Effect of Microstructure and Texture on the Edge Formability of Light Gauge Strip Steel

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Twin roll casting (TRC) of low-carbon ultra-thin cast strip (UCS) steel has the potential to deliver significant economic advantages over steels made by conventional processes. Novel microstructures containing intragranularly nucleated acicular ferrite may be produced in UCS products manufactured by the CASTRIP® process, however, their mechanical property performance has yet to be fully evaluated. This study compared the mechanical properties and edge formability of UCS steel produced by the CASTRIP process to steel produced by conventional hot rolled (HR) and cold rolled and continuously annealed (CR & CA) process routes. The results revealed an apparent paradox, where recent UCS steel produced by the CASTRIP process was shown to have lower total tensile elongation values, yet higher edge formability than products manufactured by the conventional process routes. The discrepancy was largely attributable to the influence of acicular ferrite microstructure on steel plasticity. Acicular ferrite decreased macro-plasticity by reducing the contribution of yield point elongation (lower volume fraction of pro-eutectoid ferrite) and post uniform elongation (strain localisation to softer pro-eutectoid ferrite) components of the total tensile elongation value. Conversely, the high homogeneity of the acicular ferrite microstructure enhanced micro-plasticity of the UCS steel produced by improving resistance to void initiation and propagation during forming. This study also showed that plastic anisotropy can significantly influence the edge formability of low-carbon sheet steel, with failure occurring along the direction where the resistance to through thickness thinning was the least.

KEY WORDS: acicular ferrite; hole expansion test; normal anisotropy; CASTRIP® technology; ultra-thin cast strip; tensile properties; edge formability.

1. Introduction

Direct strip casting (DSC) of low-carbon light gauge sheet steel 1 mm to 2 mm in thickness has the potential to deliver significant economic advantages over conventionally produced steels. DSC reduces the number of processing steps required to turn liquid steel into strip, and thereby reduces capital, energy and operational requirements. There are two competing DSC processes: Twin roll casting (TRC); and single roll belt casting. TRC has gained momentum as the more favoured method of DSC low-carbon steel with the commencement of commercial operations at the CASTRIP® facility at Nucor Steel’s Crawfordsville Indiana plant1) and the recent commissioning of a second CASTRIP® facility by Nucor Steel at its Arkansas plant. The CASTRIP process, similar to all twin roll casting operations, uses two counter-rotating rolls to create two individual shells that are formed into a continuous sheet at the roll nip. The solidification fundamentals are substantially different to current continuous casting processes, with the CASTRIP process operating in a regime of heat transfer that is an order of magnitude higher than conventional slab and thin-slab casting processes.1) Further details regarding the production of plain low-carbon ultra-thin cast strip (UCS) steel sheet produced by the CASTRIP process can be found elsewhere.1–5)

Integrated steel plants produce 200–250 mm thick continuously cast slabs, requiring high total reductions and multi-pass rolling to produce light gauge steel strip 1.5–2.0 mm thick. Repeated recrystallisation during rolling refines the austenite grain size which, in combination with accelerated cooling on the runout table promotes high ferrite nucleation rates and a resultant fine ferrite grain size.6,7) Accordingly, one of the main strengthening mechanisms for conventional low- carbon steel sheet is ferrite grain size refinement. In contrast, the relatively limited inline hot reduction available with the CASTRIP DSC process restricts austenite grain refinement, resulting in relatively large austenite grains before transformation. The large austenite grains however can be combined with accelerated cooling on the runout table, and an appropriate type, size, and dispersion of non-

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metallic inclusions to promote particle-stimulated nucleation of acicular ferrite. The size range and composition of the inclusions in the UCS produced by the CASTRIP process, that can induce acicular ferrite formation, can now be optimally engineered by virtue of careful control of the dissolved oxygen levels at the start of solidification, other alloying additions and processing parameters. The Mn–Si deoxidised steelmaking practice used in making UCS by the CASTRIP process provides the necessary flexibility for setting appropriate dissolved oxygen levels. The final microstructure produced may consist of large irregular grain boundary ferrite, some Widmanstätten ferrite and acicular ferrite. The microstructure produced is considerably different from the fine grained polygonal ferrite and pearlite constituents found in conventional hot and cold rolled steel products. Acicular ferrite has a fine interlocking structure which is recognised as being beneficial to both strength and toughness properties. However, to gain wide market acceptance for low-carbon steel manufactured by DSC processes, the formability characteristics of these novel microstructures and resulting sheet steel products need to be further understood.

Many products manufactured from light gauge low-carbon strip produced by conventional strip production processes fail during forming by localised necking or cracking at a sheared edge. The edge formability of the steel is thus of considerable practical importance and can conveniently be measured by the hole expansion test. In this test a hole is made, usually by a punch, and expanded by a conical dome until a through thickness crack is observed. Various metallurgical factors have been associated with improved edge formability. Correlations have been made with various parameters derived from uni-axial tensile tests. A recent review of the published data on these correlations found the tensile strength and total elongation are the most important tensile test parameters influencing edge formability. In addition, the degree of local cold work generated in the punched hole edge region can influence the edge formability. Microstructural features such as the presence of elongated manganese sulphide (MnS) inclusions, microstructural banding and the relative hardness of microstructural constituents have also been implicated in reducing the edge formability of steel. In view of these factors, fine and homogeneous microstructures such as those composed largely of tempered martensite and bainite have been found to have superior edge formability. Low-carbon sheet steel containing significant proportions of acicular ferrite, such as that which can be produced in UCS steel, therefore might also be expected to have improved edge formability due to the high microstructural homogeneity of the phase. Nonetheless, the edge formability of low-carbon steel sheet containing acicular ferrite has hitherto not been fully evaluated.

The aim of this study was to determine the mechanical properties, in particular the edge formability, of UCS steel products that can be produced by the CASTRIP process. The results are compared to conventionally processed strip steels of similar strength and strip thickness produced by conventional hot rolling (HR) and cold rolling and continuously annealing (CR & CA) process routes. The metallurgical factors that influence edge formability are discussed.

2. Experimental Procedure

2.1. Chemical Composition and Steel Processing Conditions

Seven commercial quality sheet steel grades were used in this investigation. These were: two ultra-thin cast strip (UCS) steel grades produced at Nucor Steel’s CASTRIP plant in Crawfordsville Indiana, three conventional hot rolled (HR) and two conventional cold rolled and continuously annealed (CR & CA) steel grades produced at BlueScope Steel’s integrated Port Kembla Steelworks. Figure 1(a) summarises the thermo-mechanical history for the steels used in this investigation.

A minimum of five coils of each steel type were produced and sampled for testing. To avoid any potential strip edge region effects the test coupons for all the hole expansion tests and mechanical property assessments were taken from the central region of the strip width. The hole expansion ratio is thought to be influenced by sample thickness. To reduce any thickness related effects, the gauge was restricted to 1.5 ± 0.05 mm. An expectation was made for the comparative grade 300 CR & CA (thickness = 1.2 mm) which was not usually manufactured at a 1.5 mm gauge. Additionally, all specimens used had comparable surface roughness measurements.

The typical chemical compositions of the steels investigated are shown in Table 1. UCS steels (275 UCS and 380 UCS) were manufactured by a silicon/manganese deoxidation steelmaking practice. The higher strength of the 380 UCS steel was achieved through a lower cooling temperature (520°C compared to 650°C) and higher manganese content. Comparative HR and CR & CA steels were produced from two conventional aluminium killed, plain carbon-manganese compositions. The first was a low-carbon, low manganese steel with aim 0.05 wt% carbon and

![Fig. 1. Schematic representation of (a) the thermo-mechanical history of the three processing routes investigated: ultra-thin cast strip (UCS), hot rolled (HR), cold rolled and continuously annealed (CR&CA); (b) normalising heat treatment and (c) spheroidising annealing heat treatment.](image)
0.25 wt% manganese (200 HR, and 300 CR & CA), while the second had an aim of 0.15 wt% carbon and 0.75 wt% manganese (300 HR, 360 HR and 350 CR & CA). The 360 HR steel was produced from the later composition by using a lower cooling temperature.

Additional heat treatments were applied to separate out the effect of microstructural constituents and non-metallic inclusions on the edge formability, Figs. 1(b) and 1(c). Normalising heat treatments at 950°C for 20 minutes were applied to the UCS steel to change the microstructure from acicular ferrite to ferrite and pearlite. Also, spheroidising heat treatments of 700°C for 5 hours were applied to spheroidise the existing iron carbide phase of the microstructure of the CR and CA steels.

2.2. Hole Expansion Tests

Hole expansion tests were conducted in accordance with ISO16630.23) A 10 mm hole was made in a 150×150 mm sheet of the material using a punch. A 60° diameter conical shaped die was then used to expand the hole until a through thickness crack was observed. During expansion the edges were clamped. No lubricant was applied to the specimen. Figure 2 shows a schematic representation of the process. The hole expansion ratio \( \lambda_p \) was taken as the change in hole diameter divided by the original hole diameter.

\[
\lambda_p = \frac{d_f - d_i}{d_i} \times 100\% \quad \text{(1)}
\]

where \( d_i \) and \( d_f \) are the initial and final diameters of the punched hole.

Three hole expansion tests were conducted per coil and the results averaged. Measurements of the initial and final hole diameters were made in two perpendicular (longitudinal and transverse to rolling) directions. The location of the through-thickness cracks around the circumference of the expanded hole was also recorded relative to the rolling direction of the strip. The maximum strain occurs at the edge of the hole and to ensure reproducibility the burr was always facing away from the tool, placing it on the outside of the edge where the strain was greatest.

2.3. Mechanical Property Testing

Mechanical properties were measured on a screw driven hydraulic testing machine using standard dog bone shaped tensile specimens with an 80 mm gauge length. A non-contact video extensometer system measured longitudinal and lateral strain. Two cross-head speeds were used. An initial cross-head speed of 0.1 mm/s was used until 5% strain had been reached. The cross-head speed was then increased to 1 mm/s for the rest of the test. Mechanical property measurements were made on all steel grades in the longitudinal, transverse and diagonal orientations relative to the rolling direction.

The normal anisotropy defined as \( r = \frac{\varepsilon_{\text{width}}}{\varepsilon_{\text{thickness}}} \) was determined from the uniaxial tensile tests. The normal anisotropy properties of most rolled steels are known to vary with orientation to the rolling direction and materials are often characterised by the average normal anisotropy, \( r_m \), and the planar anisotropy, \( \Delta r \). These are related to \( r \) by the following equations:

\[
r_m = \frac{r_0 + 2r_{45} + r_90}{4} \quad \text{(2)}
\]

\[
\Delta r = \frac{r_0 + r_90 - 2r_{45}}{2} \quad \text{(3)}
\]

where \( r_0, r_{45} \) and \( r_{90} \) are the \( r \) values determined along the rolling direction, at 45° to the rolling direction and transverse to the rolling direction respectively.

2.4. Metallographic Examination

Specimens for metallographic examination were etched in 3% Nital solution to reveal the cross section parallel to the rolling direction. The volume fraction of microstructural constituents was determined using standard point counting techniques while the average grain size was determined using the circular intercept method as specified in AS 1733–1976.20) Specimens were repolished and etched in dilute hydrochloric acid in order to reveal the non-metallic inclusion distribution. Fracture surfaces were imaged using a JEOL scanning electron microscope (SEM).

3. Results

3.1. Edge Formability

The \( \lambda_p \) results for the seven steels produced by the three process routes are presented as a function of the tensile strength in Fig. 3 and the total elongation in Fig. 4. Each data point represents the longitudinal tensile result and average \( \lambda_p \) per coil. Lines of best fit were drawn through the comparative steels (solid) and UCS steels (broken). When compared at similar strength and total elongation, the rela-
tive order of the hole expansion performance in the steels produced and tested was UCS > HR > CR & CA steel. $\lambda_p$ of the non UCS comparison grades decreased significantly with increasing tensile strength and decreasing total elongation. In contrast, $\lambda_p$ of the UCS steel decreased only slightly with increasing tensile strength and decreasing total elongation. Normalising the UCS steel did not appreciably change $\lambda_p$, however it did increase the total elongation and decrease the tensile strength, Figs. 5 and 6. Spheroidising the 350 CR & CA steel significantly improved the $\lambda_p$, slightly improved the total elongation and reduced the tensile strength.

3.2. Tensile Properties

Figure 7 shows a plot of total elongation versus tensile strength for all the steel grades made and examined. Each data point represents the longitudinal tensile strength value from each coil. For clarity, the normalised UCS and spheroidised CR & CA results were averaged and summarised in one data point. Lines of best fit were drawn through the centre of the comparative steels (solid) and UCS steels (broken). The total elongation of the UCS steels was found to be lower than comparable strength conventional steels. The difference in total elongation between the conventional steels and the UCS steels was smaller at lower strength levels.

Further analysis was performed to determine the reason for the lower total elongation of the UCS steels. Total elongation can be separated into the various contributing components, namely the yield point elongation (YPE), uniform elongation (UE), and the post-uniform elongation (PUE), Fig. 8. This analysis showed that the lower total elongation of the UCS steel was largely due to lower YPE and PUE. The UE components for all the steel grades evaluated were found to approximately lie along the same TS-UE trend line.

3.3. Microstructures

Typical microstructures are shown in Fig. 9 and the major microstructural features found in each steel are summarised for the lower total elongation of the UCS steels. Total elongation can be separated into the various contributing components, namely the yield point elongation (YPE), uniform elongation (UE), and the post-uniform elongation (PUE), Fig. 8. This analysis showed that the lower total elongation of the UCS steel was largely due to lower YPE and PUE. The UE components for all the steel grades evaluated were found to approximately lie along the same TS-UE trend line.
in Table 2. The microstructures of the UCS steels recently produced were found to be composed of relatively large irregular shaped pro-eutectoid ferrite grains, with a very small amount of grain boundary pearlite, fine acicular ferrite and a small proportion of Widmanstätten ferrite. The microstructure of the 275 UCS steel produced contained a significantly smaller volume fraction of acicular ferrite and a larger ferrite grain size than the microstructure of the 380 UCS steel produced. Normalising the UCS steels resulted in the formation of more traditional ferrite/pearlite microstructures.

The microstructure of the 200 HR steel consisted of uniform polygonal ferrite with a small volume fraction of pearlite. The addition of more carbon and manganese led to a smaller polygonal ferrite grain size and a greater volume fraction of pearlite in the 300 HR steel. Small regions of cementite aggregates were observed, suggesting that some transformation to bainite had occurred. The accelerated cooling experienced by the 360 HR steel resulted in a significant increase in the amount of bainite and a further reduction in the ferrite grain size.

Ferrite recrystallisation imposed by the cold rolling and continuous annealing process route, resulted in the microstructure of the 300 CR & CA steel comprising larger irregular ferrite grains than in the 200 HR steel. Spheroidal and stringer type carbides were observed at the grain boundaries as well as within the ferrite grains. The higher carbon and manganese content of the 350 CR & CA steel resulted in a finer ferrite grain size and a significantly higher proportion of carbides than the 300 CR & CA steel. The spheroidising
heat treatment diminished the large carbides and transformed them into small spheroidal carbides. Microstructural banding was evident, to some extent, in all the conventionally processed steels manufactured from 0.15 wt% carbon and 0.75 wt% manganese steel. In contrast, microstructural banding was not observed in either of the UCS steels or the conventional steel produced from the 0.05 wt% carbon and 0.22 wt% manganese chemistry. This is presumed to be due to their lower alloying content resulting in lower degrees of segregation and, in the case of the UCS steel produced, the higher solidification rate limiting segregation during casting.

Examples of the non-metallic inclusions found in the conventional continuously cast and UCS steels are shown in Fig. 10. The size distribution of inclusions found in the comparative HR and CR & CA steels were comparable due to similar deoxidation practices, ladle treatments, concentration of elements O, N, Al & S, casting practices and total rolling reductions. The inclusion types and distributions consisted of elongated MnS stringers and aligned clusters of alumina concentrated on and around the centreline of the strip. In contrast, inclusions formed in 275 USC and 380 UCS were discrete, globular manganese silicates, typically 0.5–20 μm in size. Composition of these inclusions has been found to vary depending on their size, with the proportion of manganese deceasing and the proportion of silicon increasing with decreasing inclusion size. This finding is consistent with the relative oxidation potential of these elements. Due to the inclusion engineering practice, the relatively fast solidification rate as well as the limited in line hot reduction, the inclusions remain relatively spherical and homogeneously distributed throughout the steel strip.

### 3.4. Micro-Plasticity

The failure mechanism and its relationship to microplasticity was investigated by examining the fracture surfaces of tensile test specimens and the through-thickness cracks in the hole expansion test. To highlight the major differences between steel types, the fracture surfaces of the high strength UCS steel exhibiting a high $\lambda_p$ (380 UCS) were compared to a similar strength, conventional processed, steel with a low $\lambda_p$ (350 CR & CA), Fig. 11. For the tensile test specimen fracture surfaces, all specimens exhibited many small equiaxed dimples showing the failure occurred mainly by ductile fracture. Large dimples, which at higher magnification appeared faceted, were observed in the fracture surface of tensile specimens of the 350 CR & CA steel. The concentration of the large dimples became more numerous towards the centre thickness of the strip. Only a few isolated large dimples were observed in the frac-

### Table 2. Microstructural features of the steels. Acicular Ferrite (AF), Widmanstatten Ferrite (WF), Pro-eutectoid Ferrite (PF), Pearlite (P), Bainite (B), carbide (C).

| Steel          | Volume fraction microstructural constituents | Description of secondary phases | Ferrite grain size (μm) | Microstructural banding | $\lambda_p$ |
|----------------|---------------------------------------------|---------------------------------|-------------------------|-------------------------|------------|
| 275 UCS        | 5–40% AF, 0–3% WF, 60–95% PF, <1% P         | fine carbide network precipitated between ferrite plates | 18–23                   | no observable banding   | 89%        |
| 380 UCS        | 60–95% AF, 5–40% PF                         | fine carbide network precipitated between ferrite plates | 13–18                   | no observable banding   | 87%        |
| normalised 275 UCS | 1.8% P balance PF                     | Pearlite                         | 17.2                    | no observable banding   | 84%        |
| normalised 380 UCS | 2.0% P balance PF                     | Pearlite                         | 16.4                    | no observable banding   | 83%        |
| 200 HR         | 2.7% P balance PF                         | Pearlite                         | 9.2                     | no observable banding   | 94%        |
| 300 HR         | 12.1% P and/or B balance PF               | pearlite with some small cementite aggregates | 6.3                     | moderate banding        | 35%        |
| 360 HR         | 21.5% B and/or P balance PF               | cementite aggregates with a small amount of degenerated pearlite | 4.0                     | light banding           | 36%        |
| 300 CR & CA    | 2.6% carbide balance ferrite              | spherical and stringer form carbides | 9.6                     | no observable banding   | 66%        |
| 350 CR & CA    | 14.7% carbide balance ferrite             | spherical and stringer form carbides | 6.0                     | moderate banding        | 21%        |
| spheroidised 350 CR & CA | 13.3% carbide balance ferrite         | spheroidal carbides               | 6.2                     | moderate banding        | 76%        |

![Fig. 10.](image) Typical non-metallic inclusions at the strip centre line in (a) ultra-thin cast strip (UCS), (b) conventionally cast hot rolled (HR) and cold rolled and continuously annealed (CR & CA) steel.
ture surface of the tensile specimen in the 380 UCS steel examined.

Two hole expansion specimens where chilled in liquid nitrogen and the splits cracked open to permit closer examination of the surfaces of the through thickness cracks normal to their fracture surface. The fracture surface of both hole expansion specimens exhibited many small equiaxed dimples indicating the failure mode during hole expansion was ductile fracture. As with the fracture surfaces of the tensile specimens the fracture surface of the 350 CR & CA hole expansion specimen had more large faceted dimples than the 380 UCS specimen. Furthermore, deep fissures were observed on the fracture surfaces of the 350 CR & CA hole expansion specimen, while these were absent in the 380 UCS steel product. These results show that the 380 UCS steel produced displayed a significantly greater amount of micro-plasticity than the 350 CR & CA steel. A similar trend in micro-plasticity was observed for the steels with intermediate $\lambda_p$.

The failure mechanism was further investigated by sectioning the through thickness cracks which lead to failure in the hole expansion tests, Fig. 12. In all specimens micro-voids were observed ahead of the crack tip due to decohesion between the second phase carbides and non-metallic inclusions and the ferrite matrix. The relative density of the micro-voids for the three steel types produced was found to be CR & CA > HR > UCS.

3.5. Texture Analysis

The influence of crystallographic texture on the edge formability was investigated by determining the location of the through-thickness crack around the circumference of the expanded hole, relative to the rolling direction of the steel, which lead to the failure in the hole expansion test, Fig. 13. The crack location results were compared to the $r$, $r_{\text{min}}$ and $\Delta r$ values. The results revealed a strong correlation between the crack location and the orientation of the minimum $r$-value ($r_{\text{min}}$). In the HR grades, the $r$-value was highest along the diagonal but equally low in the longitudinal and transverse to the rolling directions. The vast majority of the cracks in the HR grades were oriented with their long axis parallel to the longitudinal direction. The $r$-value was lowest along the diagonals in the CR & CA grades and the bulk of the cracks occurred along the diagonals. The UCS material was found to be almost isotropic in nature with the $r$-values along the three directions being close to unity. The $r$-value was slightly higher along the diagonal. The orientation of the cracks in the UCS steel produced was, in contrast, nearly evenly distributed around the circumference of the expanded hole, with a slightly lower frequency of cracks observed at 45° to the rolling direction.

To further investigate the influence of crystallographic texture, the influence of $r_{\text{min}}$ on $\lambda_p$ was determined for all the

![Fig. 11. SEM image of the fracture surfaces of 350 CR & CA and UCS 380 specimens. (a) & (b) tensile specimen viewed normal to the fracture surface. (c) & (d) hole expansion specimens viewed normal to the fracture surface. Deep fissures on the surface of (c) indicated by block arrows.](image1)

![Fig. 12. Cross section of the through thickness cracks which lead to failure in the hole expansion tests for (a) 380 UCS, (b) 360 HR and (c) 350 CR & CA. Magnified region marked by the box for (d) 380 UCS, (e) 360 HR, and (f) 350 CR & CA. Micro-void initiation at carbide/matrix interface indicated by arrows.](image2)
steels produced, Fig. 14. Two populations of data were observed, with the $\lambda_p$ in both populations increasing with increasing $r_{\text{min}}$. The two populations of data were related to the volume fraction and morphology of second phase. One group comprised steels with higher $\lambda_p$, which were characterised by low volume fractions of carbide and pearlite such as: 200 HR, 300 CR & CA, and normalised UCS product; plus steels with a homogeneous distribution of fine carbides such as: 275 & 380 UCS and spheroidised 350 CR & CA. The other population, with lower $\lambda_p$, was characterised by steels with high volume fractions of carbide and pearlite such as: 300 HR, 360 HR and 350 CR & CA. Importantly, the orientation exhibiting the lowest $r_{\text{min}}$ did not generally coincide with the orientation displaying the lowest yield strength.

Fig. 13. Proportion of through thickness cracks which lead to failure along a given orientation compared to the normal anisotropy ($r$), average normal anisotropy ($r_{\text{av}}$) and planar anisotropy ($\Delta r$) for the three steel types investigated.

Fig. 14. Influence of the lowest normal anisotropy value ($r_{\text{min}}$) of the normal anisotropy measured longitudinal ($r_0$), transverse ($r_90$) and diagonal ($r_{45}$) to the rolling direction, on the hole expansion ratio ($\lambda_p$).

4. Discussion

4.1. Formation and Properties of Acicular Ferrite

An understanding of the formation, structure and properties of acicular ferrite is necessary to appreciate the effect acicular ferrite has on the mechanical properties and edge formability. The sequence leading to the formation of the acicular ferrite microstructure is shown schematically in Fig. 15.25) The first transformation product to form during the decomposition of coarse grained austenite is a thin layer of allotriomorphic or grain boundary ferrite, which decrates the austenite boundaries. Further cooling promotes the particle-stimulated nucleation of acicular ferrite plates within the austenite grains. Subsequent nucleation events may stimulate other plates to nucleate sympathetically and therefore a one to one correlation with the number of nucleation events and inclusions is not expected.26) As many ferrite plates nucleate at essentially the same time, nearby plates impinge upon each other creating the characteristic fine interlocking plate morphology. Growth of the plates occurs by a displacive transformation with diffusion of carbon occurring after transformation.27) This leads to a very fine and homogeneous distribution of carbides within and in between the acicular ferrite plates. Transformation temper-
4.2. Mechanical Property Balance

The UCS steel products studied were found to have lower total elongation values compared to equivalent strength conventionally processed steel strip grades, Fig. 7. This difference between the steel types increased substantially at higher strengths. In contrast, the edge formability of the UCS steels produced was similar to the edge formability of the lower strength HR and CR grades and was superior to the higher strength HR and CR grades. While the ductility of steel strip is commonly measured in terms of total elongation, it is now clear that this measure does not provide a fair indication of edge formability performance for the case of the UCS steel grades studied.

The comparatively lower total elongation values of UCS can be explained by the presence of the stronger acicular ferrite phase in the microstructure and its effect on the stress-strain curve. Total elongation reflects the aggregation of the elongations arising from the various segments of the uniaxial tensile test, such as YPE, UE and PUE. The largest difference was observed in the PUE region which is considerably lower for the UCS steel produced, Fig. 8. During deformation the stronger acicular ferrite phase promotes strain localisation to the softer grain boundary pro-eutectoid ferrite, resulting in localised failure. PUE is of little practical significance for overall formability since it essentially corresponds to a local “necking” or thinning phenomenon just prior to fracture. Another contributing factor to the lower total elongation values measured for the UCS steels product was the reduced YPE, particularly at higher strength levels. This behaviour is due to the presence of acicular ferrite which lowered the volume fractions of pro-eutectoid ferrite in the UCS steels produced. Pro-eutectoid ferrite is well known to promote YPE in steel. This analysis shows that total elongation is not a good measure of local ductility when comparing steel produced with widely different microstructures and highlights the need to use more specific measures of ductility such as the hole expansion test to assess the edge formability of steel products.

4.3. Metallurgical Factors Effecting Edge Formability

In general decreasing the tensile strength and increasing the total elongation were found to increase the edge formability within each steel grade, Figs. 3 and 4. Many authors have made this observation and its interpretation is quite simple. Ductile materials are able to deform plastically to greater strains before the onset of plastic instability and splitting occurs, compared to less ductile materials. Tensile strength and total elongation have recently been shown to be the dominant strip ductility factors for predicting the edge formability of a wide range of steel type studied. However, in this investigation, when compared at a given strength or total elongation level a considerable difference among the studied steel grades was clearly observed with the relative order being UCS > HR > CR & CA. Macroscopic mechanical property measurements are therefore alone not able to fully explain the relative order of steels made by the three production routes. The explanation must therefore lie in the underlying microstructure and crystallographic texture.

It is known that a large difference in the hardness of microstructural constituents (i.e. low microstructural homogeneity) significantly degrades the edge formability. It has been shown that during punching de-cohesion occurs between the harder phase and the matrix. Upon subsequent expanding the voids grow and link up, reducing the local ductility of the steel, and thereby lowering the edge formability. This mechanism has been used to explain why microstructures composed of a significant proportion of tempered martensite and bainite have superior edge formability characteristics. Similarly, heavy microstructural banding also reduces the edge formability as this leads to anomalous regions of high hardness.

The relative edge formability of the three steel types studied can be largely explained from a similar view point. Acicular ferrite has high microstructural homogeneity due to: (a) a small effective grain size; (b) very fine distribution of small carbides; and (c) the higher dislocation density of the ferrite further lowers the effective difference in hardness between the matrix and the fine carbides. The absence of microstructural banding due to the high solidification rates of the TRC process further improves the microstructural homogeneity of the UCS steel products. This higher homogeneity of the UCS steels produced enhances resistance to de-cohesion at the carbide/matrix interface during punching and the subsequent expansion which improves microplasticity, and accordingly the edge formability, Fig. 12. Conversely, continuous annealing of CR & CA steel results in the formation of large carbides which promote de-cohesion between the carbide / matrix interface thereby lowering the edge formability of these steels. Strengthening of the HR steel was achieved by a combination of increased carbon and manganese contents and lower cooling temperatures. The lower transformation temperature reduced the pearlite colony size and allowed some transformation to bainite, which like acicular ferrite exhibits microstructural homogeneity, providing the HR steel strip with superior edge formability compared to the formability for the CR + CA steel.

Elongated MnS and aligned alumina phases have also been reported to deteriorate the edge formability due to the large difference in hardness compared to the steel matrix. In order to separate out the effect of non-metallic inclusions and the microstructural constituents, normalising and spheroidising treatments were performed, Figs. 5 and 6. Normalising of the UCS steel product created traditional ferrite and pearlite microstructures and resulted in similar mechanical properties and edge formability to conventional HR steel of comparable strength. Likewise, the spheroidising treatment broke up the large irregular carbides into smaller spherical carbides which substantially improved the edge formability. This behaviour shows the microstructure of the UCS steel products, particularly the morphology of the carbides, and not the presence of aligned non-metallic inclusions was the overwhelming factor which influenced the edge formability. However, the absence of such aligned non-metallic inclusions is well recognised as being generally beneficial to the formability of sheet steel, such as in bending and stretch forming processes.
In addition to microstructure, the edge formability was influenced by the underlying crystallographic texture. The texture is determined during processing by at least four processes: (1) initial solidification; (2) recrystallisation; (3) plastic deformation; and (4) transformation. It has been shown that certain textures improve the \( r_m \) and \( \lambda_r \) of steel, which have been shown to be important indicators of formability.\(^{29,30}\) In particular, increasing volume fractions of the \(<111>\>/ND\) texture (\( \gamma \)-fibre) improves \( r_m \) but not at the expense of increasing \( \lambda_r \). While the presence of the \(<110>\>/RD\) texture (\( \alpha \)-fibre) is thought to be detrimental to \( r_m \).

Both the UCS steel grades studied showed a good combination of \( r_m \) and \( \lambda_r \) compared to the other processes routes. The near unity \( r_m \) values observed along the three test directions (0°, 45°, 90° to the rolling direction) could be attributed to the large, almost through-half thickness columnar austenite grains which exist prior to transformation. This has been shown to result in a large volume fraction of transformed ferrite with the \( \gamma \)-fibre orientation.\(^{31}\) Austenite deformation and recrystallisation processes during hot rolling generally reduce the \( r_m \) and increase \( \lambda_r \) of both UCS steel grades\(^{37}\) and conventional steel types.\(^{32}\) During cold rolling the \( \gamma \)-fibre and \( \alpha \)-fibre are strengthened.\(^{29}\) Recrystallisation during continuous annealing further promotes the \( \gamma \)-fibre at the expense of the \( \alpha \)-fibre texture improving \( r_m \). However, other texture components also form, which lead to a significant increase in \( \lambda_r \).\(^{31}\)

The stress state at the edge of the punched hole during the expansion operation has been shown by finite element analysis to be predominantly uni-axial tension.\(^{33,34}\) This is supported by the equiaxed dimples observed in Fig. 11. Accordingly, steel with a high resistance to through thickness thinning should lead to an increase in \( \Delta \rho \), as less local thinning would occur at the edge of the hole for the same overall strain level. It is important to note here that for all other contributing factors being equal, it is not the \( r_m \) value that would be the most important during hole expansion, but rather \( \min_r \) along any given direction. However, only limited studies have previously been performed to experimentally characterise the suspected improved edge formability characteristics with increasing normal anisotropy.\(^{35}\) In this study a strong relationship between the location of the cracks and where the resistance to through thickness thinning was smallest was indeed observed, Fig. 13. Furthermore, after accounting for the morphology and distribution of carbides, a reasonable correlation was observed between \( r_m \) and \( \rho_m \), Fig. 14. This revealed that the crystallographic texture influenced the edge formability of the carbon steels investigated. Therefore control of texture, particularly increasing the minimum \( r \)-value, or accounting for the distribution in local anisotropy in the design of forming operations should lead to improved edge formability in low-carbon steel. In the case of UCS products produced by the CASTRIP process, the nearly isotropic behaviour should allow for the random orientation of the sheet in forming operations.

5. Conclusion

Twin roll casting (TRC) of ultra-thin cast strip (UCS) produced by the CASTRIP\(^{\circ} \) process is able to generate novel microstructures consisting of significant proportions of acicular ferrite. This study investigated the mechanical properties and edge formability as measured by the hole expansion test of commercially produced UCS sheet steel product. The results were compared to those from conventional hot rolled (HR) and cold rolled and continuously annealed (CR & CA) structural steel grades of similar strength. From the results obtained the following conclusions were made.

The UCS steels produced had lower total elongations than conventionally processed strip steels of similar strength. This was explained by the high dislocation density of acicular ferrite reducing the yield point elongation (lower volume fraction of pro-eutectoid ferrite) and post uniform elongation (strain localisation to softer grain boundary ferrite) contributions to total elongation. In spite of the lower total elongation, the practically relevant uniform elongation regions of the stress-strain curves of all three production routes were similar when compared at the same strength levels.

Within each process route, the edge formability was found to increase with decreasing tensile strength and increasing total elongation. However, when compared at similar strength and total elongation levels the edge formability of steel sheet produced via the three process routes used to produce the steels were considerably different with the relative edge formability being UCS > HR > CR & CA. This highlights the importance of using specific measures of ductility, such as the hole expansion test, when comparing the formability of sheet steel from different processes routes.

The relative difference in edge formability among the three types of steel produced for this study was primarily accounted for by differences in microstructural homogeneity. In particular, improved edge formability was associated with: a fine and homogenous distribution of carbides; small effective grain size; low microstructural banding, and a small difference in hardness between microstructural constituents. High microstructural homogeneity gives greater resistance to void initiation and propagation during punching and the subsequent expanding operation.

The location of the through thickness cracks which lead to failure in the hole expansion test were generally located with their long axis parallel to the direction where the resistance to through thickness thinning was the least. This revealed that increasing the normal anisotropy (or \( r \)-value) in particular the minimum normal anisotropy along any given direction, can improve the edge formability of low-carbon steel sheet.

The superior edge formability of the UCS steels produced and evaluated in this study can be attributed to the high microstructural homogeneity of the associated microstructure and the high degree of uniformity of the normal anisotropy, (which is evidenced by a relatively higher minimum normal anisotropy) and low planar anisotropy compared to those of conventionally produced steels.

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