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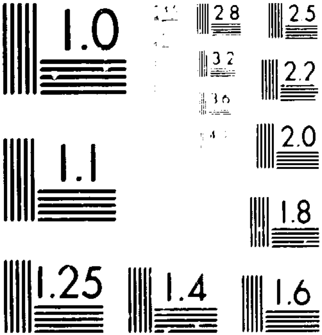

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DEVELOPMENT OF CHEAPER CRYOGENIC STEELS AND HIGH STRENGTH MARAGING STEELS*

by

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INTRODUCTION

Most of the recently developed high-strength and cryogenic steels are expensive since they contain nickel as a major alloying element. Manganese is a cheaper alternative to nickel and produces similar effects upon the austenite to ferrite transformations (2) which could allow cheaper alternatives to nickel steels.

Whether Fe-Mn alloys can be used as a basis for cryogenic high-strength steels, however, will depend on the mechanical properties that can be achieved. Earlier work has (2,3) indicated that although comparable strength levels could be obtained in Fe-Mn alloys to those of Fe-Ni alloys, such alloys were very brittle. This brittleness occurred at the prior-austenite grain-boundaries and was thought to be due to temper brittleness. Subsequent work by Freeman (25) and Gabbitas (26) confirmed these findings but no insight was obtained into the nature of the embrittlement mechanism.

The present studies were undertaken to identify the nature of this embrittlement in ferritic iron-manganese alloys and determine methods of improving the low-temperature toughness of these alloys.

THE EFFECT OF THERMAL CYCLING TREATMENT.

Lath martensite forms the basic microstructure of 9% Ni cryogenic steels and 18% Ni maraging steels (1). The same microstructure is obtained in Fe-8.0% Mn alloys at all cooling rates (2) and therefore may form a cheaper base for alternative steels. However, the Fe-8.0% Mn alloys suffer from grain boundary embrittlement (3). Auger spectroscopy has shown recently that the embrittlement is due to segregation of Mn and N to prior austenite grain boundaries (4). This paper reports the results of attempts to improve the impact toughness of the material studied in (4) by thermal cycling treatments (5-9).

EXPERIMENTAL PROCEDURE AND RESULTS

The composition of the alloy studied is given in table 1.

TABLE 1 Composition of alloy K1525

Mass	ppm												
% Mn	Ti	Cr	Mo	C	N	Si	S	Ni	Al	P	Sn	Sb	As
8.10	-	20	-	40	30	115	100	27	-	60	50	10	27

Transformation points determined by dilatometry at heating and cooling rates of 50^{o}C/Min are given in table 2.

TABLE 2 Transformation Temperatures of alloy K1525

$M_s = 360 \pm 5^oC$	$M_f = 366 \pm 5^oC$	$A_s = 677 \pm 5^oC$	$A_f = 723 \pm 5^oC$

From these transformation temperatures the thermal cycling treatment shown in figure 1 was devised. The holding temperatures and times were selected on the basis that the austenitising temperature should be low enough to minimise grain growth, while the temperature of holding in the $(\alpha+\gamma)$ phase region should be high enough to maximise the extent of diffusional transformation to low manganese ferrite and high manganese austenite (reverted austenite).

After two complete cycles, the austenite grain size was reduced from 80-90um. The impact transition temperature determined from sub-standard Charpy V-notched specimens of 5x10 mm section, was reduced from +115°C to -60°C, figure 2.

During holding in the two-phase ($\alpha+\gamma$) region, reverted austenite froms which may subsequently transform to a lath martensite and/or martensite on cooling, depending on the composition of the reverted austenite formed in the two-phase region. The proportion of phases in the alloy after heat treatment were determined by X-ray diffraction (10) (11) using line intensity measurements from 4 peak combinations. The average values are given in table 3 and thought to be accurate to better than ± 1%.

TABLE 3 Phase analysis of alloy after heat treatment

Phases	Heat treatment					
	1h 1,000°C water quench (a)	(a) + 1A (b)	(b) +1B (c)	(c)+2A (d)	(d) +2B (e)	(e) + 15 mins at - 196°C
γ	) <2%	) <2%	5.8%	4.1%	8.0%	6.0%
ϵ	) 98%	) 98%	28.0%	4.0%	24.6%	26.3%
α			66.9%	91.9%	67.4%	67.?%

It is evident from figure 4 that the reverted austenite forms mainly at the prior austenite grain boundaries and to a lesser extent at the inter-lath boundaries, as shown in the dark field electronmicrograph figure 7.

After thermal cycling, the nature of the brittle fracture changed from intergranular to that shown in figure 5; where fracture was mainly by cleavage with ductile regions apparently corresponding to the grain boundary regions which originally consisted of reverted austenite.

Previous work (12) has shown that alloy K1525 in the initial heat treated condition (ie (a) rapidly embrittles on ageing at 450°C. On tensile testing at -78°C at a strain rate of 0.5 min^{-1} the reduction of area value dropped to zero after 5 minutes ageing at 450°C. The thermally cycled material was therefore subjected to the same tensile test after ageing at 450°C to see if embrittlement could be induced in these specimens. The results are shown in figure 3 and the corresponding X-ray phase analysis in table 4.

TABLE 4 *Phase analysis after ageing thermally cycled material (1A + 1B + 2A + 2B) at 450° C*

Phases	Thermally Cycled (Th.Cy)	Th.Cy+10 mins 450°C WQ	Th.Cy+1h 450°C WQ	Th.Cy+2h 450°C WQ	Th. Cy+2h 450°C WQ + 15 mins - 78°C
γ	8.0	12.5	9.5	10.6	6.0
ϵ	24.7	11.8	14.5	8.8	9.5
α	67.3	75.7	76.0	80.6	84.5

The peculiar stress/strain curves obtained are thought to arise from deformation induced transformation of $\gamma \rightarrow \epsilon$ martensite and/or α martensite or $\epsilon \rightarrow \alpha$ Such phenomenon has been observed in TRIP steels (13) and increases the toughness of the steel.

DISCUSSION

Holden et al (2) and more recently M.J Schanfein et al (14) have reported on the excellent impact toughness of Fe-Mn alloys containing ($\gamma+\epsilon$) phases. Clearly the improved impact toughness of the present alloy can also be attributed to the introduction of these ductile phases into the microstructure as well as to grain refinement. M.J. Schanfein et al (14) report that the DBTT is lowered by 1.3°C per volume % ($\gamma+\epsilon$). Applying this figure to the present results suggests that of the total shift of 175°C in DBTT. $\sim$45°C is due to the presence of ($\gamma + \epsilon$) phases and $\sim$130°C due to grain refinement. This latter figure, (130°C) would appear to be rather large for grain refinement alone (15) and indicates a synergistic interaction between grain refinement and the presence of ($\gamma + \epsilon$) phases. From figures 2 and 7 of Roberts' work (15) the reduction in prior austenite grain size from 80-90 um to 10-15um in the present alloy, corresponds to a shift in DBTT 50°C).

The exact role of the ($\gamma + \epsilon$) phases in reducing embrittlement is not clear. It has been suggested (16) that:-

(a) Austenite may act as a sink for impurities, in this case N, reducing embrittlement during heat treatment (17)

(b) The ductile phases ($\gamma + \epsilon$) may act as crack arresters blunting the propagation of brittle cracks (18-21).

(c) Transformation of austenite to α-martensite and/or ϵ-martensite may occur during impact testing improving toughness (22,23,24). Evidence for this is provided by figure 3.

Present work on this and other alloys is aimed at establishing the relative importance of such parameters.

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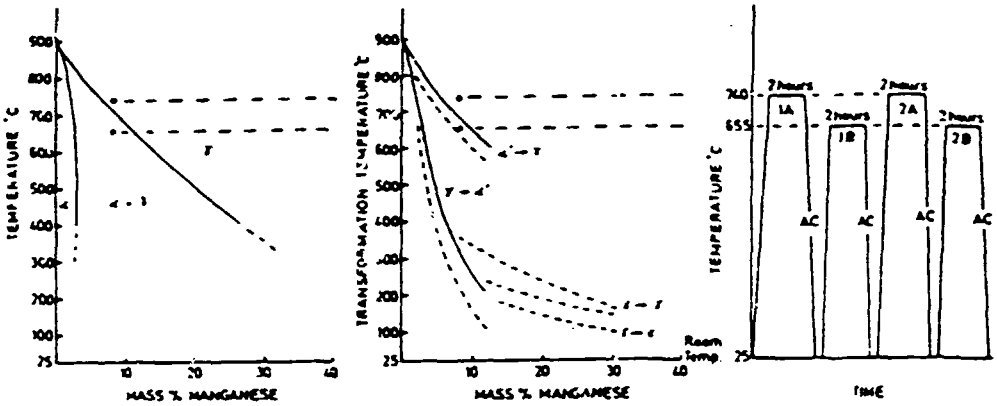

Figure 1 Schematic representation of thermal cycling treatment given to alloy

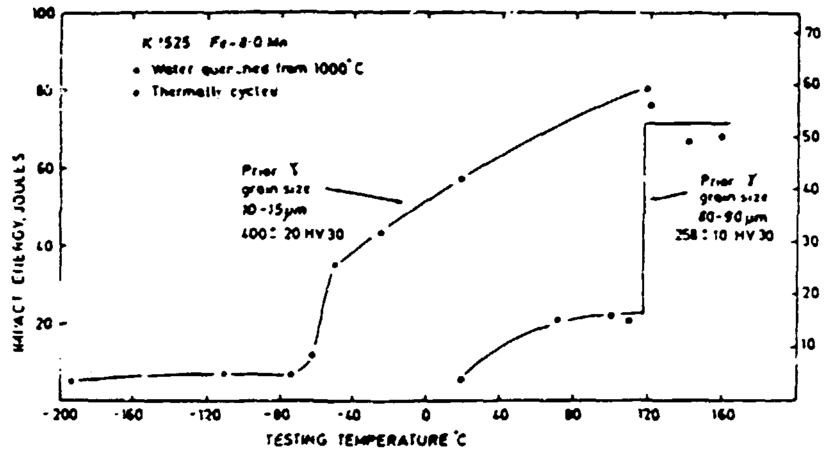

Figure 2 Effect of thermal cycling on impact toughness of alloy

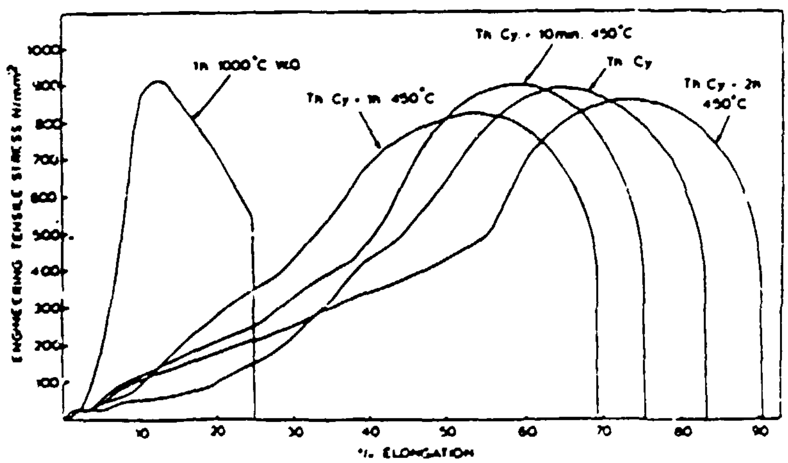

Figure 3 Stress-strain curves of alloy after various heat treatments. Tested at -78°C and a strain rate of 0.5 min^{-1}

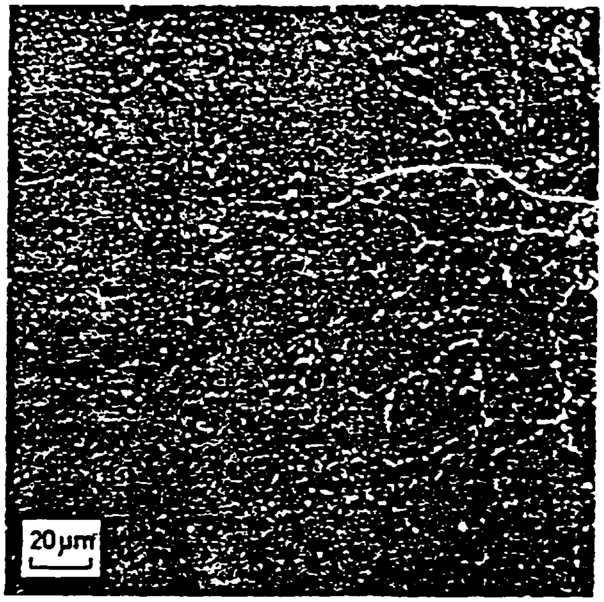

Figure 4 Optical micrograph of alloy K1525 after thermal cycling

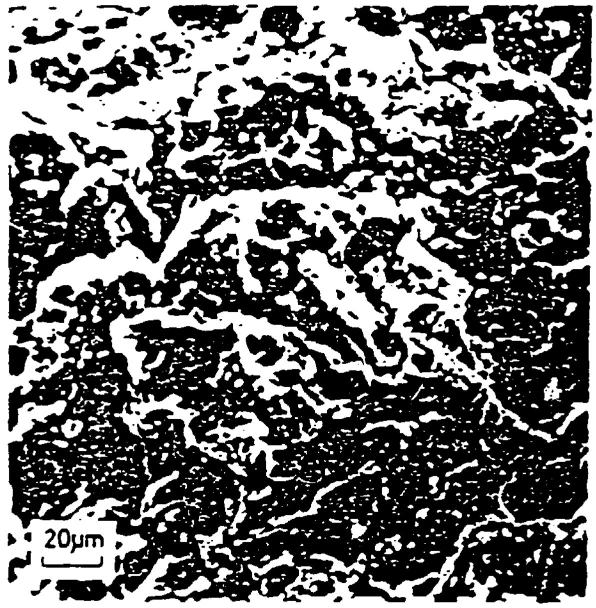

Figure 5 Scanning electron micrograph of brittle fracture of alloy K1525 after thermal cycling

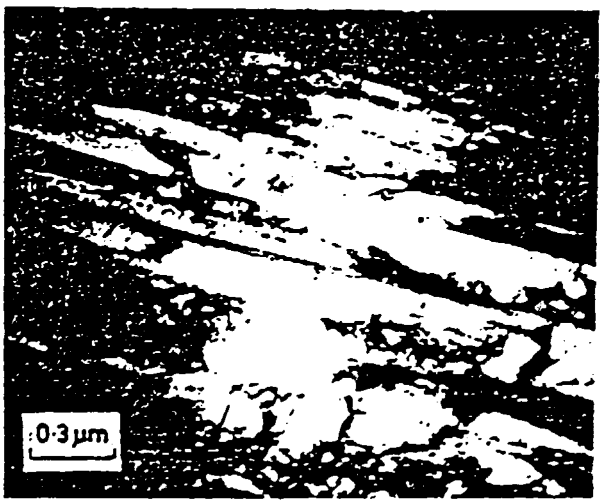

Figure 6 Bright-field transmission electron micrograph of alloy K1525 after thermal cycling

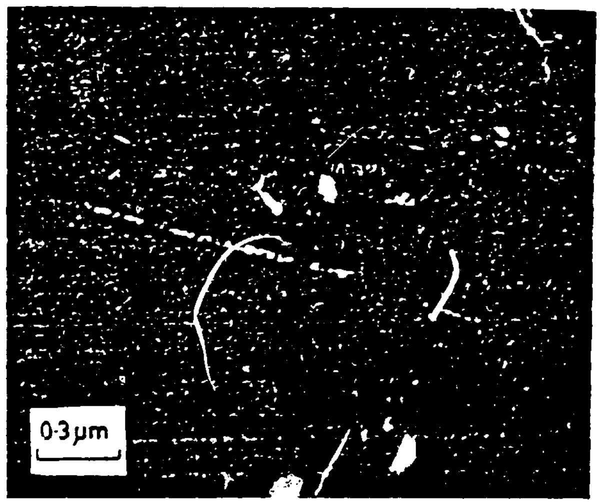

Figure 7 Dark-field image of figure 6 using (200) γ austenite reflection, illustrating inter-lath formation of reverted austenite.