Materials on Wheels: Moving to Lighter Auto-bodies

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Abstract—The weight of the automobile body has been the serious concerned as it has negative effect on the fuel consumption and pollution emission. Steel materials have been applied as body parts since the beginning of the automotive industry development. Continuous research and development in term of composition, melting and refining, heat treating, etc., resulted in variety of many high quality materials such as Dual -Phase (DP) Steel, Transformation-Induced Plasticity Steel (TRIP) , Interstitial Free (IF) Steel., etc. Non ferrous and reinforce polymers are also applied as alternatives for lighter weight and ease of fabrication. Historical development of these materials with technical implication is discussed with the aim for better understanding of materials application for body parts.

Index Terms— Steel sheet materials, automobile, light weight materials, formability

I. Introduction

The drive to reduce the weight of auto-bodies began during the mid-1970's when the price of oil was significantly raised by the oil producers. This accelerated the development and application of sheet steels with higher strength levels than the traditionally used conventional mild (low carbon) steels in order to save weight by the use of thinner gauge sheet material for body pressings. Since then a number of higher strength grades such as Low Carbon High Strength Low Alloy (HSLA) Steels, Rephosphorized, Bake Hardening, Dual-Phase (DP) and Transformation Induced Plasticity (TRIP) steels, etc. have been increasingly used as structural automotive body parts. Their use has also served to reverse the trend of vehicles becoming heavier with each new model as a result of added refinements for customer comfort and to meet improved and stricter safety requirements.

Experience in the earlier applications of these higher strength steels showed that their processing and formability was different to that of the conventional mild steels used for auto body pressings. Some high strength grades could only be used for relatively shallow pressings and, in order to give consistent performance, all grades required modifications to design of the body part, press tooling, forming conditions, welding and finishing operations.

Towards meeting the need for improved fuel efficiency and better overall performance, from 1975 to 1990 the proportion of lighter weight materials i.e. polymers, rubbers and aluminium (Al) alloys used in motor cars gradually increased. Over that period, in a typical mid-range car, the percent volume proportion of steel used decreased from 33 to 22%, that of cast iron fell from 8 to 2%, while Al use increased in volume from 6 to 13%, polymers increased from 22 to 37% and rubbers from 22 to 37% [1]. Increased use was made of Aluminium-base casting alloys to replace cast irons in power train and transmission castings, and of wrought Al alloys in sheet form as alternatives to steel for body panels. In the UK for example, the weight of equivalent family saloon models was reduced from around 1300 kg in 1975 down to just over 1000 kg until the mid-late 1980s when increasing customer demands for extra refinements and government safety and environment legislation meant that, in spite of some use of thinner gauges of steel, lighter alloys, and more polymers, etc., the weights of cars began to increase. Forecasts in the mid 1980's suggested that conventional mild steels would continue to be used as the main car body material until tooling design and production techniques were sufficiently well developed to enable the full exploitation of higher strength steels [2]. At that time the general comment on replacement of steel by aluminium was that *"it* halves *the weight at double the cost"*.

Nevertheless, interest in aluminium alloy as a body material continued to grow due in part to the establishment in 1993 of the United States Automotive Materials Partnership LLC (USAMP) which was part of the Partnership for a New Generation of Vehicles (PNGV) initiative between the US Dept. of Energy and Chrysler, Ford and General Motors. The initial general aims included the development of prototypes of low emission, fuel efficient family saloon cars that could provide the same space and comfort as conventional models, and not least, at the same price. PNGV was succeeded in 2003 by the FreedomCAR and Vehicle Technologies Program (FCVT) under which the materials focus was on light-weighting through use of dissimilar materials and on the problems of non-destructive evaluation of both structures and joints [3]. At the same time as PNGV

was being started the Al producers, seeing future market opportunities, increased their efforts for the development of Aluminium Intensive Vehicles (AIVs).

To address the increasing competition from Al-Alloys 35 of the main steel producers from 18 countries joined together in a project to develop a lightweight steel body structure under the Ultra Light Steel Auto Body (ULSAB) Consortium in 1994. ULSAB together with Porsche Engineering Services (PES) set out to design and build a lightweight steel autobody that met strict performance and cost criteria. One aim of the study was to show the capability of the HSLA steels in achieving body weight savings without the need for downsizing, and also in improving safety, comfort and overall performance [4].

Conventional Aluminium killed mild steels (AK steels) that were traditionally used have yield strengths of 150-190 MPa. The ULSAB project defined 2 main groups of higher strength steels as (a) High Strength with Yield Strengths of 210-550 MPa, and (b) Ultra-High Strength with Yield Strengths greater than 550 MPa. The sheet thicknesses used ranged from 0.65 to 2.00 mm. All of the sheet used for ULSAB was either electrolytically or hot dip double-sided zinc coated reflecting consumer and legislative demands for extended corrosion warranties. It must be noted that right up to the 1990s, and particularly during the 60s and 70s, the major complaint from customers was inadequate corrosion resistance of body parts (due to both unsuitable design and poor protection) which led to expensive repair work or early end of vehicle life, in spite of the engine and other parts still being in good condition [5].

In 1994, higher strength steel grades accounted for up to about 50% of the body weight of the latest cars but the ULSAB work showed that there was a potential for their 90% usage in auto-bodies [4]. Weight savings of up to 36% were said to be achievable when compared to the heaviest benchmark vehicle. In particular weight savings were achieved with improved structural integrity by the use of tailored blanks, which matched the strength and gauge of the steel to design requirements. Laser welded tailored blanks made up 50% of the structure, the rest being mainly conventional pressings with some sheet & tube hydroforming and steel sandwich forming. The use of such blanks enabled mass that did not contribute to performance to be removed and also allowed economies of production by reducing the total number of parts and associated tooling, reducing the number of spot welds and of assembly steps. Laser welding of the tailored blanks provided much higher dimensional accuracy than that previously achieved in steel bodies. The use of tube hydroforming enabled the production of thinner walled side roof rails with sheet hydroforming used to form the roof panels. Steel sandwich material with an inner core of polypropylene was used to press the spare tire tub

and for the dash panel. This sandwich material could be processed in the same way as steel sheet but required joining by adhesive bonding and riveting instead of welding.

In response to progress in Al alloy application, e.g. the Acura NSX–the first production car to have an all-Al body saving 200 kg compared to a steel body [6], and in the use of fibre reinforced polymers (FRP) for closures (bonnet boot lid and doors) [7], the ULSAB project was followed by the ULSAC project on closures, the ULSAS project on suspensions, and the overall ULSAB-AVC (Advanced Vehicle Programme) [8, 9]. For the last 10 years light-weighting and other development work has focused on the future need for electric and hybrid vehicles that reduce greenhouse gas emissions, the steel producers initiating the Future SteelVehicle (FSV) programme in 2008 [10-12]. During the extensive R&D carried out by the steel industry the processing and performance of a wide variety of higher strength sheet steels has been examined leading the industry to provide comprehensive guidelines to their application [13]. The range of properties available in these steels is illustrated by the strength-ductility relationships shown in Figure I Steels that contain significant alloy additions and contain 2 or more phases are often referred to as Advanced High-Strength steels (AHSS) in order to differentiate them from the conventional grades.
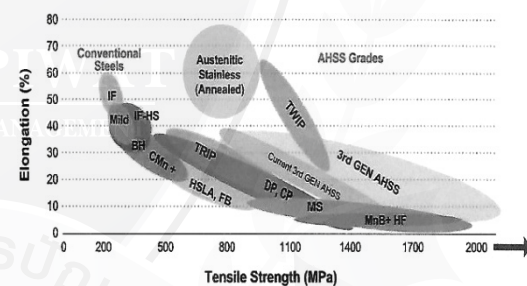

Fig. I. Tensile strength – ductility relationships in steels for use in auto-body parts [3, 9, 13]

This review article considers some key aspects in the physical metallurgy and processing of the various higher strength steels that may be used in auto-bodies and comments on the competition from light alloys, such as aluminium and magnesium, and polymer-based composites.

II. Formability of low carbon steels

Most steel auto-body parts are formed by pressing operations which involve a combination of stretching and deep drawing. Stretching behavior is controlled by the uniform elongation of the steel as measured in a tensile test since the limit of deformation is the onset of a localized region of thinning (necking) in the sheet. The greater the ability of the steel to undergo work (strain) hardening then the greater is its

resistance to necking during stretching deformation. Work hardening capacity can be described by the Work (Strain) Hardening Exponent (n) of the steel [14], where n is related to true stress (and true strain (ε) by the empirical equation:

$$\sigma = K . \varepsilon^n$$

where K is the strength coefficient.

The true strain at the point when necking begins at the end of uniform elongation is numerically equal to the value of n. Hence the higher the value of n the greater is the strain to which the steel can be stretched before local thinning occurs. Values of n for typical steel grades are listed, together with tensile properties, in Table I For conventional AK, Interstitial Free (IF) and HSLA steels the n values are taken as constant with respect to the amount of deformation and they are normally determined for the range of 10 to 20% elongation. Dual phase (DP) and Transformation-Induced Plasticity (TRIP) steels demonstrate different work hardening behaviour such that their n values are found to vary with amount of deformation [13]. The DP steels have microstructures consisting essentially of ferrite + islands of martensite which contain small amounts of bainite and retained austenite. Hence they exhibit different deformation behaviour to the essentially ferritic grades (AK, IF). In particular, DP steels undergo greater work hardening at engineering strains below 8%. Hence for DP grades n values are also quoted for the 4 to 6% engineering strain range [13, 15].

Table I.
SOME EXAMPLES OF TYPICAL MECHANICAL PROPERTIES
AND N VALUES [13].

GRADE	YS (MPa)	UTS (MPa)	El (%)	YPE (%)	n-value (10%-UE)	n-value (4-6%)
IF	163	282	49.0	-	0.240	-
DDS	163	314	42.9	-	0.230	-
CMn440	354	455	39.0	3.6	0.156	
HSLA340	377	443	28.4	2.8	0.170	-
SF600	490	615	28.6	3.2	0.148	-
HSLA480	487	604	28.0	2.2	0.128	
HSLA550	641	672	13.5	3.3	0.040*	-
DP600	340	602	29.0	-	0.170	0.208
T600	393	631	34.0	-	0.230	-
DP800	459	838	17.2	-	0.115*	0.168
T800	475	822	26.0	0.6	0.240	-
DP1000	726	1003	10.8	-	0.060*	0.08

Compared to DP steels TRIP steels contain higher amounts of retained austenite in their microstructures. This austenite transforms to martensite during deformation giving transformation induced plasticity such that higher n values are obtained throughout deformation and the danger of localized thinning is reduced during stretch forming [16]. For the DP, TRIP and other Advanced High Strength Steels (AHSS) data from plots of instantaneous n value against engineering strain must be used when predicting forming limit diagrams and in computer modeling of forming behavior [13, 15, 16].

Deep drawing behavior depends on tooling design and on the Plastic Strain Ratio r. The r value is controlled by the crystallographic texture (preferred orientation) that is developed during production of the sheet. The value of r is given by:

$$r = \frac{true\ strain\ in\ width\ direction}{true\ strain\ in\ thickness\ direction}$$

This is normally expressed as a mean value, r_m, from determinations using sheet tensile specimens taken at 0, 45 and 90° to the rolling direction. If the crystallographic texture in the sheet is such that it minimizes through the sheet thickness plastic strain then the sheet will be resistant to thinning and hence deep drawability is increased. For high formability an r_m value of nearly 2.0 is required. When $r_m < 1$ the sheet will have poor resistance to thinning on forming. In addition, for minimum earing tendency (i.e. uniform plastic flow in all directions in the plane of the sheet) during forming then the difference, Δr, between r values at 0, 45 and 90° should be as small as possible.

In low carbon (C) sheet steels for pressing the aim is to achieve the maximum amount of {111} texture [17, 18]. This means that the {111} planes in the ferrite are parallel to the rolling plane. High formability is achieved in the conventional continuously cast AK steels by close control of composition and hot & cold rolling + sub-critical annealing schedules. Batch annealing of cold rolled material, with a slow heating rate, allows controlled precipitation of Aluminium Nitride (AlN) at sub-boundaries in the un-recrystallized ferrite matrix which promotes the development of the "cube on corner" annealing texture. This is {111}<110> and gives r_m values of 1.8 to 2.0 and hence a high degree of formability during subsequent pressing. Continuous annealing, in which the cold rolled coils are uncoiled and passed through an annealing chamber involves faster heating rates and shorter times resulting in finer recrystallized grain sizes and hence higher yield strengths but with less favorable textures for formability. However, steels with very low (< 0.01%) C and N levels can be produced to give good formability even after continuous annealing [19]. These are called Interstitial Free (IF) steels and are processed by ladle decarburization under vacuum degassing conditions. Small additions of titanium and/or niobium are added to tie up interstitial C and N as carbides/carbonitrides resulting in lower yield strengths than for continuously annealed AK steels but providing potentially superior formability especially for the production of more difficult complex forms [20]. After forming the shape of a pressing must be maintained and there should be minimal tendency for springback, which has to be counteracted by over-bending parts. Springback causes variation from part shape design and it is caused by the elastic recovery of the pressing upon unloading when parts are removed from the tooling. It increases with

increasing yield stress and decreasing sheet thickness of the steel being formed. Hence springback is more severe in thin gauge high strength grades than in conventional low C steel [21]. A further problem is that springback in AHSS steels, DP and TRIP grades, is observed to be time-dependent at ambient temperatures: this is not the case in conventional mild and HSLA steels [22]. The AHSS grades show similar springback behavior to formed aluminium alloy sheet with time-dependent shape changes being proportional to log time ,however the final amount of change is about ⅓ that for aluminium. In general, it can be summarized that higher strength steels present greater formability problems since in addition to springback they tend to have lower ductility, and less favorable n and r_m values.

III. Strengthening in low c automotive sheet steels

The main strengthening mechanisms in low C steels include ferrite grain refinement, solid solution hardening of ferrite by Manganese (Mn), Phosphorous (P), Silicon (Si), etc., dispersion hardening of ferrite by carbide & carbonitride precipitates, and not least by the introduction of controlled amounts of martensite, bainite and retained austenite as in DP and TRIP grades. Additional minor strengthening can also be obtained by strain ageing, dislocation substructure hardening and texture hardening. Figure II shows typical tensile stress- strain curves for a mild steel, a conventional HSLA steel and a DP steel.
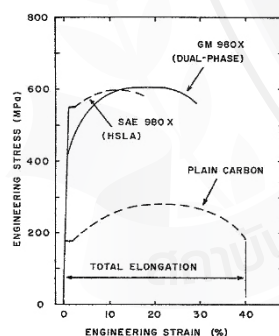

Fig. II. Example of tensile engineering stress-strain relationships for plain C (mild), HSLA & DP grades. Yield behaviour in the plain C and HSLA steels is discontinuous [30, 31]

Relatively small amounts of alloying elements can be added to low C (mild) steels to give dispersion strengthening via formation of very small niobium, titanium, or vanadium carbide/carbonitride precipitates, and to encourage the formation of fine ferrite grains [23-29]. The alloy carbides influence the austenite grain size and shape during hot rolling by pinning austenite grain boundaries and by controlling the recrystallisation behavior of the austenite. Nb is believed to be the most effective addition in retarding austenite recrystallisation since it not only pins grain

and sub-boundaries in the austenite by forming Nb carbonitride precipitates but also the Nb in solid solution exerts a solute drag effect on interstitial atoms and lattice defects delaying recovery [25-27]. The carbides limit recrystallisation such that "pancake" shaped grains of austenite are retained after hot rolling. This distorted austenite structure has a large grain boundary area available for subsequent nucleation of ferrite during the austenite to ferrite transformation on cooling and hence leads to fine ferrite grain sizes. In addition, some fine carbonitrides precipitate in the ferrite giving precipitation strengthening. As controlled hot rolled products the yield strengths obtained can be as high as 550 MPa. Similar fine ferrite structures and carbide/nitride precipitation are responsible for strengthening in the cold rolled reduced grades in which the precipitates also influence ferrite recrystallisation during annealing. The distribution of strengthening precipitates can be significantly affected by process variables during hot rolling notably temperature control and coiling temperature. Such variation influences the effects of annealing after cold rolling during both batch and continuous treatments. For example, higher coiling temperatures tend to give coarser precipitate particle dispersions leading to coarser mean final grain sizes and wider scatter in grain sizes and in mechanical behavior. Continuous annealing, as for mild steels, gives finer grain sizes and also enables inter-critical annealing + controlled cooling to be applied to produce DP and TRIP grades [31-33].

If a steel can be processed to produce a "dual phase" microstructure a significant amount of additional strengthening is gained from work hardening effects. This extra hardening is due to the presence of about 20-25% by volume of islands of martensite surrounded by a matrix of fine ferrite grains. In these islands small amounts of bainite and residual austenite are also present. Dual phase (DP) steels can be produced by intercritical annealing or by controlled hot rolling. Originally these steels were produced by continuous annealing in the austenite + ferrite (720-780°C) range where there is 15-25% by volume of austenite islands [31]. This austenite will contain segregated carbon atoms (up to about 0.3-0.5% C) and if sufficient Mn, Si and Cr is present the austenite will have sufficient hardenability to transform essentially to martensite on cooling. Some bainite and retained austenite will also be present in association with this martensite. DP steels normally contain 0.06-0.15wt%C and 1.5-3wt%Mn + small additions of Cr and Mo to prevent pearlite and excessive bainite formation with Si to form ferrite and Nb + V for structural refinement and precipitation hardening. Thicker gauge material is now usually produced by controlled hot mill processing of C-Mn grades by spray quenching after hot rolling [27, 31].

It can be seen from Figure II that DP steel has a lower yield strength than conventional HSLA grades and that it also exhibits continuous yield behavior due to the presence of highly mobile dislocations in the ferrite. This high density of mobile dislocations coupled with the deformation of the "martensite" islands is responsible for the initial high rates of work hardening. The result is improved formability while the absence of discontinuous yielding avoids the need for temper rolling prior to pressing.

As mentioned earlier, the textures that are developed during processing of DP steels and in conventional HSLA grades are not as favorable for deep drawability as in AK steels or other mild steels. For pressings where formability is the major concern, then solid solution hardened grades of AK steel are available. Phosphorus, silicon and manganese are used to solid solution strengthen the ferrite. P, in particular, has a strong solid solution hardening effect in ferrite. In the Rephosphorized AK steel for each 0.01% P addition yield strength is increased by about 10MPa with little effect on deep drawability. In steel P level is normally kept below 0.03wt%. However a P content of up to 0.1wt% is used in the Rephosphorized steels to provide yield strengths of 260-300 MPa. Solid solution hardening from P combined with the strengthening achieved during the low temperature baking treatment used to cure paint finishes is the basis for the Bake Hardening (BH) steels. In these grades C level is controlled between 0.010-0.015% to allow sufficient free C atoms to diffuse to dislocations generated during temper rolling and press forming. Such diffusion is possible at the curing temperatures (about 170^oC for 20 minutes) used in the paint bake ovens. The dislocations are then pinned resulting in a strength increase, due to strain ageing, of around 40 MPa. The bake hardening effect is less significant if the ferrite is not fine grained and/or if there is significant work hardening during forming. Yield strength levels from 270–340 MPa can be achieved by using Ti and Nb additions to control the amount of C and N in solid solution [34]. An added advantage of DP and other AHSS steels is that, depending on composition and forming conditions, they can also demonstrate BH strengthening.

In DP steels ductility was found to be influenced by the amount of retained austenite that is present in the martensite islands [35]. This led to the development of the low alloy TRIP steels in which composition and processing are controlled to provide a fine grained ferrite matrix with dispersed areas of martensite, bainite and at least 5% by volume of retained austenite. As introduced above, transformation of the additional retained austenite to martensite (compared to DP grades) provides a high degree of work hardening over the full strain range not just up to 4-6%. At equivalent strength levels (yield strength of 350 MPa/ UTS 600 MPa) a DP steel has an engineering strain of around 25% while a TRIP steel has 32% [13].

TRIP steels are produced by intercritical annealing followed by cooling to an isothermal transformation temperature to allow transformation of some of the austenite fraction to acicular (bainitic) ferrite [36-38]. TRIP grades are typically based on 0.2%C-1.5%Mn steel. Additions of 1.2-1.5%Si or 2%Al are used to prevent M_3C cementite carbides forming in the bainitic ferrite, the C atoms then becoming concentrated in the untransformed austenite.

The DP and TRIP grades have high capacity for energy absorption and hence have become increasingly used for parts requiring crash-worthiness such as safety cage components (B-pillar, engine cradle, floor panel, front sub-frame, etc.). Due to their lower C contents DP steels have better weldability than TRIP and also give better surface quality. The higher Si content in TRIP steel can cause defects in cast surfaces and also problems due to oxidation during processing and galvanizing [39]. It is possible to avoid problems due to high Si by producing low Si TRIP-assisted multiphase steels, e.g. 0.16wt%C-1.3wt%Mn-0.4wt%Si [38]. This steel is processed in the same way as a DP steel but is isothermally treated and then spray quenched to give a multiphase structure of ferrite, bainite, martensite and 10% residual austenite. Multiphase strengthening is assisted by a TRIP effect during subsequent forming.

A number of interacting factors are believed to influence the mechanical and other properties of the DP, CP and TRIP multi-phase steels. These include the ferrite grain size, secondary phase morphology, volume fractions of each phase (especially austenite), austenite stability, relative strengths of each phase, precipitation hardening, etc.

The DP, CP, TRIP and martensitic steels are quite often referred to as the 1St Generation of AHSS. The 2nd Generation includes the Austenitic Stainless Steels (ASS) and high Mn Twinning Induced Plasticity (TWIP) steels.

The TWIP steels are based on fully austenitic microstructures that are stabilized by the inclusion of up to 30%Mn and up to 9%Al in the steel composition [40-43]. The high solute content in the austenite imparts a low stacking fault energy such that slip mechanisms are restricted and plastic deformation takes place by twinning. High rates of work hardening are produced via a combination of deformation twinning, dynamic strain ageing and the FCC→HCP transformation of austenite to ε-martensite. The use of the 2nd Generation steels remains limited due to the high cost of alloying and due to problems such as obtaining consistent properties and the tendency for delayed cracking. The latter, environment assisted cracking (EAC), tends to occur in unstable austenitic stainless steels when high residual tensile stresses are present (as in a deep drawn part), and when some martensite is presence together with hydrogen [43]. One source of hydrogen

is from electrolytic galvanizing in coating to protect against corrosion.

Extensive R&D is continuing towards the development of a 3rd Generation of AHSS, which can provide, at low cost, a required property combination of UTS at 1200MPa + 30% ductility [e.g. 29, 32, 39, 44, 45]. Much of this work focuses on gaining improved understanding of how composition and processing parameters influence mechanical properties and formability. The key areas of development include Quenching and Partitioning Processing (Q&P), higher Mn TRIP-steels, and nano-structured steels.

In Q&P treatment austenite is quenched to a temperature QT below the martensite start (M_S) temperature to produce a controlled amount of initial martensite and then held at a partitioning temperature PT for a certain time to allow C atoms to diffuse into the untransformed austenite, and hence control austenite stability when the steel is cooled to ambient temperature [45, 46]. This treatment is illustrated in Figure 3. As in conventional TRIP steels Si is added to prevent cementite formation. Up to 4% of alloying elements (Mn, Si, Ni & Mo) can be used to control microstructure. Q&P treated steel is reported to be used for A- and B- pillar reinforcement [47]. Q&P treatment is also applied to intercritically annealed steel where the initial microstructure consists of austenite + ferrite.
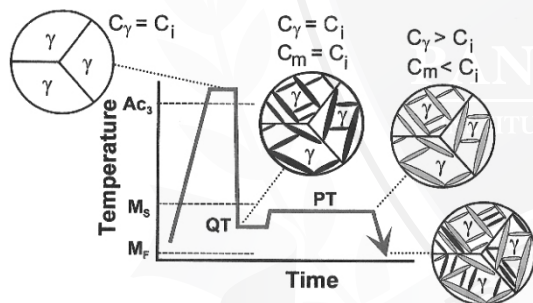

Figure III. Outline of Q&P treatment of a fully austenite matrix to obtain fine mixtures of martensite and residual austenite, Ci, Cm and Cγ represent C levels in the initial alloy, in martensite and in γ-austenite [45, 46].

Medium Mn-low C steels containing 5-7wt% Mn can be intercritically annealed to produce microstructures of very fine grained ferrite and austenite, the stability of the austenite phase being increased by enrichment with Mn during annealing [48].

An alternative processing method for medium Mn steels is being developed by making use of the "Ghost Pearlite" effect [49]. This is the fine-scale chemical patterning of Mn in the austenite produced by austenitising a pearlite structure that was formed at 500-600°C. Such pearlite has a strong partitioning of Mn into the cementite. On cooling this pattern enriched austenite only partly transforms during isothermal treatment to form a layered microstructure of ferrite/martensite and metastable residual austenite on a size scale similar to that of the original pearlite. This fine structure can be tempered to give UTS levels of 1600-2100 MPa and 7-10% elongation. This is comparable to the properties that can be obtained from the nano- or super-bainites outlined below, however the chemical patterning effect involves much shorter treatment times than those needed to produce nano-bainites.

Very fine bainitic structures have been developed in 0.8C-1.5Si-2Mn-1Cr-0,25Mo steels using low temperature isothermal treatment at 200°C, however holding times of up to 96 hours may be needed. The fine platelets of bainite ferrite are only 20-40 µm in thickness hence the ferrite + austenite mixture produced has been termed nanostructured bainite [50]. Although there is interest in nano-complex phase structures the slow rate of structure formation of nano-bainite seems much more suited to thick sectioned parts rather than sheet metal forms. However, isothermal transformation time may be reduced by alloying with Al and/or Co [51]. Nano-structured steels are considered to have microstructures containing a very high density of strong interfaces rather just contain a small amount of nano-sized precipitates [50].

Nano-lath martensitic structures can be produced by Q&P treatment followed by tempering, i.e. Q-P-T. Tensile strengths over 2000 MPa with elongation over 10% have been observed in a medium C steel (0.48C-1.19Mn-1.18Si-0.98Ni-0.21Nb) with nano-micro-structures of lath martensite, thin films of residual austenite and fine dispersed Nb-containing carbides [52].

A further group of interest are the Press hardening steels (PHS). C-Mn steels alloyed with up to 0.005wt%B are austenitised and then subjected to hot-stamping in an internally cooled die-set. The cooling rate in the die designed to be sufficient to allow transformation to martensite giving strengths of around 1500 MPa [53, 54]. These are used as reinforcements for the roof, door, windscreen upright and b-pillar, etc.

The various HSLA and AHSS steels can be made to meet mechanical property requirements by use of different chemical compositions depending upon the experience of the steel plant used for production a nd the various process variables during casting and subsequent thermo-mechanical treatments. Such differences give microstructural and texture variations that can lead to inconsistent formability, weldability, corrosion resistance and surface finish. In particular batch to batch inconsistency can present significant problems in formability [55, 56].

IV. Competition from light alloys and reinforced polymers.

As mentioned in the introduction continually increased competition for steel from aluminium alloys

was one of the main reasons for ULSAB and the subsequent R&D programmes such as FSV. Aluminium sheet alloys have been in use for many years in motor cars [1, 6, 57-60] with examples ranging from the very early Rolls Royce models with an Al body on a timber (ash) frame to the 1954 Panhard Z1 in the USA and then later to a number of Rover cars, the Land Rover and Range Rover series in the UK [1]. In 1989 the first all-Al monocoque body was used for the Honda Acura NSX in which a 40% saving in bodyweight was made by the replacement of steel by aluminium alloys when using conventional design and assembly, effective use being made of extruded sections for structural subframe and sills [6]. This was followed by the Jaguar Sport XJ220 in 1992, and subsequently by the Audi A8 , the Ford AIVs, e.g. Taurus P2000, through to current Mercedes-Benz S class and Jaguar XK models [57]. These and other developments have resulted from the efforts of the Aluminium industry to advance aluminium vehicle technology (AVT) in order to make full advantage of the relative low density of their material.

The main alloys used have been from the 5XXX (Al-Mg) series and from the heat treatable 6XXX (Al-Mg-Si) series. Relative to steel, Al alloys have lower elastic modulus, lower ductility with smaller values for R_m and greater springback. To achieve satisfactory formability and stiffness sheet thicknesses have to be increased so that the potential 65% weight saving from density differences cannot be fully achieved. However, Al alloys can be readily produced in forms other than sheet, notably in extruded profiles with complex cross-section designs and thin-walled die-castings of intricate shape.

The Audi A8 in 1994 used a complete Al body frame made of Al extruded sections joined by Al base cast nodes with bonded Al body panels. Superplastic forming of Al is another process that has become used for body parts, e.g. In the late 1990's Morgan sports cars began to be fitted with one piece wings from superplastically formed aluminium (SPF). The classic shape of the sports car was maintained but the previous 3 piece steel wings were replaced by the SPF wings with improved corrosion resistance & reduced paint protection costs as well as weight saving [61]. Many companies are now using SPF Al parts for bonnet & boot inner frames, floor pans, and quarter panels, etc. The SPF offers greater design freedom, rapid prototyping and reduced tooling costs and strengthens the competition from aluminium.

The extent of the competition is shown by the use of Al for the body of the 2012 series 4 Range Rover which is made up of, in mass %: 37% 5xxx Al alloy sheet, 37% 6xxx Al alloy sheet, 15%Al base cast parts, 6%Al extrusion and only 4% HSS + 1%PHS. The increased use of Al gave a 39% saving in weight over the previous model [57]. The use in the Range Rover shows that Al has higher potential for use in larger vehicles such as pick-ups and SUVs [60] than

in smaller cars, except for electric vehicles. Al alloy has long been used for bodies in large commercial vehicles such as trucks and buses, e.g. from 1954 in the London Transport "Routemaster" double-deck bus.

Lighter than Al, Magnesium base alloy could also be considered as autobody sheet materials. However, in a critical assessment, it has been suggested their use is unlikely without research to improve mechanical properties, formability, joining methods and corrosion resistance and to reduce cost [62]. Mg base die-cast parts can and are being used for certain body parts in preference to welded steel fabrications (pressings + tubes) e.g. for cock-pit cross beams in Range Rover and Jaguar models [63]. The first Mg base chassis component was vacuum die cast in Mg-4Al-4Ce alloy and was used for the engine cradle in the 2006 Chevrolet Corvette. [58, 64]. Mg alloy diecastings have also been used for inner panels for doors and liftgates and for roof frames for convertibles [65]. The use of die cast thin section structural parts is more promising than use of wrought alloys, the latter being said to require further developments [62, 66]. Finite element based study has shown that a Mg auto-body structure giving equivalent stiffness to that in steel or Al could be respectively 60% and 20% lighter [67].

Fibre reinforced polymers with glass (GFRP) or carbon fibres (CFRP) must also be considered since they can be rapidly processed into complex shapes and by appropriate design can give equivalent impact resistance to steels. They have been used mainly for specialist sports cars and cab, panels and roof parts for commercial vehicles. Mixed Al alloy/CFRP designs have been used for chassis construction [57]. CFRP can offer attractive combinations of strength and weight but because it is difficult to use in mass production it is limited to very low-volume specialist sports cars. Even for low volumes, at 30 cars/day, Ferrari is using Al rather than CFRP [68].

In achieving an optimized balance between weight saving, performance (safety, longevity, etc.) and cost multi-material bodies need to be built containing combinations from AHSS, cast, extruded or sheet form Al alloys, die-cast Mg alloys and fiber reinforced polymers. Hence hybrid-joining technology is just as important as the base material. Materials and process selection has to consider spot and laser welding, riveting, clinching, high speed nailing, friction stir welding, adhesive bonding , etc. with regard to joint integrity and the ease of robot operation [69-71]. For example, the problem of joining Al to B containing PHS is being tackled by combining punctual joining methods such as self-pierce riveting with adhesive bonding to improve bond shear strength [71].

To minimize life cycle environmental impacts in materials selection, attention must also be paid to sustainability including emissions from materials production and the many problems associated with recycling vehicles at the end of their life. To this end

an increasing number of life cycle analysis studies are being reported [e.g. 72-75]. The results of such complex studies not only depend on the models/ parameters/data accuracy, etc. used in the study but also on the specific circumstances in production of both raw materials and the vehicles, use of the vehicles and end of life recycling/re-use. Nevertheless it is recently claimed that the use of AHSS produces lower greenhouse gas emissions than light weighting with Al. [76].

V. Thai Situation

As an international manufacturing base for automobile industry, the domestic body parts supplier in Thailand, with highly dependent on the overseas major makers, have to employ any raw materials that complied with the specification of the makers. After nearly five decades of development, there are at least three to four medium to large domestic companies that can supply body parts with almost all steel specification. There is a company that starts to supply aluminium body parts to an electric vehicle manufacturer in USA last year. However, some Japanese car makers developed their in-house press parts for special steel sheet forming. The steel specification includes JSH400W, JAC 270D and E, JACUN440H, JAC340H, and JAC440P. The thickness ranges from 0.70 mm for hood outer, 1.00 mm for Plat Form (CREW), to 8 mm for Cross Member: Trunion, etc. These steel materials come in as steel sheet to be press forming in the factory.

One of the major hurdles for the development of Thai industry that depend on the high quality iron and steel materials is the lack of integrated steel manufacturer. The strong political will and understanding of policy makers of industry development will be the key for economic expansion of Thailand in the years to come.

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