Processing Alternatives for Optimized Strength and Toughness Balance of ASTM A709 HPS 70W Composition

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Abstract

ASTM A709 HPS 70W grade plates are intended for bridge construction and are usually processed in quenched and tempered condition. In an attempt to extend its application in tank car manufacture some alternate processing routes were explored for enhanced toughness and strength properties. The processing was primarily aimed at strengthening by grain size refining through controlled rolling and toughness control through low temperature heat treatment such as tempering. Slabs of HPS 70W composition were hot rolled to plates of different thicknesses up to 1” and then heat treated to impart improved mechanical properties. Normalized rolling followed by tempering was found to result in significant improvement in toughness and strength properties. Experimental findings indicating best combination of heat treatment have been presented. Results demonstrated possible opportunities for use of this steel for tank car applications.

Introduction

Railroad tank cars are used to transport petroleum products, liquefied gases and hazardous chemicals such as chlorine and anhydrous ammonia. Cars that are used to carry hazardous materials are designated as pressure cars and demand enhanced resistance against puncture, dent, and rupture due to impact. Primary material property requirement is increased fracture toughness. Conventionally ASTM A516-70 and AAR TC 128B steels are used for railroad tank cars. However, with the need for increased safety, it is perceived that TC 128B steel will not be able to significantly reduce fracture probabilities and steels needs to be developed with comparable strength and higher toughness [1]. In this perspective, there is a need to evaluate alternatives to TC 128 Grade B with comparable strength and improved puncture resistance. Alloy designs are focused on (i) reducing the carbon content to lower carbon to improve weldability along with base metal and HAZ toughness, (ii) lowering sulfur to increase upper shelf energy and (iii) alloying to maintain the minimum required strength and improve toughness.

One of the possible candidate steels for future tank car application is ASTM A709 Grade HPS 70W. “High Performance Steels” were developed with minimum yield strengths of 70 ksi (485 MPa) and 100 ksi (690 MPa) for bridge construction [2-4], and are so designated as they combined higher strength, good toughness characteristics, and reduced preheat requirements for welding. ASTM A709 HPS 70W steel is a “high performance steel” with a minimum yield strength of 70 ksi (485 MPa) and it is produced in either “quenched and tempered” or “thermo-mechanical control processed” condition for bridge construction. Chemical composition and mechanical property requirements for this steel are given in Table 1 in comparison to AAR TC 128B [3]. Table 2 is a summary of mechanical property data of plates produced at BH, some with TMCP and others with Q&T route [3]. This summary indicates that very similar characteristics can be achieved in this steel with either process. Arcelor Mittal USA regularly produces this steel in plate form at Burns Harbor. Based on chemical composition and obtained mechanical properties, this steel provides an excellent research opportunity to explore various hot rolling process variations on the development of structure property relations. It is theorized that a combination of hot rolled processing and a subsequent mild heat treating process will result in significant property improvement and therefore should be studied to examine the suitability for “critical service” tank car manufacturing application.

In the present study therefore, the composition of ASTM A709 HPS 70W was chosen as starting material. Slabs from this composition were hot rolled using different combinations of finish rolling temperatures and subsequently heat treated to result in structure-properties that may be attractive to some segments of tank car manufacturing as an alternate to TC-128B.

Table 1 Chemistry and mechanical property requirements for tank TC 128B and HPS 70W steels.

Requirements		HPS 70W	AAR TC 128B
Chemistry (wt.%)	C	0.11 max	0.24 max
	Mn	1.10-1.35	1.00-1.65
	P	0.020 max	0.025 max
	S	0.006 max	0.015 max
	Si	0.30-0.50	0.15-0.40 (≤3/4”); 0.15-0.50 (>3/4”)
	Cu	0.25-0.40	0.35 max
	Ni	0.25-0.40	No limit
	Cr	0.45-0.70	No limit
	Mo	0.02-0.08	No limit
	V	0.04-0.08	0.080 max
	Al	0.01-0.04	0.015 – 0.60
	N	0.015 max	0.010 max
	Cu+Ni+Cr+Mo		0.65 max
Mechanical Properties	YS, ksi	70 min	50 min
	TS, ksi	85-110	81-101
	Elongation (2”), min	19 %	22 %
	CVN, ft-lb	L 25/35 @-10F	T 15 @ -30F

Table 2 Mechanical property in 0.5-2” thick HPS 70W plates produced at BH plate mill.

Process Route	YS, ksi	TS, ksi	LCVN @ -25 F, Ft-lb
TMCP	78.6	96.0	129
Q&T	83.9	95.9	141

Material and Experimental Procedures

Three slabs from three different heats of HPS 70W composition were selected for the experimental study. The chemical compositions of the slabs are given in Table 3 which are almost similar. Composition of a TC128B plate is also shown in Table 3 for mechanical property comparison. The slabs were hot rolled to three different final thicknesses. Table 4 shows the transformation and weldability characteristics of the steels. Chemistry design of the HPS 70W composition is such that the Ar_3 transformation temperature is significantly lowered due to the Cu-Ni-Cr-Mn additions. A low Ar_3 provides sufficient workable range for austenite conditioning and also inhibits coarsening of transformed ferrite grains during cooling through γ–α transformation range. Three different processing routes were selected for the three slabs in the present study to examine structure-property development.

1. slab A was hot rolled with a controlled finish temperature of 1300 ^{o}F – this low temperature rolling ensures strengthening through introduction of dislocation substructure or subgrain formation in the ferrite.
2. slab B was hot rolled to a controlled finish temperature of 1600 ^{o}F (termed as normalized rolling) to take advantage of austenite grain refining just above Ar_3 temperature and
3. slab C was hot rolled and then normalized at 1650 ^{o}F for 1 hour – to examine property development through conventional hot rolling and processing.

A finish rolling reduction of 50-60% was ensured for all the slabs. The hot rolling details are given in Table 5.

Table 3 The chemical composition of HPS 70W steel (wt. %)

Slab	C	Mn	P	S	Si	Al	Ti	Cu	Ni	Cr	Mo	V	N
A	0.09	1.26	0.012	0.004	0.36	0.022	0.002	0.30	0.36	0.63	0.046	0.070	0.006
B	0.09	1.25	0.013	0.005	0.40	0.032	0.002	0.31	0.37	0.56	0.049	0.065	0.005
C	0.08	1.22	0.011	0.004	0.40	0.037	0.003	0.31	0.36	0.55	0.054	0.065	0.004
TC 128B	0.22	1.39	0.01	0.004	0.33	0.027	0.003	0.019	0.01	0.17	0.058	0.056	0.006

Table 4 Transformation and weldability characteristics

Steel	Ar_3, ^{o}F [5]	Ae_3 (from Thermo-Calc), ^{o}F	CE*	P^*_{cm}
A	1357	1512	0.49	0.23
B	1366		0.48	0.23
C	1369		0.47	0.23

$$CE = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15} \;;\; P_{cm} = C + \frac{Mn + Cu + Cr}{20} + \frac{Mo}{15} + \frac{V}{10} + \frac{Si}{30} + \frac{Ni}{60} + 5B$$

Table 5 Hot rolling parameters used for the present study

Plate	Thickness, in.	Reheat Temp, ^{o}F	Finish rolling Temp., ^{o}F	% Reduction during FR
A	1”	2250	1300	57
B	0.625”	2250	1600	50
C	0.75”	2250	~1750 ^{o}F	50

The hot rolled plates A & B were further tempered at 1050 and 1200 ^{o}F for 1 hour and plate C was normalized at 1650 ^{o}F for 1 hr before tempering at the above temperatures. Samples were collected from as-rolled, normalized and tempered plates of A, B and C for mechanical property and microstructure evaluation. Additionally, plate samples were collected from hot rolled plates and tempered at temperatures of 840, 930, 1020, 1110, 1200 and 1247 ^{o}F for one hour to determine tensile and Charpy impact property evolutions with tempering temperature.

Metallographic samples were prepared from hot rolled and tempered plates to examine microstructural changes using optical and scanning electron microscope.

Results

Mechanical Properties of Plates

Table 6 summarizes the tensile properties of full thickness plate samples after hot rolling and tempering at 1050 and 1200 ^oF for one hour. Yield stress values were obtained at 0.5 % extension under load (EUL). Plate A, which was finish rolled at 1300 ^oF recorded a very high yield stress value. Both the yield and tensile strength decreased with increasing tempering temperature. In contrast, the yield stress increased significantly for plates B and C after tempering with a small decrease in the tensile strength. The yield stress values changed marginally with increasing tempering temperature.

To further examine the present steels' response to tempering temperatures, tensile samples were tempered at temperatures of 840, 930, 1020, 1110, 1200 and 1247 ^oF and then tested till fracture. Figure 1 summarizes the yield stress, tensile stress and elongation (measured with 1" extensometer) values for the plate sample B tested at different tempering temperatures.

The yield stress and elongation values showed a marginal but gradual increase with tempering temperature from 840 to 1247 ^oF. The tensile strength drops with the increase in tempering temperatures.

All the plate samples manifested a continuous yielding behavior when tested after hot rolling. The yield point returned after tempering in all the samples.

Table 6 Tensile properties of plates after TMCP and tempering treatment.

Plate	Thickness	Finish Rolling ^oF	Processing	Tensile Properties		
				YS (0.5% EUL), ksi	UTS, ksi	El(8"), %
A	1"	1300	As-Rolled	90.5	95.5	20
			Tempered at 1050 ^oF for 1 hr.	80	94.2	18
			Tempered at 1200 ^oF for 1 hr.	77.5	89.4	19
B	0.625"	1600	As-Rolled	60.1	87.5	21
			Tempered at 1050 ^oF for 1 hr.	70.6	84.6	21
			Tempered at 1200 ^oF for 1 hr.	71.4	82.6	23
C	0.75"	~ 1750 ^oF	As-Rolled and Normalized	46.8	78.4	30
			Tempered at 1050 ^oF for 1 hr.	57.4	74.9	25
			Tempered at 1200 ^oF for 1 hr.	60.2	75.4	30
TC 128B	0.618"	~1730 ^oF	As-Rolled, Normalized and Stress Relieved	60.1	82.0	22

Charpy V-notch impact toughness was assessed for temperatures of 0, -30 and -50 ^oF for the hot rolled and tempered plates. Tables 7&8 present longitudinal and transverse CVN impact energy values for the as-rolled plates and plates tempered at 1200 ^oF for one hour after hot rolling. It is seen from Tables 7&8 that in all the plate samples tempering increased the impact toughness values significantly at all tested temperatures compared to that obtained in as-rolled plate samples. Maximum toughness values were obtained for the normalized and tempered plate sample C at all tested temperatures. These Charpy specimens could not be fully broken even at -30 ^oF. Plate samples from A and B also revealed a high degree of notch toughness after tempering even at a test temperature of -50 ^oF compared to the as-rolled condition.

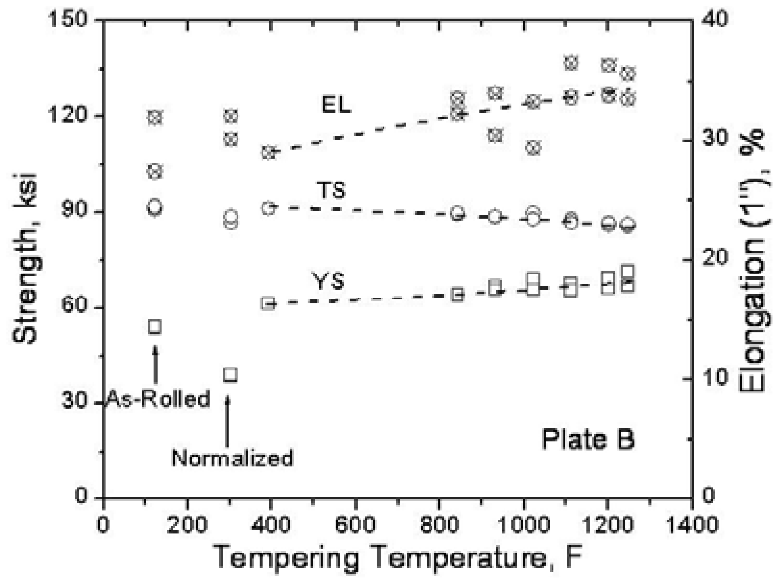


Figure 1 Influence of tempering temperatures on the transverse yield (0.5 % strain), tensile strength and elongation (1" GL) values for plate sample B.

Table 7 Longitudinal CVN impact energy results for plate samples in different processing conditions.

Plate	Thickness, in.	Test Temp., ^{o}F	As-Rolled Plate, Longitudinal CVN, ft-lbs			Tempered at 1200 ^{o}F for 1 hr., Longitudinal CVN, ft-lbs		
			1	2	3	1	2	3
A	1	0	156	168	171	206	188	190
		-30	120	142	142	194	190	191
		-50	164	150	166	183	200	180
B	0.625	0	92	98	94	190	152	116
		-30	60	56	45	178	144	142
		-50	49	29	40	101	136	148
C	0.75		After normalizing at 1650 ^{o}F Longitudinal CVN, ft-lbs			Tempered at 1200 ^{o}F for 1 hr., Longitudinal CVN, ft-lbs		
		0	120	97	81	278	280	282
		-30	46	51	46	282	266	272
		-50	76	74	64	270	258	256

Table 8 Transverse impact CVN energy results for plate samples in different processing conditions.

Plate	Thickness, in	Test Temp., ^{o}F	As-Rolled Plate, Transverse CVN, ft-lbs			Tempered at 1200 ^{o}F for 1 hr., Transverse CVN, ft-lbs		
			1	2	3	1	2	3
A	1	0	126	124	124	148	144	144
		-30	98	77	95	119	134	128
		-50	102	48	81	134	130	28
B	0.625	0	58	71	62	120	117	119
		-30	42	43	40	117	126	128
		-50	35	38	36	73	66	97
C	0.75		After normalizing at 1650 ^{o}F Transverse CVN, ft-lbs			Tempered at 1200 ^{o}F for 1 hr., Transverse CVN, ft-lbs		
		0	76	42	68	286	280	228
		-30	47	52	19	220	290	286
		-50	77	22	60	302	302	306

The Charpy impact notch toughness values obtained in plate sample A after different tempering temperatures are plotted in Figure 2. It is seen that the impact toughness initially increases with tempering temperature and remains almost constant for tempering temperatures between 1000 °F to 1200 °F. This was also evident in the toughness results of samples taken from plate B and C as shown in Figure 3 for a test temperature of 0 °F.

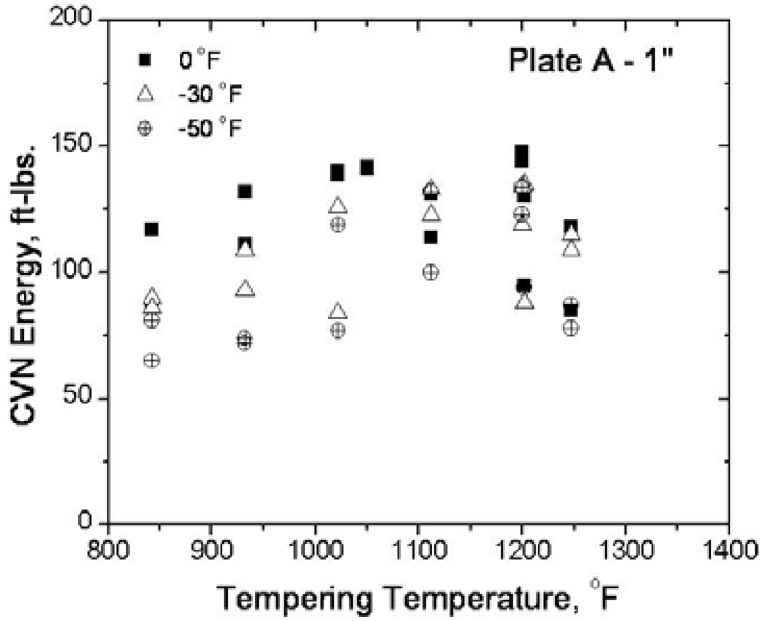


Figure 2 Charpy V-notch transverse impact energy values for Plate A as a function of tempering temperature for three different test temperatures.

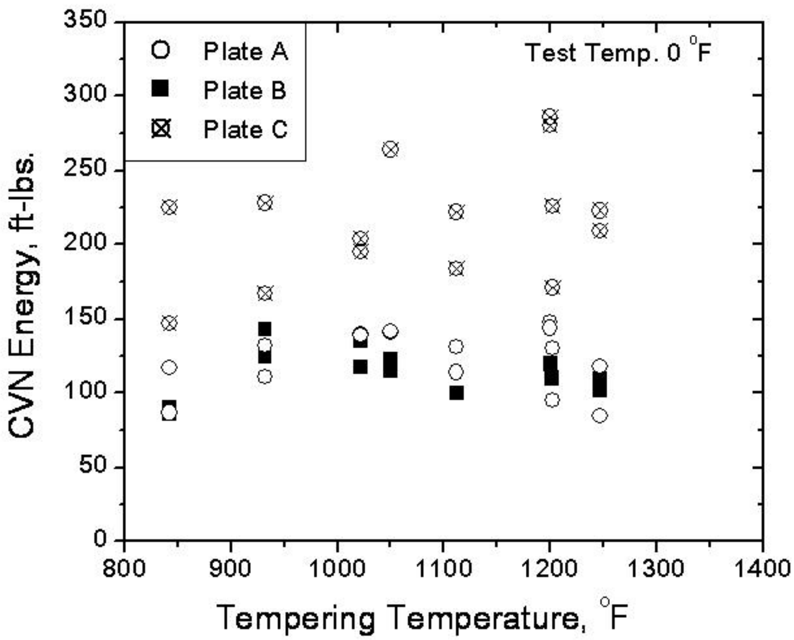
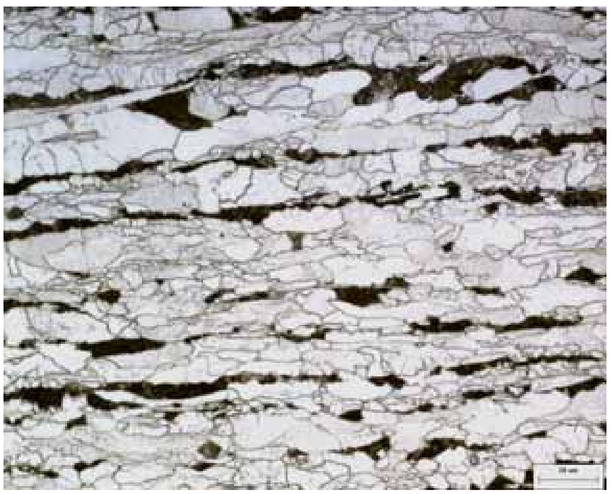


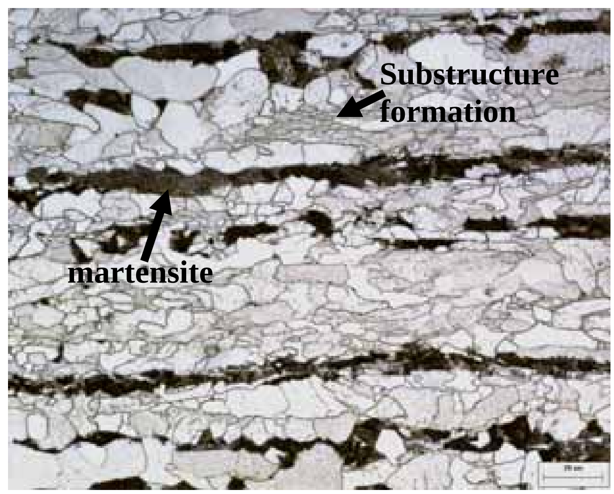
Figure 3 Charpy V-notch transverse impact energy values for Plate A, B and C as a function of tempering temperature for test temperature of 0 °F.

Microstructural Evolution in Hot Rolled and Tempered Plates

Figures 4(a)-(d) show microstructures of plate sample A in as-rolled condition and after tempering at 1200 ^{o}F for one hour. Figures 4(a)-(b) reveal the microstructure near surface and at the center of the plate sample through the thickness. Significant subgrain formation with elongated fine ferrite grains could be observed both near the surface and at the center plate thickness. Banded structures consisting of martensite and pearlite were found running parallel to the rolling direction. However, microstructure near the surface was mainly ferrite, bainite and martensite. Scanning electron micrographs in Figures 4(c)-(d) show the microstructural features before and after tempering at 1200 ^{o}F. The substructure is existent after tempering. Martensite is fully tempered and increased frequency of carbide precipitation was found at sub-grain boundaries. SEM/EDS microanalysis revealed these carbides to be Fe-carbides with occasional presence of Cr in solution.



(a)



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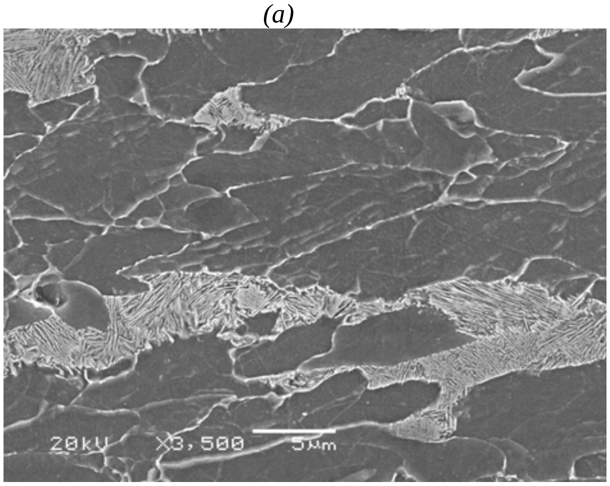

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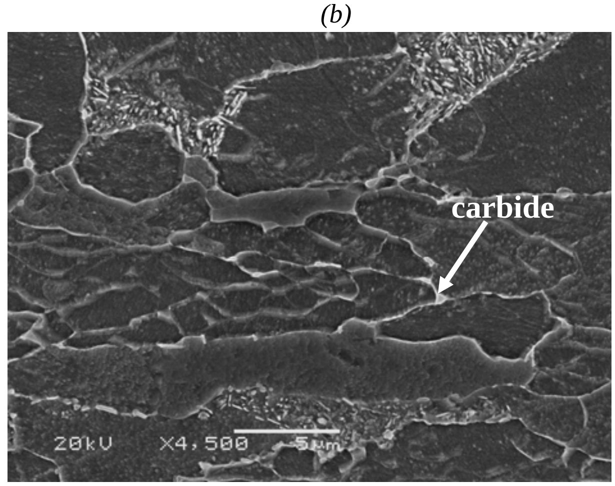


(d)

Figure 4 Microstructures of plate A viewed along the longitudinal(rolling) direction after thermomechanical processing at 1300 ^{o}F and after tempering. (a) microstructure near surface in the as-rolled sample and (b) microstructure near center of plate sample. Significant substructure formation could be observed at surface and at the center of plate. Banded structure consisted of martensite, pearlite and bainite in many instances, (c) subgrain structure before tempering and (d) subgrain coalescence after tempering. Carbide precipitation observed at sub grain boundaries and martensite constituents are fully tempered(d).

Figures 5(a)-(d) present the microstructures of plate sample B before and after tempering. Figures 5(a)-(b) reveals the microstructure near surface and center of the plate sample respectively. Second phase constituents are mainly carbide, bainite and martensite near the surface. Microstructure at the center of the plate revealed parallel bands of martensite with bainite. Ferrite is mainly polygonal with an average grain size of 8 µm. Figures 5(c)-(d) represent microstructure at the center of the plate sample before and after tempering. Martensite constituents are fully tempered and grain boundary carbide precipitations can be observed after tempering.

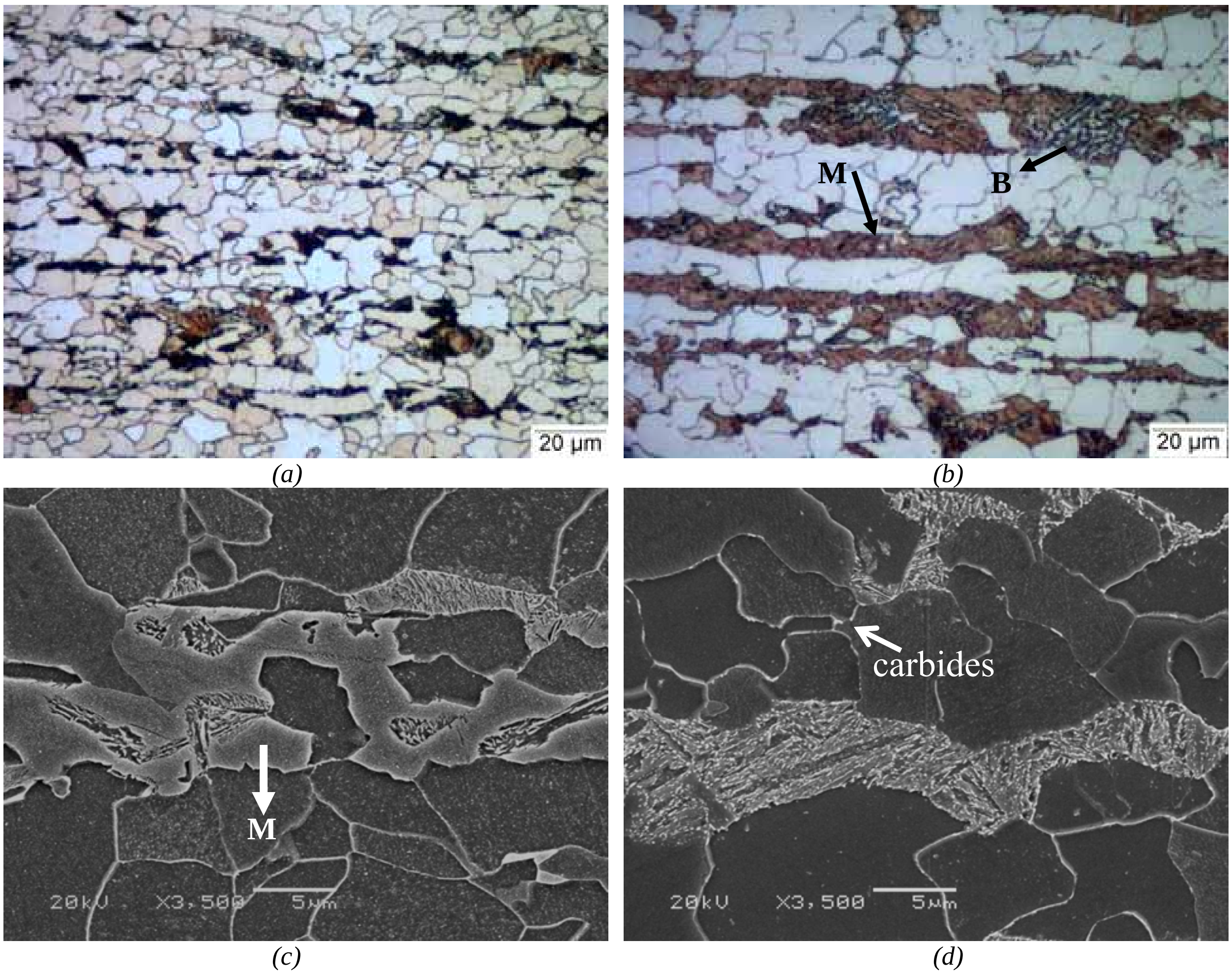


Figure 5 Microstructures of plate B finish rolled to 1600 ^oF viewed along the rolling direction. (a) polygonal ferrite grain structure near surface and (b) martensite bands at the center of plate in through thickness section. (c) SEM micrograph showing martensite and bainite constituents in as-rolled sample and (d) tempered martensite and grain boundary precipitation of carbides after tempering at 1200 ^oF. Etchant: 4% Nital and 10% aqueous sodium metabisulfite solution.

Microstructure development in plate C after normalizing and tempering are shown in Figures 6(a)-(d). Polygonal ferrite structure was found in the plate C sample after normalizing. The average ferrite grain size was 14 µm and was the largest among all the three plates investigated. The second phase constituents are mainly martensite and bainite. Banded martensite/bainite structure could be observed at the center region of the plate in through thickness section. However, the martensite bands are not as thick as seen in plate sample B. Figures 6(c)-(d) represent the microstructure before and after tempering.

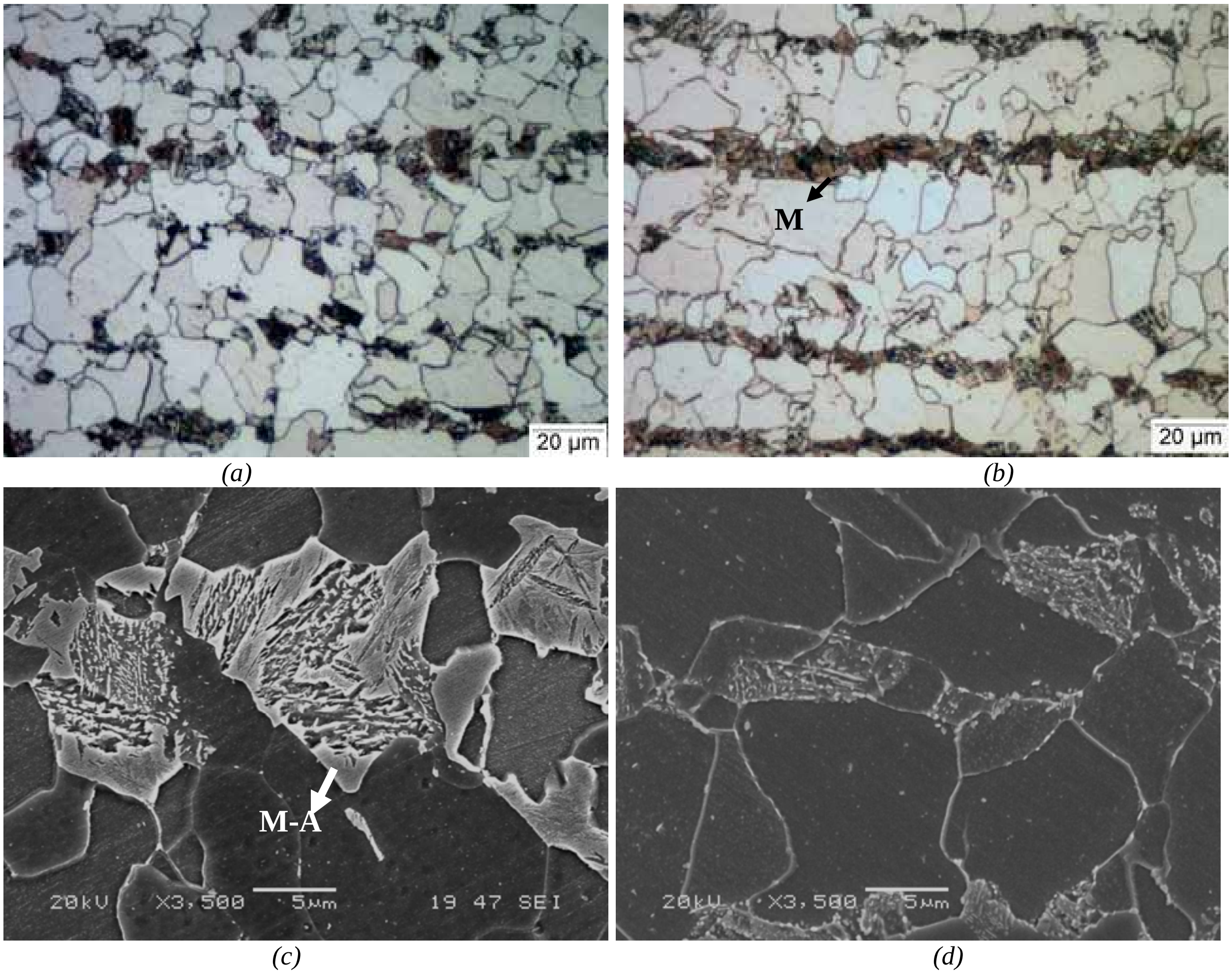


Figure 6 Microstructure of plate C normalized at 1650 ^oF followed by tempering at 1200 ^oF viewed along the rolling direction. (a) microstructure near surface showing polygonal ferrite grains and martensite and bainite as second phase constituents, (b) microstructure near center of the plate in through thickness section showing bands of martensite and bainite.(c)microstructure before tempering and (d) microstructure after tempering at 1200 ^oF. Carbide precipitation seen at the adjacent areas of tempered martensite (d). Etchant : 4% Nital and 10% aqueous metabisulfite solution. M-A: Martensite-Austenite, M: Martensite

Discussion

Presence of martensite and bainite constituents in all the plates after hot rolling is due to alloying that brought in a significant decrease in the transformation temperature. Banding is caused by the microsegregation of alloying elements and is usually expected of such a composition or alloy steel. Highest fraction of martensite was observed in plate B as the finish rolling was completed just above Ar_3 temperature. Martensite and bainite constituents influenced a continuous yielding in the flow curve of all the plate samples when tested in as-rolled condition.

An increase in the yield strength values after tempering could therefore be possibly due to a combination of the following contributory factors:

1. precipitation of Fe-carbides due to aging and
2. revelation of actual yielding strength of the steel which was masked by the pre-yielding microstrain due to mobile dislocation generation by martensite constituents [6].

The steel composition mentioned in Table 3 neither fosters the possibility of a secondary hardening component to the matrix nor does it promote age-hardening by Cu-precipitation type as in the case of Cu-bearing high strength steels [7]. A significant strengthening through carbide precipitation can be expected due to aging during tempering [6] as all the microstructures indicated a faster transformation kinetics during cooling and contained significant fraction of martensite. Precipitation is expected at the interface boundaries and in subgrain boundaries in case of plate A. Figure 7 reveals a large increase in the yield strength could be achieved in plate B after tempering at as low as 390 ^{o}F (200 ^{o}C) for one hour. The onset of return of yield point and flow behavior of the sample tempered at 390 ^{o}F (200 ^{o}C) indicated clearly the following:

- an early stage of precipitation (Fe-carbides) hardening,
- martensite constituents still able to significantly strain harden the ferrite matrix and
- no loss in tensile strength due to the low temperature aging.

These results suggested that the possible strengthening mechanism at higher temperatures of tempering is mainly due to Fe-carbide precipitation at interface boundaries and subgrain boundaries.

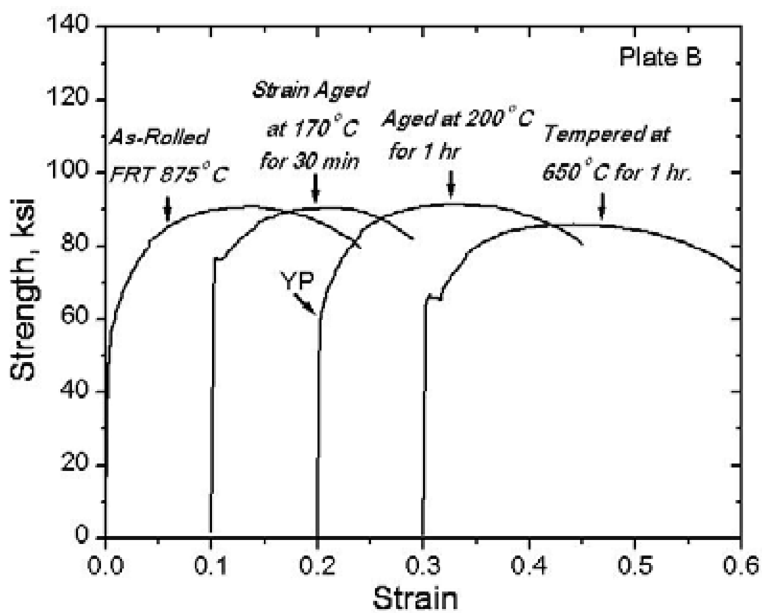


Figure 7 Increase in the yield strength due to low temperature and high temperature aging in plate B. A continuous flow behavior is indicated for as-rolled sample. The strain axis is offset by 0.10 for each sample for clarity.

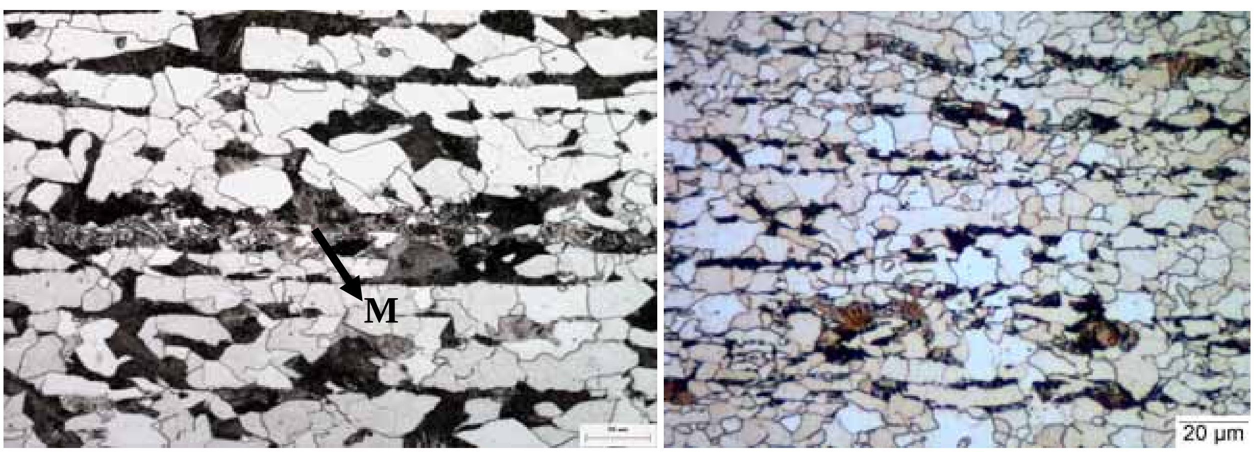
Return of yield point for the sample tempered at 1200 ^{o}F (650 ^{o}C) indicated tempering of the martensite constituents. The gradual loss of tensile strength with increasing tempering temperature (Table 6) is due to the loss of martensite strength.

The loss of as-rolled yield strength of plate A after tempering is due to annihilation of dislocation substructure with a concurrent increase in strength due to precipitation during aging.

Bake hardening tests for the plate sample B were performed to examine the aging sensitivity of the present steel in strained conditions. Tensile samples from plate B were prestrained 2% and then baked at 338 ^{o}F (170 ^{o}C) for 25 min. Figure 7 presents the flow curves for the strain aged sample B together with the flow curve of the sample tempered at 1200 ^{o}F for 1 hr. It is seen from Figure 7 that large increase in yield strength could be achieved due to strain aging. This finding is noteworthy in the sense that components formed at room temperature out of these plates will record a significant increase in the yield strength after forming and stress relieving operations. The strain aging results further confirms the potential strength increment that can be realized in the present steel through combinations of TMCP and post rolling heat treatments.

From the summary of mechanical properties achieved for different hot rolling and tempering combinations it is seen that the best combination of strength and toughness properties are achieved for plate A which was finish rolled to 1300 ^{o}F and tempered at 1200 ^{o}F. It was also observed that tempering within the temperature range of 1000 ^{o}F to 1200 ^{o}F did not significantly vary the strength or toughness properties. Plate B also demonstrated attractive strength and toughness combine albeit little lower than plate A. Plate C though manifested highest notch toughness properties but revealed a lower tensile properties compared to either plate A or B. Low temperature finish rolling is a concern for plate mills specially for production of wider plates which is usually a requisite for tank car manufacturing. Plate C requires higher production costs as both normalizing and tempering operations are involved. Processing route for plate B emerges as the most attractive option for production of plate with a minimum yield strength of 70 ksi and Charpy V-notch impact energy of over 100 ft-lb at -50 ^{o}F. The mechanical and impact toughness values of the present steels are compared with a TC 128B plate steel produced at BH. As indicated in Table 6, higher strength properties are obtained in plate B than the TC 128B steel. A microstructural comparison of normalized TC 128B steel and tempered plate B steel is given in Figure 8. TC 128B microstructure reveals comparatively larger ferrite grain size than that in the plate B steel. Pearlite fraction is more and significant banding is observed.

Figure 9 presents the CVN impact energy values for different test temperatures for tempered plates A, B and C and normalized and stress-relieved TC 128B steel made at Burns Harbor (composition given in Table 3). Low temperature toughness results for the present steel is better than TC 128 B steel. The present steel can therefore be presented as a suitable candidate for tank car manufacturing for toughness-critical applications.



(a) *(b)*

Figure 8 Microstructural comparison between (a) a normalized TC 128B steel and (b) tempered plate B steel at similar magnification. Martensite constituents are seen in TC 128B microstructure. Pearlite colonies are larger in size and fraction is higher than the plate B.

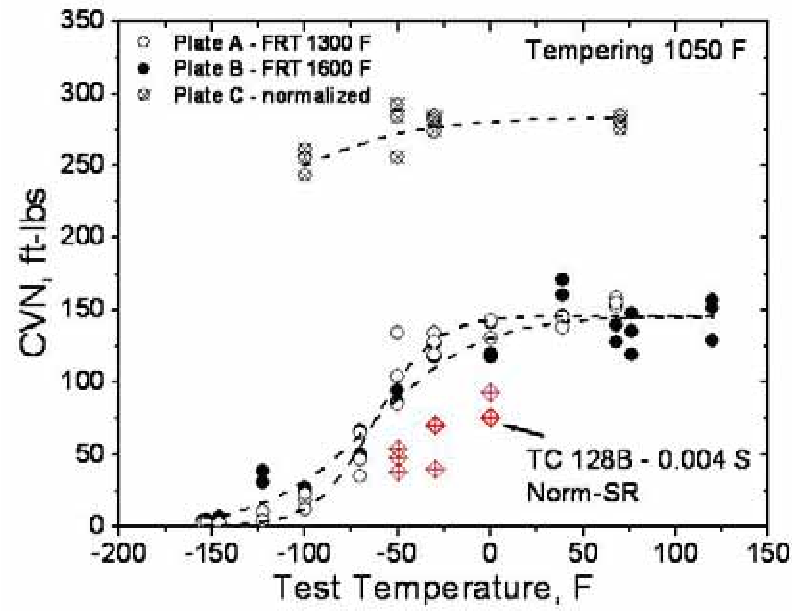


Figure 9 *CVN impact energy values for plates A, B and C for different test temperatures. Transverse CVN impact energy values for a low-Sulfur TC 128B steel(Table 3) have also been presented for comparison.*

Conclusions

Thermo-mechanical processing of HPS 70W steels with tempering within 1000-1200 ^{o}F provided excellent combination of strength and CVN impact toughness properties. A continuous yielding behavior was observed for all the hot-rolled plates. Tempering caused a significant increase in impact toughness and yield strength for the as-rolled steel. The experimental results demonstrated that both the minimum yield strength and CVN impact energy may be greatly exceeded by the present steel composition and processing compared to that usually obtained in TC128B steel. Superior CVN impact toughness with a minimum yield strength of 60 ksi can be achieved. The steel processing demonstrated a significant strengthening potential through strain aging. Components formed at ambient temperature will have a significant increase in yield strength during stress relieving operations. This finding assumes specific significance for tank car manufacturing.

The properties obtained in the HPS 70W steel through different processing combination have been compared with that of TC 128B steel and have been found to be superior. The new processing opens up opportunities for this steel to be considered for future critical tank car manufacturing.

Recommendations for Future Work

In order to extend the applicability of this steel for new frontiers, following future work is suggested/under consideration:

- study the effect of hot forming and cold forming of tank car heads on the toughness and mechanical property development,
- weldability studies to determine the effect of post weld treatments on the HAZ and weld metal toughness.
- assess performance behavior of these plates during actual tank car head forming

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References

[1] R. D. Sims and J. P. Hughes - Steels for Tank Cars, *43rd MWSP Conf. Proc.*, ISS, Vol. XXXIX, 2001, p 825-836.

[2] E. M. Focht and T. W. Montemarano – Development of High Performance Steels for Bridge Construction, *CARDIVNSWC-TR-61-95/16*, August 1995.

[3] High Performance Steels for Bridges: HPS 70W – Product Brochure, Arcelor Mittal website: http://www.mittalsteel.com/documents/en/Inlandflats/ProductBrochure/high%20performance%20bridge%2070%20ksi.pdf.

[4] A. D. Wilson, J. H. Gross, R. D. Stout, R. L. Asfahani and S. J. Manganello – Development of an Improved HPS-100w Steel for Bridge Applications, *Proc. from Materials Solutions Conference*, ASM International, 7-9 October, 2002, Columbus, OH. 2002, p 32-42.

[5] C. Ouchi, T. Sampei and I. Kozasu – The Effect of Hot Rolling Condition and Chemical Composition on the Onset Temperature of γ–α Transformation after Hot Rolling, *Trans. ISIJ*, 1982, Vol. 22, pp. 214-222.

[6] T. Waterschoot, A. K. De, S. Vandeputte and B. C. De Cooman – Static Strain Aging Phenomena in Cold-Rolled Dual-Phase Steel, *Metall. Mater. Trans. A*, Vol. 34A, March 2003, p. 781-791.

[7] A. Ghosh, B Mishra, S. Das and S. Chatterjee- Microstructure, Properties and Age-Hardening Behavior of a Thermomechanically Processed Ultralow-Carbon Cu-Bearing High Strength Steel, *Metall. Mater. Trans. A*, Vol. 36A, March 2005, p 703-713.