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Design of cold rolled and continuous annealed carbide-free bainitic steels for automotive application

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ABSTRACT

Advanced high strength steels for automotive applications were designed to achieve a carbide-free bainitic microstructure after conventional thermo-mechanical processing and a continuous annealing treatment. The microstructure obtained consists of ferrite laths interwoven with thin films of untransformed retained austenite. The sufficiently tough matrix and the control of the heterogeneity in the microstructure will allow an optimum combination of strength, ductility, and formability to be achieved. The designed steels reached far higher uniform elongations than that in commercial dual phase steels and martensitic steels with the same range of ultimate tensile strengths. Their formability was found to be appropriate for the production of final parts after cold-stamping or cold-forming. On the other hand, the yield strength/ultimate tensile strengths ratio was found to remain roughly constant (~0.7). The reduction of area value did not seem to change as a function of overaging temperature, but the V-bending angle and the hole expansion ratio (cut-edge stretching ability) decreased significantly at the bainite holding temperature increases.

1. Introduction

Potential future car technologies include new energy sources and new materials, which are being developed in order to make vehicles more sustainable, safer, more energy efficient, and less polluting. With rising oil prices, the future of cars is leaning towards increased fuel efficiency, energy savers, hybrid vehicles, battery electric vehicles, and fuel cell vehicles. Auto-makers are focused on weight reduction as a key enabling technology for lowering fuel consumption in combustion engine vehicles and for extending the range—and the size—of alternative vehicles. It is known that the amount of electricity needed to power an electric vehicle is mostly driven by vehicle weight. The heavier the car, the bigger the battery required. Larger capacity batteries cost more and weigh more, take longer to charge, require a greater level of manufacturing capacity to produce, and increase the demand for new and expensive high voltage charging infrastructure.

Large reductions in weight are often thought of as requiring radical changes, such as all aluminum bodies, or carbon–fiber composites that are sometimes featured in concept vehicles and high end production models. Although these materials are used selectively today and are likely to see broader application in the future, it is important to be aware of the substantial, near-term opportunities from less costly available technologies utilizing engineering plastics and lightweight metals, as well as steel and iron.

The Future Steel Vehicle (FSV) Program has recently featured steel body structure designs that reduce mass by more than 35% over a benchmark vehicle and reduce total life cycle emissions by nearly 70%. The FSV program has brought more advanced steels and steel technologies to its portfolio, including more than 20 new Advanced High Strength Steels (AHSSs) grades, representing materials expected to be commercially available in the 2015–2020 technology horizon. The FSV material portfolio includes the *first generation* of AHSS, i.e., steels that possess primarily ferrite-based microstructures such as dual phase (DP), transformation-induced plasticity (TRIP) [1], complex-phase (CP), and martensitic (MART) steels; and the *second generation* of AHSS, i.e., austenitic steels with high manganese contents, which include steels that are closely related to austenitic stainless steels, such as twinning-induced plasticity (TWIP) steels [2]. All these type of steels reach into GPa-strength levels and are the newest in steel technology offered by the global industry.

Recently, there has been increased interest in the development of the *third generation* of AHSS, i.e., steels with strength-ductility combinations significantly better than exhibited by the first generation AHSS, but at a cost significantly less than required for second

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generation AHSS. The third generation of AHSS will include materials with microstructures consisting of a high strength phase (e.g., ultra-fine grained ferrite, martensite, or bainite) and a significant amount of ductile austenite that improves the work-hardening of the complex composite structure by TRIP effect.

Approaches to the development of the third generation of AHSS have been recently performed [3–7]. A new process referred to as *quenching and partitioning* (Q&P), has recently been shown to be a unique processing route for the production of high strength martensitic steels with significant amounts of retained austenite [3]. The materials are rapidly cooled to a specific quench temperature between M_S and M_F to create controlled fractions of martensite and austenite. This step is followed by a thermal treatment at the partitioning temperature, at which carbon migrates from martensite to austenite to increase the austenite stability resulting in higher austenite fractions at room temperature after cooling. In the Q&P process, formation of iron carbides is intentionally suppressed, and the austenite is stabilized rather than decomposed. The sensitivity of the Q&P process to alloy content and processing temperatures has been explored, and it has been shown that the Q&P process works as a viable route to produce third generation AHSS in a variety of alloy systems. Of particular interest is the strength/ductility performance of a 0.2C–1.6Mn–1.6Si (wt.%) alloy processed with Q&P heat treatment [4].

Alternatively, design methodologies based on diffusionless bainite transformation theory have been applied to develop steels with a carbide-free bainitic (CFB) microstructure consisting of a mixture of bainitic ferrite, retained austenite, and some martensite [5–8]. Using thermodynamics and kinetics models, CFB steels with a 0.2 and 0.3 wt.% carbon content were designed and manufactured following a conventional hot rolling practice. The designed steels present significant combinations of strength and ductility, with tensile strengths ranging from 1300 to 1800 MPa and total elongations of over 14%.

In terms of in-use properties, first generation of AHSS, such as low alloy TRIP steels, are excellent in stretch formability due to their large uniform elongation, but they are generally inferior in stretch flangeability, as in the expansion of a pierced hole. This is related to the presence of hard and soft phases in their microstructure [9]. Although stretch flangeability is also known to decrease with increasing strength, the effect of micro-scale uniformity is noteworthy – the more uniform the microstructure, the better the stretch flangeability. Bendability can also be construed to be similar to stretch flangeability, as a fracture due to a large local deformation [10], and can be adjusted, for example, by the scatter of hardness (i.e., the distribution of hard and soft phases in the microstructure) for 980 MPa-class steel. Accordingly, the improvement of bendability can be achieved by the same concept as in the improvement of stretch flangeability.

These weak points corresponding to the formability of first generation of AHSS may be overcome by replacing the polygonal ferrite matrix in TRIP steels by a lath-like ferrite matrix [11]. The CFB steel is expected to achieve extremely good stretch-flangeability due to its uniform fine lath structure. On the other hand, the heterogeneities of hardness due to the presence of martensite in this advanced bainitic microstructure will allow this steel to reach a good, deep drawability [12]. In this sense, the main objective of this work is to design cold rolled and continuous annealed AHSS with a CFB microstructure for manufacturing of uncoated (i.e. bare) or electrogalvanized (EG) products for the automotive industry, overcoming property limitations of AHSS commercially available for cold-stamping or cold-forming products. In more detail, the engineering requirements are for an ultimate tensile strength (UTS) higher than 1000 MPa, a yield stress/ultimate tensile stress (YS/UTS) ratio of 0.7, a combination of strength and total elongation (UTSxTEl) higher than 15,000 MPa%, and a hole expansion (HE) ratio higher than 30%. The sufficiently tough matrix and the control of the heterogeneity in the microstructure will enable an optimal ductility that is comparable to drawing steels while keeping a suitable bending performance.

2. Theoretical calculations and alloy design

An alloy design procedure [5] based on phase transformation theory alone was successfully applied to design steels with a CFB microstructure. An increase in the amount of bainitic ferrite is needed in order to avoid the presence of large regions of untransformed austenite, which under stress, decompose to brittle martensite.

2.1. Outline of phase transformation models in steels

The bainite transformation progresses by the diffusionless growth of tiny platelets known as ‘sub-units’ [13]. The excess carbon in these platelets partitions into the residual austenite soon after the growth event. Diffusionless growth of this kind can only occur if the carbon concentration of the residual austenite is below that given by the T_o curve. The T_o curve is the locus of all points, on a temperature versus carbon concentration plot, where austenite and ferrite of the same chemical composition have the same free energy [14]. It follows that the maximum amount of bainite that can be obtained at any temperature is limited by the fact that the carbon content of the residual austenite must not exceed the T_o curve of the phase diagram. The design procedure avoids this difficulty in two ways: by adjusting the T_o curve to greater carbon concentrations with the use of substitutional solutes such as Mn and Cr and by controlling the mean carbon concentration [15,16].

Bainite is formed below the T_o temperature when $\Delta G^{\gamma\rightarrow\alpha} < -G_{SB}$ and $\Delta G_m < \Delta G_n$, where $G_{SB} \cong 400\ \mathrm{J\ mol^{-1}}$ is the stored energy of bainite [14] and $\Delta G^{\gamma\rightarrow\alpha}$ is the free energy change accompanying the transformation of austenite without any change in chemical composition. The first condition therefore describes the limit to bainite growth. The second condition refers to nucleation; thus, ΔG_m is the maximum molar Gibbs free energy change accompanying the nucleation of bainite. G_n is a universal nucleation function based on a dislocation mechanism of the kind associated with martensite [17]. The temperature dependence of G_n is independent of chemical composition; together with the growth condition, the function allows the calculation of the bainite start temperature, B_S, from a knowledge of thermodynamics alone.

Apart from controlling the T_o curve and B_S temperature, substitutional solutes also affect hardenability, which is an important design parameter to avoid transformations such as proeutectoid ferrite and pearlite. For this purpose, thermodynamic and kinetics models developed to allow the estimation of isothermal and continuous transformation diagrams, from the knowledge of the chemical composition of the steel concerned, were used in the alloy design process [18–22]. There are other output parameters, such as the martensite and Widmanstätten start temperatures. An extensive description of all the models used on the design procedure was reported elsewhere [5].

2.2. Earlier developments of CFB steels

Edmonds and co-workers [15,16,23–25] showed that CFB is, in principle, an ideal microstructure from many points of view. An example of the microstructure of CFB is presented in Fig. 1; instead of the classical structure of bainitic ferrite laths with interlath carbide, it consists of bainitic ferrite laths interwoven with thin films of untransformed retained austenite. In particular, the steel has a high resistance to cleavage fracture and void formation due to