1. Introduction

Cold headability is the capability of a cylindrical piece of material to be shaped into the head of a bolt, screw, or other cold-formed part without cracking. This property is material dependent and can be influenced by many factors such as chemical composition, surface condition, and microstructure. The effect of microstructure upon the cold headability of a medium carbon steel (1036M) was investigated. Six different microstructures were produced by various heat treatments. Drop weight tower (DWT) testing methods, previously developed at McGill University, were used on samples of these materials. Visual inspection, metallographic and SEM analysis were carried out to detect cracks on the surfaces of tested samples and to identify their causes. The axial and circumferential strains of the tested samples were measured and the strains at which cracks were initiated were used to assess the headability. The cold headability was found to be particularly sensitive to the microstructure and was greatest in the completely spheroidized structures. This indicates that DWT testing is a valid method for evaluating the cold headability of metallic materials.

KEY WORDS: drop weight testing; cold headability; medium carbon steel.

2. Experimental

2.1. Material

In order to eliminate the effect of chemical composition, a single steel was used in this project. The material selected was a medium-carbon grade 1036M steel, commonly used for the cold heading of fasteners; the chemical composition is listed in Table 1.

Cold headability can be affected by many factors including the chemical composition, surface condition and microstructure of the material. In the present work, a medium carbon Cold Heading Quality (CHQ) steel, grade 1036M, was investigated and the microstructure varied using different heat treatments. The specific objectives were to verify the validity of the DWT to distinguish between different microstructures, as well as to determine which microstructure had the greatest cold headability.

| Table 1. Chemical composition of 1036M CHQ steel (mass%). |
|----------------------------------------|
| C | Cu | P | S | Si | Ca | Ni | Cr | Mo | Sn | Al | N |
|---|----|---|---|----|----|----|----|----|----|----|---|
| 0.24 | 0.04 | 0.009 | 0.015 | 0.24 | 0.2 | 0.09 | 0.25 | 0.30 | 0.012 | 0.004 | 0.0065 |
Two as-received materials were investigated: one was in the as-hot rolled state (HR), the other in a basic heat treated state (HT). The latter had been annealed at 745°C for 4 h, then cooled to 500°C in 5 h. As cold headability can be affected by the surface condition of the material, all samples were smoothed by machining prior to testing.

2.2. Heat Treatment

To produce additional microstructures for investigation, four other material states (HT-I, II, III, IV) were produced, as summarized in Table 2. HT-I was designed to yield a completely spheroidized structure, whereas the HT-II, III and IV treatments were designed to partially spheroidize the materials to varying degrees. All samples were annealed in a standard muffle furnace using a natural atmosphere.

The microstructures of each material type were determined using standard metallographic techniques. The volume fraction of pearlite in each sample was measured using an image analysis (CLEMEXTM) system. The Vickers microhardnesses (at a load of 50 g) of both the ferrite and pearlite colonies in each material were obtained by averaging measurements from multiple points on undeformed samples.

2.3. Drop Weight Testing (DWT)

The DWT samples were of cylindrical shape with a diameter of 5.2 mm. An aspect ratio of 1.24 was chosen for all the microstructures, whilst ratios of 1.0 and 1.6 were also employed for the HR and HT materials.

The configuration of the DWT machine is shown in Fig. 1. Varying the drop height causes a change in strain rate; thus the height was kept constant and the masses were varied. Previous work revealed that a drop height of 1.5 m was suitable for fracturing a similar medium carbon (1038) steel. Therefore, this height was used and the drop weight was increased from 11.35 kg in increments of 2.3 kg until the sample failed. Once a rough weight limit was established, the weight was decreased at intervals of 0.6 kg to determine more exactly the minimum weight at which cracking occurred.

2.4. Fracture Evaluation

The cold headability was evaluated in terms of the surface strain at which the first crack was initiated. This strain was determined by measuring the distance between grid lines on the surface of the sample, as shown in Fig. 2; it was then calculated using the following relationship:

\[ \varepsilon_a = \ln \frac{H}{H_0} \]

\[ \varepsilon_a = \ln \frac{w}{w_0} = \ln \frac{D}{D_0} \]

Note that determination of whether a sample had cracked or not is a subjective measurement; there is no industry standard available to evaluate the surface quality of cold headed parts. Thus, a stereomicroscope was employed to inspect these surface cracks and the presence of a ghost line crack at a magnification of ×25 was used to define the initiation of failure in a tested sample. The impact energy was also calculated from the drop weight and height.

A JEOL JSM-840A scanning electron microscope was used to further analyse the cracks and fracture surfaces of tested samples. Energy Dispersive Spectrum (EDS) analysis was employed to qualitatively determine the chemical composition of the particles observed on these fracture surfaces.

3. Results and Discussion

3.1. Microstructures

The microstructures of the 1036M steel in the two as-received states are shown in Figs. 3(a) and 3(b). The as-hot rolled (HR) state, which was rapidly cooled on the STELMORTM conveyor, consisted of fine lamellar pearlite and ferrite, whilst the heat treated (HT) material had a slightly coarsened structure.

The laboratory heat-treated microstructures are shown in Figs. 4(a) to 4(d). HT-I consisted of equiaxed pearlite and ferrite, as the hot rolled pearlite was transformed into austenite above the Ac₁ temperature (approximately...
720°C), and then transformed back into pearlite during cooling.

HT-II consisted of spheroidized cementite distributed in a ferrite matrix. Because annealing was carried out below the Ac1 temperature (715°C), no phase transformation occurred, but due to the elevated temperature, lamellar cementite in the pearlite was converted into spheroidal cementite. HT-III and HT-IV, annealed for shorter times than HT-II, were partially spheroidized to lesser degrees.

### 3.2. Drop Weight Test Results

The minimum strain at which cracking initiated was taken as the fracture strain and these are listed in Table 3 for all six microstructures at an aspect ratio of 1.24.

#### Table 3. DWT fracture strains and critical impact energies for the various structures.

|          | Aspect Ratio | Axial Strain | Circumferential Strain | Critical Impact Energy (J) |
|----------|--------------|--------------|------------------------|---------------------------|
| HR       | 1.24         | 1.52 - 1.70  | 0.77 - 0.88            | 423                       |
| HT       | 1.24         | 1.59 - 1.63  | 0.82 - 0.84            | 354                       |
| HT-I     | 1.24         | 1.75 - 1.89  | 0.89 - 0.93            | 567                       |
| HT-II    | 1.24         | 1.84 - 1.95  | 0.89 - 0.99            | 592                       |
| HT-III   | 1.24         | 1.76 - 1.81  | 0.88 - 0.93            | 480                       |
| HT-IV    | 1.24         | 1.58 - 1.84  | 0.8 - 0.94             | 440                       |

**Fig. 3.** Microstructures of 1036M steel, ×2000.

**Fig. 4.** Microstructures of the laboratory heat treated steels, ×2000.
lower limit was defined as the strain at which the first crack appeared, whereas the upper limit was the maximum strain at which no crack was found.

It can be seen that the as-hot rolled material had the lowest fracture strains, whilst the HT microstructure had the lowest impact fracture energy, although the fracture strain was not significantly greater than that for the HR material. This indicates that the industrial (HT) heat treatment significantly lowered the toughness of the material, without significantly improving its formability. The HT-II material had the highest fracture strains and the highest impact energy among the six tested materials. This indicates that the industrial (HT) heat treatment significantly lowered the toughness of the material, without significantly improving its formability. The HT-II material had the lowest fracture strains and the highest impact energy among the six tested materials. This demonstrates that the complete spheroidization heat treatment best improves the cold headability. HT-I and HT-III had similar fracture strains, but the HT-I impact energy was higher than that of HT-III. This indicates that the complete spheroidized microstructure has a lower deformation resistance than a microstructure containing lamellar pearlite.

The upper and lower limits of cracking were used to define the possibility of failure, as shown in Fig. 5. Strains below the lower limit were deemed to be in a "safe" zone, whilst strains above the upper limit were in a "fracture zone" and these limits were separated by a "danger zone". It is evident that the larger the safe zone, the better the cold headability of the material. Also, the HR microstructure, which contains large amounts of lamellar pearlite, has the smallest safe zone, whereas HT-II has the largest.

The DWT results for the HR and HT samples tested using different aspect ratios are shown in Fig. 6. In general, the fracture strain increased with increasing aspect ratio. The greater the aspect ratio in a heading operation, the lower the stress concentration at the free surface and hence the greater the cold headability. This is in agreement with the theory of workability limit diagrams. Therefore, comparisons of the cold headability should be conducted on specimens of similar aspect ratio.

### 3.3. Effect of Microstructure

The tested microstructures consisted of a two-phase mixture of ferrite and pearlite or spheroidized cementite. The volume fraction of each phase in the mixture influences the strength and formability. In general, the hard phase (pearlite) contributes to the material's strength and hardness, whereas the "soft" phase (ferrite or spheroidized cementite) contributes ductility and toughness. Therefore, decreasing the pearlite fraction will improve the cold headability. The volume fractions of pearlite in the six materials are listed in Table 4; the HR structure had the highest content and HT-II the lowest.

Goods and Brown suggested a relationship between the critical strain to nucleate a void and the volume fraction of second phase particles:

$$\varepsilon_{cr} = \frac{1}{b} \left( \frac{\sigma_i - \sigma_R - \sigma_0}{\mu} \right)^2 \left( 1 + 3 f_v + \frac{\varepsilon_{pl}^2}{1.8} \right)^2 \ldots (3)$$

where $\varepsilon_{cr}$ is the critical strain to nucleate a void, $\sigma_i$ the interfacial strength, $b$ the Burgers vector, $\sigma_R$ the Orowan stress, $f_v$ the particle volume fraction, $\sigma_0$ the hydrostatic tension, $r$ the particle radius and $\mu$ the shear modulus. It is evident from this relationship that increasing the second phase volume fraction ($f_v$) results in a decrease in the critical strain required to nucleate a void ($\varepsilon_{cr}$).

Figure 7 shows the relationship between the fracture strain and the volume fraction of pearlite. As the pearlite fraction was decreased, the fracture strains (axial and circumferential) increased. This is because the interfacial areas of the two phases are the regions of high stress concentration. Void nucleation occurs at the sites of high stress.

| HR  | HT  | HT-I | HT-II | HT-III | HT-IV |
|-----|-----|------|-------|--------|-------|
| 78  | 63  | 48   | 43    | 52     | 75    |

| Fraction of Pearlite (%) | Axial Strain | Circumferential Strain |
|--------------------------|--------------|------------------------|
| 35                       | 45           | 55                     | 65          | 75    | 85      |

Figure 7. Fracture strain vs. volume fraction of pearlite.
concentration. Raising the volume fraction of pearlite, therefore, results in an increase in the density of void nucleation sites and hence diminishes the cold headability.

The stress concentration at the interface during deformation results from the difference in mechanical properties between the two phases. When this concentration exceeds a critical value, voids nucleate on these interfaces or at second phase particles. Thus, the smaller the difference in ductility between the two phases, the better the cold headability.

In order to determine the difference in strength, and hence ductility, between the two phases, microhardness analysis was used and the results are illustrated in Fig. 8(a). Note that some uncertainty may be involved in these values, because the dimensions of some of the spheroidal cementite particles were sometimes below the resolution of the test indenter. HR, which was rapidly cooled, had the highest hardnesses of both pearlite and ferrite, as well as the largest disparity between them. After heat treatment, the microhardnesses of both phases decreased to different degrees, as did the disparity values. The HT-II material had the smallest microhardness difference between phases.

As expected, the fracture strain decreased with an increase in the microhardness disparity, as shown in Fig. 8(b). It is evident that higher disparity values result in higher stress concentrations at the phase interfaces and therefore an increase in the probability of nucleating microvoids and fractures.

The sizes and shapes of second phase particles are also factors that affect the characteristics of fracture. Fisher and Gurland reported that voids in spheroidized carbon steel

![Fig. 8.](image)

(a) Microhardness differences between undeformed pearlite and ferrite; (b) relationship between microhardness difference and fracture strain.

![Fig. 9.](image)

SEM micrographs of heat treated steels, ×2000.

(a) HT-II  
(b) HT-III  
(c) HT-IV
were generally associated with particles whose dimensions were of greater than average size and that they rarely formed at very small particles. Non-equiaxed or irregularly shaped cementite particles were often subject to internal fractures. Particles situated at ferrite grain boundaries were favored sites for the nucleation of voids. The fracture stress of a second phase particle may be calculated using an energy balance approach. It has been assumed that when the elastic energy stored in a second phase particle exceeds the surface energy for a newly formed void, fracture can occur at the particle. The relationship between the critical stress for particle cracking (σc) and particle size (d) was provided by Van Stone et al.11)

$$\sigma_c = \left( \frac{6\gamma E}{q^2 d} \right)^{1/2}$$ .................................(4)

where E is the particle Young’s modulus, γ the surface energy of the particle and q the particle stress concentration factor.

Trufiakov et al.13) revealed that the fracture toughness of two-phase steels was determined by the volume fraction and particle size of the second phase component. They gave the relationship between the crack resistance and the second phase fraction and size as the following:

$$\delta = \frac{A}{V_{sp}\sqrt{h_{sp}}}$$ .................................(5)

where δ is the crack resistance, Vsp the volume fraction of second phase, hsp the size of the second phase and A is a constant.

From the above discussion, it is evident that increasing the dimensions of the second phase will decrease the critical stress required to cause particle cracking, as well as the crack resistance and so deteriorate the formability.

The HT-II, III and IV materials, which were annealed at the same temperature but for different holding times, varied in the size distribution of the carbides present, as can be seen in the SEM images of Fig. 9. The pearlite was spheroidized to different degrees in these microstructures. Here the “spheroidization ratio” is defined as the ratio of the number of spheroidal cementite particles to the total number of cementite particles including those of lamellar form.14)

Several researchers14,15) have employed the “aspect ratio” of the cementite to evaluate the degree of spheroidization. The aspect ratio is defined as the length ratio of the long axis to the short axis of a cementite particle. When this ratio is less than a critical value, the particle is deemed to be spheroidized. Das et al.14) chose 2 as this criterion, whilst Chattopadhay and Sellars15) chose 5.

Such a quantitative analysis of spheroidization was not carried out in the present work. Nevertheless, it is qualitatively evident that the aspect ratios of almost all the cementite particles in the HT-II and III materials are less than 5, as shown in Figs. 9(a) and 9(b), which indicate that complete or nearly complete spheroidization took place. The aspect ratios of most of the cementite particles are greater than 5 in the HT-IV material, Fig. 9(c). Because the second phase aspect ratio was the highest, the headability of the HT-IV material was the lowest of the four steels heat treated in the laboratory.

3.4. Crack Analysis

In the present work, the samples tested by DWT displayed both longitudinal and shear cracks at the point of initiation. The first appearance of cracking on a sample of each of the tested materials is illustrated in Fig. 10. These cracks appear to be aligned longitudinally on the barrel surface. However, in the less ductile materials, these “longitudinal” cracks are seen to consist of segments of shear cracks, Fig. 11.

Such widely opened fractures occurred in the HR, HT and HT-IV materials, even at moderate strains. These all contained non-equiaxed pearlite. No widely opened cracks were found in the HT-II and III grades. A widely opened fracture on an HR sample is illustrated in Fig. 11(a). The fracture edges are widely separated on the barrel surface and several internal cracks of a shear nature extend from the root of the fracture along the different directions. Figure 11(b) shows a widely opened fracture on another HR sample with similar crack morphology and again the shear character of the internal cracks is evident.

A similar fracture on the surface of an HT-I sample is...
displayed in Fig. 11(c). This fracture was widely opened on the barrel surface and a zigzag (shear) crack can be seen to characterize the root of the fracture.

SEM observations of the fracture surfaces revealed large amounts of dimple-like depressions separated by ridges that are the remnants of ligaments. The dimples, which were distributed randomly, were mainly equiaxed and of different sizes. Fig. 12(a) illustrates some of the dimples on the fracture surface of an HR sample. Some particles are observed to be lodged in the bases of the dimples and most of these are elongated.

The presence of dimples on the fracture surfaces indicates that the failures of the DWT samples took place by ductile fracture controlled by a microvoid coalescence mechanism. The microvoids were initiated at the second phase particles or at interfaces between the matrix and the particles as a result of the stress concentrations discussed above. These second phase particles can be inclusions, particles of spheroidal cementite, or an entire pearlite colony, where the latter consists of alternate cementite and ferrite platelets. The dimples were formed as a result of the growth and coalescence of the microvoids.

Second phase particles play important roles in fracture development. They lead to the initiation of fracture because of the stress concentrations caused by the differences in ductility between the inclusions and the matrix material. Particles were often found at the bottom of dimples, such as those of the HR sample in Fig. 12(a). These were analysed

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**Fig. 11.** Widely opened fractures on HR samples (aspect ratios 1.24, ×25).

**Fig. 12.** Cracks (arrowed) on DWT samples, ×25.
The large particles fail by cracking while particles, then grow and coalesce during further flow of the broken by the crack that penetrates through the colony. The pearlite colony, located at the root of the fracture, was pearlite colony on the fracture surface of an HR sample. A small crack could be seen, at low magnification, on the free surface of the sample and the crack was quite shallow. It was evident that this was initiated just below the surface and, at higher magnification, the fracture was seen to originate at the surface of a sample can act as a surface defect and then initiate a surface crack in this way.

A pearlite colony can also act as a stress concentration site and cause fracture during DWT. Figure 13(b) displays a pearlite colony on the fracture surface of an HR sample. The pearlite colony, located at the root of the fracture, was broken by the crack that penetrates through the colony. During plastic deformation, voids nucleate at the harder particles, then grow and coalesce during further flow of the matrix metal.\(^6\)\(^,\)\(^11\) The large particles fail by cracking while the smaller ones fail by interfacial decohesion.\(^11\) That is why unbroken small particles were present at the fracture surface, while the pearlite colonies were cracked.

The growth and coalescence of voids depend on the flow instability of the matrix metal. During stable plastic flow, there is little potential for ductile fracture. Only when the tensile or shear instability condition of the matrix is reached does ductile fracture progress rapidly. The strain in the matrix from the moment of instability to fracture depends primarily on the proportion of second phase particles.\(^6\)\(^,\)\(^9\) The greater the second phase fraction, the lower the fracture strain. The HR microstructure had the highest proportion of the second phase; therefore it displayed the lowest fracture strains. By contrast, the HT-II material had the lowest fraction of the second phase and the highest fracture strain.

Although a few sulphide particles were found on some of the fracture surfaces, EDS analysis revealed that most of the second phase particles responsible for fracture in the DWT tests were carbides. In spheroidized microstructures, such as the HT-II and HT-III, the dimensions of the carbide particles were small as were their aspect ratios. This resulted in lower stress concentrations and higher ductility of the matrix. In relatively un-spheroidized microstructures, such as those of the HR, HT and HT-IV samples, large amounts of the cementite were lamellar and of uneven dimensions. The platelet colonies caused high stress concentrations during DWT; this resulted in early fracture. This is also the reason why microstructures containing spheroidal cementite possess better cold headability than those containing pearlite.

4. Conclusions

1) The drop weight test is able to differentiate between the cold headability of samples containing different microstructures. Therefore it is a valid method for assessing the suitability of materials for cold heading.

2) The cold headability of medium carbon steel is sensitive to differences between the mechanical properties of the ferrite and the pearlite colonies. It can therefore be improved by reducing this disparity by means of heat treatment.

3) Cold headability is also sensitive to the inclusion and second phase particle content of the microstructure. When such particles are present near the external surface of a material subjected to cold heading, they act as surface defects and lead to the nucleation of cracks.

4) According to the present results, a totally spheroidized microstructure (the HT-II) has the best cold headability amongst the six tested microstructures.

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