Ultra Fine Ferrite Formation through Dynamic Strain Induced Transformation - A Review

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Abstract :- In this paper we study grain refinement in steels at ultrafine level i.e. ultrafine ferrite formation by dynamic strain induced transformation(DSIT). DSIT is essentially a process of spontaneous dislocation induced discontinuous decomposition of metastable austenite under deformation to nano sized carbides and ultrafine ferrite. Post transformation deformation of ferrite envisages a continuous dynamic recrystallisation accomplished through continuous increase in misorientation across the boundaries due to defect accommodation; in the process a ferrite grain ,after completion of its transformation growth on account of inevitable impingement with similarly growing grains and being continuously deformed constitutes elongated micro volumes separated by high angle misorientation(ferrite submicrob and).Continued deformation entails the attainment of boundary migration enthalpy minimum which prohibits further reduction in the width of ferrite submicrobands ;crosswise emanation of new subboundaries result in the division of the original ferrite grains into very fine near polygonal micro volumes of high angle boundaries.

Keywords: **- DSIT, TMCP, DIFT, ECAP, HPT, ARB, ESBD, subboundaries, submicroband, low angle boundaries, high angle boundaries.**

I. INTRODUCTION TO DSIT (DYNAMIC STRAIN INDUCED TRANSFORMATION)

DSIT is essentially a process of spontaneous dislocation induced discontinuous decomposition of metastable austenite under deformation to nano sized carbides and ultrafine ferrite. In DSIT austenite to ferrite transformation is given to take place during the course of deformation. The idea of DSIT centres on the creation of high density of intragranular nucleation sites within highly strained metastable austenite so as to produce ultrafine ferrite grains in the final microstructure. Hudgson et.al.[1] postulated three critical factors for the formation of UFF grains during SIT:

(i) A high shear strain,

(ii) A heavy under cooling,

(iii) An appropriate deformation temperature.

II. DIFFERENCE WITH TMCP (THERMO MECHANICAL CONVENTIONAL PROCESS)[2]

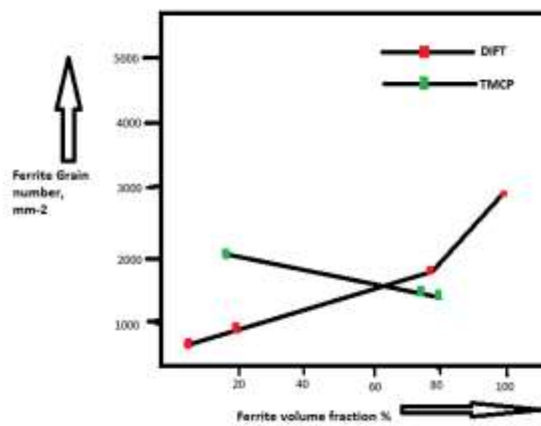

Fig1. Comparison of ferrite grain number between DIFT and TMCP

Fig.1 compares the ferrite grain number per unit area as a function of ferrite volume fraction of DIFT and TMCP. It is clear that the grain number increases during DIFT, but it decreases during TMCP. Therefore, DIFT is a nucleation controlling process, i.e., the transformation is mainly accomplished through the continuous nucleation process, whereas TMCP is a grain growth and coarsening dominant process. The growth of ferrite grain is inhibited to a great extent during DIFT due to the rapid and repeated nucleation of ferrite grain at γ/α interface, which results in a final finer grain size than that obtained by TMCP.

III. PROCESS DESCRIPTION

Various processes have been carried out on various alloys of low carbon steel to attain ultrafine ferrite grains by heating samples to austenitisation temperature and deform them by various methods at various deformation temperatures followed by predefined cooling rates and media. Some of the processes are as follows:

(a) Warm deformation tests were performed on a torsion rig and samples of C-Mn-Si with other alloying elements like S,P and Al were heated to austenitisation temperature in inert atmosphere. Then samples were deformed at 875^0 C and 810^0C.

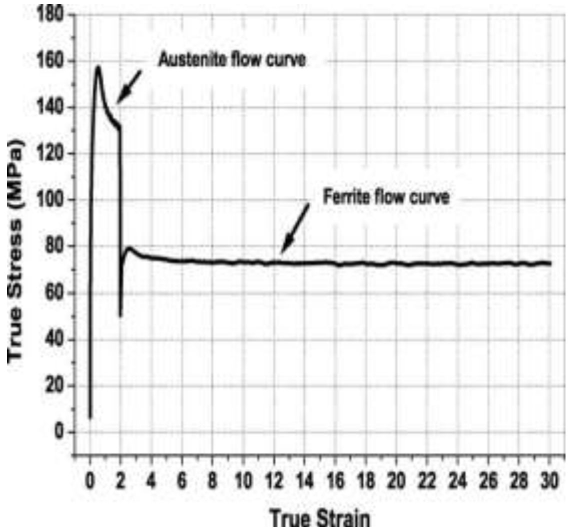

Fig2. A typical true stress–true strain flow curve of austenite and large strain warm deformed ferrite at 876 and 810 ◦C, respectively, with strain rate of 0.1 s−1

(b) A five pass rolling was performed on a sample of micro alloyed Nb-V-N low carbon steel after heating above austenitising temperature upto1200^{0}C and soaking for 30min. Various deformation temperature were selected for every pass but the last three passes were at same temperature having varying time gap before deformation and called DIFT rolling.

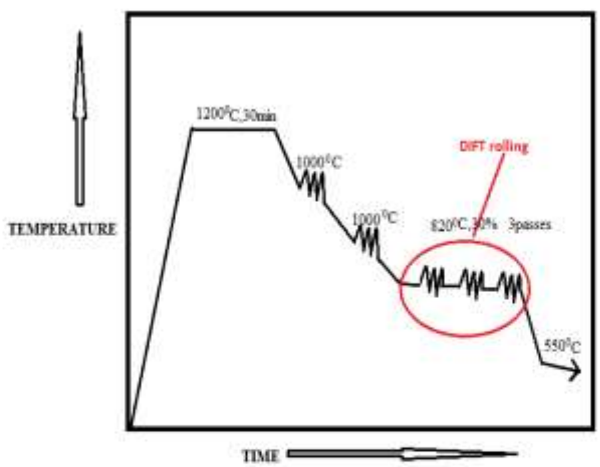

Fig3. Deformation Schedule for the laboratory rolling

(c) Hot compression tests were performed on a sample of C-Mn-Si with other alloying elements, using a Gleeble 1500 hot simulation testing machine to attain ultrafine structure.

(c) Samples of C-Mn and C-Mn-Nb-Ti were warm torsioned in the intercritical range (between Ac_1 and Ac_3) cooled to room temperature and subsequently annealed at 800^{0}C to attain ultra fine grain structure.

(d) A C-Mn-V steel was used to study ultra fine ferrite formation (1-3µm) through dynamic strain induced transformation using hot torsion experiment

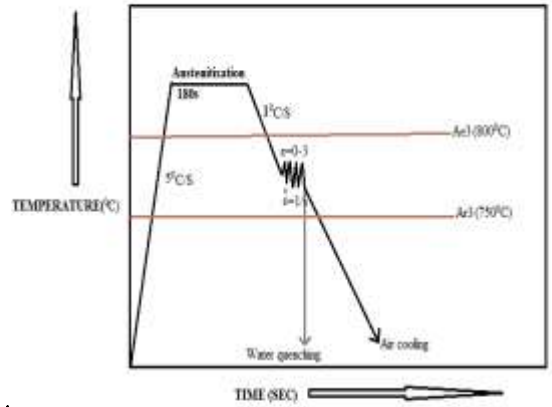

Fig4. Schematic representation of the thermomechanical processing schedules.

IV. CHARACTERIZATION TECHNIQUE

Characterization technique involve microstructural analysis which is carried out by electron back scattered diffraction (EBSD) and shows misorientation pattern of grain boundaries, also low angle boundaries (LAB's) and high angle boundaries (HAB's) are shown by the technique in various methods to attain UFF by dynamic strain induced transformation.

In EBSD analysis specimens are prepared in a standard manner. The EBSD measurements were carried out using a Leo field emission gun scanning electron microscope (FEG/SEM) equipped with a HKL technology. The FEG/SEM is operated at an accelerating voltage of 20kv. Area mapping for 210µmX150µm is performed with a step size of 0.3µm on a square grid so that total number was 350,000.

V. IMPORTANT RESULTS

(a) Continuous dynamic recrystallisation(CDR_X) transformed the microstructure into stable fine equiaxed ferrite structure, in the early stage of deformation, having HAB's suitable for the grain boundary sliding during high ductile deformation upto strain of 30.

(b) Ultra fine grained V-Nb-N microalloyed steel with a grain size of 1.5µm was produced by DIFT rolling.

(c) By warm torsion test on C-Mn and C-Mn-Nb-Ti samples the martensite decompose discontinuously as the precipitation reaction occurs preferentially, starting at the prior austenite grain boundaries and occasionally on the edges of the martensite laths. The kinetics of discontinuous type precipitation reaction is controlled initially by carbon diffusion, then at longer tempering time by the diffusion of carbide forming elements(Nb,Ti).

(d) The principle reason for refining the ferrite structure is the nucleation rate of the new phase, which is a result of increasing dislocation density in austenite just before the start of phase transformation. The large strain produces more deformation defects such as deformation bands. These defects are beneficial in increasing the preferential nucleation sites of ferrite. The nucleation probability largely increase with increasing of stain and this leads to the very fine strain induced ferrite grains.

(e) Deformation at 845^0C with the strain of 0.8and strain rate of $0.1s^{-1}$ produces the ultrafine, equiaxed and homogeneous ferrite grains of about 2μm with little amount of proeutectoid deformed ferrite.

(f) During hot torsion test of C-Mn-V steel the minimum strain required to produce UFF was 2. This strain is called critical strain for UFF formation ($\varepsilon_{c,UFF}$). The result suggests that the coarsening was significantly reduced with an increase in cooling rate upto 5^0C/S. The results suggests that the cooling rate higher than 9^0C/S is required to significantly decrease the coarsening rate of ferrite, although it will encourage the formation of non equilibrium phase.

(g) Results shows that there are two critical strains. When straining in Ae_3 -Ar_3 region, the first is for the start of DSIT(i.e. $\varepsilon_{c,DSIT}$) while the second corresponding to ultra fine ferrite formation(i.e. $\varepsilon_{c,\ UFF}$) in the final microstructure.

(h) For coarse grained austenite, the ferrite grain size was reasonably homogeneous with more than 90% ferrite grains less than 2μm in diameter at quench temperature of 775^0C.

For fine grained austenite , the population of ferrite grain sizes less than 2μm significantly decreased from 95% at the quench temperature of 775^0C, to 40% at room temperature.

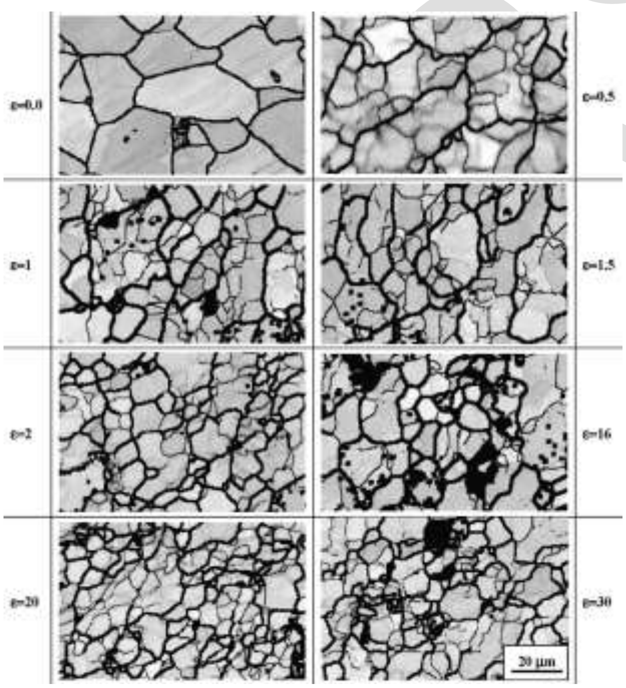

Fig5. The EBSD maps showing LABs (3◦ < θ <15◦, thin lines) and HABs (15◦ < θ, thick lines) in plain low carbon steel samples deformed with different strain at810 ◦C and ε◦ = 0.1 s−1 (the shear direction is vertical and the radial direction is normal to the page).

For the fine grained austenite, the DSIT ferrite grain distribution was more homogeneous as compared with coarse austenite.

VI. GAP IN UNDERSTANDING

Different researchers have reported fascinating structural changes during the course of DSIT, unfortunately in majority of the cases the description of the different aspects of this transformation process has been dealt in individual static frames. It is probably why many a places one may observe lack in harmony with the reported observations when DSIT is viewed in totality.

Report on quantitative assessment of the impact of such deformation on the magnitude of energy of nucleation and the critical nucleus size of the transformation product is not extensively documented.

It is known that DSIT involves heterogeneous nucleation of ferrite at the defect sites. Too much focus on the manner of enhancement of nucleation sites and evolution of means to avoid wiping out of some of them due to transformation growth seems to have disregarded the fact that simple statistical rule would preclude any expectation that all the defect sites created in austenite are of sufficient effectiveness.

Viewed on the aspects of transformation mechanisms one is to notice that dynamics of DSIT is less attended. Beladi et.al.[3] demonstrated serrated grain boundaries during DSIT but did not consider ledge mechanism as the operative one for ferrite transformation at highly perturbed austenite grain boundary.

Reasons were ascribed to faster nucleation of ferrite in DSIT but strain rate effect seems to have missed out serious attention. The boundary mobility of either ferrite-ferrite or ferrite-austenite has not been considered from first principals and integration with the possible causes of evolution of structure actually observed in DSIT has not been appreciably reported till date.

VII. TRENDS IN DSIT

In recent years several studies on grain refinement of ferrite have been conducted by different methods like[4]:

(a) Equal Channel Angular Pressing (ECAP)

(b) High Pressure Torsion (HPT)

(c) Accumulative Roll Bonding (ARB)

in order to optimize the relationship between mechanical properties and microstructure of steels. ECAP has demonstrated the capability to produce a great refinement of the steel microstructure. Alternative methods such as ARB and martensitic steel rolling were presented later by Saito et.al.[5] and Ueji et.al.[6] ,respectively. Some of these processes, however, are disadvantageous either due to their complexity or to their inapplicability to large scale/volume.

Brittle to Ductile transition (BDT) behaviour was investigated in low carbon steels deformed by an accumulative roll bonding process[7].

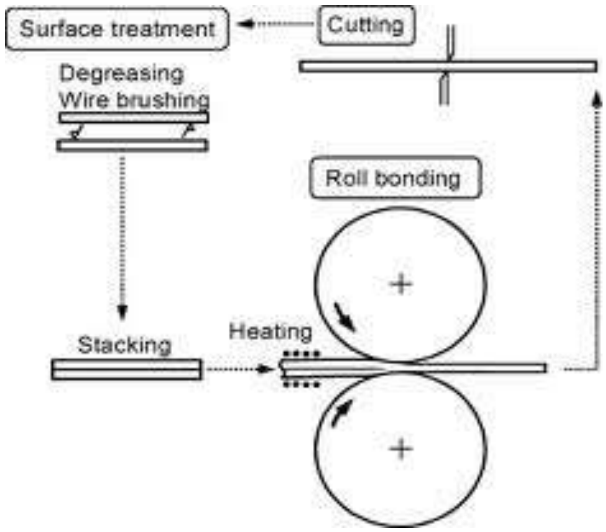

Fig 6. Accumulative roll bonding process

Both macroscopic and microscopic microstructural changes in a plain low carbon steel with a mixed structure of ferrite and pearlite were examined during equal channel angular pressing(ECAP)[8] .

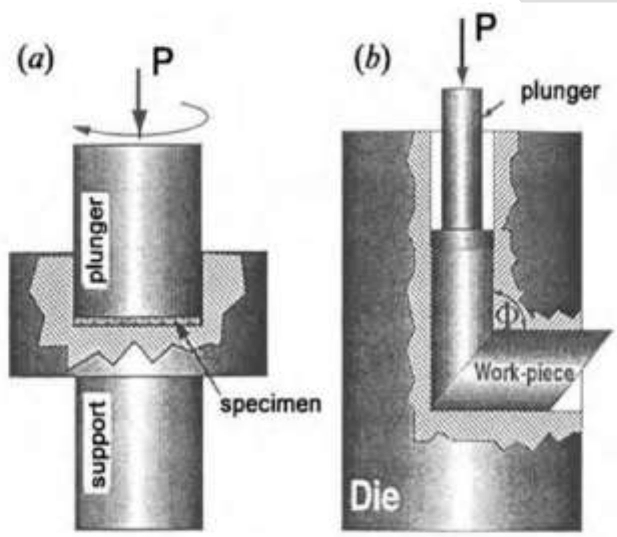

Fig7. Equal channel angular pressing process

The structural and phase transformations which take place in low carbon steel during deformation by high pressure torsion (HPT) and subsequent heating have been studied using transmission electron microscopy and X-ray structural analysis methods. Whatever the initial state ,be it ferrite-pearlite or martensite (obtained by quenching from 950^{0}C and 1180^{0}C), HPT at room temperature leads to the formation of a nanosized oriented grain subgrain structure[9] .

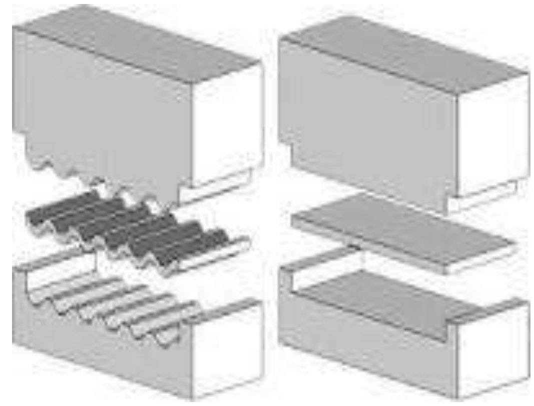

Fig 8. High pressure torsion process

VIII. CONFLICITY IDEA

In spite of doubtless capability of dynamic transformation to produce ferrite grains of around one micron size as demonstrated by majority of researchers in this area, there remains several unknowns and apparent ambiguities in respect of phenomenological aspects of phase transformation involved in DSIT. There is no general theory which can satisfactorily explain the entirety of DSIT phenomenon.

The transient character of phase transformation in steel which is a function of instantaneous composition of austenite, its transformation temperature, residence time at the transformation temperature and the substructural conditions has not paid sufficient attention.

Let us hypothetically refine the grain to below the value say 1 micron up to 0.2 micron. It is immediately apparent that the driving force for grain growth becomes 15MPa, a value higher than for refinement through dislocation activity (i.e. 10MPa). Thus grain coarsening will set in to bring about a balance.

IX. PROPOSED MECHANISM OF DSIT[10]

The DSIT ferrite dynamically nucleate on the prior austenite grain boundaries at an early stage of deformation. While the strain for a final microstructure of ultra fine ferrite was $\varepsilon_{c,UFF}$, the actual start of strain induced transformation, $\varepsilon_{c,DSIT}$, was much lower in this case a strain of 0.5 for the coarse initial austenite. At higher strains Intragranular Ferrite (IG) grains started to nucleate in the prior austenite grain interiors and the prior austenite grain boundaries were completely decorated by DSIT ferrite grains. At intermediate strains the ferrite grains forms a layer with full impingement along the prior austenite grain boundary. The volume fraction of intragranular ferrite gradually increases with an increase in strain to $\varepsilon_{c,UFF}$. For the fine prior austenite grain size, the DSIT of ferrite occurred at a lower strain (i.e. $\varepsilon_{c,DSIT}$=0.1). Here, ferrite forms entirely at the prior austenite grain boundaries at an early stage of straining. When the austenite grain boundaries is decorated by ferrite, intragranular ferrite nucleation occurred on intragranular defects (deformation bands) at $\varepsilon_{c,UFF}$ (strain of 1.8). DSIT ferrite grains were seen as elongated rafts through the microstructure. For the final initial austenite microstructure the distribution of ferrite grains was much more homogeneous at the critical final volume fraction was reached. In addition, the coarsening occurs not only in the IG ferrite grains but also in grain boundary ferrite. However, the ferrite coarsening mostly occurs for the IG ferrite rather than grain boundary ferrite grains for the coarse austenite grain. The results suggests that the coarsening behaviour is mostly controlled through impingement of ferrite grains at an early stage of ultra fine ferrite transformation, rather than the pinning effect of carbide particles.

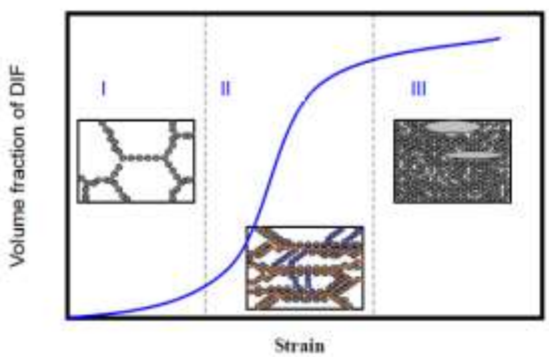

Fig9. Three stage microstructural evolution of DIFT

X. CONCLUSION [11-16]

DSIT is essentially a process of spontaneous dislocation induced discontinuous decomposition of metastable austenite under deformation to nano sized carbides and ultrafine ferrite. Post transformation deformation of ferrite envisages a continuous dynamic recrystallisation accomplished through continuous increase in misorientation across the boundaries due to defect accommodation; in the process a ferrite grain ,after completion of its transformation growth on account of inevitable impingement with similarly growing grains and being continuously deformed constitutes elongated micro volumes separated by high angle misorientation(ferrite submicroband).Continued deformation entails the attainment of boundary migration enthalpy minimum which prohibits further reduction in the width of ferrite submicrobands ;crosswise emanation of new subboundaries result in the division of the original ferrite grains into very fine near polygonal micro volumes of high angle boundaries. Driven by the need to assume lower energy configuration, self organisation process of those submicrovolumes sets in by mechanical rotation. Continued refinement leads to a situation of no boundary movement as characterised by least dispersion in topological classes among the grains of varying number of triple junctions. Beyond the steady state value which is till date found to be around one micron, any deformation induced grain refinement is opposed by the overriding influence of driving energy for grain growth. This steady state value is virtually a state of dynamic equilibrium between the deformation induced refinement and the concurrent dynamic grain growth. What remains to be answered is the reason for coincidence, if at all that is there, of grain locking by a specific topology and the attainment of dynamic equilibrium between recrystallisation and grain growth. Dynamic recrystallisation of prior austenite is not a necessary prerequisite of DSIT (though it is advantageous in respect of attaining uniformity and steady state situation relatively earlier) as would be evident from the fact that matter of post transformation grain refinement and attainment of dynamic equilibrium is exclusively intraferritic; it is for this reason it was noticed by Beladi et.al. that the finally attained ferrite grain size was independent of prior austenite grain size and coarse grained austenite gave rise to higher extent of refinement by default, as the final size of DSIT ferrite is fixed otherwise.

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