FRACTURE TOUGHNESS OF PLASTIC-MOLD STEELS: DEPENDENCE UPON HEAT TREATMENT AND MICROSTRUCTURE.

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ABSTRACT

The fracture toughness of low-alloy steels significantly depends upon the microstructure, i.e. upon metallographic constituents amounts and distribution and other parameters like carbides distribution, grain size, etc. Almost all the available studies and experimental data concern single constituent microstructures, whereas specific industrial applications use steels containing variable fractions of tempered martensite and bainite, as well as ferrite-pearlite, due to the production cycle and especially to the heat treatment used. In these cases it is particularly difficult to estimate the fracture toughness on the basis of the available experimental data; thus, it isn't possible in practice to evaluate (and minimize) the crack propagation risk.

Molds for plastic automotive components are a peculiar example, being produced from pre-hardened large blocks. The fracture toughness of the core mixed microstructures, that can occupy parts of the mold face, being sought, K_{Ic} tests have been performed as a function of the sample position in a 1.2738 steel block, and results interpreted at the light of fracture morphology and local microstructure.

1 FRACTURE TOUGHNESS REVIEW OF MIXED-MICROSTRUCTURE LOW-ALLOY STEELS

Fracture toughness (either K_{Ic} or J_{Ic}) of mixed microstructures as encountered in slack quenching and tempering of HSLA steels haven't received any attention. Appreciable reductions of notch strength and notch ductility of slack quenched and tempered AISI 1340, 2340, and 5140 steels in respect to the fully quenched and tempered condition were reported more than 50 years ago (Sachs et. al. [1]). Recently (Zhang et. al. [2]), K_{Ic} tests were performed at -80°C upon A533B steel samples having bainite and auto-tempered martensite (B/M) mixed microstructure, deriving from austenite with 120 and 8 µm grain size. Microstructures consisting of 30/70% B/M (g.s.120 µm) yielded 60 MPa√m, whereas the 45/55% B/M (g.s. 8 µm) ones yielded 55 MPa√m; simple upper bainite yielded 32 - 46 MPa√m and simple auto-tempered martensite 89 - 92 MPa√m (for 120 - 8 µm respectively). These data give a clue to the detrimental effect of bainite, but do not allow defect allowance calculations as regards very large blocks of low alloy quenched and tempered steels. The fracture toughness of mixed microstructures consisting of tempered martensite-bainite and pearlite-ferrite, that can occur at the core of the blocks, has not been previously investigated.

2 LARGE PLASTIC MOLDS FOR THE AUTOMOTIVE INDUSTRY: APPLICATIONS, STEEL GRADES, PRODUCTION CYCLE, CRITICAL PROPERTIES

Large forged steel blooms (typically 1x1 m section and more than 1 m length) are used to fabricate plastic molds, in turn employed to form automotive components, such as bumpers and dashboards, made of thermoplastics, usually polypropylene or reinforced ABS.

Mold design is usually based on previous experience; a complete stress analysis is seldom performed and no fracture mechanics defect allowance procedure is normal in the mold industry. Yet, some producers reported macroscopically brittle mold failures during service.

The stress pattern applied to the molds in service arises from the polymer's injection pressure and from the thermal gradients, and could be enhanced by notch effects and by defects of various origin (particularly weld bed depositions effected without proper heat treatment). Stresses may be significantly raised by abnormal operations, e.g. incomplete extraction of already molded objects.
Each mold is expected to produce a few millions of pieces in its life, corresponding to the production run of one car model, thus fatigue effects should also be considered.
Wear induced by the reinforced resins flow may be very severe and may be an additional cause for crack nucleation, with the flowing resin infiltrating cracks and acting as a wedge.
The most commonly used steel grade is 1.2738 (40CrMnNiMo8-6-4, UNI and DIN standards), an heat-treatable, 0.4% C, high-hardenability, low-alloy steel (Table 1).
For economic and logistic reasons, the traditional production cycle (rough machining, heat treating and finishing) has been abandoned and is commonly substituted by the following steps:

- in the steel mill: casting of steel ingot, forging in order to obtain 1x1 m (or larger) sections, dehydrogenization, oil quenching and tempering, eventually followed by stress relief;
- in the commercial warehouse: sawing to requested dimensions and removal of rough surfaces;
- in the mold-machining shop: chip-removal and/or electrical-discharge machining to the mold shape, grinding with or without polishing in selected areas, local surface treatments; upon request, shape corrections using weld depositions.

ESR refining of the steel is usually not possible due to dimensional limits of ESR plants.
The section of the bloom is usually comparable to the section of the original ingot (because larger ingots are not feasible), thus a sensible reduction ratio is usually obtained through repeated forging steps, each consisting of elongation followed by compression along the axis of symmetry of the ingot. Each step achieving a limited reduction ratio (1.5 is a possible value), total deformation is much less than that obtained in rolling and not comparable in the effects.
Depending on the size, dehydrogenization can last a few days, whereas austenitizing, tempering and stress relieving times can be 1 to 2 days each. Usually, blooms are austenitized in the 840-880°C range, quenched in oil (use of salt baths has also been reported [3]) and tempered in the 550-600°C range to a final 330-300 HB hardness. Heating stages are usually executed in air.
Since forging yields a rough shape with deep decarburations occurring during heating, external material removal may be up to 20 mm (plus scale). Furthermore, blooms may be sawn to requested size (often asymmetrically); blooms for bumper molds are usually sawn to yield a U shape.
In large oil-quenched blooms different microstructures occur at increasing depths; martensite decreases and bainite increases with distance; pearlite with some ferrite appears at prior austenite grain boundaries, further increasing towards the center. Tempering affects all the microstructures; any of them can be found at the mold face, where notch effects and welding depositions defects are often present. Thus, the relevant steel properties, particularly fracture toughness, fatigue and wear resistance, should be studied as a function of the microstructure and thus of the position in respect to the quenched surfaces. First results of such an undertaking are here presented.

3 EXPERIMENTAL

3.1 Specimen preparation

Samples were cut from a commercial bloom used for the production of a bumper mold. The thermal and mechanical history of the bloom (not precisely known) is believed to be consistent with the usual and previously described production cycle. The bloom was delivered to the mold shop showing machined surfaces; nevertheless, on the basis of the as-delivered dimensions, of the usual production cycle and of some metallographic results, it has been possible to estimate the relative position of the original forged and heat treated surfaces, that are defined *nominal surfaces*.

Table 1 - Composition limits of 1.2738 steel and composition of examined heat (wt. %).

	C	Cr	Mn	Ni	Mo	Si	S	P
1.2738 - 40CrMnNiMo8-6-4	0.35-0.45	1.8-2.1	1.3-1.6	0.9-1.2	0.15-0.25	0.2-0.4	<0.03	<0.03
Examined Bloom	0.36	2.0	1.5	1.1	0.25	0.25	0.001	0.008

The chemical analysis (as provided by the steel mill, Table 1) satisfies the 1.2738 standard.
The original bloom was a 2420(L) x 1140(T) x 1000(S) mm parallelepiped (nominal dimensions). The L direction defines the long ingot forging axis, whereas S and T directions are thought almost indifferent in respect to the casting and forging procedures and define only a conventional reference system. The *depth* of a point is defined as its distance from the nearest nominal surface.
Four oversize blanks were cut from the positions shown in Fig. 1. O and O1 blanks were kept in the conditions determined from the heat-treatment of the original bloom. NT and QT blanks, WxB=41x25 mm, were re-austenitized at 870°C; NT was then normalized, QT oil-quenched; both were thereafter tempered at 560°C for 4h.
A K_{Ic} specimen and a metallographic specimen were machined from each blank. Further metallographic specimens as well as a cylindrical 7 mm diameter tension specimen were machined from the broken halves of each K_{Ic} specimen. Metallographic surfaces were perpendicular to the T direction. The orientations and the depth of some relevant points are specified in Table 2.

3.2 Results

3.2.1 Optical metallography
In re-heat-treated blanks NT and QT, Fig. 2, the microstructure was tempered martensite (troostite), homogeneous over the whole 41x25 mm section.
The as-received O blank was examined at the ends and at mid-point in respect to its larger dimension (the latter region was 3 mm ca. from the K_{IC} fracture surface). The microstructure was tempered bainite and martensite; increasing the depth from 31 to 217 mm, more tempered bainite was found, with some prior austenite grain boundary constituents appearing at the larger distance.
The as-received O1 blank showed a large dependence of the microstructure upon the depth. The end closest to the nominal surface (149 mm depth, Fig. 3a) showed a microstructure similar to that of the O blank mid-point. In the middle (280 mm) and at the furthest end (393 mm) of the same O1 blank, prior austenite grain boundaries constituents increase in volume (Fig. 3b).

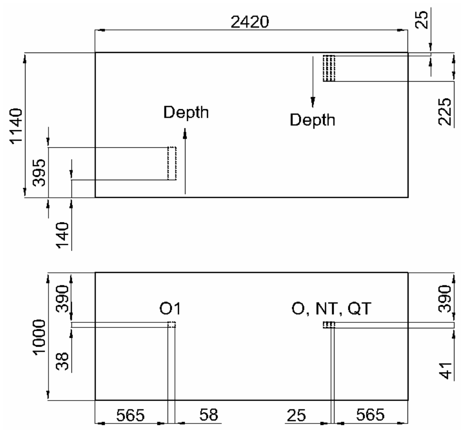


Fig. 1 - Positions of examined blanks in respect to the bloom's nominal surfaces.

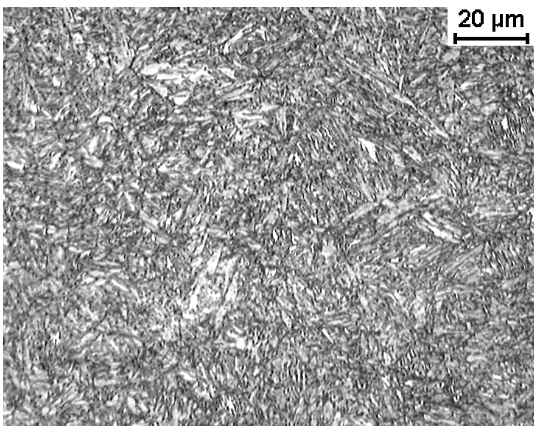
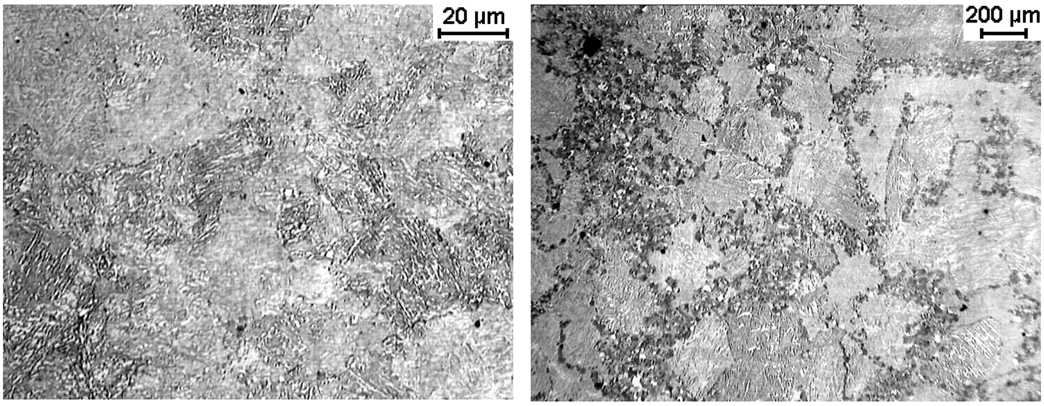


Fig. 2 – Microstructure of QT blank. Tempered martensite. 1% Nital etch.



(a) (b)

Fig. 3 – O1 blank microstructures: (a) 149 mm depth, tempered martensite and bainite; (b) 280 mm, tempered martensite and bainite with pearlite and some ferrite appearing at prior austenite grain boundaries. 1% Nital etch.

Detailed observations at 15 mm ca. depth intervals indicate that pearlite-ferrite appeared first in isolated spots at 197 mm depth and as a continuous constituent beginning from a depth of 212 mm. The K_{Ic} fracture surface was in this mixed-microstructure region.

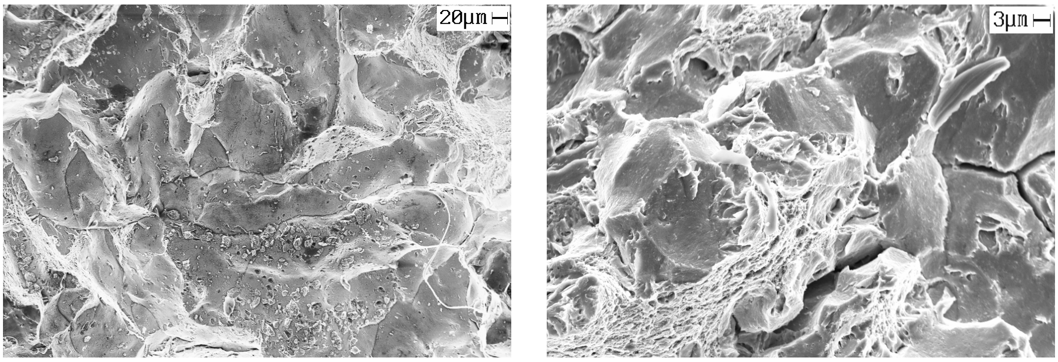
3.2.2 Mechanical testing

The results of all the mechanical tests (performed at room temperature) are presented in Table 2. Hardness resulted in the usual range for pre-hardened 1.2738 steel, but was moderately higher in the re-heat-treated samples, because they were tempered in the lower end of the usual range for a much shorter time and because their as-quenched microstructure was probably harder (fully martensitic instead of martensite and bainite). A remarkable difference was observed in tensile results between the as-received (average YS and UTS: 693 and 904 MPa; hardening exponent, n: 0.1 ca.) and the re-heat-treated specimens (970, 1111 MPa and 0.06, respectively).
Fracture toughness tests were performed using actuator displacement control. K_{Ic} tests (ASTM Std. E399) were performed upon O and O1 specimens using a 5 µm/s cross-rate, corresponding to a 0.28 MPa√m/s stress intensity rate and a 4.5 min total test time for the O1 specimen and to 0.44 MPa√m/s and 2 min for the O specimen. K_{Ic}/J_{Ic} tests were performed on QT and NT specimens using the single-specimen unloading-compliance technique (ASTM Std. E1820), with a rate in the 3-6 µm/s range, corresponding to 0.38-0.53 MPa√m/s.

Table 2 - Mechanical properties of the examined bloom.

Specimens	Hardness				Pre-cracked 3 point bend							Tension (T orientation)			
	HRC after quench		HB	HV	orien.	fracture surface depth	W	B	K_q	B,a min	J_q	depth of reduced section ends	YS	UTS	n
	-		-	-	-	mm	mm	mm	MPa√m	mm	kJ/m^2	mm	MPa	MPa	-
O	-		287	307	TS	117	41	25	50$^{(1)}$	13	-	154 - 189	696	888	0.10
O1	-		287	290	TL	272	58	38	76	30	-	319 - 354	690	920	0.11
NT	52	53	330	373	TS	-	41	25	123$^{(1,2)}$	40	83$^{(3)}$	-	966	1110	0.06
QT	54		330	377	TS	-	41	25	119$^{(2)}$	38	72$^{(3)}$	-	973	1112	0.06

(1) $K_{max}/K_q>1.1$ (2) $2,5\,(K_q/s_y)^2 > a$ or B (3) insufficient stable crack propagation



(a) (b)

Fig. 4 - Fracture surfaces of O (a) and QT (b) specimens; intergranular rupture with different facet size and facet morphology.

The only valid fracture toughness data pertains to the O1 specimen (K_{Ic} = 76 MPa√m), with the O specimen yielding not valid results (K_q = 50 MPa√m), owing to K_{max}/K_q being larger than 1.1, and the re-heat-treated specimens having insufficient dimensions (see notes on table 2). For the latter ones calculated critical J values (not perfectly valid) are consistent with respective K_q values. Thus, a K_{Ic} value of the order of 120 MPa√m ca. can be safely hypothesized for both NT and QT conditions. It can be concluded, even with some uncertainty, that the fracture toughness of specimens cut from the original bloom is much lower than that pertaining to specimens of the same compositions re-heat-treated to yield a fully tempered martensite microstructure.

3.2.3 SEM observations of fracture surfaces

NT and QT tension fracture surfaces were traveled by radial ridges with shear slip lateral faces and showed very limited external cone zones. O and O1 tension fracture surfaces had a well developed cup-and-cone morphology, with central regions characterized by ductile fracture with void sheet coalescence between primary voids.

The fracture propagation surfaces of the K_{Ic} specimens yielded the following results:

- *O1*: mixed, intergranular with quasi-cleavage areas suggesting fracture through pearlite colonies; typical intergranular facet about 100 µm wide.
- *O:* (Fig. 4a): prevalently intergranular with rounded edges and curved facets (typically 100 µm wide) showing some segregation effects.
- *NT and QT:* (Fig. 4b): prevalently intergranular (sharp edges and more planar facets, typically 10 µm wide, without any segregation) with limited ductile fracture areas.

In QT/NT K_{Ic} specimens a large blunting occurred in the fracture initiation region at the end of the fatigue pre-crack, with the formation of a well defined fan of slip lines (Firrao et. al. [4]) and multiple shear slip mode II fractures initiating at the blunted notch root. Blunting and shear slip fractures were much less evident in the O1 specimens, and very limited in the case of the O specimen.

4 DISCUSSION AND CONCLUSIONS

4.1 Metallography

Based upon the observations of the present bloom (not fully reported above) and of another bloom and upon comparing CCT diagrams and expected cooling rates, the microstructures occurring at increasing depths in a large plastic mold steel bloom are estimated as follows: tempered martensite (surfaces or edges); tempered martensite and bainite; tempered martensite/bainite with an increasing fraction of pearlite-ferrite at the previous austenite grain boundaries (inhomogeneous on a 0.1 mm length scale); tempered bainite and pearlite-ferrite; full pearlite-ferrite (core).

4.2 Fracture

The dominant mechanism of crack propagation appears to be brittle intergranular. One in-service failure occurred in a bumper mold confirmed this result. The difference between O/O1 and QT/NT specimens resides in the intergranular facets (those pertaining to O/O1 specimens being much larger and less straight) and can be related to different previous austenitic grain sizes: very long austenitizing times, experienced in the large block from where O/O1 specimens were cut, can be seen as the cause of austenitic grain growth. The intergranular rupture can be in both cases related to embrittling elements migrating to the prior austenite grain borders: both one-step and two-step embrittlement (if stress-relieving is applied after tempering) are conceivable (Briant [5]).
The difference in grain size on the fracture surface cannot be taken as a justification of the large K_{Ic} difference between the two groups of results (50/76 vs.119/123). Moreover, previous results on as quenched AISI 4340 steel by Firrao et al. [6] and by Ritchie et al. [7] show that intergranular rupture fracture toughness increases by increasing the austenitic grain size. Thus, the reported differences in fracture toughness levels among the various microstructures appear to be due to the energy consumption associated with the slip lines deformation and multiple fracture along them.
The difference in fracture toughness between the as-received O/O1 and the re-heat-treated QT/NT microstructures confirms the detrimental effect of uncompleted quenching; this fact and the difference in fracture toughness between the two specimens taken from the as-received bloom also confirms the dependence of the toughness upon the position inside the block.

5 ACKNOWLEDGEMENTS

L. Vitrano, H.C.M. Stampi SpA, for steel procurement; E. Stagno, M.G. Ienco, M.R. Pinasco for microstructure interpretation; E. Gianotti, Ferioli & Gianotti SpA, for NT and QT heat-treatments. Financial support by Italian Ministry for Education, University and Research (PRIN 2003091205).

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