Effect of Intercritical Heat Treatment on the Microstructure and Mechanical Properties of Medium Mn Steels

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In the present work, the effects of intercritical annealing parameters on the microstructure and cold rollability (deformation rate and ratio) of “3rd Generation Advanced High Strength Steels (AHSS)” were studied. Hence, this paper discusses the formation of microstructures with different volume fractions of ferrite, martensite, bainite and retained austenite (RA). Two novel microstructures have been created, based on two levels of manganese (Mn): (i) ferrite plus martensite nucleated in austenite microstructure (FMNA structures), using Mn levels of 5 to 7 wt% and (ii) ferrite plus retained austenite duplex structure (FADP steels) for a Mn level of 10 wt%. In general, the ductility is a function of the amount of retained austenite and the strength is highly dependent on the martensite level.

KEY WORDS: Transformation Induced Plasticity (TRIP); Twinning Induced Plasticity (TWIP); Strain Induced Transformation (SIT); Stacking Fault Energy (SFE); Retained Austenite (RA).

1. Introduction

Over the last few decades, Advanced High Strength Steels (AHSS) have been very attractive to the automotive industry as a structural material for weight reduction and formability. These materials have a superior combination of strength and ductility, coupled with a relatively complex chemistry and microstructures. Nowadays, most of the AHSS possess a relatively high volume fraction of metastable austenite, which transforms under mechanical load.¹⁾ In particular, rapid work hardening occurs due to strain induced martensite or twins formed from the metastable (*i.e.* retained) austenite, which also increases the work hardening of surrounding ferrite; this gives rise to high uniform elongation by delaying the onset of necking. A volume expansion due to this transformation also contributes to increased ductility.^{2,3)} Therefore, it has been claimed that high manganese steels with either twinning or strain induced martensitic transformation offer extraordinary mechanical properties, which leads to improved crash worthiness properties for better passenger safety.^{4,5)}

The extraordinary mechanical properties of high manganese (20–25 wt%) steels are dependent on the composition and processing conditions, which are problematic.^{6,7)} High manganese content and oxidation behaviors of these materials are practical processing issues. Moreover, the competition with the lightweight materials makes it necessary to reduce the costs (*i.e.* amount) of alloying additions.⁸⁾ Hence, medium manganese (5 to 10 wt%) steels are currently one

of the materials being developed to harness the promise of 2nd generation steels.⁹⁾

In the present work, the effect of intercritical annealing parameters on the microstructure and consequently cold rolling properties of medium manganese AHSS were studied. Cold rolling studies and mechanical testing were performed to examine their potential as sheet products.

2. Experimental Procedures

2.1. Material

The steel ingots used throughout this work were supplied by CANMET-MTL (Hamilton, Ontario, Canada). Each melt was done in an induction furnace and cast in 100 kg ingots. The chemical compositions (in wt%) are given in **Table 1**. The rationale for these compositions is given in reference.¹⁰⁾ After casting, the ingots were annealed for homogenization and were cut and hot rolled to 5.2 mm thickness plates by TUBITAK – MRC (Gebze, Kocaeli, Turkey). This was followed by a heat treatment and cold rolling step down to approximately 2 mm sheets in McGill University Steel Processing Laboratory conditions.

Table 1. The chemical composition of steel samples.

Sample Nr.	C (%)	Mn (%)	Si (%)	Al (%)	Mo (%)	Fe
S-1	0.12	4.98	3.11	3.05	0.05	bal.
S-2	0.19	4.96	3.09	2.99	0.03	bal.
S-3	0.22	7.15	3.11	3.21	0.05	bal.
S-4	0.20	10.02	3.17	3.19	0.06	bal.

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2.2. Annealing

Figure 1 shows the annealing schedule before cold rolling at different times (Annealing time (t_{an}), from 10 to 30 min and temperatures (T_{an}) ranging from 900 to 1 200°C, which is in the intercritical area (*i.e.* the austenite plus ferrite two phase region). The specimens were cooled to room temperature by water quenching (WQ) and air cooling (AC). The average cooling rate for air cooling and water was about 13 and 83°C/sec respectively. A very low cooling rate (FC) of 0.5°C/s was also conducted on specimens annealed at 1 100°C.

Initially, the main objective was to maximize the retained austenite with controlled deformation induced transformation behavior. Therefore, the heat treatment sequence was designed to generate a ferrite plus austenite structure by an intercritical anneal. The annealing was followed by various cooling rates designed to create a wide variety of microstructures, as listed in **Table 2**.

The annealing temperatures for T_{an-1} were determined as 950°C for S1, 1 020°C for S2, 1 155°C for S3 and 1 240°C for S4 according to the highest FCC content of phase diagrams, as can be seen from **Fig. 2** and **Table 3**, which are calculated using the FactSage software with FSSel database.¹¹⁾

The second annealing temperature (T_{an-2}) was chosen as 900°C for all the steel compositions to generate finer polygonal ferrite and more stabilized retained austenite during air cooling. Finally, to evaluate the effect of grain size on the cold rolling properties of the two phase microstructures, high-temperature annealing (T_{an-3}) followed by slow cool-

ing (0.5°C/s) to room temperature, was conducted on the cold-rolled steel to obtain a coarse-grained ferrite plus pearlite microstructure.

2.3. Cold Rolling

Annealed plates were cold rolled at different deformation ratios with various reheating steps. Details of the rolling procedure and deformation process in each step before reheating are given in **Table 4**. The rolling direction was the same as for the hot rolling without reversing and a lubricant was applied on the rolls. The rolling speed was constant and calculated approximately around 35 revolutions per minute (RPM). Each reheating process was done for the steel sam-

Table 2. Detailed parameters used in various annealing process.

Sample Nr.	Ann. Nr	T_{an} (°C)	t_{an} (min)	Cooling	Microstructure
Steel 1	T_{an-1}	950	10	WQ	F + M + RA
Steel 1	T_{an-2}	900	15	air	F + M/B + RA
Steel 1	T_{an-3}	1 100	30	FC	F + P/ B + RA
Steel 2	T_{an-1}	1 020	15	WQ	F + M + RA
Steel 2	T_{an-2}	900	15	air	F + M/B + RA
Steel 2	T_{an-3}	1 100	30	FC	F + P/ B + RA
Steel 3	T_{an-1}	1 150	10	WQ	F + M + RA
Steel 3	T_{an-2}	900	15	air	F + M/B + RA
Steel 3	T_{an-3}	1 100	30	FC	F + P/ B + RA
Steel 4	T_{an-1}	1 200	10	WQ	F + M + RA
Steel 4	T_{an-2}	900	15	air	F + M/B + RA
Steel 4	T_{an-3}	1 100	30	FC	F + P/ B + RA

T_{an} : annealing temperature; t_{an} : annealing time; WQ: water quenching; FC: furnace cooling; F: ferrite, M: martensite, B: bainite, R. A.: retained austenite, P: pearlite

Table 3. FCC% content of steel compositions at elevated temperatures.

Sample Nr.	T_{an-1} (°C)	FCC (%)
Steel 1	950	12
Steel 2	1 020	25
Steel 3	1 150	42
Steel 4	1 200	62

Fig. 1. Heat treatment schedule.

Fig. 2. Isopleths phase diagrams of two phase region ($\alpha + \gamma$) with the intercritical annealing temperature of highest FCC content (a) S-1 and (b) S2, S3 and S4.

ples to give maximum deformation ratio (before crack formation).

2.4. Characterization

In this study, characterization of the samples was made by means of microstructural examinations and mechanical tests. The microstructural examinations were carried out on a Nikon L150 optical microscope, Philips XL-30 field emis-

sion scanning electron microscope (FE-SEM), Bruker D8 X-ray diffractometer and Philips CM200 200 kV transmission electron microscope (TEM). The volume percentage of the phases were calculated with the help of Clemex Captive image analyze software and an XRD intensity correlation method.¹²⁾ During the scan, the XRD detector position (2θ) was 30° and the apparatus was operated at 35 kV accelerating voltage and 45 mA beam current.

For optical microscopy, the following etchants were used to reveal the microstructure and differentiate the phases clearly.

- 2% Nital
- 2% Nital followed by 10% aqueous sodium metabisulfite ($\text{Na}_2\text{S}_2\text{O}_5$)
- LePera's Etchant (equal portions of 1% aqueous $\text{Na}_2\text{S}_2\text{O}_5$ and 4% picral)

Etching by sodium metabisulfite makes ferrite gray, bainite or martensite black, and retained austenite white.^{13,14)} With LePera's Etchant, martensite or retained austenite appears white, bainite and pearlite appears black, ferrite appears grey or yellow (due to carbon concentration of sur-

Table 4. Parameters of the cold rolling process.

Steels	S1	S2	S3	S4
D. R. (%) $\varepsilon_t = \left(\frac{t - t_0}{t_0} \right) \times 100$	10	9	21	34
N.P.	19	14	21	24
R.H.	4	4	2	2
Total Reduction ($\pm 5\%$) $\approx$	40	35	45	60

D. R.: Deformation ratio before reheating - ($\pm 5\%$); N. P.: Number of passes before reheating

R.H.: Total number of reheating stages

Fig. 3. The microstructures of annealed samples.

face).¹³⁾ In most cases, although grain boundaries are not strongly etched, phases were clearly observed.

The mechanical tests were done by means of tensile tests. Tensile test samples were machined according to ASTM E8 sub-size standard and performed on a hydraulic MTS machine with a strain rate 0.01 mm/sec.

3. Result

3.1. Microstructure

The microstructures of the steels after T_{an-1} , T_{an-2} and T_{an-3} annealing are given in **Fig. 3**. The volume percentage of the retained austenite was calculated by using image anal-

Fig. 4. Typical XRD patterns of (a) S2 which represents FMNA structure and (b) S4 for FADP structure.

Fig. 5. The microstructures of cold rolled samples.

yses techniques and XRD intensity correlation method. Typical XRD patterns of the given microstructures presented in **Fig. 4** and the quantified results of phase fractions, including those obtained by microscopy, are listed in **Table 5**.

For S1, S2 and S3, after T_{an-1} , ferrite, martensite and retained austenite are observed. After annealing with T_{an-2} , the grain size is much finer as a result of the lower annealing temperature and, mainly result of the lower cooling rate, a very small amount of bainite forms, which is also, nucleated inside the austenite grains. However, it can be easily seen from the micrographs that the cooling rates of both annealing schedules have led to considerable levels of martensite during cooling from the two phase region. Quantitatively (Table 5) there is not much difference between these two heat treatments, but there is a tendency towards a higher retained austenite for T_{an-2} . In the case of T_{an-3} , the slow cooling rate has generated coarse pearlite as opposed to martensite and the ferrite volume fractions are higher.

Steel S4 is quite different in that T_{an-1} leads to a duplex microstructure, which consists only of ferrite and austenite. In addition to this, a limited amount of annealing twins was detected in the austenitic phases. After T_{an-2} , the grain size is again much finer, and the duplex microstructure persists but apparently with more annealing twins. The T_{an-3} treatment seems to be no different from the T_{an-2} treatment. As will be discussed later, this seems to indicate that the austenite is much more stable in S4, which could be due to the Mn level. Quantitatively, T_{an-1} yields the highest volume fraction of retained austenite.

The microstructures of cold rolled samples are given in **Fig. 5**, typical XRD patterns of the given microstructures presented in **Fig. 6** and the quantitative values are listed in **Table 6**. The important feature is the behavior of the retained austenite. In the case of S1, S2 and S3, the retained austenite transformed to martensite (either α' or ε) due to the cold rolling procedures, except for S3 in the T_{an-1} condition, in which there appears to be no strain induced transformation of the retained austenite. For, the retained austenite for S1 and S2 in the T_{an-1} condition, the retained austenite is approximately halved, after cold rolling. For T_{an-2} , S1, S2 and S3 all undergo a strain induced transformation, with more retained austenite transforming to martensite.

For S4, cold rolling also transformed some of the retained austenite to martensite, as shown in Fig. 6 and Table 6, with more transformation occurring for T_{an-2} . Moreover, defor-

mation twins in the austenite were seen in all cold rolled samples, with twins and martensites occasionally seeming to co-exist in the same austenite grain.

Finally, **Fig. 7** show the microstructures of tensile tested samples. The deformation due to tensile testing seems to have less influence on the final microstructures, in contrast to the cold rolled specimens. But the quantified results of phase fractions (**Table 7**) show that in some cases austenite has transformed (to martensite). Typical XRD patterns of the given microstructures can be seen in **Fig. 8**.

3.2. Tensile Test Results

Figure 9 shows typical stress -strain diagrams of uniaxial tensile tests after annealing treatments of hot rolled samples. The overall mechanical properties measured by these tensile

Table 5. The volume fraction of phase for steel compositions after annealing process.

Steel Nr.	T_{an-1}			T_{an-2}			T_{an-3}		
	F _v %	A _v %	M _v %	F _v %	A _v %	M _v + B _v %	F _v %	A _v %	P _v %
S-1	74	15.2	10.8	76	17	4	84	—	16
S-2	63	9.1	27.9	60	14	24	86	—	14
S-3	51	17.1	31.9	57	20.5	22.5	87	1.7	11.3
S-4	43	53	4	48	47	5	47	47.5	5.5

*F_v%; Ferrite volume fraction; A_v%: Austenite volume fraction; M_v%: Martensite volume fraction; B_v%: Bainite volume fraction; P_v%: Pearlite volume fraction

Table 6. The change of RA volume fraction after cold rolling process.

	T_{an-1}		T_{an-2}		T_{an-3}	
	Before C.R.	After C.R.	Before C.R.	After C.R.	Before C.R.	After C.R.
	A _v %	A _v %	A _v %	A _v %	A _v %	A _v %
S-1	15.2	7.2	17	4.3	—	1.3
S-2	9.1	4.8	14	2.6	—	—
S-3	17.1	13.7	20.5	5.7	1.7	2.1
S-4	53	32.7	47	23.2	47.5	28.4

*F_v%; Ferrite volume fraction; A_v%: Austenite volume fraction; M_v%: Martensite volume fraction; B_v%: Bainite volume fraction; P_v%: Pearlite volume fraction

Fig. 6. Typical XRD patterns of cold rolled (a) S3 which represents deformed FMNA structure and (b) S4 for deformed FADP structure.

Fig. 7. The microstructures of tensile test samples.

Fig. 8. Typical XRD patterns of tensile test samples; (a) S1 represents for FMNA structure and (b) S4 for FADP structure.

tests are given in **Table 8** by means of maximum strength (σ_{UTS}) and total elongation ($\% \epsilon_{total}$) values. In general, it was observed that increasing annealing time and temperature decreased the strength of these materials significantly, but only slightly increased the ductility. After annealing at T_{an-1} and T_{an-2} , a tensile strength of 800–900 MPa and total

elongation of about 10 to 30% was obtained. Tensile strength decreased strongly after T_{an-3} for all steel compositions to 500–700 MPa but the total elongation remained largely unchanged (15–25%).

Table 7. The change of RA volume fraction after tensile test.

	T _{an-1}		T _{an-2}		T _{an-3}	
	Before T.T.	After T.T.	Before T.T.	After T.T.	Before T.T.	After T.T.
	A _v %	A _v %	A _v %	A _v %	A _v %	A _v %
S-1	15.2	6.2	17	8.6	—	—
S-2	9.1	7.5	14	5.8	—	—
S-3	17.1	12.5	20.5	17.3	1.7	1.6
S-4	53	42.3	47	45.1	47.5	30.9

*F_v%; Ferrite volume fraction; A_v%; Austenite volume fraction; M_v%; Martensite volume fraction; B_v%; Bainite volume fraction; P_v%; Pearlite volume fraction

Table 8. Tensile test results of annealed samples.

	Hot Rolled Tensile		T _{an-1} + Tensile		T _{an-2} + Tensile		T _{an-3} + Tensile	
	UTS	% ϵ_{total}	UTS	% ϵ_{total}	UTS	% ϵ_{total}	UTS	% ϵ_{total}
S-1	855	5.07	799	17	858	15	674	21
S-2	1008	3.06	828	17	941	11	561	16
S-3	1044	2.2	868	20	928	15	497	18
S-4	827	6	868	31	873	23	685	25

Fig. 9. Engineering stress – strain diagrams of steel samples; (a) S1, (b) S2, (c) S3 and (d) S4.

Fig. 10. The FactSage prediction for the change of (a) carbon and (b) manganese content in (c) austenite as a function of the temperature (depicted with EPMA results of C and Mn content in retained austenite).

4. Discussion

4.1. Effect of Composition on the Microstructure

The compositional differences appeared to have generated two classes of microstructure, a multiphase structure that comprises ferrite, martensite/bainite and retained austenite (S1, S2 and S3) and a duplex retained austenite plus ferrite microstructure (S4). The key difference seems to be the increase of Mn from that of S3 (about 7% Mn) to S4 (about 10% Mn).

Figure 10 represents the FactSage prediction of the change of austenite and its carbon and manganese content as a function of the temperature for the steel compositions. The FactSage predictions for the equilibrium structures certainly indicate a significant increase in austenite at corresponding annealing temperatures (Fig. 10), but this does not in itself explain the different room temperature microstructures, *i.e.* the relatively strong stability of the austenite formed during annealing at this level of Mn. The C levels of the austenite formed are important because these control the stability of the austenite; a higher C level indicates an austenite more likely to be retained. However, the FactSage predictions for C in the austenite during annealing, which are also illustrated in Fig. 10(a), show that S4 has the lowest C at all temperatures, but it has the highest retained austenite. On the other hand, the predicted Mn levels in the austenite at the annealing temperatures is much higher in S4 than the other alloys, and it may be that this is the critical alloying element related to retained austenite levels (Fig. 10(b)).^{9,10,15}

The levels of retained austenite in S1, S2 and S3 are similar to those of other medium manganese (5 to 7%) containing steels,^{10,15,16} in which significant quantities of retained austenite were not produced.

In order to validate the above predictions from FactSage, C and Mn levels in the retained austenite of S2, S3 and S4 were measured for T_{an-1} and T_{an-2} by EPMA using WDS detectors. The results are plotted in Figs. 10(a) and 10(b). Although T_{an-2} is followed by air cooling, which may lead to some partitioning of the elements during cooling, the general trend due to temperature and composition will be maintained. The measured results indicate that the FactSage predicted trends are matched, therefore reinforcing the hypothesis that Mn is the critical element in obtaining the significant increase in retained austenite from S3 to S4.

4.2. Effect of Annealing Temperature and Cooling Rate on the Microstructure

When the effects of T_{an-1} and T_{an-2} are compared for S1, S2 and S3, in contrast to the water quenched specimens, air cooled samples seem to be little affected, with regard to the volume fraction of the second phase, with change in intercritical annealing temperature. This is because, with decreasing intercritical annealing temperatures at T_{an-2} , although the volume fraction of austenite decreases, the carbon content of this austenite increases (Fig. 10), leading to more stabilized retained austenite with less martensitic transformation occurring during cooling. It can be seen from the EPMA - WDS elemental mapping of C for S2 after T_{an-1} and T_{an-2} annealing (Fig. 11), that the thermally transformed phase nucleating inside the retained austenite islands can be differentiated from retained austenite. There is a much greater difference in C levels between the RA and the nucleated phase within the austenite in the case of T_{an-2} . This suggests that the phase formed in the austenite due to T_{an-2} , *i.e.* after air cooling as opposed to quenching, is bainite rather than martensite.

For both air cooling and water quenching, the volume fraction of the secondary phases (austenite + martensite) was nearly constant irrespective of the annealing temperatures. Thus the microstructure of T_{an-1} and T_{an-2} resembles a TRIP steel phase formation, but with martensite mostly substituting for the bainite in T_{an-1} , and the martensite/bainite surrounded by austenite (Fig. 12). On the other hand, for the T_{an-3} annealing treatment, the very slow cooling rate leads to pearlite, no martensite and slightly coarser grains. In addition to this, in the case of S-1 and S-2, when the austenite is quenched from a low temperature, the non-equilibrium concentration of carbon in solid solution is higher, which should increase the SFE in the FCC phase.

In case of S-4, all annealing schedules resulted in a duplex microstructure of ferrite and austenite, with a limited number of annealing twins and very small amount of martensite in the austenitic phases. The most interesting aspect of S4 is that the different cooling rates did not make any significant difference on the amount of retained austenite. As can be seen from Fig. 13, S4 retained a significant level of austenite at room temperature even after furnace cooling. Note that FactSage predicts no austenite at equilibrium at room temperature. Moreover, the retained austenite volume

Fig. 11. The BSE images and EPMA-WDS carbon mapping of S2 after (a), (b) T_{an-1} annealing and (c), (d) T_{an-2} annealing.

Fig. 12. The microstructure of (a) classic TRIP steel composition (0.27C, 1.4Mn, 1.4Si),¹⁷⁾ (b) candidate 3rd Generation AHSS composition (0.23C-5.35Mn-3.5Al-2.89Si)¹⁵⁾ and (c) S3 composition (0.22C-7.15Mn-3.21Al-3.11Si) at T_{an-1} conditions.

Fig. 13. The EBSD map of S4 after T_{an-3} annealing: Inverse pole figure and phase distribution (ferrite and retained austenite).

fractions of S4 after T_{an-1} , T_{an-2} and T_{an-3} were more or less close to each other as was seen in Table 5.

4.3. Effect of Annealing Temperature and Cooling Rate on the Mechanical Properties

The results of tensile tests conducted on hot rolled and annealed plates and are given in Table 8 to show the evolution of mechanical properties at each processing stage. It can be seen that the mechanical properties of steels are not only connected to the increase of retained austenite also related with the transformed martensite volumes. As a general trend the higher the martensite content the higher is the strength but the lower ductility. On the contrary, with increasing Mn and the retained austenite content, higher ductility was observed.

In particular, the maximum elongation observed correlated with retained austenite as well as Mn (wt%) content, with S4 (FADP microstructure) demonstrating reasonably high strength. The highest strength was obtained with S2 (FMNA microstructure) in T_{an-2} annealing, which has the highest martensite volume fraction. The trend of decreasing strength and increasing ductility generally corresponds to increasing martensite volume fraction and decreasing Mn (wt%) content as well as retained austenite volume fraction (Fig. 14).

Indeed, the reason for the high elongation values in S4 or FADP structure is believed to be the strain induced transformation take place during deformation (either mechanical testing or cold rolling) (Fig. 5). On the other hand, when T_{an-1} and T_{an-2} compared to each other, for FMNA structure, the relationship between strength/ductility and structure might be due to a difference in the carbon levels of the martensites and/or bainite; *i.e.* there was a higher carbon level in the specimen intercritically annealed and air cooled samples (Fig. 11). Hence it can be claim that the stability of retained austenite mostly related with Mn content during thermal processing and as well as C content during deformation.

4.4. Effect of Cold Rolling

Cold rolling led retained austenite to strain induced mar-

Fig. 14. The change of (a) strain and (b) stress as a function of retained austenite volume fraction.

Fig. 15. Deformation twins in S4 after T_{an-1} annealing and cold rolling; (a) bright field TEM image and (b) weak-beam dark field TEM image with SAED pattern oriented along [011] zone axis (FCC).

tensitic transformation plus twinning in S4 microstructure. At cold rolling of S4 deformation twins are already present in more than 50% of the grains (Figs. 5 and 15 as well). Grains without mechanical deformation twins contain a high density of planar dislocation. It is believed that, the increase in the deformation ratio of S4 related with the formation of deformation twins in retained austenite grains. Since twin boundaries act as strong barriers to slip propagation, twinning sustain extra strain hardening effect.¹⁷⁾ Therefore, it

could be predicted that a gradual formation of deformation twins in retained austenite is necessary to improve cold rollability of medium Mn steels.

In general, T_{an-1} has shown better deformation ratio comparing to the T_{an-2} annealing process, although, S1, S2 and S3 did not reached the desired deformation ratios on both annealing processes. The increase in martensite volume fraction decreased the deformation ratio of these steels due to increasing volume fraction of harder phases. However, total reduction in thickness was increased to 30 to 40% with several reheating processes in each rolling schedule due to given deformation in each pass (Table 4).

5. Summary

In this study, intercritical heat treatments of these medium Mn steels have generated two distinct novel microstructures:

- (i) ferrite plus martensite nucleated in austenite microstructure (FMNA structures) (alloys S1, S2 and S3)
- (ii) ferrite plus retained austenite duplex structure (FADP steels) (alloy S4)

The FMNA structures are formed with steels having Mn levels between 5 and 7 wt%, by quenching from intercritical temperatures at which the austenite varied between about 25 and 45 vol%. The FADP structure was generated from steel with a Mn addition of about 10 wt%, intercritically annealed at a temperature at which the austenite was about 50 vol%.

The FMNA structures are heat treatable and could be converted into a tempered martensite (or partitioned structure) – FTMNA, or a TRIP type structure, replacing the martensite with bainite – FBNA structure.

The novelty of structures in this study can be explained by the stability of retained austenite of these alloys. All studies about medium Mn steels suggested that the general microstructure of 3rd generation AHSS consists of three components: martensite and austenite, which are separate phases, both inside a ferrite matrix.^{3,8,9,16,18–22} However, the FMNA microstructure consists of martensite nucleated *within* austenite, and the FADP structure is ferrite plus retained austenite with no other non-equilibrium phase (*i.e.* martensite or bainite) present.

Generally, the strength and ductility of these steels are affected by several microstructural characteristics, in particular, volume fraction of the retained austenite, its carbon content, plasticity of the metastable phases during deformation, and the ferrite characteristics (grain size). Among these factors, the first three are believed to play the most impor-

tant role in controlling the medium Mn steel property. Finally, the results show that the ductility of these medium Mn steels increases with the retained austenite fraction of the microstructure, and the strength increases with martensite volume fraction.

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