually, the ductility will decrease with a decrease in T and an increase in $\dot{\epsilon}$ [2].

Regardless of the hot working process, the deformation is guaranteed to significantly change the microstructure of the material. The deformation of the microstructure is driven by the deformation of the grains in their entirety. This grain deformation is governed by dislocation movement in the microstructure, which as stated earlier can move via glide, cross glide, and climb within the material. During the hot working of a material, strain accumulation in the microstructure can lead to strain hardening of the material. To better understand what happens to a materials microstructure during hot working, there are few key concepts to be discussed.

Recovery (RV)

Recovery (RV) is defined as relaxation process with the following characteristics [6]:

- Recovery involves no incubation period
- The recovery rate at the beginning of recovery is large, but decreases till it is near zero
- Recovery has a limiting value that varies with temperature, the higher the temperature, the higher the limiting value and the shorter the time needed to reach the limiting value
- The greater the amount of deformation, the larger the initial recovery rate.

Decreasing the grain size helps accelerate the recovery process

A secondary definition of recovery is: “all the microstructural changes ‘which do not involve the sweeping of deformed structure by a migrating high-angle boundary.’” [16] In the recovery region, low-angle grain boundaries and point imperfections are absorbed by dislocations while polygonization occurs causing an alteration in the distribution and density of line defects [16, 17]. Because the grain boundaries remain fixed in this region, RX does not occur, and the metal still becomes softer [2, 17]. During recovery, deformed grains lower their stored energy by removing or rearranging the defects within their crystal structure. This reduction in dislocation density leads to a decrease in material yield strength and an increase in material ductility [13].

The rate of recovery is dependent on temperature. At low temperatures, RV is driven by dislocation translation. At medium temperatures, RV is caused by dislocation

movement and redistribution [6]. At a high enough temperature, roughly $> 0.3T_M$, the atoms within the crystal structure have enough vibrational energy to enable rapid vacancy migrations, thus the mechanism of recovery is the climb of edge dislocations. Also, at the same temperature, the screw segments of the dislocation are able to cross-glide to accompany the edge segments of the dislocations. The increased mobility of the dislocations is driven by the attractive forces between other dislocations and allows them to annihilate one another. This annihilation causes the dislocations to arrange themselves into arrays of similar dislocations with low energy. As dislocations form arrays of edges and screws, they increase the misorientation of the surrounding regions. This small increase in misorientation leads to the formation of subgrains, and arrays of subgrains are known as sub-boundaries. For metals with cell structures, formed via densely tangles of dislocations, the recovery mechanisms of climb and annihilation rearrange the tangles into neatly formed arrays and the cell structure transforms into sub grains [2]. Figure 6 shows the process of recovery during annealing [2].

Fig. 6. The stages of the changes in substructure that occur because of recovery during annealing. Images progress from left to right.

As shown in the images above, recovery begins with the thick-walled cell structures (Panel a). The process of annealing causes dislocations to annihilate each other,

thus thinning out the sub-boundaries (Panel b). Further annealing causes the cells to grow (Panel c), and after an extended period of time, the subgrains enlarge (Panel d). The process of annealing restores the ductility of the worked metal [2].

Dynamic Recovery (DRV)

One of the problems of recovery is that during further deformation strain accumulates within the crystal structure and causes strain hardening. When dislocation density is low within the subgrain structure, only dynamic recovery (DRV) will occur. Dynamic recovery, recovery that takes place during deformation, helps alleviate the problem of strain hardening by allowing the dislocations to rearrange themselves in the subgrains of the microstructure into low-energy arrays. The flow curve in Figure 7 shows the four stages of flow stress that occur due to hot forming [1]. The steady-state regime is where the curve begins to level out and the stress and strain are at steady-state. This steady-state regime is a characteristic feature of hot working deformation and provides evidence of a constant dislocation density of a balance of strain hardening and DRV mechanisms. The rate of DRV during the steady-state regime is equal to the rate of hardening [13]. Also, the reduction in flow stress exhibited by the metal is evidence of DRX [6].

Fig. 7. A typical flow stress curve at an elevated temperature.

As seen in the image above, there are four stages of a flow stress curve: stage I the work hardening stage, stage II is the transition stage, stage III is the softening stage, and stage IV is the steady stage.

During DRV the grains elongate leading to the creation of increased GB area GB serrations while subgrains remain equiaxed. The subgrain position reveals that the walls repeatedly migrate, disintegrate, and reform as strain proceeds. Optical microscopy has historically revealed that the subgrains become larger, more equiaxed, more sharply defined as temperature is increased. At a high enough level of DRV, sufficient DRV dislocation build up is prevented enough to inhibit DRX nucleation [13].

Cold-working

Just like hot working, cold working is governed by the homologous temperature of the material. Cold working operates at a homologous temperature less than 0.4 ($<0.4T_M$). However, cold working is applied to all industrial metals at roughly 20°C

regardless of T_M . At 20° , grain boundaries act as effective barriers to dislocation movement [16]. While the concept of deformation is common between both processes, cold working deformation occurs with or without the presence of annealing – a process that involves a controlled and extended heating and cooling procedure of a material. Cold working is a common starting point for examining the plastic behavior of a material because of its simplicity, extensive amount of background knowledge behind cold working, and the minimal dislocation movement. The deformation in cold working begins through interaction of in-grown dislocations; their increased density triggers strain hardening [15]. Figure 8 showcases the substructure behavior during cold deformation [2]. Table 1, in Appendix A, shows an overview comparison between hot-working and cold-working [15].

Fig. 8. The progression of the change in substructure during cold working. Images progress from left to right.

Prior to cold-working, the dislocations are arranged in a simple network (Panel a). Applying a light strain to the metal causes large cells to form (Panel b). Increasing the strain causes the cells to reach their smallest size possible (Panel c). Further strain and deformation of the metal induces an increase in cell wall thickness (Panel d). This

progression proves that the metal's substructure develops and progresses due to an increase in strain.

Compression Testing

Tension, torsion and shear, and compression are three traditional forms of mechanical processing. When combined with heat, these testing methods are extremely efficient at creating metal deformation. Compression testing is essentially the reverse of tensile testing; the material is squashed as opposed to being pulled apart. Figure 9 displays how a material behaves while being subjected to a compression test [3]. During compression testing, the material experiences a reversible elastic region, followed by a plastic region. The continued compression in the plastic region gives way to a permanently deformed, and shorter, cylinder. Reloading of the material will prove that its yield stress is greater than that of the original starting form [18]. Hot compression has been found to be suitable for hot-working because the stress system produced is similar to those found in deformation processing without the arise of instabilities, such as necking. The area increase of the specimen is inversely proportional to the decrease in height, thus the force for deformation increases rapidly and requires strong testing machinery [19]. Despite the conceptual simplicity of compression testing, the friction between the specimen and the testing machine faces (platens). In axisymmetric compression the friction between the platens and the specimen leads to greater deformation in the midsection than in the constrained ends. The friction at the platens thus results in a barrel-shaped specimen upon unloading of the specimen. The friction can be reduced significantly by the use of lubricants. The constraints caused by friction increase the flow

stress of the specimen. However, this effect reduces as the diameter decreases relative to the height. The real flow stress can be found by taking the flow stresses of a number of specimens of different diameters and extrapolating to a diameter of zero. However, when the height of the specimen exceeds the diameter by a factor of two, the problem shifts from barreling to collapse by buckling [18, 19]. For the current study, all hot-compression tests were performed using a Gleeble 1500 thermo-mechanical simulator.

Fig. 9. Graphical and visual representation of the metal behavior during compression testing. Testing images progresses from left to right.

Stress Relaxation and Creep

In the field of materials science and engineering, a material that possesses viscoelasticity is a material which displays a combination of viscous flow and elastic

behavior upon undergoing deformation. The viscosity of a material is characterized as the ability of the material to resist shear flow while exhibiting a linear strain behavior under an applied stress. An elastic material will deform when subjected to an applied load and almost instantaneously return back to its original form. There are two main types of viscoelastic behavior that will be described in detail: creep and stress relaxation.

Creep

Creep is defined as a slow plastic deformation caused by a fixed stress at an elevated homologous temperature, roughly around $>0.5T_M$. [20]. More specifically, creep is classically associated with time-dependent plastic behavior under a constant stress at an elevated temperature [16]. The reason creep can only occur at an elevated homologous temperature is because atoms move more readily and the immediate elastic deformation is higher [20]. The strain during creep is measured as a function of time. There are three stages of creep, as seen in Figure 10 [16]. Primary creep, or Stage I, occurs when the creep rate – plastic strain rate – changes with an increase in plastic strain, time, or temperature and stress. In stage I, the metal will make relatively rapid plastic adjustments at points of stress along grain boundaries, and applying a constant strain rate to the metal causes it to harden and increase its flow stress. Eventually, the initial plastic adjustments give way to a strain rate decrease or stress increase that is constant over a range of strain. This secondary stage, Stage II, is known as steady-state creep. Here, the strain rate decreases or stress increases to a constant value over a range of strain that leads to cavitation and/or cracking that decreases the flow stress and area of the metal. The area-reduction at a constant load will accelerate the strain rate until rupture occurs. This gives

way to tertiary creep, or Stage III. In stage III creep, further increases of the strain rate or decreases the flow stress can eventually induce cracking and/or cavitation [16, 20]. The deformations at higher strain rates, characteristic of creep, result in a more rapid generation of dislocations. At the steady state balance, the subgrain size is smaller and the flow stress is higher. The substructure change due to creep are very similar, as seen in Figure5, but larger [2].

Fig.10. Graphical representation of the three stages of creep at a constant σ (Panel a) and $\dot{\epsilon}$ (Panel b)

Stress Relaxation

Stress relaxation is defined as the decay of stress at constant strain by atomic molecular rearrangement [5]. In a viscoelastic material, the stress decreases under constant strain because the molecules are able to gradually flow by one another. Since molecular movement is a driving force for stress relaxation, it is no surprise that temperature has an effect on stress relaxation [21]. In Figure 11, are the results of a stress relaxation test obtained by Karjalainen using hot-deformed austenite [22].

Fig. 11. Results from a stress relaxation test using hot-deformed austenite

As seen in the figure above, the values of 0.15 and 0.5 are the strain values determined experimentally by Karjalainen. These values are below and above the critical strain necessary for the nucleation of DRX in a low carbon steel; the 0.15 curve represents the SRX behavior and the 0.50 curve represents the post-dynamic softening [22]. Another thing to note from the image above is that there are three stages of stress relaxation. The first stage, a linear region in the log-time scale, is caused by the recovery in work-hardened austenite. The second stage, as is indicated by a fast decrease in stress level driven by the RX in austenite. The final stage owes its linear behavior to the recovery in softened austenite [7, 11, 22-24]. From the results of a stress relaxation test, a degree of RX as a function of time can be produced using the mixture rule and by fitting the function with the Avrami equation. Figure 12 is an example of the degree of RX graph of hot-deformed austenite [22].

Fig. 12. RX fraction vs. time graph fitted with the Avrami equation

Gleeble 1500

For this current work, all hot-working compression testing is performed on the Gleeble 1500 thermo-mechanical simulator. The Gleeble 1500 is a thermo-mechanical testing device that has been utilized to solve a multitude of material problems. Such problems include, but are not limited to, formability, studies of hot ductility, mechanical properties of the weld HAZ, continuous cooling transformation diagram, stress relief cracking, precipitation kinetics, basic weldability studies, strain age cracking, constitutional liquation, liquid metal embrittlement, and solidification of austenitic steel

weldments. The versatility of the Gleeble 1500 is attributed to the dedicated minicomputer that can be programmed to simulate almost any thermo-mechanical process. The minicomputer also allows for closed loop control of both the thermal and mechanical operations, and a variety of other experimental parameters. These include, but are not limited to, heating rate, quenching rate, stroke, stroke rate, force, temperature, chamber pressure, strain, and strain rate.

Heating of the samples used in the Gleeble is obtained via flow of low-frequency alternating current through the sample. Heating rates of the Gleeble can be as high as 10,000⁰C (18,032⁰F) per second. A fine wire thermocouple is percussion welded at approximately the mid-length of the surface of the sample. The sample closes the circuit between the thermocouples, essentially forming an electrical junction; this helps create a more reproducible hot junction. The thermocouple attachment allows for the feedback signal that makes closed loop control of the thermo-mechanical process possible.

The samples, which can be treated as a uniform cylindrical bar with internal generation of heat, are placed into a controlled-atmosphere chamber, Figure 13, which can be evacuated to pressures as low as 10⁻⁷ Torr. Inside this chamber are the jaws of the Gleeble. These jaws are equipped with water-cooled aluminum and copper jaw inserts. Using the copper jaw inserts, which are of the wedge type, allow for closer contact for heat and current at the surface.

Attachment area for
thermocouple wires.

Jaws of the Gleeble 1500. Testing
specimens are carefully placed in between
the jaws.

Fig. 13. A schematic of the jaw lay out of inside the chamber of the Gleeble 1500.

As heat is supplied to the sample via electrical current flowing into it, heat loss occurs through three mechanisms. The first is conduction along the length of the sample; heat is transferred to the water-cooled jaws. The second mechanism occurs via convection in the atmosphere of the chamber; heat is removed and dissipated to the surroundings through mixing. The third mechanism of heat loss is radiation that is emitted by the sample [25].

Industrial Relevance

As mentioned earlier, hot-working of metals is of great importance to the field of metallurgy because of its ability to improve the strength of metals. One of the common forms of hot-working steel is hot-rolling. Hot rolling lends its popularity to the fact that the machinery for hot-rolling is readily available, it produces samples that are to be used for further deformation, and because it produces a higher tonnage of samples than all

other hot-working methods combined. The problem with this method is that it is difficult to maintain the rolling temperature of small samples and it is difficult to provide high strain rates and short rest intervals to the samples at the final stands of the mill. Also, it provides flow stress equations that cannot be directly associated with the dislocation structure in a region of the crystal structure during deformation [15]. To improve the results of hot-rolling methods such as multipass rolling at high temperatures and rapid cooling at the final stage, have been implemented to influence RX events of austenite. Effective use of these new methods requires extensive knowledge of the phenomena that takes place during RX controlled rolling. However, mill work is expensive and time consuming. Thus, establishing a physical model that accurately and quantitatively describes the RX kinetics during deformation is warranted [26]. The use of stress relaxation to simulate the kinetics of SRX, DRX and MDRX is not novel to the industry. In multiple works using various metals Karjalainen , an Emeritus professor at the University of Oulu, was able to successfully apply the method of stress relaxation for establishing quantitative modeling equations for DRX, SRX, and MDRX kinetics [7, 10, 11, 22-24, 26]. Like Karjalainen, the objective of this work will be to demonstrate that hot compression testing on the Gleeble 1500 using 1018 Steel system can supply data for metal working.

CHAPTER 2

Experimental

Axisymmetric hot compression tests were performed using 10 mm x 15 mm 1018 steel cylinders, see Figure 14. Chromel-Alumel thermocouple wires were attached onto the surface of the steel samples at mid-length via percussion welding; Figure 15 shows the exact location of thermocouple attachment. Tantalum squares and graphite circles were attached at the ends of the steel samples. The tantalum squares acted as diffusion blockers while the graphite circles were used for lubrication. The graphite circles were punched out using a “Precision Brand Punch and Die Set.” Figure 16 shows an example of how the graphite circles were obtained. “Silver Goop” was used to provide electrical continuity of the sample and hold the inserts in place. The hot compression tests were performed using a Gleeble 1500 thermo-mechanical simulator. The Gleeble 1500 is commonly used in the metallurgic industry to perform a number of thermo-mechanically controlled processes. Data acquisition was performed using Origin8.1, data was then transferred over to Microsoft Excel to generate the graphs.

REVISIONS			
SYM	DESCRIPTION	DATE	
R.1	MOVED DRAWING FROM PART # R971212C U.D. TOLERANCES & FINISHING NOTES	AGD 7/28/00 AGD 8/21/00

DYNAMIC SYSTEMS INC. (D.S.I.) RESERVES THE RIGHT TO MAKE CHANGES AT ANY TIME IN ORDER TO IMPROVE DESIGN, AND SUPPLY THE BEST PRODUCT POSSIBLE. INFORMATION CONTAINED IN THIS PUBLICATION IS BELIEVED TO BE RELIABLE AND ACCURATE, HOWEVER RESPONSIBILITY IS NOT ASSUMED FOR ITS USE, NOR FOR INFRINGEMENT OF PATENTS OR RIGHTS OF OTHERS THAT MAY RESULT FROM THEIR USE. NO LICENSE IS GRANTED BY IMPLICATION OR OTHERWISE UNDER ANY PATENT OR PATENT RIGHT OF DSI OR OTHERS.

MATERIAL AS SPECIFIED	DRAFTSMAN H.R.	DATE 12/12/97	DSI POESTENKILL N.Y 12140	
	CHECKER		TITLE SPECIMEN - FLOW STRESS 10 mm DIA. X 15 mm LG.	
	ENGINEER			
FINISH BREAK ALL CORNERS 0.25 mm (0.010 in.) TO 0.40 mm (0.015 in.) RAD. 16 ELSEWHERE	UNLESS OTHERWISE SPECIFIED DIMENSIONS ARE IN INCHES TOLERANCES ARE .XX ± 0.13 mm .XXXX ± .XXX ± 0.025 mm (0.001 in.) ANG. ± (0.001 in.)		SIZE B	SMC203 REV. 001
P:\SPECIMEN\SMC203.DWG		SCALE 2/1	SHEET 1 OF 1	

Fig. 14. The engineering drawings for the AISI 1018 steel samples.

Fig. 15. One of the 1018 steel samples after percussion welding. The thermocouple wires are attached at mid length and along the same equatorial plane. This is to ensure a uniform thermal gradient.

Fig. 16. The punch and die set used to create the graphite circle inserts. The size of the graphite circle inserts was 5/8”.

The samples were heated up to 900°C and 1000°C at a heating rate of 8.18°C/s and held at temperature for 200s. Once the holding time was reached, the 1018 steel was compressed at the inputted stroke distances. The total stroke distances used were 5.0 mm and 2.0 mm and produced true strain values of -0.405 and -0.143, respectively. The velocity of the stroke was 0.25 mm/s or 2.5 mm/s thus producing strain rates of 0.01 or 0.1, respectively. The data acquisition was performed at 2000 Hz. Table 2, in Appendix A, provides a list of the controlled design parameters of the experiment.

Constitutive Equations

Equations provide a quantitative means of understanding how the microstructure of a metal behaves upon deformation. For this study on stress relaxation and RX, there are three main equations involved: the Zener–Hollomon parameter, the Avrami equation, and a flow stress equation. One of the most important parameters in hot working is the Zener–Hollomon parameter, Z , also known as the temperature compensated strain rate [12].

Equation 1:

$$Z = \dot{\epsilon} e^{\frac{Q_{app}}{RT}}$$

- $\dot{\epsilon}$ is the strain rate (s^{-1})
- Q_{app} is the apparent activation energy for deformation
- R is the gas constant
- T is the deformation temperature ($^{\circ}K$)

The stress experienced by the microstructure at steady-state stress, where DRX occurs, can be related to Z through various equations. Generally speaking, steady-state stress is an uncommon measurement to obtain, instead, just like the objective of this work, peak stress is recorded. One of the equations used to relate Z to steady state stress is the low stress law.

Equation 2:

$$\sigma = AZ^{\alpha}$$

This equation was developed for creep situations, but has been used in various empirical models for hot working. Other equations such as the high stress law and unified law also relate Z to steady-state stress, but due to the uncommon use and complexity of these laws, the low stress law is the equation used in this work. The use of any of the laws requires the incorporation of strain hardening. In this work, the power hardening law, the simplest model, which is suitable for strains up to 0.5, is incorporated into the low stress law.

Equation 3:

$$\sigma = \sigma_0 \varepsilon^a Z^b$$

By expanding this equation, the flow stress equation is obtained. This equation relates the max or peak stress back to Z .

Equation 4:

$$\sigma = \sigma_0 \varepsilon^a \dot{\varepsilon}^b e^{\frac{bQ_{app}}{RT}}$$

This form of the flow stress equation has gained popularity in online modeling of hot working. There are two main reasons for the wide use of this equation form. First, the model parameters of this equation are easily obtained and controlled. Secondly, the equation becomes linearized by taking the logarithms of both sides.

Equation 5:

$$\ln \sigma = \ln(\sigma_0) + \varepsilon \ln(a) + \dot{\varepsilon} \ln(b) + \left(\frac{bQ_{app}}{R} \times \frac{1}{T}\right)$$

By linearizing the flow stress equation, the unknown variables, a , b , & Q_{app} can be solved using the statistical method of multiple linear regression.

Multiple Linear Regression (MLR)

Upon establishing the constitutive equations necessary to quantify the maximum flow stress during the peak compressive force applied on the 1018 steel samples, the statistical method of Multiple Linear Regression (MLR) was used to provide a solution. A simple linear regression is used to estimate the relationship between a response variable, Y , and a single explanatory variable, X given a data set that includes observations for both variables for a particular population. Conceptually, MLR is similar to a simple linear regression where it is ultimately attempting to analyze the association between quantitative variables; however MLR is used for predicting the values of multiple response variables given a set of, p , explanatory variables ($x_1, x_2, \dots x_p$) [27]. Equation 6, known as the multiple regression model, describes the relationship between the dependent variable and the explanatory variables.

Equation 6:

$$y_i = \beta_0 + \beta_1 x_{1i} + \beta_2 x_{2i} + \dots \beta_p x_{pi} + e_i$$

- β_0 is the constant term or population intercept
- β_1 to β_p are the population slopes of x_1 to x_p ; this means that β_1 to β_p are the coefficients relating the p explanatory variables to the variables of interest
- e_i is the residual of the regression

The Multiple linear regression can be thought of as an extension of simple linear regression, with p explanatory variables. The term “linear” comes from the assumption that Y is directly related to a linear combination to “ p ” explanatory variables.

The relationship between a linear regression and correlation still applies when dealing with MLR; this analysis does not allow for causal inferences. Multiple linear regression does allow us to analyze how a set of explanatory variables is associated with the dependent variable in question.

The null hypothesis H_0 is that the slope of the explanatory variable β_p equals 0. This shows that there is not relationship between the explanatory variable and the response variable. The alternative hypothesis, H_A is that the slope of the explanatory variable, β_p does not equals 0. This shows that there exists a relationship between the explanatory variable and the response variable. In other words, the equations for the null hypothesis and alternative hypothesis are as follows:

$$H_0: \beta_p = 0$$

$$H_A: \beta_p \neq 0$$

The hypothesis test can also be performed on each predictor variable j :

$$H_0: \beta_j = 0 \ (j = 1, \dots, p)$$

$$H_A: \beta_j \neq 0$$

There are number of other statistics and tests involved in MLR [27]:

- s^2 : estimated error variance of the model (also s_e^2)

- R^2 : the coefficient of determination; proportion of variation in Y explained by using X as a predictor in the model

$$\circ R^2 = \frac{\text{explained variation}}{\text{total variation}} = \frac{SSR}{SST} = \frac{SST - SSE}{SST} = 1 - \frac{SSE}{SST}$$

- $SSE = \text{Sum of Squared Residuals} = \sum (y_i - \hat{y}_i)^2$, $df = n - 2$
 - The sum of squared deviations between each observed response (y) and the regression model's prediction for that response ($\hat{y}$). Figure 17 is a graphical representation of SSR. The downward sloped line is model's prediction for that particular response, and the dots with straight lines connected to the slope represent the actual data points and how far they deviate from the model's regression [28].

Fig. 17. Graph of sum of squared residuals

- $df = \text{Degrees of Freedom} =$
 $\text{sample size} - \# \text{ of parameters estimated}$
 - $SST = \text{Total Sum of Squares} = \sum (y_i - \bar{y})^2$, $df = n - 1$

- The sum of squared deviations between each observed response(y) and the overall mean of the response variable($\bar{y}$). This represents the overall variation in the response variable. Figure 18 is a graphical representation of SST. The horizontal line represents the overall mean of the response variable, and the dots with straight lines connected to the horizontal line represent the actual data points and how far they deviate from the mean [28].

Fig. 18. Graph of total sum of squares

- SSR = Sum of Squares Regression ; Explained Sum of Squares=

$$\sum (\hat{y}_i - \bar{y})^2, df = 1$$

- The sum of the squared deviations between each explanatory variables predicted response ($\hat{y}$) from the regression and the overall mean response($\bar{y}$). Figure 19 is a graphical representation of SSR. This graph is equivalent to

combining the SST and SSE graph to compare the difference in error between the two models [28].

Fig. 19. Graph of sum of squares regression

R^2 always goes up when a new X variable is added, even though the new X variable is significant. Therefore, when comparing models in MLR it is essential to always use R^2 -adj because it does not automatically increase with the addition of a new X variable. Use R^2 -adj to assess the fit of the regression

- R^2 -adj: R^2 adjusted for the number of predictors in MLR

- $R^2_{adj} = 1 - \frac{MSE}{MST} = 1 - \frac{s^2_{\epsilon}}{s^2_y}$

- $MSE = \frac{SSE}{n-p-1} = s^2_{\epsilon}$

- The variance of the residuals (e). Estimates the unknown variance of the regression errors (ϵ).
- $p = \#$ of explanatory variables.

- $MST = \frac{SST}{n-1} = s_y^2$

- The variance of the response variable Y

R^2 is used ultimately to test to linear relationship between the response variable and its explanatory variables. By using R^2 -adj. during MLR ensures a more accurate test of the relationship between the response variable to the multiple predictor variables.

- Global F-Test: overall significance of the model as a whole. Answers the question of whether or not the regression equation is a better predictor than the mean. This tests whether or not any explanatory variables are linearly associated with the response variable. Thus, the hypothesis test goes as follows [27]:
 - H_0 : None of the explanatory variables are linearly associated with the response variable
 - $H_0: \beta_1 = \beta_2 = \dots \beta_p = 0$
 - H_A : At least one of the explanatory variables is linearly associated with the response variable
 - The F-test Statistic
 - $F = \frac{MSR}{MSE}$
 - MSR : The regression sum of squares divided by the degrees of freedom within the model
 - Rejection criteria
 - Reject model p-value is $< \alpha$

By linearizing the flow stress equation, as mentioned earlier, an MLR in Microsoft Excel was applied to the data and the unknown values for a , b , & Q_{app} were obtained.

CHAPTER 3

Results

Graphical Results

Figure 20 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 5 mm and a velocity of 2.5 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.1.

The maximum compressive force is 3941.1 lbf with a peak stress of 32,374 psi.

Fig. 20. Force as a function of time for 1018 steel at 900°C

Figure 21 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 1000°C with a total stroke of 5 mm and a velocity of 2.5 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.1.

The maximum compressive force is 2561.5 lbf with a peak stress of 20,672 psi.

Fig. 21. Force as a function of time for 1018 steel at 1000°C

Figure 22 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 2 mm and a velocity of 0.25 mm/s; thus producing a true ϵ of -0.143 and a true $\dot{\epsilon}$ of 0.01.

The maximum compressive force is 2101.9 lbf with a peak stress of 17,266 psi.

Fig. 22. Force as a function of time for 1018 steel at 900°C

Figure 23 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 1000°C with a total stroke of 2 mm and a velocity of 0.25 mm/s; thus producing a true ϵ of -0.143 and a true $\dot{\epsilon}$ of 0.01.

The maximum compressive force is 1277.2 lbf with a peak stress of 10,491 psi.

Fig. 23. Force as a function of time for 1018 steel at 1000°C

Despite being performed at different temperatures, these graphs show case the same thing. They reveal both the maximum compressive force and thus the maximum stress experienced by the 1018 Steel during hot-compression test. The maximum stress obtained through hot-compression testing is equivalent to the peak flow stress as seen in Figure 7. Upon establishing the peak flow stress, the linearized form of the Zener-Hollomon parameter can be used to establish the Q_{app} of the deformation process.

See Appendix B for all graphs created during testing.

MLR

After linearizing Equation 4 to obtain Equation 5, an MLR was performed on equation 5 using Microsoft Excel. Below is the Excel output of the MLR. See Table 3 in Appendix B for the input of the MLR.

SUMMARY OUTPUT

<i>Regression Statistics</i>	
Multiple R	0.9909
R^2	0.9819
$R^2 - adj$	0.9683
Standard Error	0.0618
Observations	8

ANOVA					
	<i>df</i>	<i>SS</i>	<i>MS</i>	<i>F</i>	<i>Significance F</i>
Regression	3	0.8300	0.2766	72.34	0.0006
Residual	4	0.0152	0.00382		
Total	7	0.8453			

	<i>Coefficients</i>	<i>Standard Error</i>	<i>t Stat</i>	<i>P-value</i>	<i>Lower 95%</i>	<i>Upper 95%</i>
Intercept	0.6228	0.5425	1.1480	0.3149	-0.8835	2.1293
XVariable 1	0.2764	0.0420	6.5804	0.0027	0.1597	0.3930
XVariable 2	0.1686	0.0189	8.8795	0.0008	0.1159	0.2213
XVariable 3	6360.2	652.98	9.7402	0.0006	4547.2	8173.2

Using equation 5, listed below for as a reference, an MLR was performed to compute the values of the unknown variables a , b , & Q_{app} .

$$\ln \sigma = \ln(\sigma_0) + \varepsilon \ln(a) + \varepsilon \ln(b) + \left(\frac{bQ_{app}}{R} \times \frac{1}{T}\right)$$

From the results, the computed values of a , b , & $\frac{bQ_{app}}{R}$ are 0.276, 0.168, and 6360, respectively.

To verify that these coefficients are sufficient values, the Global F-test and hypothesis tests for each of the individual slopes must be looked at using the output of the MLR.

Global F-Test

The “Significance F” or Global F-test for this MLR is listed below

- $H_0: \beta_a = \beta_b = \beta_{Q_{app}} = 0$
- H_A : At least one of the slopes $\neq 0$
- $\alpha = 0.05$,

According to the “ANOVA” section, the “Significance F” yields a p-value of 0.000610404.

$$p - value = 0.0006 < \alpha$$

With a p-value of 0.0006, we reject the null hypothesis and confirm that at least one of the explanatory variables used in the MLR model is linearly associated with the response variable.

Also, by looking at the “ $R^2 - adj$ ” value of 0.968, it is safe to say that there is a strong linear association between the multiple explanatory variables

Now that it is understood that at least one of the coefficients of the explanatory variables is linearly associated to the response variable, individual hypothesis tests are performed to each of the slopes to see which one is linearly associated to the response variable.

Hypothesis Test for “X Variable 1”

Hypothesis test for “X Variable 1”:

- $H_0: \beta_a = 0$
- $H_0: \beta_a \neq 0$
- $\alpha = 0.05$

Using the “P-value” for “X Variable 1” we obtain a value of 0.00276.

$$0.002 < \alpha$$

With a “P-value” less than α , the null hypothesis is rejected and it is confirmed that “X Variable 1” possesses a relationship to the response variable.

Hypothesis Test for “X Variable 2”

Hypothesis test for “X Variable 2”:

- $H_0: \beta_a = 0$
- $H_0: \beta_a \neq 0$
- $\alpha = 0.05$

Using the “P-value” for “X Variable 2” we obtain a value of 0.00089.

$$0.00089 < \alpha$$

With a “P-value” less than α , the null hypothesis is rejected and it is confirmed that “X Variable 2” possesses a relationship to the response variable.

Hypothesis Test for “X Variable 3”

Hypothesis test for “X Variable 3”:

- $H_0: \beta_a = 0$
- $H_0: \beta_a \neq 0$
- $\alpha = 0.05$

Using the “P-value” for “X Variable 3” we obtain a value of 0.00062.

$$0.00062 < \alpha$$

With a “P-value” less than α , the null hypothesis is rejected and it is confirmed that “X Variable 3” possesses a relationship to the response variable.

CHAPTER 4

Discussion

Graph Behavior

Given the working temperatures of 900°C and 1000°C and the low carbon content of 0.18% carbon, the 1018 Steel samples are well in the austenite range, Figure 1.

Karjalainen has been able to establish constitutive models that quantitatively describe the RX behavior of the microstructure of a variety of steels using temperatures above 900°C because this is when austenite RX occurs.

Figure 24 is an inverse of Figure 20; the absolute values of the compressive forces were plotted against the testing time. This graph is characteristic of a typical flow stress graph as mentioned earlier, Figure 7. The linear climb is representative of Stage I of flow stress curve. The work hardening rate plays a stronger effect on the material microstructure than the dynamic recovery rate. As the compressive force is continuously applied to the specimen, the stress rises linearly but slows down as the peak is approached and the transition stage or Stage II is reached. In stage II, softening is induced by the mechanisms of DRV and DRX. Stage II occurs very rapidly and quickly transitions into Stage III of the flow curve. Here, the mechanisms at play are the decreasing work hardening rate and the increased softening rate due to induced by DRV and DRX. As the flow stress curve begins to level out, it enters Stage IV, the steady-state stage. Here, a balance between the hardening and softening rate is obtained and DRX is the main phenomena present. These four stages occur in each of the 8 “Force Vs. Time”

graphs. As mentioned earlier, data acquisition occurred at 2000 Hz; the curve is not a line graph, but a scatter plot.

Fig. 24. The inverse of the Force Vs. Time graph.

Despite the commonalities that exist between the 8 graphs, there are also some differences to be noticed. These differences are caused by the design parameters implemented for each of the tests. One important difference is the effect of working temperature on maximum compressive force; the maximum compressive force is higher at 900°C than at 1000°C. This is expected because at 900°C the steel is less ductile, and applying the same deformation distance, or Total Stroke Length, at the lower working temperature requires more force. Along with possessing a greater maximum force, the lines at 900°C are generally steeper than those at 1000°C because a greater force at the same strain rate will result in a steeper line.

Another design parameter that affected maximum compressive force was the true strain. The true strain was calculated by the following equation:

Equation 7:

$$True \varepsilon = \ln \frac{L_o}{L_f}$$

- L_o is the original length of the specimen
- L_f is the final length of the specimen upon deformation

Two true strain were used, -0.405 and -0.143. The steel specimens subjected to the lower strain of -0.143 produced significantly smaller maximum compressive forces; $F_{avg} = 1923$ for $True \varepsilon = -0.143$ and $F_{avg} = 2826$ for $True \varepsilon = -0.405$. As the equation above shows, $True \varepsilon$ is solely dependent on the length change experienced by the specimen. Therefore, by requiring the steel to deform at a smaller distance will require less compressive force to be exerted by the Gleeble 1500.

True $\dot{\varepsilon}$ played a more significant role in the shape of the curve than the compressive force. True $\dot{\varepsilon}$ was calculated by the following equation:

Equation 8:

$$True \dot{\varepsilon} = \frac{v(t)}{2} \left(\ln \frac{L_o}{L_f} \right) \times \frac{1}{dh}$$

- $v(t)$ is the velocity of the jaws of the Gleeble 1500
- d is the diameter of the sample

- h is the height of the sample

The two *True* $\dot{\epsilon}$, values were 0.01, and 0.1 and the higher the value, the sharper and steeper the curve. This increased steepness of the curve is induced by the increase in velocity; the jaws move faster and cause deformation over a smaller amount of time.

MLR Results

The MLR performed on equation 5 established the values of a , b , & Q_{app} , where a is the strain exponent, b is the strain rate exponent, and Q_{app} is the apparent activation energy of the hot-deformation process. From the MLR, the values of a , b , & Q_{app} are 0.28, 0.17, and 6360, respectively. Because a & b are exponents, their results require no transformation. The value for Q_{app} requires an extra calculation; because of the way Equation 5 is set up, we must multiply by R , the Universal Gas Constant of $8.314 \text{ J mol}^{-1} \text{ K}^{-1}$, and divide by b . Thus, the actual value for $Q_{app} = 311.04 \text{ kJ mol}^{-1} \text{ K}^{-1}$.

In works performed by Karjalainen with varying types of steels, he obtained values ranging from 0.10 – 0.13 for the strain rate exponent, b . In other works Karjalainen was able to obtain values of 2.7 – 4 for the strain exponent, a . Lastly, the Q_{app} for hot-working of steel has been experimentally determined to be within the range of 273 – 400 $\text{kJ mol}^{-1} \text{ K}^{-1}$. The difference in values can be caused by a number of factors. For this work, 1018 is a relatively low-carbon steel in the austenite region. Because it is an austenitic steel, it is relatively soft, thus the Q_{app} for hot-compression is expected to be relatively low. Also, the difference in the values lies in the difference in testing methods used. In Karjalainen's works, he uses various tests such as torsion testing, double and multi-pass deformation, single and multi-stage rolling.

The experimentally determined Q_{app} value of $311.04 \text{ kJ mol}^{-1} \text{ K}^{-1}$ can be interpreted as the amount of energy required in the temperature dependent process of hot-compression to cause changes to the microstructure of 1018 steel. The maximum compressive stresses obtained with the different design parameters, is also the maximum flow stress of 1018 steel at the respective design parameters. Thus maximum flow stress represents the amount of stress required to keep the metal flowing and continue the deformation process.

Future Work

The objective of this work was to demonstrate that hot compression testing on the Gleeble 1500 using 1018 Steel system can supply peak stress data for metal working. However, as previously mentioned, the Gleeble 1500 thermo-mechanical simulator has been used in the industry for a number of formability tests. Karjalainen has used the Gleeble to perform stress relaxation on various types of metals and used constitutive equations to provide data for metal working. In addition to peak stress data hot-compression testing can yield stress relaxation testing as seen in Figure 25.

Fig. 25. Force as a function of time, with an extended X-axis to show stress-relaxation location

In this figure, from 132s onward, immediately after peak compressive force, stress relaxation takes place. Further investigation using hot-compression testing with the Gleeble can provide additional data for metal working. In this region of the graph, constitutive models for percent recrystallization %RX, recrystallization time t_{RX} , and grain size/grain growth d_s can be computed.

For %RX models, Avrami-type equations can be used to calculate the percentage of the material that has undergone DRX or SRX respectively. These equations provide a quantitative look into the softening caused by either DRX or SRX kinetics [12].

Equation 9:

$$X_{DRX} = (1 - \exp[-k \left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{ss} - \varepsilon_c} \right)]^n$$

- k is a material constant

- ε is the strain
- ε_c is the critical strain for the start of DRX
- ε_{ss} is the strain where steady-state region is reached
- n is the Avrami-type exponent

Equation 10:

$$X_{sRX} = 1 - \exp(-kt^n)$$

- t is the holding time
- k and n are constants

For the t_{RX} model, it is common to use $t_{50\%}$ because of its ease of calculation. This equation can be derived from the %RX models. This equation provides a quantitative value for when the material will undergo 50% RX [9].

Equation 11:

$$t_{0.5} = A\varepsilon^{-p}\dot{\varepsilon}^{-q}d_0^r \exp\left(\frac{Q_{app}}{RT}\right)$$

- $A, p, q, \text{ and } r$ are experimental constants
- d_0 is the original grain size of the material before deformation
- R is the universal gas constant
- T is the absolute working temperature in $^{\circ}\text{K}$

Furthermore, the RX influence on grain size/grain growth can be established upon further investigation during stress relaxation. This relationship provides a quantitative look at the effects of RX kinetics on the grain size/grain growth of the microstructure [7].

Equation 12:

$$d_{RX} = A\left(\frac{d_0}{\varepsilon}\right)^{0.67}$$

- A is a constant
- d_0 is the original grain size before deformation
- ε is the strain
- The exponent 0.67 is an experimentally determined value specific to the hot-deformed austenite under investigation

These three provide an in-depth and quantitative look at the driving forces and kinetics involved during hot-working and stress relaxation. This work looked at the relationship between maximum compression and flow stress through the transformation of the Zener-Hollomon parameter. With further investigation of hot-compression on 1018 steel combined with the use of the previously mentioned models, it is possible to obtain a greater understanding of changes that occur within the microstructure due to the kinetics of RX.

CHAPTER 5

Conclusions

The hot-compression tests on the Gleeble 1500 thermo-mechanical simulator provided the data for 1018 Steel which shows that:

1. The Gleeble is an effective tool to calculate peak stress in compression
2. Peak stress for a given strain and strain rate decreases with an increase in temperature
3. Peak stress at a given temperature decreases with decreased strain rate
4. An equation for flow stress was developed using the Zener-Hollomon parameter, ϵ , $\dot{\epsilon}$, and Q_{app} . Q_{app} was calculated to be $52.9 \text{ kJ mol}^{-1} \text{ K}^{-1}$
5. Stress relaxation data provided in the test could potentially be used to further investigate other working operations

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APPENDIX A

Tables

Table 1. *Overview comparison of hot-working and cold- working*

Comparison of Hot and Cold Working

Hot Working (HW) $>0.5T_M$ ($^{\circ}\text{K}$)

Cold Working (CW) $<0.4T_M$

1. Objectives

a. Large change in dimensions b. Replace cast structure with more uniform one	Precise final dimensions Produce selected final structure and properties

2. Operations

Preheating—deformation—cooling Multistage: many stages or passes without reheating	Deformation with or without final annealing Multistage usually with occasional intermediate annealing

3. Equipment

a. Large scale because of 1a b. Rough tooling Tooling wear high—rough oxidized surface of high T workpiece c. Lubrication poor (also coolant) Friction high	Strong because of 5a, b Smooth, precise tooling Tooling fatigue life short—high stresses Lubrication good Friction low

4. Starting Stock

Cast material, large dimensions Surface rough, conditioned to cut-out defects, possibly scarfed Microstructure Large columnar grains, segregation inside and between grains, large inclusions, blowholes, weak surfaces	HW or CW + annealed, small HW pickled microrough CW + anneal clean smooth Equiaxed fine grains, homogeneous Directional properties due to fibrous stringers or preferred crystal orientation

5. Forces and Work

a. Yield stress low b. Maximum flow stress σ cd work hardening rate low, increases as $\dot{\epsilon} \uparrow$ decreases as $T \uparrow$, friction high Force: low due to 5a, b Work: up due to large areas 3a up due to 1a, 3b, 6 Energy per Unit Volume For Soft Product Thermal: high Mechanical electrical/E: low For Hard Product Quenching: low If heat treatment: high	High due to 9—increases indefinitely High strain hardening rate almost independent of $\dot{\epsilon} \uparrow$ small decrease with T, friction low High due to 5a, b Down due to small contact areas Down due to 3b, 6 Mechanical: high Thermal: high for annealing Thermal: low no anneal

6. Ductility

High in limited range of T and $\dot{\epsilon}$ (usually $\downarrow$ as $T \downarrow$, $\dot{\epsilon} \uparrow$)	Low unaffected by $\dot{\epsilon}$ ($<10^3 \text{ s}^{-1}$)

Table 1 (Continued). Overview comparison of hot-working and cold- working

	<i>Substructure within Grains</i>
Deformation of grains entirely dependent on dislocation motion	Same
Glide and cross glide and climb	Glide and some cross glide
Dislo. density $\rho_d \sim \epsilon$ at low ϵ saturates $\sim 10^{15}$ m/m ³ due to DRV	$\rho_d \sim \epsilon$ linear at low ϵ ρ_d goes from 10^{11} to 10^{17} m/m ³
$\sigma \sim \sqrt{\rho_d}$ —saturates at σ_s	$\sigma \sim \sqrt{\rho_d}$ —increases to fracture Very little
	<i>Dynamic Recovery, DRV</i>
Dislocations arranged in subgrains	Dislocations form (i) pile ups or (ii) cell structure
	<i>Cell Size d_c</i>
$\sigma \sim 1/d_{sj}$; $d_s \sim 1/\sqrt{\rho_d}$ $d_s \uparrow$ as $T \uparrow \epsilon \downarrow$ Steady state d_s independent of ϵ Static recovery: $\rho_d \downarrow d_s \uparrow$ slowly between stages or during cooling	$\sigma \sim 1/\sqrt{d_c}$, $d_c \sim 1/\sqrt{\rho_d}$ d_c almost independent of T , $\dot{\epsilon}$ d_c decreases as $\epsilon \uparrow$, walls thicken Occurs in Al; α -Fe annealing Stress relieving
	<i>Dynamic Recrystallization, DRX</i>
New soft grains nucleate above $\epsilon_c \sim$ deformed as they grow to re-establish substructure, DRX repeats $D_s \sim d_s$; $D_s \sim (1/\sigma)^{0.8}$	Does not occur Useful increase as T raised, limited by need to avoid oxidation or breakdown of lubricant
T must be less than T_M even with heat of working—incipient melting, burning at GB—hot short High reductions common	 Usually small reductions

Table 1 (Continued). *Overview comparison of hot-working and cold- working*

7. Product Properties	
Homogenization due to heating and working (1b)	Homogeneous from previous HW elongated grains
Fibrous grains, inclusions, directional properties	Preferred crystal orientation strong direction properties
Retain as-worked structure by quenching: $\sigma_y = 25\text{--}50\%$ of CW	Strain hardened (1b)
<i>Slow Cooling</i>	<i>Anneal</i> at $0.4\text{--}0.6T_M$ (optional)
Static recovery: $\rho_d \downarrow$, $d_s \uparrow$	
Static recrystall.: $D_g \downarrow$ as $\epsilon_T \uparrow$, $\dot{\epsilon} \uparrow$, $T \downarrow$	Recrystallization: $D_g \downarrow$ as $\epsilon \uparrow$
Hall-Petch relationship . . .	$\sigma = \sigma_0 + K_y D_g^{1/2}$
Very slow cooling, grain growth	$D_g \uparrow$ as $T \uparrow$ and $D_g^2 \uparrow$ as $t \uparrow$

8. Dimensions—Surface	
Surface oxidized; pickle to clean; scarf to smooth, remove defects	Smooth, unoxidized
Dimensions: fair, however, machining frequently not needed	Good—little machining
Shape control: moderate	Good
Deforms in handling until cooled	Flatten coiled by leveling
Straighten by stretching, leveling	
<i>Springback</i> : slight since $E \downarrow$ as $T \uparrow$	Equals σ/E , especially in bending

Table 1 (Continued). Overview comparison of hot-working and cold- working

9. Internal Structure and Mechanisms

<i>Grains and Macrostructure</i>	
Grain strain = object strain	Grains elongate continuously with ϵ
Soft inclusions elongate into fibres	Mostly hard inclusions, may crack
Some hard inclusions open up pores	Particle matrix decohesion
a. Grain boundaries become serrated	No
b. Dynamic recrystallization—(below)	No
<i>Multistage/Final Cooling</i>	<i>Annealing</i>
c. Static recovery SRV	SRV little utilized industrially
<i>Static Recrystallization</i>	
d. SRX, new soft equiaxed grains	SRX to restore low σ , ductility, and refined grains
Grain refinement, $D \downarrow$ as $\epsilon_{\text{pass}} \uparrow$	
<i>Grain Boundary Sliding</i>	
1–10% of total ϵ ; $\epsilon_{\text{GB}} \uparrow$ as $T \uparrow$ $\dot{\epsilon} \downarrow$	Almost none
GB cracking, pores, fissures	No, only if brittle phase

Table 2. *Design parameters for hot-compression testing*

L_0 mm	D mm	L_{final} mm	T ($^{\circ}C$)	Total Stroke	V ($\frac{mm}{sec}$)	True ε	True $\dot{\varepsilon}$
15	10	10	900	5.0 mm	0.25	-0.405	0.01
15	10	10	1000	5.0 mm	0.25	-0.405	0.01
15	10	10	900	5.0 mm	2.5	-0.405	0.1
15	10	10	1000	5.0 mm	2.5	-0.405	0.1
15	10	13	900	2.0 mm	0.25	-0.143	0.01
15	10	13	1000	2.0 mm	0.25	-0.143	0.01
15	10	13	900	2.0 mm	2.5	-0.143	0.1
15	10	13	1000	2.0mm	2.5	-0.143	0.1

Table 3. *Input for MLR of Equation 5*

σ	$\ln \sigma$	$\ln \varepsilon$	$\ln \dot{\varepsilon}$	$1/T$ ($^{\circ}K$)	ε	$\dot{\varepsilon}$	T
140.1	4.9423	-0.9038	-4.6051	0.0008	-.405	0.01	900
223.2	5.4080	-0.9038	-2.3025	0.0008	-.405	0.1	900
119	4.7791	-1.9449	-4.6051	0.0008	-.143	0.01	900
171.7	5.1460	-1.9449	-2.3025	0.0008	-.143	0.1	900
104	4.6443	-0.9038	-4.6051	0.0007	-.405	0.01	1000
149.1	5.0044	-0.9038	-2.3025	0.0007	-.405	0.1	1000
72.33	4.2812	-1.9449	-4.6051	0.0007	-.143	0.01	1000
103.7	4.6417	-1.9449	-2.3025	0.0007	-.143	0.1	1000

APPENDIX B

Graphs

Figure 26 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 1000°C with a total stroke of 5 mm and a velocity of 2.5 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.1.

The maximum compressive force is 2561.5 lbf with a peak stress of 20,672 psi.

Fig. 26. Force vs. Time for Steelcm2.d01

Figure 27 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 5 mm and a velocity of 2.5 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.1.

The maximum compressive force is 3941.1 lbf with a peak stress of 32,374 psi.

Fig. 27. Force vs. Time for Steelcm2.d02

Figure 28 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 1000°C with a total stroke of 2 mm and a velocity of 0.25 mm/s; thus producing a true ϵ of -0.143 and a true $\dot{\epsilon}$ of 0.01.

The maximum compressive force is 1277.2 lbf with a peak stress of 10,491 psi.

Fig. 28. Force vs. Time for Steelcml.d02

Figure 29 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 2 mm and a velocity of 0.25 mm/s; thus producing a true ϵ of -0.143 and a true $\dot{\epsilon}$ of 0.01.

The maximum compressive force is 2101.9 lbf with a peak stress of 17,266 psi.

Fig. 29. Force vs. Time for Steelcml.d03

Figure 30 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 2 mm and a velocity of 2.5 mm/s; thus producing a true ϵ of -0.143 and a true $\dot{\epsilon}$ of 0.1.

The maximum compressive force is 3014.9 lbf with a peak stress of 24,766 psi.

Fig. 30. Force vs. Time for Steelcml.d04

Figure 31 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 5 mm and a velocity of 0.25 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.01.

The maximum compressive force is 1831.5 lbf with a peak stress of 15,045 psi.

Fig. 31. Force vs. Time for Steelcml.d05

Figure 32 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 900°C with a total stroke of 5 mm and a velocity of 0.25 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.01.

The maximum compressive force is 2473.7 lbf with a peak stress of 20,320 psi.

Fig. 32. Force vs. Time for Steelcm.d05

Figure 33 is the graph of the compressive force (lbf) as a function of time (s). This compression test was performed at 1000°C with a total stroke of 5 mm and a velocity of 2.5 mm/s; thus producing a true ϵ of -0.405 and a true $\dot{\epsilon}$ of 0.1.

The maximum compressive force is 1832.4 lbf with a peak stress of 15,052 psi.

Fig. 33. Force vs. Time for Steelcm.d06

Figure 34 below is the Stress(MPa)-Strain graph of test run “Steelcm2.d01”.

G:\quiksim\dsi\steelcm2.d01

Fig. 34. Stress (MPa) vs. Strain for Steelcm2.d01

Figure 35 below is the Stress(MPa)-Strain graph of test run “Steelcm2.d02”.

G:\quiksim\dsi\steelcm2.d02

Fig. 35. Stress (MPa) vs. Strain for Steelcm2.d02

Figure 36 below is the Stress(MPa)-Strain graph of test run “Steelcml.d02”.

Fig. 36. Stress (MPa) vs. Strain for Steelcml.d02

Figure 37 below is the Stress(MPa)-Strain graph of test run “Steelcml.d03”.

Fig. 37. Stress (MPa) vs. Strain for Steelcml.d03

Figure 38 below is the Stress(MPa)-Strain graph of test run “Steelcml.d04”.

G:\quiksim\dsi\steelcml.d04

Fig. 38. Stress (MPa) vs. Strain for Steelcml.d04

Figure 39 below is the Stress(MPa)-Strain graph of test run “Steelcml.d05”.

Fig. 39. Stress (MPa) vs. Strain for Steelcml.d05

Figure 40 below is the Stress(MPa)-Strain graph of test run “Steelcm.d05”.

G:\quiksim\dsi\steelcom.d05

Fig. 40. Stress (MPa) vs. Strain for Steelcm.d05

Figure 41 below is the Stress(MPa)-Strain graph of test run “Steelcm.d06”.

Fig. 41. Stress (MPa) vs. Strain for Steelcm.d06