Columnar grain development in C–Mn–Ni low-alloy weld metals and the influence of nickel

ZHUYAO ZHANG, R. A. FARRAR
Department of Mechanical Engineering, University of Southampton, Southampton SO17 1BJ, UK

This paper discusses the development of columnar grains in as-deposited C–Mn–Ni(-Mo) low-carbon low-alloy weld metals and the influence of the alloying elements, particularly nickel. It was found that the austenite columnar grain size prior to the γ–α transformation of the weld metals was mainly controlled by the alloying contents rather than HAZ grains adjacent to the fusion boundaries, although the latter determined the size of the initial columnar grain size. The addition of nickel initially depressed the prior austenite grain size and subsequently dramatically coarsened it. This was related to the nickel equivalent (Ni_{eq}) of the weld metals and the peritectic reaction during the solidification process. Small columnar grains were associated with a Ni_{eq} between 3.4 and 6.2% which resulted in a peritectic reaction when the weld melt solidified, whilst a Ni_{eq} higher than 6.2% produced very large columnar grains because the weld pool would directly solidify into austenite and have a subsequent continuous growth.

1. Introduction

In practical fabrication processes, low-carbon low-alloy weld deposits normally undergo solidification from the liquid state into δ-ferrite, which subsequently transforms to austenite, and then α-ferrite, bainitic or martensitic structures (with very rapid cooling conditions). The final weld metal microstructure is dependent upon many contributory factors associated with these processes.

It has been appreciated for some time that, besides the welding heat-input and chemical compositional effects, the structure of the prior columnar austenite grains is also critical in determining the final transformation microstructure of low-carbon low-alloy as-deposited weld metals [1–8]. Direct dilatometric and microstructural evidence has demonstrated that prior austenite grain size can influence considerably the γ–α transformation kinetics by varying the critical transformation temperatures and microstructural transformation rates, and thus the final microstructures and morphologies [6–8].

In low-carbon low-alloy weld metals, austenite grains normally form from the original δ-ferrite boundaries, and grow in a columnar manner. The austenite columnar grains formed are usually smaller than δ-ferrite columnar grains [9], indicating that more than one nucleation event occurs at the particular δ-ferrite grain boundary during the transformation. Usually, the austenite grains will cross the primary δ-ferrite grain boundaries. However, sometimes the δ and γ boundaries can be coincident, and under certain circumstances, the γ columnar structure may be appreciably larger than the prior δ-dendrite spacing [10, 11]. If the carbon level or the alloying element contents are sufficiently high, the austenite can be the first solid phase to form in the weld pool, and the γ columnar structure (which nucleates epitaxially at the fusion boundaries like δ-ferrite grains [12–15]) will grow directly from the melt. This process, however, is complicated in low-alloy welds, and because of experimental limitations there have not been any direct investigations of the δ–γ transformation. In the literature, it has been recognized that the development of the prior austenite columnar grains in weld metals is influenced by various factors, such as the oxygen content (thus the inclusion population and size distribution), the grain structure in the weld heat-affected zone (HAZ), and particularly the alloy element content, although the detailed contribution of the different factors or individual elements are either in debate or unclear. Referring to the role of weld metal chemical composition, it has been reported that carbon and some alloying elements might influence the columnar grain development and the prior austenite grain size [16–21]. The earlier work by Evans [16] and Taylor and Evans [17] showed that in C–Mn weld deposits, an increase in the carbon content led to a reduction of prior austenite grain size of the as-deposited weld metals. This was confirmed by the work of Svensson and Gretoft [18]. These authors and others also found that elements such as manganese, aluminium and titanium could also affect the γ grain development, but sometimes their results appeared inconsistent [16–20]. The effect of nickel was reported by Jones [21]. By adding nickel (from 0.5% to approximately 5.0%) in a series of C–Mn weld metals, he observed a significant influence on the columnar grain growth. The trends were that increased nickel led to large grains, although there were no quantitative results provided in the paper.

These results indicate that alloying elements have a considerable influence on the columnar grain development and the consequent prior austenite grain size, but there is a lack of systematic investigations in this field. Detailed studies are obviously needed to further our understanding of the matter. The current work, employing C–Mn–Ni and C–Mn–Ni–Mo low-alloy weld metals, attempts therefore to establish the direct effect of some alloying elements like nickel and manganese in influencing prior austenite columnar grain structure in weld deposits.

2. Experimental details

2.1. Weld metals and the preparation

The current investigation used either manual metal arc (MMA) or inert gas tungsten (TIG) welding, two series of C–Mn–Ni welds of two different oxygen levels, namely intermediate and low oxygen contents (coded as A0–A2, B1–B3, C1 and T1, T3), and one series of C–Mn–Ni–Mo welds (coded as M1 and M2) were employed.

The MMA welds were produced using metal powder electrodes (4 mm diameter), with the process conditions being 145 A, 18 V and a nominal heat input of 1.2 kJ mm^{-1}. Using three beads per layer (two beads for the first layer), the total number of runs was 23. The TIG weld deposits were prepared with 2.4-mm diameter wire and shielded by pure argon gas (99.99% Ar), the welding parameters being 180 A, 13 V and a nominal heat input of 1.8 kJ mm^{-1}. The welds were deposited with three beads per layer (two beads for the first layer) using a total of 50 beads. The interpass temperature was held at 150 °C. The welding was carried out on BS970-070M20 plate and in the horizontal position. The thickness of the plate was 19 mm.

In order to investigate the columnar grain growth at the earliest possible stage, a series of in-situ quenching experiments were carried out. Two welds (B1 and B3) were prepared by welding a single bead on a $10 \times 51 \times 300$ mm plate and quenching whilst the welding was still going on.

2.2. Chemical composition of the welds and metallography

The chemical composition of the as-deposited weld metals was examined by a Bausch & Lomb ARL-3560 Optical Emission Spectrometer using samples containing the full section of the weld (the actual target size was 16 mm in diameter) and checked by Quantovac methods. The oxygen and nitrogen contents of some welds (A1 and B1, M1 and M2, T1 and T3) were assessed by a Leco TC-136 machine. The chemical compositions of the weld metals are listed in Table I.

Metallographic specimens were extracted from each welded joint across the section. The specimens from the in-situ quenched welds were extracted longitudinally. After mounting in bakelite, they were polished to a 1 μm diamond finish, and then etched in 2% Nital. Optical examination was carried out on a Neophot II microscope. The prior austenite grain size measurement was obtained using the linear intercept method [22] and the grain size (i.e. the columnar grain width) was measured on the polygonal ferrite boundaries [23]. Sizing was done over a standard 100 mm linear line and the measurement was carried on 36 random fields for each individual weld.

3. Results and discussion

3.1. Columnar grain size of the welds and its variations with composition

The average columnar grain width of the weld metals is listed in Table II.

The data obtained show a very interesting change in the columnar grain size (width) with the variations of the nickel and manganese contents. In the MMA C–Mn–Ni welds, at higher manganese content ($\sim$1.6%), increasing the nickel content effectively depressed the prior austenite grain. When the nickel

TABLE I Chemical compositions of weld metals (wt %)

Weld Code	AWS specification	C	Mn	Si	S	P	Ni	Ti	Mo	Al	Cu	O	N
A0	A5.5/E7018-G	0.04	**1.57**	0.46	0.010	0.013	**0.03**	0.016	< 0.01	0.004	0.05	nd	nd
A1	A5.5/E8016-G	0.05	**1.68**	0.50	0.011	0.015	**0.95**	0.019	0.01	0.007	0.05	0.0254	0.0134
A2	A5.5/E8016-C1	0.04	**1.67**	0.49	0.009	0.013	**2.48**	0.018	< 0.01	0.005	0.06	nd	nd
B1	A5.5/E8018-C1	0.04	**0.85**	0.50	0.008	0.013	**2.56**	0.021	< 0.01	0.005	0.06	0.0310	0.0100
B2	A5.5/E8016-C2	0.06	**0.67**	0.53	0.008	0.013	**3.32**	0.024	< 0.01	0.006	0.06	nd	nd
B3	EC	0.05	**0.78**	0.66	0.008	0.015	**5.53**	0.025	0.01	0.007	0.07	nd	nd
C1	EC	0.03	**0.36**	0.55	0.008	0.013	**5.58**	0.023	< 0.01	0.006	0.08	nd	nd
M1	A5.5–81/E9016-G	0.06	**1.50**	0.37	0.009	0.015	**0.96**	0.018	0.23	0.003	0.06	0.0205	0.0054
M2	A5.5–81/E11016-G	0.05	**1.75**	0.31	0.009	0.016	**2.73**	0.020	0.20	0.004	0.06	0.0275	0.0069
T1	A5.28/ER80S–Ni1	0.08	**0.96**	0.53	0.014	0.010	**0.90**	< 0.002	0.02	0.010	0.17	0.0027	0.0115
T3	EC	0.09	**1.05**	0.21	0.005	0.010	**3.32**	< 0.001	0.02	0.011	0.20	0.0048	0.0174

bold type = design changes.
EC = experimental consumables.
nd = not determined.

TABLE II Columnar grain size of the as-deposited weld metals

Weld Code	Type	Average columnar grain width (μm)
A0		100
A1	MMA	110
A2		61
B1		69
B2	MMA	64
B3		215
C1	MMA	180
M1	MMA	115
M2		102
T1	TIG	94
T3		187

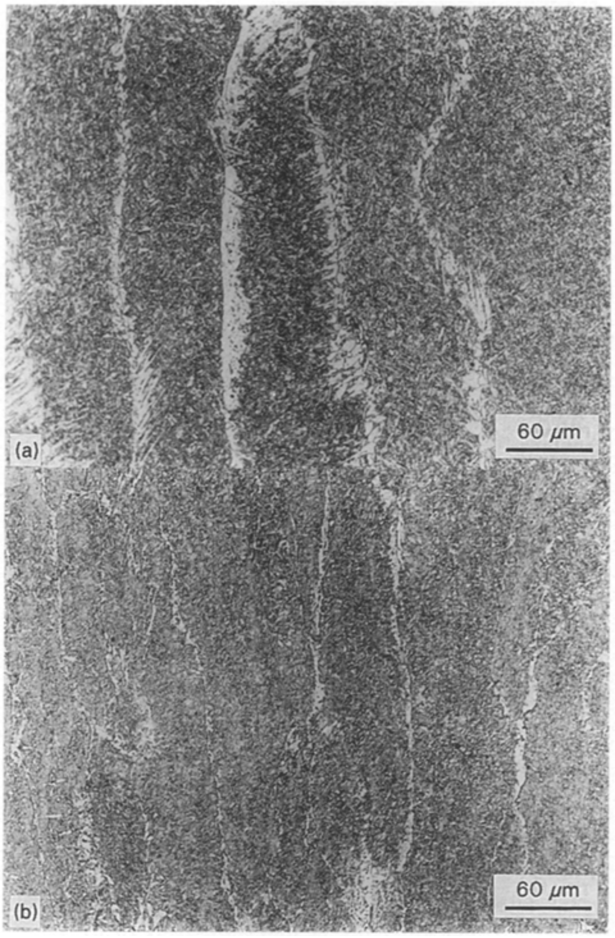

Figure 1 Columnar grain structure of the MMA C–Mn–Ni as deposited top beads (1.6% Mn) (a) A0 weld (0% Ni) (b) A2 weld (2.5% Ni).

concentration changed from 0 to 2.5%, the average width of the columnar grains at the top beads reduced by 39% (i.e. from 100 μm to 61 μm). With a lower manganese level (~0.7%) and 2.5% or 3.5% nickel contents, the width of the austenite grains were still refined to less than 70 μm. But when the nickel content increased to 5.5%, the columnar size dramatically increased by more than three times to greater than 200 μm. These columnar grain size changes are clearly illustrated in Figs 1 and 2, which demonstrate the microstructures of the as-deposited weld metals with changing nickel content at different manganese levels. In the MMA C–Mn–Ni–Mo welds, an increase in the nickel content also resulted in a reduction in the columnar size. In the case of lower oxygen content (TIG) deposits, a similar increase in the columnar grain size was observed and very large grains were produced when the nickel content was increased from 1.0% to 3.5%, as shown by Fig. 3.

From the data obtained, the effect of manganese on the columnar size is also obvious. The extent of its effect on the C–Mn–Ni welds depends synergistically upon the nickel level. At the 2.5% Ni level, a reduction of 0.9% in manganese (from 1.6% to 0.7%) increased the grain size by 13%, while at higher nickel level (5.5%), a drop of manganese from 0.7% to 0.3% resulted in a 16% depression of the grain size.

3.2. The columnar grain growth in the C–Mn–Ni weld metals

The data in Table II shows that the columnar width of the welds with the highest nickel content was much larger than those with a lower nickel content. In order to investigate the columnar grain development and growth behaviour of the weld metals with different nickel contents, detailed optical observation and quantitative assessment were carried out on the in-situ quenched welds.

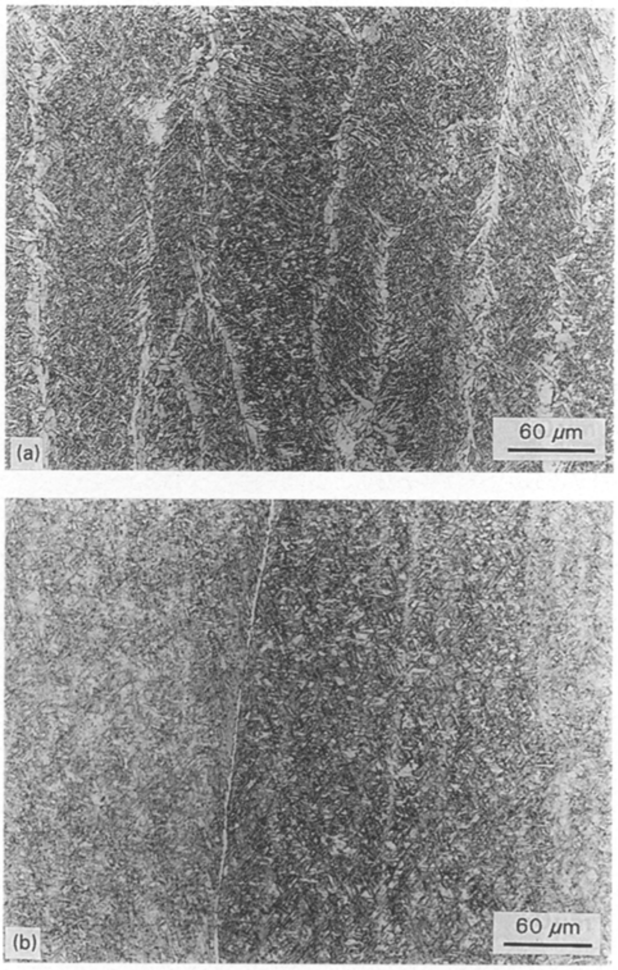

Figure 2 Columnar grain structure of the MMA C–Mn–Ni as deposited top bead welds (0.7% Mn). (a) B2 weld (3.5% Ni); (b) B3 weld (5.5% Ni).

Figs 4 and 5 illustrate the complete columnar macrostructures of the MMA C–Mn–Ni welds with either low (2.5%) or high (5.5%) nickel contents (Manganese level: 0.7%) which were obtained from the in-situ quenched B1 and B3 welds. These provide a complete picture of the columnar grain development from the fusion boundary between the HAZ regions and the welds. Table III lists the measured average columnar grain width corresponding to typical regions as marked in the figures, i.e. A, the regions in which the first columnar grains grew epitaxially out from the HAZ coarse grains; B, regions close to A, where the columnar growth was considered to be controlled by the characteristics of the weld metals rather than the grain size of the HAZ; C, upper centre of the welds, where the columnar grains reached their maximum size.

The results in Table III show that the columnar grain development of the two C–Mn–Ni welds was significantly different and an increased addition of nickel significantly promoted the expansion of the columnar grains. These are consistent with the quantitative results listed in Table II. In the first stage of the columnar grain development, i.e. when the grains nucleated epitaxially from the HAZ grains, the size of the grains formed in the welds were controlled by the HAZ grains. Detailed micrographs show that each columnar grain corresponded to a HAZ grain in both the low and high-nickel content welds (Fig. 6). As the grains grew into the centre of the welds, however, the situation became dramatically different. The columnar grain size of the lower nickel weld remained identical and the columnar grain width in the final region (C) was just a few microns larger than that of region A. With a higher nickel content deposit, on the other hand, the columnar width in region C was about 3.6 times of that in the initial region (A). These indicate that the columnar grains in the B3 weld experienced a significant lateral expansion during the formation process, whilst those in the B1 weld apparently did not.

Comparing the grain sizes in the different regions of the low and high-nickel content weld metals is also interesting. In the region A, the sizes of the two welds were almost the same; the grains in the region B of the B3 weld were about twice that of the B1 weld; in region C, the largest grain size difference was reached: the columnar grain width of the B3 weld was 132 μm larger than that of the B1 weld,

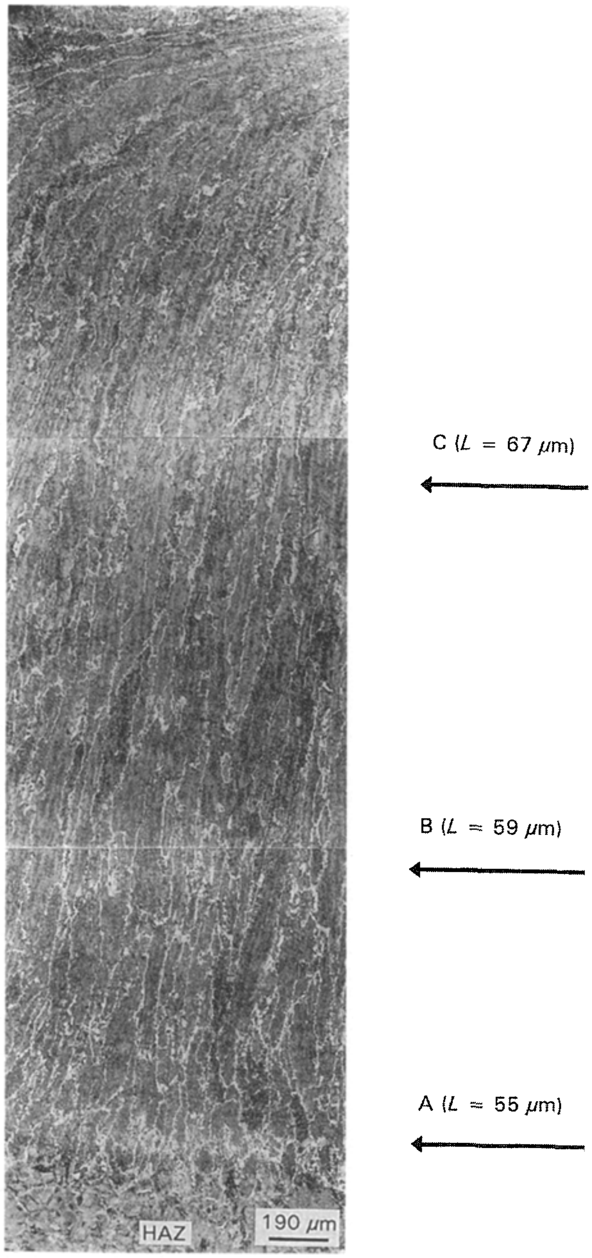

Figure 4 Columnar grains of the low nickel weld B1: 0.7% Mn, 2.5% Ni).

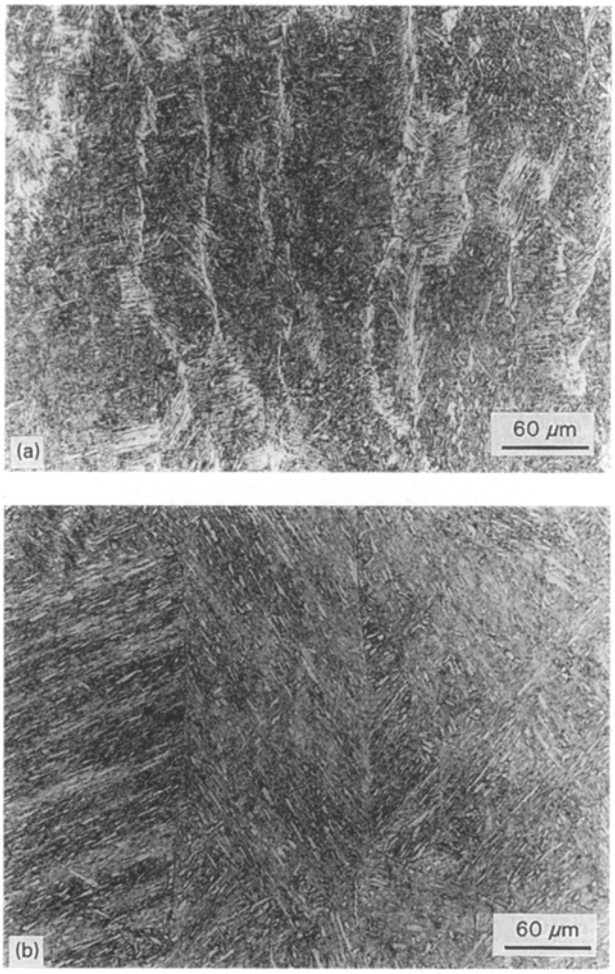

Figure 3 Columnar grain structure of the T1G C–Mn–Ni as-deposited top bead welds. (a) T1 weld (1% Ni); (b) T3 weld (3.5% Ni).

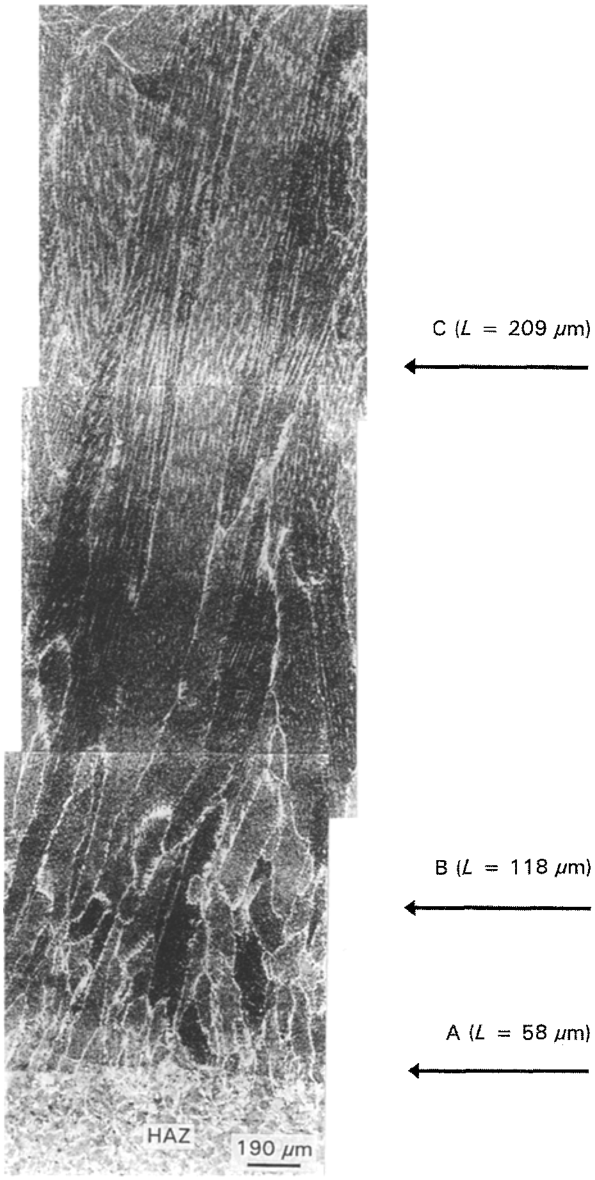

Figure 5 Columnar grains of the high nickel weld (B3: 0.7% Mn, 5.5% Ni).

TABLE III The columnar grain size in different regions of the welds

Weld Code	Type	Average columnar grain width (μm)		
		Region A	Region B	Region C
B1	MMA	55	59	67
B3	MMA	58	118	209

as illustrated by plotting the average grain size as a function of the distance from the fusion boundaries (Fig. 7). This indicates the profound influence of the nickel content on the weld metal columnar grain development.

3.3. The weld columnar grain development and the influence of nickel

The results described in the previous sections show that the variations of the columnar grain size due to the changes in alloying contents, particularly the nickel content, are very significant and the changing

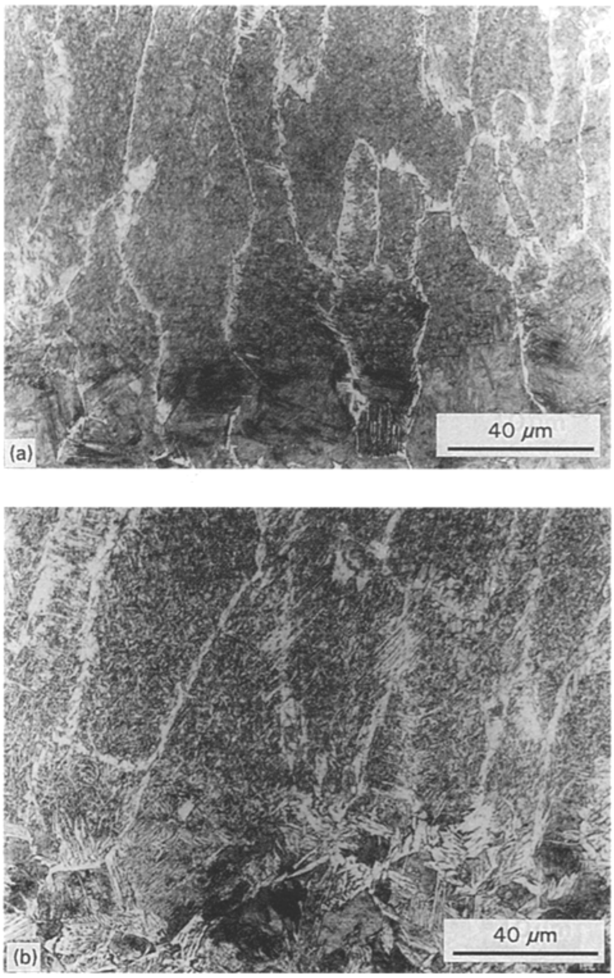

Figure 6 Columnar grains corresponding to the coarse grains of the HAZ. (a) Lower nickel weld B1 (2.5% Ni); (b) higher nickel weld B3 (5.5% Ni).

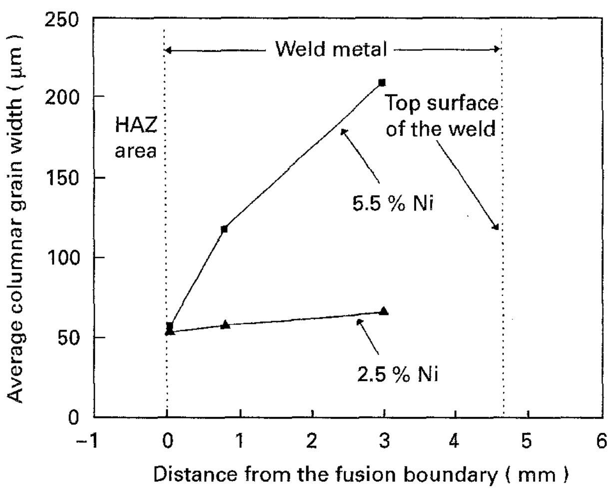

Figure 7 Effect of the nickel content on the columnar grain development (the B1 and B3 in-situ quenched welds, manganese level: 0.7%).

trends reversed as the nickel content progressively increased. In the intermediate oxygen level C–Mn–Ni welds, an increase in the nickel content initially refined the columnar grains. A similar refinement effect was observed in the MMA C–Mn–Ni–Mo welds when the nickel content changed from 1.0% to 2.5%. But when the nickel content was increased to its highest level, 5.5%, the grain size dramatically expanded to more than 200 μm. In the low oxygen content welds, a

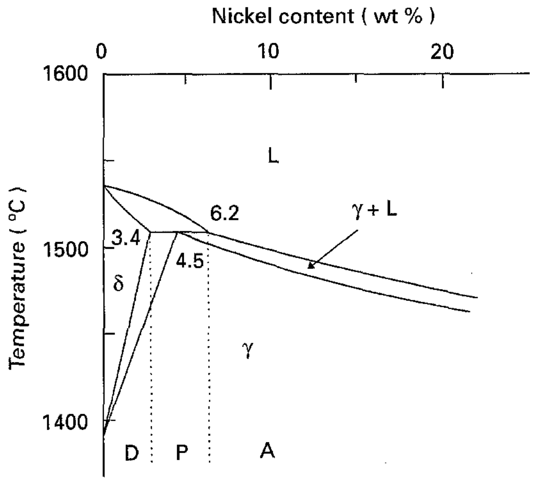

Figure 8 Fe–Ni equilibrium phase diagram (high temperature section) [24] (D: primary δ-ferrite solidification, P: peritectic effect, A: primary austenite solidification).

similar unusual large columnar grain structure occurred when the nickel content reached 3.5%: the average width of the columnar grains of the T3 weld was measured as high as 187 μm. These indicate that a high nickel content is associated with columnar grain coarsening. The effect of manganese appeared to depend upon the level of nickel; at low nickel contents, increasing manganese depressed the grain size whilst at high nickel level (5.5%), increasing the manganese level reinforced the effect of nickel, i.e. encouraged the formation of large columnar grains.

In order to obtain a better understanding of this phenomenon, the high-temperature section of the Fe–Ni equilibrium phase diagram is shown in Fig. 8 [24]. From the diagram, it can be seen that with increasing nickel content, the solidification sequence and the $\delta \rightarrow \gamma$ transformation process change: the simple δ-ferrite regime progressively decreases and when the nickel content is between 3.4 and 6.2%, a peritectic reaction will occur during the solidification process; when the nickel content is higher than 6.2%, liquid metal will solidify directly into austenite. These different solidification regimes are indicated as D, P and A respectively in the Fe–Ni phase diagram shown in Fig. 8. These changes in the solidification mode could have influenced the formation process of the columnar grains in the weld metals. In order to use this equilibrium phase diagram approach and since the weld metals studied were not a pure Fe–Ni alloy, the "nickel equivalent" ($Ni_{eq} = 30C\% + Ni\% + 0.5Mn\%$ [25]) which is normally used for the Schaeffler diagram was used. Table IV lists the corresponding Ni_{eq} for the weld metals studied.

The data in Table IV shows that the nickel equivalents of those weld metals (i.e. B3, C1 and T3) in which the columnar grain sizes were unusually large are all beyond the maximum nickel content, 6.2%, for the peritectic reaction to occur. It is therefore very likely that the liquid metal in the B3, C1 and T3 weld pools solidified directly to austenite without experiencing the peritectic reaction or $\delta \rightarrow \gamma$ transformation. This allowed the resultant columnar γ grains to grow extensively. Fig. 9 clearly illustrates the trend in the way the Ni_{eq} value influences columnar grain size of the C–Mn–Ni (and C–Mn–Ni–Mo)* MMA weld metals. The figure is plotted using the data of the present studies and the data reported by Harrison and Farrar [2, 23]. It shows that increasing Ni_{eq} initially refines and subsequently coarsens the columnar grains, the smallest grain size being produced in the weld metal with a Ni_{eq} between 3.4% to 6.2%. The largest grain size, on the other hand, will be expected when the Ni_{eq} is greater than 6.2%. From the figure, it can be clearly seen that the columnar grain width of the weld metals is distributed in three different size groups, namely intermediate, small and extra-large sizes, in accordance with their nickel equivalent values, and these three concentration ranges of the size groups are consistent with the Ni_{eq} regions corresponding to the different solidification modes (i.e. D, P and A regions, respectively). This strongly suggests that the addition of nickel influenced the columnar grain size by varying the weld pool solidification modes, i.e. from simple δ-ferrite solidification to peritectic reaction and then

TABLE IV Nickel equivalents of the weld metals studied

Weld Code	Type	Nickel equivalent, Ni_{eq} (%)
A0		1.99
A1		3.34
A2		5.35
B1	MMA	4.19
B2		5.49
B3		7.43
C1		6.96
M1	MMA	3.51
M2		5.11
T1	TIG	3.78
T3		6.55

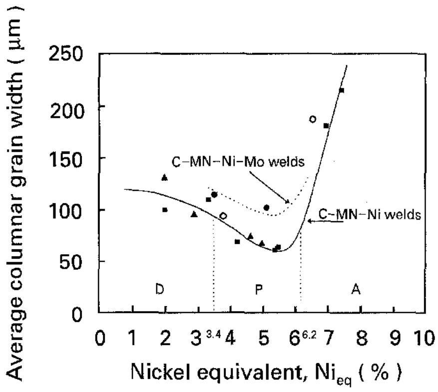

Figure 9 Influence of nickel equivalent on the columnar grain size (D: primary δ-ferrite solidification, P: peritectic effect, A: primary austenite solidification) (▲: data after Harrison and Farrar [2, 23], ■ ●: current results (MMA), ○: current results (TIG)).

*In reality, the molybdenum acts as a negative function in the Schaeffler Ni_{eq} formula.

simple γ solidification with a continuous increase in nickel content.

Further evidence supporting this conclusion comes from the results obtained from the in-situ quenched welds B1 and B3. As shown in Figs 4, 5 and 6, in the initial stages of the development of the columnar grains (i.e. nucleating on the fusion boundary at HAZ), the width of the grains in the two welds were very similar and the nucleated grains corresponded to the adjacent HAZ grains (i.e. epitaxial growth). It may therefore be concluded that at this stage the columnar grain size was determined by the size of the HAZ grains. As the columnar grains grew into the weld pools, however, the difference between higher and lower nickel welds emerged. Extra large columnar grains were found to be associated with the higher nickel content, as the solidification became dominated only by the weld pool chemical composition. These indicate that at the same welding conditions, the columnar grain size is controlled by the alloying content and confirm the previous conclusion, i.e. the high nickel content resulted in large columnar grains whilst a lower nickel content produced smaller columnar grains.

Comparing the columnar structure of the B1 weld with that of the B3 deposit reveals a further difference. From Fig. 5, it can be seen that the structure of the high nickel content weld consisted mainly of large complete columnar grains which developed continuously from the very early stage until they reached the vicinity of the top surface of the weld. In the lower nickel weld as shown in Fig. 4, on the other hand, the columnar structure appeared to be composed of a series of sequential, small and narrow grains. Obviously, these reflect the difference in their solidification features. Due to the presence of a temperature gradient within the welding pool, the solidification process would proceed from the fusion boundary to the upper surface of the weld. This means that for the B1 weld metal with a Ni_{eq} of 4.19%, the $L \rightarrow \delta$ transformation, peritectic reaction (i.e. $L + \delta \rightarrow \delta$), and $\delta \rightarrow \gamma$ transformation would occur sequentially and at different moments at various locations within the weld pool, and this would result in a large number of small grains. In the case of the B3 weld in which the Ni_{eq} was much larger than 6.2%, a simple growth process of the γ grains from the nucleation area into the welding pool would happen and a structure with fewer but much larger columnar grains then would be expected. Fig. 10 schematically illustrates the processes.

To sum up, in the C–Mn–Ni (and C–Mn–Ni–Mo) low-carbon weld metals, the smallest columnar grains are associated with the peritectic reaction where the Ni_{eq} is between 3.4 and 6.2%, whilst with a Ni_{eq} lower than 3.4%, intermediate columnar grains will be expected as the result of simple δ-ferrite solidification, and at a Ni_{eq} higher than ~6.2%, extra-large columnar grains will be produced because the simple austenite solidification mode will lead to a continuous growth of the columnar grains formed in the welds. These different columnar grain sizes will then influence the subsequent $\gamma \rightarrow \alpha$ transformation and will result in different final microstructures [5–7] and the mechanical performance of the weld metals. A recent investigation has revealed that very large columnar grains promoted intercolumnar fracture and thus contributed a deleterious effect to the weld metal toughness [27].

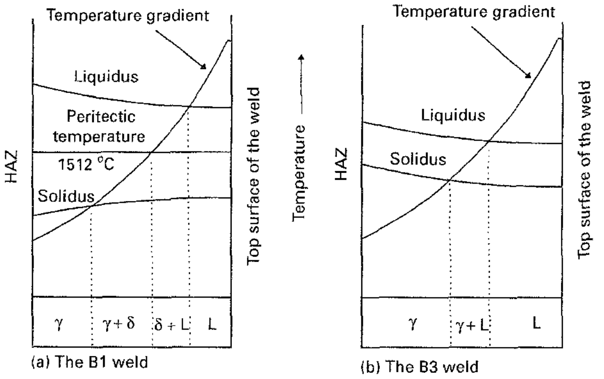

Figure 10 The difference of the solidification modes between the B1 and B3 weld ($Ni^{B1}_{eq} = 4.19\%$, $Ni^{B3}_{eq} = 7.43\%$)

4. Conclusions

The columnar grain development of the as-deposited low-carbon low-alloy C–Mn–Ni and C–Mn–Ni–Mo weld metals and the influence of nickel were investigated. The following conclusions were obtained.

1. The as-deposited columnar grain size prior to the $\gamma \rightarrow \alpha$ tansformation of the weld metals varied significantly. This was mainly controlled by the alloying element content of the welds, namely nickel and manganese. The effect of HAZ grains adjacent to the fusion line is very limited and operates within a small region.
2. The addition of nickel initially depressed the prior austenite columnar grain size and subsequently dramatically coarsened it. This is related to the nickel equivalent (Ni_{eq}) of the weld metals and the concurrent solidification process variations in the weld pool. Small columnar grains were associated with a Ni_{eq} between 3.4 and 6.2% which would result in a peritectic reaction when the weld melt solidified, whilst a Ni_{eq} higher than 6.2% would generate very large columnar grains as the liquid in the weld pool would directly solidify into austenite and have a continuous growth afterwards. If the Ni_{eq} was lower than 3.4%, the solidification in the weld pool would be a simple $L \rightarrow \delta$-ferrite process and intermediate prior austenite columnar grains would result.
3. The variations in the prior austenite columnar grain size will consequently influence the $\gamma \rightarrow \alpha$ transformation kinetics, the final weld microstructure and the mechanical performance of the weld metals.

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References

1. O. GRONG and D. K. MATLOCK, *Int. Metal Rev.* **31** (1986) 27.
2. P. L. HARRISON and R. A. FARRAR, *Metal Constr.* **19** (1987) 392R.
3. H. K. D. H. BHADESHIA, L.-E. SVENSSON and B. GRETOFT, *Acta Metall.* **33** (1985) 1271.
4. N. A. FLECK, O. GRONG, G. R. EDWARDS and D. K. MATLOCK, *Weld. J. Res. Suppl.* **65** (1986) 113s.
5. C. B. DALLAM and D. L. OLSON, *ibid.* **68** (1989) 198s.
6. R. A. FARRAR, ZHUYAO ZHANG, S. R. BANNISTER and G. S. BARRITTE, *J. Mater. Sci.* **28** (1993) 1385.
7. ZHUYAO ZHANG, PhD thesis, University of Southampton, Southampton (1994).
8. G. THEWLIS, *Mater. Sci. Technol.* **10** (1994) 110.
9. D. J. WIDGERY and G. S. SAUNDERS, *Weld. Inst. Res. Bull.* **16** (1975) 277.
10. G. S. BARRITTE, PhD thesis, University of Cambridge, Cambridge (1982).
11. R. C. COCHRANE, *Welding in the World* **21** (1/2) (1983) 16.
12. W. F. SAVAGE, C. D. LUNDIN and A. H. ARONSON, *Weld. J. Res. Suppl.* **44** (1965) 175s.
13. W. F. SAVAGE and A. H. ARONSON, *ibid.* **45** (1966) 175s.
14. W. F. SAVAGE, E. F. NIPPES and J. S. ERICKSON, *ibid.* **55** (1976) 213s.
15. W. F. SAVAGE, *Welding in the World* **18** (5/6) (1980) 89.
16. G. M. EVANS, *Weld. J. Res. Suppl.* **62** (1983) 313s.
17. D. S. TAYLOR and G. M. EVANS, *Metal Constr.* **15** (1983) 438.
18. L.-E. SVENSSON and B. GRETOFT, *Weld. J. Res. Suppl.* **69** (1990) 454s.
19. G. M. EVANS, *Oerlikon-Schweißmitt* **48** (124) (1990) 15.
20. F.-C. LIAO and S. LIU, *Weld. J. Res. Suppl.* **71** (1992) 94s.
21. K. W. JONES, MSc Thesis, Cranfield Institute of Technology, Cranfield (1986).
22. F. B. PICKERING, paper presented at conference *Quantitative and Qualitative Metallurgy*, 17–18 July 1975, Sheffield, The Institute of Metallurgical Technicians, Sheffield.
23. P. L. HARRISON and R. A. FARRAR, *J. Mater. Sci.* **16** (1981) 2218.
24. M. HANSEN, *Constitution of Alloys* (McGraw-Hill Company Inc., New York, 1958) p. 678.
25. J. F. LANCASTER, *Metallurgy of Welding* (George & Unwin, London, 1980).
26. P. L. HARRISON, PhD Thesis, University of Southampton, Southampton (1983).
27. ZHUYAO ZHANG and R. A. FARRAR, Microstructure and toughness of C–Mn–Ni low alloy weld metals and the influence of alloying elements, to be published. Submitted to *Welding Journal Research supplement*, American welding society, January 1995.

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