CHAPTER 7. REQUIREMENTS FOR ADVANCED PERFORMANCE ARMOUR STEELS

7.1 Introduction

The optimisation of the microstructures and mechanical properties of the armour plates in a quenched and tempered condition has been undertaken with the intention of improving the ballistic performance of the steels A, B, C and D currently produced and used for military, civil and security purposes. Standardised ballistic testing of plates of new armour steels with 6 mm thickness assessed the improved resistance against ballistic impact. The reliability of the 6 mm armour plates was assessed by the ability of thinner plates ranging from 4.7 to 5 mm to withstand the direct impact of 5.58 mm rounds fired by a R4 rifle at zero degree obliquity from a distance of 30m at muzzle velocities ranging between 930 and 1050 m/s, as specified currently for civil and military applications with armour plates thicker than 8.5 mm. Since armour plates with thickness of 8.5 to 20 mm are currently being used to obtain the same ballistic performance, this makes the new alloys a significant improvement in weight reduction. It represents a reduction in weight of the plate used in manufacturing light vehicles for security and combat purposes from 66.3 kg/m^2 to less than 35.1 kg/m^2, a possible reduction of almost 50% in weight of the protected areas in these vehicles. This will reduce fuel consumption and increase the maximum speed attainable in these vehicles and will also open new global markets for the local steel industry. Thirteen advanced performance armour plate steels have been tested successfully in the conditions mentioned above. Through thin foil transmission electron microscopy, scanning electron microscopy and X-ray diffraction techniques from carefully sectioned samples at and near the impact areas after ballistic testing, it was confirmed that the success of these new alloys lies in their ability to undergo Transformation Induced Plasticity or "TRIP" of retained austenite upon impact together with a transition from twinned plates to slipped lath martensite that consume a significant part of the kinetic energy of the fired rounds.

Three groups of armour steels were identified by considering together the martensite start temperature, the volume fraction of retained austenite and the morphology of the martensite.

7.2. <u>**Group 1:**</u> consists of armour steels containing 1 to 6% volume fraction of retained austenite. Steels of this group have a higher ability to withstand rounds from R4 rifles fired from a distance of less than or equal to 30 metres from the plate. The martensite start temperatures of these steels range from 130 to 210°C. The morphology of the martensite formed in this temperature range is twinned plates containing nodules of retained austenite in the plate interfaces and on grain boundaries. Thin foil Transmission Electron Microscopy of the regions of the 4.5 to 5.0 mm thick plates of Group 1 deformed upon multiple ballistic impacts reveals the combination of two operating mechanisms in resisting ballistic perforation. The kinetic energy of the fired rounds is absorbed in the impacted regions in two ways:

- The transformation of the retained austenite to martensite by the well known TRransformation-Induced Plasticity or TRIP effect; and
- through reaustenitisation subsequent to the heating up of the plate martensite by entropy trapping inside the so called "adiabatic region" at the centre of the impacted zone and the formation of new slipped martensite upon being cooled down by the surrounding material. It needs to be recognized,

however, that the reaustenitised material probably has a higher M_s temperature than originally as little carbide dissolution may take place under the dynamic conditions of shock wave propagation

7.3. Group 2: comprises armour steels whose martensite start temperatures are between 210 and 280°C. This Group 2 armour steels has a combination of high strength, high Charpy-V impact energy at -40°C and a high hardness. The yield strengths of the steels in this group are higher than 1500 MPa, the ultimate tensile strength higher than 2000 MPa, the elongations on 50 mm gauge length are higher than 7% at room temperature, the Charpy-V impact energies of sub-sized specimens is above 13 Joules at -40°C and the hardness values are higher than 570 BHN.

The combination of high mechanical properties has inspired the design of armour steels in the past. The current martensitic armour steels such as steels A, B, C and D belong to Group 2. The minimum thickness of the plates of these armour steels required to withstand the R4 rifles fired in the same conditions than above, is 8.5 mm. The volume fraction of retained austenite in these plates is less than 1% and the martensite formation is generally accompanied by slip of dislocations. Some regions of the martensite contain twins on a background of laths. The coexistence of the laths and the twinned regions within the martensite was clearly observed in the three-dimensional surface relief AFM images and was also observed by thin foil transmission electron microscopy. The retained austenite in these steels is present as thin films along the lath interfaces and does not appear to undergo any TRIP. It remains untransformed upon ballistic impact, contrary to the nodular austenite located on plate interfaces and grain boundaries of the Group 1 armour steels.

7.4 Group 3: comprises armour steels whose martensite start temperatures are higher than 280°C. They contain less than 1% retained austenite and have the highest Charpy-V impact energy at -40°C and % elongation at room temperature. Their yield and ultimate tensile strength are intermediate between those of Group 1, the lowest, and those of Group 2, the highest. Armour steels of Group 3 are very sensitive to auto-tempering upon quenching.

7.5. General requirements of high performance armour steels
The location and the morphology of the retained austenite in all of these steels seem to be a function of the martensite start temperature. The TRansformation Induced Plasticity effect appears to be more effective when the martensite is formed at lower temperatures rather than at higher temperatures. Furthermore, nodular retained austenite seems to transform more readily by the strain-induced mechanism. Further investigation is needed for a better understanding of the conditions that determine the location and the morphology of the retained austenite in these martensitic armour steels.

The mechanical properties and ballistic performance of martensitic armour steels are strongly dependent on the chemical composition and the heat treatment cycle. The tensile properties are difficult to measure in untempered conditions for the armour steels of Groups 1 and 2 because of their brittle behaviour. On the other hand the martensitic armour steels of Group 3 are ductile from auto-tempering during quenching and their tensile properties are measurable in the as-quenched condition. The steels of Group 3 also have a high Charpy-V impact energy of sub-sized specimens in the untempered condition. The tensile properties as well as their impact energy at -40°C of all of these armour steels are improved by tempering between 200°C and 250°C for times less than 1 hour. Tempering above

300°C for the same time leads to a reduced tensile strength due to the precipitation of coarse cementite particles in the martensitic matrix.

Thin foil transmission electron microscopy of the tempered steels confirmed the well-known effect of Silicon in retarding the decomposition of the supersaturated martensite and the formation of cementite. Martensitic armour steels containing 0.8 to 1.2%Si may resist over-tempering at 300 to 350°C without forming coarse cementite that is detrimental to the mechanical properties as well as to the resistance against ballistic perforation. Thin carbides start precipitating along the twinned plate interfaces upon tempering of these armour steels, which is in agreement with the results published by Kelly [100], who has confirmed that the cementite thus formed obeys the Bagaryaski orientation relationship with the ferrite.

7.6. Design philosophy of high performance armour steels

From the comparison between mechanical properties and ballistic performance of the 13 steels tested in the conditions described in Chapters 4 and 5, it appears that the highest ballistic performance is reached by heat treatment conditions that are contrary to those required for the combination of the highest mechanical properties for a given chemical composition of the steel. This observation is in agreement with the published results on the ballistic performance of armour steels in general and, thereby, constitutes the basis for the development of advanced performance armour steels. This observation is, therefore, in contradiction to the current design philosophy for armour steels used hitherto. Neither the high tensile properties nor the high Brinell hardness number of the steel are ideal criteria within themselves in predicting the ballistic performance of armour plates or the dynamic resistance of structures to impact loading [6,7, 12, 13, 14, 15, 16, 17, 18]. Instead, the ratio of the yield strength to the ultimate tensile strength of the material and the volume fraction of retained austenite contained in the martensite are an improved indication of the steel's ability to resist localised yielding that favours ballistic perforation. The steels whose YS/UTS ratio is lower than 0.6 and contain 1% to 7% of retained austenite in twinned martensite, appear to present a low tendency to localised yielding upon high velocity impact loading.

This behaviour may be quantified through a function that includes the diameters of the impact affected zones around the incidence point after ballistic testing. However, it is difficult to establish a direct relationship between mechanical properties and ballistic performance of armour steels. The differences between mechanical properties and ballistic performance for a given armour steel and the reason why an armour steel plate performs ballistically better than another, should rather be explained in terms of their microstructures as well as their response to high velocity dynamic loading. Indeed, published data have reported measured dynamic tensile stresses in steel plates subjected to ballistic impacts of as high as 28 GPa [16,17,18]. The dynamic measurable maximum stress is, therefore, 12 times higher than the ultimate tensile strength of current high strength steels. This fact renders the current criteria that are largely based on the absolute values of the yield or the ultimate tensile strength in predicting the ballistic performances of armour steels, inappropriate.

In this study the resistance to localised yielding by the ballistic impact has been quantified by the diameters of three concentric zones formed around the incidence point. It appears that the diameter of the inner zone 1 is almost equal for all the steels and is slightly larger than the fired round's diameter. On the other hand the diameters of the outer zone 3 are

larger for those steels whose volume fraction of retained austenite is higher and their yield strength to ultimate strength ratios are lower. The relationship between the volume fraction of retained austenite in the martensitic steel, the diameter of the outer zone 3 and the thickness of the plate may then be used in defining a criterion for its ballistic performance. The yield strength to ultimate tensile strength ratio (YS/UTS), which also appears to be a function of the volume fraction of retained austenite in the martensitic steel through the TRIP effect, may be introduced as a dependent variable in this relationship. The uncertainty in predicting the ballistic performance of the steels considered in this study using the Ballistic Performance Index (BPI) proposed by Srivathsa and Ramakrishnan [6,7] may be explained by the lack of a relationship between the independent variable, i.e. the volume fractions of the phases present in the microstructure and the dependent variable, expressed as the yield strength to ultimate tensile strength ratio. The Ballistic Parameter BP has, therefore, been defined in Chapter 5 to account for the effect of the volume fraction of retained austenite and the thickness of the steel plates on the diameters of the outer zone 3 of the ballistic impact-affected regions, which indicate the resistance to localised yielding due to impact loading upon high velocity ballistic impact. It has been observed in this study that the combination of a high austenitisation temperature ranging between 870 and 950°C together with a low tempering temperature ranging between 170 and 200°C, is favourable for a low YS/UTS ratio and this provides a high resistance to localised yielding of the plates in thicknesses smaller than 5 mm. When the Silicon content of the armour steel is high the plates may be tempered at 350°C and still have a high ballistic performance. Contrary to this design philosophy, a low austenitisation temperature of 800 to 850°C and a high tempering temperature of 250 to 300°C, generally improve all the tensile properties and the sub-zero impact energy but reduce the steel's resistance to ballistic perforation.

7.7 Localised microstructural features of impact loading

Shock induced transformations and transitions occur respectively inside zones 1 and 3 of the impact affected regions. Quasi-adiabatic conditions prevailing inside zone 1 are favourable for the conversion of kinetic energy into heat. The localised temperatures in those zones are high enough to induce melting and welding of the fired round's material onto the steel plates, dynamic reaustenitisation of the twinned plate martensite and subsequent formation of slipped lath martensite together with auto-tempering. The new lath martensite forms from austenite with likely a higher M_s than originally as it is unlikely that carbides will fully dissolve during the brief and dynamic temperature rise. The Vickers micro-hardness profiles across the sections of the impacted regions show global hardening of these regions but with relative softening of zone 1 compared to zones 2 and 3. Inside zone 3 many dislocations are produced. Thin foil transmission electron microscopy of zone 3 shows dislocation pile-ups at twinned plate interfaces, demonstrating the high resistance of the twin interface to shear by slip of dislocations. New martensite is formed by a strain-induced mechanism of the retained austenite, i.e. the TRIP effect. No retained austenite was detected by X-ray diffraction of the impact-affected regions after ballistic testing. Rather, thin foil transmission electron microscopy of the same regions revealed new untwinned martensite that was formed by transformation of nodular retained austenite located in plate interfaces.

Scanning electron microscopy of the three-dimensional cracks revealed the character of dynamic cracking with microscopic cracks propagating along the grain boundaries. This observation may be a consequence of the high resistance provided by the twinned plate interfaces against the movement of dislocations throughout the martensite crystals and throughout the grains. Former parent austenite grain boundaries are then less resistant for

the propagation of microcracks induced by the transmission of that part of the kinetic energy of the fired round that is converted into potential energy of the compressive and tensile stress waves crossing the plates. From the simulation using an adaptation of the stress distribution model in space and time upon impact loading proposed by Hopkinson and modified by Taylor [5] and taking into account the likely true fracture stress as reported in the literature, the fracture of the untempered plates is predicted to occur 11 to 20 cm away from the incidence point for an incidence velocity of the fired rounds between 930 and 950 m/s. Tensile stresses near to 12 GPa may be developed within the armour plate after three or four reflections of the tensile stress wave from the edges of the plates, which were 30 cm long and 20 cm wide. The profiles of the shear bands across the fractured sections of the untempered plates suggested a vibratory dynamic response of the plates to the ballistic impact loadings that excited some natural frequencies within the plates according to their harmonics. The sizing of the plates to be used should then consider a lowest natural frequency of the plate structure larger than the highest firing frequency of the R4 to avoid synchronisation. Smaller and thinner plates may be the remedy for mechanical resonant failures.

7.8 Martensite characterisation in these advanced performance armour steels

The martensite was characterised using the lattice parameters of the martensite and of the retained austenite as measured by X-ray diffraction, as inputs in predicting the crystallographic features by the Bowles and Mackenzie model of the Phenomenological crystallographic Theory of the Martensitic Transformation. This was found to be in good agreement with the predictions of the existence of three plateaux distinguishing the martensite formed by an internal twinning mechanism, $\{2 \quad 2 \quad 5\}\gamma$ butterfly martensite formed by a slip mechanism in the austenite and $\{1 \quad 1 \quad 1\}\gamma$ lath martensite formed by a slip mechanism in both the austenite and the ferrite. The three formation mechanisms of the martensite were observed by qualitative AFM analysis of the surface relief accompanying the martensite formation.

- Surface relief of armour steels in Group 2 whose martensite start temperatures ranged from 210 to 280°C, were irregular N-shaped which revealed no character of the IPS. The three-dimensional AFM images and the thin foil TEM of these steels showed the coexistence of twinned plates and dislocated lath martensite. Thermodynamic data reported by Morozov et al [21] suggest that this transition in morphology is likely to occur between 232 and 284°C for Fe-C-X systems.

- The surface relief accompanying the formation of martensite at temperature above 300°C in Group 3, indicated high plastic strain accommodation. However, relief deterioration is also possible in the case of this relatively high Ms temperature through thermal smoothing and oxidation.

- The Bowles and Mackenzie model of the PTMT predicts a crystallographic orientation between the martensite and the austenite close to the Kurdjumov-Sachs orientation relationship in the armour steels of Group 1.

For the other two groups the PTMT predicts a larger deviation from the Kurdjumov-Sachs orientation relationship, suggesting the existence of an influence from the relative crystallographic orientation between the martensite and austenite crystals within the plates

or laths on the ability that grain boundaries have in resisting spallation and crack propagation by dynamic loading. Indeed, backscattered scanning electron microscopy in Figures 6.2 and 6.3 show big differences between the plates and laths formed within the grains that may have some influence on the microscopic geometric configuration of the grain boundaries. It is well known from the theory of formation and growth of martensite that the growth of plates and laths is arrested at parent austenite grain boundaries. The microscopic configuration of the grain boundaries will then be the result or consequence of the formation history of the martensite inside the grains.

An effective control of the state of the grain boundaries and of the subsequent ballistic performance or of the mechanical properties, depending on the application, may then be achieved by controlling what is happening inside the grains through appropriate design of the chemical composition of the armour steel and by applying the relevant heat treatment.

7.9. Proposed revised specification for advanced performance armour steels

The specifications for the advanced performance plate armour steels may be revised as follows:

- The optimum chemical composition should range between:

Element	%C	%Mn	%Si	%Mo	%Cr	%Ni	%Cu	%P	%S
Range in weight percentage	0.38 - 0.43	0.4 - 2.0	0.4 - 1.2	0.4 - 0.6	0.4 - 1.5	1.0 - 4.5	<0.2	<0.005	<0.005

- The martensite start temperature of the steel should be lower than 210°C;

- The volume fraction of retained austenite in plate martensite should be higher than 1%;

- The heat treatment should consist of austenitisation at temperatures between 850 and 950°C for less than 1 hour, followed by water quenching to room temperature;

- Tempering should be undertaken at temperatures ranging from 150 to 180°C for 20 to 60 minutes when the Silicon content is lower than 0.6%. The tempering temperature may be raised to 300°C when the Silicon content is higher than 1%;

- The design methodology should be based on the YS/UTS ratio which should preferably be below 0.6 as well as on a value of the Ballistic Parameter of 0.018 to 0.060 to predict the ballistic performance;

- Small size and thin armour plates should be preferred to reduce the risk of mechanical resonance of the armour plate with the firing frequency;

- Manganese sulphide and coarse carbides are detrimental to ballistic performance as well as to mechanical properties of the tempered plates; and

- Under these conditions armour steels with a Brinell hardness of 475 BHN and a Charpy impact energy at -40°C as low as 10 Joules, are acceptable.

Chapter 7: Requirements For Advanced Performance Armour Steels