ABSTRACT

Chromium and nickel are generally used as alloying elements in iron base alloys for corrosion resistance and elevated temperature applications. A survey of recent literature indicates a renewed interest in the development of Fe-Mn-Al alloys as substitutes for Cr-Ni stainless steels with a view to conserve and avoid the use of more expensive elements like chromium and nickel. The aim in the alloy design is to obtain optimum compositions of the Fe-Mn-Al alloy system with regard to corrosion resistance and high temperature properties. The austenitic structure, stabilised by manganese, would permit plastic deformation with relative ease whereas the aluminium in solid solution would diffuse to the surface of the alloy to form an aluminium rich protective oxide scale during high temperature exposure. Austenitic wrought Fe-Mn-Al-Si alloys are reported to possess good mechanical properties upto 600^{o}C and good oxidation resistance upto 900^{o}C. However, most of the investigations on Fe-Mn-Al alloys have so far concentrated on assessing its characteristics for wrought applications. The present investigation was therefore conducted on Fe-(26-32)Mn - (2-9)Al - (1.5-2.3)Si - (0.25-0.94)C alloys to evaluate the following characteristics for cast applications :

(i) Casting and solidification,

(ii) Non-isothermal and isothermal phase transformations,

(iii) Oxidation resistance in the temperature range $700^{o} - 1000^{o}C$,

(iv) Machinability.

The object of the investigation was to identify alloy compositions which would offer an optimum combination of castability, phase stability, oxidation resistance and machinability.

Experimental Procedure

The alloys were melted under lime (65%) - fluospar (35%) flux cover in a 1.5 kg. magnesia crucible of an induction furnace and in a basic lined 25 kg. arc furnace. The induction furnace melts were cast in the form of 10 mm x 10 mm x 200 mm rods in both resin bonded and sodium silicate bonded silica sand moulds. About 15 kg. of each arc furnace melt was cast in the form of a 75 mm diameter x 250 mm long rod and a standard keel block sample in resin bonded moulds. The remaining portion of the melt, free from slag, was treated with 0.5% misch metal in a ladle and cast in the form of a keel block. The experimental techniques adopted to investigate the characteristics mentioned above included : (i) optical and scanning electron microscopy, (ii) hardness measurements, (iii) heat treatment, (iv) X-ray diffractometry for phase identification, (v) magnetic permeability measurements,

(vi) isothermal oxidation studies by continuous mass gain measurements, (vii) electron microprobe analysis and (viii) elevated temperature and conventional machining. The principal findings of the investigation are summarised below.

A) Casting and Solidification Characteristics

The Fe-Mn-Al alloy castings suffered mould-metal reaction to a varying degree. In general the severity of the reaction decreased (a) with increase in aluminium content in the melt, (b) on substitution of sodium silicate bonded sand by resin bonded sand and (c) on misch metal treatment of low aluminium (2% Al) alloys. In alloys containing 2-5%Al, fluid iron silicate and manganese silicate of varying compositions were formed by mould-metal reaction. The fluid slag which also entrapped sand grains later solidified on the casting surface. In the higher aluminium and misch metal treated alloys, mould-metal reaction was prevented by the formation of a protective film of alumina and cerium aluminate respectively on the casting surface. Cast Fe-Mn-Al alloys with more than 5% Al suffered hot tearing in thick sections ($\sim$75 mm). Solid state shrinkage was minimum in the 5% Al alloy.

The as cast microstructures of Fe-(26-32)Mn - (2-10)Al - (0.25-0.9)C - (1.5-2.3)Si alloys consisted of austenite, ferrite, a complex carbide [$Al(Fe,Mn)_3C_x$] and an intermetallic ($Mn_{12}Si_7Al_5$) phase. The presence of conspicuous austenite dendrites and the large volume percent of austenite in the alloys suggest that the general sequence of transformations during solidification was as follows.

$$L \dashrightarrow \gamma + L \dashrightarrow \gamma + (\gamma+\alpha) \dashrightarrow \gamma+\alpha$$

B) Phase Transformations :

Phase transformations during furnace cooling from $1100^{\circ}C$ as well as during isothermal ageing in the temperature range 500°-$800^{\circ}C$ was investigated. In either case an intermetallic $Mn_{12}Si_7Al_5$ phase and a complex carbide $Al(Fe,Mn)_3C_x$ phase precipitated respectively at the grain boundary and inside the grains by solid state decomposition of the austenite phase. In addition, an ordering reaction leading to the formation of an ordered Fe_3Al phase occured at $500^{\circ}C$ and below. The extent of precipitation depended largely on the alloy composition and it could be appreciated from hardness measurements. Alloys with 5%Al were least susceptible to phase transformations. A 5%Al, 0.3%C alloy did not undergo any transformation either during furnace cooling or on ageing. Increasing the carbon content to 0.9% at 5%Al level resulted in only a small increase in the ageing tendency. The 7.5%Al and 10% Al alloys however suffered extensive phase transformations, alloys 2B and 3B (7.5% and 10% Al with 0.6%C) being conspicuous exceptions. The sequence of transformations depended on the heat treatment schedule. On continuous furnace cooling, the intermetallic $Mn_{12}Si_7Al_5$ phase precipitated first between 1100°C and 900°C. Carbide precipitation started below 900°C although extensive precipitation occured only below 700°C. On ageing at 800°C, only the intermetallic phase precipitated in some of the alloys. At 600°C, the first stage of ageing involved the precipitation of the carbide phase within the austenite grains which was followed by separation of intermetallic $Mn_{12}Si_7Al_5$ phase at grain boundaries. At 500°C, the ordering reaction occured first. This was followed by precipitation of carbide and intermetallic phases successively. In 10%Al alloys, the two stages of precipitation overlapped. The precipitation of either an intermetallic phase or carbide depleted the austenite matrix with respect to manganese. As a result, the austenite matrix could no longer hold the entire complement of aluminium and carbon in solid solution which, in turn, promoted the next stage of precipitation. An increase in the solute elements aluminium or carbon therefore increased the susceptibility to transformations.

(C) <u>Oxidation Resistance</u>

Fe-30Mn - (2-9)Al - (2.0-2.3)Si - (0.55-0.94)C alloys were found to have good oxidation resistance in the temperature range 700° - $900^{\circ}C$. Oxidation resistance improved and the oxide scale thickness decreased (a) with increase in aluminium percentage (b) on addition of misch metal to the alloys. Alloys treated with misch metal had good oxidation resistance even at $1000^{\circ}C$. The effect of misch metal addition on improvement of scale adherence and oxidation resistance was most conspicuous in case of 2% Al alloys. Kinetic treatment of oxidation data (log mass gain/unit area vs. log time) indicated that at the initial period of oxidation parabolic law was followed by all the alloys. During later period of oxidation, higher aluminium alloys (> 7% Al) (with and without misch metal addition) deviated from the parabolic behaviour and finally the process of oxidation stopped completely. In the higher aluminium alloys (7-9%Al) major constituent of the outer layer of the adherant oxide scale was alumina. In misch metal treated samples Ce_2O_3 and $AlCeO_3$ were also present. From the available evidences it may be concluded that the initial scale was formed by oxidation of manganese. MnO was later replaced by aluminium oxide by substitution reaction. Morphology of the oxide scale was nodular in most of the oxidized samples. In the misch metal treated 2% Al alloy, a transition from nodular to granular growth pattern occured on increasing the temperature from 800° - $1000^{\circ}C$.

(D) Machinability

The machinability study was conducted on the Fe-32Mn - 5.5Al - 2.3Si - 0.55C alloy. The experimental variables were (a) cutting speed, (b) cutting temperature. In general, the alloy did not have satisfactory machinability at ambient temperature due to work hardening tendency. However, machinability improved considerably on preheating the tool/stock contact surface by oxyacetylene flame to 600°C. This was evaluated by cutting force measurements as well as by examination of chip characteristics such as appearence, microstructures, hardness measurements, chip reduction co-efficient and side flow of chips.