Internal ductile failure mechanisms in steel cold heading process

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Abstract

The occurrence of internal ductile failure in cold-headed products presents a major obstacle in the fast expanding cold heading (CH) industry. This internal failure may lead to catastrophic brittle fracture under tensile loads despite the ductile nature of the material. Comprehensive testing and investigation methodologies were used to this work to reveal the complicated interplay of process and material parameters contributing in the initiation and propagation of internal ductile failure in six CH quality AISI steel grades.

The metallurgical and microscopic investigations showed that internal ductile failure occurs progressively by void nucleation and growth mechanisms with increasing plastic strain inside the highly localized adiabatic shear bands (ASBs). The void nucleation occurs by decohesion at second-phase particles, inclusion–matrix interfaces, grain boundaries and by particle or inclusion cracking. Therefore, the number and morphology of any inclusions and second-phase particles are key factors in material formability.

The metallurgical investigations showed that under compressive loading conditions, the nature of the metal flow pattern promotes different rates of material flow around the inclusions and stringers which supports decohesion and void nucleation since the early stages of deformation. At advanced stages of deformation, the metal flow pattern contributes to the ASB localization in supporting void growth and coalescence along the band leading to narrow void sheets.

All tested materials in this work experienced ductile failure by void nucleation and coalescence, forming cracks along the ASBs. The ductile failure of each material was the result of the contribution of all the mechanisms of void nucleation at the inclusion–matrix interface, second phase–matrix interface and at the grain boundaries. However, the level of contribution of each mechanism in the final ductile failure varied depending on material properties and their microstructure.

1. Introduction

The occurrence of any failure is a major limitation governing the limits of any forming process. Therefore, understanding the failure mechanisms and the complex interplay of process and material parameters in the failure occurrence in metal forming operations has attracted the attention of many researchers for more than six decades. The knowledge of different failure mechanisms is crucial to improve product quality and design methodologies.

Barrett (1997) considered in his report about the fasteners that the cold heading (CH) process is one of the most important multi-stage metal forming processes because of its many advantages over machining of the same part, including high productivity for complex final shapes, minimum material waste and increased tensile strength from cold working. Chitkara and Bhutta (2001) who studied the near-net shape heading of splines and solid spur gear forms reported that the decision to use the CH process to manufacture certain products depends on the complexity in the shape of the head and the material employed and the possibility of internal and external failure occurrence.

Currently, the CH industry favors using faster headers, reducing the number of manufacturing stages, and producing high-strength fasteners without final heat treatment. To achieve these goals, modified process designs are required which result in higher strains and strain rates, which cause two familiar types of ductile defects in the cold-headed products. The first is the external oblique or longitudinal crack caused by the exhaustion of the material ductility. The second failure was reported by Bai and Dodd (1992) in their study for the adiabatic shear band (ASB) phenomenon. They reported that the ASBs may lead to internal crack. Okamoto et al. (1973) stated in their study of different forming processes that this type of failure may result in splitting of the fasteners’ heads as shown in Figs. 1–3, respectively.

Many researchers have focused on surface defects. Cockroft and Latham (1968) and Lee and Kuhn (1973) studied this defect in upsetting and presented different criteria to predict it.
(2000) studied the surface defect in CH process and reached to a conclusion that Cockroft–Latham criterion can be used to predict the surface defect in CH process with a reasonable accuracy. In addition to the large number of studies that covered this type of failure extensively, it can be visually inspected and accounted for in the process design of the component for removal by means of machining or trimming. Thus far, internal ductile failure due to the ASB phenomenon in CH process has not received the same atten-

Fig. 1. Cold-headed industrial bolt: (a) without defect, using material 1, (b) shank of the defective bolt (material 2), (c) separated industrial bolt head due to an internal crack (material 2), and (d) magnified detail of the ductile crack of the separated industrial bolt head (material 2).

Fig. 2. Sectioned bolts revealing material flow inside the bolt head after etching with Fry's reagent (cupric chloride 36 g; 145 ml hydrochloric acid; 80 ml water): (a) without defect bolt (material 1), (b) with defect bolt (material 2).
tion. The clear need for an in-depth understanding of the internal ductile failure mechanisms which limits the development of new CH designs and the improvement of cold-headed product quality motivated this work.

To have a full understanding of the complicated parameters and mechanisms contributing in the initiation and propagation of the internal ductile failure, the first part of this paper will provide a detailed introduction covering important details about the CH process and the ASB phenomenon followed by examples of ductile failure in cold-headed parts. The second part of this paper will focus on the experimental and metallographic procedures used to investigate the failure initiation and propagation. The last part of this paper presents the detailed result and discussion.

1.1. Cold heading process

The term heading refers to cold, warm, or hot working of metal employing a static or dynamic impact compressive force causing the metal at one end of the rod or blank to form a head. Yoo et al. (1997) defined the cold heading (CH) process as a forming process that is performed without an external heat source and it involves applying a force to the free end of a metal workpiece contained between a die and a punch, by one or several blows of the punch. The CH process is a high-rate deformation process with strain rates exceeding $10^2$ s$^{-1}$. Kawashima (1992) reported that the CH process is used to produce a large variety of components such as fasteners, studs, small shafts, and other machine parts. The automotive, construction, aerospace, railway, mining and electrical product industries are major consumers of such parts.

Matsunaga and Shiwaku (1980) concluded in their study of manufacturing CH quality wire rods and wires that material properties and process parameters significantly affect the heading results. Moreover, the CH material must meet two sets of requirements. The first set concerns the material's ability to be formed without failure while the second set is linked to the properties of the final product. Sarruf et al. (1999) stated that in order to form without failure, CH wire should be free from surface defects and internal inclusions. Nickoleotopulos et al. (2001) reported that the chemical composition of CH wires plays a significant role in the material ductility and the final properties of the products. Low- and medium-carbon steels (0.06–0.30% C), are commonly used in the CH process, as they offer good ductility. Also, Nickoleotopulos et al. (2001) referred that the CH material microstructure is usually comprised of a ferrite matrix with varying amounts of lamellar pearlite. A spheroidized microstructure provides the desired ductility for CH of complex geometries. The required final strength properties are usually achieved through alloy addition and/or heat treatment (quenching and tempering).

1.2. ASB phenomenon

Wright (2002) defined the ASB as a two-dimensional, narrow, nearly planar region of very large shearing that occurs in metals and alloys experiencing intense dynamic loading (as in the CH process). When the band is fully developed, the two sides of the region are displaced relative to each other. However, the material still retains full physical continuity from one side to the other. The thickness of the most heavily sheared region might be few tens of microns or less, and its length might extend many millimeters or centimeters. Generally, ASBs form under impact loading at high-strain rates (higher that $10^2$ s$^{-1}$) and high strains. With increasing plastic deformation, the work hardening (from increases in strain and strain rate) results in an increase in the flow stress. However, most of the plastic work (90–95%) is converted into heat causing a local temperature increase and a flow stress decrease due to thermal or work softening. Thus, a competing mechanism between the work hardening and the thermal or work softening commences and continues in the deformation zone. As the work hardening mechanism dominates over the work softening at the beginning, an increase in the flow stress occurs and with continuing deformation, the thermal or work softening mechanism can progressively dominate over work hardening increases, which then triggers unstable deformation. Wright (2002) also reported that this instability condition will force the deformation to localize into a narrower band that through further localization can lead to final failure. Cowie et al. (1989) showed in their study of microvoid formation during shear deformation of ultrahigh-strength steels that this failure initiates and develops by progressive nucleation, growth and coalescence of microvoids and microcracks. A sharp drop in the load-carrying capability for the deformed zone will associate with microvoid formation causing what is known as the microvoid softening. Once the failure starts to develop, the combined work softening effect of the thermal and microvoid softening mechanisms will compete with the work hardening effect within the band.

Zukas (1990) reported that there are two types of ASBs, namely the deformed adiabatic shear bands (DASBs) and the transformed adiabatic shear bands (TASBs). DASBs occur in materials that do not undergo phase transformation, or when the local temperature

Fig. 3. Cold-headed brass bolt experienced ductile failure in service along the ASB: (a) and (b) sectioned failed bolt, (c) crack surface in a failed bolt (Reitz, 2005) Reprinted with permission of ASM International (www.asminternational.org).
within the band is not high enough to cause phase transformation. The damage and fracture process in DASBs involves a number of metallurgical events involving different steps between void nucleation and crack propagation. The initial phase of damage coincides with void nucleation at inclusion edges or at grain boundaries due to the interaction of local stresses and dislocations phenomena. The damage in the material is then driven by the accumulated plastic strain and affected by the stress triaxiality. Hambli (2001) reached an important conclusion that deformation under compressive (negative) stress triaxiality increases material formability in comparison with deformation under tensile loading state, i.e. positive stress triaxiality. Zukas (1990) declared that the competition between the stress triaxiality levels reached in the specimen and the high-plastic strains will determine the failure site.

Daridon et al. (2004) reported that the localization of plastic deformation in narrow shear bands is a major damage mechanism that leads to ductile failure during high-strain rate deformation. Lee et al. (2004) reported that internal cracks occur within the shear band, and become the sites of eventual failure. Venugopal et al. (1997) concluded that this can take place when the material is subsequently subjected to dynamic impact loading during forming or in service.

TASBs occur in materials that undergo phase transformations, and are often found in high-strength steel in locations where the critical temperature for the transformation of ferrite to austenite (Ac₃) is exceeded. After high-strain rate forming, the rapid heat dissipation from the ASB to the surrounding matrix material results in a fast decrease in temperature below the Ms limit. This sudden change in temperature leads to austenite transformation into martensite, yielding the TASBs. TASBs may fracture in a brittle manner in directions normal to the local tensile stress. If the matrix is sufficiently ductile, brittle fracture will be limited to the band, but sometimes the fracture may extend to the matrix. Rogers (1983) stated that even if the TASB does not fracture, it remains a brittle fracture path in the middle of the ductile matrix and might lead to catastrophic failure. Moreover, Klepaczko et al. (1988) have

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**Fig. 4.** As-received testing materials microstructure: (a) 1008 steel, (b) 1018 steel, (c) 1038 steel, (d) 1541 steel, (e) 8640 steel, and (f) DP steel.
indicated that shear bands either with or without phase transformations, are considered as a site of fracture initiation.

1.3. Examples of internal ductile failure in cold-headed products

The following paragraphs present cold-headed products failed by internal ductile fracture along the ASBs. Fig. 1 shows in detail industrial bolts were made of two different 1018 steels from two different CH steel providers (material 1 and material 2) using the same CH sequence. The industrial bolt made of material 1 (Fig. 1(a)) did not exhibit internal cracks while the second bolt made of material 2 failed by a crack resulting from the head splitting when a low-tensile load was applied. Fig. 1(b)–(d) shows the details of the crack inside the head of the bolt made of material 2. The metallographic study of the sectioned bolts made of the material 2 (Fig. 2(b)) confirmed the nucleation, growth and coalescence of microvoids leading to microcracks.

Reitz (2005) investigated the failure of brass bolts after 12 years in service. The microscopic investigation revealed that the bolts failed by macrocracks along the highly deformed band inside the bolt’s head (Fig. 3).

In general, the ductile crack occurred inside a highly localized region of the ASB. The occurrence of ductile failure in material 1 using the same CH multistage design in the part in Fig. 1 raises important questions about the ASB phenomenon and how it leads to such a catastrophic failure. The ductile failure that took place inside bolt heads made of material 1 presents clear evidence that the current rules of thumb are neither effective nor adequate in fulfilling the needs of the new design developments in the CH industry. They do not provide enough guidance to fulfill the needs of new design developments in the CH industry. Therefore, the CH industry is in need of systematic and theoretical studies to understand the failure mechanisms and the different parameters contributing to failure initiation and propagation in cold-headed products.

These examples of failed cold-headed products reveal the apparent lack of knowledge about the internal ductile failure mechanism and the role of the ASB phenomenon in the presence of these defects in cold-headed products.

2. Experimental

2.1. Material selection and characterization

The CH quality AISI steel grades chosen for this research are: 1008, 1018, 1038, 1541, 8640 and dual phase (DP) steel which cover low- and medium-carbon content steels and also provide different CH quality steels ranging from best to poorest CH formability steels. The 1018, 1038, 1541, and 8640 steels were received in rod form from Ivaco Rolling Mills (Ontario, Canada), while the 1008 steel was provided from the undeformed part of the bolts made of material 1.

In addition, DP steel was also included in the test materials as it possesses unique properties which show great potential in the CH industry. Persson (1986) reported that the DP steel is intended to be cold headed directly from the “green rod” and reaches the final properties of the cold-headed products without heat treatment.

Optical microscopy revealed a microstructure consisting of ferrite matrix and lamellar pearlite for all steel grades except the DP steel which consisted predominantly of ferrite grains and a martensitic second phase. The as-received microstructure of all the testing materials is shown in Fig. 4. Table 1 lists the chemical composition of the selected material.

Quasi-static tensile tests were performed on specimens machined to ASTM-E8 standard at an approximate strain rate of 0.002 s⁻¹, the force and the axial displacement values were recorded. The engineering stress-engineering strain curves were constructed from the load–elongation measurements. These curves were used to calculate the true stress–true strain curves (Fig. 5) for all testing materials.

2.2. Testing material cleanliness

Cleanliness is a measure of the inclusion content and inclusion types that are deleterious to processing performance and final product properties.

Dodd and Bai (1987) indicate that the inclusions refer to non-metallic particles that are held mechanically in the material matrix. Such particles include oxides, sulfides or silicates. The presence of these inclusions strongly influences the ductile failure mechanism. Thus control of cleanliness and second-phase particle shape can influence material ductility.

The inclusion rating process for this study was performed using automatic image analysis inclusion ratings in accordance with microscopic method A (worst fields) of E45 of the ASTM standard. Worst field is a class of rating in which the specimen is rated for each type of inclusion by assigning the value of the highest severity rating observed of the inclusion type anywhere on the specimen surface.

Method A requires a survey of a polished surface square area of 160 mm² at 100× magnification parallel to the longitudinal axis of the wire. The severity level of the worst fields and the series type of four inclusion types (A, B, C, and D) should be reported for every specimen depending on Tables 2 and 3. Table 2 presents the

<table>
<thead>
<tr>
<th>AISI steel grade</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>Si</th>
<th>S</th>
<th>Cu</th>
<th>Ni</th>
<th>Cr</th>
</tr>
</thead>
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<tr>
<td>1008</td>
<td>0.09</td>
<td>0.4</td>
<td>0.018</td>
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<td>0.01</td>
<td>0.02</td>
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<tr>
<td>1018</td>
<td>0.16</td>
<td>0.69</td>
<td>0.006</td>
<td>0.006</td>
<td>0.22</td>
<td>0.04</td>
<td>0.08</td>
<td>0.08</td>
</tr>
<tr>
<td>1038</td>
<td>0.38</td>
<td>0.82</td>
<td>0.007</td>
<td>0.002</td>
<td>0.19</td>
<td>0.04</td>
<td>0.08</td>
<td>0.08</td>
</tr>
<tr>
<td>1541</td>
<td>0.37</td>
<td>1.41</td>
<td>0.011</td>
<td>0.018</td>
<td>0.22</td>
<td>0.16</td>
<td>0.08</td>
<td>0.09</td>
</tr>
<tr>
<td>8640</td>
<td>0.40</td>
<td>0.91</td>
<td>0.009</td>
<td>0.005</td>
<td>0.25</td>
<td>0.14</td>
<td>0.42</td>
<td>0.43</td>
</tr>
<tr>
<td>DP</td>
<td>0.088</td>
<td>1.69</td>
<td>0.009</td>
<td>0.011</td>
<td>0.62</td>
<td>0.04</td>
<td>0.07</td>
<td>0.07</td>
</tr>
</tbody>
</table>

Table 1

Chemical analysis of the testing materials (wt%).
Table 2
Total inclusion length or number for minimum severity level numbers (method A) (ASTM E45-97).

<table>
<thead>
<tr>
<th>Severity level</th>
<th>Type A</th>
<th>Type B</th>
<th>Type C</th>
<th>Type D</th>
</tr>
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<tr>
<td>0.5</td>
<td>3.7</td>
<td>1.7</td>
<td>1.8</td>
<td>1</td>
</tr>
<tr>
<td>1</td>
<td>12.7</td>
<td>7.7</td>
<td>7.6</td>
<td>4</td>
</tr>
<tr>
<td>1.5</td>
<td>26.1</td>
<td>18.4</td>
<td>17.6</td>
<td>9</td>
</tr>
<tr>
<td>2</td>
<td>43.6</td>
<td>34.3</td>
<td>32.0</td>
<td>16</td>
</tr>
</tbody>
</table>

Table 3
Inclusion width and diameter parameters (method A) (ASTM E45-97).

<table>
<thead>
<tr>
<th>Inclusion type</th>
<th>Thin series (T)</th>
<th>Heavy series (H)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Width min.</td>
<td>Width max.</td>
</tr>
<tr>
<td>Type A</td>
<td>2</td>
<td>4</td>
</tr>
<tr>
<td>Type B</td>
<td>2</td>
<td>9</td>
</tr>
<tr>
<td>Type C</td>
<td>2</td>
<td>5</td>
</tr>
<tr>
<td>Type D</td>
<td>2</td>
<td>8</td>
</tr>
</tbody>
</table>

minimum severity level number for each inclusion type and Table 3 provides the inclusion dimensions for the thin (T) and heavy (H) series according to method A.

This microscopic method places inclusions into several composition-related categories: sulfides, oxides, and silicates. The typical chemical types of inclusions in Type A are sulfide stringers, while Type C are silicate stringers. Type B are oxide stringers while Type D is a globular oxide. Method A requires length measurement of Type A inclusions, the total stringer length of Type B and C inclusions, and the number of Type D inclusions. Type D globular oxide may not exceed an aspect ratio of 5:1.

The inclusion rating test results for the six testing materials are displayed in Table 4 and the morphologies of the inclusion types are shown in Figs. 6–8.

Table 4
Inclusion rating test results for each testing material using ASTM E45-97 method A.

<table>
<thead>
<tr>
<th>AISI grade</th>
<th>A</th>
<th>B</th>
<th>C</th>
<th>D</th>
</tr>
</thead>
<tbody>
<tr>
<td>1008 steel</td>
<td>None</td>
<td>None</td>
<td>None</td>
<td>D1</td>
</tr>
<tr>
<td>1018 steel</td>
<td>None</td>
<td>None</td>
<td>C1</td>
<td>D1</td>
</tr>
<tr>
<td>1038 steel</td>
<td>None</td>
<td>A1</td>
<td>None</td>
<td>D1</td>
</tr>
<tr>
<td>1541 steel</td>
<td>A1</td>
<td>None</td>
<td>None</td>
<td>D1</td>
</tr>
<tr>
<td>8640 steel</td>
<td>A1</td>
<td>None</td>
<td>None</td>
<td>D1</td>
</tr>
<tr>
<td>DP steel</td>
<td>None</td>
<td>None</td>
<td>C2</td>
<td>D1</td>
</tr>
</tbody>
</table>

* Cleanliness tests were performed by Ivaco Rolling Mills, L’Orignal, Ontario.

Fig. 6. Inclusion type A1 in the testing materials.
Zhang and Thomas (2002) reported that the inclusion size is very important because large inclusions are the most harmful to the mechanical properties of CH materials. Sometimes a catastrophic defect is caused by a single large inclusion in an entire steel heat. Thus, producing clean steel requires not only controlling the mean inclusion content in the steel but also avoiding inclusions larger than the critical size.

2.3. Tests

The drop weight compression test (DWCT), as described by Nickoletopoulos (2000), facilitates the testing of the CHP and is capable of generating ASBs during upset testing. The DWCT machine, as illustrated in Fig. 9, consists of a tower enabling interchangeable weight plates to be dropped from heights up to 2.4 m. The die set configuration (Fig. 10) rests on a central column that is welded to the base of the DWCT machine. Specimens for CH were machined with a tolerance of 0.02 mm from as-rolled rod material to a cylindrical configuration of 5.3 mm in diameter with aspect ratios of 1.6 and 1.8. A series of tests is performed by varying the weights until internal fracture is found at different locations inside the ASB.

A guided cylindrical-pocket die set was designed for the DWCT machine. A sleeve guides the dies and reduces die movement during testing. Air vents in the sleeve prevent pressure build-up. An impact plate is used on top of the upper die to prevent direct impact with the die. Fig. 10 presents a schematic assembly of a test specimen which is placed between the upper and lower dies. This assembly rests atop the force sensor inside a cylindrical pocket.

2.4. Sample preparation

The cold-headed samples were sectioned transverse to the heading direction using a high-concentration diamond wafering blade on a slow speed (200 rpm) cut off saw that was operated with an oil coolant. Sectioned specimens were mounted in Bakelite and prepared for metallographic examination using automated techniques for grinding and polishing. Specifically, the specimens were ground with successive papers of SiC from 280 grit to 600, followed by rough polishing using a 9 μm diamond suspension with
an alcohol based lubricant on a silk cloth and final polishing with 3 and 1 μm on a nylon and porous pad, respectively. After ultrasonic cleaning, the specimens were etched by immersion in a 2% nital solution.

Microstructural examination was performed using optical light microscopy with polarizing capabilities and image analysis software to examine the overall characteristics of the ASBs and the voids and cracks initiated in the interior or along the ASBs. High-resolution imaging for examining the topography of the voids and cracks in the ASBs was performed using scanning electron microscopy operated in secondary electron (SE) mode and equipped with an X-ray detector (EDAX) for elemental analysis of secondary phases. Mapping the topography of the voids and cracks in the ASBs was performed using an Olympus OLS1200 laser scanning confocal microscope (LSCM) equipped with an argon laser (488 nm) and objective specifications of up to 100× with a 6× optical zoom. As the image produced by scanning the x- and y-axes with a spot laser beam in the LSCM originates from a shallow focus depth, for the DP steel specimens a series of precisely focused optical images were acquired along the z-axis of the microscope. The successive images were then overlapped using specialized reconstruction software to obtain a three-dimensional extended focus image of the cross-sectional view that contained both height and intensity information. Specifically, the topographic features in the extended focus image of the ASB structure enabled the characterization of the size of the width and depth of the voids and cracks.

3. Results and discussion

In general, microscopic examination and FE analysis of the DWCT specimens at various regions of each specimen during different deformation levels showed that the deformation concentrates in the ASBs (Sabih et al., 2005, 2006a,b). The microstructure of the DASBs is laminar consisting of highly elongated grain layers of ferrite and pearlite or second-phase particles (Fig. 11). Voids and cracks found along the ASBs inside DWCT specimens (Fig. 12) coincided with the cracking that occurs inside the head of the industrial bolt shown in Fig. 1. This finding agrees with that of Qiang and Bassim
Ductile fracture occurs progressively by void nucleation and growth mechanisms with increasing plastic strain. Different mechanisms of ductile failure have been reviewed in many works: Dodd and Bai (1987), Dodd and Atkins (1983) and Garrison and Moody (1987). All these reviews consider the most important factor affecting the material formability is the number and morphology of inclusions and second-phase particles.

Dodd and Atkins (1983) found in their study of flow localization in shear deformation of void containing and void-free solids that ductile failure occurs following this sequence of events:

1. Nucleation of voids occurs by decohesion at second-phase particles or at inclusion–matrix interfaces, or by particle or inclusion cracking. Dodd and Hartley (1991) concluded that in certain cases, these voids can nucleate at grain boundaries or grain boundary triple points.
2. An increase in the volume fraction of voids with increasing applied plastic strain.
3. At a critical strain, plastic deformation is localized along shear bands between the voids.

Masheshwari et al. (1978) studied quality requirements for steel CH grades and concluded that the second-phase particles and inclusions, such as heavy silicate, sulphide, alumina and globular oxide inclusions, should not exceed the acceptable limits of the cold heading material. Thomson (1990) pointed out that these particles and inclusions act as stress raisers and can support microvoid nucleation, growth, and coalescence, leading to ductile fracture.

Meyers et al. (2001) reported that the localization of the ASBs is an important deformation mode that causes ductile fracture in metals. Wei et al. (2006) concluded in their study of mechanical behavior and dynamic failure under uniaxial compression that the thermal and geometric softening and material flow patterns are responsible for dynamic failure under compressive loads.

Ductile failure caused by ASBs within the DWCT specimens under compressive loads is a complicated process which depends on the interplay of many factors. The following sections will detail the ductile failure mechanisms, the void nucleation, coalescence and crack initiation mechanism. Furthermore, the interplay of inclusion types and content, material microstructure, material flow pattern and the localization phenomenon will be discussed along with the impact of these factors on the occurrence of ductile failure.

3.1. Void initiation mechanism

A careful reading of the literature shows that there is a debate about the issue of void nucleation under a compressive state of stress. Petersen et al. (1997) reported that ductile failure is unlikely to occur under a compressive stress state. Teng et al. (2007) studied the transition from adiabatic shear banding to fracture and reported that ductile fracture under high-negative stress triaxiality cannot occur. On the other hand, many researchers have found that inclusions or inter-metallic particles serve as void nucleation sites under compressive loads: Puttick (1959) showed in his study of ductile fracture that microvoid initiation and growth is caused by the debonding of second-phase particles produced by deformation in the surrounding matrix. Cox and Low (1974) found that void nucleation and growth occurs more readily at the larger inclusions.

Rogers (1960) explained the role of the small and large particles and inclusion in the initiation and growth of voids leading to a ductile fracture. Similarly, Gurland (1972) also explained the effect of...
the inclusions on the nucleation, growth, and coalescence of microscopic voids causing fracture of ductile solids. Horstemeyer and Gokhale (1999) modeled void–crack nucleation process for ductile metals with second-phase particles. They concluded from the modeled compression, tension, and torsion experiments that the void–crack nucleation can take place under compressive loading condition.

Nemat-Nasser et al. (2001) also reported that cracks of this kind, which are produced by local defects, can be created by a local tensile stress state under an overall compressive applied stress. They also found that the occurrence of tensile stresses in the vicinity of the inclusions and/or second-phase particles create the classical conditions for void nucleation and growth. Thus, fracture initiation in ASBs under compressive loading of the DWCTs is highly probable.

The present findings are supported by previous work by Dodd and Atkins (1983) showing that void growth is sensitive to a hydrostatic stress state. Furthermore, they found that voids nucleate at second-phase particles or inclusions within the ASB under compressive hydrostatic stresses. Under compressive loading conditions with a local tensile stress state, both void nucleation and growth can occur in steels, albeit at a lower rate in comparison to the dominating failure mechanisms of void formation and propagation under tensile loading. Park and Thompson (1988) have shown that void nucleation occurs at both sides of the tensile pole of a particle under compressive loads. Horstemeyer and Gokhale (1999) reported the possible occurrence of void nucleation even under compression loads as tensile states exist for compression loadings.

3.2. Effect of material flow pattern on void nucleation

Void nucleation around second-phase particles and non-metallic inclusions is highly influenced by the nature of the material flow during deformation. Sarruf et al. (1999) reported that the metal flow pattern during the CH process supports the void nucleation mechanism at non-metallic inclusions and the surrounding material. Meyers et al. (2001) reported that the material flow pattern during shear localization is an important factor in sub-grain breakage and rotation. Tvergaard (1981) reported that this mechanism is responsible for void nucleation due to brittle cracking or decoherence of inclusions.

Detailed microscopic examination of various regions of the DWCT specimens showed that the deformation is concentrated in the localized central ASBs (CASB) with well-defined flow contours starting from the upper dead zone (UDZ) and the lower dead zone (LDZ), as illustrated in Fig. 13. Specifically, there is a sharp transition in the behavior of the material flow as observed by the contour lines which traverse from the dead zones to the CASB (Fig. 14(a)), to the CASB edge at the mid-axis (Fig. 14(b) and (c)). In the deformed region within the DWCT dies, the material flow behavior showed elongated semi-elliptical contours that increasingly concentrated
Fig. 15. Excessive deformation at the region between the UDZ, the LDZ and the CASB resulting in microscopic crack formation at the boundary between the top and bottom dead metal zones and at the inflection point of material flow (DP steel, aspect ratio of 1.6, drop height of 2.4 m and drop weight 31.6 kg).

Fig. 16. A comparison of void density inside the ASB of two specimens (DP steel, aspect ratio of 1.8 and drop height of 2.4 m) that were deformed with a drop weight of (a) 38 kg and (b) 31.5 kg. The dark heights on the surface represent the void locations and their inverted depth.

3.3. Void nucleation mechanism at the inclusions–matrix interface

Siruguet and Leblond (2004a,b) concluded that the void nucleation, growth and coalescence mechanism is highly dependent on the stress state under which deformation takes place. Fig. 17 displays possible void nucleation mechanisms under different stress states in order to have a better understanding of the ductile failure occurring inside the ASB. When the stress triaxiality is high (tensile loads/positive), voids grow in the directions of the tensile loads (Fig. 17(b) and (c)), and the enclosed inclusions do not influence their growth. Moreover, it has been reported by Pardoen and Delannay (1998) that the initially spherical voids will change to ellipsoidal voids under tensile triaxiality.

Siruguet and Leblond (2004a,b) showed that under compressive loads from all directions, or negative stress triaxiality, voids can be subjected to compression in one or several directions, thereby changing their shape. Also, if they are still in contact with the enclosed inclusions in these directions, inclusions prevent their shrinkage, a process known as void locking illustrated in Fig. 17(d). Pardoen and Delannay (1998) reported that under low-compressive stress triaxiality, the ductile matrix may flow around the inclusion resulting in the failure of the interface forming small voids at the sides of the inclusion (Fig. 17(e)). Conversely, under high-compressive loads, the fracture of the inclusion is possible (Fig. 17(f)).

In the case of shear banding under compressive loads, the formation of microvoids has been reported by Pirondi and Bonora (2003) and Sabih et al. (2005, 2006a). They indicated that the microvoids have smaller dimensions than the tensile cases. They also concluded that voids under compressive loads do not increase their dimensions but stretch and eventually coalesce by internal necking, known as microvoid sheeting (Fig. 17(g)).

Evidence of void and crack initiation and propagation at the non-metallic inclusions and stringers sites was put forward in the work of Vora and Polonis (1976). They found that the weakly bonded particles to the matrix had elongated in the direction of material flow. Once decohesion between the matrix and the inclusion or second-phase particle occurred, the nucleated voids were observed to grow with increasing strain from an initial spherical shape to an ellipsoidal form or sheet form, as illustrated in Fig. 17(h). Boyer et al. (2002) modeled void growth under different stress states and concluded that void volume increases under compressive loads if these voids are filled with inclusions.

Meyers et al. (2001) noticed that the different rates of material flow around the inclusion supports decohesion and void nucleation mechanisms between the matrix and the inclusion or the second-phase particle as shown in Fig. 17(i).

In DWC testing, the deformation takes place under compressive stress triaxiality over almost the entire specimen. The ASB occurs inside the DWCT as a result of the softening mechanisms and the shear deformation between the edges and the center of the band. The combined compressive and shearing deformation results in ductile failure inside the band. The void nucleation and coalescence follows some of the illustrated mechanisms in Figs. 17 and 18.

Fig. 18 displays some metallographic observations of the void nucleation mechanisms around the spherical inclusions (Type D). Fig. 20(b)–(f) show the steps of the void sheeting process inside the CASB of the 1541 steel. This process starts when a void nucleates at the edges of the inclusion due to the material flow around it. With more deformation, the inclusion changes its shape from spherical to ellipsoidal and the void propagates along the material flow direction forming a void sheet (Fig. 18(e)). This void sheet occurs...
Fig. 17. Illustrations of the void and crack nucleation mechanisms around the inclusions under different stress states: (a) inclusion inside the undeformed material (no void); (b) under multiaxial tensile loads, void growth in all directions; (c) under uniaxial tensile load, void nucleation in the load direction; (d) under multiaxial compressive loads, void shrinks on the inclusion (void locking); (e) under uniaxial compressive load, void nucleation at the sides of the inclusion; (f) breakage of inclusion under multiaxial compressive loads; (g) void sheet nucleation and growth at the sides of the inclusion under compressive and shear loads; (h) fractured or elongated weak inclusion along the material flow direction under combined compression and shear stresses; (i) decohesion between the inclusion and the matrix due to rotation of inclusions caused by different rates of material flow around the inclusion.

in a manner similar to the void sheeting mechanism illustrated in Fig. 17(h).

The metallographic inspection showed that some of the small voids change their shape to an ellipsoidal geometry, while the large inclusions were not affected by material deformation and continued to remain in their spherical shape. However, it was noticed that void sheets initiated at the vicinity of the big inclusions and propagated along the material flow direction (Fig. 18(g)). These findings are supported by the results of the ductile damage finite element model used by Boyer et al. (2002) to predict void growth under normal mean stress.

Due to the different rates of material flow around the inclusions, the big spherical inclusion rotates in the direction of the high-flow rate. This results in decohesion at the interface between the inclusion and the surrounding matrix. These findings are supported by similar observations made by Meyers et al. (2001), who reported that the material flow pattern during shear localization is an important factor in the breakage and rotation of inclusions.

Sabih et al. (2005, 2006a) showed in their metallographic investigations and FE simulations that at low-plastic strains (less than 1) and low-negative stress triaxiality, voids were observed to initiate around the elongated inclusions (Fig. 19(b)); Bandstra et al. (2004) reported void nucleation at the non-metallic inclusion–matrix interface at small strains. In this work, inclusions were also observed to fracture during deformation (Fig. 19(c)). According to the microscopical examination of fracture surfaces made by Broek (1973), this can cause stress concentrations promoting ductile fracture.

Type A and C (long inclusion) stringers are the favored places for void nucleation and growth within the ASBs. The metallographic study of the DWCT specimens showed that the void sheeting process took place by decohesion at the stringer–matrix interface in several steps. The void sheeting mechanism around the stringers is the result of the material flow and strain localization within the ASB. As shown in Fig. 14, the stringers move along the material flow direction and change their alignment by approximately 90°. This sharp change in the flow direction at the early stages of deformation triggers the decohesion process at the stringer–matrix interface at the boundaries of the CASB (Fig. 14(b) and (c)).

When these stringers join the localized CASB, the softening mechanisms and the high-strain rate deformation within the band support the void sheeting process along the stringers sides and around the spherical inclusions (Fig. 14(f) and (g)). Fig. 20(a) displays the sharp change in the stringers' direction which supports
Fig. 18. Changes in the inclusion morphology and void sheet nucleation mechanism around inclusions at different deformation levels in 1541 steel: (a) spherical inclusion inside the undeformed material, (b) inclusion shape changed from a spherical to an ellipsoidal form inside the DASB, (c) elongated inclusion inside the DASB, (d) void sheet initiated at the edges of the elongated inclusion, (e) void sheet growth inside the DASB, (f) a large number of void sheets inside the highly deformed TASB, (g) void sheet propagating ahead of the inclusion, and (h) rotation of the inclusion.
Fig. 19. The morphology of inclusions: (a) spherical inclusion in as-received (undeformed) DP steel, (b) elongated particles with adjacent voids initiated at low-plastic strains, and (c) fractured elliptical inclusion.

Fig. 20. Void sheet mechanism around the stringers: (a and b) 1038 steel, aspect ratio of 1.8, drop height of 2.4 m and drop weight 26.2 kg, (c) 1018 steel, aspect ratio of 1.6, drop height of 2.4 m and drop weight 28.5 kg, (d) 8640 steel, aspect ratio of 1.6, drop height of 2.4 m and drop weight 29.5 kg, and (e) 1541 steel, aspect ratio of 1.6, drop height of 2.4 m and drop weight 23.4 kg.
the void sheets inside the DWCT of 1541 steel. Usually, the voids nucleate around the stringer, forming the void sheet inside the deformed region. With further deformation, the void sheet, which is considered as another softening mechanism, propagates along the localized CASB (Fig. 20(a) and (b)).

In addition to void sheeting around the stringers within the CASB, metallographic inspection revealed that these stringers are the source of ductile failure inside the region deformed within the DWCT dies (Fig. 20(a) and (c)). In this region, the material flow pattern forces the stringers to take the shape of semi-elliptical contours that become increasingly concentrated towards the mid-axis of the specimen: Fig. 20(a), (d) and (e). The stringers and the surrounding material at these locations deform under low-compressive stress triaxiality in comparison with the center of the CASB. Under this stress state, and with further deformation, cracks initiate at the tip of the semi-elliptical stringers due to the stretch-like deformation. This stretching occurs because the flow of the semi-elliptical sides is restricted, due to the friction with the die faces, while the material continues to push the tip of the semi-elliptical stringers away from the specimen center.

3.4. Void nucleation mechanism at the grain boundaries and the second phase–matrix interface

As presented in Section 2.1, the materials tested are ferritic–pearlitic steels, except for the DP steel which has a ferrite–martensite microstructure. The pearlite grains consist of cementite lamella in a ferrite matrix. This type of microstructure makes the materials subject to a new source of void nucleation during plastic deformation under compressive loading.

![SEM micrographs display the void nucleation at the ferrite–cementite interface, at the grain boundaries and around the inclusions inside the DASB: (a) lamellar pearlite (ferrite–cementite) in the undeformed zone (b) elongated pearlite in the direction of material flow, (c) void nucleation at the ferrite–cementite interface, (d) void nucleation at the grain boundaries, (e) void nucleation at the ferrite–cementite interface and at the grain boundaries, and (f) void nucleation around an elongated inclusion (1038 steel, aspect ratio of 1.8, drop height of 2.4 m and drop weight 33 kg).](image-url)
Cox and Low (1974) referred in their study of ductile fracture of AISI 4340 steel that this failure is caused by void sheets composed of small voids nucleated at the matrix–cementite (second-phase particles) interface. Gurland (1972) studied the formation of cracks and voids at second-phase particles (carbides) under tension, compression and torsion. Hahn and Rosenfield (1975) reported that void growth and coalescence was observed the cracked inclusions and the grain boundaries. Nieh and Nix (1980) explained void growth and coalescence at grain boundaries. In conclusion, non-metallic inclusions, second-phase particles, and grain boundaries are considered to be the best candidate sites for void nucleation and ductile failure.

Tanguy et al. (2006) have also reported that the presence of second-phase particles, such as lamellar cementite or martensite, are a source of void nucleation. Chen et al. (2002) showed that lamellar cementite has a delayed deformation rate as compared to ferrite. Valiente et al. (2005) explained that the lamellar cementite of a pearlitic grain has a much higher stiffness than ferrite. Valiente et al. (2005) showed that due to microstructural aspects, the difference in the local flow behavior can lead to cementite elongation and fragmentation. This triggers void nucleation at the discontinuities produced by the breakage of cementite and by decohesion of atomic bonds at the interface between the matrix and inclusion or second-phase particles. The voids grow, assisted by plastic deformation, until they link up. Fragmentation of cementite within ASBs has also been reported by Chen et al. (2003). Argon (1976) found that voids were more easily nucleated at the larger cementite particles than at the smaller ones.

Beyond the early stages of deformation in DWCT specimens, metallographic study by SEM microscopy showed that the voids nucleated at the cementite–ferrite, the pearlite–matrix interface and at grain boundaries (Fig. 21).

Within the matrix, the cementite inside the pearlite grains far off the shear band have a lamellar shape as shown in Fig. 21(a). Lamellar cementite inside the pearlite on the shear band side is elongated along the material flow direction (Fig. 21(b)). At the high-deformation zone close to the CASB, broken lamellar cementite was noticed, and clear void nucleation took place at the ferrite–cementite interface (Fig. 21(c)).

Voids and microcracks nucleated at ferrite–pearlite interfaces close to the localized plastic deformation zone of the CASB (Fig. 21(d)). Closer to the highly localized plastic deformation within the CASB, more void nucleation and cracking along the grain boundaries along the ASB was observed (Fig. 21(e)). These voids and microcracks, along with the voids around the inclusions (Fig. 21(f)) contribute to ductile cracking along the ASBs.

3.5. Ductile failure characterization in the testing materials

The metallographic study of the DWCT specimens showed that all materials experienced ductile failure by void nucleation and coalescence, forming cracks along the CASBs. The ductile failure of each material was the result of the contribution of all the mechanisms of void nucleation at the inclusion–matrix interface, second phase–matrix interface and at the grain boundaries, as previously discussed. However, the level of contribution of each mechanism in the final ductile failure varied depending on material.

Ductile failure along the ASB inside the DWCT occurred under compressive loads. The voids and cracks found in the failed zones have the narrow, stretched, elongated shape or the void sheet shape along the highly localized regions of the ASB.

In general, it is possible to classify the ductile failure into two groups. The first group consists of materials that experienced very

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Table 5
Inclusion rating result with the minimum inclusion length or number according to the minimum severity level numbers (method A) (ASTM E45-97).

<table>
<thead>
<tr>
<th>AISI steel grade</th>
<th>A1 Min. length (mm)</th>
<th>C1 Min. length (mm)</th>
<th>C2 Min. length (mm)</th>
<th>D1 Min. number</th>
</tr>
</thead>
<tbody>
<tr>
<td>1008</td>
<td>None</td>
<td>None</td>
<td>None</td>
<td>D1 4</td>
</tr>
<tr>
<td>1018</td>
<td>None</td>
<td>C1 7.6</td>
<td>None</td>
<td>D1 4</td>
</tr>
<tr>
<td>1038</td>
<td>A1 12.7</td>
<td>C1 7.6</td>
<td>None</td>
<td>D1 4</td>
</tr>
<tr>
<td>1541</td>
<td>A1 12.7</td>
<td>C1 7.6</td>
<td>C2 32</td>
<td>D1 4</td>
</tr>
<tr>
<td>8640</td>
<td>A1 12.7</td>
<td>C1 7.6</td>
<td>C2 32</td>
<td>D1, D1H 4</td>
</tr>
<tr>
<td>DP</td>
<td>A1 12.7</td>
<td></td>
<td></td>
<td>D1 4</td>
</tr>
</tbody>
</table>

DH1 is a heavy series of Type D1 inclusion, the inclusion width range is between 8 and 14 μm. The minimum width of the other inclusion types is 2 μm.

Fig. 22. Ductile failure in DWCT specimens of 1008 steel: (a) a long crack along the CASB (aspect ratio of 1.6, and drop weight 37.6 kg), (b) void nucleation in the DASB (aspect ratio of 1.6, and drop weight 26.2 kg).
large cracks, including 1008 steel and 1018 steel. The second group consists of the material that experienced short ductile cracking along the ASB. This group includes 1038 steel, 1541 steel, 8640 steel and DP steel.

The first group is composed of soft steels with low-inclusion content. Table 5 shows that the only inclusion type in the undeformed 1008 steel is the spherical inclusions (Type D1), while the undeformed 1018 steel contains the long stringer or inclusion Type C1 in addition to Type D1 inclusions.

Both steels experienced very long cracks which extended for more than 3 mm along the mid-axis of the CASB, as shown in Figs. 22(a) and 23(a) for 1008 steel and 1018 steel, respectively. The SEM metallographic inspection of the cracked specimens revealed that void nucleation and coalescence took place at the boundaries of the highly elongated grain within the CASB in contribution to the microcracks at the ferrite–cementite interface. The metallographic investigation of the role of the inclusions type in the 1008 steel (inclusion Type D1) showed that voids nucleated along the inclusions leading to shorter cracks in comparison to the crack initiated at the grain boundaries (Fig. 22(b)).

Similar conclusions can be made with respect to the 1018 steel DWCT specimens. The metallographic investigation showed that all the void nucleation mechanisms contributed to the ductile cracks found within the highly localized deformed CASB in the DWCT specimens made of 1018 steel (Fig. 23). Void nucleation took place around the inclusions (Fig. 23(b)), and at the interface between both types of inclusions (Type C1 and Type D1) and the matrix. The microscopic observation revealed that at the early stages of deformation, void sheets and short cracks extend along the grain boundaries in addition to microcracks at the ferrite–cementite interface. With further deformation, the coalescence of these void sheets and cracks form bigger cracks that extend over 3 mm along the center of the CASB (Fig. 23(a)). The void nucleation along the Type C1 inclusion is readily apparent in Fig. 20(c).

The materials in the second group can be characterized by the occurrence of the TASBs that are accompanied with short cracks. These cracks were noticed to be within, or at, the sides of the TASB zone and the DASB zone of the band. As in the first group, all of the void nucleation mechanisms contributed to the ductile cracks found within the highly localized deformed and the transformed part of the CASB. However, the long inclusions (Type A1, C1 and C2) played an important role in ductile failure initiation as previously discussed. The presence of inclusion type C2 in 1541 steel, 8640 steel and DP steel means that stringers with a total minimum length of 32 mm are present in the matrix. As discussed previously, this long inclusion is the favored location of void nucleation by the decohesion mechanism at the inclusion–matrix interface. With further straining inside the localized ASB, these voids coalesce around the stretched long inclusion inside the band. Figs. 20(a), 20(b) and 24(a) display the long cracks which initiated within the CASB of the 1038
Fig. 24. Crack initiation along the TASB: (a) 1038 steel, aspect ratio of 1.8, drop height of 2.4 m and drop weight 33 kg, (b) 8640 steel, aspect ratio of 1.8, drop height of 2.4 m and drop weight 29.5 kg.

Fig. 25. Void coalescence in the ASB region of deformed DWCT specimen of DP steel: (a) 500× and (b) 1000× (aspect ratio of 1.8, drop height of 2.4 m and drop weight 35.6 kg).

Careful investigation of the microstructure at the side of the CASB of the second group materials reveals that the elongated ferrite and pearlite grains also helped in the nucleation of narrow and stretched void sheets along the localized band. Nevertheless, the role of the long inclusion clearly plays a much larger part in ductile failure.

4. Conclusion

The occurrence of internal ductile failure in cold-headed products presents a major obstacle in the fast expanding CH industry. The current design rules of thumb failed in avoiding this type of failure which results in reducing the material’s capability to carry loads. This internal failure may lead to catastrophic brittle fracture under tensile loads despite the ductile nature of the material.

The lack of specialized studies regarding the mechanisms of internal ductile failure in cold-headed products and the failure of the current CH design procedures to fulfill the needs of new developments in the expanding CH industry motivated this research. A comprehensive testing and investigation methodology was
used to reveal the complicated interplay of process and material parameters contributing to the initiation and propagation of the internal ductile failure in six CH quality AISI steel grades.

The metallurgical and microscopic studies showed that internal ductile failure occurs progressively by void nucleation and growth mechanisms with increasing plastic strain inside the highly localized CASBs. The most important factor affecting material formability is the number and morphology of any inclusions and second-phase particles. Moreover, the thermal and geometric softening mechanisms and material flow pattern are responsible for dynamic failure under compressive loads.

It was confirmed that ductile failure occurs by a void nucleation mechanism which occurs by decohesion at second-phase particles and/or at inclusion–matrix interfaces, by particle or inclusion cracking, or at grain boundaries. The localization process inside the CASBs supported the void growth and coalescence along the band forming narrow void sheets. Under compressive loading conditions, the void nucleation, growth and coalescence mechanism occurred due to the local tensile stress state in the vicinity of the inclusions and/or second-phase particles.

The metallurgical investigations showed that the nature of the metal flow pattern during the CH process, and the different rates of metal flow around the inclusions, support decohesion and void nucleation mechanisms between the matrix and the inclusion or the second-phase particle in the highly localize CASBs.

The void sheeting process starts when a void nucleates at the edge of the inclusion due to the material flow around it. With more deformation, the inclusion changes its shape from spherical to elliptoidal and the void propagates along the material flow direction forming a void sheet.

Under combined compressive and shearing deformation, the large inclusions continued to retain their spherical shape. However, it was noticed that void sheets initiated in the vicinity of the large inclusions and propagated along the material flow direction. Type A and C long inclusions are the favored places for void nucleation and growth within the CASBs. The void sheeting process took place by decohesion at the stringer–matrix interface in several steps. The sharp change in the flow direction at the early stages of deformation triggered the decohesion process at the stringer–matrix interface. The thermal softening and the high-strain rate deformation within the localizing band support the void sheeting process along the stringers’ sides and around the spherical inclusions. With further deformation, the void sheet propagates along the localized CASB.

In addition to non-metallic inclusions, the second-phase particles and grain boundaries are considered to be the best candidate sites for void nucleation and ductile failure. The difference in the local flow behavior, due to microstructural aspects, can lead to cementite elongation and fragmentation. This triggers void nucleation at the discontinuities produced by the breakage of cementite and by decohesion at the interface between the matrix and inclusion or second-phase particles.

The microstructure of the DASBs is laminar, consisting of narrow, stretched, and highly elongated grain layers of ferrite and pearlite or second-phase particles. Void sheets and cracks are found in the failed zones along the ASBs.

All materials tested experienced ductile failure by void nucleation and coalescence, forming cracks along the ASBs. The ductile failure of each testing material was the result of the contribution of all the mechanisms of void nucleation at the inclusion–matrix interface, second-phase–matrix interface and at the grain boundaries. However, the level of contribution of each mechanism in the final ductile failure varied depending on the material.

In summary, a comprehensive experimental and metallurgical study of the ASB phenomenon and the associated ductile failure were introduced in this study to uncover the complex interplay of different material and process parameters controlling the failure mechanisms within the ASBs in cold heading process.

The ductile failure initiation under compressive loads is a debated subject of many works in the literature. However, this detailed study reached important findings that uncovered the complex interplay of the different mechanisms of ductile failure initiation and growth under compressive loads. All these findings were based on detailed supporting experimental and metallurgical evidences about these mechanisms.

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