Fe-Mn-Al-C ALLOY STEELS – A NEW ARMOR CLASS

R.A. Howell, R.J. Gerth

Program Executive Office Ground Combat Vehicles
Tank Automotive Research Development and Engineering Center
Warren, MI
Tel.: (586) 282-6346
E-Mail: ryan.a.howell2.mil@mail.mil, richard.j.gerth.civ@mail.mil

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SUMMARY

Fe-Mn-Al-C alloy steels are being ballistically characterized by the Tank Automotive Research Development and Engineering Command (TARDEC) and the Army Research Lab (ARL) as a Class V Rolled Homogeneous Armor (RHA) Grade under MIL-A-12560. This paper summarizes the prior body of work regarding ballistic investigation leading up to the current Class V development. Physical and metallurgical characterization with limited ballistic tests has occurred by ARL[1,2]. Fe-Mn-Al-C alloys were investigated since they possess lower density (6.5 to 7.2 g/cm^3) versus standard steel chemistries (7.8 g/cm^3)[3], tensile strengths from 600 to 2,000 MPa, and elongation to failure as great as 70%. These alloys are near non-magnetic, and they can exhibit high-energy absorption within a chemistry range of Fe-10-30Mn-5-15Al. High strain rate testing has shown that as strength increased with strain rate, ductility exceeded 30% at 1000 s^{-1}, indicating that ballistic evaluation would be favorable, even in a cast form[3,4,5]. A ballistic test and report was conducted on a Fe-13Mn-10Al-1C wrought alloy in 1979[2], an extensive investigation into a cast Fe-30Mn-9Al-1Si-0.9C-0.5Mo alloy was conducted between 2006 and 2010[1], and a wrought form of the Fe-30Mn-9Al-1Si-0.9C-0.5Mo alloy was ballistically tested in 2008 which has not been previously reported.

The 1979 wrought investigation examined cold worked solution treated material against the Class I requirement for MIL-A-12560 Rolled Homogeneous Armor (RHA) acceptance curves. The cast alloy was a quenched and aged alloy tested with respect to Class II military performance specification MIL-PRF-32269 but was also measured against RHA standards as was the wrought form of the Fe-30Mn-9Al-1Si-0.9C-0.5Mo alloy. The goal, in all cases, was to develop a lighter steel alloy utilizing existing infrastructure and manufacturing means such that a dimensionally constant, but lower weight material, could be used anywhere RHA or P900 were employed without exorbitant costs to achieve the weight savings.

MELT PROCEDURES AND PROCESSING BEST PRACTICES

Fe-Mn-Al-C steel making does not require exotic procedures specialized equipment even though it is a very high alloy steel. Less than 80% of composition was iron in the 1979 investigation and less than 60% of the composition was iron in the later trials[1,2]. There are best practices that should be followed, though, in Fe-Mn-Al-C steel production to maximize cleanliness and material properties. Fe-Mn-Al-C alloys have been produced utilizing existing steel making equipment with slight modifications to standard melt practices. The material has been made in both open air induction furnaces and vacuum furnaces. For open air induction melting furnaces, MgO refractory linings are utilized as much as possible. Alumina and silica refractory exchange oxygen with the liquid steel due to the melt's high Al content, and therefore should not be used. This exchange degrades furnace linings and increases inclusion content in the steel. Lab scale ingots for rolling have been produced in vacuum induction furnaces. This production method does not require cover gas or a synthetic slag to protect the melt from atmospheric oxidation. For casting mold material, olivine sand based molds or preheated investment shell molds have been used to minimize reactivity. Steel heats that were melted utilizing open air furnaces also require a heavy flow of Ar cover gas to minimize atmosphere reactivity. The high Al content does benefit the liquid melt in that there is not a requirement to add a deoxidizing agent to the melt prior deslagging and tap. In either open air furnaces or in the vacuum furnaces, oxygen typically measured less 6 ppm.

Fe-Mn-Al-C alloy steel is sensitive to P content[6], therefore melt stock and steel making procedures guarded against any alloy P contamination. Best practices included the use of high purity electrolytic iron and electrolytic manganese. Removing P can be achieved by heavy Ca treating and argon stirring. Small additions (<1 wt.%) of misch metal also negate P. Other sources of P that have been considered and substituted were any refractory glaze, refractory binder, or mold binder agent that contained P. A later section describes in detail the mechanical property degradation for high P containing Fe-Mn-Al-C steel.

The high Al content reduces the liquidus and solidus temperatures such that the melt temperature is commensurate with cast iron. But additions of Si also causes reductions in the liquidus and solidus temperatures as is shown in Table 1 for a nominal Fe-30Mn-9Al-XSi-0.9C-0.5Mo alloy[7]. The addition of Si is a typical foundry practice to increase fluidity of a steel alloy and has the additional benefit of minimizing ferrite content.

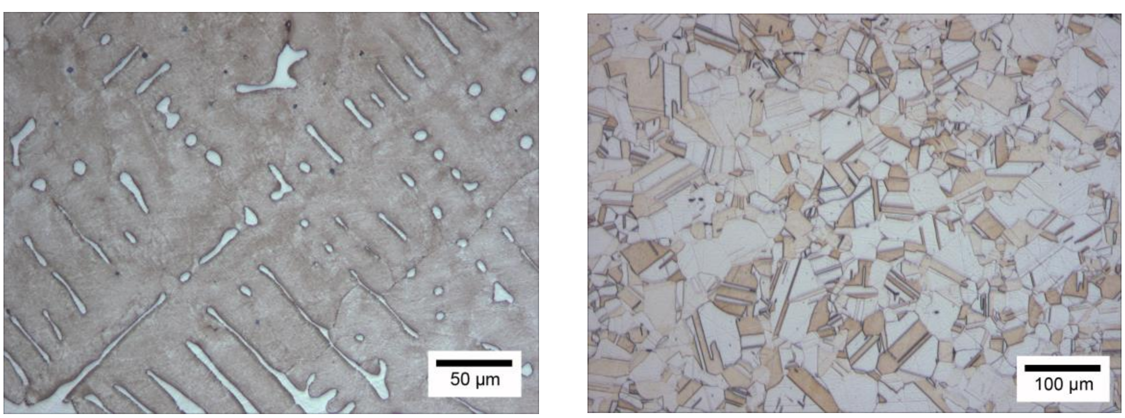
Table 1. Liquidus and Solidus Temperatures as Determined by Thermal Analysis and Differential Thermal Analysis

Si Content (weight %)	Thermocouple Data (°C)			Differential Thermal Analysis (°C)		
	Liquidus	Solidus	ΔT	Liquidus	Solidus	ΔT
0.3	1395	1333	58	1376	1309	67
0.82	1353	1309	50	1365	1312	53
1.36	1343	1247	96	1326	1268	58
2.24	1317	1275	42	1306	1242	64

PROCESS, MICROSTRUCTURE, AND PROPERTY RELATIONSHIPS

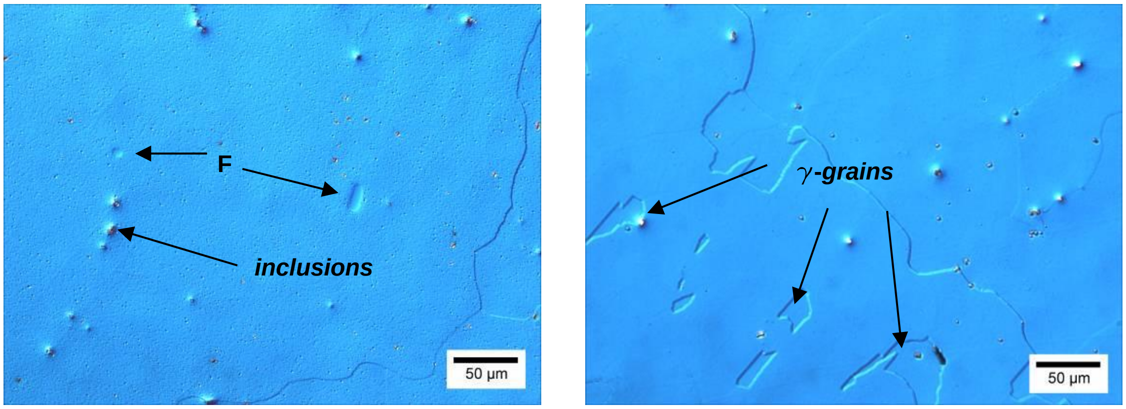
The Fe-Mn-Al-C steel chemistries described in this paper are best utilized in a solution treated or age hardened condition. The as-cast microstructure of Fe-Mn-Al-C alloys are primarily an austenitic matrix. Interdendritic regions are formed of ordered ferrite with a Kappa carbide structure precipitated throughout the matrix upon slow cooling or age hardening as can be seen in Figure 1a. The wrought microstructure contains equiaxed grains of 50 microns or less in size as shown in Figure 1b. The refined austenite grains contain twin boundaries produced during recovery from the rolling process. The matrix composition varies with the heat treatment, and the alloys described in this paper reflect three different heat treatment conditions to attain properties: 1) solution treat, water quench, and cold rolled for the 1979 wrought trial 2) solution treat, water quench, and age hardened for the cast material and 3) solution treat, hot rolled, water quench, and age hardened for the 2008 wrought trial.

The highest attainable toughness and ductility properties are dependent upon maximizing the volume fraction of austenite. The largest austenite volume fraction is achieved by rapid quenching from solution treatment temperatures above 900 °C as shown in Figure 2 for a cast microstructure. Rapid quenching is required in order to avoid the nose of the Kappa carbide transformation curve as shown in Figure 3 for a 1.25% Si containing alloy[8]. Rapid quenching minimizes grain boundary precipitate formation, thus increasing toughness by reducing brittle grain boundary phases. The regions in Figure 3 are divided to include a mapping of the grain structure morphology. Chemical modulation occurs in area 1 followed by homogeneous κ-carbide precipitation in Area 2. Area 3 corresponds to both grain boundary formation of κ-carbide and a DO_3 or B2 precipitate. Area 4 represents a discontinuous reaction product of grain boundary κ-carbide, DO_3 and/or B2 precipitates. All equilibrium phases are present in Area 5; and for alloys without silicon, precipitation of β-Mn is included as Area 6. As expected, higher temperatures result in faster kinetics. Aging at 500°C and below greatly extends the time until grain boundary precipitation occurs.



(a) (b)

Figure 1. An as-cast nominal Fe-30Mn-9Al-1C-0.5Mo alloy is shown in (a) and a wrought microstructure of the same chemistry is shown in (b).



(a) (b)

Figure 2. Microstructure of Fe-Mn-Al-C steels with 0.82 % Si solution treated at (a) 900°C showing the formation of small ferrite (F) islands and (b) at 1000°C that produced a fully austenitic structure upon quenching. The images were taken using a differential interference contrast technique, thus causing the steel to appear blue in the image.

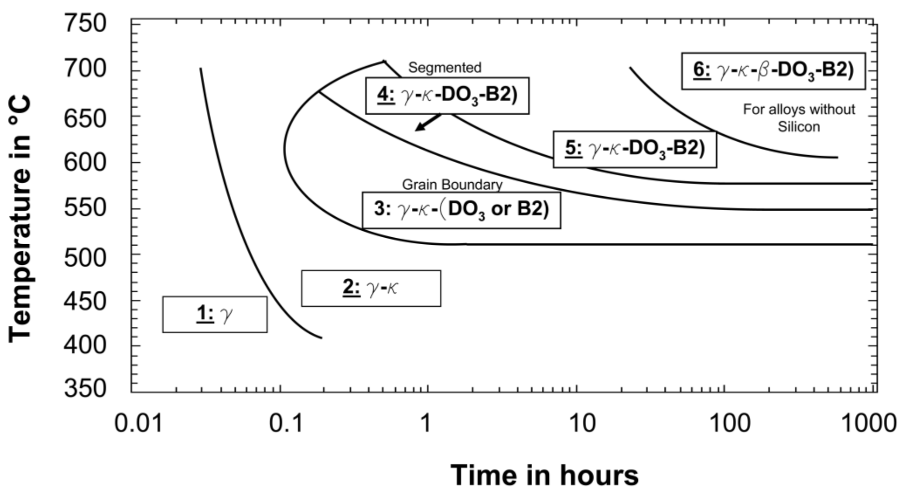


Figure 3. Time temperature transformation for a 1.25 wt.% Si modified Fe-Mn-Al-C alloy is subdivided as follows: 1) chemical segregation and zone formation, 2) matrix κ-carbide nucleation and growth, 3) heterogeneous matrix κ-carbide formation and grain boundary κ-carbide and DO_3 or B2 precipitation, 4) continued growth and decomposition forming discontinuous segmented phases of κ-carbide, DO_3, and B2, 5) final equilibrium phases, and 6) β-manganese for non-silicon containing alloys.[8]

1979 WROUGHT MATERIAL INVESTIGATION

The wrought Fe-13Mn-10Al-1C material[2] investigated was cold rolled to 0.272 inches thick with an areal density equal to 9.32 lbs/ft^2. Hardness of the cold rolled material measured 357 Brinell Hardness (BHN). Typical RHA values at an equal areal density are in the range of 341-388 BHN. The wrought Fe-13Mn-10Al-1C alloy was estimated to have a yield strength (YS) of

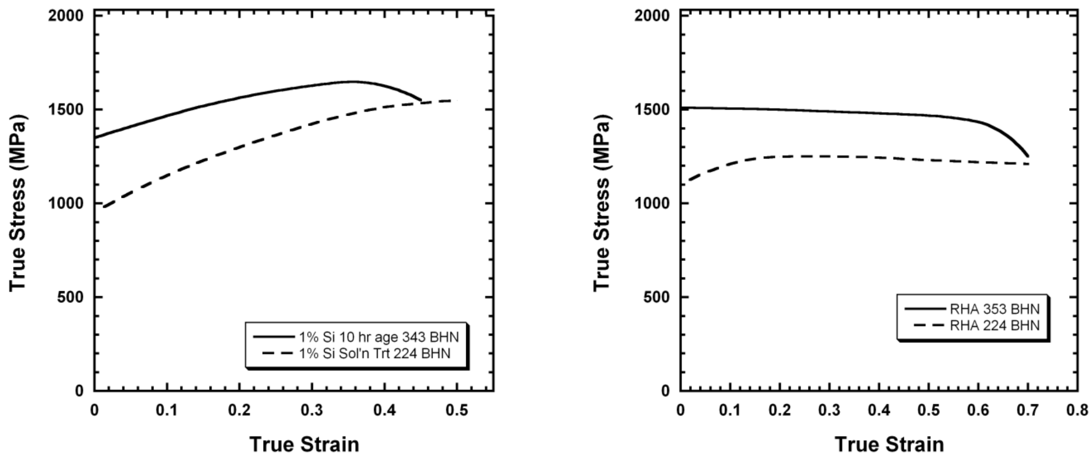
109 ksi with an ultimate tensile strength (UTS) of 151 ksi at 30% elongation based on the measured hardness. RHA typical tensile values for the hardness range previously listed are 165-180 ksi YS, 180-220 ksi UTS, and 13-16% elongation. V50 was determined and compared to for the 7.62mm M80 ball munition at 0 degrees obliquity. The wrought Fe-13Mn-10Al-1C alloy achieved a V50 value of 2592 ft/s. The RHA value at the same areal density is 2395 ft/s. The wrought Fe-Mn-Al-C alloy mass efficiency was 1.08.

P900 CAST INVESTIGATION

The standard Class II cast perforated homogenous armor, or P900, was an age hardenable cast alloy that is tunable within the range specified for both the P900 PRF and RHA specifications[1]. The heat treat condition of this alloy were tested in a T6 solution treated and aged condition. Measured density, hardness, and strain rate strength data for cast Fe-30Mn-9Al-1Si-0.9C-0.5Mo and RHA are shown in Table 2. The density was measured to be 13% less than RHA. The specific strengths exceed RHA by greater than 10%, and the strain sensitive work hardening behavior of the cast Fe-30Mn-9Al-1Si-0.9C-0.5Mo alloy versus measured values for RHA are shown in Figure 4. This cast alloy attained mass efficiencies of 1.08 and 1.01 when tuned and compared to the 0.30 cal APM2 acceptance limit for MIL-PRF-32269 and class 1 RHA respectively. The alloy has to be tuned to the upper end of the hardness range to achieve these efficiencies in cast form. Table 3 shows the pertinent test data from the 0.30 cal APM2 test compared to RHA for the cast Fe-Mn-Al alloy in the T6 condition. Phosphorus (P) content is deleterious to properties in these high Mn steels, thus higher V50 efficiencies are possible for lower P content.

Table 2. Comparison of specific compressive strengths of two heat treated alloy conditions between the cast 1% silicon containing Fe-Mn-Al-C alloy and RHA tested at 3000 s^{-1} strain rate.

Alloy	Density	Specific Strength 1% Si – BHN 224 RHA – BHN 224	Specific Strength 1% Si – BHN 343 RHA – BHN 352
1% Si	6.7 g/cm^3 (0.24 lbs/in^3)	231 MPa/ρ (933 ksi/ρ)	246 MPa/ρ (995 ksi/ρ)
RHA	7.8 g/cm^3 (0.28 lbs/in^3)	173 MPa/ρ (646 ksi/ρ)	192 MPa/ρ (775 ksi/ρ)



(a) (b)

Figure 4. The Fe-Mn-Al-C alloy (a) shows work hardening in the solution treated and aged condition at a strain rate of 3000 s^{-1}. Rolled homogeneous armor (b) does not strain harden to failure. Ultimate strength occurs between 0.1 and 0.2 strain followed by decreasing stress to fracture beyond 0.6 strain.

Table 3. Cast Fe-Mn-Al T6 alloy compared to HDNS and V50 acceptance limits for 0.30 cal APM2 within the MIL-PRF-32269 Class 2 and MIL-A-12560 Class 1 specifications at an areal density that translates to 0.530 inches thick normal steel density from the acceptance charts.

Property	Cast T6 Alloy	MIL-PRF-32269	MIL-A-12560
HDNS	351 BHN	302-352 BHN	363-400 BHN
V50	2549 ft/s	2374 ft/s	2525 ft/s

WROUGHT Fe-30Mn-9Al-1Si-0.9C-0.5Mo

The wrought age hardened Fe-Mn-Al-C alloy was a vacuum induction melted alloy. The ingot was hot rolled to a thickness of 0.41 inches and water quenched. The plate was then aged at 530 °C for 10 hours to a hardness of 340 BHN. The chemistry was not measured after processing and the P content remains unknown. The density measured 13% less than RHA which translates to a mass equivalent thickness of 0.36 inches. Ballistic equivalency comparisons to Class I RHA is shown in Table 4. On a mass equivalency basis, the wrought material achieved a mass efficiency of 1.1. Volume equivalent efficiency is 1.02. By comparison, titanium based armor alloys have volume efficiencies less than 1 even though their mass efficiencies are well above 1. The significance of the volume efficiency for Fe-Mn-Al-C based steels are that for weight savings applications, a direct substitution is possible for RHA or other high strength low alloy (HSLA) steels while attaining a 13% weight savings. This direct substitution means that the physical geometry and component interfaces of RHA or other HSLA steel components remain the same when converted to Fe-Mn-Al-C alloy steel while achieving a 13% weight reduction. The cost savings/cost avoidance achieved from maintaining component and interface geometries is not insignificant.

Table 4. Wrought Fe-Mn-Al-C T6 alloy compared to HDNS and V50 acceptance limits for 0.30 caliber APM2 MIL-A-12560 Class 1 steel at equal mass and also at equal thickness for a constant volume comparison.

Property	Wrought T6 Alloy	MIL-A-12560 (0.36" thick – mass equivalent)	MIL-A-12560 (0.41" thick – volume equivalent)
HDNS	340 BHN	363-400 BHN	363-400 BHN
V50	2218 ft/s	2011 ft/s	2175 ft/s

CUTTING AND MACHINING CONSIDERATIONS

The Fe-Mn-Al-C alloy family is a derivative of the Hadfield class of high Mn steels. The high work hardening nature of these steels requires that care has to be taken when cutting, machining, and processing these materials. Blunt carbide or aluminum oxide cutting tools are recommended for all machining operations. To prevent tool wear and machine tool damage, slow speeds and feeds are recommended, but a thick depth of cut is possible. These alloys can be flame, abrasive and band saw cut. However, any saw blade must be capable of cutting austenitic stainless steels and band saw blades must be carbide tipped. Traditional tools steels, abrasive cut-off blades, and metal cutting band saw blades are incapable of processing these materials and must not be used.

WELDING

To date, no welding has been formally conducted on plate materials. Initial lab investigations on cast material suggest that standard austenitic stainless steel wire for gas metal arc welding (GMAW) is sufficient to bond plate material together. Two cast plates of 0.006% P containing alloy of a nominal Fe-30Mn-9Al-1Si-0.9C-0.5Mo composition at 0.6 inches thick in an solution treated and aged condition were prepared and welding with a 304 stainless steel based Metal Inert Gas (MIG) weld. Ar was used as the cover gas. The plates were chamfered at 45° for a butt joint configuration. The plates were not preheated. Three passes were required to fill each side of the butt weld. The result was a crack free weld joint. No mechanical testing was conducted as this was an initial inspection with respect to the viability of using Fe-30Mn-9Al-1Si-0.9C-0.5Mo alloy in a welded configuration.

CURRENT AND FUTURE EFFORTS

For the current phase of development, a vanadium (V) modified alloy is being investigated and quantified as a Class V MIL-A-12560 alloy. The V modified alloy is required for plate in order to boost strength of the parent material in thick section and in the heat affected zone (HAZ) of any welded section. The alloy target chemistry is Fe-28Mn-8Al-1Si-0.9C-0.55Mo-0.55V. The reduction of Mn (by 2 wt.%) and Al (by 1 wt.%) from the previously tested material is necessary to improve toughness and ductility while the V will increase strength, toughness, and provide necessary grain refinement and grain pinning during thermomechanical processing. The alloy will still be capable of a 10% weight reduction due to the 1% Si addition. The combined effects should easily couple for a mass efficiency greater than 1.1.

Future research should be directed at higher strengths and density reductions for wrought processed material. Target chemistries should incorporate higher Al, lower Mn, Ni additions, Nb substitutions for V, and possibly W alloy additions. An initial recommended target chemistry for research is Fe-20Mn-5Ni-10Al-1Si-0.8C-0.5Mo-0.5V. The Mn reductions are needed to increase ferrite phase structures, but the Ni is needed to toughen those brittle phases. Higher Al will further reduce the density, but a reduction in C is needed to prevent a high volume fraction of ordered brittle phase structure. Further additions of Al should be sought to understand the maximum level of Al that the matrix can tolerate and still support material processing and desired properties.

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