Formation of Ferritic Products during Continuous Cooling of a Cu-bearing HSLA Steel

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A continuous cooling transformation (CCT) diagram was constructed by a combination of dilatometry and metallographic methods for a Cu-bearing HSLA steel, which is a low carbon low alloy variant of the ASTM A710 type structural steel. It was found that the decomposition of austenite was significantly depressed to lower transformation temperatures resulting in a prominent transformation region for bainitic structures, at temperatures intermediate between those of diffusional products and the displacive transformation to martensite. Polygonal and quasi-polygonal ferrite were observed to grow across and eliminate the prior austenite grain boundaries at relatively low cooling rates. At a cooling rate ranging from 0.35 to 20°C/s, the structure was characterised by a mixture of quasi-polygonal ferrite, Widmanstätten side-plate ferrite, and bainitic structures associated with minor dispersed islands of martensite and/or retained austenite which were dark etching on preparation for optical microscopy. This microstructure develops by the following processes. The Widmanstätten side-plate ferrite nucleates from the ferrite grain boundary allotriomorphs at the early stage of transformation, together with the bainitic ferrite plates which nucleate directly at the prior austenite grain boundaries. On further cooling, the neighbouring plates of Widmanstätten and bainitic ferrite each tended to coalesce and the volume of untransformed austenite decreased and the shapes of the enclosed volumes evolved into residual islands between the ferrite plates. Provided the cooling rate was greater than 20°C/s, the bainitic ferrite plates nucleated directly at the prior austenite grain boundaries, and the plate morphology was revealed by regions of elongated retained austenite or its decomposition products. At the fastest cooling rate obtained by dilatometry (~375°C/s), the structure was largely characterised by a mixture of bainitic ferrite and martensitic packets surrounded by retained austenite films. Dilatometric and metallographic examination of the martensite and bainitic ferrite formed on rapid cooling failed to find a clear microstructural distinction between the two products. The packets of bainitic ferrite plates were generally nucleated directly from the prior austenite grain boundaries, whereas the martensite was characterised by thinner ferritic units with a higher dislocation density. There also appeared to be a larger number of variants of lath packets and apparent intragranular nucleation in the case of martensitic ferrite.

KEY WORDS: Cu-bearing low alloy steels; CCT-diagram; polygonal ferrite; quasi-polygonal ferrite; Widmanstätten ferrite; granular bainite; bainitic ferrite; lath martensite.

1. Introduction

Although the decomposition of austenite is one of the oldest and most well understood research areas in physical metallurgy, industry continues to demand innovative higher strength steels in order to lighten structural parts. A systematic alloy design trend of high strength combined with higher toughness has led to the development of low and/or ultra low carbon steels. These engineering requirements are mainly driven by plate and sheet steel applications which, in addition to improved strength and toughness, have resulted in increased weldability due to the lower carbon content. A decrease in carbon content promotes a good combination of toughness and weldability by minimising the volume fraction of carbide containing microconstituents, but generally at the cost of strength and hardness.

To compensate for this lower strength, Cu-bearing steels are an attractive class of HSLA steels in which copper can be added to the steel composition in an amount appropriate precipitation hardening by formation of copper rich particles during an aging treatment.1-4) Hot rolled steels have been substantially improved by microalloying and thermomechanical controlled processing in recent years.5) Copper bearing high strength low alloy (HSLA) steels have similarly been developed into commercially viable types of HSLA steels. Such alloys are considered to have a good balance of strength and other mechanical properties such as ductility, toughness, fatigue resistance, and workability by forming, welding and machining. As a result, the carbon content of this type of HSLA steels shows a downward trend in carbon to about 0.02%. In contrast to the equiaxed or polygonal ferritic microstructures of conventionally hot rolled plate steels, the ferritic microstructures formed by decomposition of austenite in such low carbon thermomechanical control
processed steels often assume non-equiaxed morphologies. The temperature range of formation of the non-equiaxed morphologies of ferrite is intermediate between those at which austenite transforms to equiaxed ferrite/pearlite and to martensite, and therefore is the same as that in which bainite forms in medium carbon steels. However, the low and ultra-low carbon ferritic microstructures which form at such intermediate temperatures are different in variety and form from classical bainite structures, defined for higher carbon contents.

In general, a mixture of ferritic phases with different morphologies and properties is to be expected on transformation over a wide range of temperatures during continuous cooling. The morphological identification and classification system for these structures have been generally regarded as difficult compared to the ferrite-carbide aggregates formed in higher carbon steels. These non-classical microstructures usually have a variety of mixed phases consisting of intermediate stage transformation products between those of purely diffusional (ferrite and pearlite) and displacive martensitic (martensite) phases in association with minor islands of carbon enriched austenite and/or its transformation products (microphases), depending on the level of carbon partitioning in the remaining austenite and the thermal cycle. There are many difficulties and problems in identifying the phases and sometimes serious confusion in the terminology of the microstructures is apparent. Furthermore, it is still unclear if bainite and martensite, which are clearly distinguishable in medium carbon steels, can be distinguished in very low carbon low alloy steels as well. This paper deals with an experimental study of the characterisation of transformations and microstructures in a Cu-bearing low carbon steel in which austenite transforms to equiaxed ferrite/pearlite and therefore is the same as that in which the temperature of the first-stage expansional change.

2. Material and Experimental Procedure

The steel used in the present study was CR HSLA80 steel developed by BHP Steel, SPPD, Port Kembla, Australia. The chemical composition of this steel is given in Table 1. The steel composition was selected from a series of laboratory produced trial ingots which were subjected to various hot rolling schedules and ageing treatments. On this basis, a commercial heat was produced by continuous casting. Slabs of the steel were subsequently hot rolled from 1200°C by a thermomechanical controlled processing route and then aged at 550°C.

Continuous cooling decomposition of austenite was investigated using a high-speed Theta dilatometer. Hollow cylindrical specimens (l: 13 mm; O.D.: 4.9 mm; I.D.: 3.5 mm) were cut transversely from the as received steel plate. The samples were austenitised at 1200°C for 5 min in a vacuum of about 10⁻⁴ torr and subsequently cooled at various rates. The specimens cooled at a rate slower than 20°C/s followed a constant cooling path down to 200°C with a temperature deviation of no more than 10°C from the linear cooling curve over the transformation region. In the cases of dubious data, double checks were made to enhance the reliability of the test.

The optical microstructures were observed after polishing and etching specimens in 2.5% nitral. Hardness measurements were made with a 10 kg load and are reported as Vickers hardness numbers. Thin foil specimens were prepared using a twin jet polishing Struer Tenupol system operating at 60 V in a solution consisting of 95% glacial acetic acid and 5% perchloric acid at a temperature of about 12°C. Observations were carried out in a JEOL 2000 FX operated at 200 kV.

3. Results

3.1. Continuous Cooling Transformation Behaviour

3.1.1. Dilatometry Curves

In order to construct the complete CCT-diagram for the low carbon plate steel, the cooling rates were varied from 0.1 to ~375°C/s. When the cooling rate was slower than 0.35°C/s, the dilatation curve exhibited a major dilatational change consistent with the formation of polygonal ferrite. For a cooling rate between 0.35 and 3°C/s, the dilatation curve exhibited a two-stage expansion as shown in Fig. 1(a). In order to determine accurately the associated transformation start temperature of various microstructural components, the specimens were subjected to a series of interrupted quenching tests to freeze the microstructure at various temperatures during continuous cooling transformation. For example, the specimens were cooled at a rate of 1°C/s, before interrupted quenching at various temperatures during the transformation. Based on the microstructural observations, polygonal and/or quasi-polygonal ferrite was found to form firstly at about 725°C which is consistent with the temperature of the first-stage expansion change. The formation of intermediate ferritic transformation products started at about 625°C and ceased at about 528°C, respectively. Both temperatures are compatible with the start and finish temperature of the second-stage dilatational change.

At a moderate cooling rate (3 to 20°C/s), the dilatometric curve exhibited only one major dilatational change as shown in Fig. 1(b). Based on the interrupted quenching tests and microstructural observations, the dilatational change corresponded to the formation of quasi-polygonal ferrite, Widmanstätten ferrite, and bainitic structures associated with minor dispersed islands of martensite and/or austenite. However, the associated derivative expansion curve of d(Δl/l₀)/dT versus temperature (T) revealed details of the transformation start temperatures of various ferritic products more clearly than the original (Δl/l₀)-T curve. Figure 2 shows a typical derivative dilatation curve as a function of temperature for the specimen continuously.

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
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<th>Ni</th>
<th>Cu</th>
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<tr>
<td>Nb</td>
<td>Ti</td>
<td>P</td>
<td>S</td>
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<td>0.013</td>
<td>0.012</td>
<td>0.003</td>
<td>0.0075</td>
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Table 1. Chemical composition (in weight percent) of the CR HSLA80 steel investigated.
cooled at 5°C/s. Two steps are observed in the dilatational change. Based on interrupted quenching tests and microstructural observations, the final expansional change is attributed to the decomposition of remaining carbon enriched austenite pools entrapped by the ferritic matrix of the intermediate transformation products. At a cooling rate greater than 20°C/s, the specimen did not follow a constant cooling path after the transformation had begun because the rate of latent heat evolution was so large that the regulated helium flow was not able to remove the heat fast enough. In this case, the cooling rate becomes non-linear, but the expansional change has been attributed to the major intermediate ferritic transformation products. For the fastest cooling rate obtained (~375°C/s), the dilatation curve shows a major catastrophic expansional change at relatively low transformation temperature (Fig. 1(c)), which is largely associated with martensitic transformation.

3.1.2. Continuous Cooling Transformation Diagram

Figure 3 shows the continuous cooling transformation (CCT) diagram for the low carbon HSLA steel. As can be seen, the CCT-diagram is characterised by multilayer transformation curves. The transformation start and finish temperatures rise gently with decreasing cooling rate. The major transformation curves lie in the temperature range between 750 to 350°C, which are significantly depressed because of the presence of the strong austenite stabilising elements of Cu, Ni, and Mn. The features of microstructural evolution over this temperature range include polygonal (αp) and quasi-polygonal ferrite (αq), Widmanstätten ferrite (αw), granular bainite (αB), bainitic ferrite (α°B) and lath martensite (α'M). The CCT-diagram shows a prominent region associated with an intermediate structure, frequently referred to as bainitic structure. This region is terminated by a boundary field at about 300°C (M°MA) associated with the decomposition to martensite of small carbon-enriched austenite pools entrapped at the boundaries of the interlath or interpacket bainitic structures formed at an earlier stage of transformation. The suppression of M°MA to lower temperatures by de-
creasing cooling rate emphasises significant partitioning of carbon in the remaining austenite islands which accompanies polymorphic and intermediate ferritic product formation at higher temperatures, and the effect of carbon enrichment on martensite formation. However, the degree of carbon partitioning in the remaining austenite regions may be variable, and some transformation to martensite may occur at temperatures above the indicated $M_a$ curve. It is interesting to note that in the CCT-diagram, the $M_a$ curve is distinct from the $M_s$ temperature. The $M_s$ temperature, which was estimated using the formula derived by Andrews,	extsuperscript{17} indicates the temperature at which austenite of the alloy matrix begins to transform to martensite, while the $M_s$ curve represents the temperature at which small amounts of carbon enriched austenite begin to transform to martensite.

Various cooling rates utilised for drawing the CCT-curves are shown in the diagram. The Vickers hardness value for each individual cooling rate is shown at the end of the respective curve.

3.1.3. Hardness and Optical Microstructures

Figure 4 shows examples of typical microstructures with the associated hardness obtained from the application of continuous cooling transformation over a wide range cooling rates. The microstructure of a water quenched specimen for comparison is also shown in Fig. 4(a) which is characterised by many packets of laths and parallel alignment of the individual bainitic or martensitic units within the packet. The microstructures formed at the fastest cooling rate possible by dilatometry in this study are presumed to consist of martensite plus bainitic ferrite, since the corresponding hardness (335 HV10) is lower than that obtained after direct water quenching (350 HV10). The structures associated with the sharp drop in hardness are shown in Figs. 4(b) and 4(c). These structures are characterised by a mixture of martensite and bainitic ferrite, with the bainitic type structure being revealed by small elongated islands of martensite and/or retained austenite. Furthermore, at a higher cooling rate the ferritic crystals become finer, and the martensite/austenite constituent becomes thinner and more uniformly distributed in the bainitic matrix.

As the cooling rate was reduced below 20°C/s, polygonal and quasi-polygonal ferrite formed at the prior austenite grain boundaries became increasingly common. At 5°C/s the amount of grain boundary ferrite is small and the structure has a part plate-like and a part granular appearance, Fig. 4(d). For 1°C/s, quasi-polygonal ferrite dominates, but plate-like granular regions form in remaining austenite volumes, Fig. 4(e). Figure 4(f) shows a microstructure formed in the most slowly cooled specimen, which consisted of a large volume fraction of polygonal ferrite, and dark contrast regions which are mainly pearlite or degenerate pearlite.

3.1.4. Development of Bainitic Structures during Continuous Cooling Transformation

In order to investigate in detail the transformation mechanism of bainitic structures, microstructural observations were carried out on a series of interrupted quenching specimens from temperatures between the transformation start and finish. The cooling rates of 5 and 15°C/s were chosen, for which granular bainite “αB” and bainitic ferrite “α°B” were respectively predominant. The specimens were cooled continuously at 5 and 15°C/s after being austenitised at 1200°C for 5 min, and then interrupted quenched from 620, 580, 555°C for 5°C/s, and 565, 540 and 515°C for 15°C/s. The transformation start and finish temperatures were 670 and 490°C for 5°C/s, and 630 and 435°C for 15°C/s, respectively.

Figure 5 shows the microstructural evolution in the specimens continuously cooled at 5°C/s. As can be observed in Fig. 5(a), decomposition down to 620°C produces only a grain boundary ferrite allotriomorphs which have a generally faceted appearance consistent with αq. At a later stage of transformation, Widmanstätten side-plate ferrite grows from these ferrite allotriomorphs as can be seen in Fig. 5(b). Widmanstätten ferrite side-plates formed at these “high” transformation temperatures are relatively wide. It is inferred that side plates of the same orientation growing from grain boundary ferrite coalesce by sidewise growth, forming wider plates. As transformation proceeds (Fig. 5(c)), the structure assumes the appearance of granular bainite “αB” with elongated or more globular pools of untransformed austenite between the plates or groups (packets) of plates. The amount of austenite decreased as the temperature fell. This transformational process for the formation of granular bainitic structure was dominant in the slower cooled specimens.

As cooling rate was increased, the tendency for nucleation and growth of bainitic ferrite from the prior austenite.
ite grain boundary increased. Figure 6 illustrates the microstructural development in step-quenched specimens cooled at 15°C/s. Some grain boundary allotriomorphs nucleated at the early stage of transformation in some regions and Widmanstätten ferrite side-plates grew from these ferrite allotriomorphs into the austenite grains as in (a) and (c), together with the nucleation and growth of bainitic ferrite plates directly at the prior austenite grain boundaries as in (b) and (d). As transformation progressed, the amount of bainitic structure (granular bainite “αB” and bainitic ferrite “αΦB”) increased. Parts of the neighbouring plates of Widmanstätten and bainitic ferrite tended to coalesce, and the pools of remanent austenite became finer and thinner.

3.2. Electron Microscopy of Continuously Cooled Microstructures

The microstructure of the specimen cooled at 5°C/s, as revealed by optical microscopy consisted of a mixture of quasi-polygonal and Widmanstätten side-plate ferrite together with intermediate ferritic structures (granular bainite “αB” and bainitic ferrite “αΦB”) associated with minor dispersed dark contrasting regions of MA constituents (Fig. 4(d)). Consistent with the observations from optical microscopy, TEM micrographs revealed ferritic crystals either in the form of parallel ferritic crystals separated by low angle boundaries (Fig. 7(a)), or islands of a second constituent in a relatively featureless ferritic matrix (Fig. 8(a)). An example of grain boundary and Widmanstätten side-plate ferrite formed at a high transformation temperature is also shown in Fig. 7(b). Figure 8(b) is a bright field micrograph in which the island microstructure is revealed in detail. Twinned martensite is observed in the island, because of partitioning of carbon to austenite as ferrite grows with falling temperature.

As cooling rate was increased, the higher thermodynamic driving force for the decomposition of austenite resulted in finer and more well defined packets of ferritic lath nucleat-
ed directly from the prior austenite grain boundaries (Fig. 9). As can be observed, the ferritic transformation product appeared in a large group or packet of parallel units with individual laths being separated by low angle boundaries. Within these packets, the boundaries between ferritic laths were usually fairly straight and quite well defined. Figures 10(a) and 10(b) for the specimen cooled continuously at 160°C/s are higher magnification bright field and dark field micrographs of fine ferritic laths sandwiched between a thin filmlike phase. The dark field image micrograph was taken using the (200)A austenite reflection, and confirmed that the filmlike phase is retained austenite. The significant stabilization of retained austenite films in the rapidly cooled specimens may indicate that stabilization is partly due to the carbon and nitrogen partitioning. The high Cu, Ni and Mn contents of the alloy together with residual compressive strain may also mechanically stabilise the austenite. No carbides were detected at the lath boundaries.

The fastest dilatometric cooling rate (~375°C/s) showed a mixture of microstructural components. Examples of low magnification bright field micrographs of fine ferritic lath-like microstructures are shown in Fig. 11. The sheaflike in-
Fig. 12. TEM bright field micrographs showing the changes in fine structure with cooling rate. Specimens cooled at: (a) 5°C/s; (b) 160°C/s; (c) water quenched.

Fig. 13. Effect of cooling rate on the thickness of ferritic laths in specimens cooled at various rates.

Fig. 14. Effect of cooling rate on the number of ferritic packets per austenite grain (Np/grain) in specimens cooled at various rates.

tragranularly nucleated ferritic units, characterised by high dislocation density are normally attributed to low carbon lath martensite (Fig. 11(a)). Although the growth of both martensitic and bainitic ferrite does not proceed across the prior austenite grain boundaries, the large group of thin ferritic crystals which appears to have nucleated directly at the prior austenite grain boundary in Fig. 11(b), is likely to be bainitic ferrite, as discussed in Sec. 4.4.

Figure 12 presents typical structural changes as a function of cooling rate. As can be seen, the thickness of the ferrite laths decreases and the density of dislocations within the ferritic units increases as the cooling rate increases. Bainitic ferrite, or lath martensite or a mixture of both of these forms of ferrite becomes dominant at the higher cooling rate. The thickness of the ferrite laths as a function of cooling rate was also measured using electron micrographs (Fig. 13). It is evident that increasing the cooling rate gradually reduces the thickness of the ferrite laths, but there was a significant change for cooling rates faster than 30°C/s. Figure 14 shows changes in number of ferritic packets per austenite grain (Np/grain) as a function of cooling rate. As can be observed, the number of packets are slightly increased with cooling rate lower than 30°C/s. The dramatically change in clustering of ferritic packets at cooling rate faster than 30°C/s is partly associated with martensitic transformation.

4. Discussion

4.1. Polygonal and Quasi-polygonal Ferrite

Based on the dilatometric experiments, interrupted quenching tests and metallographic observations, it was found that polygonal ferrite (αp) generally nucleates and grows at the prior austenite grain boundary (Figs. 4, 5). This transformation process is associated with compositional change which proceeds by diffusion and produces a reconstructive transformation with the normal sigmoidal isothermal transformation characteristic. During reconstructive transformation, the atoms cross the γ/α interface in an uncoordinated manner accomplishing the required lattice change and simultaneously causing a minimisation of the strain energy. The growth mode is controlled by the long range diffusion of solute atoms with local equilibrium being established at the α/γ interface. Since the formation of polygonal ferrite involves a diffusional rearrangement of iron atoms at relatively high temperatures, this process does not result in accumulation of transformational strain in the ferrite because thermally activated relaxation can occur. As a result, the polygonal ferrite grain interiors are generally substructure-free and therefore insensitive to etching (white grains). The final structure is defined by ferrite grain boundaries, and dark etching regions of pearlite or degenerate pearlite (Fig. 4(f)).
At somewhat lower transformation temperatures the ferrite growth mode becomes apparently anisotropic and the final shape seems irregular and often has jagged grain boundaries (Figs. 4(e), 5). This more irregular shaped ferrite has been termed quasi-polygonal ferrite (qB) by the Japanese Bainite Committee,12 idiomorphic ferrite (aB) by Bhadreshia,20 and “massive ferrite” by Wilson21 and Edmonds.22 Although the formation of polygonal and quasi-polygonal ferrite did not lead to a clear dilatometric resolution on continuous cooling transformation (Fig. 1(a)), the step-quenching experiments revealed that the transformation start temperature of quasi-polygonal ferrite was just below that of typical polygonal ferrite and that the formation of both forms of ferrite continued down to the B_s temperature (Figs. 3, 4, 5). It is likely that the formation of quasi-polygonal ferrite can start on pre-existing allotriomorphic polygonal ferrite at the austenite grain boundaries once the temperature has fallen to a level where ledge growth is energetically and kinetically favoured over the motion of high angle boundaries.

4.2. Widmanstätten and Bainitic Ferrite

Widmanstätten ferrite (aW) was observed to form by anisotropic (side-plate) growth from allotriomorphic ferrite on the austenite grain boundaries (Figs. 4(d), 5, 7(b)). In contrast, bainitic ferrite (aB) nucleates directly at the prior austenite grain boundaries (Figs. 4(c), 6, 7(a)). The “acicular” shapes of aW and aB crystals, which minimise strain energy generated during transformation, differentiate these forms of ferritic phases from polygonal and quasi-polygonal ferrite which are produced by reconstructive transformation across disordered boundaries by long range diffusions.

In this low carbon plate steel, aW was formed at relatively high transformation temperatures in conjunction with the polygonal and/or quasi-polygonal ferrite formation (Figs. 4(d), 5, 7(b)). On the other hand, when Widmanstätten and bainitic transformation occur at higher temperatures, carbon can diffuse long distances either during or after the transformation into the remaining austenite regions. The carbon enrichment of the residual austenite pools is clearly shown by the depression of their martensite transformation start temperature (curve M_s in Fig. 3) and by the twinned martensite in the island structures (Fig. 8(b)). Although this phenomenon can be interpreted as support for the diffusional model of Widmanstätten and bainitic transformation in low carbon steels,23 other workers24,25 have proposed a diffusionless transformation for Widmanstätten ferrite formation with respect to substitutional solutes, with interstitial diffusion of carbon into remanent austenite.

The Widmanstätten and bainitic structures are characterised by a plate or lath shape associated with a relatively high dislocation density (Figs. 7, 9). Although the substructure of Widmanstätten ferrite shows slightly lower dislocation density than that of bainitic ferrite, it has been often observed that Widmanstätten ferrite shows a consistent crystallographic orientation relationship and well established habit plane close to {111}_bcc, which is close to that observed for bainitic ferrite. However, the nucleation sites for the two structures are completely different, e.g. Widmanstätten ferrite grows from the grain boundary allotriomorphic ferrite produced by a completely diffusional process (Figs. 4(d), 5(b), 7(b)), whereas the bainitic ferrite is thought to nucleate diffusionlessly at low carbon regions in the vicinity of austenite grain boundaries (Figs. 4(c), 8, 7(a), 9). Furthermore, there are characteristic differences between the Widmanstätten and bainitic ferrite, with the main differences being a larger size of individual subunits, a greater width and a lower dislocation density in the Widmanstätten ferrite.

In the present work, the formation of Widmanstätten and bainitic-type ferrite structures did not show a clear dilatometric resolution on continuous cooling transformation (Fig. 1(b)), and the observed aW seemed to occur mostly in company with bainitic type transformation (Figs. 4(d), 6). Moreover, the transformational process is associated with the formation of twinned martensite or martensite-austenite constituent in island regions (Fig. 8(b)), indicating that a significant level of carbon enrichment in the untransformed austenite pools occurs during or after the formation of the Widmanstätten and bainitic ferrite. This indicates that both of these types of ferritic phases probably form by the same transformation mechanism.

4.3. Granular Bainite (aB) and Bainitic Ferrite (aB)

The experimental results show that the austenite in the low carbon steel investigated transforms to a largely ferritic microstructure over a wide range of cooling rates (Figs. 3, 4). The ferritic matrix assumes various morphologies and substructures which are a function of cooling rate and temperature range of formation. For intermediate cooling rates, development of the intermediate transformation product can be described in terms of the following stages: lath-like bainitic crystals form in the early stage of transformation, and subsequently the resultant carbon enriched austenite pools entrapped by the bainitic matrix transform fully or partially to high carbon martensite at lower temperatures on continuous cooling (Figs. 5, 6). A possible explanation for the fact that the residual austenite regions adjacent to the bainitic crystals remain untransformed at the reaction temperature is that their carbon concentration increases as a consequence of transformation to such an extent that the formation of bainitic structure eventually becomes thermodynamically impossible. If the carbon concentration of the remaining austenite regions is large, the formation of bainitic ferrite ceases at an early stage of the reaction, and so the volume fraction of the residual phases is larger (Fig. 4).

By examining in detail the dilatometry and metallography of ferritic transformation products obtained for a wide range of cooling rates, it can be seen that the formation temperature range of the allotriomorphic ferrite and bainitic structures was clearly separated on the dilatation curves (Fig. 1(a)), and that the prior austenite grain boundaries are preserved when intermediate ferritic products are formed (Figs. 4(c), 6, 7(a)). These phenomena imply that a polymorphic transformation mechanism other than reconstructive transformation is operative during nucleation and growth of these ferritic products. In contrast, the formation of polygonal and quasi-polygonal ferrite occurs by diffusion because the product phase grows across and eliminates
the prior austenite grain boundaries (Figs. 4, 5). In addition, the ferritic matrix of the intermediate transformation products contains a relatively high density of dislocations (Figs. 7, 8, 12), which are a product of the inherent volume change on $\gamma \rightarrow \alpha$ transformation.

From the kinetic point of view, it was found that the formation of the two types of bainitic structures did not show a clear dilatometric resolution during continuous cooling transformation (Fig. 1(b)). This implies that the formation of these bainitic products occurred by the same transformation mechanism. However, by interrupted quenching when bainitic transformation had just begun during continuous cooling, it was found that the granular bainitic structure was formed at higher temperatures (Figs. 5, 6). It was also found to be slightly softer than the bainitic ferrite. Therefore, the differences between $\alpha B$ and $\alpha' B$ are in degree rather than in kind, with the laths being wider and slightly less dislocated and the residual islands, coarser and less elongated in $\alpha B$, which forms at a lower cooling rate or at higher temperatures (Figs. 4, 12, 13).

4.4. Bainitic Ferrite and Lath Martensite

In the present investigation, martensite and bainitic ferrite could be distinguished by a quite different etching susceptibilities as shown by optical micrographs (Fig. 4). Since the bainitic transformation occurs at a higher temperature compared to the martensitic transformation, carbon can diffuse to a greater extent either to the remaining austenite pools or to the boundary between laths. When this structure is etched, the boundaries of the retained austenite islands or its decomposition products etch deeply, giving the overall appearance of a plate shaped ferritic matrix with a superimposed dispersion of dark contrasting “particles”. The martensitic transformation is characterised by clusters of very fine ferritic laths which form at lower temperatures (Figs. 11(a), 12(c), 14). Since the carbon distribution in the martensitic structure is more uniform, it etches more uniformly (Fig. 4(a)).

At a fast cooling rate, the austenite decomposition may change from bainitic to martensitic transformation, but this could not be detected by means of dilatometry since the associated curve shows only a single exponential change over the bainitic and martensitic transformation temperature range (Fig. 1(c)). This indicates that the kinetic properties of bainitic ferrite are similar to those of martensite. TEM examination also indicated that the bainitic ferrite crystals exhibited a constant crystallographic orientation relationship with the parent austenite phase, in a similar fashion to that of low carbon lath martensite (Fig. 10). This observation is consistent with the displacive view of the bainitic transformation. Therefore, the overall experimental evidence indicates that carbon atoms partition by diffusion either during or after the bainitic transformation, while the crystallographic properties are much more easily understood in terms of displacive theory. These aspects of bainitic transformation have stimulated significant discussion regarding the mechanism of the bainite formation in modern low carbon HSLA steels.

However, on the basis of the overall dilatometric and metallographic observations, it is evident that the transformation start temperature of the bainitic ferrite is significantly higher than that of the martensitic transformation, and so the bainitic ferrite is softer than the martensite (Figs. 3, 4). Furthermore, the bainitic ferrite structure is characterised by fine and extremely long parallel ferritic laths which nucleate directly at the prior austenite grain boundary (Figs. 4(c), 7(a), 9, 11(b)), while the martensitic structure appears to develop by bursts of nucleation and growth of ferritic lath packets of different variant clusters within the austenite grains (Figs. 4(a), 11(a), 12(c), 14). The formation of these microstructural features is associated with the thermodynamical driving force during bainitic and martensitic transformation. In the bainitic ferrite structure, the laths in a given packet are almost all in similar crystallographic orientations and the laths are therefore separated by low-angle boundaries (Figs. 7(a), 11(b), 12(a)). In contrast, smaller packets of laths, with each packet consisting of variants of a numbers of normals similar habit plane axis clustered around a (111) are observed in the martensitic structure (Figs. 11(a), 12(c)). This is reasonable because rapid quenching induces high macroscopic stresses and stress gradients which will interact with variant selection in such a way as to partially accommodate the quenching stresses. The lack of thermal activation to dissipate transformational strains is counteracted by formation of small packets of differing variant clusters and the limited thickening of laths. The larger numbers of packets within a prior austenite grain also reflect the accommodation of strain accumulated over large distances through the austenite grain (Figs. 11, 12, 13, 14).

A small but significant volume fraction of the high temperature austenite phase was stabilised at room temperature in the rapidly cooled specimen (Fig. 10). As well as a significant amount of fcc stabilising elements such as Cu, Ni and Mn in the alloy, the high austenitising temperature (1200°C) would also contribute to the stabilisation of austenite through complete or partial dissolution of stable carbides (TiC, NbC) and some of the nitride (TiN), providing the necessary interstitials in solution. As a result, the interstitial carbon and nitrogen are rejected from supersaturated bainitic ferrite to untransformed austenite either during or after the bainitic transformation and then concentrates in untransformed austenite, leading to significant stabilisation of interlath retained austenite films. Thus it is reasonable to conclude that the large groups of ferrite laths of similar orientations with interlath austenite films are bainitic ferrite type structure (Figs. 9, 10, 11(b)).

It is interesting to point out that the bainitic ferrite structure formed in the low carbon steel on continuous cooling is similar to medium carbon upper bainite in the sense of parallel ferrite laths alternating with a minor interlath residual phase (Figs. 7(a), 10). However, in the case of the low carbon steel investigated, the second “phase” located at the interlath or intersheaf boundaries is not cementite as in classical upper bainite, but one or more of a series of residual phases or phase mixtures such as retained austenite, martensite, and martensite-austenite constituent. Nevertheless, this bainitic or lath ferrite structure with interlath or interpacket residual phase is certainly the low carbon variant of upper bainite and is the most common ferritic product observed for intermediate continuous cooling transformations.
5. Conclusions

Decomposition of austenite during continuous cooling has been studied in a low carbon (0.055% C), Cu-bearing HSLA steel containing significant additions of manganese, nickel and copper. The conclusions are as follows:

(1) The continuous cooling transformation (CCT) diagram for the low carbon steel was constructed by a combination of dilatometric and metallographic observations. The CCT-diagram exhibited multilayer transformation characteristics which involved curves for the evolution of various ferritic products and a curve for the subsequent decomposition of remaining carbon enriched austenite pools at lower temperatures.

(2) The polygonal and quasi-polygonal ferrite, which formed at high temperatures during slow cooling, were observed to grow across and eliminate the prior austenite grain boundaries. These ferritic transformation products were characterised by relatively insensitive to etching (white grains).

(3) Between the formation temperatures of polygonal/ quasi-polygonal ferrite and martensite, the intermediate transformation products can be classified as the ferritic products: Widmanstätten ferrite (αB), granular bainite (αB) and bainitic ferrite (α′B). Both kinetic and metallographic data showed that the bainitic transformation products (αB, α′B) formed by two typical stages of transformation: (i) the formation of ferrite laths; and (ii) the subsequent decomposition of retained carbon enriched austenite regions to martensite at lower transformation temperatures. The carbon enriched austenite decomposition products typically consisted of martensite and austenite (MA) constituents.

(4) The granular bainitic structure was characterised by more globular dispersed islands of martensite and/or austenite constituents in a relatively non-acicular ferritic matrix. This microstructure is considered to be formed by the coalescence of wide Widmanstätten and bainitic ferrite crystals.

(5) The high temperature austenite phase was stabilised at room temperature as films around ferritic laths produced on rapid continuous cooling (160–375°C/s) from a high austenitising temperature (1 200°C). The lath ferrite structures formed in association with interlath continuous retained austenite films are considered to be bainitic ferrite because of their higher transformation start temperature than the calculated Ms temperature, the characteristic feature of large groups of parallel ferritic laths which nucleated directly at the prior austenite grain boundaries, and the lower hardness and lower dislocation density compared with lath martensite.

REFERENCES