

2004

The influence of microalloying elements on the hot ductility of thin slab cast steel

Kristin R. Carpenter
kristinc@uow.edu.au

Recommended Citation

Carpenter, Kristin, The influence of microalloying elements on the hot ductility of thin slab cast steel, PhD thesis, Department of Materials Engineering, University of Wollongong, 2004. <http://ro.uow.edu.com.au/theses/161>

NOTE

This online version of the thesis may have different page formatting and pagination from the paper copy held in the University of Wollongong Library.

UNIVERSITY OF WOLLONGONG

COPYRIGHT WARNING

You may print or download ONE copy of this document for the purpose of your own research or study. The University does not authorise you to copy, communicate or otherwise make available electronically to any other person any copyright material contained on this site. You are reminded of the following:

Copyright owners are entitled to take legal action against persons who infringe their copyright. A reproduction of material that is protected by copyright may be a copyright infringement. A court may impose penalties and award damages in relation to offences and infringements relating to copyright material. Higher penalties may apply, and higher damages may be awarded, for offences and infringements involving the conversion of material into digital or electronic form.

The Influence of Microalloying Elements on the Hot Ductility of Thin Slab Cast Steel

A thesis submitted in fulfilment of the requirements for the
award of the degree

of

DOCTOR OF PHILOSOPHY

From

THE UNIVERSITY OF WOLLONGONG

By

KRISTIN CARPENTER

B.E (MATL)

DEPARTMENT OF MATERIALS ENGINEERING

2004

CANDIDATE'S CERTIFICATE

This is to certify that the work presented in this thesis is original and was carried out by the candidate in the Department of Materials Engineering, the University of Wollongong and has not been submitted to any other university or institution for a higher degree.

.....

Kristin Carpenter

Table of Contents

Table of Contents	I
List of Figures	VII
List of Tables	XIV
Synopsis	XV
Acknowledgements	XVIII

CHAPTER 1: LITERATURE REVIEW: THIN SLAB CASTING TECHNOLOGY	1
1.1 Introduction: Thin Slab Casting Technology	1
1.2 Benefits of Thin Slab Casting	2
1.3 Metallurgical Considerations of Thin Slab Casting	4
1.3.1 Liquid Steel Quality	4
1.3.2 Caster Design	5
1.3.3 Metallurgical Aspects	6
1.3.4 Thin Slab Casting with Liquid Core Reduction (LCR)	7
1.4 Rolling Aspects of Thin Slab Casting	10
1.4.1 Tunnel/Holding Furnace	10
1.4.2 Metallurgical Considerations During Rolling of Thin Cast Slabs	10
1.5 Thin Slab Casting and Hot Direct Rolling Processes	12
1.5.1 CSP- Compact Strip Production	12
1.5.2 ISP- In-line Strip Production	15
1.5.3 “Danieli” Thin Slab Caster	16
1.5.4 TSP- Tippins-Samsung Process	18
1.5.5 Conroll (VAI)	19
1.5.6 QSP- Quality Strip Production	20
1.6 Concluding Remarks	22

CHAPTER 2: LITERATURE REVIEW: HOT DUCTILITY	23
2.1 Introduction	23
2.2 Origin of Transverse Cracks During Continuous Casting	25
2.2.1 Role of Oscillation Marks in the Formation of Transverse Cracks	25
2.2.2 Effects of Chemical Composition on Oscillation Marks and Transverse Cracks	28
2.2.3 Influence of Carbon on Transverse Cracking	29
2.3 Relevance of the Hot Tensile Test to the Problem of Transverse Cracking	31
2.4 The Hot Ductility Curve	34
2.4.1 Intergranular Failure: Role of Ferrite films	35
2.4.2 Intergranular Failure: Precipitation Free Zones (PFZ)	37
2.4.3 Intergranular Failure: Grain Boundary Sliding	37
2.5 Effects of Microstructural Variables on Hot Ductility	39
2.5.1 Grain Size	39
2.5.2 Precipitation	40
2.6 Effect of Composition on Hot Ductility	42
2.6.1 Influence of Carbon on Hot Ductility	42
2.6.2 Effects of Sulphur on Hot Ductility	43
2.6.3 C-Mn-Al – Influence of AlN on Hot Ductility	46
2.6.4 C-Mn-Al-Nb - Influence of Nb on Hot Ductility	48
2.6.5 C-Mn-Al-Ti- Influence of Ti on Hot Ductility	50
2.6.6 Influence of Ti on Nb Microalloyed Steels	53
2.6.7 Role of Inclusions/Residuals on Hot Ductility	54
2.7 Effect of Test Variables on Hot Ductility	56
2.7.1 Cooling Rate	56
2.7.2 Strain Rate	57
2.7.3 Test Direction	57
2.7.4 Thermal History	58

2.8	Dynamic Recrystallisation and Deformation Induced Ferrite	61
2.8.1	Deformation Induced Ferrite	61
2.8.2	Dynamic Recrystallisation Verses Deformation Induced Ferrite	62
2.8.3	Modelling of Dynamic Recrystallisation	63
2.8.4	Commercial Implications on the Formation of Deformation Induced Ferrite and Dynamic Recrystallisation	64
2.9	Concluding Remarks	65
CHAPTER 3: EXPERIMENTAL		67
3.1	Introduction	67
3.2	Materials and Compositions	67
3.3	Tensile Testing: Gleeble Thermomechanical Simulator	69
3.3.1	Solution Treatment Procedure	71
3.3.2	Melting and Solidification Procedure (Direct Casting)	72
3.3.3	Simulation of Thermal Oscillations During Direct Cast Gleeble Tests	74
3.3.4	Determination of Reduction of Area (RA)	77
3.3.5	Radial Temperature Gradient Measurements	78
3.4	Metallography	80
3.4.1	Determination of Dendritic Segregation	82
3.4.2	Determination of Homogenisation Time	83
3.5	Scanning Electron Microscope (SEM)	84
3.6	Transmission Electron Microscope (TEM)	84
3.7	Dilatometry	86
3.8	Determination of Young's Modulus and Yield Stress	87
CHAPTER 4: RESULTS		90
4.1	Hot Ductility Curves	90

4.1.1	C-Mn-Al (Base Composition)	90
4.1.2	C-Mn-Al-Nb (Nb steel)	91
4.1.3	C-Mn-Al-Ti (Ti Steel)	92
4.1.4	C-Mn-Al-Nb-Ti (Nb-Ti Steel)	93
4.1.5	Summary of Hot Ductility Data	94
4.1.6	Hot Ductility Results from Cyclic Thermal History Tests	95
4.1.7	Comparisons of Hot Ductility Curves Under Solution Treatment Conditions	96
4.1.8	Comparisons of Hot Ductility Curves Under Direct Cast Conditions	97
4.2	Stress Strain Curves	98
4.2.1	Stress-Strain Curves- Occurrence of Dynamic Recrystallisation	100
4.3	Austenite Grain Size	102
4.4	Segregation Pattern and Determination of Secondary Dendrite Arm Spacings (SDAS)	103
4.4.1	Homogenisation Time	105
4.5	Metallographic Examination of Fracture Surfaces	106
4.5.1	C-Mn-Al	110
4.5.2	C-Mn-Al-Nb	110
4.5.3	C-Mn-Al-Ti	110
4.5.4	C-Mn-Al-Nb-Ti	111
4.5.5	Cyclic Temperature Oscillation Tests	111
4.6	SEM	112
4.6.1	SEM- Cyclic Temperature Oscillation Tests	116
4.7	TEM Results	118
4.7.1	C-Mn-Al	118
4.7.2	C-Mn-Al-Ti	120
4.7.3	C-Mn-Al-Nb	125
4.7.4	C-Mn-Al-Nb-Ti	128
4.7.5	TEM Particle Size Distribution Histograms	135
4.7.6	Influence of Particle Size on Hot Ductility	144

4.7.7	Influence of Interparticle Spacing on Hot Ductility	145
4.8	Dilatometry and Transformation Temperatures	146
4.9	Young's Modulus and Yield Stress	148
5.0	DISCUSSION	150
5.1	Influence of Precipitation on Hot Ductility	151
5.1.1	Influence of Precipitate Size on Reduction of Area (RA)	151
5.1.2	Influence of Volume Fraction of Precipitation	153
5.1.3	Influence of Location of Precipitation	153
5.2	Influence of Alloying Additions	154
5.2.1	Base Composition- C-Mn-Al	155
5.2.2	Influence of Niobium (Nb)	158
5.2.3	Influence of Titanium (Ti)	161
5.2.3.1	Influence of Dynamic Recrystallisation on the Ti Steel Under Solution Treatment Conditions	162
5.2.4	Influence of Niobium-Titanium (Nb-Ti)	166
5.2.5	Comparisons of Hot Ductility Under Solution Treatment Conditions	167
5.2.6	Comparisons of Hot Ductility Under Direct Cast Conditions (as-cast)	168
5.3	Influence of Cooling Rate	170
5.3.1	C-Mn-Al	170
5.3.2	C-Mn-Al-Nb	173
5.3.3	C-Mn-Al-Ti	173
5.3.4	C-Mn-Al-Nb-Ti	174
5.4	Influence of Temperature	177
5.4.1	Influence of Cyclic Thermal Oscillations on Hot Ductility of Nb-Ti Steels	179
5.5	Limitations of Solution Treatment and Direct Cast Laboratory Hot Tensile Tests	182
5.5.1	Influence of Segregation and Homogenisation Time	184

6.0	CONCLUSIONS	186
6.1	Recommendations and Suggestions for Future Work	189
6.2	Publications	191
7.0	REFERENCES	192
APPENDIX A: INTERPRETING ABNORMAL DATA POINTS AND EXPERIMENTAL SCATTER		206
APPENDIX B: HISTOGRAMS- PARTICLE SIZE DISTRIBUTION		210

List of Figures

Figure 1.1: Comparison of the layout for conventional and thin slab, direct rolling processing routes.	2
Figure 1.2: Thermal histories of conventional (thick) and thin cast slabs during casting.	6
Figure 1.3: Comparison of SDAS for conventional, 230mm thick slab and thin slab, 50mm thick [2].	7
Figure 1.4: Infinitely variable segment adjustment via position-controlled hydraulic cylinders for LCR at Thyssen Krupp Stahl.	8
Figure 1.5: Nucor's initial CSP plant layout.	13
Figure 1.6: Layout of CSP plant with twin strand casting at Thyssen Krupp Stahl's integrated works at Duisburg.	15
Figure 1.7: Layout of the ISP plant in Arvedi, Cermona.	15
Figure 1.8: Basic layout of the Tippins-Samsung Process (TSP).	18
Figure 1.9: Process configuration of Trico's medium thickness slab caster and compact hot strip mill.	21
Figure 2.1: Schematic diagram of a typical continuous casting machine and its inherent stresses [47].	24
Figure 2.2: Three main mechanisms for the formation of oscillation marks, A) Overflow, B) Overflow and remelting and C) Meniscus bent back..	26
Figure 2.3: Effect of heal time on oscillation mark depth.	28
Figure 2.4: The formation of thin sections in the shell wall in the mould for a) fine grain and b) Coarse grain steel.	30
Figure 2.5: Effect of carbon content on the depth of oscillation marks.	31
Figure 2.6: Schematic diagram showing the three characteristic ductility regions on a hot ductility curve. Reproduced from Ref [47].	34
Figure 2.7: Intergranular thin films of ferrite formed in C-Mn-Al-Nb steel when tensile tested at 800°C. Note the intergranular cracking along the ferrite films.	35
Figure 2.8: Intergranular microvoid coalescence of low alloy steels by deformation in low temperature γ region a-c, and α/γ duplex phase region, d-f. Where; a, d dynamic precipitation and strain concentration	

within soft layers along γ grain boundaries in initial stages of deformation; b, e microvoid formation by decohesion of precipitate/matrix interfaces; c, f coalescence of microvoids, resulting in ductile intergranular fracture of austenite.	36
Figure 2.9: Schematic models showing three scenarios that can lead to the formation of wedge cracks by grain boundary sliding; arrows indicate sliding boundary and sense of translation.	38
Figure 2.10: Hot ductility curves for 0.65%C steel showing effect of grain size.	40
Figure 2.11: Influence of particle size on reduction of area values for Nb, Ti and Nb-Ti steels in single-phase austenite (reproduced from Comineli [73]).	41
Figure 2.12: Change of activation energy for dynamic recrystallisation with carbon content, reproduced from Crowther et al. [75].	43
Figure 2.13: Effect of sulphur content on the minimum reduction of area for two cooling rates of 1°C/s and 4°C/s.	44
Figure 2.14: Hot ductility curves for high and low S, C-Mn-Al steels. The solution treatment temperature was 1430°C.	46
Figure 2.15: The precipitation-time-temperature (PTT) diagram for undeformed austenite (static precipitation), 5% pre-strain and dynamic precipitation for 0.05%C, 0.035% Nb steel.	49
Figure 2.16: Representative diagram depicting the general thermal histories simulated in references [61, 62, 103, 104].	61
Figure 2.17: Schematic diagrams showing (a) how the width of the ductility trough could be controlled by dynamic recrystallisation (DRX) and (b) how increasing the strain rate can reduce the depth and width of the trough.	64
Figure 3.1: Position of steel block samples taken from 230mm as-cast slab, prior to hot rolling.	68
Figure 3.2: Specimen set-up for tensile testing in the Gleeble machine.	70
Figure 3.3: Schematic representation of placement of thermocouples on tensile specimens.	71
Figure 3.4: Schematic diagram showing two cycles employed during the tensile tests (a) Solution treatment and (b) Direct casting.	72

Figure 3.5: Schematic diagram showing the thermal cycle used to simulate the thermal pattern experienced near the surface a thin slab during continuous casting, where Cycle 1 had an average cooling rate of 200K/min with oscillations of $\pm 50^{\circ}\text{C}$, Cycle 2 had an average cooling rate of 200K/min with oscillations of $\pm 100^{\circ}\text{C}$ and Cycle 3 had an average cooling rate of 100K/min with oscillations of $\pm 50^{\circ}\text{C}$	75
Figure 3.6: Plant data (F.E.M) of temperature profiles for a continuously cast slab, taken from the NS/BHP medium thickness caster. Slab dimensions were 90mm x 950mm and caster speed was 5.0M/min.	76
Figure 3.6: Temperature profile of surface and centre temperature measurements without Nextel sleeving.	79
Figure 3.7: Temperature profile of surface and centre temperature measurements with Nextel sleeving.	80
Figure 3.8: Thermo-cycle used to find $\gamma \rightarrow \alpha$ and $\alpha \rightarrow \gamma$ transformation temperatures by dilatometry.	87
Figure 4.1: Hot ductility curves generated for C-Mn-Al steel from solution treatment (Sol) and direct cast (Melt) conditions at cooling rates of 100K/min and 200K/min.	91
Figure 4.2: Hot ductility curves generated from the Nb steel for solution treatment (Sol) and direct cast (Melt) conditions at cooling rates of 100K/min and 200K/min.	92
Figure 4.3: Hot ductility curves generated from the Ti Steel for solution treatment (Sol) and direct cast (Melt) conditions at cooling rates of 100K/min and 200K/min.	93
Figure 4.4: Hot ductility curves generated from the Nb-Ti Steel for solution treatment (Sol) and direct cast (Melt) conditions at cooling rates of 100K/min and 200K/min. Included in Fig 4.4 are the RA points for the three cyclic thermal patterns, Cycles 1-3, generated at 900°C for the Nb-Ti Steel.	94
Figure 4.5: Comparison of reduction of area (RA) values at 900°C for different test conditions for the Nb-Ti steel	96
Figure 4.6: Hot ductility curves for solution treatment conditions for all four steels at cooling rates of 100K/min and 200K/min.	97

Figure 4.7: Hot ductility curves for direct cast conditions for all four steels at cooling rates of 100K/min and 200K/min.	98
Figure 4.8: Stress-strain curves as a function of temperature for the Ti steel at 100K/min under solution treatment conditions.	99
Figure 4.9: Stress-strain curves as a function of temperature for the Ti steel at 100K/min under direct cast conditions.	99
Figure 4.10: Stress-strain curves at 1000°C for indicated composition under solution treatment conditions and direct cast conditions.	101
Figure 4.12: Dendritic pattern after normalising where a) solution treatment conditions (Nb Steel), b) direct cast conditions (Nb-Ti Steel) and c) transition zone at edge of hot zone (Nb-Ti Steel).	104
Figure 4.13: Representative microstructures of fracture surfaces a) predominately ferrite (700°C), b-c) intergranular cracking along ferrite films at 750°C and 800°C,	107
Figure 4.13: (cont.) d) 850°C, showing ferrite films, e) 850°C, no ferrite films, f) Intergranular cracking in martensite (transformed from parent austenite) at 900°C	108
Figure 4.13: (Cont.) Intergranular cracking in martensite (transformed from parent Austenite) at (g) 950°C and (h) 1000°C, i) Voiding in martensite (transformed from parent austenite) at 1000°C.	109
Figure 4.14: Typical fracture surface, in the longitudinal direction, displaying sharp intergranular cracks, region A and large open cracks, region B.	111
Figure 4.15: Representative SEM photos of the fracture surface, where; a) Intergranular with ductile voiding at 700°C (Nb-Ti Sol), b) Intergranular failure via microvoid coalescence 750°C (Ti Melt), c) Higher mag of (b).	113
Figure 4.15: (Cont.) d-e) Intergranular failure via decohesion 800°C (Nb Melt), f) Mixed mode- intergranular failure and ductile voiding 900°C (Nb-Ti Melt).	114
Figure 4.15: (Cont.) g) Mixed failure mode where ductile mode is predominant 900°C (Ti Sol) and h) High temperature, high ductile fracture displaying deep voids 900°C (Al Sol).	115

Figure 4.16: SEM image of fracture surfaces for a) Nb-Ti direct cast conditions at 900°C and b) Nb-Ti direct cast- cyclic temperature oscillation test at 900°C	117
Figure 4.17: Typical morphology of rectangular precipitates (note rounding of corners) found under solution treatment conditions at temperatures 950°C and above, for Ti steel.	121
Figure 4.18: Typical morphology of fine circular precipitates found at 900°C and 850°C under solution treatment conditions for Ti steel, photo taken at 900°C at cooling rate of 200K/min.	121
Figure 4.19: Typical morphology of Fine TiN precipitates found under direct cast conditions, for Ti steel at 900°C. Precipitate size <30nm.	123
Figure 4.20: Typical morphology of fine circular Nb precipitates found at 850°C under direct cast conditions, for Ti steel. Precipitate size <15nm.	123
Figure 4.21: Typical morphology of fine Nb precipitates, formed in a row, taken at 900°C under direct cast conditions for Nb steel.	126
Figure 4.22: Typical morphology of rectangular and circular precipitates found under solution treatment conditions, taken at temperature 1000°C for Nb-Ti steel.	129
Figure 4.23: Typical morphology of star shaped (crucifix) precipitates found at 1000°C under solution treatment conditions for Nb-Ti Steel.	129
Figure 4.24: Typical morphology of fine precipitates found between 950°C and 850°C under solution treatment conditions, Nb-Ti Steel. (Photo taken at 950°C)	130
Figure 4.25: Typical morphology of fine precipitates found between 950°C and 850°C under direct cast conditions, Nb-Ti Steel. (Photo taken at 900°C)	130
Figure 4.26: Precipitate displaying caps (possibly higher in Nb than the core) on each corner of a larger rectangular precipitate, taken at 850°C under solution treatment conditions, Nb-Ti Steel.	131
Figure 4.27: Histograms displaying the number of particles per unit area as a function of size distribution for Nb-Ti, direct cast conditions (900°C), where a) 100K/min and b) 200K/min.	136

Figure 4.28: Histograms displaying the number of particles per unit area as a function of size distribution for Nb-Ti, thermal temperature oscillations, where a) cycle 1, b) cycle 2 and c) cycle 3.	137
Figure 4.29: Histograms displaying the number of particles per unit area as a function of size distribution for Nb-Ti, thermal temperature oscillations, where a) quenched at 900°C (T_{\min}) and b) quenched at 1100°C (T_{\max}).	139
Figure 4.30: Histograms displaying the number of particles per unit area as a function of size distribution for Nb-Ti, solution treatment conditions 100K/min, where a) 950°C, b) 900°C and c) 850°C.	141
Figure 4.31: Histograms displaying the number of particles per unit area as a function of size distribution for Nb-Ti, solution treatment conditions 200K/min, where a) 950°C, b) 900°C and c) 850°C.	143
Figure 4.32: Reduction of area (RA) as a function of average particle size (nm) for precipitates in Nb and Nb-Ti specimens in the single-phase austenite region for direct cast specimens. Included in the figure is the influence of precipitation in the two-phase region, 750-800°C, on RA.	145
Figure 4.33: Reduction of area as a function of interparticle spacing (1/IP) for the Nb-Ti steel, under direct cast conditions.	146
Figure 4.34: Yield stress as a function of test temperature for the Nb-Ti steel for all thermomechanical conditions.	149
Figure 5.1: Reduction of area as a function of average particle size (nm) for precipitation (at a test temperature of $\geq 850^{\circ}\text{C}$) in Nb and Nb-Ti specimens.	151
Figure 5.2: Precipitation arranged in a row, probably indicating the former position of an austenite grain boundary. (Nb-Ti steel, direct cast 975°C, 200K/min).	154
Figure 5.2: Differences in the stress-strain curves between solution treatment and direct cast conditions for C-Mn-Al specimens at 700, 800 and 900°C, where the arrows indicate the differences in stress of the curves at the same temperature.	157

Figure 5.3: Stress-strain curves for Ti specimens under solution treatment conditions at 900°C, where two curves, corresponding to specimens with high ductility, displayed evidence of dynamic recrystallisation (marked with arrows).	163
Figure 5.4a: Grain delineation at the edge of the ‘hot zone’ in a Ti specimen tested at 900°C, 100K/min, with low hot ductility. Note the large grain size.	164
Figure 5.4b: Grain delineation at the edge of the ‘hot zone’ in a Ti specimen tested at 900°C, 100K/min, with high hot ductility. Note the fine grain size.	164
Figure 5.5: Interparticle spacing as a function of temperature with the corresponding RA values plotted on the secondary Y-axis. Values were obtained from the Nb-Ti steel under direct casting conditions.	167
Figure 5.5a: Microvoid coalescence, typical of a C-Mn-Al specimen tested at 850°C at 200K/min (100x).	172
Figure 5.5b: Discrete voids, typical of a C-Mn-Al specimen tested at 850°C at 100K/min (50x).	172
Figure 5.6: Stress-strain curves for the Nb-Ti steel showing the changes in the stress-strain behaviour between 100K/min and 200K/min at 800°C, 900°C and 1000°C.	174
Figure 5.7: Interparticle spacing as a function of temperature for the cooling rates of 100K/min and 200K/min for the Nb-Ti steel under solution treatment conditions.	176
Figure 5.8: Typical example of clustering seen in Nb-Ti specimens (900°C) tested under solution treatment conditions, 200K/min.	176
Figure A: Hot ductility curves for the Nb-Ti steel showing experimental scatter and two abnormal data points.	206
Figure B: Stress-strain curves for 1000°C and 985°C, Nb-Ti steel 200K/min direct cast conditions, for normal and abnormal tensile tests.	207

List of Tables

Table 3.1: Chemical compositions of alloys.	69
Table 3.2: Transformation temperatures calculated from the chemistry of the respective steels.	73
Table 4.1: Summary of the hot ductility data taken from the reduction in area as a function of temperature curves	95
Table 4.2: Average austenite grain size measured as equivalent diameter.	102
Table 4.3: Measured and calculated Secondary Dendrite Arm Spacings (SDAS)	105
Table 4.4: Diffusion data for solutes in austenite, the diffusion coefficients of these elements at 1330°C in austenite and the homogenisation times.	105
Table 4.5: Summary of the mean particle size and particle number density of selected steels and thermomechanical treatments.	119
Table 4.7.2a: TEM results, C-Mn-Al-Ti 100K/min- Solution Treatment	122
Table 4.7.2b: TEM results, C-Mn-Al-Ti 200K/min- Solution Treatment	122
Table 4.7.2c: TEM results, C-Mn-Al-Ti 100K/min- Direct Cast	124
Table 4.7.2d: TEM results, C-Mn-Al-Ti 200K/min- Direct Cast	124
Table 4.7.3a: Nb Steel 100K/min- Direct cast	127
Table 4.7.3b: Nb Steel 200K/min- Direct cast	127
Table 4.7.4a: Nb-Ti 100K/min- Solution Treatment	132
Table 4.7.4b: Nb-Ti 200K/min- Solution Treatment	132
Table 4.7.4c: Nb-Ti 100K/min- Direct cast	133
Table 4.7.4d: Nb-Ti 200K/min- Direct Cast	134
Table 4.7: Interparticle spacing for the microalloyed steels at 900°C and 950°C, for direct cast conditions.	146
Table 4.8: Summary of Transformation Temperatures.	147
Table 4.9: Young's modulus values for all steels determined theoretically.	148

Synopsis

Experiments were performed on a Gleeble 3500 Thermomechanical Simulator to study the hot ductility behaviour of C-Mn-Al steel and the influence of Nb, Ti and Nb-Ti additions. The simple hot tensile test has been shown to correlate well to the problem of transverse cracking. Therefore, the principle aim of this research is to gain a greater understanding of transverse cracking during the straightening of continuously cast slabs. In particular, attention was paid to thin slab casting conditions.

Hot tensile test specimens were either solution treated or melted *in-situ* (direct cast) and cooled to the deformation temperature. Solution treatment tests simulated conventional casting, where slabs are cooled to room temperature then reheated prior to rolling. Direct cast tests simulated hot direct rolling conditions, where slabs are rolled directly after casting without being cooled below the austenite to ferrite transformation. Specimens were cooled to the deformation temperature at two cooling rates, 100K/min and 200K/min. The cooling rate of 100K/min corresponds to the average cooling rate experienced for a conventionally cast slab, 250mm in thickness. The cooling rate of 200K/min corresponds to the average cooling rate for thin-cast slabs, 50mm in thickness.

The development of the combination of thin slab casting with hot direct rolling requires hot ductility work to be performed under direct cast conditions and at higher cooling rates. Surface quality is of the utmost importance in thin slab casting so the elimination of transverse cracking is of prime economic importance. There are significant differences between as-cast (direct cast) and reheated (solution treatment) microstructures. In particular, changes in precipitate behaviour, austenite grain size, and the relationship between segregation and the position of austenite grain boundaries was investigated. An attempt has been made to determine what influence these differences in microstructure have on hot ductility.

Niobium bearing steels were selected for the reason that there are still problems with Nb steels regarding transverse cracking. Furthermore, there have been contradictory reports on the effects of Ti additions on the transverse cracking behaviour of Nb steels. There is

evidence from commercial practice that indicates that small additions of Ti improve the transverse cracking susceptibility of Nb steels. However, laboratory results generally show Ti additions have little influence or even a detrimental effect on hot ductility. Disparities in the thermal history simulated in laboratory tests to actual conditions near the surface of a continuously cast slab is the most likely reason for this discrepancy. Therefore, the influence of more closely simulating the thermal history conditions near the surface of a continuously cast slab was evaluated for the Nb-Ti steel.

Experimental work involved metallographic and scanning electron microscopy examination of the fracture surface. Transmission electron microscopy was used to determine precipitation characteristics. Tensile tests were conducted to determine mechanical properties, where reduction in area (RA) was used as a measure of ductility. The dendritic structure for direct cast and solution treatment specimens was revealed using a heat treatment procedure (normalising). Particle size was correlated to reduction of area for precipitates in the single-phase austenite temperature region. It was shown that particles below 15nm were detrimental to hot ductility. The relationship between interparticle spacing and reduction of area was also determined.

Microalloying additions to C-Mn-Al steels significantly widen the ductility trough but the depth remains similar. Low ductility was found at higher temperatures in the microalloyed steels due to intergranular failure as a result of grain boundary sliding in the austenite. Grain boundary sliding was favoured by the slow strain rate and was enhanced by fine microalloyed nitrides and/or carbides. Fine particles can pin austenite grain boundaries, allowing sufficient time for cracks to link together, ultimately causing intergranular fracture. Increasing the cooling rate generally lowered ductility further by promoting finer precipitation. The trough depth is similar in all steels as the formation of thin ferrite films controls ductility at the minimum trough position. The formation of thin films of ferrite allowed strain to concentrate in the softer ferrite phase and intergranular failure occurred due to microvoid coalescence.

Direct cast conditions always led to lower ductility compared to solution treatment conditions. This is explained in terms of differences in the microstructure, namely, grain size, segregation and precipitation. It is recommended that direct cast conditions should

be used to determine hot ductility behaviour as it more accurately simulates continuous casting conditions.

It was found that simulating the thermal history near the surface of a continuously cast slab, as opposed to cooling directly to the deformation temperature, improved ductility of the Nb-Ti steel. This improvement in ductility was attributed to the thermal history providing favourable conditions for coarsening of NbTi(C,N).

Acknowledgements

The research work reported in this thesis was carried out under the supervision of Professor Brendon Parker and Mr Chris Killmore, who was the associate supervisor. I express my appreciation for their guidance, encouragement and constant support during the course of this work.

I acknowledge the support of Professor Rian Dippanaar who assisted with supervision of my work during 2002-03 when Professor Parker relocated to the University of New South Wales in January 2002.

I thank Mr. Chris Killmore, Mr. Paul Kelly and Mr. Les Moore (BHP, Central Research Laboratories, Product Development Group) for their guidance. Their assistance in interpreting the data was of great help and their input in to the experimental program retained the projects industrial relevance. I also acknowledge Mr. Stuart Laird (BHP, Central Research Laboratories, Product Development Group) for his assistance in providing me with steel slab samples

I express my gratitude to the Australian Research Council and BHP Steel, Port Kembla for providing financial support. I also thank BHP Steel for providing me with the steel and the analysis of the chemical composition.

I thank Mr. Greg Tillman for his assistance in metallographic preparation and optical microscopy, Mr. Nick Mackie for his help in electron microscopy, Mr. Bob DeJong and Dr. Priya Manohar for their help with the Gleeble 3500 Thermomechanical simulator and their assistance in developing the experimental program. I also thank Dr. Priya Manohar for his help in reviewing my work. I thank Professor Paul Munroe (University of NSW, Australia) for his assistance in electron microscopy at the University of New South Wales.

Finally, I thank my parents, Mr. Trevor Carpenter and Mrs. Susan Carpenter for their loving support and encouragement during the progression of my thesis.