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# Dynamic Recrystallization Behavior of Ni-Containing Austenite-Based Fe-Mn-Al-C steels

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## 1. INTRODUCTION

Formerly relevant only for being a replacement for conventional stainless steel, Fe-Mn-Al-C steels have stood out as materials capable of combining tensile strength, ductility, and low density to outstanding extents. In the pursuing of new avenues of energy-efficiency optimization, lightweight Fe-Mn-Al-C steels have emerged as a prominent material for transportation and military industry given its lower density in comparison to general carbon steels. The density reduction is chiefly a function of aluminum (Al) content, which typically ranges from 5 to 12 wt.%; it results from both the lattice expansion caused by the large atomic radius of the Al in solid solution and the lower atomic mass of Al in comparison with iron [1,2].

Depending on the chemical composition, Fe-Mn-Al-C steels can assume a fully austenitic microstructure; in that case, their mechanical behavior relies on their deformation mechanism, which, in turn, is determined by the stacking fault energy (SFE). Stacking faults in a face-centered cubic (FCC) crystal structure can be understood as an interruption in the order of the stacking sequence of {111} closed-packed planes; these faults can be introduced during solidification and deformation and up to two layers of interruption can occur [3]. In a FCC crystal structure, a perfect dislocation can break up into two partial dislocations; a repulsive force between them ensues, leaving a stacking fault that extends over the distance between the partial dislocations, namely, dissociation width [4]. This width dictates the mobility of the dislocations and the mechanical behavior accordingly.

High-SFE steels have narrower dissociation width, which facilitates the cross-slip and climb of dislocations; in that case, dislocation glide of perfect and partial dislocations can accumulate along slip bands – microband-induced plasticity deformation mechanism. [3-5]. Characteristic of low-SFE steels with wider dissociation width, transformation-induced plasticity (TRIP) leads to the formation of martensite since the stacking fault is characterized by local hexagonal-closed packed sequences. The SFE is a composition-dependent parameter, whereby the deformation mechanism can be adjusted by controlling the chemical composition. In this regard, Al is well known for increasing the SFE; thus, Al can stabilize the austenite by increasing the resistance to TRIP. In fact, as demonstrated by Chen et al. [6], even additions of 1 wt.% of Al could inhibit the TRIP deformation mechanism.

A strengthening mechanism in austenite-based Fe-Mn-Al-C steels occurs through the coherent, intragranular precipitation of κ-carbide, a nano-scale precipitate having a perovskite crystal structure – $E2_1$ in the Strukturbericht notation. The interaction between these carbides and dislocations occurs through Orowan bypassing mechanism, promoting the refinement of the slip band spacing, thereby enhancing the strain hardening rate; this, in turn, effects high strength and strain hardening capability [7]. Intragranular, non-nucleated κ-carbides can form during the cooling, even at fast rates (quenching), of austenite from solution treatments at above 900°C as a result of the spinodal decomposition and subsequent ordering reaction of austenite [8].

The spinodal decomposition divides the high-temperature austenite (γ) into two low-temperature γ phases: the C- and/or Al-depleted γ' phase and the C- and/or Al-rich γ'' phase. Cooling to lower temperatures promotes a short-range ordering reaction in the γ'' phase, which transforms into the $L1_2$-ordered meta-stable phase, Al and Fe/Mn atoms located at the corner and face center sites, respectively. Thereafter, upon further ordering of C atoms, κ-carbide, the low-temperature, stable phase with $E2_1$ structure, forms [8,9]. At very low temperature, $L1_2$-ordered phase nuclei form in the austenite accompanying the ordering reaction. During quenching, however, particularly at low temperatures, there isn't much time for the diffusion of solute atoms and the growth of $L1_2$-ordered phase nuclei [11], explaining the suppression of κ-carbide through fast cooling rates.