THE EFFECTS OF DEFORMATION TEMPERATURE ON THE
FORMABILITY OF AUSTENITIC STAINLESS STEEL SHEET

by

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ABSTRACT

The effects of temperature and alloy stability on the sheet formability of commercial austenitic stainless steels were evaluated. Forming limit diagrams were developed for Types 301, 304, and 305 stainless steel tested at temperatures between 15 and 60 °C using a laboratory punch-stretch forming press. Temperature control was maintained to ± 3 °C with a special heating and cooling system adapted to a limiting dome height test frame. Surface strains were measured on stretch-formed sheets using standard circle grid techniques and the extent of strain-induced martensite was monitored at each measured grid circle with a commercial ferrite scope.

The strain-induced transformation to martensite increased with decreased deformation temperature and was proportional to the imposed effective strain. Small differences in deformation temperature (10 to 20 °C) were sufficient to supply the thermodynamic driving force necessary to modify the $\gamma \rightarrow \alpha'$ transformation. For Type 301, the strain limits increased when martensite formation was suppressed at temperatures above 25 °C. Whereas the higher stability alloys, Types 304 and 305, showed improved formability under test conditions which promoted martensite, i.e. at temperatures below 25 °C. The material properties required for formability were different depending on which deformation mode predominated: drawing, plane strain, or stretching. The stress (or strain) state affects the martensite transformation where more martensite forms in drawing or stretching than in plane strain, for a given effective strain.
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The primary requirement for sheet metal products is an ability to withstand complex forming operations without fracture or the appearance of surface defects. Sheet metal formability will be limited by the amount of strain endured before instability and the distribution of strain. Forming processes may simply involve bending or entail more complicated processes such as spin forming, deep drawing, or press forming. In the most common press forming applications, a flat blank is formed into a finished shape between a matched set of dies. The mechanics of this operation can be separated into three principal deformation modes: drawing, plane strain, and stretching.

Drawing is defined by extension in one direction in the plane of the sheet, accompanied by simultaneous contraction in the other two orthogonal directions, or in terms of principal true strains: \( \varepsilon_1 > 0, \varepsilon_2 < 0, \) and \( \varepsilon_3 < 0. \) An aluminum beverage can is a good example of deep-drawing. Plane strain is defined by elongation in one direction, no change in the perpendicular planar direction, and contraction in the thickness direction; \( \varepsilon_1 > 0, \varepsilon_2 = 0, \) and \( \varepsilon_3 < 0. \) Stretching involves extension in two perpendicular directions within the plane of the sheet and contraction in the thickness dimension; \( \varepsilon_1 > 0, \varepsilon_2 > 0, \) and \( \varepsilon_3 < 0. \) A good example of stretching is the press forming operation for automotive body panels.
Stretching limits can be determined experimentally and represented on a forming limit diagram (FLD), sometimes referred to as a Keeler-Goodwin diagram [1, 2]. This technique involves printing a repeating pattern of small circles on the surface of a test blank prior to deformation. During forming the circles are distorted into ellipses or larger diameter circles (balanced biaxial stretching). Measurements of the major and minor axes of each ellipse are used to calculate the major and minor strains plotted on the FLD. Ellipses are classified as “failed” if they are cracked or egg-shaped (necked) or “safe” if uniformly deformed. This technique was developed by Keeler [3] and is referred to as a circle grid analysis.

A typical FLD for low-carbon steel is shown in Figure 1.1 [4]. In this figure, the forming limit is represented by a Keeler-Goodwin band. This band separates the failure zone from the safe zone by an approximate 10% safety margin. The forming limit may also be represented as a forming limit curve (FLC) separating the highest safe limiting strains from the lowest failed strains. The basic premise of the diagram is that sheet metal parts can be successfully formed if surface strains are kept below the FLC. Above the FLC, stretch formed parts will likely experience plastic instability and failure.

Forming limit curves can be determined by a number of experimental methods. The most common techniques utilize the uniaxial tensile test, flat bottom punch [5], hemispherical dome punch [6], the hydraulic bulge test, or a combination of these methods. Hecker [7] recognized the problems associated with the older Erichsen and Olson cup tests [8], which were: 1) insufficient penetrator size, 2) inadvertent drawing-in
Figure 1.1 A typical FLD for low-carbon steel with the deformation modes for drawing, plane strain, and stretching indicated by distorted circles [4].
through the flange, and 3) inconsistent lubrication. As a result, Hecker introduced a modified method of assessing sheet metal formability. The modified cup test uses a 101.6 mm (4.0 in) hemispherical punch and a circular draw-bead. Sheet specimens are locked in place by the draw-bead and stretched to failure over the hemispherical punch. Ghosh [9] extended this method by using progressively narrower blank widths. This allowed for increasing amounts of lateral draw-in and examination of a larger portion of the FLD. Ghosh measured the dome height at maximum load and the minor strains ($\varepsilon_2$) in the necked or fractured region. The limiting dome height values were normalized with respect to the punch radius, then plotted against the minor strains. This method was termed the limiting dome height (LDH) test.

In this study, the LDH test configuration was used to investigate the formability of Types 301, 304, and 305 austenitic stainless steels. These alloys are considered metastable with respect to a strain-induced transformation [10-12] from fcc austenite to bcc martensite ($\gamma\rightarrow\alpha'$). The kinetics of this phase transformation will affect the formability of 300 series stainless steels depending the alloy chemistry, deformation temperature, and plastic strain [13-15] with lesser contributions from stress state [14, 16-19] and grain size [20-22].

The formability of Types 301, 304 and 305 was investigated by manipulating the $\gamma\rightarrow\alpha'$ transformation. The extent of martensite transformation was modified by controlling the deformation temperature during LDH testing. After experimental work
was completed, the overall formability was assessed by differences in the forming limit curves, volume fractions of martensite, and the distributions of strain.

1.1 Austenitic Stainless Steels

Austenitic stainless steels are frequently chosen for applications which require corrosion resistance, good formability, a range of strength levels, and/or aesthetic appearance [23]. Stainless steels are used extensively in food processing, chemical plants, textile mills, for architectural trim, and other applications where corrosion resistance is required [23]. The “stainless” designation comes with an addition of the 11% or more Cr needed to form a protective chromium oxide surface layer.

Type 304, with 18 wt. % Cr and 8 wt. % Ni, is the most common stainless alloy with applications ranging from passenger rail car siding to kitchen sinks. Type 301 was developed with lower Cr-Ni specifications for applications which require higher work-hardening. Whereas, Type 305 was designed with higher Cr-Ni for deep-drawing applications which require lower flow strength.

As mentioned previously, 300 series austenitic alloys are considered metastable with respect to the transformation from austenite to martensite. This transformation can occur by three different mechanisms relative to the temperature and stress state in which transformation initiates. A schematic representation of the interrelationship between stress and temperature is shown in Figure 1.2 for the \( \gamma \rightarrow \alpha' \) transformation [24]. On cooling to the \( M_s \) temperature, spontaneous transformation occurs at existing nucleation
Figure 1.2 Schematic representation of the interrelationship between stress and temperature for the $\gamma \rightarrow \alpha'$ transformation [24].
sites. Slightly above the $M_s$, stress-assisted transformation takes place at the same sites with the addition of applied stress. At the temperature designated $M_s^\sigma$, the stress required for transformation reaches the yield strength of the austenite. Upon yielding of the austenite, new potent nucleation sites are generated as a result of plastic strain. This stimulates the strain-induced transformation on these new sites at lower stresses than for the stress-assisted nucleation. Above the $M_d$ temperature, no such transformation occurs.

The transformation from austenite to $\alpha'$ martensite during plastic deformation is generally thought to occur by a $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ sequence by way of a transition $\varepsilon$ martensite phase. Using transmission electron microscopy (TEM), Venables [25] and Lagneborg [26] observed the heterogeneous nucleation of $\alpha'$ martensite at the intersection of microscopic shear bands or where the $\varepsilon$ martensite crystals intersect grain boundaries or twin boundaries in the austenite. Mangonon and Thomas [27] confirmed these observations based on evidence that $\varepsilon$ formed independently of $\alpha'$ and that $\varepsilon$ was detected within the $\alpha'$ phase.

1.2 Factors Affecting $\gamma \rightarrow \alpha'$ Transformation

Five factors have been identified which affect the deformation-induced transformation from $\gamma$ austenite to $\alpha'$ martensite: alloy chemistry, deformation temperature, plastic strain, stress (strain) state, and the austenite grain size. These factors
are discussed in terms of their effect on the thermodynamic stability of the austenite, i.e. the stability of the austenite against transformation to martensite.

The effect of alloying elements on austenite stability can be understood by examination of the constitutive $M_{D30}$ equation developed by Angel [13]:

$$M_{D30} (°C) = 413 - 462(C + N) - 9.2(Si) - 8.1(Mn) - 13.7(Cr) - 9.5(Ni) - 18.5(Mo) \ [1.1]$$

where elements are in weight percent. $M_{D30}$ represents the temperature where 0.3 true strain will induce a 50% transformation to martensite. Alloying elements are classified as either ferrite formers or austenite stabilizers based on the tendency of a single element, when added to iron, to enlarge or reduce the austenite phase field (gamma loop) [28]. Carbon, nitrogen, manganese, and nickel are considered austenite stabilizers with respect to delta ferrite formation. Whereas silicon, chromium, and molybdenum are (delta) ferrite formers. Although categorized separately, each element listed in Equation 1.1 increases the stability of the austenite with respect to spontaneous, stress-assisted, or strain-induced transformations [13, 28].

The effects of stress and temperature on deformation-induced martensite have been the subject of many investigations over the past 45 years. Early work from Angel [13] was supplemented by Hecker et al. [14] and modeled by Olsen [24]. Figure 1.3 shows the volume fraction of martensite vs. true strain for Type 304 deformed in uniaxial tension between -188 and 50 °C. For a given temperature, the volume fraction of martensite increased with plastic strain. For a given strain, the volume fraction of
Figure 1.3 Volume fraction of martensite vs. true strain for Type 304 deformed in uniaxial tension in the temperature range from -188 to 50 °C. Solid lines are original data from Angel [13] dashed lines are from Hecker et al. [14], and the dotted extrapolation is from Olson [24].
martensite decreased with an increase in test temperature. Similar results were shown by Peterson [29] for Types 301, 304, and 305 tested in uniaxial tension between -60 and 125 °C. Recent work by Talyan [30] examined the effects of adiabatic heating on the volume fraction of martensite formed in uniaxial tension for three different strain rates. Figure 1.4 presents the effects of both strain and strain rate on the sample temperature and shows that the formation of martensite was suppressed at high strain rates ($10^2 s^{-1}$ and $10^3 s^{-1}$) due to increased adiabatic heating and higher starting temperatures.

The effects of stress (strain) state on the $\gamma \rightarrow \alpha'$ transformation have been examined by several authors. Patel and Cohen [16] found that the transformation was promoted by shear stresses and either promoted or suppressed by normal stresses, depending on the orientation of the applied stress with respect to the volume change. Powell et al. [17] showed that more martensite formed in uniaxial tension than in compression. Using TEM measurements, Murr et al. [31] showed that more shear band intersections were formed in biaxial tension per unit effective strain than in uniaxial tension. From this evidence, they concluded that more martensite would form in balanced biaxial tension than for uniaxial tension. Both Diani and Parks [19] and DeMania [18] showed that more martensite is produced in uniaxial tension than for plane strain, for a given equivalent strain.

The austenite grain size has been shown to affect martensite transformation by various researchers. Nohara et al. [20] noted that the $M_{D30}$ equation developed by Angel [13] should include grain size as well as composition. Results from Nohara et al. showed
Figure 1.4  Changes in specimen temperature and percent martensite with strain for Type 304L tensile tests conducted in room temperature air [30].
that the amount of strain-induced martensite increased with increasing austenite grain size. Raman and Padmanabhan [22] attribute the increase in martensite formation to higher fractions of annealing twins found in larger grains, where twins are recognized as preferred nucleation sites for \( \alpha' \) martensite. Research from Tomimura et al. [21] indicated that spontaneous martensite transformations were significantly affected by grain size. Whereas, the strain-induced transformation were only slightly affected.

In this study, four factors which affect the \( \gamma \rightarrow \alpha' \) transformation are examined, excluding grain size effects. Experimental materials were chosen with a wide range of austenite stability attendant to their alloy compositions. The thermodynamic stability of the austenite was also examined by systematic changes in deformation temperature during stretch-formability testing. In addition, differences in strain and stress (strain state) were investigated with respect to martensite formation as these variables change with variations in dome height and sample width.
2.0 PURPOSE OF STUDY

2.1 Research objectives

The objectives for this study were to investigate the microstructural changes which underlie the forming limits for metastable austenitic stainless steels. These alloys are considered metastable with regard to the strain-induced phase transformation from austenite to martensite during plastic deformation. The extent of transformation depends principally on alloy chemistry, plastic strain, and deformation temperature, with lesser contributions from stress state and austenite grain size. With this in mind, variations in alloy composition and testing temperature were used to control the amount of strain-induced martensite formed during the stretch-formability testing. Afterwards, the overall formability was assessed by differences in forming limit curves, volume fractions of martensite, and strain distributions.

2.2 Industrial Relevance

Past research has shown the formability of austenitic stainless steels to be significantly affected by the stability of the austenite. When commercial drawing and stretching operations are conducted at ambient temperature, the actual temperature experienced by the part may increase above room temperature, owing to the effects of friction and adiabatic heating. Thus, a complete understanding of the effects of
temperature on austenite stability, in conjunction with the known role of alloying elements, would be required to optimize and control formability.

Formability results were represented on forming limit diagrams (FLDs), which have proven useful in diagnosing actual and potential problems in sheet metal forming [40]. During die tryouts or in production, surface strains can be analyzed by a circle grid analysis and compared with the FLDs. Once potential problem areas have been identified, modifications to the tooling, lubrication, or material can be made to insure reproducible stampings.

2.3 Experimental Design

Experimental design concerns the questions of why one experimental material, method, or condition was chosen over another. This section attempts to provide reasonable arguments for the following three questions:

- Why were Types 301, 304, and 305 chosen?
- Why was the limiting dome height test system used?
- Why was formability examined in the temperature range between 15 and 60 °C?

Common commercial grade materials were chosen for their relevance to stainless steel users. In addition, commercial materials should be more consistent than laboratory heats regarding chemistry, texture, and surface quality. Types 301, 304, and 305 were selected based on significant differences in austenite stability. The relative austenite
stabilities of these alloys are ranked by their respective \( M_{D30} \) temperatures calculated from the constitutive equation developed by Nohara [20]:

\[
M_{D30} \ (°C) = 551 - 462(C + N) - 9.2(Si) - 8.1(Mn) - 13.7(Cr) - 29(Ni + Cu) \\
- 18.5(Mo) - 68.0(Nb) - 1.42(\text{ASTM GS } - 8.0)
\]  

[2.1]

The relative austenite stabilities calculated from Equation 2.1 are: 7.4, -29, and -103 °C for Types 301, 304, and 305, respectively. The low stability alloy, Type 301, transforms to martensite, to some extent, with plastic strain at room temperature. Whereas, the high stability alloy, Type 305, is essentially resistant to strain-induced transformation at room temperature. The expected transformation behavior of Type 304 is intermediate to Types 301 and 305.

Stretch-formability testing was conducted on a standard servo-hydraulic load frame equipped with the limiting dome height tooling (LDH). A wide range of strain paths can be examined with the hemispherical dome test, from drawing to balanced biaxial stretching. This is accomplished by varying the sample width (lateral constraint) following the method developed by Ghosh [9].

Lubrication at the punch-specimen interface was provided by thin polyethylene sheets. This lubrication condition was chosen for three reasons: 1) to minimize the friction coefficient (minimize frictional heating), 2) to provide some degree of thermal insulation between the punch and sheet, and 3) to insure a reproducible dry lubrication method. A 0.212 mm/s (0.5 in/min) punch velocity was used in each hemispherical dome
test. This stroke rate was selected because it provided a balance between excessive adiabatic heating at faster rates and the duration of temperature control at slower rates.

Stretch-formability testing was conducted in the temperature range between 15 and 60 °C (59 and 140 °F). This temperature window was sufficient to investigate both the low and the high stability alloys. These alloys were designed for room temperature forming applications and as a result show diminishing marginal returns (low dome heights and lower forming limits) below 15 °C and above 60 °C.
3.0 EXPERIMENTAL METHODS

The first section of this chapter introduces the experimental materials including alloy chemistry, mill processing, and mechanical properties. The second section describes the stretch-forming system, sample preparation, and testing details. The third section presents the design and use of a temperature control system and the final section describes the applicable measurement techniques.

3.1 Sheet Steel Characterization

3.1.1 Steel Chemistry

The experimental materials were supplied by Allegheny Ludlum [32] as commercial grade, cold rolled and annealed sheets. Composition and sheet thickness information are summarized in Table 3.1. The primary differences in chemistry between these three alloys are the chromium, nickel, and carbon contents. Chromium content in excess of 11 weight percent imparts the “stainless” corrosion resistant surface film which identify stainless steels. Nickel and carbon are considered austenite stabilizers. Although chromium is categorized as a ferrite former, all three elements extend the temperature range over which austenite is thermodynamically stable [28].
### Table 3.1 - Composition by Weight Percent and Thickness of Types 301, 304, and 305.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Cr</th>
<th>Ni</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>301</td>
<td>16.82</td>
<td>6.21</td>
<td>0.135</td>
<td>1.81</td>
<td>0.34</td>
<td>0.028</td>
<td>0.0003</td>
<td>0.37</td>
</tr>
<tr>
<td>304</td>
<td>18.26</td>
<td>8.06</td>
<td>0.057</td>
<td>1.84</td>
<td>0.42</td>
<td>0.029</td>
<td>0.0003</td>
<td>0.49</td>
</tr>
<tr>
<td>305</td>
<td>18.65</td>
<td>11.81</td>
<td>0.035</td>
<td>0.85</td>
<td>0.49</td>
<td>0.032</td>
<td>0.0008</td>
<td>0.32</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Cu</th>
<th>Nb</th>
<th>V</th>
<th>N</th>
<th>Co</th>
<th>W</th>
<th>Thickness mm (in)</th>
</tr>
</thead>
<tbody>
<tr>
<td>301</td>
<td>0.46</td>
<td>0.024</td>
<td>0.079</td>
<td>0.066</td>
<td>0.10</td>
<td>0.026</td>
<td>0.508 (0.020)</td>
</tr>
<tr>
<td>304</td>
<td>0.40</td>
<td>0.008</td>
<td>0.076</td>
<td>0.062</td>
<td>0.13</td>
<td>0.018</td>
<td>0.533 (0.021)</td>
</tr>
<tr>
<td>305</td>
<td>0.45</td>
<td>0.010</td>
<td>0.105</td>
<td>0.021</td>
<td>0.073</td>
<td>0.027</td>
<td>0.508 (0.020)</td>
</tr>
</tbody>
</table>

#### 3.1.2 Metallographic Procedures

Materials were prepared for metallography and examined in the thickness, transverse, and longitudinal orientations. The as-received microstructures were revealed with an electrolytic etch using a 10% oxalic acid solution. For the longitudinal orientation, micrographs were obtained by etching for 30 seconds at 5 V and 400 mA, backpolishing on 0.05 alumina, then a final etch for 60 seconds at 4 V and 300 mA. For the thickness or transverse orientation, micrographs were obtained by etching for 30 seconds with 5 V and 150 mA, backpolishing with a 0.05 micron alumina, followed by a final etch for 75 seconds with 4 V and 150 mA. Grain boundaries were revealed with an
electrolytic etch using a 60% nitric acid solution. Satisfactory results were obtained with 40 to 90 seconds, 1 V, and 100 mA.

3.1.3 Mechanical Properties

Mechanical properties were evaluated from tensile bars oriented 0° to the rolling direction. Tensile blanks were sheared from sheet material and machined according to ASTM E-8 specifications [33]. All tensile testing was conducted in ambient air at 25 °C with an engineering strain rate of $1.67 \times 10^{-3} \text{ s}^{-1}$.

3.2 Stretch-Formability Testing

A stretch-forming press was designed and built for the Advanced Steel Processing and Products Research Center (ASPPRC) in 1986 by D.A. Burford in conjunction with his Ph.D. dissertation [34]. For this study, a forming press was equipped with the limiting dome height (LDH) tooling set. An advantage to using the LDH configuration with a hemispherical punch was that failure modes ranging from drawing ($\varepsilon_1 = -2\varepsilon_2$) through plane strain ($\varepsilon_2 = 0$) to balanced biaxial stretching ($\varepsilon_1 = \varepsilon_2$) could be examined with a single set of tooling.

Schematic diagrams of the forming press, mounted on a 445 kN (100 kip) servo-hydraulic load frame, and the LDH tooling are shown in Figures 3.1 and 3.2. The clamping dies illustrated in Figure 3.2 bolt directly into the upper and lower die seats shown in Figure 3.1. Each test panel, positioned between the clamping dies, was held in
place with a 400 kN (90 kip) force applied by four Enerpac hydraulic cylinders. The hemispherical punch mounts onto the lower actuator with a 90 kN (20 kip) load cell. Once the test panel was fixed in place, the sheet was stretched to failure with a punch velocity of 0.212 mm/s (0.5 in/min).

3.2.1 Sample Preparation

Sample widths and geometries were varied to obtain failure modes ranging from drawing to balanced biaxial stretching. Two different sample geometries were used, rectangular and “hour glass”. The hour glass geometry, designed by Cooke [35], was used where rectangular samples had the tendency to fail at a stress riser near the die radius. Rectangular sheet samples were sheared 178 mm (7 in) in the rolling direction in the following widths: 178, 152, 140, and 127 mm (7, 6, 5.5, and 5 in). Hour glass shaped samples were sheared 178 mm in the rolling direction and water jet cut in 76 and 51 mm widths (3 and 2 in). Duplicate samples were tested for each width and geometry.

Five layers of polyethylene sheet were used to lubricate the punch-specimen interface. This lubrication scheme helped to minimize the frictional heating. The lubrication condition should affect the region of the forming limit diagram investigated, i.e. by decreasing the friction coefficient between the punch and sheet, the stretching capability increases [28].
Figure 3.1  Schematic diagram of the stretch-forming press designed and built by Burford [34].
Figure 3.2 Schematic diagram of the limiting dome height (LDH) test configuration.

Clamping Dies with Draw Bead

Stretched Sheet

5.08 cm radius Hemispherical Punch

13.26 cm

10.57 cm
A pattern of four 2.54 mm diameter circles in a 6.35 mm square (0.1 in circles, 0.25 in square) was applied to the surface of each sample by electrochemical marking. This pattern provided a reference grid used to locate specific circles for measurement. Optimal etch contrast was achieved with a Lectroetch V45A power source, 53NC electrolyte, 40 Ohmites, and 40 seconds.

Photographs of a Type 301 LDH sample are shown in Figure 3.3. As shown in the aerial perspective, the 178 x 178 mm (7 x 7 in) flanged region was wrinkled and deformed slightly out of square by application of the circular drawbead. The fracture location is shown on left side of the dome and in the close-up profile view. The numbered locations were designated for surface strain and thickness strain measurements near the fracture. Strain measurements are discussed in Section 3.4.1.

3.3 Temperature Control System Design

For this project, a temperature control system was added to the existing forming press to enable stretch-formability testing at temperatures ranging from 15 to 60 °C. A photograph of the forming press equipped with the temperature control system is shown in Figure 3.4. The MTS console, data acquisition, and thermocouple readouts are pictured on the left with the load frame and forming press shown on the right. In the close-up view of the forming press shown in Figure 3.5, the steel clamping dies (A) are shown with copper heat exchangers (B) and a stretch-formed sheet sample (C). In this photograph, the upper die assembly has been raised so that the test details can be seen.
Figure 3.3 Photographs of a 178 x 178 mm (7 x 7 in) LDH sample: a) an aerial perspective and b) close-up view of the fracture.
Figure 3.4  Photograph of the stretch-forming press mounted on a 445 kN (100 kip) MTS servo-hydraulic test frame. The PC controller and thermocouple acquisition system are shown on the left.

Figure 3.5  Photograph of the LDH test system: (A) steel die platens, (B) copper heat exchangers, and (C) a stretch-formed sheet sample with thermocouples attached.
During a punch-stretch test, the sample contacts the upper and lower clamping dies and the hemispherical punch. It is therefore necessary to control the temperature of these three components in order to approximate an isothermal test condition. The following two sections describe the equipment used to maintain the temperatures of the dies, punch, and test specimen. Once the test details have been explained, the capability of the test system is examined based on calibration experiments.

3.3.1 Low Temperature System

The equipment used for the low temperature experiments is presented in the simplified block diagram shown in Figure 3.6. First, the temperature of the clamping dies was maintained with contacting heat exchangers in series with a recirculating methanol bath. Three to four hours were required to reach thermal equilibrium. Next, the punch was removed, refrigerated, and replaced (multiple punches facilitated efficient testing). Then the sheet sample was locked in place between the clamping dies and a refrigerated air stream was situated directly above the sample. The temperatures of the sample, clamping dies, and punch were monitored and adjusted until an equilibrium test temperature was achieved within ±1 °C (30 minutes to 1 hour). Once an approximate isothermal condition was established, the 5 minute LDH test was conducted.
Figure 3.6 Block diagram of the low temperature control system.

Low Temperature System

Air Compressor → Liquid Nitrogen + Methanol Bath → Shower Head → Sample → Hemispherical Punch → Freezer

Heat Exchangers

Pump

Refrigerated Cooling Coils → Methanol + Water Bath
3.3.2 Elevated Temperature System

The elevated temperature system is shown in Figure 3.7. Analogous with low temperature system described above, the temperature of the clamping dies was maintained with a recirculating oil bath in series with heat exchangers. Prior to each test, the removable punch was preheated in a furnace and replaced. Once the component temperatures were established within ±1 °C, the sample was locked in place and a variable intensity heat lamp was situated directly above the sheet specimen. Again, the temperatures of the sample, clamping dies, and punch were monitored and adjusted to achieve an isothermal condition before initiating the 5 minute LDH test.

3.3.3 Calibration Experiments

Type K thermocouples were spot-welded to the surface of each sample to enable real-time temperature acquisition. As shown in Figure 3.8, thermocouples were arranged in positions designated as pole, mid-point, and free span. The pole corresponds with a position on the sheet which is in contact with the punch for the duration of a test, whereas the mid-point contacts the punch at some intermediate dome height. The position identified as the free span does not contact the punch throughout the duration of the test.

Figure 3.9 shows the temperature measurements acquired during stretch-formability testing of Type 304. The target temperatures were 15, 25, 45, and 60 °C. As shown in the figure, the temperature fluctuation is ±3 °C for sheets deformed at 0.212 mm/s (0.5 in/min). The characteristic shape of the curves is similar for all four
Elevated Temperature System

Figure 3.7 Block diagram of the elevated temperature control system.
Figure 3.8 Aerial view of a rectangular LDH sample with the circular draw bead superimposed. Thermocouples are spot welded to the sheet surface in the locations marked with an X.
Figure 3.9  Temperature measurements acquired during stretch-formability testing of Type 304. Thermocouple locations are designated in Figure 3.8.
temperatures tested. Similar temperature profiles were also observed for Types 301 and 305. At the start of the test, all three thermocouples stabilized within 3 °C. With the initiation of the test, the punch contacts the sheet at the pole. At some intermediate time, the punch contacts the mid-point. As a result, the pole, mid-point, and punch temperatures equalize, increasing slightly throughout the duration of the test due to sliding friction and the heat generated from plastic strain. The free span remains free from contacting the punch. As a consequence, the temperature recorded at the free span was attributed solely to heating produced by plastic deformation.

Once the test systems were calibrated with spot-welded thermocouples, self-adhesive thermocouples [36] were substituted. This provided two advantages: quick, ease of attachment and the elimination of possible crack initiation sites coincident with the spot-welds. The temperature measurements were similar using either the welded thermocouples or self-adhesive thermocouples.

3.4 Measurement Techniques
3.4.1 Strain Measurement

Surface strains were measured with a LECO 2001 image analyzer at approximately 15X. Deformed circles were generally of elliptical shape unless a near biaxial stretch condition existed, in which case, the distorted circle simply had a larger diameter than an undeformed circle. The major axis of each ellipse was oriented radial
with the pole of the dome-shaped sample with the minor axis oriented 90° to the major axis.

Each etched ellipse had an inner and an outer dimension defined by a line thickness. Duplicate measurements were taken from the inside edge across the major (or minor) axis to the outside edge using the contrast between the etched border and the unetched surface. A total of four separate measurements were taken across each axis and averaged to obtain a single value. If any two of the four measurements differed by more than 5% the circle was considered unmeasurable.

Circles were designated either “safe” or “failed”. “Safe” circles were those circles adjacent to the failure site which were intact and deformed uniformly. “Failed” circles were necked (egg-shaped) or cracked through the outside perimeter of the etch, but measurable. Circles which were cracked through the interior perimeter were excluded from measurement.

A statistical analysis was conducted to assess the 95% confidence interval for major and minor surface strain measurements. Six samples were evaluated, representing each sample width and geometry described in Section 3.2.1. The precision interval for 95% probability is shown in Figure 3.10. Each data point represents the mean value of fifteen individual measurements and the error bars correspond to the interval which 95% of the measured values are expected to fall. Variations in the length of each error bar are indicative of the measurement problems associated with circle grid analysis, e.g. inconsistent etch, low contrast, or circle distortion.
Figure 3.10  The precision interval associated with 95% confidence limits for a hypothetical forming limit curve. Error bars represent the interval within which any single strain measurement is expected to fall.
In most cases, sheet thickness was calculated from measured surface strains assuming constant volume. Otherwise, thickness strains were measured with a Panametrics Model 25DL ultrasonic thickness gage equipped with a M208 probe. Thickness values were used to correct for the error associated with ferromagnetic measurements of thin sheets (discussed in Section 3.4.3). Reported values for martensite volume fraction differed by less than 2% using either measured or calculated thickness values.

3.4.2 Martensite Measurement

Martensite content was measured with a Fisher Feritscope MP3C ferrite meter equipped with a EGAB 1.3 probe (1.25 mm probe diameter). The Feritscope measures the volume fraction of ferro-magnetic martensite relative to the non-magnetic austenite. This instrument is calibrated for the magnetic response to ferrite. As a result, each ferrite number (FN) must be converted to a corresponding volume fraction of martensite (VFM). Calibration curves were constructed by Peterson to convert FN to VFM for Types 301 and 304 stainless steel [29]. The curve-fitting equations used to calculate martensite content are reproduced below. Use the corrected FN values discussed in Section 3.4.3 for Equations 3.1 through 3.4.

Type 301, FN ≤ 4.0: \[ VFM = 0.169FN^{0.432} \] [3.1]
\[ VFM = 0.00725(FN + 25) - \frac{8}{FN + 25} - \frac{125}{(FN + 25)^2} \]

Type 301, FN \geq 4.0:

\[ - \frac{50}{(FN + 25)^3} - \frac{10^4}{(FN + 25)^4} + 0.535 \]

Type 304 or 305, FN \leq 1.0:

\[ VFM = 0.188FN^{0.257} \]

Type 304 or 305, FN \geq 1.0:

\[ VFM = 0.011(FN + 15) - \frac{4}{FN + 15} - \frac{8}{(FN + 15)^3} \]

\[ - \frac{4000}{(FN + 15)^4} + 0.32 \]

3.4.3 Thickness Correction for Ferrite Meter Readings

Sheet material thickness influences the measured ferrite number (FN_{meas}) below a thickness of 2 mm (0.08 in), i.e. FN_{meas} values decrease with decreasing sheet thickness. FN_{meas} values were multiplied by a correction factor (CF) to approximate the value representative of “thick” material (\geq 2 mm). FN_{meas} \times CF = FN_{corr} (corrected ferrite number). The thickness correction curve shown in Figure 3.11 was developed from a thickness correction curve in the Fisher manual and data obtained by stacking multiple thin stainless steel sheets of equivalent martensite content.
Figure 3.11 A thickness correction curve used to correct ferrite numbers (FN) measured on sheet material less than 2 mm thick. The variables used in the curve-fitting equation are thickness (T) and correction factor (CF).
4.0 RESULTS AND DISCUSSION

The first section of this chapter presents the microstructural and mechanical properties of the as-received experimental materials. The second section examines the forming limit diagrams constructed for each alloy and forming temperature. In section three, the volume fractions of martensite are correlated with effective strain and strain path. The fourth section presents the forming limit curves, grouped by deformation temperature and grouped by alloy designation. In section five, the differences in strain distribution are examined as a function of deformation temperature and strain hardening. In the final section, the locations and orientations of fracture are discussed for each alloy and test temperature.

4.1 Material Characterization

4.1.1 Light Microscopy

Figures 4.1 through 4.3 show the microstructures of the as-received, cold rolled and annealed austenitic stainless steels. For each alloy, the annealed condition was verified by the presence of straight-sided annealing twins and equiaxed grains in each of the three orientations shown. The micrographs show fully recovered, recrystallized structures with similar grain sizes between the three alloys. Considering the comparable grain size and the removal of prior dislocation structures, the differences in formability between these three alloys were principally attributed to compositional effects.
Figure 4.1 Three-dimensional photomicrograph montage of the as-received, cold rolled and annealed microstructure of Type 301. Light micrograph, 10% oxalic acid.
Figure 4.2 Three-dimensional photomicrograph montage of the as-received, cold rolled and annealed microstructure of Type 304. Light micrograph, 10% oxalic acid.
Figure 4.3 Three-dimensional photomicrograph montage of the as-received, cold rolled and annealed microstructure of Type 305. Light micrograph, 10% oxalic acid.
4.1.2 Grain Size

Table 4.1 lists the ASTM grain size numbers and the mean intercept distances for Types 301, 304, and 305. The grain size was determined from through-thickness sections oriented along the coil edge using the three-circle procedure, ASTM E 112 [37]. As shown in Table 4.1, the grain sizes of all three alloys were similar, although Type 305 had slightly larger grains. Grain size has been shown to affect alloy stability by several authors [20-22] who showed that reducing the grain size increased the austenite stability.

Examination of grain size coefficient in Equation 4.1, developed by Nohara et al. [20], indicates that the relative potency of grain size as an austenite stabilizer is small compared with alloying additions, especially with respect to the interstitial elements, carbon and nitrogen.

\[ M_{D30} (^\circ C) = 551 - 462(C + N) - 9.2(Si) - 8.1(Mn) - 13.7(Cr) - 29(Ni + Cu) \\
- 18.5(Mo) - 68.0(Nb) - 1.42(\text{ASTM GS - 8.0}) \]  

[4.1]
4.1.3 Mechanical Properties

Recent work by Peterson [29] has provided extensive tensile property data for Types 301, 304, and 305 including strain rate and temperature effects. As a result, limited tensile testing was performed only to verify Peterson’s data. Tensile properties were evaluated from longitudinal tensile bars tested in ambient 25 °C air with an engineering strain rate of $1.67 \times 10^{-3} \, \text{s}^{-1}$. Table 4.2 lists the tensile properties for the commercial steels used in this study and those steels tested by Peterson. Although alloy specifications and steel suppliers were identical, the steels used in each study were obtained from separate heats.

Similar results were obtained for Types 301 and 304 tensile bars oriented 0° to the rolling direction in both Peterson’s study and the current program. Values obtained for yield strength and ultimate tensile strength varied within 5%, corresponding with similar differences in uniform and total elongation. Given the similarity between Peterson’s results and those obtained in this study, the mechanical property data collected by Peterson for Types 301 and 304 should be applicable for use in the current investigation.

The mechanical property data obtained for Type 305 tensile bars oriented 0° to the rolling direction differed from the results shown by Peterson. The yield strength and ultimate tensile strength of the current material was lower (27% and 17%, respectively) corresponding with increased uniform and total elongations. Approximately half the difference in yield strength could be expected from a difference in austenite grain size. Peterson’s Type 305 had a smaller grain size of ASTM 10.0 compared with 7.6 for the
Table 4.2 – Tensile Properties for Types 301, 304, and 305 Tested in Ambient 25 °C Air with an Engineering Strain Rate of $1.67 \times 10^{-3} \text{ s}^{-1}$.

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Uniform Elongation (%)</th>
<th>Total Elongation (%)</th>
</tr>
</thead>
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<tr>
<td>301</td>
<td>331</td>
<td>869</td>
<td>73</td>
<td>79</td>
</tr>
<tr>
<td>304</td>
<td>290</td>
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<td>64</td>
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<tr>
<td>305</td>
<td>214</td>
<td>572</td>
<td>55</td>
<td>64</td>
</tr>
</tbody>
</table>

**Tensile Properties of Commercial Steels Tested in Peterson’s Study**

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Uniform Elongation (%)</th>
<th>Total Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>301</td>
<td>313</td>
<td>822</td>
<td>76</td>
<td>84</td>
</tr>
<tr>
<td>304</td>
<td>300</td>
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<td>67</td>
</tr>
<tr>
<td>305</td>
<td>291</td>
<td>634</td>
<td>51</td>
<td>60</td>
</tr>
</tbody>
</table>

**Tensile Properties of Commercial Steels Tested in Peterson’s Study**

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Uniform Elongation (%)</th>
<th>Total Elongation (%)</th>
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<td>280</td>
<td>747</td>
<td>85</td>
<td>95</td>
</tr>
<tr>
<td>304</td>
<td>289</td>
<td>656</td>
<td>64</td>
<td>74</td>
</tr>
<tr>
<td>305</td>
<td>288</td>
<td>618</td>
<td>58</td>
<td>73</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% Offset Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Uniform Elongation (%)</th>
<th>Total Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>301</td>
<td>289</td>
<td>769</td>
<td>81</td>
<td>90</td>
</tr>
<tr>
<td>304</td>
<td>288</td>
<td>618</td>
<td>58</td>
<td>73</td>
</tr>
<tr>
<td>305</td>
<td>292</td>
<td>624</td>
<td>57</td>
<td>67</td>
</tr>
</tbody>
</table>
Type 305 tested in this study. A Hall-Petch analysis showed that 53% of the 77 MPa difference in yield strength could be accounted for by grain size. The difference in yield strength was calculated from the Hall-Petch equation [38],

\[ \sigma_y = \sigma_0 + k_y d^{1/2} \]  

\[ \Delta \sigma_y = k_y (d_1^{1/2} - d_2^{1/2}) \]

where \( d_1 \) and \( d_2 \) represent the mean intercept distances (in millimeters) for the two grain sizes and \( k_y = 1.25 \text{ kg mm}^{-3/2} \) [21]. The remaining 36 MPa could easily be reflected by differences in alloy composition, well within the alloy specification.

In this section, the as-received microstructures were examined, grain sizes were compared, and tensile properties were evaluated for Types 301, 304, and 305. The micrographs show fully annealed structures with similar grain sizes for all three steels. Based on the microstructures observed, differences in mechanical properties and formability should be primarily attributed to differences between alloy chemistries. The tensile properties evaluated in this study showed good correlation with results from Peterson for Types 301 and 304. This validated the use of mechanical property data from Peterson for these two materials. Differences in strength and corresponding elongation were noted for Type 305 between this study and Peterson’s indicating the need for judicious use of Peterson’s results for Type 305.
4.2 Forming Limit Diagrams

Forming limit diagrams (FLDs) are presented in Figures 4.4 through 4.7 for Types 301, 304, and 305 tested at 15, 25, 45, and 60 °C. Sheet specimens were stretched to failure using the limiting dome height (LDH) test configuration and the temperature control system discussed in Sections 3.2 and 3.3. Both safe and failed data points are included in the figures so that the shape of the forming limit curves (FLCs) may be interpreted.

In general, the FLCs shown in Figures 4.4 through 4.7 do not conform to the characteristic V-shaped curve observed for many ferritic sheet steels [39], e.g. Figure 1.1, with the exception of the FLC shown in Figure 4.5c for Type 305 tested at 25 °C. A V-shaped FLC has a minimum value of major strain near the plane strain axis ($\varepsilon_2 = 0$) with higher strain limits for both drawing (left side of the FLD) and stretching (right side). Typically, the slope of the FLC on the stretching side is lower for austenitic stainless steels than for low carbon steels [40].

The low temperature FLDs are shown in Figure 4.4 for each alloy tested at 15 °C. The straight-line FLCs drawn for Types 301 and 304 show reduced biaxial ductility which is indicative of unstable alloys which form extensive martensite with strain (specific volume fractions of martensite are quantified in Section 4.3). In comparison, Type 305 shows a significant increase in stretching formability which is probably the result of strain-induced martensite formation at high positive biaxial strains. In this case, martensite formation is beneficial because the formation occurs in highly strained
Figure 4.4 Forming limit diagrams for Types a) 301, b) 304, and c) 305. Test conditions: 15 °C, 0.212 mm/s punch velocity, and polyethylene lubrication, 80 microns thick.
Figure 4.5  Forming limit diagrams for Types a) 301, b) 304, and c) 305. Test conditions: 25 °C, 0.212 mm/s punch velocity, and polyethylene lubrication, 80 microns thick.
Figure 4.6  Forming limit diagrams for Types a) 301, b) 304, and c) 305. Test conditions: 45 °C, 0.212 mm/s punch velocity, and polyethylene lubrication, 80 microns thick.
Figure 4.7  Forming limit diagrams for Types a) 301, b) 304, and c) 305. Test conditions: 60 °C, 0.212 mm/s punch velocity, and polyethylene lubrication, 80 microns thick.
regions, e.g. in an incipient neck area, strengthening that area locally and additional strain would be forced to spread to adjacent areas which are less strained. Thus, local deformation and necking would be homogenized and failure would be delayed.

Figure 4.5 shows the FLDs for each alloy tested at room temperature, 25 °C. Types 301 and 304 showed improved stretching ability compared with the 15 °C results shown in Figure 4.4. In this case, the excess martensite formed below room temperature was detrimental to the stretching behavior of Types 301 and 304. This can also be seen in the low temperature tensile results from Peterson [29] where the tensile elongation was reduced at -25 and -60 °C. Again, Type 305 displays the best biaxial stretching ability for the three alloys tested at 25 °C.

The FLCs shown in Figure 4.6 exhibit horizontal forming limits on the right side of the FLDs. The tight grouping of data points shown for Types 304 and 305 instills confidence in the FLCs for these two alloys. Whereas, the data point scatter shown for Type 301 makes placement of the FLC more ambiguous. Comparison of Figure 4.5b with Figure 4.6b shows that the stretchability of Type 304 improved with higher deformation temperature and increased austenite stability for testing conducted at 45 °C compared with 25 °C. Further comparisons between 45 °C and 25 °C FLDs show slightly enhanced stretching for Type 301 and no significant change for Type 305.

The FLDs for each alloy tested at 60 °C are shown in Figure 4.7. Each alloy displayed limited biaxial stretching (low values of positive minor true strain) which was the result of low dome heights and fracture in the free span. Both fracture location and
dome height are reviewed further in Section 4.6. Although subtle differences exist in the shapes of FLCs, the overall formability was similar for all three alloys tested at 60 °C. The formability of each alloy should become more similar as the deformation temperature approaches the \( M_D \) (the temperature above which no martensite will form) compared to the highly metastable behavior observed at low temperatures.

The formability of Types 301, 304, and 305 can be quantified and compared by considering the maximum values of major true strain corresponding with specific minor true strain values on the FLCs. Figure 4.8 summarizes the strain limits for each material, minor true strain value, and test temperature. For Type 301, the limiting major true strains in drawing \((\varepsilon_2 = -0.1)\) and plane strain \((\varepsilon_2 = 0)\) increased with forming temperature up to 45 °C, then decreased at 60 °C. For stretching \((\varepsilon_2 = 0.1)\), the strain limits increased with temperature up to 60 °C. The increase in formability with temperature may result from the concomitant reduction in the volume fraction of martensite (VFM) and a shift in martensite formation to higher strains.

For Type 304, the limiting major true strains in drawing, plane strain, and stretching first decreased, then increased as temperature was increased to 60 °C. The initial decrease may be attributable to a decrease in the VFM. At higher temperatures, the return of improved formability may be associated with a more optimal martensite formation rate, i.e. just enough transformation to diffuse neck formation, thereby increasing uniform elongation.
Figure 4.8 Limiting values of major true strain ($\varepsilon_1$) corresponding with minor true strain values ($\varepsilon_2$) displayed on the forming limit curves for Types (a) 301, (b) 304, and (c) 305.
For Type 305, the limiting major true strains decreased for drawing as
temperature was increased to 60 °C. In plane strain and stretching, strain limits decreased
from 15 to 45 °C, presumably due to decreased VFM. Then, limiting strains increased
from 45 to 60 °C, suggesting the action of some alternate mechanism which affects the
formability of the fully austenitic structure at 60 °C.

Three observations were made from the forming limit data presented in this
section. First, in order to attain good formability, it may be beneficial to promote or
suppress martensite formation (relative to room temperature behavior) depending on the
alloy composition and strain path. Second, the material properties required for good
formability are different depending on which deformation mode predominates: drawing,
plane strain, or stretching. For example, Type 304 showed better drawability when
martensite was promoted at 15 °C and improved stretchability if martensite formation
was suppressed at 60 °C. Third, Type 305 displayed similar strain limits in plane strain
and stretching when the $\gamma \rightarrow \alpha'$ transformation was either promoted at 15 °C or suppressed
at 60 °C. This implies the action of two competing mechanisms: transformation
plasticity at low temperature versus work hardening the fully austenitic structure at
elevated temperature.

4.3 Strain-Induced Martensite Formation

The volume fraction of martensite in each limiting dome height (LDH) sample
was measured with a commercial ferrite meter. Individual measurements were taken
only at locations which were designated as safe points on the forming limit diagrams (FLDs) presented in Section 4.2. Figures 4.9 through 4.11 show the volume fractions of martensite (VFM) for each safe point, denoted by various symbols. In addition, lines of constant effective true strain are included to help determine the relationship between effective strain and martensite formation. Regions which are identified by letters pertain to the discussion which follows in this section.

Effective strain lines shown in Figures 4.9 through 4.11 were constructed using the Von Mises effective strain equation [41],

\[ \bar{\varepsilon} = \sqrt{\frac{2}{3} (\varepsilon_{\text{major}}^2 + \varepsilon_{\text{minor}}^2 + \varepsilon_{\text{thickness}}^2) } \],

which simplifies complex states of strain (or stress) into a single parameter. As a result, strain measurements obtained through different strain paths, e.g. drawing, plane strain, or stretching, may be compared on an equivalent strain basis. Lines of constant effective true strain were calculated assuming constant volume.

In general, the VFM correlates with the magnitude of effective strain. An increase in effective strain increases the amount of martensite formed for a given deformation temperature and alloy composition. For example, consider the 25 °C testing of Type 305 shown in Figure 4.11b. Martensite was not formed below 0.6 effective strain, 1 to 9% martensite formed between 0.6 and 0.7 effective strain, and 10 to 19% martensite formed above 0.7 effective strain. Similar trends are evident for the 301 material shown in Figure 4.9 and for 304 shown in Figure 4.10.
Figure 4.9 Volume fraction of martensite (VFM) measured in Type 301 LDH samples tested at: a) 15 °C, b) 25 °C, c) 45 °C, and d) 60 °C. Data points denote the amount of strain-induced martensite produced during forming. Dashed arcs represent lines of constant effective true strain.
Figure 4.10  Volume fraction of martensite (VFM) measured in Type 304 LDH samples tested at a) 15 °C, b) 25 °C, c) 45 °C, and d) 60 °C. Data points denote the amount of strain-induced martensite produced during forming. Dashed arcs represent lines of constant effective true strain.
Figure 4.11  Volume fraction of martensite (VFM) measured in Type 305 LDH samples tested at a) 15 °C, b) 25 °C, c) 45 °C, and d) 60 °C. Data points denote the amount of strain-induced martensite produced during forming. Dashed arcs represent lines of constant effective true strain.
The simple relation between martensite formation and effective strain can be complicated by the underlying stress state effects. Several authors [14, 17-19] have examined the influence of stress (strain) state on the $\gamma \rightarrow \alpha'$ transformation. Their results indicate that more martensite is formed in uniaxial tension than in plane strain [18, 19] and more martensite forms in biaxial tension than uniaxial tension [14]. The results presented here also indicate that more martensite forms in uniaxial tension (drawing) or biaxial tension (stretching) than in plane strain ($\varepsilon_2 = 0$) for an equivalent level of effective strain. For example, in Figure 4.9b, 50 to 59% martensite was formed in region D near plane strain, whereas 60 to 69% martensite formed in region E for the same level of effective strain. Also, in Figure 4.9d, more martensite formed in regions G (drawing) and I (stretching) than for region H near plane strain.

The temperature dependence associated with strain-induced martensite formation can be seen in Figure 4.10 for Type 304. For a given level of effective strain, the VFM increased with a decrease in temperature. Small differences in deformation temperature (10 to 20 °C) supply the necessary thermodynamic driving force to affect the $\gamma \rightarrow \alpha'$ transformation. With decreasing temperature, the austenite gains sufficient energy from undercooling to overcome the barriers of nucleation and growth for the bcc nuclei in a foreign fcc matrix [42]. Consider the boxed regions in Figure 4.10. These regions were designated for a similar strain path and effective strain. For region K, 30 to 39% martensite was formed at 25 °C, whereas 50 to 59% martensite formed in region J when
the deformation temperature was reduced only 10 °C. Similar trends are evident for comparisons with regions L or M.

In Section 4.2, the formability of Type 301 was shown to improve with increased deformation temperature, presumably by reducing the formation of martensite. Other studies [15] have shown that the extent of uniform elongation in uniaxial tension depends on the "timing" of the $\gamma\rightarrow\alpha'$ transformation. If transformation occurs "early" at low strains, strength increases rapidly resulting in reduced total elongation. In a more optimal situation, transformation occurs "late" at higher strains, suppressing neck formation and extending the total elongation. The benefits of limiting the austenite transformation can be seen in Figures 4.9a and 4.9b. Referring to regions A and C, when the VFM increased with a decrease in deformation temperature from 25 to 15 °C, the magnitude of effective strain was reduced. Also, by comparing region B with F, the biaxial ductility of Type 301 decreased with an increase in the VFM.

In summary, the amount of martensite formed was proportional to the magnitude of effective strain, for a given temperature and alloy. For an equivalent effective strain, more martensite formed in drawing or stretching than in plane strain and the VFM increased when deformation temperature was reduced. Also, the formability of Type 301 was shown to improve by suppressing the formation of martensite. Conversely, the formability of Types 304 and 305 improves with increased martensite formation as the deformation temperature was reduced below room temperature.
4.4 Forming Limit Curves

The formability of austenitic stainless steels will be affected by the deformation temperature and by alloy stability as these variables relate to the formation of strain-induced martensite. With this in mind, the forming limit curves (FLCs) from Section 4.2 have been reproduced here with the data points removed. FLCs are grouped by test temperature in Figure 4.12 and by alloy designation in Figure 4.13.

Figure 4.12 shows the FLCs obtained at different test temperatures for each of the three alloys. Consider the low stability alloy, Type 301 shown in Figure 4.12a. Excessive martensite forms (60 to 80%) at 15 °C, reducing the ductility of Type 301 for all strain paths represented on the diagram. When the deformation temperature was increased to 25 °C, the drawability and biaxial ductility increased the strain limits approximately 10% and 20%, respectively. At 45 °C, the biaxial ductility again increased, while drawing remained unchanged. No apparent benefit was realized by increasing the test temperature to 60 °C.

Next, consider the intermediate stability alloy, Type 304 shown in Figure 4.12b. The material behavior on the stretching side (right side of the diagram) mimics the results shown for Type 301, i.e. biaxial ductility improved when the γ→α' transformation was suppressed at higher deformation temperatures. Conversely, the drawability decreased with reduced VFM as the test temperature was increased to 60 °C.

In Figure 4.12c for Type 305, the stretching behavior differed from the results shown for Types 301 and 304. In this case, the biaxial ductility of Type 305 decreased as
Figure 4.12  Forming limit curves obtained at four test temperatures for Types a) 301, b) 304, and c) 305. Test conditions: 0.212 mm/s punch velocity and polyethylene lubrication, 80 microns thick.
Figure 4.13  Forming limit curves for Types 301, 304, and 305 tested at a) 15 °C, b) 25 °C, c) 45 °C, and d) 60 °C. Test conditions: 0.212 mm/s punch velocity and polyethylene lubrication, 80 microns thick.
temperature was increased from 25 °C to 45 °C. Similar FLCs were obtained at 15 °C
and 60 °C, which indicates that comparable formability could be obtained by either work
hardening the austenite or by the combination of work hardening coupled with the
transformation plasticity.

In Figure 4.13, the FLCs are grouped by alloy designation at each of the four test
temperatures. The relative stability of these alloys was predicted from the constitutive
\( M_{D30} \) equation proposed by Nohara et al. (Equation 4.1) [20]. Based on these
calculations, Type 305 had the highest austenite stability, Type 304 exhibited some
intermediate stability, and Type 301 displayed the least stability.

The FLCs presented in Figure 4.13 indicate that the material properties required
for good drawability are not quite the same as those required for good stretchability.
Figure 4.13a, shows that at 15 °C the highest drawing limits were recorded for Type 304,
whereas Type 305 performed best in biaxial tension. At room temperature, Type 301
showed the highest drawability and Type 305 exhibited the highest stretchability, similar
to the 15 °C results. As the test temperature was increased to 45 °C, Type 301
maintained superior drawability with Type 304 showing improved stretchability. At
60 °C, Type 301 showed the highest overall formability.

An examination of the temperature dependent FLCs emphasizes the differences in
formability attributed to variations in martensite content. By changing the deformation
temperature, the martensite formation rate could be manipulated to obtain maximum
uniform ductility in a given strain path. Given the difficulty associated with controlling
the forming temperature, appropriate alloy selection may be a more logical alternative. In this case, austenitic stainless steels would be chosen based on austenite stability and the rate of work hardening required.

4.5 Strain Distribution

Metastable austenitic stainless steels exhibit extensive uniform elongations as a result of a unique work-hardening mechanism. In addition to the typical strain-hardening which results from dislocation interaction and multiplication, the strain-induced transformation from austenite to martensite contributes to the strain distribution; and if transformation occurs at high strains, necking is suppressed and ductility increases. Through-thickness strain distributions are shown in Figures 4.14 through 4.16 for Types 301, 304, and 305 LDH samples tested to failure. In each figure, thickness strains and corresponding volume fractions of martensite (VFM) are plotted against the original distance from the pole. The “original distance” corresponds with circle grid dimensions prior to deformation. Measurements were taken along two lines oriented parallel and perpendicular to the sheet rolling direction. These are designated longitudinal (L) and transverse (T) in the following discussion.

The strain distribution capability of each alloy can be quantified by the thickness gradient measured across the centerline of a LDH sample. For example, the Type 301 25T curve (25 °C transverse, open square symbols) shown in Figure 4.14a demonstrates an excellent through-thickness strain distribution which is uniform across the dome from
Figure 4.14  Through-thickness strains and volume fractions of martensite in 178 x 178 mm LDH samples of Type 301. Measurements were taken along two centerlines oriented parallel and perpendicular to the RD. Open and solid symbols indicate the transverse and longitudinal orientations, respectively.
Through-thickness strains and volume fractions of martensite in 178 x 178 mm LDH samples of Type 304. Measurements were taken along two centerlines oriented parallel and perpendicular to the RD. Open and solid symbols indicate the transverse and longitudinal orientations, respectively.
Figure 4.16 Through-thickness strains and volume fractions of martensite in 178 x 178 mm LDH samples of Type 305. Measurements taken along centerlines oriented parallel and perpendicular to the RD. Open and solid symbols indicate the transverse and longitudinal orientations, respectively.
-40 to 40 mm at a value of approximately 45% thinning. A broad horizontal curve shows the absence of localized necking which would be consistent with good formability. Conversely, the 60T and 60L curves shown in Figure 4.14b indicate poor strain distribution with steep strain gradients and marked peaks corresponding with the necked regions at -40 and 40 mm. As expected, the point of maximum thinning also coincides with the fracture location for each sample measured. Fracture locations are presented and discussed in Section 4.6.

The thinning distributions appear to correlate directly with the VFM for each sample. Figures 4.14 through 4.16 show the shape and magnitude of the thinning curves for a particular test temperature correspond directly with the associated VFM curves. For each alloy, the height of the thinning curves corresponds with the dome heights attained. Dome heights are listed in Table 4.3 for the strain distribution samples. For example in Table 4.3 – Dome Heights for the Through-Thickness Strain Distribution Samples.

<table>
<thead>
<tr>
<th>Testing Temperature (°C)</th>
<th>Dome Height (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>301</td>
</tr>
<tr>
<td>15</td>
<td>52</td>
</tr>
<tr>
<td>25</td>
<td>56</td>
</tr>
<tr>
<td>45</td>
<td>58</td>
</tr>
<tr>
<td>60</td>
<td>52</td>
</tr>
</tbody>
</table>
Figure 4.15a, the thinning curves for Type 304 are ranked 25, 45, 15, and 60 °C (from highest to lowest percent thinning) corresponding with 58, 56, 55, and 50 mm dome heights.

In Figures 4.14a and 4.14b, Type 301 strain distribution curves show greater thinning in the transverse orientation compared with longitudinal curves. Also, the strain distribution was more uniform in the transverse orientation than for the longitudinal samples, specifically for the 15T and 25T curves compared with 15L and 25L data. The fact that transverse curves show greater uniform thinning which suggests a textural effect, possibly linked to the nucleation of martensite. An analysis of the plastic strain ratios (r-values) from Peterson’s Type 301 [29] corroborates the differences in through-thickness flow strengths shown here; i.e. the average r-value measured parallel to the rolling direction (RD) was greater than the r-value taken transverse to the RD, $r_{\alpha} = 1.02$ and $r_{90^\circ} = 0.72$.

The 45T, 60L, and 60T curves shown in Figure 4.14b have different curve shapes from the 15T, 15L, 25T, and 25L curves shown in Figure 4.14a. The higher temperature curves exhibit a characteristic “M” shape marked by steeper strain gradients and locally high thinning near the -40 and 40 mm positions. The appearance of the M-shaped curves corresponds with low VFM found in the higher temperature samples. This suggests that the formation of strain-induced martensite in amounts greater than 50% will increase the strain distribution ability of Type 301.
Figure 4.15 shows the strain distribution and VFM results for Type 304 LDH samples. Contrary to the results shown for Type 301, no apparent differences exist between curves measured in the longitudinal or transverse orientations which suggests more isotropic texture in the Type 304 material. M-shaped curves were recorded for the 60T and 60L samples. Presumably, the low VFM (around 0.25), was sufficient to distribute the thinning across the width of the 45 °C samples and alter the M-shaped strain distribution.

Figure 4.16 shows the strain distribution and martensite content for Type 305 LDH samples. Both longitudinal and transverse curves comply with the M-shaped curve description for all test temperatures. This suggests that departure from the M-shaped strain distribution may require sufficient \( \gamma \rightarrow \alpha' \) transformation not shown in the more stable Type 305.

The results presented in Figures 4.14 through 4.16 indicate that the amount of strain-induced martensite directly affected the way through-thickness thinning strain was distributed. For those samples with lower VFM, M-shaped strain distributions were recorded. At higher VFM, LDH samples exhibited more uniform strain distribution across perpendicular centerlines characterized by thinning curves with a broad plateau. Departure from the M-shaped strain distribution required a sufficient amount of martensite: 50% for Type 301 and 25% for Type 304.

Any discussion about strain distribution or strain gradients would be incomplete without mention of the strain hardening exponents (n-values). The relative degree of
strain distribution in steels increases with n-value; i.e. high strain gradients are associated with low n-values [43]. Strain hardening data from Peterson [29] is reproduced in Figure 4.17 to show the variation of n-values with strain and temperature. In general, the n-values increase with strain to a maximum then decrease to a value equal to the uniform true strain.

For Type 301, the n-values at 25 °C and 55 °C diverge at 0.2 strain and continue to deviate with increased strain. The n-value for 25 °C reaches a maximum value at 0.7 versus 0.5 for 55 °C. The wide separation of n-values for the two temperatures shown is consistent with the difference in VFM that forms during the tensile test: 70% martensite at 25 °C compared with 35% at 55 °C. Also, the greater work-hardening capability at 25 °C is consistent with the uniform distribution of thinning and the greater dome heights observed in the 25 °C LDH specimens. The n-values for Type 304 were similar for the two temperatures examined, although maximum n-values occurred at different strains. The n-values obtained at 25 °C peaked at 0.4 engineering strain while n-values at 55 °C peaked at 0.5.

For Type 305, the n-values increased from 0.1 to 0.2 strain then remained relatively constant or decreased slightly with increased strain. For this material, the n-values saturate early in the strain cycle, at low strains. As a result, the material which is undergoing deformation will not work-harden significantly and deformation may proceed in the area that deforms first. Thus, deformation can localize more readily than in the case of an ever increasing n-value, e.g. Type 301. The general inability of Type 305 to
Figure 4.17  Variation of the strain hardening exponent with strain for Types a) 301, b) 304, and c) 305 tested at 25 °C and 55 °C with an engineering strain rate of 1.67 x 10^-3.
increase its work-hardening rate with strain is consistent with the localized thinning shown in Figure 4.16 (M-shaped curves).

N-values have been shown to vary with increased strain and martensite content. For this reason, a single n-value may be inadequate to describe the strain distribution ability of these samples. Rather, an average n-value would more accurately reflect the material behavior. For example, the n-values shown in Figure 4.17b for 25 °C and 55 °C are similar in magnitude but the average n-value at 25 °C is greater. Thus, the strain distribution at 25°C was more uniform than at 60 °C for the Type 304 shown in Figure 4.15a. Similar comparisons can be made for Types 301 and 305 n-values and their corresponding strain distributions.

4.6 Fracture Location and Orientation

The orientation and location of fracture in limiting dome height (LDH) samples was influenced by sample geometry, deformation temperature, and/or alloy grade. Figures 4.18 through 4.20 show schematic drawings of each LDH sample (from an aerial perspective) with the failure locations indicated. These figures include the complete test matrix for Types 301, 304, and 305 specimens tested at four temperatures. Fracture locations are designated in five classes: near the drawbead (Class I – failure ⊥RD) and (Class II – //RD); midway between the drawbead and pole (Class III – ⊥RD) and (Class IV – //RD), or centered across the pole (Class V – ⊥RD). This classification scheme was derived from the method used by Peterson [29].
Figure 4.18  Effects of deformation temperature and sample geometry on the location and orientation of fracture in Type 301 LDH samples.
<table>
<thead>
<tr>
<th>mm</th>
<th>178 x 178</th>
<th>178 x 152</th>
<th>178 x 140</th>
<th>178 x 127</th>
<th>178 x 76</th>
<th>178 x 51</th>
</tr>
</thead>
<tbody>
<tr>
<td>(in)</td>
<td>(7 x 7)</td>
<td>(7 x 6)</td>
<td>(7 x 5.5)</td>
<td>(7 x 5)</td>
<td>(7 x 3)</td>
<td>(7 x 2)</td>
</tr>
</tbody>
</table>

### 15 °C Deformation Temperature

- **Class IV**
- **Class IV**
- **not tested**
- **Class III**
- **Class V**
- **Class V**

### 25 °C Deformation Temperature

- **Class IV**
- **Class IV**
- **Class III**
- **Class III**
- **Class V**
- **Class V**

### 45 °C Deformation Temperature

- **Class IV**
- **Class IV**
- **Class III**
- **Class III**
- **Class V**
- **Class V**

### 60 °C Deformation Temperature

- **Class I**
- **Class II**
- **Class I**
- **Class I**
- **Class V**
- **Class V**

---

**Figure 4.19**  Effects of deformation temperature and sample geometry on the location and orientation of fracture in Type 304 LDH samples.
Figure 4.20 Effects of deformation temperature and sample geometry on the location and orientation of fracture in Type 305 LDH samples.
LDH samples were locked in place by a 142 mm (5.6 in) diameter drawbead. Samples which were narrower than 142 mm experienced more elongation in the longitudinal direction due to the limited width constraint. As a result, all of the 140, 127, 76, and 51 mm samples (5.5, 5, 3, 2 in samples) failed by cracking perpendicular to the rolling direction. In the fully constrained samples, 178 and 152 mm width (7 and 6 in), fracture orientations were observed both parallel and perpendicular to the rolling direction. The variation in fracture orientation for the fully constrained samples was presumed to be a function of the metallurgical state rather than sample geometry.

Thickness strain distributions were presented in Section 4.5 for 178 x 178 mm (7 x 7 in) samples of each alloy and temperature. Those strain distribution curves were useful in understanding fracture locations because failures occur in areas which thin preferentially. For example, consider the M-shaped strain distribution curves shown in Figure 4.14b for Type 301. Both the 45T and 60T (45 and 60 °C transverse) curves had peaks near the ±40 mm positions which coincide with the Class II fractures for those samples shown in Figure 4.18. Also in Figure 4.18, there was a transition for fully constrained samples from Class I and II fractures at 45 and 60 °C to Class V fractures at the pole for 15 and 25 °C samples. Both the 15L and 25L curves shown in Figure 4.14a depart from the M-shaped strain distribution to one where maximum thinning occurs near the pole, which accounts for the fracture location. The longitudinal curves also display steeper strain gradients near the pole which is consistent with the fracture orientation perpendicular to the rolling direction.
Figure 4.19 shows the effects of temperature and sample geometry on the fracture location in Type 304 LDH samples. There was a transition for all the rectangular samples from Class I and II fractures at 60 °C to Class III and IV fractures at 15, 25, and 45 °C. The reason for this transition can be understood from a comparison between the thickness strain distributions shown in Figure 4.15a with the failure locations shown in Figure 4.19. The strain distribution at 60 °C differed from the more uniform thinning shown at the lower test temperatures. As in Type 301, M-shaped strain distributions are consistent with Class I and II fractures while the more uniform strain distributions resulted in Class III and IV fractures.

Figure 4.20 shows the effects of temperature and sample geometry on the fracture location in Type 305 LDH samples. Fracture locations for the fully constrained samples correspond with peaks in the strain distribution curves, analogous with the discussion above. In Figure 4.16, the 15 and 60 °C strain distribution curves showed thinning peaks around ±35 mm which corresponds with the Class II fractures shown in Figure 4.20. Also in Figure 4.16, the 25 and 45 °C curves showed thinning peaks closer to the pole which coincides with the Class IV fractures shown in Figure 4.20. The occurrence of Class II fractures at 15 and 60 °C or Class IV fractures at 25 and 45 °C is consistent with the data presented in Figure 4.8. The major true strain ($\varepsilon_1$) was highest at 15 and 60 °C in plane strain ($\varepsilon_2 = 0$) and stretching ($\varepsilon_2 = 0.1$) while $\varepsilon_1$ diminished at 25 and 45 °C for the same stress state. It would be expected then that the fracture location would be closer to the pole at these two intermediate temperatures, as observed.
In summary, fracture locations generally coincided with the strain peaks shown in the through-thickness strain distribution curves presented in Section 4.5. Samples which were not fully constrained (less than 142 mm (5.6 in) wide) fractured perpendicular to the rolling direction. Fully constrained LDH samples failed either parallel or perpendicular to the rolling direction. Non-uniform M-shaped strain distributions caused Class I and Class II fractures. Relatively high values of thinning at the pole, coupled with adjacent strain gradients resulted in Class V fractures, e.g. 15L and 25L Type 301 samples. Fully constrained samples of Type 304 and 305 generally failed parallel the rolling direction, coincident with the location of maximum thinning along the transverse orientation.
5.0 SUMMARY AND CONCLUSIONS

The formability of Types 301, 304, and 305 austenitic stainless steels has been evaluated. Formability was assessed by differences in the forming limit curves (FLCs), volume fractions of martensite (VFM), and strain distributions. Based on the experimental results presented in Section 4.0, the following observations were made.

The strain-induced transformation from austenite to martensite had a profound effect on the punch-stretch formability of Types 301, 304, and 305. In order to attain an optimum balance between cold-worked strength and formability, it may be necessary to promote or suppress martensite transformation depending on the alloy composition (austenite stability) and specific forming requirements. For Type 301, the strain limits increased when martensite formation was suppressed by increasing the forming temperature. Whereas the higher stability grades, Types 304 and 305, showed improved formability by thermally assisting the transformation at low temperature.

The material properties required for good formability were different depending on which deformation mode predominated: drawing, plane strain, or stretching. Consider the room temperature FLCs shown in Figure 4.13b. Each alloy outperformed the other two alloys by withstanding higher limiting strains, depending on the strain path; i.e. Type 301 performed best in drawing, Type 304 in plane strain, and Type 305 in stretching. Also, the particular stress state (strain path) modifies the martensite transformation, i.e.
more martensite forms in drawing or stretching than in plane strain for a given effective strain.

An examination of the VFM formed at different deformation temperatures emphasizes the differences in formability attributed to variations in martensite content. The VFM increased as the forming temperature decreased and was proportional to the effective strain. By changing the deformation temperature, the martensite formation rate could be manipulated to achieve maximum uniform ductility for a given alloy and strain path. Alternatively, austenitic stainless steels could be selected for each application based on austenite stability and the rate of work hardening required.

For metastable austenitic alloys, strain hardening comes from two sources: dislocation interaction and multiplication in the austenite and the strain-induced transformation to martensite. Both mechanisms contribute to the variation in n-value with strain. For this reason, a single n-value may be inadequate to describe the strain distribution capability of these alloys. Rather, an average n-value appears to reflect the material behavior more accurately.

The results from through-thickness strain measurements indicate that the extent of transformation to martensite directly affects the way strain was distributed. At low VFM, distributions with steep strain gradients and marked peaks were observed. At higher VFM, plastic strain was more uniformly distributed across the hemispherical dome of each LDH sample.
Based on the experimental results, the following conclusions can be made with respect to forming limits, volume fractions of martensite, and strain distribution.

5.1 Forming Limits

- Forming limits can be extended by adjusting the deformation temperature which will alter the martensite transformation.

- Material properties required for good formability depend on the predominant deformation mode; i.e. drawing, plane strain, or stretching.

- Suggested forming temperatures for the temperature range investigated are given in Table 5.1:

Table 5.1 – Optimum Forming Temperatures for Types 301, 304, and 305.

<table>
<thead>
<tr>
<th>Forming Temperatures</th>
</tr>
</thead>
<tbody>
<tr>
<td>Type 301</td>
</tr>
<tr>
<td>45 °C*</td>
</tr>
<tr>
<td>*60 °C for stretching</td>
</tr>
</tbody>
</table>

5.2 Volume Fractions of Martensite

- Formation of strain-induced martensite is proportional to the effective strain.

- Martensite formation increases with decreasing deformation temperature.

- For an equivalent level of effective strain, more martensite forms in drawing or stretching than in plane strain.
• Strain-hardening exponents increases with increasing volume fractions of martensite.

5.3 Strain Distribution

• Martensite formation improves the strain distribution by increasing the n-value.

• The strain distribution ability of each alloy can be quantified by the thickness gradient measured across the centerline of a LDH specimen.

• At high temperature (60 °C) and low martensite content, strain distribution is non-uniform with steep strain gradients and marked strain peaks. Poor strain distribution reduces the formability as measured by dome height.

• At low temperature (15 °C) and high martensite content, strain is uniformly distributed. Uniform strain distribution increases the dome height except where excessive martensite reduces ductility, e.g. >60% for Type 301.

• Sufficient martensite transformation is required to achieve uniform strain distribution, e.g. 50% for Type 301 and 25% for Type 304.

• Differences in the through-thickness strain distributions (thinning strain) with sample orientation suggests a preferred crystallographic texture for Type 301.
6.0 FUTURE WORK

Considerable understanding has been acquired during this research project with respect to punch-stretch formability testing. During the experimental work, several system modifications were recognized which could improve limiting dome height (LDH) testing. Also, alternative test methods could be developed which would more closely approximate industry conditions. During the analysis, it became apparent that more information about strain evolution and martensite formation could be gained through incremental dome height testing. These observations suggest potential areas for future research which are discussed in the following three sections.

6.1 LDH Test System

Two modifications are suggested to improve the LDH test system. First, develop a detection system capable of sensing the onset of necking (plastic instability) with a quick enough response time to stop the test prior to fracture. Roamer [44] designed an adequate system for LDH testing conducted in load control, although similar systems would be useful for testing in stroke or strain control. Second, a tooling improvement is suggested for LDH testing with the temperature control system. In the current program, the hemispherical punch was removed before each test and heated/cooled separately, then replaced prior to testing. An integral fluid channel could be machined inside the
hemispherical punch to enable circulation of a heating/cooling fluid. This would improve testing efficiency and temperature control.

6.2 Alternative Test Methods

An ideal laboratory testing system which would enable stretch-formability testing at specific temperatures would involve immersing the sample, punch, and fixtures in an stirred isothermal bath. Given the experimental difficulty of establishing such a system, two alternative methods are proposed. First, temperature sensitive alloys could be investigated with hydroforming technology by controlling the temperature of the fluid. Second, both temperature behavior and strain path dependence could be investigated by integrating fluid circulation and variable draw beads into a matched set of dies. Both hydroforming and press forming have considerable relevance with respect to actual industrial processes.

6.3 Incremental Dome Height Testing

Generally, the limiting dome height test involves stretching the sheet specimen until plastic instability is detected or failure. Considerable insight could be gained into the distribution of strain over time and the corresponding martensite formation rates by examining test specimens at incremental heights. Alternatively, real-time measurements of surface strains and martensite content during stretch-formability testing could provide similar information.
7.0 REFERENCES


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