

## AUSTEMPERING AT CONVENTIONAL SINTERING TEMPERATURES

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### ABSTRACT

Last year the advantage of austempering after high temperature sintering was described. It was shown that, with austempering, a true hardness equivalent to wrought steels is possible.

However there is added cost involved with high temperature sintering. Therefore the natural question that evolves is what if you sinter at a conventional sintering temperature of 1120°C and austemper. This paper describes the tensile properties that result from sintering at 1120°C and austempering according to the cycle described in the previous paper. The properties that are obtained are compared with properties that resulted from the high temperature sinter and austemper cycle. Finally, differences in the respective properties are described. Ways to improve the properties are explored.

### INTRODUCTION

Last year we demonstrated the use of pressure quenching in a partial pressure/vacuum furnace to achieve hardenability (1). However, to many, high temperature sintering is too expensive for the applications that they have in mind. In other words, the concept of a belt furnace that simply acts continuously at a temperature of 2050°F (1120°C) to produce parts in what is defined as sintering for 30 minutes at temperature is the desired norm. Pressure quenching can still work under these conditions. The question becomes, how much will we lose in the quality of the parts if we conform to the requirements of conventional belt sintering. It is the purpose of this paper to demonstrate that austempering after sintering works, even if we confine our thinking to a 2050°F (1120°C) sinter for 30 minutes to simulate conditions that are experienced in a belt furnace. The question then is what is being sacrificed by submission to these conditions.

## PROCESSING CONDITIONS

The furnace used for the present work was the same furnace that was used for the high temperature sintering work. This furnace is equipped so that small quantities of useful gases may be introduced for specific reasons at various stages of processing. For instance, for the initial stages of lubricant removal, a wet argon atmosphere is employed to provide a source of oxygen to react with the hydrocarbon, and also to prevent the emerging vapor from the lubricant reaction with the oxygen from depositing on the furnace components. To accomplish removal of lubricant, the temperature was arrested at 1000°F (540°C) for a half hour to ensure removal of lubricant from the specimens.

Once the lubricant had been reacted and exhausted from the furnace, the wet argon was voided from the furnace and replaced with hydrogen at partial pressure. The hydrogen served as a reductant to reduce the oxygen film on the water atomized iron alloy powders. The temperature was elevated to 1450°F (790°C), and maintained for 15 minutes to permit the graphite to diffuse into the iron particles.

The temperature was then increased to 2050°F (1120°C) to sinter the specimens. This temperature is considered to be the one most commonly employed by parts producers. Since most parts producers maintain the belt speed to permit the parts to sinter for 30 minutes at temperature, we limited our sintering time to 30 minutes. Normally we would not do so because, not only are we limiting ourselves by the selection of the lower sintering temperature, but we are also limiting the diffusion time required to provide bonding and necking of the compacted particles. However, except for sintering atmosphere, which was a partial pressure of hydrogen, we wished to simulate commercial sintering practice as closely as possible. After the sintering cycle was complete, the power was diminished, and the furnace was permitted to cool to 1500°F (815°C). The hydrogen was removed from the furnace and the furnace was pressurized to 2 bar helium pressure. The temperature was further reduced to 800°F (425°C) and maintained for one hour in nitrogen to austemper the specimens. The specimens were then cooled to ambient. A typical cycle is included as Figure 1.

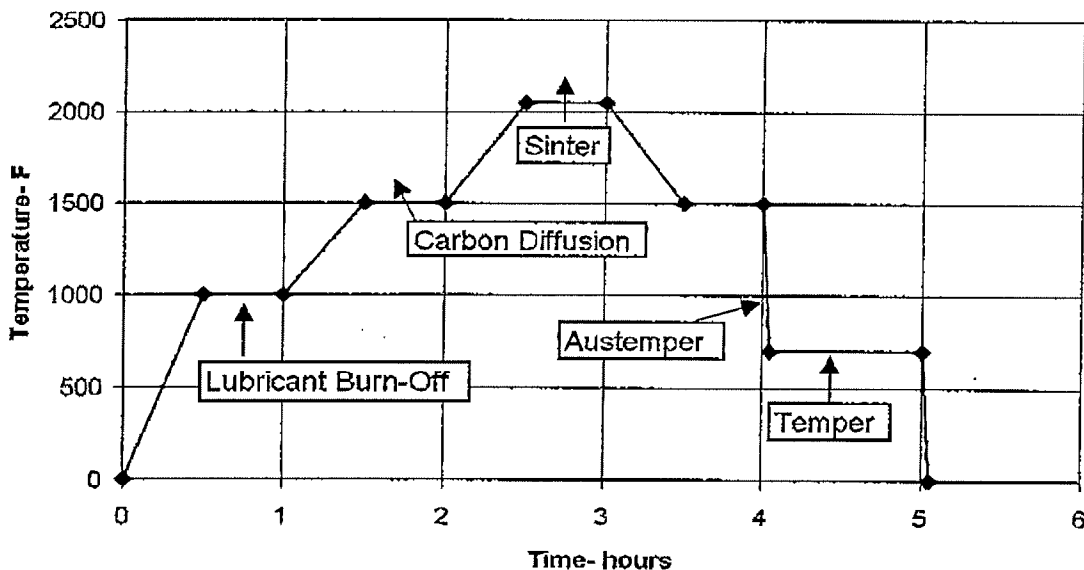
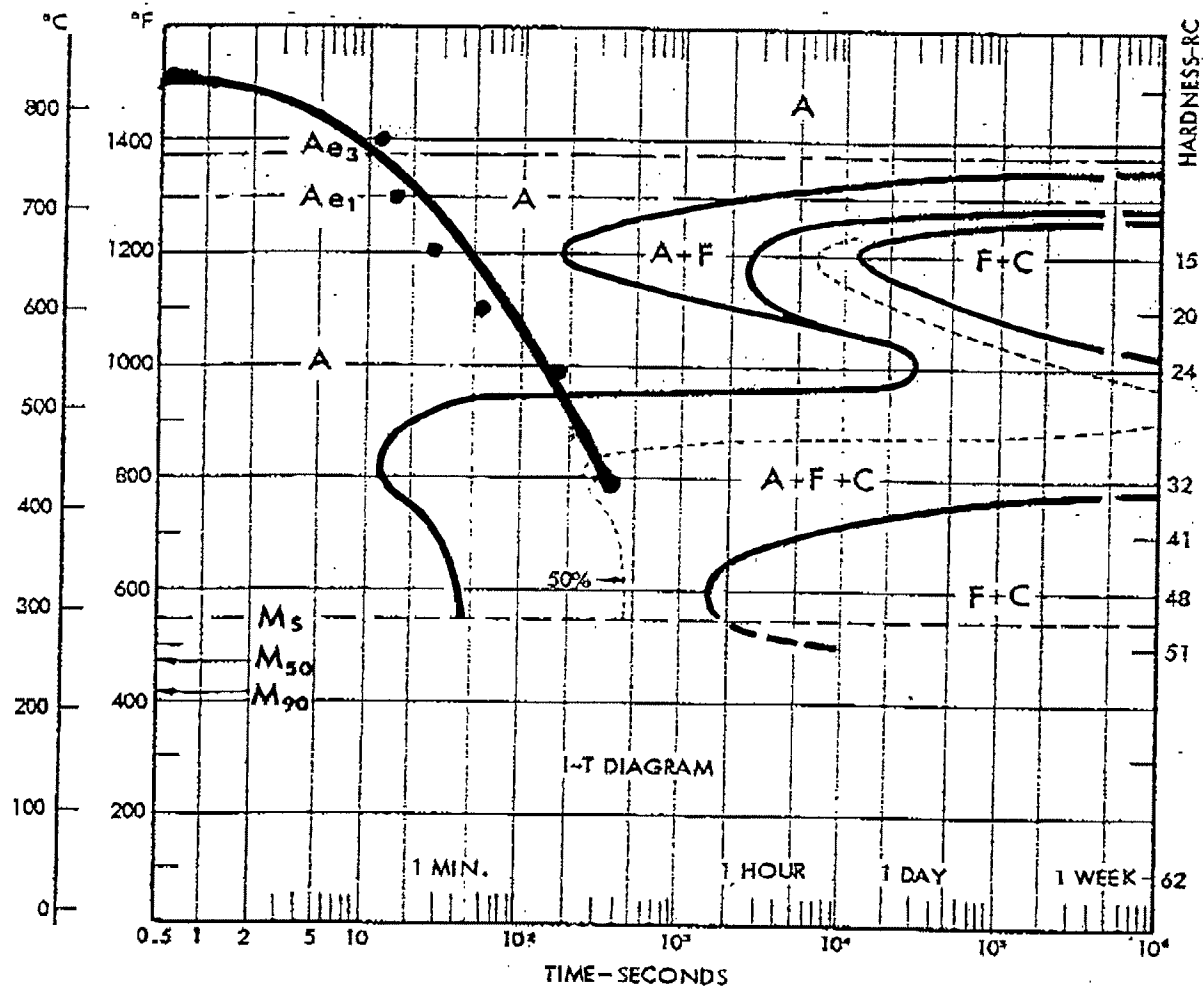


Figure 1. Thermal cycle.

The key to hardenability is the quench rate. The quench rate varies according to the alloying elements that are included within the steel. A typical cooling curve for a low-alloy steel is included as Figure 2.



**Cooling Curve of Austempered Impact Specimens**

Figure 2. Cooling curve of austempered specimens.

Superimposed on the cooling curve is the quench rate that was timed for the 2 bar helium quench. Note that the quench rate was designed to permit the temperature to be arrested where bainite is expected as the transformation product. Owing to the inertia of the quench, the cooling rate was actually slightly faster. However, we adjusted the temperature so that a temper of 800°F (425°C) was provided for one hour.

PROPERTIES OF CONVENTIONALLY SINTERED SPECIMENS

For this study tensile and transverse rupture specimens were included. A total of five different alloys were included within the study. Three alloys are commercially available, and two are experimental. Therefore to avoid commercialism, the alloys are simply identified as Alloy 1 through Alloy 5. The tensile properties for each of the alloys were determined from sintered dogbone specimens. After fracture, sections of these specimens were used to determine the carbonitrogen, oxygen, and sulfur levels. A half section of the tensile specimen was used to determine the sintered density, using Archimedes' Principle. An additional cross section of the broken tensile specimen was used to define the microstructure and to determine the true hardness by microhardness measurements. The sintered transverse rupture specimens were used to determine the expansion from die size.

Three carbon levels were included within the study. Sufficient graphite was admixed in each case to provide carbon concentrations after sintering of 0.4 m/o, 0.6 m/o or 0.8 m/o. An addition of 0.5 m/o ethylene bisstearamide was admixed to provide lubrication. All specimens were die compacted at 50 TSI (680 MPa). Only the high pressure was considered because we expected to lose some of the advantage in properties that would be gained from high temperature sintering. It was therefore considered impractical to further degrade the properties by adding additional porosity to the study.

The green properties are included in Table I. As is normal, density decreases as the graphite addition is increased. Expansion from die size is normal for these alloys.

Table I. Green Properties of Low Alloy Steel Compacts

Alloy	0.5 m/o Graphite	
	Green Density-g/cm <sup>3</sup>	Expansion From Die Size-%
1	6.97	0.43
2	6.95	0.42
3	6.82	0.38
4	6.88	0.38
5	6.88	0.39
	0.7 m/o Graphite	
1	6.90	0.39
2	6.92	0.44
3	6.79	0.44
4	6.86	0.36
5	6.84	0.40
	0.9 m/o Graphite	
1	6.89	0.41
2	6.88	0.38
3	6.77	0.44
4	6.83	0.38
5	6.83	0.36

As in our prior work, the aim is to achieve a true hardness of HRC 40. This has been shown to be what could be expected in relation to a fully dense structure (see Figure 3).

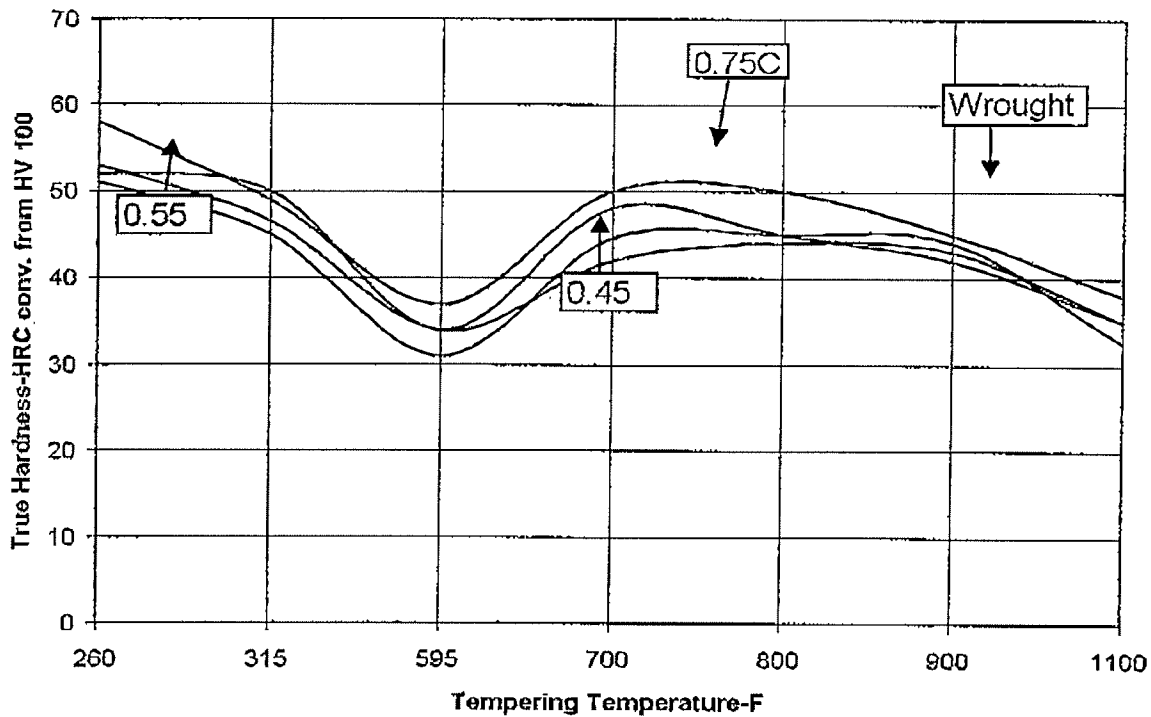


Figure 3. Tempering curve of FL-4605 type alloys.

The properties that were achieved using the simulated conventional sinter are presented in Table II. When conventionally sintered, the only alloys that achieved this goal were several of those that contained 0.8 m/o carbon. Therefore, if our goal is to match properties of high temperature sintered products or fully dense products, it is a given that a steel containing 0.8 m/o carbon is required. If we are comparing these properties with those of conventional wrought products, there is a large danger of retained austenite, which in turn leads to embrittlement.

Table II. Tensile Properties of Austempered Low Alloy Steels Sintered at 1120°C

Property	0.5 Graphite				
	Alloy 1	Alloy 2	Alloy 3	Alloy 4	Alloy 5
Ultimate Tensile Strength-psi	100,400	100,400	120,000	119,000	110,300
Apparent Hardness-HRC	10.5	14	17	12	9.5
True Hardness-HRC	32	28	23	29	35
Sintered Density-g/cm <sup>3</sup>	7.14	6.95	6.88	6.86	6.82
Dimensional Change-%	0.12	0.09	0.07	0.41	0.47
Carbon-m/o	0.43	0.39	0.40	0.43	0.41
Oxygen-m/o	0.030	0.063	0.036	0.024	0.025
Nitrogen-m/o	ND	ND	ND	ND	ND

Property	0.7 Graphite				
	Alloy 1	Alloy 2	Alloy 3	Alloy 4	Alloy 5
Ultimate Tensile Strength-psi	119,500	107,700	126,000	118,500	96,100
Apparent Hardness-HRC	13	16.5	20.5	13.5	12
True Hardness-HRC	37	32	29	32	36
Sintered Density-g/cm <sup>3</sup>	6.94	6.95	6.84	6.83	6.83
Dimensional Change-%	0.11	0.13	0.13	0.42	0.51
Carbon-m/o	0.61	0.60	0.61	0.61	0.58
Oxygen-m/o	0.038	0.048	0.035	0.023	0.021
Nitrogen-m/o	ND	ND	ND	ND	ND
Property	0.9 Graphite				
	Alloy 1	Alloy 2	Alloy 3	Alloy 4	Alloy 5
Ultimate Tensile Strength-psi	104,400	99,900	134,800	114,300	105,200
Apparent Hardness-HRC	14.5	17	20.5	16	14
True Hardness-HRC	40	34	36	41	41
Sintered Density-g/cm <sup>3</sup>	6.92	6.96	6.83	6.81	6.79
Dimensional Change-%	0.13	0.13	0.13	0.47	0.54
Carbon-m/o	0.82	0.72	0.78	0.79	0.77
Oxygen-m/o	0.64	0.040	0.038	0.027	0.029
Nitrogen-m/o	ND	ND	ND	ND	ND

However, let us see what this means in respect to the tensile strength and ductility of these steels. A comparison of the properties of two of the low alloy steels is included as Table III.

Table III. Comparison of Tensile Properties of Low Alloy Steels With Sintering Temperature

Alloy	Graphite	Ultimate Tensile Strength		Elongation-%		Apparent Hardness-HRC	
		1120°C	1260°C	1120°C	1260°C	1120°C	1260°C
A	0.5	119,400	130,900	-	3	12	14
A	0.7	118,500	150,200	-	3	13.5	19
A	0.9	114,300	154,200	-	4	16	23
B	0.5	110,300	118,800	-	2	9.5	7
B	0.7	96,100	133,700	-	2	12	14
B	0.9	105,200	136,100	-	2	14	19

First, let us dispose of the ductility consideration. As is readily seen, there is no ductility in the steels that received the conventional sinter. This is because the temperature was lower and the sintering time was restricted to 30 minutes at temperature. It is evident that diffusion has been severely restricted. Now let us consider the tensile strength. Again, as a result of the limiting conditions, the tensile strength is severely limited. Note that these alloys have a true hardness of 40 HRC or greater if they contain 0.8 m/o carbon. Therefore, the transformation product should be bainite and the properties should be equivalent to the high temperature sintered alloys. Alloy composition therefore cannot be a factor in the degradation of these properties.

Now let us consider the sintered density. All of the alloys that we studied had sintered densities of 6.8 to 6.9 g/cm<sup>3</sup>. If we consult MPIF Standard 35, we find that all of the alloys studied comply with the tensile strengths that are listed for the low alloy steels at this density level. In addition, one steel stands out as exhibiting increased tensile strength as the graphite addition is increased. Therefore, if Standard 35 is considered the norm, a 2 barr helium quenched low alloy steel satisfies the criteria for heat treated properties for this steel, and it is probable that nothing further can be done to improve these properties without additional changes in chemistry, increase in density or change in sintering conditions (2). If we consider sintering conditions, in a belt furnace the optimum conditions would be to move the belt as slowly as possible and consider boosting the sintering temperature to 2100°F (1150°C). The partial pressure sinter does not have these limitations and can be held at temperature as long as commercially desirable or at temperatures recognized for high temperature sintering for only slight additional cost. Other investigators are considering development of technology to increase the green density of compacted low alloy steel parts. Of course any improvement in this direction would result in improvement in properties of 2 barr helium quenched low alloy steels. Low alloy steel chemistry is a wide-open subject. Only now is chromium being considered as a hardening element in low alloy steels. Other possibilities exist, even for development of improved wrought low alloy steels, as well as powder metallurgy alloys.

In a paper that we previously presented, we simply quenched and stress relieved a series of low alloy steels (3). In this study we suggested that the embrittlement had restricted the tensile properties of the steels and that, if we had austempered the steels, the tensile properties would be improved. Last year we showed that this was true for high temperature sintered and 2 barr helium quenched austempered steels. We therefore suggest that if additional time at conventional sintering temperatures is permitted, additional diffusion will result, rounding and strengthening connecting surfaces of adjoining particles.

Now let us examine the dimensional change that has occurred as a result of processing. Note in Table II that shrinkage occurred in alloys 1 through 3 in relation to the expansion that had resulted from compaction. However, alloys 4 and 5 did not shrink appreciably after compaction. One might consider that lack of shrinkage may have resulted because surface oxides were not reduced. However, Table II indicates that the contamination of these two alloys was the lowest. Why then did not these alloys also shrink as much as the other three? This remains an ongoing subject for continuing research. Despite all of these factors, alloys 4 and 5 containing 0.4 or 0.8 m/o carbon exhibited tensile properties that exceeded the tensile properties of alloy 1 and alloy 2.

Last, let us consider carbon control. In each case our intent was to end up with a nominal carbon content within  $\pm 0.05$  m/o of predicted values. The aims were 0.4, 0.6 and 0.8 m/o carbon. Note that in most cases, the control of the carbon content was within the limits of tolerance. Again, we used a partial pressure of hydrogen as our furnace atmosphere, whereas a considerably greater flow of even nitrogen would be required for a belt furnace. Cost of hydrogen can therefore be equalized, resulting in much tighter control of the carbon content of the sintered and heat-treated alloys.

The microstructures and the pertinent properties of the alloys studied in relation to the carbon content are included as Figures 4,5 and 6. All appear to be bainitic, consistent with the thermal treatment that was provided. Some retained austenite may be present within the alloys that contain 0.8 m/o carbon, but not enough to cause the brittleness that was experienced.

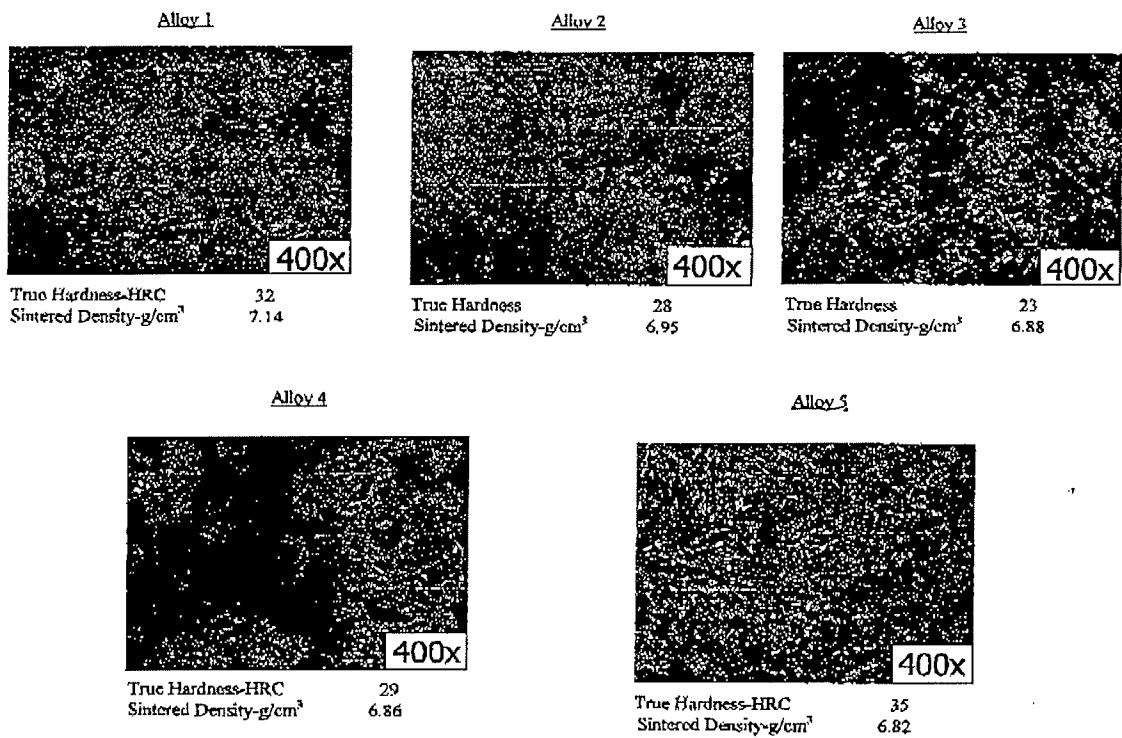


Figure 4. Microstructure of austempered sinter hardenable alloys containing 0.5 m/o graphite.

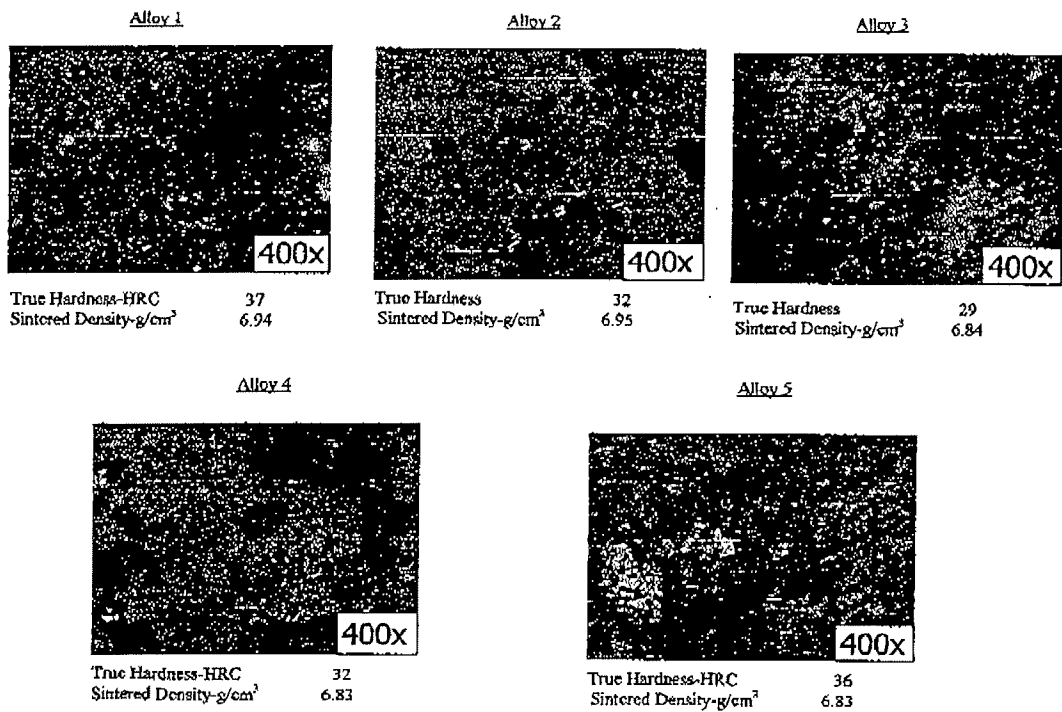


Figure 5. Microstructure of austempered sinter hardenable alloys containing 0.7 m/o graphite.



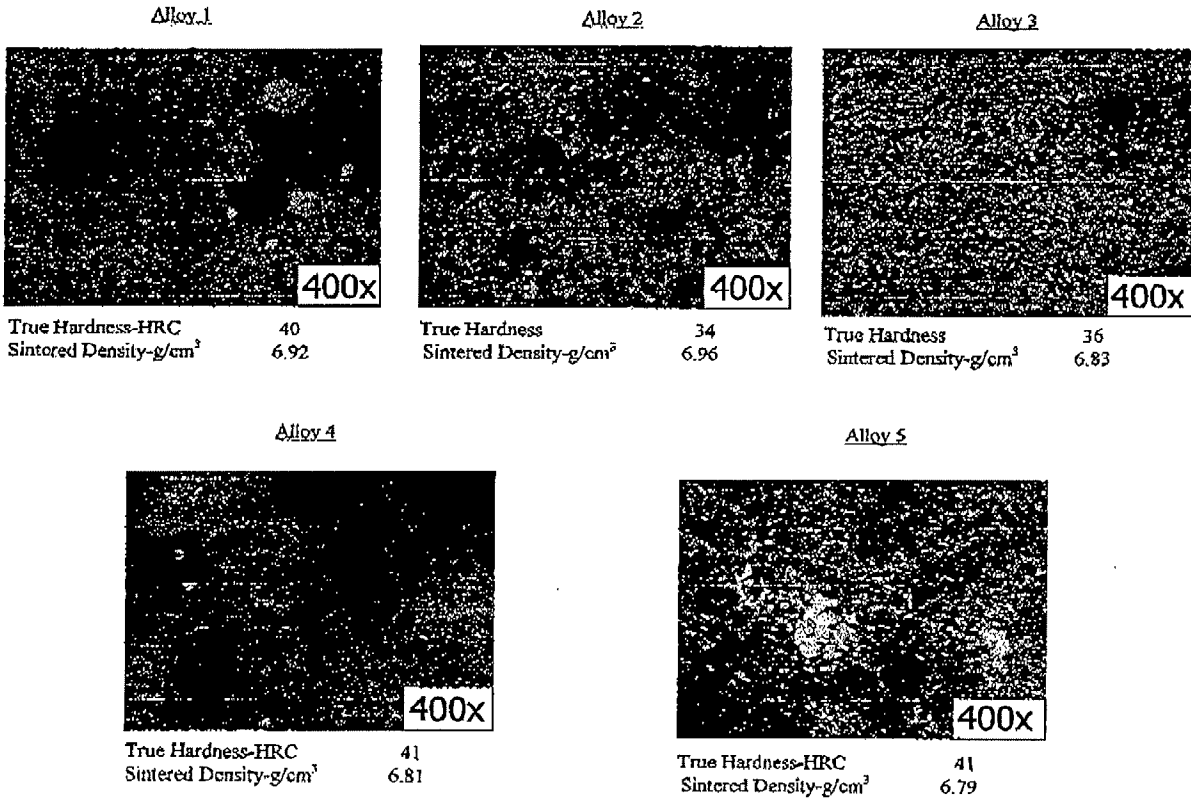


Figure 6. Microstructure of austempered sinter hardenable alloys containing 0.9 m/o graphite.

SUMMARY

The data show that austempering will produce heat treated low alloy steels with equivalent or improved tensile properties compared with heat treated properties presented in Standard 35 (2). Although the steels are similar in properties, the steels are brittle, owing to limited diffusion and necking that result from the sintering conditions that were imposed. It is suggested that, with little additional cost, the same thermal processing can be provided to provide a high temperature sinter to enhance diffusion and densification with an overall improvement in the mechanical properties of the alloys. Also there is much more additional research that can be accomplished to provide alloys with improved properties at conventional sintering temperatures. Carbon control is also within at least  $\pm 0.05$  m/o.

## REFERENCES

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